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Yannis P. Korkolis
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In cooperation with:



Preface

The International Conference on High Speed Forming (ICHSF) has developed into one of the leading international forums for research and industrial exchange in the field of impulse-based metalworking. Since the first conference in 2004 in Dortmund, initiated by Professor Matthias Kleiner, the biannual meeting has continuously brought together scientists, engineers, manufacturers and industrial operators, to discuss the latest developments in high speed forming technologies and their applications.

We are especially pleased that the 10th ICHSF in 2025 is once again being held as an in-person event in Dortmund following the disruptions caused by the COVID-19 pandemic. Personal interaction has always been an essential part of the conference, enabling not only the presentation of new scientific results but also direct discussions and the exchange of practical experience between academia and industry.

Beyond the technical presentations, the conference serves as a platform for open dialogue between researchers and industrial practitioners, by bringing together 47 participants: 76% from universities and research institutes, and the remaining from industry. The ICHSF 2025 continues its tradition of fostering worldwide international collaboration and advancing the understanding and application of high speed forming technologies, with 72% of the participants coming from Europe, 21% from Asia, and 6% from the Americas

We are particularly happy to see the continued strengthening of the international impulse forming community. Following the last conference held outside Europe in Daejeon, South Korea in 2014, the next ICHSF will further expand this global network when it takes place in Mumbai, India in 2028.

We would like to express our sincere gratitude to all authors, co-authors, reviewers, members of the scientific committee, sponsors, and participants for their valuable contributions and support. Their commitment makes this conference possible and ensures the continued growth of the high speed forming community.

Dortmund, August 2025

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Chapter 1

Impulse forming

Strain rates in electrohydraulic forming of thin stainless-steel sheet metals

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Abstract

Impulse forming technologies can be used for shaping thin sheet metal parts, e.g. bipolar plates for hydrogen production. Thereby, the energy transmission behaviour to the sheet metals depends on actual process characteristics. In electrohydraulic forming with exploding wire, shockwaves are expected to cause transient pressure fields onto the forming area. This transient force or pressure action realises the punch while the die is a rigid part of the tool. Hence, one part of the forming tool does not feature a defined geometry. To understand the impact of such kind of punches the resulting strain field was mapped during electrohydraulic forming experiments. Different explosion wire materials were used to form 100 µm thin EN 1.4404 / AISI 316L stainless-steel sheet metals. The displacement was measured in-process with two laser triangulation sensors at different positions. This time and space resolving data was taken to represent the initial pressure and resulting strain distribution. The in-process measurements enable further optimisation of the process settings.

Keywords

Electrohydraulic Forming, Energy Transmission, In-Process Measurement

1 Introduction

Electrohydraulic forming can be used for shaping thin sheet metal parts, e.g. for bipolar plates as used in hydrogen production. While the die is a rigid part of the tool like in conventional stamping, the punch is realised by a transient pressure field that is built up

instantaneously by an electrical underwater discharge. Various industrial applications have been demonstrated in recent years (Cantergiani et al., 2018; Golovashchenko et al., 2014). Electrical underwater discharge can be carried out with or without an explosion wire, although recent studies have observed an increase in pressure and repeatability when using an explosion wire (Eguia et al., 2011). In addition to the geometry, the properties of the wire material are also likely to determine the forming results and are therefore of interest for process optimisation. This was shown by experiments on the diameter of aluminium wires in electrohydraulic sheet metal forming (Langstädtler et al., 2018). A correlation was shown between the ratio of loading energy and calculated vaporisation energy and the forming result. Earlier investigations showed that the wire materials can have an influence on the explosion behaviour itself (Romanova et al., 2015). It was observed that the wire material changes the vaporisation process and the associated formation of shock waves in air (Rososhek et al., 2018). An increase in the forming result through the use of aluminium wires in electrohydraulic forming was attributed to the chemical potential of aluminium (Löffler et al., 2001). This raises the question of how the wire material influences the energy transfer and the final forming result.

A fast method for high velocity mapping via photon doppler velocimetry was introduced for foil vaporization forming (Lee et al., 2018). Here, the velocities were determined at a single bulging point for impact welding.

In this paper, physical electrohydraulic forming tests were performed with annealed EN 1.4404 / AISI 316L stainless-steel sheet metals with initial thickness of 100 μm using different exploding wire materials at different energy levels. The forming was measured in-process at two separated bulging points with two laser-triangulation distance sensors to detect the distribution of the arising pressure field.

2 Materials and Methods

2.1 Forming set-up

Electrohydraulic forming was performed with a pulsed-power machine. The pulsed-power machine consisted of a capacitor bank with a capacitance of $C = 100 \mu\text{F}$ which was loaded by a power supply to a loading voltage of $U_0 = 1.3 \text{ kV}$. This resulted in a loading energy of $E_C = 85 \text{ J}$. By short-circuiting the capacitor bank by a high current switch (ignitron NL8900, National Electronics, LaFox, IL, USA) over a thin wire, a peak current arose which vaporized the 20 mm long wire resulting in an explosion that was usable for electrohydraulic forming. Stainless-steel sheets (EN 1.4404 / AISI 316L) with dimensions of $100 \times 100 \text{ mm}^2$ and a thickness s_0 of 100 μm , were bulged with a milled die made of EN 1.4301 / AISI 304. The die enabled two simultaneous bulging operations by rectangular cut-outs of $8 \times 16 \text{ mm}^2$ with 3 mm edge radius (see **Fig. 1**). The workpiece and the die were in contact, as the gap was set to 0 mm. A round blank holder with a diameter of 70 mm was used.

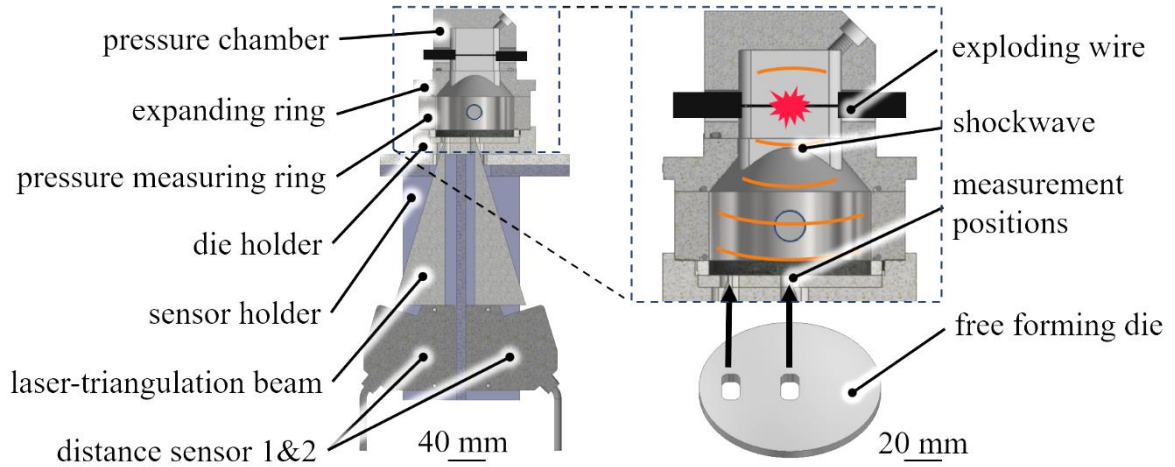


Figure 1: Forming set-up

2.2 Forming procedure

The material properties of the wires are given in **Table 1**. The loading energy E_C was set constant to 85 J. The quotient R of the loading energy and the vaporization energy E_V was kept constant to a value of $R = 1.71 \pm 0.22$. This was reached by adjusting the diameter for the different wire materials, see **Eq. 1** with Δh_v specific vaporization enthalpy, V volume of wire (depends on length and diameter) and ρ density of wire material. It is assumed that the wire can only evaporate completely when the necessary vaporization energy was provided. It should be noted here that the additional melting energy required was assumed to be negligible.

$$R = \frac{E_C}{E_V} = \frac{0.5 \cdot C \cdot U_0^2}{V \cdot \rho \cdot \Delta h_v} \quad (1)$$

Wire materials were pure aluminium (Al99.5), pure copper (Cu99) and pure tungsten (W99). The chosen wire diameter values were $d(\text{Al99.5}) = 0.35$ mm, $d(\text{Cu99}) = 0.28$ mm and $d(\text{W99}) = 0.18$ mm. This resulted in vaporization energies E_V of $E_V(\text{Al99.5}) = 57$ J, $E_V(\text{Cu99}) = 52$ J and $E_V(\text{W99}) = 44$ J and R values as $R(\text{Al99.5}) = 1.49$, $R(\text{Cu99}) = 1.63$ and $R(\text{W99}) = 1.93$, respectively. The tests for each wire material were repeated three times.

Table 1: Wire properties (matweb.com)

wire property	unit	Al99.5	Cu99	W99
melting point θ	°C	660	1083	3370
boiling point	°C	2519	2562	5555
specific vaporization enthalpy Δh_v	kJ/g	10.9	4.7	4.4

2.3 Measurement set-up

Two photoelectric sensors (LK-H157, Keyence, Osaka, Japan) were implemented for measuring the displacement h of the workpiece in the center (sensor 1) and 26 mm to the side (sensor 2) sampled every $2.55 \mu\text{s}$ (see **Fig. 2**). This arrangement enabled a two-point insight into the arising pressure field. The measurement set-up was operated with a separate electrical potential (hv-disconnected) to reduce electrical interferences.

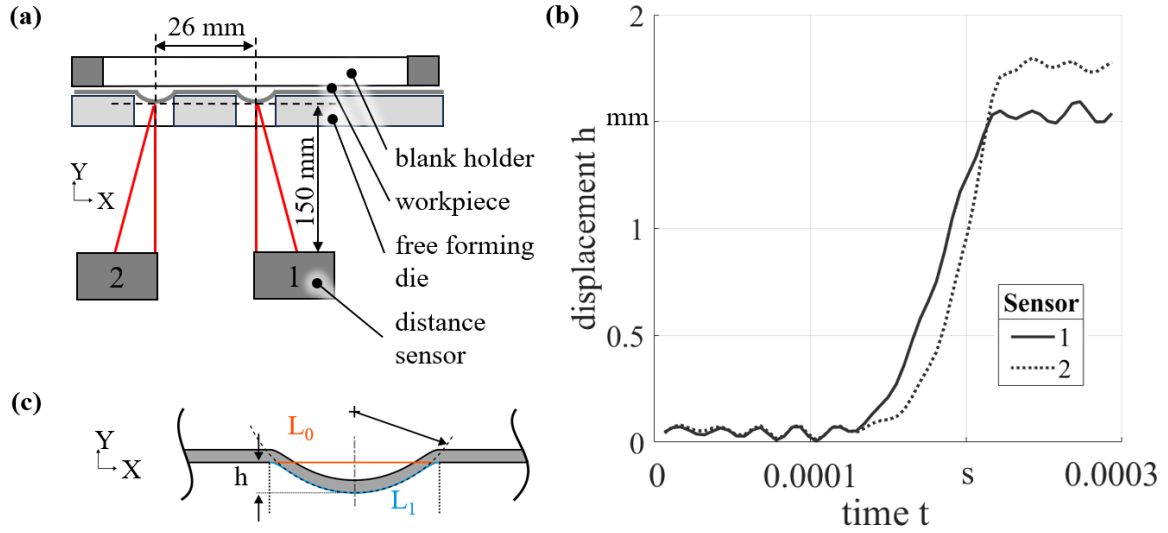


Figure 2: Measurement: (a) sketch of the set-up, (b) example (W99) measured displacement h versus time t and (c) definition of geometric measures for strain calculation

The signal of the displacement h versus time t was low-pass filtered with 50 kHz by minimum-order filter and a stopband attenuation of 60 dB. In connection with the shape in the X-Y-plane (cp. Fig. 2 (c)), the length of the surface segment L_1 was calculated from the displacement h , **Eq. 2**. To calculate the strain ε_g , the length L_1 was set in relation to the initial length L_0 of $L_0 = 8 \text{ mm}$ as given by the forming die (see **Eq. 3**). The strain rate $\dot{\varepsilon}_g$ was determined by differentiate the strain ε_g over time t . The workpiece velocity v and its maximum were obtained by derivation of the displacement h . This method was introduced with a single sensor in the application of electromagnetic forming and laser shock forming (Langstädtler et al., 2024).

$$L_1 = \frac{\arctan\left(\frac{2 \cdot h}{L_0}\right) \cdot (4 \cdot h^2 + L_0^2)}{2 \cdot h} \quad (2)$$

$$\varepsilon_g = \frac{L_1 - L_0}{L_0} \quad (3)$$

3 Results

3.1 Sheet metal displacement

The displacement varied for different wire materials while supplying the same amount of initial loading energy E_C , **Fig. 3**. The displacement curves were superimposed with an interference frequency resulting from the hv-disconnected power supply. The process time was in the range around 50-100 μ s. The lowest displacement was achieved with tungsten wires, Fig. 3a. Furthermore, differences in displacement in the center (sensor 1) and the side (sensor 2) were observed, whereas not only the maximum value, but also the curvature changed. The sheet metal starts to displace first in the center and then at the side with a delay of about 5-10 μ s.

Higher displacement h was measured for copper wires, Fig. 3b. A slight drag of the sensor 2 signal could also be detected here. It is assumed that the time difference in the displacement was caused by the presence of a non-uniform pressure propagation. This can be caused by differences in the explosion or by reflections on the chamber wall. However, the maximum displacement achieved was the same. Highest displacement was observed for aluminium wires, Fig. 3c. The increase of displacement at ≈ 2.5 mm with a change of curvature could be explained by a partial sheet metal fracture. Again, a slight drag and higher maximum value of the displacement at sensor 2 was detected.

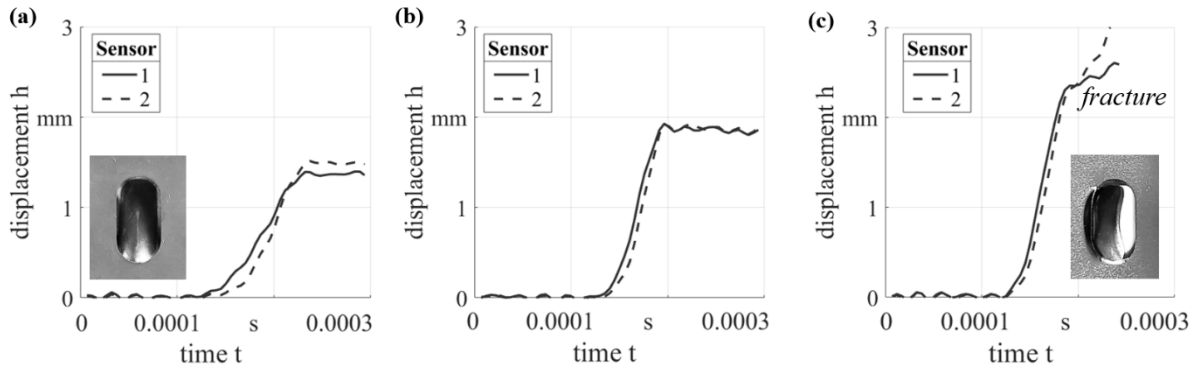


Figure 3: Example curve of displacement h versus time t for different wire materials ($E_C = 85$ J): (a) W99, (b) Cu99 and (c) Al99.5

The maximum sheet velocity values $v_{\max}$ were determined in the range of 30 m/s to 90 m/s depending on the exploding wire material. With tungsten wires (W99) the lowest values were observed, with higher values of around 30 ± 8 m/s at sensor 2 compared to around 25 ± 8 m/s at sensor 1. The fluctuations in the velocities occurred for tungsten wires were relatively high. Copper wires led to higher velocities up to 55 ± 4 m/s. In addition, less difference in maximum velocity between the measurement in the center (sensor 1) and at the side (sensor 2) and less fluctuations could be detected. The highest sheet velocities were achieved with aluminium wires with up to 80 m/s (sensor 1: 60 ± 10 m/s) with slightly higher velocity at the side (sensor 2; 50 ± 10 m/s). Furthermore, the highest fluctuations occurred for aluminium wires. In addition to fluctuations in discharge, this is also attributed to the occasional fracture and cutting of the sheets due to higher pressure peaks.

3.2 Sheet metal strain and strain rate

As expected, the estimated strain and strain rates show similar behaviour as the discussed displacement and velocity. Maximum strains were determined at 0.05 to 0.15 for the different wire materials, **Fig. 4**. The detected failure of the stainless-steel sheets using aluminium wires was detected at 0.2 - 0.25 which correlates with expected fracture strains. Achieved strain rate values were in the range of 2000 and 8000 1/s, **Fig. 5**. Analogous to the sheet velocities, the lowest values were achieved for tungsten, increasing with copper and the highest strain rates using aluminium wires. In addition, the strain rates calculated with sensor 2 tend to be higher. An increase in strain rates is justified by the fact that the wire materials cause different pressure surges and different amounts of energy. As a result, aluminium wires have the highest pressure surge, which leads to higher energy transmission and higher strain rates.

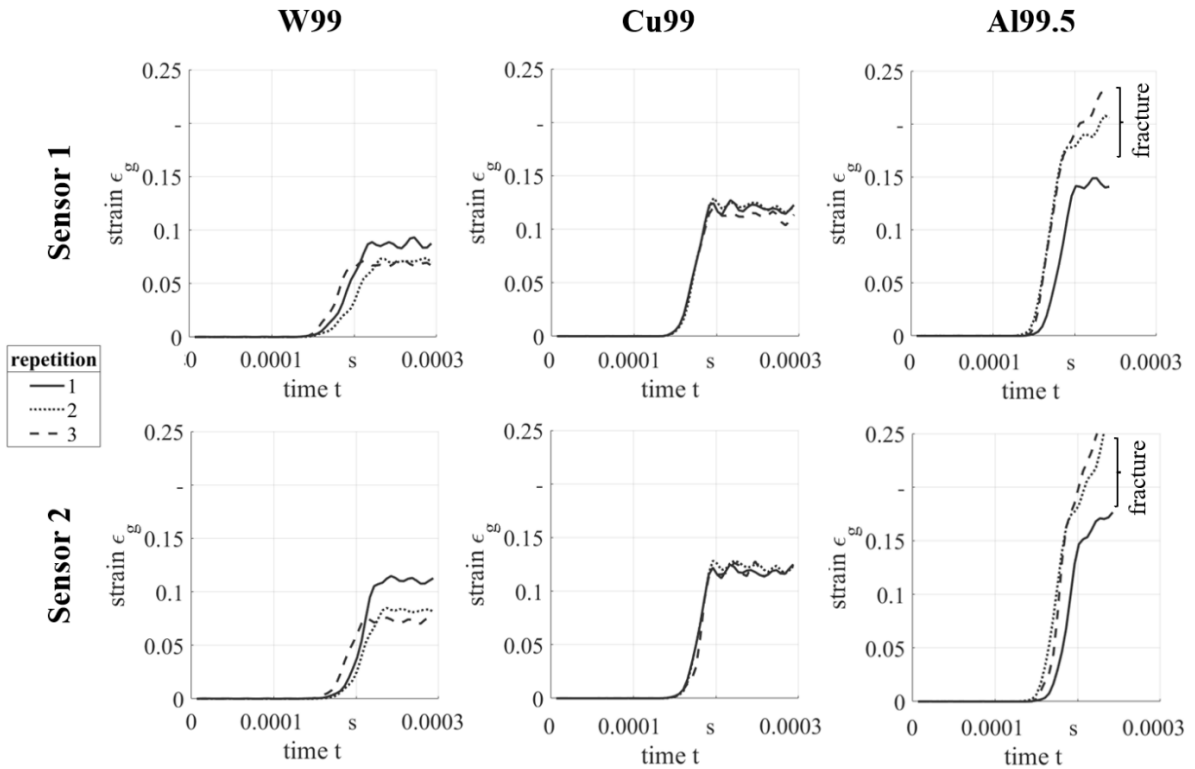


Figure 4: Strain ϵ versus time t for different wire materials and positions ($E_C = 85$ J)

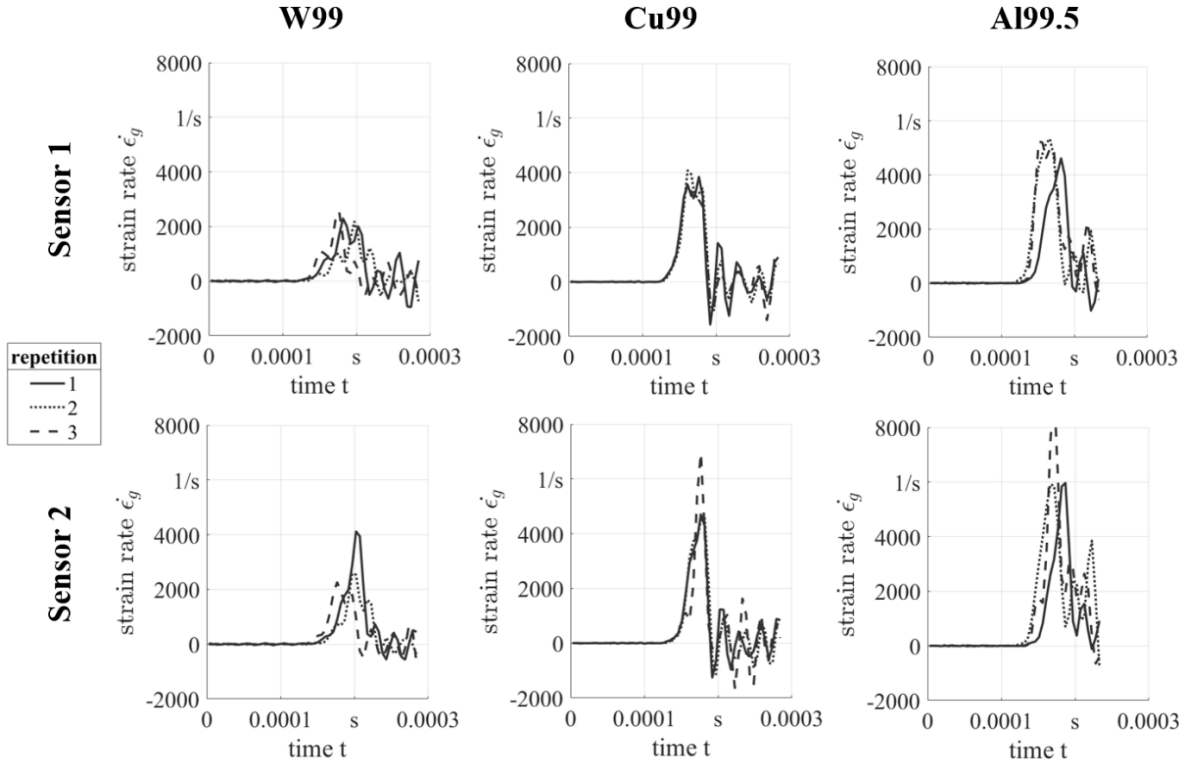


Figure 5 Strain rate $\dot{\epsilon}_g$ versus time t for different wire materials and positions ($E_C = 85 \text{ J}$)

4 Summary and conclusion

A in-process displacement measurement for investigation pressure fields of electrohydraulic forming was introduced in this paper. Interference effects on the measurement signal were reduced using a special power supply. The measuring system was equipped with two sensors to enable the measurement of two separate forming operations within the sheet plane. The tested wire materials had a significant influence on the process behavior of electrohydraulic forming thin stainless-steel sheet metals. The following conclusion could be drawn:

- Sheet metal velocities up to 80 m/s were measured
- Strain rates up to 8000 1/s were calculated
- Differences in displacement over time and maximum velocities as well as the strain and strain rates were observed depending on the wire material
- Gradients in the pressure field were detected which are related to the wire material

The gradients in the values may be caused through differences in the interacting pressure wave. Here, reflections and interferences in the pressure chamber could cause differences in the effect. It is also expected that the resulting gradients in the pressure fields could change depending on the R value which will be investigated in future work.

Aluminium wires showed the highest straining which could be related to the higher reactivity of the wire material. However, aluminium as well as tungsten wires tended to higher fluctuations and caused differences in the pressure fields. Therefore, it would be desirable to combine the positive properties of the wires such as reactivity with the stable behavior by using composite wires.

Acknowledgments

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Production of individual sheet parts using vaporizing foil actuator forming and a flexible liquid-solid tool

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Abstract

Conventional sheet metal forming is inflexible, as each part requires costly punch-and-die tools, usually produced by machining of solid steel. These tools are suitable for large batches but inefficient for individual parts, as the cost per part is high. The aim is to produce individual parts with flexibly applicable tools to also reduce the tool costs per part. The high-speed process of vaporizing foil actuator forming (VFAF) decreases the tool costs by replacing the solid steel punch with a flexible, low-cost, electrically exploding aluminum foil. Until now, also VFAF relies on conventional solid steel dies. This paper investigates the feasibility of a novel flexible liquid-solid tool to manufacture patient-specific cranial implants as an example of highly individual parts. The liquid solid-tool comprises of a thin form-giving additively manufactured (AM) polymer die which is supported by a shear thickening fluid (STF). When the sheet impacts the AM polymer, the STF behaves quasi-solid providing structural die support during forming. The new tool concept also increases the flexibility by potentially reusing a part of the tool – the STF. The results show that the new tool can be designed to successfully manufacture cranial implants with a nominal geometry deviation of ± 1 mm. The deviation lies in the range of other competing flexible processes like powder bed fusion or single point incremental forming. Hence, the liquid-solid tool can replace solid steel dies and decrease the tool costs drastically compared to the solid dies.

Keywords

Vaporizing foil actuator forming, Shear thickening fluid, Individual parts, Flexible tooling

1 Introduction

In conventional sheet metal forming the tools, such as punches and dies, are machined from solid steel and are inherently geometry-bound to one part design. Long lead times make these tools both time- and cost-consuming. While the high costs and extended lead times are usually not critical in mass production, since the expenses are distributed across a large number of parts, resulting in a low tooling cost per unit. However, they become a significant drawback for individual or low-volume production, where the tooling cost per part remains high. In recent years, several attempts were made to replace geometry-bound tools with flexible tools to produce different individual parts with the same tool and thereby decrease the tool costs per part. For example, the rubber pad forming replaces the die with a flexible rubber pad (Afteni et al., 2018). Nakajima (1969) developed a reconfigurable matrix of pins, which can move orthogonally to the tool base, in order to replace both geometry-bound tools. Single point incremental forming (SPIF) entirely avoids the usage of any specific die and punch (Leszak, 1967). However, SPIF has a low geometrical accuracy, partly due to the high elastic springback of the locally formed part, which needs to be taken into account during the process design. In contrast to this, impulse forming processes decrease the elastic springback (Baron and Henn, 1964) due to an elastic (Golovashchenko, 2009) or plastic wave front (Kamal, 2005) which travels through the thickness of the sheet and releases residual stresses and strains. Furthermore, working media- or working energy-based impulse processes replace the punch with a flexibly applicable impulse pressure due to their working principle. However, until now also in impulse forming processes solid steel dies are used. Pegel et al. (2018) showed for electrohydraulic forming, that the steel die can be replaced by a bulk additively manufactured (AM) polymer die, which can resist the impact from the sheet. Later, Hahn et al. (2021) invented a novel flexible liquid-solid tooling concept. There, a thin AM polymer die is refilled with a shear thickening fluid (STF). The STF solidifies upon impact of the sheet, thereby mechanically supporting the thin AM polymer die. A main advantage of this method is the reusability of the STF for other geometries. Hahn et al. (2021) validated this concept for electromagnetic forming (EMF) and showed the necessity of the STF. However, a key limitation of EMF is its reliance on a high electrical conductivity, restricting its applicability to a range of materials. To address this limitation, impulse forming processes that are independent of a material's conductivity, such as vaporizing foil actuator forming (VFAF), can be employed. The VFAF uses an actuator, e.g., an aluminum foil, which is connected to a capacitor bank with a resistance R , an inductance L , and a capacity C . Upon discharge, within a few microseconds, the actuator undergoes resistive heating due to the application of the high voltage $U(t)$ and electrical current $I(t)$, the foil actuator vaporizes leading to an explosive expansion. This rapid expansion generates the impulse pressure $p(t)$ that forms the sheet metal into a form-giving die via an intermediate elastomer layer (**Fig. 1**).

This paper investigates the feasibility of forming an individual part with VFAF and the new liquid-solid tool concept. The individual part is a real-world example of patient specific cranial implants published by Li et al. (2021a) (**Fig. 5 a**). In the following, the

design of the novel tool, including its needed air evacuation, is described. Then, its impact on experimental forming results is analyzed.

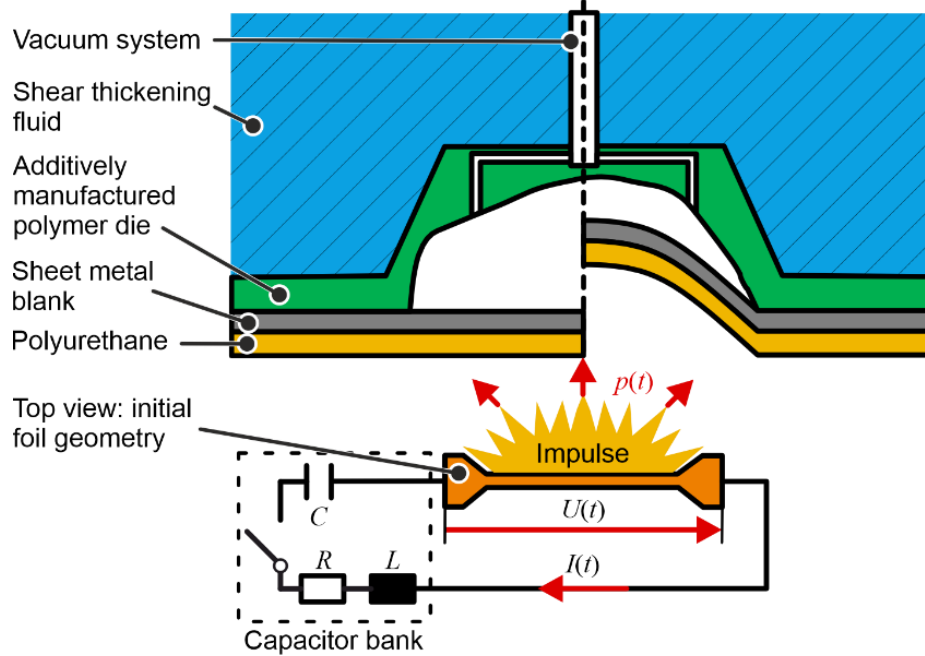


Figure 1: Schematic principle of the vaporizing foil actuator forming with the new flexible liquid-solid tool concept

2 Methodology

First, the experimental setup for the part forming is described, along with the method used to evaluate the forming results. Second, the design is presented.

2.1 Experimental setup

Principally the same experimental setup for the VFAF as in Hahn et al. (2019) and for the liquid-solid tool concept as in Hahn et al. (2021) is used. Therefore, only the most important aspects are mentioned here. The actuator is an 0.05 mm thin aluminum foil with a rectangular vaporization area. The foil is insulated with polyamide tape. The vaporization area has dimensions of $b_0 = 18$ mm (x-direction) and $l_0 = 61.7$ mm (y-direction). The longer dimension of the cranial implant is oriented along the vaporization area (y-direction). The energy input deposited until the burst E_b is calculated using MATLAB by analyzing the measured time-dependent voltage $U(t)$ and current $I(t)$ data. The burst energy w_b per vaporized foil mass m , defined as $w_b = E_b / m$, is thus determined. The discharge energy of the capacitor bank is chosen to be $E_{\text{inp.}} = 9.48$ kJ and is kept constant. The measured specific burst energy w_b is an indicator for the forming pressure. The sheet material is titanium grade 2 with a thickness $s = 0.5$ mm. The STF is a cornstarch water suspension with a cornstarch concentration of 55-wt%, which solidifies upon impact, as reported by Crawford et al. (2013). The accuracy of the formed cranial implants is evaluated by scanning the implant

using the ATOS Triple Scan system from the company Carl Zeiss AG. The scanned geometries are then compared to the nominal CAD geometry using the 3D analysis software GOM Professional 2020. The minimum and maximum surface deviations $\delta_{\min}$ and $\delta_{\max}$ are determined by the normal distance from the nominal CAD geometry to the formed sheet metal.

2.2 Die design

The die is developed using Autodesk Inventor Professional 2024 software, with a patient-specific cranial implant surface from Li et al. (2021b) (Fig. 5 a)) defining the form-giving contour of the die. The target surface for the cranial implant is extrapolated tangentially in the radial direction by 10 mm to prevent discontinuities between the target surface of the implant and the remaining die surface. Preliminary studies were carried out to determine the position of the air evacuation system to withstand the impact of the sheet. Here, only the final design is shown (**Fig. 2 a)**). The air evacuation system is positioned at $\pm 45^\circ$ to the vaporization area (**Fig. 2 b)**) with a diameter of $d = 2$ mm and a minimum die thickness of $t_{\min} = 12.9$ mm. The die is additively manufactured using fused deposition modeling (FDM) with ABS material on an UltiMaker 3. The layer thickness is set to $t_s = 0.15$ mm and a 100 % infill density is used.

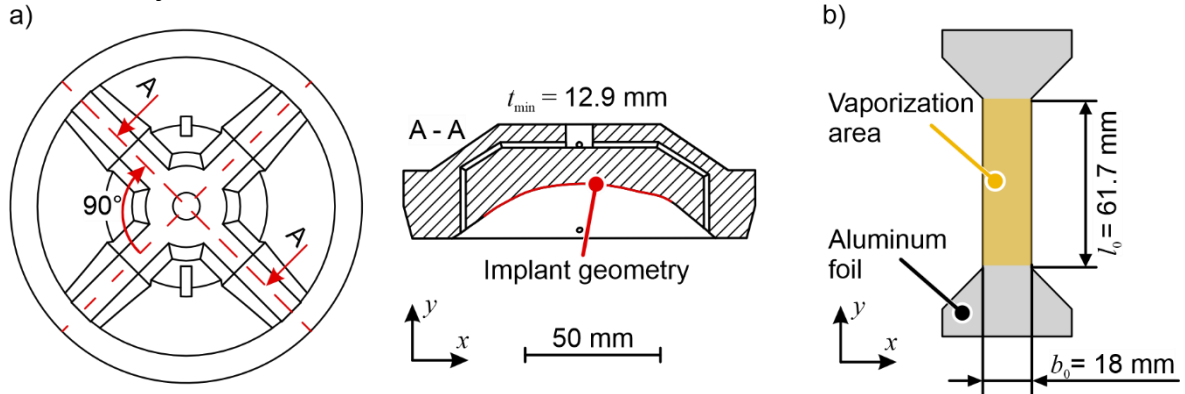


Figure 2: The position of a) the air evacuation system $\pm 45^\circ$ to the x-direction and b) the orientation of the vaporizing foil

3 Results and Discussion

This section analyzes the geometric accuracy of the formed cranial implant and addresses the influence of the STF.

3.1 Geometric accuracy after VFAF

The maximum and minimum surface deviation from the CAD geometry after forming and before laser cutting is $\delta_{\min} = -0.83$ mm and $\delta_{\max} = 0.88$ mm (**Fig. 3 a)**). Together with the preliminary experiments (not shown here), the results show a consistent trend where the surface of the formed implant is lower than the nominal CAD geometry (negative deviation)

in the direction orthogonal to the vaporization area (x -direction), while it is higher (positive deviation) along the vaporization area (y -direction).

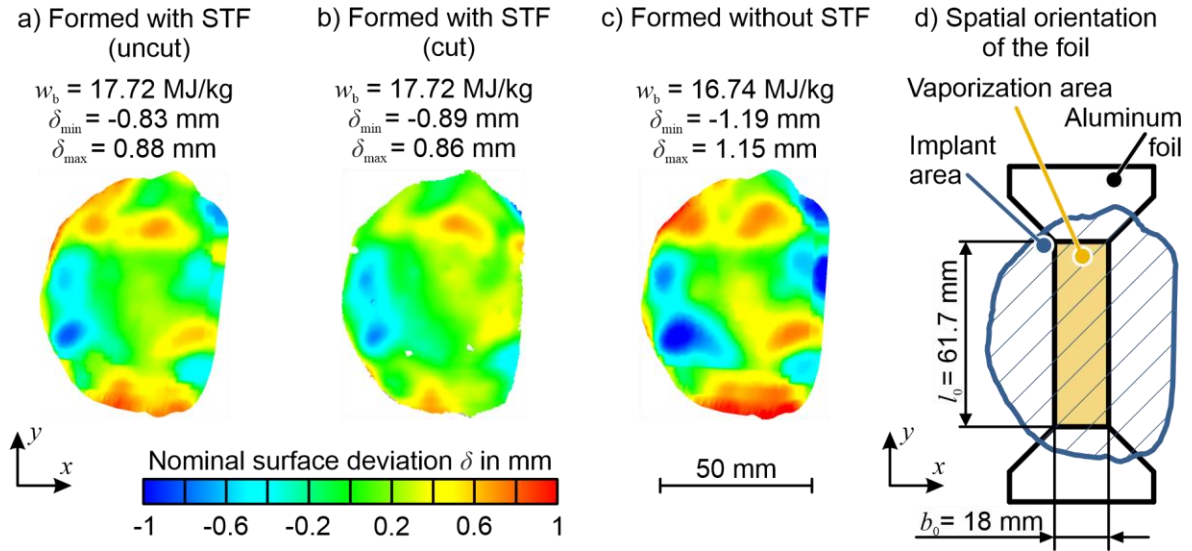


Figure 3: Nominal surface deviation from the CAD geometry after VFAF a) formed with STF (uncut), b) formed with STF (cut), c) formed without STF and, d) spatial orientation of the foil to the later implant

One explanation for these deviations is the (burst) pressure distribution. The extended vaporization zone in the longer foil direction, which is the direction of current flow, generates a correspondingly longer pressure pulse in that direction (y). This plastically accelerates a larger sheet mass fraction in this direction compared to the direction orthogonal to it (x). This causes the die to elastically flex more in the long foil direction than orthogonal to it and thus causes the observed differences in surface deviations. This is supported by numerical simulations, which are beyond the scope of this paper. A possible solution to avoid these inaccuracies could be an equal pressure distribution in both directions. After laser cutting the cranial implant out of the sheet, the minimum surface deviation increases to $\delta_{\min} = -0.89 \text{ mm}$ and the maximum deviation decreases to $\delta_{\max} = 0.86 \text{ mm}$ (**Fig. 3 b**). This change lies within the range of the measuring error, though. The necessity of the STF during VFAF is proven with an additional experiment without STF. The absence of the STF decreases respectively increases the surface deviation to $\delta_{\min} = -1.19 \text{ mm}$ and $\delta_{\max} = 1.15 \text{ mm}$ (**Fig. 3 c**). Moreover, the missing structural support during forming causes a horizontal delamination of the AM die (**Fig. 4**), so that it cannot be used for more parts. It is concluded that this horizontal delamination leads to the increased surface deviation due to the lack of structural support. The results thus show the necessity of the STF in maintaining geometric accuracy in VFAF, also for a batch size of just one.

The burst pressure has a significant influence on the surface derivation. Preliminary studies to determine the necessary burst pressure, beyond the scope of this paper, show that if the burst pressure is too low, die contact may not occur, thereby increasing surface

deviations and if the burst pressure is too high, it can lead to delamination of the AM die, which also increases surface deviations, as illustrated in Fig. 3c).

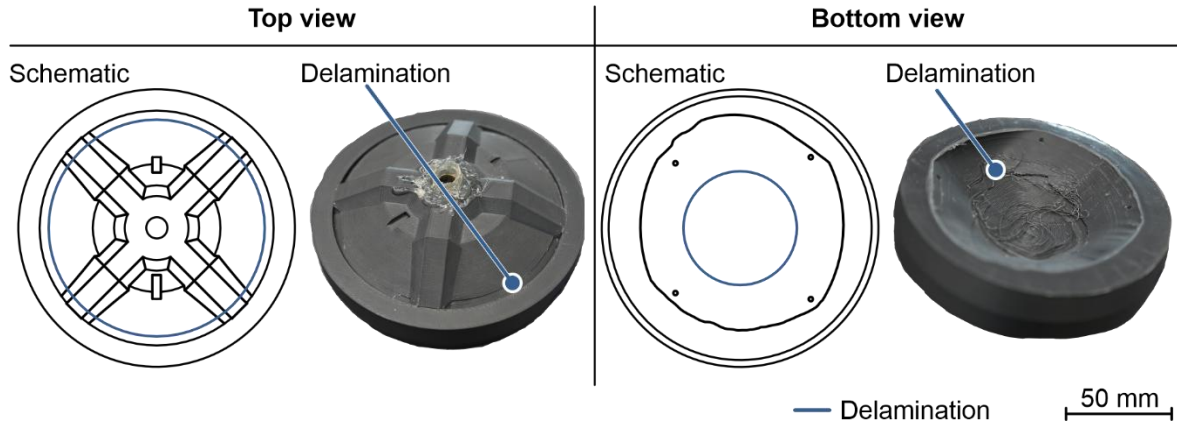


Figure 4: Visualisation of the AM die failure by delamination of the layers for a part formed without STF

3.2 Demonstrator

The achieved geometric accuracy of the formed implant (**Fig. 5 b)- c)** with $\delta_{\min} = -0.89$ mm and $\delta_{\max} = 0.86$ mm is comparable to that of additively manufactured titanium cranial implants with $\delta_{\min} = -1.11$ mm and $\delta_{\max} = 0.42$ mm (Ackermann et al., 2019) and even more accurate compared to implants produced via incremental sheet forming with $\delta_{\min} = -2.95$ mm and $\delta_{\max} = 1.83$ mm (Thakur and Chauhan, 2024). The results show that employing VFAF in combination with the new flexible liquid-solid tool enables a competitive manufacture of individual parts. Furthermore, without addressing detailed calculations, the tool cost for the novel tool concept is 64 % lower compared to a solid steel die and reduces the scrap by 81 vol. % compared to a conventionally machined solid steel die.

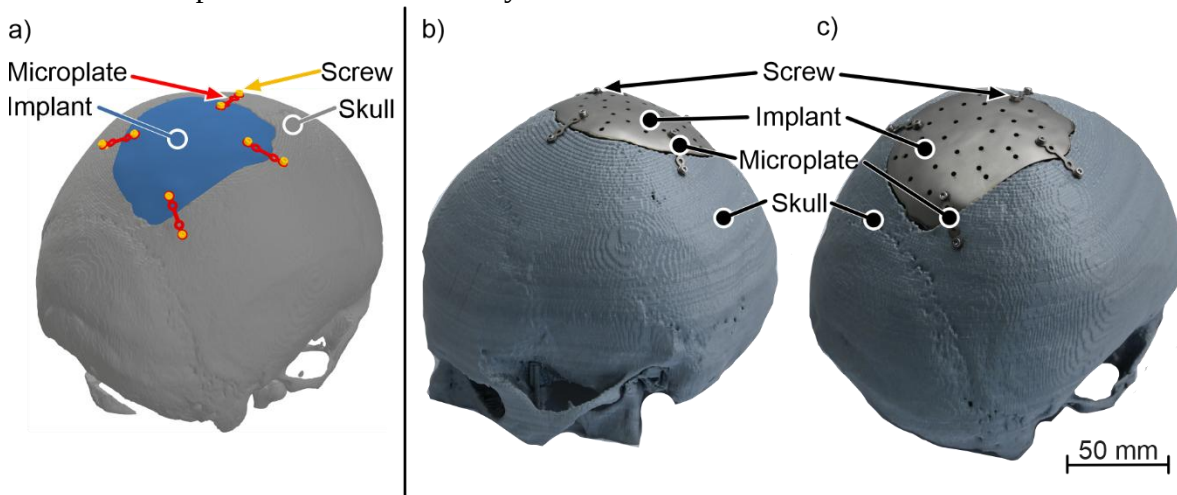


Figure 5: Patient-specific cranial implant a) nominal geometry b) - c) manufactured by VFAF with the new tool concept (for the data see Li et al. (2021b), Jindal et al. (2023))

4 Summary and Outlook

This study demonstrates the feasibility of manufacturing individual sheet parts, like patient-specific cranial implants, using VFAF in combination with the flexible liquid-solid tool concept. The usage of AM allows the design of complex air evacuation channels and a quick manufacturing of the polymer die later filled with the needed STF. The geometric accuracy of the formed implants is found to be comparable to that of directly additively manufactured implants and superior to implants produced by incremental sheet forming. In future work, a numerical model of the process that incorporates a fracture criterion for the AM polymer die could reduce the experimental as well as the time effort to design a suitable AM die.

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An Experimental Study: Improved Electrohydraulic Forming Efficiency through Wire-Free Discharge

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Abstract

This experimental research proposes improved electrohydraulic forming (EHF) methods to overcome the limitations of traditional high-speed forming and deep-drawing processes. EHF uses a working fluid to generate forming force, which lasts longer than in electromagnetic or explosive forming, allowing precise shape control. However, conventional EHF requires fastening wires and draining/refilling the working fluid for each experiment, resulting in long preparation times and inconsistent forming due to variation in wire position. To address these issues, two new methods ‘wire-cross EHF’ and ‘wire-free EHF’ were introduced and tested using a free-bulging die and 0.5 t STS430 specimens. Forming distribution was analysed by comparing forming force, forming height, and preparation time. The wire-free EHF showed lower initial forming force but achieved comparable forming height through consecutive experiments. It also demonstrated high consistency and reduced total experiment time by 4.72 times compared to the conventional EHF, which suffered from increased wire length during twisting, leading to variations in forming height and forming distribution. Overall, the wire-free EHF proved most effective in achieving consistent forming results with significantly reduced preparation time, making it a promising approach for efficient EHF process.

Keywords

Free forming, Hydroforming, Electrode

1 Introduction

In response to increasingly stringent environmental regulations, the transportation industry—including automobiles and aircraft—has been focusing on improving fuel efficiency by reducing vehicle weight [1]. To form the outer body panels of such vehicles, the deep-drawing method using presses and dies is widely applied. However, when using materials such as aluminium alloys or high-strength steels, the formability is relatively low, often resulting in discrepancies between the intended and actual formed shapes due to elastic recovery. Various studies are underway to address this issue [2].

EHF, a high-speed forming technique, has emerged as a promising alternative to enhance forming distribution in sheet metal processing [3]. The rapid forming speed generates compressive stresses that suppress void growth during deformation, thereby improving the material's extension capacity. Furthermore, unlike other high-speed forming methods, EHF utilizes hydraulic pressure as a medium, enabling a longer application of forming force. This extended pressure reduces spring back effect, improving forming accuracy. Despite these advantages, EHF has not been widely adopted in the manufacturing industry due to time-consuming preparation and relatively high costs.

This research explores two novel EHF approaches aimed at reducing experimental preparation time while maintaining consistent forming distribution. Specifically, two modified experimental setups – a wire-free configuration and a simplified wire-cross configuration – were tested and compared using 0.5 t STS430 rectangular sheets and a free-bulging die. In both methods, the discharge voltage was fixed at 6 kV. As the wire-free setup tends to generate lower forming force in a single discharge, a consecutive discharge approach was employed to achieve a comparable bulging height. The number of discharges and total time required to reach equivalent forming results were evaluated using 3D scanning results of formed specimens.

2 Experimental Setup

2.1 Conventional Electrohydraulic Forming (Using single wire)

EHF used in this research is illustrated in Fig. 1. It comprises a die, a chamber set capable of draining and refilling working fluid, and a sheet positioned between them. The die is designed with a hole within the holder region that secures the sheet, allowing for the installation of various shapes of die configurations. In this research, a free-bulging die was employed. The die has a cylindrical shape with a diameter of 120 mm and a height of 65 mm. The radius at the sheet inlet region is 15 mm. The chamber contains a pair of electrodes through which electric energy, stored in a capacitor, is discharged. It has a rounded conical shape in its side profile, with a maximum depth of 100 mm measured from the sheet. Its diameter is 150 mm, determined based on the inlet region of the die. The sheet used in the experiments was a rectangular STS430 specimen with a length of 460 mm, a width of 230 mm and a thickness of 0.5 mm.

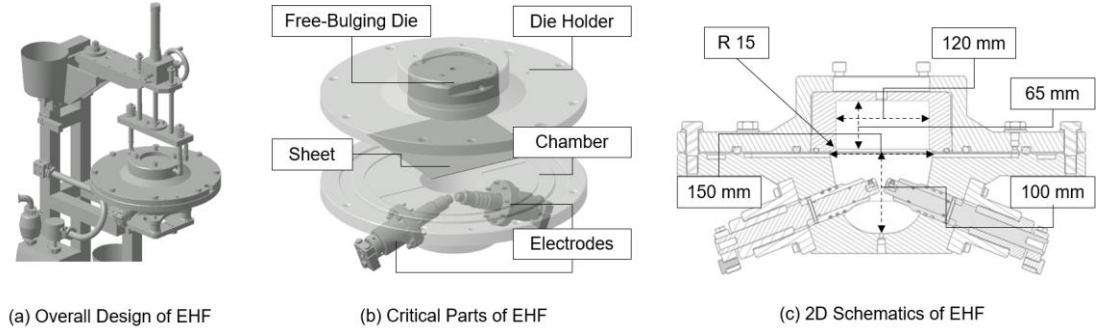


Figure 1: Apparatus of electrohydraulic forming device

In the conventional setup, a thin wire is connected between the electrodes. When the capacitor discharges, a high current flows instantaneously through the wire, ideally concentrating the electrical energy at its center. This causes the wire to explode, and the resulting high-pressure shockwave transmits forming force through the working fluid to the sheet.

The magnitude of the forming force is primarily determined by the capacitor specifications and the input voltage. However, the location of the wire also influences the effectiveness of the forming process. As shown in Fig. 2, under identical conditions (input voltage, capacitor capacity, wire material, and length), two key variables, stand-off distance (SOD) and electrode gap (EGAP) significantly affect the forming force. Experimental results confirmed that the maximum forming force in the conventional configuration is achieved at an EGAP of 20 mm and SOD of 35 mm.

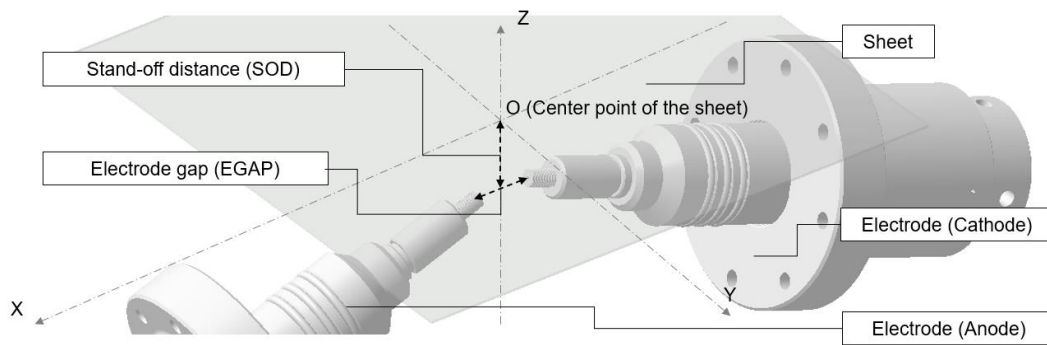


Figure 2: Definition of variables of electrohydraulic forming force

2.2 Proposed Electrohydraulic Forming Methods

To address the time-consuming nature of wire installation in conventional EHF, this research proposes two alternative methods. The first is a wire-free method, which eliminates the wire entirely by placing two electrode tips at a narrow gap. The second is a wire-cross method, where two short wire segments are connected to each electrode in a crossing configuration.

The wire-free method was inspired by in-house experimental observations which showed that electrical conduction can occur across a short gap under high-voltage discharge.

This method is like the EDM (electric discharge machining), which processes the surface of a material using electrical discharge. The wire-cross method, on the other hand, is based on the idea that a wire can fully vaporize into plasma before generating the forming force.

As illustrated in Fig. 3, the wire-free method can significantly reduce preparation time by removing the need to install wires between experiments. Meanwhile, the wire-cross method, although still requiring wire installation, allows faster and more consistent setup compared to the conventional method. Furthermore, it can generate forming forces comparable to those of the conventional setup due to the closed-circuit configuration.

To determine which of the three methods, conventional, wire-free, or wire-cross EHF is the most time-efficient, experiments were conducted to establish the optimal forming conditions that produce the maximum forming force in each case. With the input voltage fixed at 6 kV, suitable SOD and EGAP values were experimentally identified.

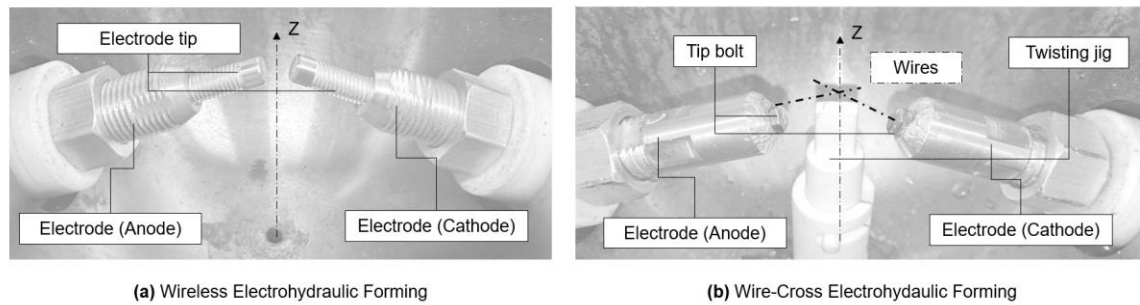


Figure 3: Two proposed electrohydraulic forming methods

2.3 Pre-Experiment for Maximum Forming Force

The wire-cross method utilizes the same wire material and length as the conventional case and thus adopts the same SOD and EGAP values. However, in the wire-free method, the electrode tips have a larger cross-sectional area to prevent electrode damage during consecutive discharges. Therefore, a separate experiment was conducted to determine the optimal electrode positioning for maximum forming force in the wire-free configuration.

Due to the 20-degree incline of the electrodes in the owned EHF device, SOD and EGAP are geometrically linked. Five experimental cases were designed by varying EGAP to investigate its effect on the bulging height and electrode wear in the wire-free method. The formed sheet's bulging height and electrode tip wear were measured accordingly.

As shown in Fig. 4, an EGAP of 7 mm yielded the highest bulging height, and was thus selected for the wire-free configuration. The results show that as EGAP increases, electrode wear decreases and bulging height increases. This is attributed to the reduced arc concentration at the electrode tip, leading to less electrode wear, and the more favorable formation and propagation of the pressure wave with increased EGAP. However, beyond an EGAP of 9 mm, bulging height begins to decrease, and electrode wear increases again. At EGAP 11 mm, no forming force was observed due to the failure of electric discharge in the open circuit, indicating a limit in electrical energy transfer from the capacitor to the working

fluid. It's because if the EGAP is too large, the discharge energy dissipates within the working fluid before reaching the sheet.

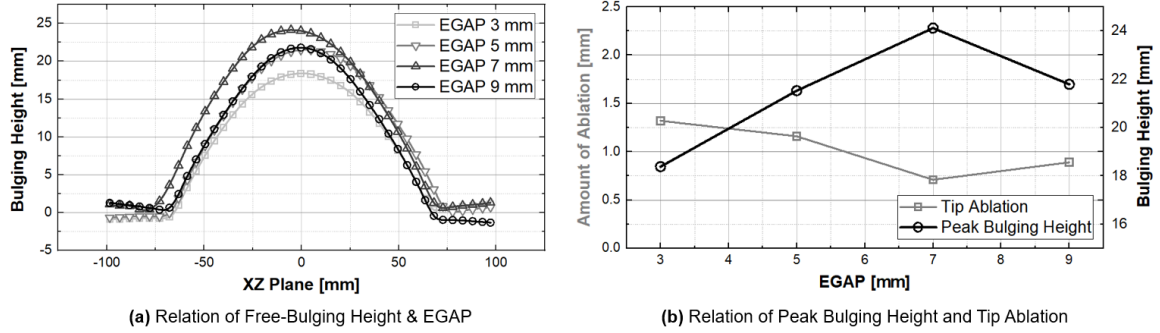


Figure 4: Relation among EGAP, amount of tip ablation and free-bulging height

3 Experimental Results

3.1 Comparison of Free-Bulging Height

Three experiments were conducted for each forming method, conventional, wire-free and wire-cross using 0.5 mm thick STS430 rectangular specimens. The bulging results were captured via 3D scanning along the plane parallel to the current flow direction. In the wire-free case, up to three consecutive discharges were applied to achieve bulging heights comparable to the wire-based methods under the same voltage input. These consecutive discharges were conducted without draining and refilling the working fluid and without separating the die and chamber. To maintain a constant EGAP, the electrode was moved forward during successive discharges, compensating for the ablation observed in Section 2.3.

Fig. 5 presents the forming results for the conventional and wire-cross EHF. Fig. 6 shows the results of one-time and two-time consecutive discharges repeated three times each for the wire-free method. The results of three-time consecutive discharges were excluded from Fig. 6 as they exceeded the forming heights obtained from the wire-based methods under the same voltage. The peak heights of each test are summarized in Table. 1

For a single discharge, the conventional method produced an average bulging height of 26.99 mm, while the wire-cross method achieved a higher average of 30.12 mm. The wire-free method yielded 23.11 mm with a single discharge, 29.81 mm with two discharges, and 32.09 mm with three discharges. These results indicate that the wire-cross EHF outperforms the conventional EHF in terms of forming height, likely due to the increased wire length in the crossed-wire configuration, which generates greater explosive energy compared to the tightly fixed wire in the conventional setup.

In contrast, the single-discharge performance of the wire-free EHF was lower due to its open-circuit configuration. However, by applying consecutive discharges, the forming height increased progressively by 6.7 mm in the second discharge and by 2.28 mm in the third, demonstrating the method's cumulative forming effect. This reduction in incremental

height is due to the increase in SOD with each discharge, which causes greater energy loss as the shockwave travels farther through the working fluid before reaching the sheet.

Moreover, the wire-free EHF exhibited more consistent forming distribution across repeated trials. In wire-based method precise central positioning of the wire is difficult to achieve manually, often resulting in off-centered bulging peaks. In the wire-free case, the peak consistently formed at or near the center of the chamber-die assembly and shifted closer to the center with each consecutive discharge, confirming improved uniformity.

Experiment	Conventional	Wire-cross	Wire-free #1	Wire-free #2	Wire-free #3
No. 1	25.77	28.01	24.12	29.48	31.70
No. 2	27.87	31.66	22.64	29.74	32.50
No. 3	27.34	30.66	22.58	30.18	32.08

Table 1: Maximum bulging height for each experimental method [mm]

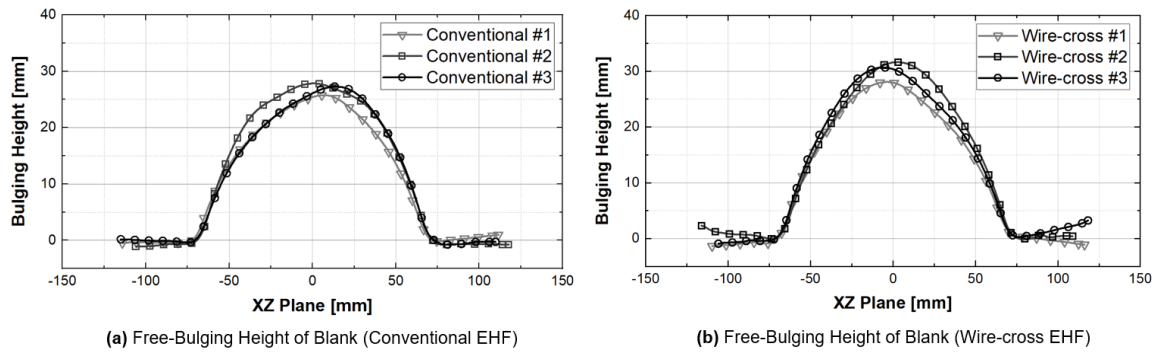


Figure 5: Experimental results with wire-using electrohydraulic forming

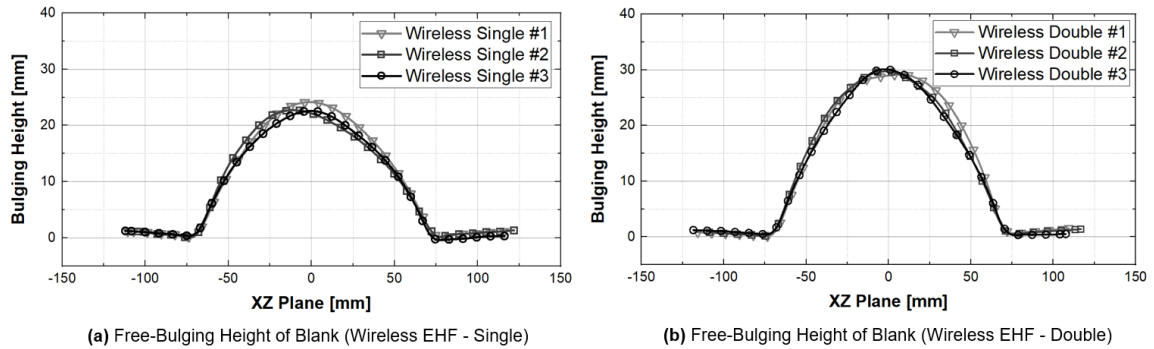


Figure 6: Experimental results with wire-free electrohydraulic forming

3.2 Comparison of Experimental Time

Fig. 7 illustrates the overall process flow and time required for each experimental method. While routine equipment maintenance, such as electrode tip replacement took minimal time, a significant portion of the time in wire-based methods was consumed in the wire installation. Accurately placing and tensioning the wire across the chamber's center with consistent

length was especially time intensive. Although the wire-cross method offers more flexibility in wire length, it still required 12.06 times more preparation time than the wire-free method.

In contrast, the wire-free method drastically reduced time requirements by eliminating the need to install wires or replace the working fluid between discharges. After each test, only the electrode needed to be adjusted to maintain the constant EGAP. Table 2 summarizes the time required for each method: 20.6 minutes for a single experiment in the conventional EHF, 14.9 minutes for the wire-cross EHF, and just 3.2 minutes for the wire-free EHF. For consecutive experiments, the time difference becomes even more significant. The conventional and wire-cross EHF must repeat all process steps, including wire installation and fluid replacement, while the wire-free method only requires capacitor recharging and electrode adjustment. Consequently, the second wire-free discharge takes only 0.4 minutes.

These results demonstrate that the wire-free method can reduce experimental time by up to 98% compared to the conventional method and 97% compared to the wire-cross method. Furthermore, the time savings increase proportionally with the number of tests confirming the efficiency and practicality of the proposed wire-free electrohydraulic forming approach.

Method	Conventional	Wire-cross	Wire-free #1	Wire-free #2	Wire-free #3
Time	20.6	14.9	3.2	3.6	4.0

Table 2: Time required for each experimental method [min]

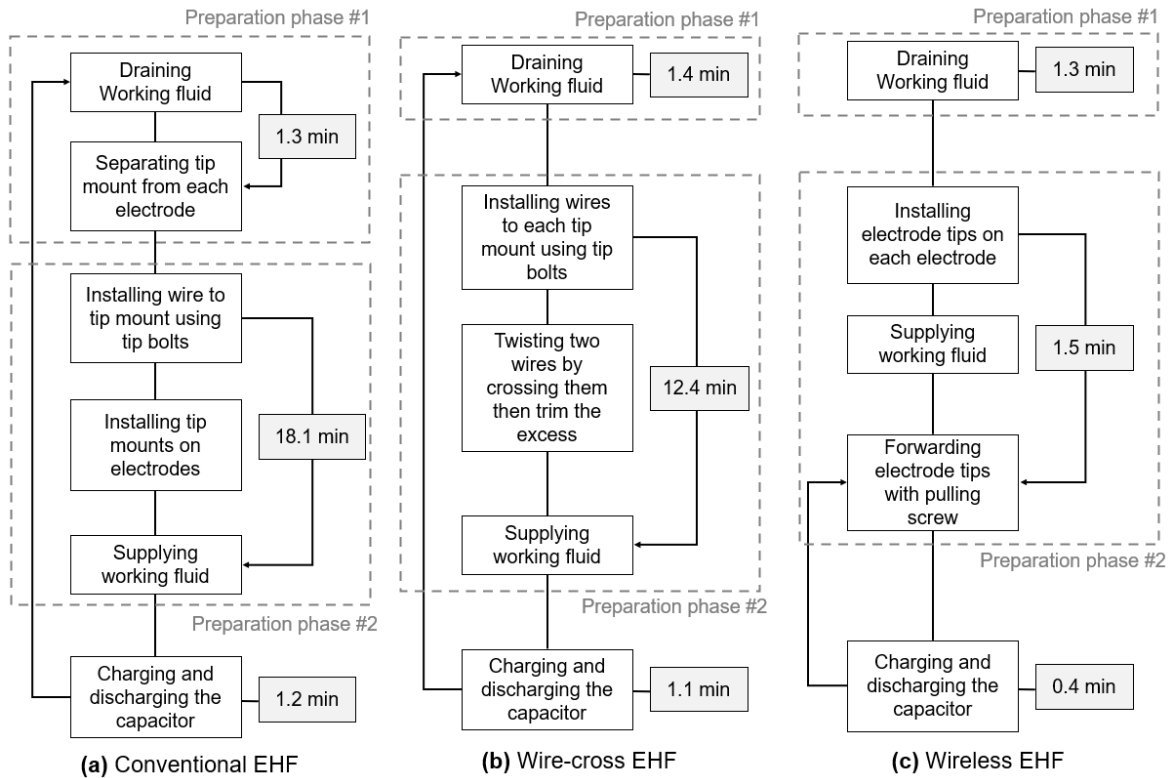


Figure 7: Working flow charts and time required for each experimental method

4 Summary

This study focused on addressing the inefficiencies of conventional electrohydraulic forming (EHF), a high-speed forming technique that offers several advantages over the traditional deep-drawing method. In particular, the focus was on reducing the overall processing time by minimizing wire installation, which is one of the most time-consuming steps in EHF.

To tackle this issue, two alternative methods: wire-cross and wire-free EHF were proposed and experimentally evaluated. Both approaches aimed to resolve the inefficiencies inherent in the conventional experiment. Among the three methods tested under identical input voltage conditions, the wire-cross method produced the highest bulging height. However, both wire-based methods demonstrated inconsistency in forming results when wires were not precisely fastened, and accurate installation remained a time-intensive process.

In contrast, the wire-free method, while yielding the lowest forming height in a single experiment, significantly reduced the experimental time. Furthermore, by applying consecutive discharges, the wire-free approach was able to achieve bulging heights comparable to those of the wire-based methods. Despite requiring multiple discharges, the total time remained substantially shorter, highlighting the wire-free methods' potential as more efficient and reliable alternative for electrohydraulic forming.

4.1 Acknowledgments

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Determination of the High Strain Rate Behaviour of a CuCrZr Alloy Using an Electromagnetic Forming Bench Test

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Abstract

A novel electromagnetic forming (EMF) bench test has been developed to characterize the dynamic behaviour of metallic materials. This paper presents the specific case of a CuCrZr wrought alloy, detailing both the experimental approach and the inverse numerical methodology. The primary advantage of the proposed method lies in its ability to calibrate a dynamic material model, which is particularly relevant for forming applications involving thin specimens, with thicknesses as low as 0.6 mm. The well-known Johnson-Cook model is used without consideration of the thermal softening term. The calibration is validated within the strain rate range of $[1-4000]s^{-1}$. It is indeed shown that this method generates a broad variation of strain rates during the 300 μs test duration. The benefits from using a strain rate sensitive law as opposed to a quasi-static one are also demonstrated. Analysis of plastic strains and peak stresses further indicates that the EMF test is particularly well-suited for ductile materials, whereas brittle materials may fracture prematurely or fail to deform sufficiently. Limitations regarding the non-uniqueness of the calibrated model and the incorporation of thermal effects are briefly discussed.

Keywords

Dynamics, Modelling, Forming

1 Introduction

High Pulsed Power (HPP) processes involve the rapid discharge of stored electrical energy (typically in the range of tens of kJ) into a specialized setup. This energy is converted into magnetic pressure, which can be harnessed for various applications such as electromagnetic forming, welding, or crimping of metallic components. Regardless of the application, both the tooling and the workpiece experience severe mechanical loadings under dynamic conditions during an HPP process. High strain rates are particularly advantageous in metal forming of thin materials. It is therefore crucial to characterize the dynamic behaviour of materials to better predict the lifetime of tooling and the final shape of a workpiece.

Depending on the range of strain rates considered, different experimental techniques are available to assess the dynamic behaviour of a metal such as the Split Hopkinson Pressure Bars test (10^2 to 10^4 s⁻¹) or the Taylor Test (10^4 to 10^6 s⁻¹). Novel techniques are being developed as well, for instance the High Current High Strain Rate Tensile testing platform (10^3 to 10^4 s⁻¹) developed by Wu et al. (2024) to study a CuCrZr alloy.

This study presents an Electro-Magnetic Forming (EMF) bench test designed to calibrate materials models under bending driven and dynamic conditions for ductile metals with high electrical conductivity. The design is inspired by the approach outlined by Golovaschenko and Mamutov (2004), later adapted by Jeanson in her thesis (2016). The EMF bench test is more user-friendly than the Split Hopkinson Pressure Bars and can achieve strain rates around 10^3 s⁻¹, making it suitable for characterizing flat materials with thicknesses ranging from 0.6 mm to 3 mm. The experimental method is coupled with an inverse numerical method, which is a common routine when it comes to material models calibration (Meuwissen et al., 1998 ; Hernandez et al., 2013).

2 Experimental Procedures

The material used in the study is a wrought CuCrZr which properties are presented in the **Table 1**. The CuCrZr alloy is frequently used in the HPP field because of its good balance between electrical conductivity and mechanical strength.

Chemical composition (wt. %)						Density (g/cm ³)	Hardness HB	Electrical Conductivity (m/Ω.mm ²)
Cu	Cr	Zr	Fe	Si	Others			
Bal.	0.77	0.067	0.037	0.017	< 0.2	8,9	118-124	46.8-47.4

Table 1: CuCrZr chemical composition and physical properties

The ambient tensile behaviour of the material is investigated in quasi-static conditions. According to the ASTM E8 recommendations, a strain rate of 2.5×10^{-4} s⁻¹ is employed and the specimen geometry is rectangular with a 50 mm gauge length and a 2 mm thickness.

The EMF bench test is presented **Fig. 1**. A dog bone specimen (thickness 2 mm, width 10mm) is held by clamps onto a coil with a polyimide film electrically insulating the specimen from the coil. The coil is designed with a slot so as the specimen acts as a short-

circuit. The underlying physical principle of the EMF bench test is depicted in the **Fig. 2**. When the electrical energy stored in a capacitor bank is discharged into the EMF set-up, which can be assimilated to an RLC system, the resulting electrical current flowing through the coil and the specimen generates a magnetic field. An electromagnetic force, known as Lorentz force, is consequently produced, deforming the specimen. The equivalent RLC parameters are specified in the Fig.1. The experiment being voltage-controlled, the voltages used in this study are 9 kV, 11 kV and 12 kV (*i.e.* respectively 3.24 kJ, 4.84 kJ and 5.76 kJ of discharged energy). A Rogowski coil measures the current at the coil terminals.

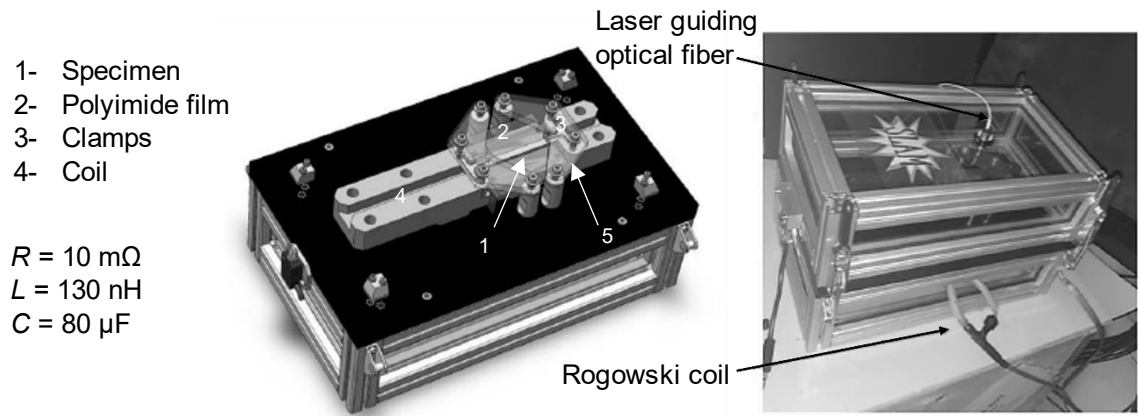


Figure 1: The Electro-Magnetic Forming bench test

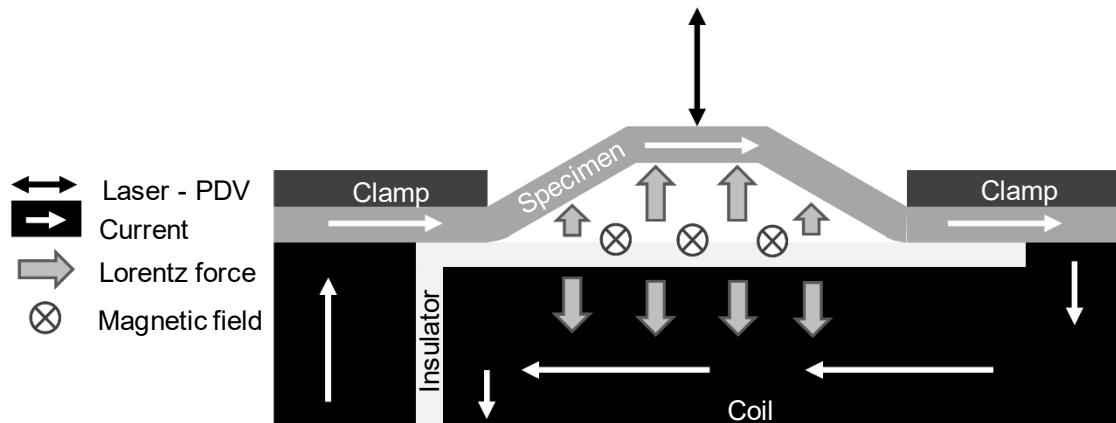


Figure 2: The Electro-Magnetic Forming principle

A Photon Doppler Velocimeter (PDV), with a laser focused on the central point (CP) of the upper gauge length of the specimen, measures the frequency shift of the laser before and after the interaction with the moving specimen. Raw data are processed using Python to calculate the main output which is the velocity of the CP as a function of time. The graph obtained can be divided into two phases:

1. A first phase with an increase of the velocity followed by a steady state. This phase corresponds to the inertial behaviour of the specimen, solely depending on the loading and the specimen characteristics (density, dimensions) ;

2. A second phase with a brief increase in the velocity value followed by a sharp decrease, both accounting for the material response due to the propagation of the plastic deformation at the CP.

The experiment is simulated using the LS-Dyna software with the corrected experimental current applied to model the loading conditions. Material law parameters are then calibrated thanks to an inverse method which compares the simulated and experimental CP velocity data. The mesh consists in five hexaedric elements through the thickness.

The **Fig. 3** proposes a summary of the general methodology developed in this study.

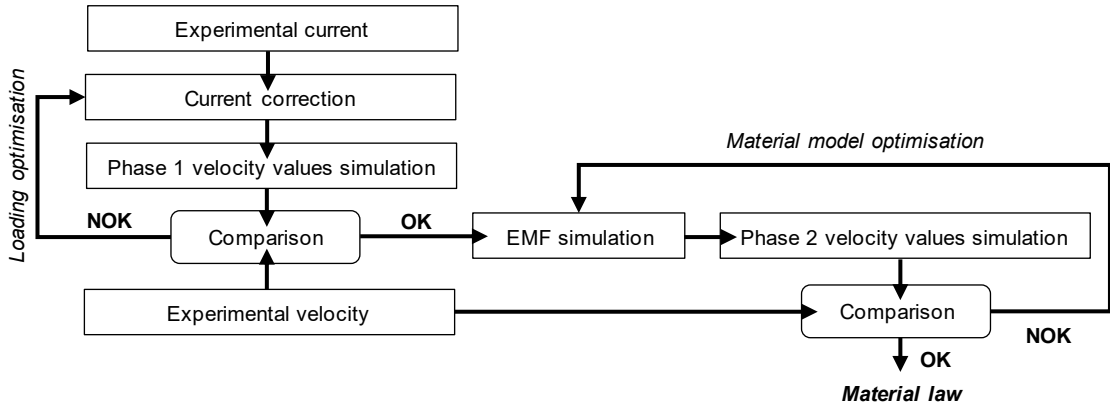


Figure 3: Identification methodology of a dynamic material law using the EMF bench test and an inverse method via LS-Dyna (inspired from Jeanson (2016)). NOK = non-OK

3 Results

3.1 Quasi-Static Tensile Behaviour

The **Fig. 4** shows the experimental results from the tensile tests. A Ludwik model is calibrated with a fairly good fitting. The general Ludwik material law is given by the **Eq. 1**.

$$\sigma = \sigma_0 + k \cdot \varepsilon_p^n \quad (1)$$

with σ the true stress, σ_0 the yield stress, ε_p the plastic strain, k and n materials constants being respectively the coefficient and the exponent of strain-hardening.

Identifying a Ludwik constitutive material law as a suitable model for the quasi-static behaviour is convenient since the Johnson-Cook (JC) model, one of the most commonly used dynamic laws, builds upon it. As expressed in **Eq. 2**, it is a multiplicative model that incorporates the strain rate sensitivity and the thermal softening through two factors applied to the Ludwik quasi-static model (Johnson and Cook (1983)).

$$\sigma = (\sigma_0 + k \cdot \varepsilon_p^n) \left(1 + C \cdot \ln \frac{\dot{\varepsilon}}{\dot{\varepsilon}_0}\right) \left(1 - \left(\frac{T - T_r}{T_m - T_r}\right)^m\right) \quad (2)$$

with $\dot{\epsilon}$ the strain rate, $\dot{\epsilon}_0$ the strain rate at reference state taken as 1 s^{-1} , T the temperature, T_r the room temperature, T_m the melting temperature, C and m material constants being respectively the strain-rate hardening coefficient and the thermal softening exponent.

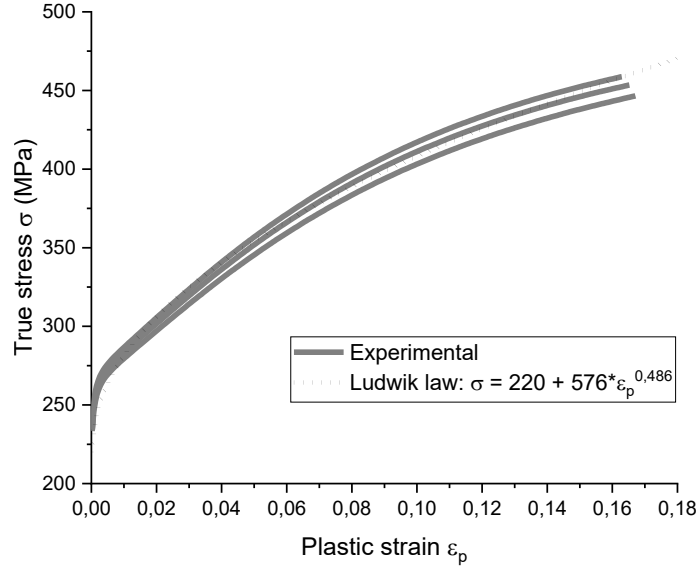


Figure 4: Quasi-static tensile behaviour at ambient temperature of the CuCrZr alloy and identification of the Ludwik model parameters

3.2 Dynamic Testing

The **Fig. 5a** shows the EMF results along with clarifications regarding the different phases of the specimen behaviour during the experiment. The maximal current values measured for each condition are also specified. The **Fig. 5b** illustrates the specimen shape during the different stages of the EMF.

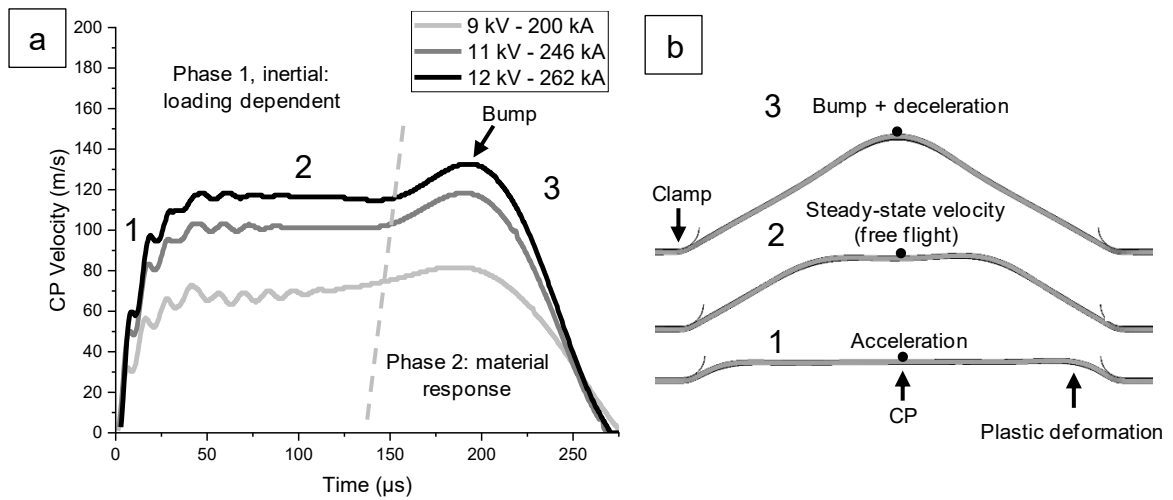


Figure 5: a. CP velocities results after EMF tests. Dashed line separate “phase 1” from “phase 2”; b. Specimen deformation with regards to the 3 stages of the EMF (LS-Dyna)

3.3 Material Model Calibration

The JC model calibrations are presented **Fig. 6**. The thermal softening is not activated ($m = 0$). The JC parameter C is first calibrated at 12 kV and then tested at 9 kV and 11 kV. A good fitting between experimental and simulated data is obtained, leading to a C value of 0.01 suitable to describe the CuCrZr behaviour during an EMF test. A comparison of the CP velocity calculated when $C = 0$, *i.e.* when no strain-rate sensitivity is considered, proves that the Ludwik law from the Fig. 4 is insufficient to describe the material behaviour.

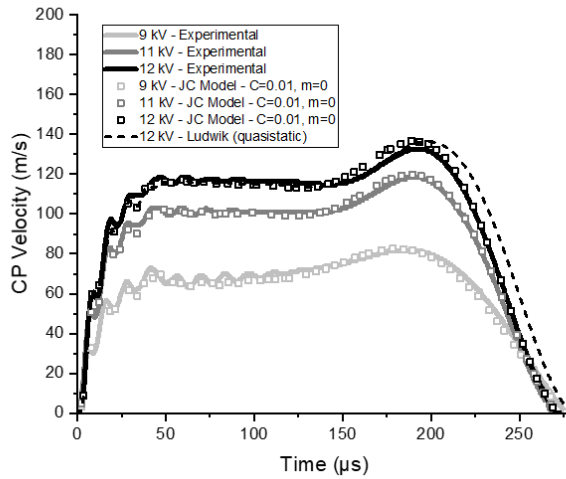


Figure 6: JC model calibration (LS-Dyna) with no thermal softening and comparison with the EMF simulated curve obtained with a Ludwik quasi-static law

Note that this work does not address the non-uniqueness of the calibrated law. Although calibration at three different voltage levels significantly narrows down the range of possible solutions, multiple combinations of JC parameters may still yield an acceptable fit to the EMF data. To address this issue, additional criteria need to be implemented. Ongoing work is also exploring alternative material models, such as the Cowper-Symonds law. Furthermore, the calibration process becomes even more complex when $m > 0$ is considered, as thermal effects arising from plastic strains and the Joule effect -each varying spatially within the specimen- must be accounted for.

3.4 Stress and Strain States

Once a material model is identified, it is possible to analyse the stress and strain states within the specimen during an EMF test. The **Fig. 7** shows the map of the maximum Von Mises stresses reached during a 12 kV test while the **Fig. 8** presents the distribution of plastic strains at the end of the same test. The areas of maximal stresses and plastic strains are located beneath the clamps. Special care must be given to prevent any necking in this area as it would compromise the validity of the experiment. Outside of this critical area, the stress and plastic strain fields are largely homogeneous across the specimen. A slight gradient is observed near

the CP where the bottom face experiences lower stresses and strains. Both results from Fig. 7 and 8 suggest that the EMF bench test is well-suited for characterizing ductile materials whereas brittle materials would likely break prematurely.

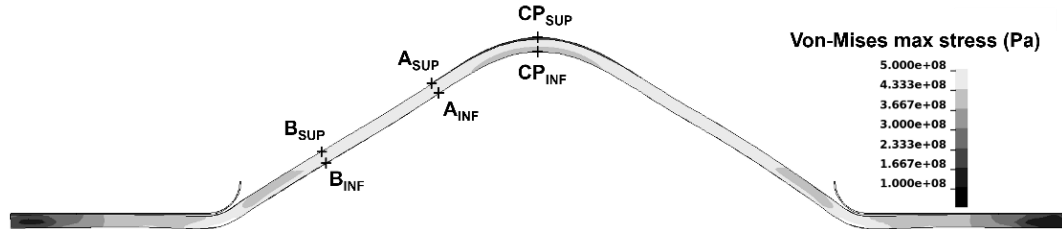


Figure 7: Maximum Von-Mises stresses during a 12 kV EMF test

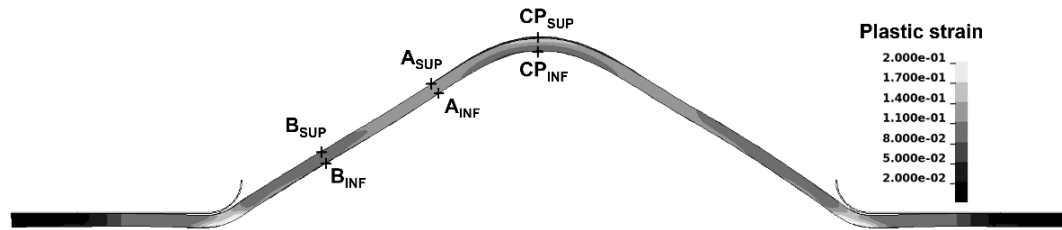


Figure 8: Effective plastic strain after a 12 kV EMF test

The **Fig. 9** shows the variation of strain rates during a 12 kV test. Six positions are analysed as indicated in the Fig. 7 and 8. Due to the propagation of the plastic deformation toward the CP, both the timing and magnitude of peak strain rates depend on the position along the length and the thickness of the sample. The top face undergoes higher strain rates, ranging from 1500 s^{-1} up to 4000 s^{-1} while the bottom face sees lower rates, between 500 s^{-1} and 1500 s^{-1} .

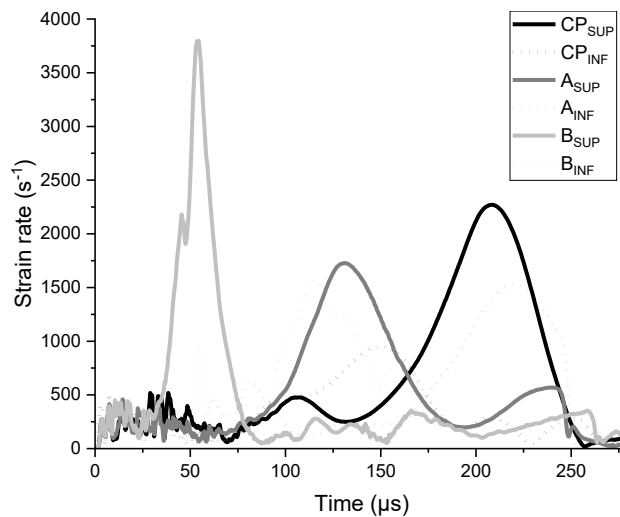


Figure 9: Strain rates variation along the specimen and during a 12 kV EMF test

A key takeaway here is that an EMF test performed at a fixed voltage does not result in a constant strain rate loading but rather in a wide range of strain rate values. Consequently, the material model identified in the previous section should be regarded as an apparent law valid within the $[1-4000] \text{ s}^{-1}$ strain rate range.

4 Conclusions

An experimental method based on electro-magnetic forming has been developed to characterize the dynamic behavior of ductile metallic materials with good electrical conductivity. The calibrated material model is validated within the strain rate range of $[1-4000] \text{ s}^{-1}$. The specific case of the CuCrZr is studied demonstrating the clear benefits from using strain-rate sensitive laws over quasi-static models. Further efforts are required to reduce the number of plausible solutions obtained via inverse identification and to incorporate the thermal effects arising from plastic deformation and the Joule effect.

Acknowledgments

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Calibration of Aluminum Sheets by High-Speed Hydroforming

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Abstract

A new, robust production system for tubular and sheet-based workpieces is presented. It combines the methodologies of high-speed forming with conventional hydroforming and is based on components and drive technology that can be easily controlled by the use of standard industrial technology. In the article the new system and its function is explained and potentials are shown. Moreover, the focus is on the experimental results of production and calibration of high quality, thin-walled sheet metal and cast components for aircraft and power electronics industry. Thin-walled metal parts are difficult to machine due to their low stiffness and potentially high residual stresses. The paper shows an application example of the new forming device to calibrate these parts by high-speed forming. Moreover, the technological properties of the parts are analysed and an outlook on further work is shown.

Keywords

Hydroforming, Calibrating, High-speed forming

1 Introduction

Hydroforming is an important production technology since many years. Complex thin-walled lightweight components based on sheets and tubes for example for automotive applications can be produced in one process step with high accuracy (Neugebauer, 2007).

For smaller production lots this technology is inefficient because of high machine and tool costs. For smaller parts and very big lots the relatively high process time is a problem.

High-speed forming methods were developed to produce sheet metal or tubular parts. Advantages of the high forming speed are among others the extension of process limits and the production of workpieces with high geometrical complexity and accuracy in one process step (Psyk et Risch, 2012).

Very well-known is the electromagnetic forming process that can be used for forming operations on tubular parts and sheet metal parts (Psyk et al, 2011). Niaraki (2019) shows a combination of electromagnetic forming and high-speed hydroforming that extends process limits of electromagnetic forming. For bigger workpieces, the energy of the electromagnetic devices is not sufficient. High-speed forming technology then is driven by explosive materials or the detonation of gas mixtures (Mynors et al. 2002, Weber et al. 2006). Further investigations use plunger-based mechanical drives for forming (Liangliang et al. 2024).

The problem of all high-speed methods is the difficult control and automation of the process by standard control systems. Moreover, tools and production system components are not robust enough for industrial application and bigger production lots. In addition, licenses are required for the use of explosives and there is a risk of misuse. In the case of electromagnetic forming, the service life of the tool coils and the dangers caused by the high voltages are relevant.

Better control possibilities and many advantages seem to have the plunger-based systems because of the more conventional drive technology. This technology is robust and components can be adjusted by mechanical engineers without special training. The problem of these processes is the limit in the production of complex large-surface parts because of the short pressure pulse. In the state-of-the-art, large surface parts are produced by quasi-static processes like deep drawing or conventional hydroforming. Thus, the combination of the methodologies of high-speed forming with conventional hydroforming is a sensible innovation. An interesting possible approach is the combination of quasi-static high pressure sheet metal hydroforming and pneumo-mechanical high-speed forming on one and the same high-speed forming machine (Djakow et al. 2016).

2 Development of a device for high-speed hydroforming

In previous work (Hermes et Böhmer 2024) an idea for a production device combining the methodologies of high-speed forming with conventional hydroforming to a High-Speed Hydroforming Process (HSHF) is shown. Based on these ideas in a first development step a robust production system that is easy to control by standard industrial components was built up in an earlier project (Hermes et al. 2018). The system can be used as well for highly automated production systems for mass production of smaller tubular and sheet metal parts as for the flexible manufacturing of workpieces with larger dimensions. The device is a hydraulic driven high speed pressure generator that can be mounted directly on the hydroforming tool. The system consists of pressure amplifier driven by an oil-based hydraulic system with a seat valve and an accumulator. A pressure of 230 MPa and a pressure rising of 20 ms was reached. The axial feed mechanism for tube hydroforming was

relatively slow, so a dynamic axial feeding in tube hydroforming (THF) was not possible. The device showed a high potential of the method but not sufficient was i.e., the slow speed of the pressure rising, the slow axial feeding for tube hydroforming to reduce thinning and the fragile rubber-based stopper of the pressure amplifier.

The second development step is shown in this paper: The aim is to increase the forming pressure and the forming speed of the system. So, a new device was designed (Figure 1).

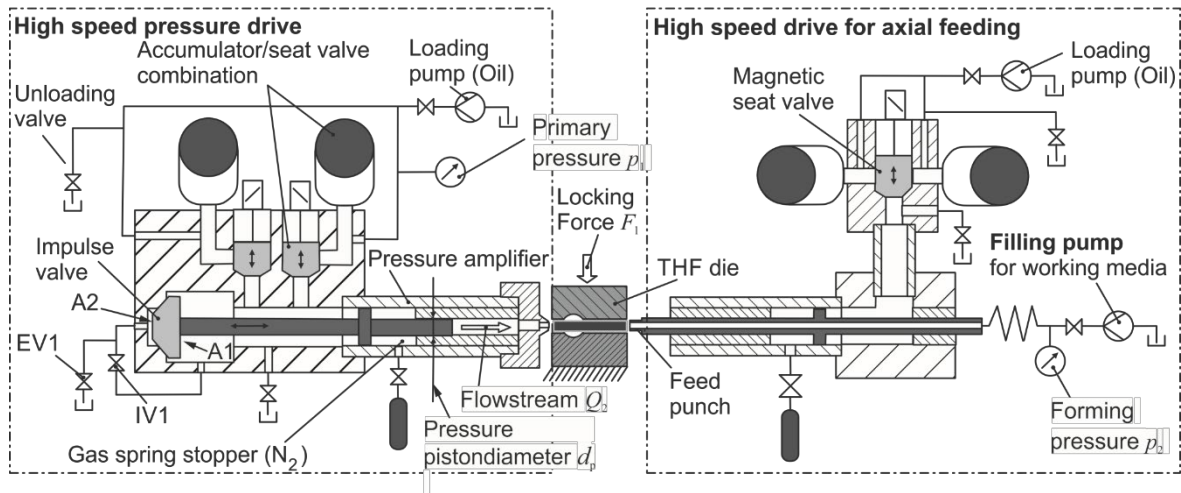


Figure 1: Actual (second) development step of the high-speed hydroforming system

The high-speed pressure drive on the left is a pump system driven by four accumulator seat valve combinations to increase the flow rate of the hydraulic loading pump. To reduce losses in speed caused by the poor opening time of the standard seat valves an impulse valve was designed that opens a big cross section at a short time by the opening of a smaller valve that creates pressure equalization. The stopper is realized by a N_2 -filled chamber that works as an adjustable gas spring and prevents piston seizing and diesel effects. It is possible to influence the maximum forming pressure by adjusting the primary pressure p_1 on the oil side and moreover by changing the pressure amplifier components by reducing the piston diameter ($d_p = \varnothing 40, 30, 20$ mm).

On the right-hand side, the function of the high-speed drive for axial feeding is shown. The punch is driven by an accumulator seat valve system as well. The filling pump for the working media (oil-water emulsion) is filled by a high-pressure pump, so quasi static experiments for comparison are possible.

The function for a tube hydroforming process is as follows: After filling the system and the workpiece with working media, the loading pumps load the pressure accumulator with a defined pressure. The seat valve of the pressure drive can be opened or closed at this step. A controlled signal opens the impulse valve and the seat valve of the axial drive.

Then the tube is axial compressed and radial expanded by the pressure increase of the working media. It is possible to use the impulse valve or the four seat valves to initiate the forming pressure pulse. All valves can be independent activated.

Due to the slow opening time (70 ms) of the standard hydraulic seat valves the impulse valve was developed by the authors. The function of the impulse valve is based on a

difference in surface area between the piston ring surface A1 and the effective piston surface in the end position state of the other side A2. If the pressure chamber is pressurized by the open poppet valves, but the A2 side is depressurized by an open relief valve EV1 and a closed initial valve IV1, the impulse valve is ready to start. If EV1 is closed and then IV 1 opened, the pressure equalization results and the piston is accelerated towards the forming side. This results in a very large cross-sectional area for the oil flow and thus a very fast impulse-like initialization of the process.

In Table 1 the technical parameters of the device are listed.

The THF tool and the axial feeding system were not used for the investigations and results that are described in the following chapter because of the focus on sheet plate calibrating.

Table 1: Technical parameters of the high-speed hydroforming system

Parameter	Value
Hydraulic oil pressure / primary pressure p_1 (pressure drive and axial feeding)	30 MPa
Forming Pressure p_2 (depend on pressure piston diameter d_p)	$d_p = \varnothing 30 \text{ mm} \rightarrow p_2 = 200 \text{ MPa}$ $d_p = \varnothing 25 \text{ mm} \rightarrow p_2 = 300 \text{ MPa}$ $d_p = \varnothing 22 \text{ mm} \rightarrow p_2 = 400 \text{ MPa}$
Maximum Flow stream in the forming Chamber Q_2	$d_p = \varnothing 30 \text{ mm} \rightarrow Q_2 = 45 \text{ dm}^3/\text{s}$
Locking force F_1	1000 kN

Advantages of the presented process and the machine compared to conventional methods with explosive drivers, the possibility of setting the forming pressure and volume precisely via the hydraulic parameters. This means that the required pump volume and forming energy can be precisely adapted to the target component. This results in high precision and energy efficiency.

The sealing of workpieces is a major challenge in order to make the forming pressure usable. The technologies described in this paper are an important topic.

3 Investigation of calibrating sheet metal plates

Focus of the following experiments are the calibrating of sheet plates mate of EN AW 1050. In Figure 2 the experimental set-up is shown.

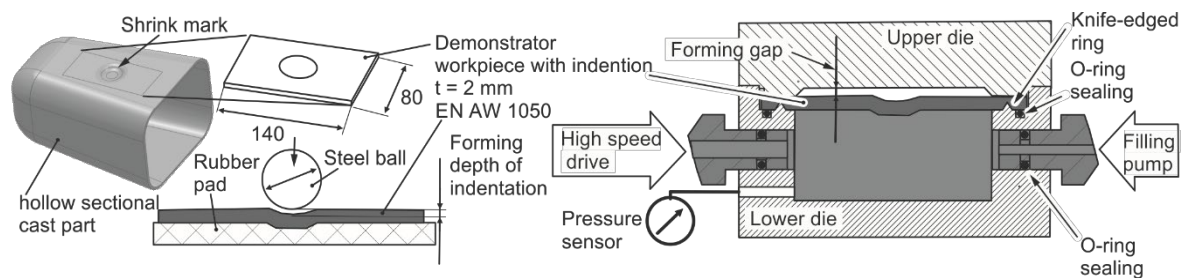


Figure 2: Experimental setup for forming defined indentations in sheet plates and high-speed die for surface calibrating

These investigations are based on an industrial request with the focus on calibrating shrink marks on hollow sectional cast parts made of light metal alloys. The idea is to calibrate the surface by an inner pressure shot initiated by the presented high speed hydroforming system. To analyse the problem, demonstrator workpieces with defined dimensions were used. For this, aluminium plates were formed by a steel ball in the surface of the workpiece as a proxy process for the shrink marks of the cast parts.

After that, the workpieces are mounted in the special high-speed die for surface calibration. The die is mounted in the high-speed hydroforming device and is prefilled from the right-hand side the working media with 0.5 MPa. From the left-hand side, the high-speed drive supplies a pressure pulse in the lower die and the sheet plate is pressed on the surface of the upper die. There was a forming gap of 0.5 mm left and the upper die has several venting holes. The sealing is an important challenge that is solved by several different mechanisms. A circumferential knife-edged ring was machined in the lower die to create a metal-metal sealing. Additionally, a circumferential nitril o-ring was used. A pressure sensor was installed in the lower die and the maximum locking force F_1 of the device with respect to the dimensions of the workpiece allows a maximum static pressure p_2 of about 90 MPa.

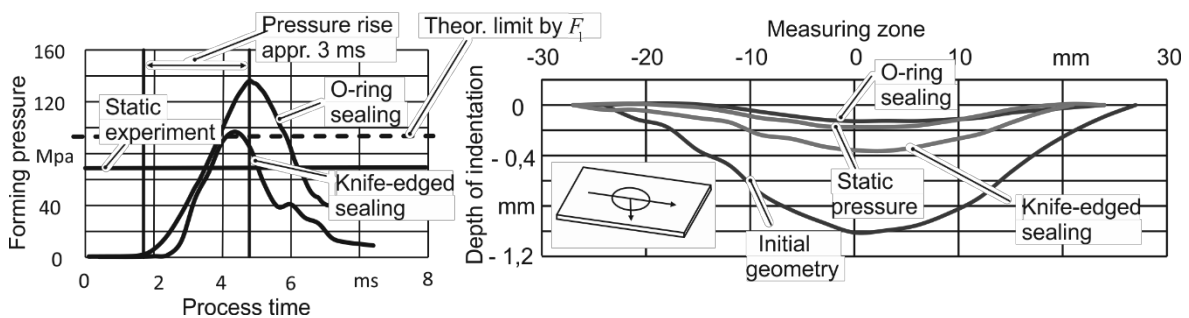


Figure 3: Pressure profiles and flatness measurement of the calibrated workpieces

In Figure 3 results of the experimental investigations are shown. On the left-hand side the measured pressure profiles show that the o-ring and additional knife-edged ring sealed experiments reach 140 MPa maximum pressure in 3 ms. Experiments without o-ring reach a maximum pressure of up to 100 MPa. The theoretical possible pressure $p_2 = 200$ MPa of the system with the mounted pressure piston of $d_p = \varnothing 30$ mm could not be reached in this forming test but the maximum pressure of 90 MPa limit defined by the maximum locking force could of the locking press ($F_1 = 1000$ kN) could be clearly exceeded. The reason for this is on the one hand the pressure loss by seizing of the sealing and on the other hand the inertia effects of the locking system (mass of the upper die and the other moving parts of the locking press). Interesting is the comparison of the results with the static experiment. The pressure was built up by the filling pump of the system, which is able to produce high pressure with a very slow flow rate. The sealing seizing could be detected between 60 and 70 MPa because the positive dynamic effects are not active at low flow rate.

On the right of Figure 3 flatness measurements of the workpieces are shown. The initial geometry formed by the steel ball shows a depth of the indentation of 1.0 mm. After the calibration the best result with the lowest depth less than 0.2 mm was reached by the experiment with the o-ring sealing. The results with the knife-edge ring sealed workpieces

lead to results less than 0.4 mm. Although the maximum pressure in the quasi-static reference-experiment is lower, the indentation is in the field of the o-ring results. Furthermore, the roughness and hardness of the quasi-static and dynamic calibration tests are in equal area, meaning these technological properties are unaffected by the dynamic forming load.

4 Conclusion and Outlook

The paper shows, that it is possible to build a hydraulic driven metal hydroforming machine based on mostly standard components that reach high speed and pressure for the forming of tubes and sheets. The presented device produces pressure pulses driven by the combination of hydraulic accumulators and an impulse valve. For tube hydroforming it is possible to feed the tube length into the forming zone by a synchronously working axial high-speed feeding system.

The aim of the presented experimental investigations in this paper is the calibration of workpieces in geometry dimensions and flatness. Sheet metal plates are calibrated by a pressure shot on the surface to reduce the depth of an indentation. The results show an improved flatness by a dynamic driven pressure shot. The sealing of the workpiece in the die resists the dynamic pressure rising better than a slow quasi-static pressure curve. Even the effective locking force of the press is higher than the theoretical because of the influence of mass inertia and the dynamic pressure generated by the very high flow rate of the forming machine. By this effect, leakages of the sealings lead later to a pressure loss.

The quasi-static reference-experiments with lower pressure have nearly the same flatness than the dynamic ones. This has to be analyzed in future. Reasons may be rebound effects that influence the dynamic results. The use of quasi-static pressure is no option for the industrial application of the process because of an unacceptable increasing of the locking force when extrapolating to bigger sized workpieces.

This work shows the beginning of research with the new device. In the field of workpieces geometry and material properties a bigger variety has to be analyzed in future. Theoretical work has to be done to generate basics for process layout and simulation. Dynamic effects, influence on materials behavior and machine stiffness have to be included in the investigations.

Simultaneously the transfer for an industrial application is planned.

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Automatic Mesh Adaptation for Magnetic Pulse Forming Simulations

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Abstract

Multiphysics coupling simulation for the electromagnetic modelling of pulse forming technology comes with several challenges: Large deformations with high strain rates, fast impact conditions, multiple phenomena with distinctive proper time scales, complex geometries, etc. In order to accurately represent the electromagnetic phenomena together with the thermomechanical effects in a reliable manner a key step is the ability to measure the numerical error intrinsic to the discretization scheme (typically within the finite element framework) and use of automatic mesh adaptation techniques. In this work, an a posteriori error estimator-based technique adapted to the electromagnetic phenomena is presented together with an anisotropic automatic mesh adaptation technique. These methods are rooted in the understanding of the underlying physical phenomena, bringing as consequence: ease of the modelling stage as numerical results are less dependent on initial mesh settings. Accelerated simulation times as computation resources are adapted automatically to increase accuracy where needed and decrease mesh density otherwise. All of this in a natively distributed-memory parallel framework which can be scaled for large problems within the simulation software FORGE®.

Keywords

Metal Forming, Finite element method (FEM)

1 Introduction

Magnetic Pulse Forming (MPF), a key high-speed forming process, enables deforming metallic components via pulsed Lorentz forces. Its contactless nature enables shaping and joining of materials often unfeasible with conventional methods (Psyk et al., 2011). MPF's ability to produce complex geometries and minimize spring back makes it especially appealing in automotive, aerospace, and electrical applications.

Simulating MPF can be computationally demanding due to the transient, nonlinear, and tightly coupled electromagnetic-mechanical interactions involved. High mesh resolution is essential near coils and workpiece surfaces, where electromagnetic fields vary sharply due to skin and proximity effects, while coarser meshes suffice in less active regions, such as surrounding air. This work presents a fully automatic anisotropic mesh adaptation strategy for finite element simulations of the MPF. By combining a posteriori error estimation with field-sensitive metric tensor construction, the method achieves high accuracy and reduced computational cost, adapting seamlessly to physical field variations and geometric complexity.

2 Error Estimators

A key component of the mesh adaptation strategy is the construction of an a posteriori error estimator that can guide local refinement or coarsening of the mesh. In this study, we adopt a residual minimization-based gradient recovery strategy to evaluate the discretization error of vector fields involved in electromagnetic simulations, such as the magnetic field or the electric field (Garcia et al., 2022). The methodology is established by formulating a global minimization problem:

$$\min \|\vec{F}_{P_1} - \vec{F}_{P_0}\|^2 \quad (1)$$

Which in variational or weak form reads:

$$\langle \vec{\psi}, \vec{F}_{P_1} - \vec{F}_{P_0} \rangle = 0 \quad (2)$$

where $\vec{\psi}$ represents the basis function of the Sobolev space $H(curl)$ used for electromagnetic fields. $\vec{F}_{P_1}$ represents an enriched (or recovered) representation of the base vector field $\vec{F}_{P_0}$.

Once the field has been recovered on the higher precision representation the error estimate is given by

$$\epsilon = \|\vec{X}_{P_1}(x_G) - \vec{X}_{P_0}(x_G)\|_{L^2(\Omega)} \quad (3)$$

where x_G represents the Gauss points coordinates of each element (e). Then, the error normalisation is given by

$$\|\epsilon\| = \frac{\epsilon_{L^2(\Omega_e)}}{\epsilon_{L^2(\Omega)}} \quad (4)$$

where $\epsilon_{L^2(\Omega_e)}$ and $\epsilon_{L^2(\Omega)}$ are the local and global error respectively. It is worth to point out that local defined error estimates generally do not take into consideration the directional effect of the anisotropy of the solution. However, a globally defined error estimator, as the one defined here, contains additional information about the solution, which allows to generate an anisotropic mesh adaptation approach.

3 Anisotropic Metric

The metric aiming to capture the anisotropic behaviour of the physical phenomenon is calculated from the Jacobian of a vector field describing it. This is achieved using the same Galerkin recovery or residual minimization recovery procedure applied in the error estimator. Let us consider a vector field $F: \mathbb{R}^3 \rightarrow \mathbb{R}^3$.

Depending on the geometrical nature of the field, two procedures can be envisioned:

3.1 Jacobian of a vector field

$$J(\vec{F}) = \vec{\nabla}(\vec{F})^T \vec{\nabla}(\vec{F}) \quad (5)$$

The procedure to calculate the metric from the Jacobian of F involves a sequence of smoothing and gradient calculation. Defining $\vec{F}_{P_1}$ at the element nodes, the gradient at the Gauss points is calculated as:

$$\vec{\nabla}\vec{F}_{P_0}(x_G) = \sum_{i=1}^{\text{nodes}} \vec{\nabla}\varphi_i \vec{F}_{P_{1i}} \quad (6)$$

where φ represents the basis functions at the nodes. Then, the gradient at the element nodes is calculated by the recovery approach:

$$\langle \varphi, \vec{\nabla}\vec{F} - \vec{\nabla}\vec{F}_{P_0} \rangle = 0 \quad (7)$$

3.2 Hessian from a scalar field

$$H(g) = \vec{\nabla}(\vec{\nabla}(g)) \quad (8)$$

The procedure to obtain the hessian from a scalar quantity is similar to the Jacobian procedure but the smoothing step has to be performed an additional time, as the gradient of the scalar field forms a first vector field which needs to be smoothed out.

3.3 Metric and normalization

The metric tensor forms the basis of the adaptive algorithm, providing the size, shape, and orientation of the new elements. The discretised metric is given as a function of the Jacobian tensor field at each node of the mesh: $M_i = M(J(x_i))$. To build this metric, first,

the Jacobian is computed following the procedure above. Symmetry and positive definiteness are guaranteed by construction ($|T|_i = T_i^T T_i$). The metric is then calculated by a filtering on the Jacobian matrix values, $\Lambda_i = \text{eigvals}(J_i)$ and $R_i = \text{eigvecs}(J_i)$. The eigenvalues link the required mesh size (h) along the eigenvector's directions with the error estimator. To avoid unrealistic metric specifications, such as infinite mesh sizes in an element, the eigenvalue is restrained applying an error tolerance (ϵ_0) and bounding the minimum (h_{min}) and maximum (h_{max}) mesh size, as follows:

$$\widetilde{\lambda}_k = \min \left(\max \left(\frac{\sqrt{\lambda_k}}{\epsilon_0}, \frac{1}{h_{max}^2} \right), \frac{1}{h_{min}^2} \right) \quad (9)$$

Finally, the metric is constructed as follows (for more details refer to):

$$M_i = R_i \widetilde{\Lambda}_i R_i^T \quad \text{with} \quad \widetilde{\Lambda}_i = \text{diag}(\widetilde{\lambda}_k) \quad (10)$$

4 Numerical Validation

Two study cases are used as demonstration examples. The first one is a well-established ring expansion semi-analytical case (Alves, 2016). The second one consists of a pancake-like configuration.

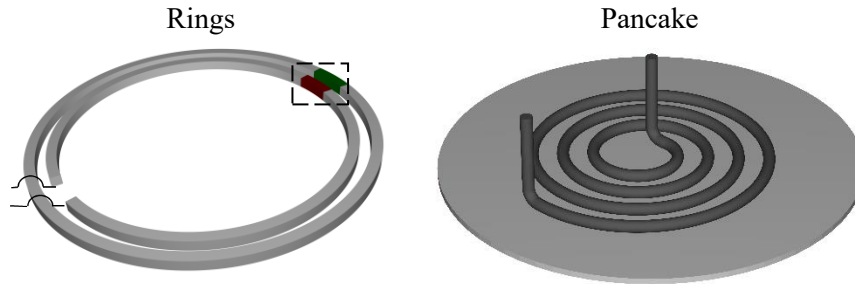


Figure 1: Magnetic Pulse Forming example cases: (left) Ring Expansion, (right) Bulging of a pancake configuration.

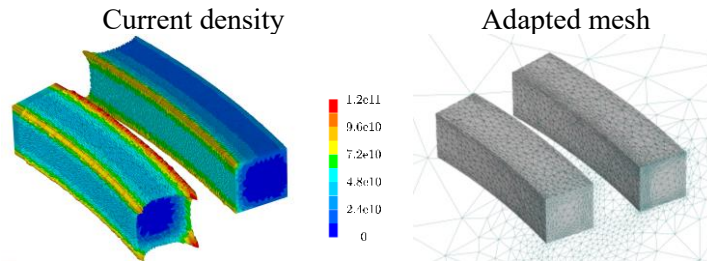


Figure 2: MPF case: Rings geometry, current density distribution [units=A/m²] and adapted mesh

Through the mesh adaptation procedure, the mesh is effectively refined and adjusted to capture the intricate electromagnetic phenomena. The conductive components, in particular, exhibit mesh adaptation in alignment with the anisotropy of the current density, effectively reducing the error evolution during computation, as illustrated in **Error!**

Reference source not found. while simultaneously reducing the overall computational time as seen in Table 1.

#Degrees of freedom		CPU time	
Initial mesh	Adapted mesh	Initial mesh	Adapted mesh
320.4 k	140 k	37 min	20 min

Table 1: CPU times for the ring expansion case

The objective of the investigation of the flat pancake configuration is to validate which metric tensor is able to capture the complexity of the magnetic field distribution in spite of the high aspect ratio of the geometry. A discharge pulsed signal with a natural frequency of 162kHz is used. The magnetic field ($\vec{H}$) is used to construct the metrics. Fig. 3 shows the comparison of the meshes and error distribution for both metrics and the initial mesh. It is observed that the Hessian metric achieves a smaller final error distribution than Jacobian. However, the difference between the degrees of freedom, shown in Table 2, is quite high in both of them, leading the metric Jacobian to require less degrees of freedom to achieve a mesh with an acceptable diminution of the error.

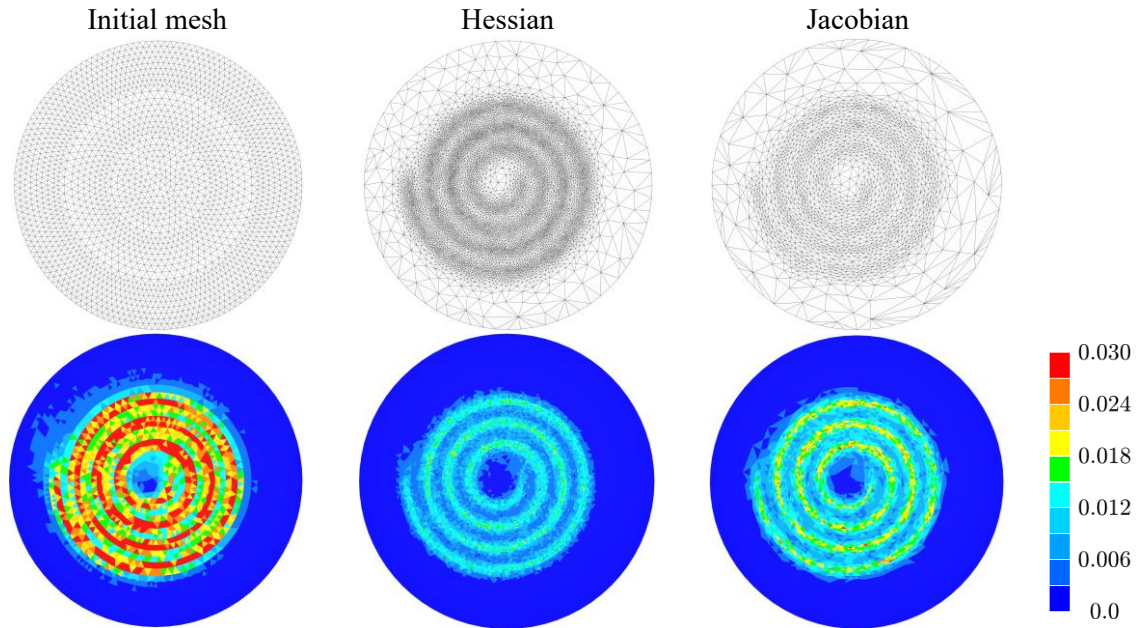


Figure 3: Magnetic pulse forming mesh and error distribution (in percentage) results for the pancake configuration.

Mesh	Initial	Hessian	Jacobian
Nodes	67805	79735	54292
elements	438928	465389	309727
Error [%]	1.7×10^{-7}	0.9×10^{-7}	1.0×10^{-7}

Table 2: Mesh size and error estimator.

Applying a cutting-plane view to the resulting mesh traversing the central plain of the circular geometry it can be seen in Fig. 4 how the methodology enables to automatically set fine mesh on the coils surface and within the workpiece near the coil. Both regions being the ones where the magnetic field reaches its peak and changes faster. The methodology will simultaneously keep a coarse mesh in regions of the air far-away such as to reduce the overall computational cost.

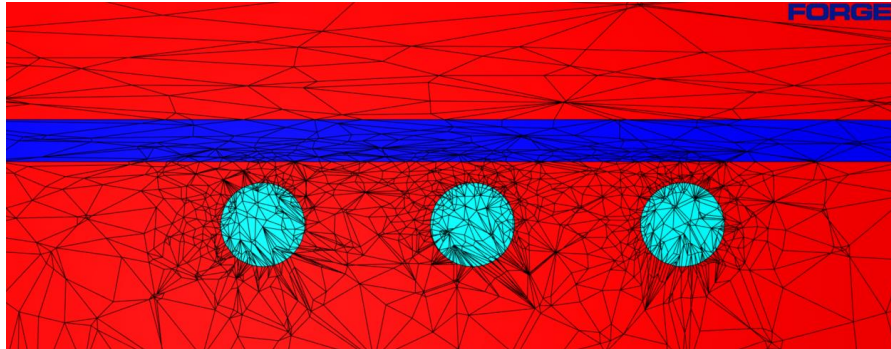


Figure 4: Cut-plane view of mesh density on the air, coil and workpiece.

5 Conclusion

This work presents an efficient and robust mesh adaptation strategy that is both generic in formulation and efficient in computational terms to tackle the complexity arising from the magnetic pulse forming simulations. By employing a posteriori error estimation and a Jacobian-based metric tensor, the methodology enables to (a) capture high-gradient electromagnetic effects (e.g., skin effect) with fewer elements (b) align mesh elements with field directions, improving accuracy (c) reduce computational cost without sacrificing fidelity.

While both Hessian and Jacobian metrics are viable, the Jacobian-based approach shows superior anisotropy representation and error reduction for vector field-driven problems. Future work will explore the extending the method to other electromagnetic forming processes such as crimping and welding.

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Finite element modelling and simulation of the multiscale crystalline structure of aluminium deposit produced by helium cold spray additive manufacturing

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Abstract

This study addresses an experimental analysis with a multiscale modelling of a pure Aluminium deposit produced by helium cold spraying. The grain morphology and the crystallographic texture of the deposit revealed a significant grain refinement along with a formation of strong metallurgical bonding. Complementing the experiments, a two-part multiscale modelling framework is developed, integrating a simulation of the particle upon impact and a crystal plasticity finite element model for analysing the consolidation and deformation mechanisms across multiple scales. Using Johnson-Cook equations and coupled Eulerian-Lagrangian approach, the particle impact model effectively captures high-velocity impact behaviour, deformation mechanisms, and thermal interactions at the particle-substrate interface. The influence of intergranular interactions and phase heterogeneities on the overall mechanical response, at the mesoscale level, is investigated. The crystal plasticity-based approach can simulate the behaviour of individual grains within a representative volume element (RVE), with a prediction of the anisotropic mechanical properties and stress-strain relationships between various grains. The results pave the way for a better understanding of the microstructural evolution and mechanical behaviour of Aluminium deposits produced by helium cold spraying.

Keywords

Cold Spray, Microstructure, Modelling, Aluminium, Additive Manufacturing, Crystal Plasticity Simulation.

1 Introduction

Additive manufacturing (AM) has emerged as a transformative technology across various industries, offering the potential for rapid prototyping, complex geometrical freedom, and reduced material waste. Among AM techniques, cold gas dynamic spray (CGDS), also known as cold spraying, stands out due to its unique capability of depositing metals, composites, and ceramics at relatively low temperatures. This process involves the acceleration of solid particles to supersonic velocities, allowing them to adhere and consolidate upon impact with a substrate, thereby forming dense, well-bonded deposits [1]. Such characteristics make CGDS an attractive method for producing coatings, repairing components, and fabricating parts in diverse fields ranging from aerospace to biomedical applications. An intriguing advancement in cold spraying is its ability to achieve a high strain rate self-consolidation, where material bonding occurs through severe plastic deformation induced by high-velocity particle impacts [2]. Aluminium and its alloys, renowned for their high strength-to-weight ratio, corrosion resistance, and thermal conductivity, are widely used in aerospace, automotive, and defence industries. However, the fabrication of Aluminium components via conventional manufacturing processes often poses challenges related to porosity, residual stresses, and microstructural inhomogeneity. CGDS offers a promising solution to these challenges through high-strain-rate self-consolidation, which facilitates the formation of dense, defect-free structures with tailored mechanical properties [3].

This study explores the microstructural evolution and modelling of Aluminium deposits produced via high strain rate self-consolidation using helium CGDS. The paper is structured as follows. First, a review of the principles of CGDS and its application to Aluminium is presented, highlighting the challenges and opportunities. Next, a series of characterization experiments are performed on helium cold spray-fabricated pure Aluminium sample. Following microstructure characterization, particle impact simulations are conducted to detail the complex features of the particle-substrate interface, particularly in its vicinity. By analysing the interplay between particle deformation, interfacial bonding, and thermal effects, we aim to elucidate the underlying mechanisms driving the formation of dense and cohesive deposits. At the same time, in order to uncover the intrinsic mechanisms of the observed behaviours, a crystal plasticity finite element model is developed to provide insights into the interplay between grains and strain rate effects. It is worth noting that the RVE model is derived from the real EBSD microstructure of the CS deposit through image processing. Crystal plasticity modelling helps in elucidating the localized deformation behaviours and stress-strain responses under various loading conditions, bridging the gap between process parameters and material performance. This work emphasizes the viability of CGDS as an advanced manufacturing technique for producing high-quality Aluminium components with tailored properties.

2 Material and fabrication

The feedstock material used in this study is a pure aluminium powder (purity $\geq 99.5\%$) commercialized by Toyal company. The granulometry ranges in between 15 and 45 μm . The high purity of aluminium ensures minimal contamination and promotes a better bonding during deposition, which is critical for the high strain rate self-consolidation process inherent to cold gas dynamic spray additive manufacturing. The deposition of the aluminium powder was carried out using a closed-loop cold spray system [4]. The substrate is Al 7075 substrate. Helium was used as the carrier gas to ensure optimal velocity for the cold spray deposition. Additionally, the process gas was preheated to 350 $^{\circ}\text{C}$ and delivered at a pressure of 28 bar.

3 Experimental procedure and results

The cold spray samples were sectioned for microstructural and mechanical analysis. The metallographic preparation included mounting, grinding, polishing, and etching using Keller's reagent [5]. Scanning electron microscopy (SEM) and optical microscopy (OM) are used to examine the microstructure, focusing on particle deformation and bonding characteristics.

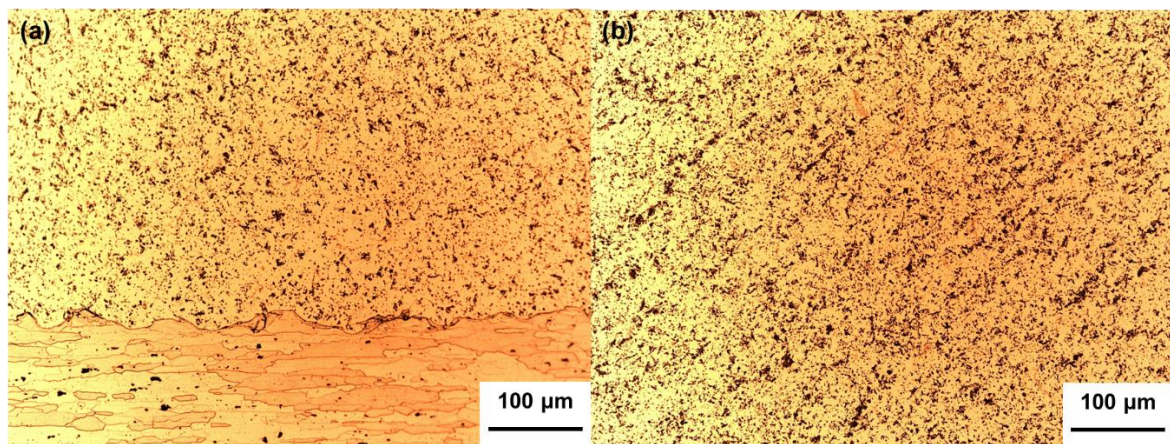


Figure 1: OM image of the helium CSed Aluminium deposit: (a) cross section; (b) top section.

The microstructural analysis of the aluminium deposit shows distinct features associated with the processing conditions. The as-deposited samples exhibited a dense microstructure with a very low porosity ($<2\%$), as observed through OM and scanning electron microscopy SEM. High-resolution SEM micrographs revealed the presence of localized bonding interfaces between the individual particles, characterized by flattened and elongated grain structures indicative of severe plastic deformation. These observations revealed a significant plastic deformation of the aluminium particles upon impact. The particle interfaces were characterized in the as-deposited state using Electron backscatter

diffraction (EBSD). The results show equiaxed grains with an average size of 2 μm , indicating substantial grain refinement compared to the original feedstock material.

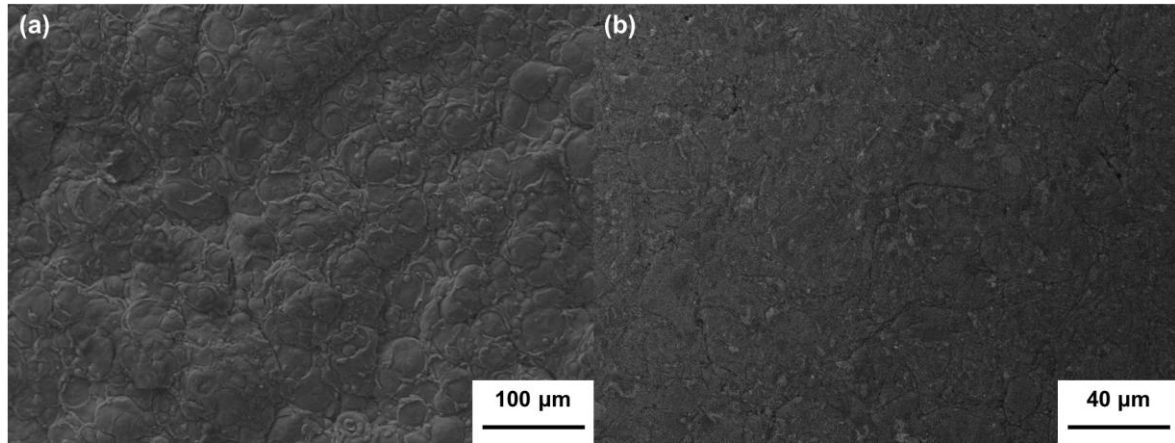


Figure 2: SEM image of CSed Aluminium deposit: (a) top section without polishing; (b) top section with polishing.

The interface between the Aluminium deposit and the substrate displays a transition zone with recrystallized or deformed grains as shown in Figure 3. The grain size varies from large to fine across the interface. The EBSD analysis provides insights into the understanding of the grain structure and recrystallization phenomena. Figure. 3 shows the cross sectional EBSD image which exhibits equiaxed or elongated grains. It shows significant grain refinement compared to the feedstock powder.

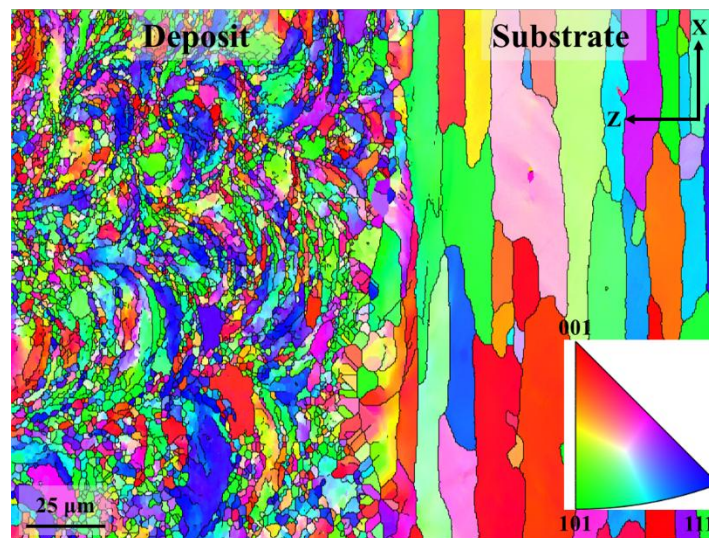


Figure 3: EBSD image of the helium CSed Aluminium deposit: cross section of the bonded interface. (Z presents building direction, X presents nozzle traverse direction)

The average grain size was reduced from approximately 25 μm in the feedstock to sub-micron levels ($<1 \mu\text{m}$) within the deposit. The formation of equiaxed grains at the immediate

vicinity of the interface reveals a thermal recrystallization process upon impact, due to plastic work produced by the several plastic deformations. The fine but elongated grains are attributed to the dynamic recrystallization due to the high strain rates deformation at the interface upon impact.

4 Modelling of particle impact behaviour and results

The finite element modelling of the cold gas dynamic spray process provided further insights into the high strain rate self-consolidation mechanism [6]. A model incorporating particle impact dynamics, plastic deformation, thermal evolution, and inter-particle bonding mechanisms is of interest. In this study, a finite element model is performed to simulate the stress, strain and temperature distribution during the particle impact process based on coupled Eulerian-Lagrangian approach. The uses of the Johnson-Cook plasticity constitutive law allow for accounting for the strain hardening, the strain rate sensitivity, and the thermal softening. The detailed J-C equations and parameters of pure Aluminium can be seen in the literature [7]. The impact velocity of particle is 700 m/s. A mesh size of 0.5 μm was employed for both the particle and the refined region of the substrate adjacent to the interface to ensure simulation accuracy.

The residual stress distribution is also investigated by analysing the compressive stresses near the substrate-deposit interface due to collision. The spatial and dimensional agreement between the simulated high-strain/high-temperature band and the experimentally observed dynamically recrystallized belt validates the model's ability to couple deformation and microstructure evolution during cold spray single-particle deposition. The modelling results also highlighted the critical role of thermal softening and adiabatic heating in facilitating particle bonding. A combination of localized thermal softening and adiabatic shear at the particle interfaces facilitated the metallurgical bonding.

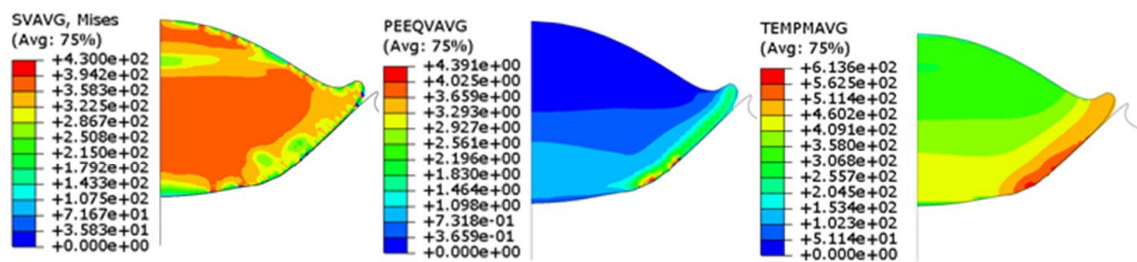


Figure 4: Simulation of single particle impact behaviour: distribution of stress, strain, and temperature, respectively.

5 Modelling of microstructural evolution and results

The modelling of microstructural evolution of due to the cold sprayed deposition depicts the inhomogeneous deformation mechanisms resulting from high strain rate self-consolidation. Figure. 5 shows the RVE modelling using the crystal plasticity framework.

From the EBSD map in Figure 5 (a) and the microstructural experiments mentioned above, it can be seen that the Aluminium deposit has a bimodal polycrystalline microstructure consisting of recrystallized ultrafine grains (UFG) and deformed coarse grains (CG). The significant size disparity between CGs and UFGs requires a multiscale modelling approach. To this end, it is essential to distinguish these domains when constructing the microstructural RVEs and conducting the simulations. We adopt an image-based RVE model, explicitly representing these domains as polycrystalline ensembles, incorporating particle boundary regions (UFG domain) as shown in Figure 5 (c). Figure 5 (d) shows the final RVE model for the Aluminium prepared by cold spraying, thus combined with the crystal plasticity equation. We analysed the impact of the topology on the plastic behaviour of the heterogeneous structure using a 2D representative volume elements ($90\ \mu\text{m} \times 60\ \mu\text{m}$) with an element size of $0.35\ \mu\text{m}$. For the FEM simulation, quadrilateral quadratic plane strain elements with reduced integration (CPE8R) were utilized in combination with a user-defined material subroutine. Periodic boundary conditions were applied in the RVE model to simulate an upward tensile loading scenario.

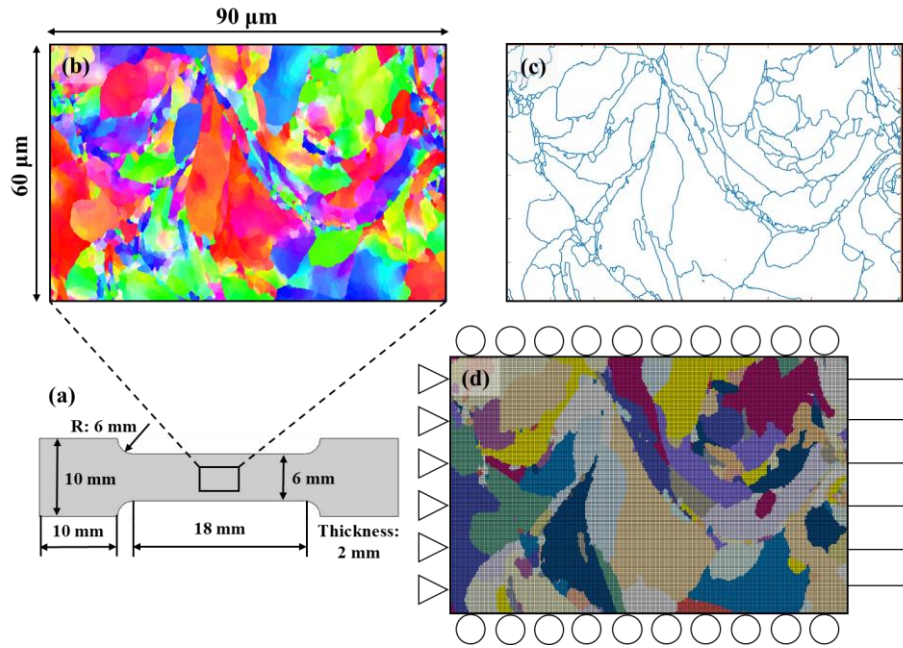


Figure 5: Process of configuration reconstruction: (a) sketch of tensile test sample with selected region for EBSD characterization; (b) EBSD IPF map; (c) grain boundary map; (d) EBSD-based RVE model with load condition.

The identification of elastic and plastic properties of materials from indentation tests has been extensively studied in the literature [8]. The classical dislocation density-based crystal plasticity theory equations can be found in the literature [9]. To enhance computational efficiency, the parameter identification for the constitutive model was conducted using the lsqnonlin algorithm in MATLAB. The target parameters were first assembled into a vector and initialized with estimated values. ABAQUS was then invoked from within the MATLAB working environment to perform simulations, and the parameters

were iteratively optimized using a nonlinear least squares approach. Once the deviation fell within an acceptable tolerance range between the experiment and simulation, the corresponding parameters were regarded as the optimal solution. The identified parameter values are presented in Table 1.

Table 1: Fitted crystal plasticity parameters of cold sprayed Aluminium.

Parameter	Variable	Unit
Reference slip rate	1435	1/s
The first obstacle profile	0.7	-
The second obstacle profile	1.3	-
Activation energy	5.00E-19	J
Lattice Boltzmann constant	1.38E-23	-
Absolute temperature	298	K
Lattice friction	37	MPa
Geometric factor in Taylor's strength	0.25	-
Shear modulus	29	GPa
Burgers vector per slip system	0.000286	μm
SSD density multiplication rate	200	1/ μm
SSD annihilation rate	1000	-
Initial dislocation density per slip system	1.4	μm^{-2}

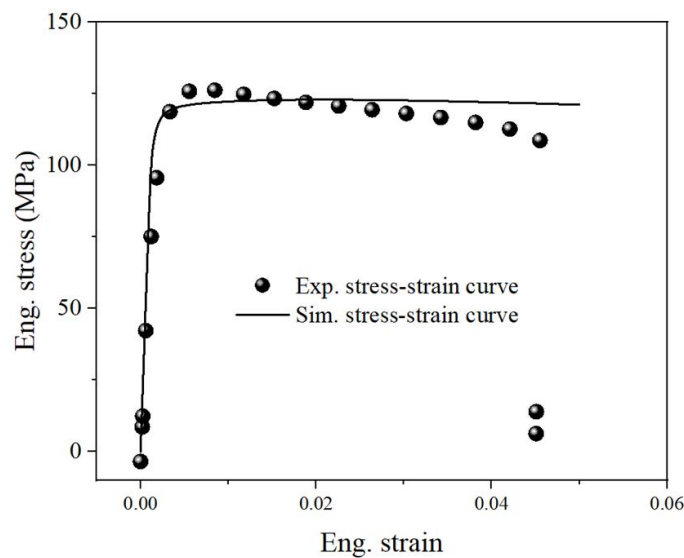


Figure 6: Comparison of tensile curve between simulation and experiment.

Virtual tensile deformation simulations of the cold-sprayed Aluminium crystals were conducted using the commercial finite element software ABAQUS/Standard, with single crystal plasticity models as material constitutive framework. Each integration point requires assigning a specific number of grains to represent a polycrystalline coarse or fine-grain material point, with each grain defined by its orientation and size. A macroscopic engineering strain of 0.05 was applied in for the virtual test. Figure 6 presents the comparison of experiment and simulation curve. The curve obtained from crystal plasticity parameter fitting aligns well with the experimental results. While the experimental curve shows a downward trend, the simulation does not replicate this behaviour. This difference may result from the simulation not accounting for process-induced defects—such as porosity, variations in inter-particle bonding, and residual stresses—that are typical of cold-sprayed materials and can impact their mechanical response.

Figures 7 and 8 present the computed stress and strain distributions based on a real microstructure-based representative volume element model using the crystal plasticity constitutive law. The results reveal significant inhomogeneity in both stress and strain fields. Notably, high stress concentrations—up to 433 MPa—are observed near particle impact zones characterized by ultrafine grains. These stresses tend to localize at particle boundaries, particularly in regions with smaller grains. The areas of high stress correspond to localized strain bands, which are critical drivers of microstructural evolution. As shown in Figure 8, the strain distribution exhibits clear localization along grain boundaries and within grains possessing specific crystallographic orientations, underscoring the influence of crystal anisotropy on the deformation behaviour. In coarse-grained regions, strain localization leads to the formation of distinct 45-degree shear bands, indicative of plastic instability. Conversely, lower strain levels are observed in mechanically harder or more constrained regions. These findings suggest that microstructures promoting stress and strain localization—such as weak inter-particle bonding or pronounced grain boundary activity—may facilitate failure initiation at particle interfaces. Overall, the non-uniform strain distribution highlights the critical role of microstructural features, including grain orientation, grain boundary interactions, and particle-particle interfaces, which are intrinsic to the cold spray additive manufacturing process.

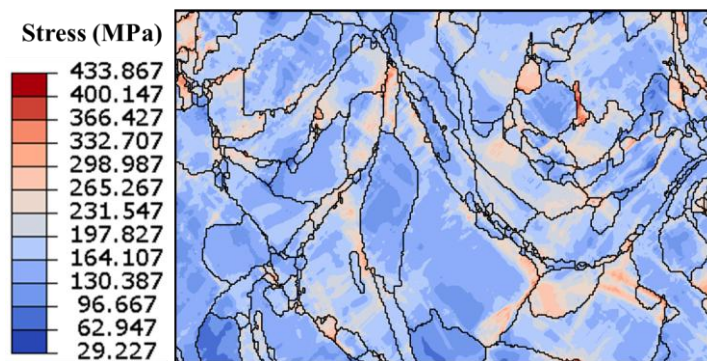


Figure 7: Contours of stress distribution.

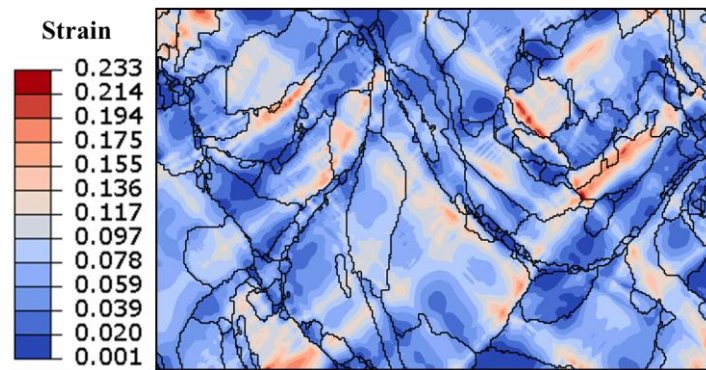


Figure 8: *Contours of logarithmic strain distribution.*

6 Conclusion

In this study, the microstructural analysis and multiscale modelling of an Aluminium deposit produced by the high-strain-rate self-consolidation in cold gas dynamic spray additive manufacturing were investigated. The combination of experimental results and modelling emphasizes the advantage of using a crystal plasticity model applied to a RVE microstructure which was built from the EBSD characterization of the deposit. The microstructure consists of fine equiaxed recrystallized grains mixed with fine elongated grains attributed to a dynamic recrystallization. This computational approach helps for understanding the stress and strain development within the fine microstructure generated by the high-speed collision of cold spraying. The results revealed stress concentrations at the boundaries between particles in the regions of fine grains due to localized strain bands. The strain distribution maps evidenced strain localization along the grain boundaries and within the grains with specific crystallographic orientations, confirming the role of crystal anisotropy in producing the deformation patterns.

Acknowledgement

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Numerical Investigation of Integrating Electromagnetic Forming with Supercritical Fluid-Assisted Polymer Injection Molding

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Abstract

Lightweight sheet metal-polymer components with micro- to nanocellular structures offer a unique combination of structural performance and functional properties. Using conventional manufacturing approaches requires complex and costly multi-stage manufacturing routes to realize such components. Moreover, achieving polymer foams with fine morphology is challenging due to the limitations in precise control over the current foaming technique. To address this limitation, this study explores the novel concept of integrating electromagnetic forming with supercritical fluid (SCF)-assisted polymer injection molding, called the electromagnetic forming injection foaming (EFIF) process. To investigate the feasibility of this integration, a combined numerical-analytical framework is developed using Abaqus finite element (FE) software and cell nucleation analytical models. Simulations are conducted and compared with experimental results to evaluate the influence of this integration on the electromagnetic stage. The deformation rates extracted from the Abaqus simulations are incorporated into a set of modified analytical models based on classical nucleation theory to estimate the pressure drop dynamics. This study provides a more accurate representation of cell formation within the EFIF process by coupling these pressure variations with the governing equations for nucleation and growth. The results demonstrate the feasibility of this integration and its potential to enable precise control over the SCF-assisted foaming process.

Keywords

Electromagnetic Forming, Supercritical Fluid-assisted Foaming, Hybrid manufacturing, Micro/Nanocellular Structures, Analytical-Numerical Modeling

1 Introduction

Multi-material design is typically employed to combine different material systems with complementary properties, resulting in enhanced performance beyond what a single material can offer (Tekkaya et al., 2014). One such multi-material approach is the combination of sheet metal and plastic, which can significantly reduce overall mass, part count, and production effort. This hybrid structure improves properties such as energy absorption and dimensional stability and allows for greater design flexibility and functional integration. Despite their advantages, the broader adoption of metal-polymer hybrids, especially in cost-sensitive sectors like automotive, remains limited due to long cycle times and high production costs. To overcome these challenges, integrated manufacturing processes such as polymer injection forming (PIF) present a viable solution (Hussain et al., 2012). In PIF, a metal blank is placed into a mold and shaped by the pressure of molten polymer injected using a standard injection molding machine, enabling cost-efficient and reliable production of sheet metal-plastic/composite components in one single operation (Farahani et al., 2018). Foaming can introduce new functional properties to multi-material parts by enhancing insulation, damping, and energy absorption capabilities. Supercritical fluid (SCF) assisted injection molding is an advanced technique that uses supercritical CO₂ or N₂ to produce high-performance foamed plastic parts. Once the polymer reaches a molten state in the injection molding barrel, SCF is injected and mixed with the melt to form a single-phase solution. This mixture is then injected into a mold at atmospheric pressure, causing a rapid pressure drop that triggers cell nucleation (Pirani et al., 2025). Key advantages of this method include reduced part weight, lower polymer melt viscosity, minimized shrinkage, and faster cooling times. To take advantage of these benefits in the PIF process, the concept of integrating SCF technology with the PIF process called SCF-assisted PIF (SFPIF), was introduced by Farahani et al. (2020) to realize lightweight, sheet metal-polymer components with microcellular structures. Increasing cell density and reducing cell size beyond microcellular structures can further enhance the properties of polymer foams, such as thermal insulation and strength-to-weight ratio. However, current large-scale production methods, particularly injection molding, struggle with poor process control and repeatability. To address these challenges, a novel method called electromagnetic forming injection foaming (EFIF) process has been introduced (Farahani et al., 2025). This new hybrid process combines two key innovations: first, it enhances the PIF technique by incorporating SCF technology, which helps overcome common PIF limitations, such as polymer melt shrinkage and high viscosity, by enabling SCF-assisted foaming during injection. The second innovation integrates an electromagnetic forming (EMF) stage, offering precise control over the foaming behavior

and resulting in finely structured micro- to nanocellular foams. In this approach, a polymer melt, pre-mixed with dissolved SCF, is injected to fill the cavity while simultaneously forming the metal sheet against the mold's cavity. Following injection, a pulsed electromagnetic field is applied to shape the metal component further and hence induce a rapid pressure drop, triggering the foaming process in the polymer (Pirani et al., 2024). The schematic of the EFIF process is shown in **Fig. 1**.

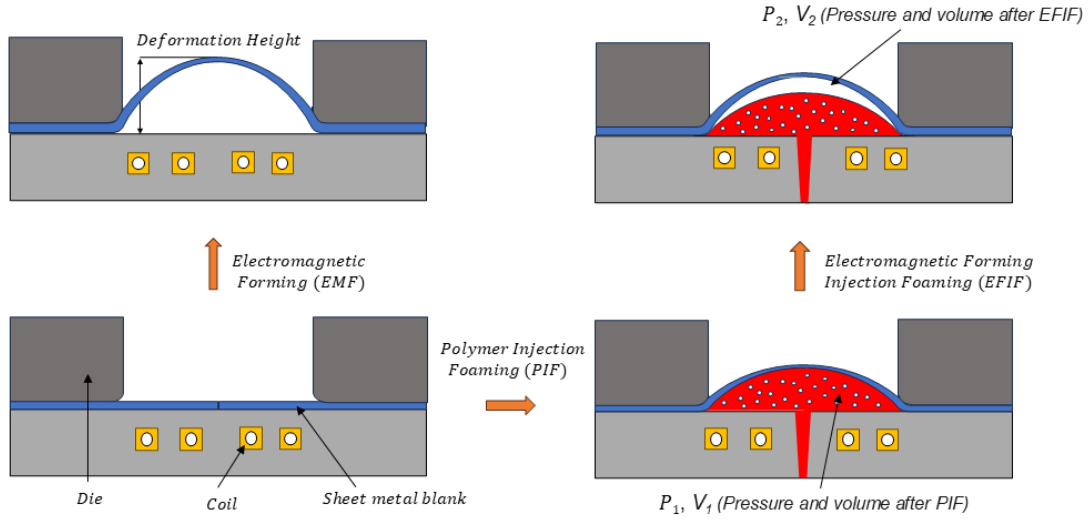


Figure 1: Schematic illustration of the electromagnetic forming injection foaming (EFIF) process

2 Numerical-Analytical Modeling

A combined methodology involving both numerical simulation and analytical modeling was adopted to investigate the feasibility of the EFIF concept. This numerical investigation aims to evaluate the effectiveness of EMF in enhancing the foaming behavior by inducing a significant pressure drop, which in turn promotes cell nucleation within the polymer matrix. The simulation framework was implemented in Abaqus 2022 using a 2D axisymmetric dynamic explicit model and was executed in two sequential steps. The first step mimicked the simultaneous filling-forming stage of the PIF process by applying a uniform load all over the deformation zone. The second step simulated the high-speed EMF process, incorporating a varying analytical pressure field derived from magnetic field equations.

The results obtained from the simulation, particularly displacement fields and deformed volume, were extracted and used as inputs for a set of equations derived from classical heterogeneous nucleation theory.

2.1 Finite Element Models

To reduce computational cost and simplify the simulation setup, a 2D axisymmetric model was developed in Abaqus 2022, considering free-forming conditions for the PIF step and circular coils for the following EMF step. A circular sheet metal blank with a thickness of 1 mm was modeled as an axisymmetric, deformable shell part, while the blank holder and die were represented as axisymmetric analytical rigid bodies. Two independent reference points

were assigned to the blank holder and die to facilitate the application of boundary conditions and loading constraints. The material properties of aluminum 6061 used for the blank are detailed in **Table 1** (Farahani, 2018).

Density (kg/m^3)	2710	Yield Strength (MPa)	253.16
Modulus of Elasticity (GPa)	68.9	Poisson ratio	0.33

Table 1: Material properties of AA 6061 used in simulation adopted from (Farahani, 2018)

The material behavior in this study was modeled using the rate-dependent Johnson–Cook (J–C) constitutive flow stress equation. The parameters used here are presented in **Table 2**.

Material	A (MPa)	B (MPa)	n	C	m	T_m (K)
AA 6061	324	114	0.42	0.002	1.34	925

Table 2: Johnson-Cook material model parameters (Doley et al., 2022)

Following the assignment of material properties and assembly of the model components, the simulation was carried out using two sequential analysis steps. The first step corresponds to the PIF process and was assigned a time increment of 0.001 seconds. The second step represents the EMF phase, simulated over a shorter duration of 0.0003 seconds to capture the high-speed nature of the process. During the first step, a uniform pressure load was applied to the top surface of the blank to emulate the mechanical pre-forming. In the second step, the electromagnetic pressure was applied as an analytical load field based on the magnetic pressure distribution described by **Eq. 1**. This equation was modified to better align with the experimental data reported by (Paese et al., 2022). It predicts the pressure distribution generated by a spirally wound, flat tool coil as a function of the coil's geometric dimensions and the transient current I (Psyk et al., 2011).

$$p_r = \frac{\mu_0 I^2 N^2}{2\pi^2 l^2} \left[\tan^{-1} \left(\frac{a_{air}^2 + r_{o,Coil}^2 - r^2}{-2a_{air}r} \right) - \tan^{-1} \left(\frac{a_{air}^2 + r_{i,Coil}^2 - r^2}{-2a_{air}r} \right) \right]^2 \quad (1)$$

Here, N represents the number of coil turns, and l denotes the coil winding width. The term a_{air} refers to the gap between the tool coil and the workpiece. The parameters $r_{i,Coil}$, and $r_{o,Coil}$ correspond to the inner and outer radii of the tool coil, respectively, while r indicates the radial position at which the electromagnetic pressure is evaluated and μ_0 is the magnetic permeability in a vacuum. For the simulation, the parameters were set as: $N = 8$, $l = 37.5$ mm, $r_{i,Coil} = 10$ mm, $r_{o,Coil} = 47.5$ mm, $\mu_0 = 4\pi \times 10^{-7} \text{N}\cdot\text{A}^{-2}$, and for the maximum current $I = 42000\text{A}$, whereby its amplitude was input as an exponentially decaying sine wave approximated from the experiments. The air gap was calculated after the PIF step, and the average value was used in step 2. A friction coefficient of 0.1 was applied, and the die and blank holder were fully constrained to fix their positions throughout the simulation. For meshing the blank, an element size of 0.1 mm was chosen to ensure sufficient resolution for capturing localized deformation and stress gradients. The blank was discretized using the CAX4R element type, a 4-node bilinear axisymmetric quadrilateral element with reduced integration and compatible with explicit dynamic analysis.

2.2 Cell Nucleation Analytical Models

Cell nucleation plays a key role in shaping the morphology and properties of polymer foams. A rapid nucleation rate is essential to achieve high cell density, typically driven by thermodynamic instability in the gas-polymer system. This instability is often triggered by a sudden drop in pressure, which sharply reduces gas solubility. More rapid pressure drops generally lead to more cell nucleation due to the increased instability. The classical nucleation theory is a fundamental framework for modeling various nucleation mechanisms (Pirani et al., 2024). In this study, the heterogeneous nucleation model was employed to analytically calculate the nucleation rate under EFIF processing conditions, as it providing better alignment with both the simulation results and experimental observations in terms of the magnitude and nucleation behavior. The heterogeneous nucleation rate is written as:

$$N_{het} = f_1 C_1 \exp(-\Delta G_{het}^* / kT) \quad (2)$$

where N_{het} is the heterogeneous nucleation rate, f_1 is the frequency factor for heterogeneous nucleation, C_1 represents the concentration of heterogeneous nucleation sites, k is Boltzmann constant, T is the absolute temperature and ΔG_{het}^* is the change in Gibbs free energy for the heterogeneous nucleation, which is expressed as:

$$\Delta G_{het}^* = \frac{16\pi}{3\Delta P^2} \gamma_{bp}^3 S(\theta_w) \quad (3)$$

ΔP represents the pressure drop inside the cavity after the injection, γ_{bp} is the surface energy at the polymer–bubble interface, and $S(\theta_w)$ is the function that depends on the wetting angle between the polymer, the gas, and the second-phase particle (Xu, 2010). To apply **Eq. 2** and calculate the theoretical heterogeneous nucleation rate, it is essential to calculate the pressure drop under EFIF processing conditions. The additional deformation induced by the electromagnetic field leads to an expansion of the cavity volume. Since the injected polymer melt is incompressible, cell nucleation and growth compensate for this volume increase, thereby initiating the foaming process (Pirani et al., 2024). Consequently, the relationship between the pressure drop and cavity volume expansion can be derived using a modified form of the ideal gas law, as shown in **Eq. 4**. Where P_1 and V_1 are the pressure and deformation volume at the end of the simultaneous filling and forming stage, P_2 and V_2 are the pressure and deformation volume after applying electromagnetic forces (see Fig. 1), and K is a constant. Pressure and deformation volume can be calculated from the simulation results.

$$\Delta P = P_2 - P_1 = KT \left(\frac{1}{V_2} - \frac{1}{V_1} \right) = P_1 \left(\frac{V_1 - V_2}{V_2} \right) \quad (4)$$

3 Experimental investigation

Experimental investigations were conducted on AA 6061 samples formed using the PIF process, with an injection volume of 25 cm³ and a thickness of 1 mm. The samples were subsequently subjected to the EMF process without applying any additional materials such

as injected polymer between the coil and sheet metal to assess their effect on the final shape, the rate of volume expansion, and the rate of pressure drop which are key factors influencing the nucleation rate. The process parameters used in the experiments were identical to those applied in the simulation studies. **Fig. 2** shows the deformation height before and after applying the EM force, along with the corresponding volume increase rate resulting from the deformation.

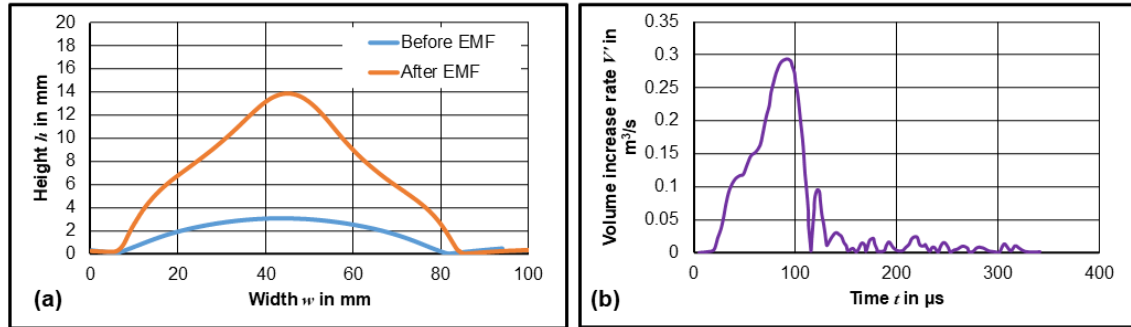


Figure 2: Experimental results for AA 6061 (a) deformation height (b) Volume increase rate

Results and Discussions

Based on the input parameters detailed in Section 2.1, the displacement results for each forming step and the comparative analysis between simulation and experimental results regarding the deformation height following EMF are presented in **Fig. 3**.

Using the displacement data, the deformed volume of the blank for each step was calculated and utilized in **Eq. 4** to determine the associated pressure drop. This value was then substituted into **Eq. 3** to calculate the corresponding Gibbs free energy change, which was finally placed in **Eq. 2** to calculate the heterogeneous nucleation rate. The input data used for this calculation and the results derived from the numerical simulations and the analytical cell nucleation models are listed in **Table 3**.

$V_1(\text{cm}^3)$	$V_2(\text{cm}^3)$	$P_1(\text{bar})$	$P_2(\text{bar})$
6.208	29.472	16	3.37
$f_1 (1/\text{sec})$	$\gamma_{bp}(\text{dynes/cm})$	$T(\text{K})$	$k(\text{J/K})$
8.2	7.85	513.15	1.380649×10^{-23}
$\frac{\Delta P}{\Delta t} (\text{bar/sec})$	$\Delta G_{het}^* (\text{J})$	$C_1(\text{sites/cm}^3)$	$N_{het}(\text{nuclei/cm}^3\text{sec})$
4.2×10^4	5.08×10^{-21}	10^{10}	4×10^{10}

Table 3: Input data and calculation results

The results demonstrate the EFIF concept's effectiveness in creating a sudden pressure drop, leading to a favorable nucleation rate essential for producing polymer foams with a micro- to nano-cellular structure. Moreover, by adjusting key EMF parameters, it is possible to control the pressure profile and optimize cell size and density. This highlights the potential of EFIF as a tunable and efficient technique for scaling the production of sheet metal-polymer parts with fine cellular structures.

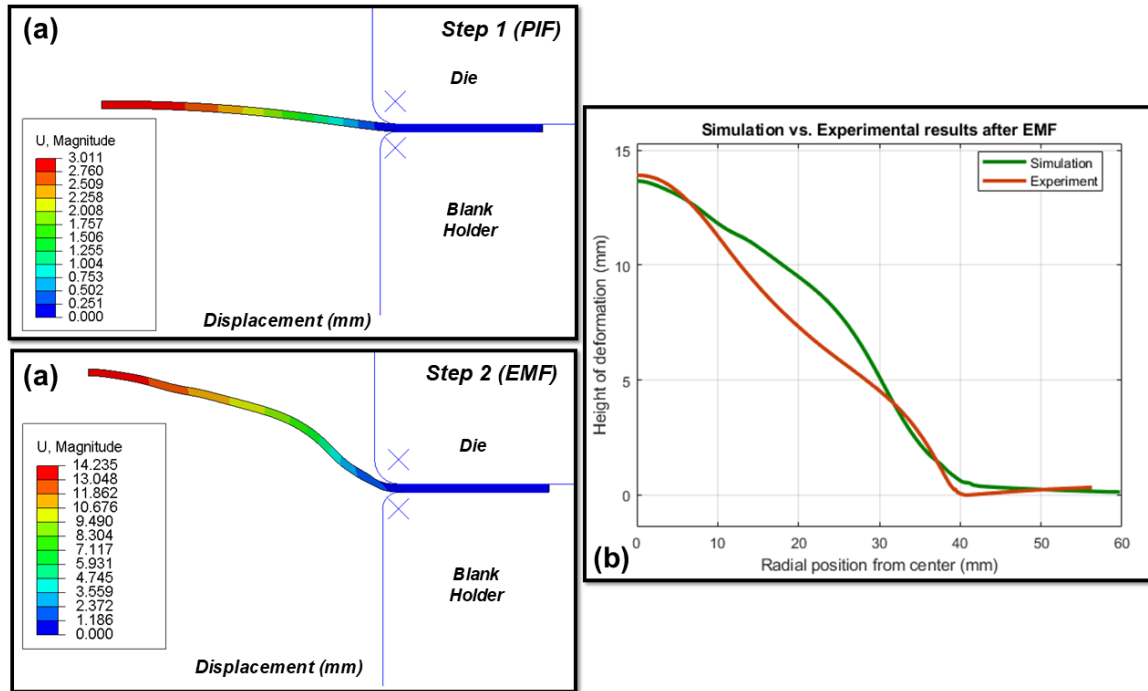


Figure 3: (a) Simulation results for each step (b) Simulation vs experiment height of deformation result

4 Conclusion and Future Work

A combined numerical and analytical investigation was conducted to evaluate the feasibility of the EFIF process. Simulation results were utilized within the analytical framework to calculate the pressure drop and cell nucleation rates, providing insight into the coupling between mechanical deformation and thermodynamic conditions involved in the EFIF concept. The findings demonstrate that this novel approach has the potential for controlling SCF-assisted foaming in hybrid manufacturing processes like the PIF process, introducing a tunable and efficient technology for scaling the production of sheet metal-polymer parts with micro to nanocellular structures.

Future explorations in this direction include investigating the effects of coil geometry, current variation, and different polymers and sheet metals on the process outcome. In addition, implementing a closed-shape cavity mold will be considered to improve control over the final shape and thickness of the foamed sheet metal part. The next phase will involve the fabrication of dedicated tooling and the execution of experimental trials to validate the analytical and numerical models, along with the EFIF process for real-world industrial applications.

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Chapter 2

Electromagnetic welding

Investigation of Electromagnetic Pulse Welding for the Automated Production of Resource-Efficient Multi-Material Joints

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Abstract

Electromagnetic pulse welding (EMPW) is a promising technology for the automated production of multi-material joints made of high-strength aluminium alloys and steels and is therefore capable of realising light-weight design by exploiting the potential of both materials. The ongoing Design2Collide research project is investigating the potential applications of EMPW in an industrial environment with the aim of reducing the limitations associated with the use of the process. Its automation capability is demonstrated by converting a state-of-the-art robotic production cell to join a demonstration assembly made of steel (DC04) and aluminium (AA6016). Welding studies to derive process windows with the best joint strengths are carried out using an EMPW system and a special model test rig for collision welding. Reproducible attainment of ample joint strength is indicated by failure of the base material AA6016 due to shearing, while the joint itself remains intact. To ensure excellent joint properties, an in-line quality control method based on active thermography is developed and integrated into the automated process. Initial stand-alone tests of the developed method with the EMPW system show promising results for feasibility and industrial qualification. The examination of the EMPW value chain is currently being

completed by the investigation of corrosion prevention and passivation systems for the pre- and post-treatment of materials and their impact on weldability, and by the improvement of process simulation. For the latter, adapted material models are essential. Respective flow curves have already been determined for the materials used at high strain rates of up to 10^4 1/s.

Keywords

Electromagnetic Pulse Welding, Multi-Material Design, Lightweight Concepts, Emission Reduction

1 Introduction

The transport sector accounts for about 21% of global CO₂ emissions, with passenger cars contributing about 38% of this share (Crippa et al., 2024; EDGAR, 2024; IEA, 2025). In addition to greenhouse gases, non-exhaust particulate emissions - especially from tyre wear and brake dust - are gaining attention as they are strongly influenced by vehicle weight (Lopez et al., 2023; OECD, 2017; Timmers et al., 2016). Reducing vehicle mass therefore helps to reduce both CO₂ emissions and particulate pollution such as microplastics (Bandivadekar et al., 2008; Beddows and Harrison, 2021; OECD, 2017). Lightweight design using multi-material design, particularly high-strength steel and aluminium, plays a key role in this effort (Tisza & Lukács, 2018).

However, joining these dissimilar materials remains a challenge. Conventional fusion welding methods, such as laser or TIG welding, often lead to the formation of brittle intermetallic phases due to different melting points, which significantly reduce the strength of the joint (Agudo et al., 2007; Atabaki et al., 2014; Carlone and Astarita, 2019; Khedr et al., 2023). As a solid-state or collision welding process, electromagnetic pulse welding (EMPW) offers a promising alternative (Agudo et al., 2007; Kapil and Sharma, 2015; Khedr et al., 2023). Currently, EMPW process design relies on iterative methods, which are costly and introduce uncertainties in mass production (Marschner et al., 2021).

The ongoing Design2Collide research project is investigating the potential applications of EMPW in an industrial environment with the aim of reducing the limitations associated with the use of the process. The entire chain for integrating the joining technology into industrial series production of multi-material components made of aluminium and steel materials is being analysed. Welding tests are carried out using an EMPW system and a special model test rig for collision welding to determine process windows and parameters with the best possible joint strength. These two systems are also used to implement measures to improve quality and extend process limits by investigating corrosion prevention and passivation systems for the pre- and post-treatment of materials and their impact on weldability. This machine-level approach will also be used to develop an in-line quality control method for subsequent integration into automated EMPW production processes. The

automation of the EMPW process will be achieved by converting a robotic production cell for joining a demonstration assembly made of steel (DC04) and aluminium (AA6016).

In parallel to the physical processes and investigations mentioned above, the EMPW process is also being analysed and further developed on the level of numerical simulation. For this purpose, adapted material models are required. Corresponding flow curves have already been determined for the materials used in the project at high strain rates of up to 10^4 1/s. The numerical representation of the collision process and the associated prediction of the joint formation will thus enable the simulation-based design and stress analysis of EMPW joined components in the future. The knowledge gained in the course of the project, both at the real process level and at the simulation level, will finally be transferred to the production of an EMPW Crash Management System (CMS) for passenger cars. The design of the EMPW-CMS will be adapted compared to a conventionally manufactured CMS in order to enable EMPW-compatible production on the one hand, and to make best use of the lightweight concepts and strength potential resulting from the EMPW technology in combination with high-strength aluminium and steel materials on the other hand.

In the following sections, this paper discusses the process and the process steps carried out in the project to adapt and qualify the robotic production cell and the demonstration assembly, both originally designed for conventional joining processes, for automated EMPW production.

2 Methodological structure

The structure of the Design2Collide research project shown in Fig. 1 is divided into the physical process research track and the process simulation track. At the end of the project, knowledge from both areas will be transferred to the production of the EMPW Crash Management System. In the following sections, this paper presents the physical process research and the path to automated EMPW production of an established component. In Fig. 1, these are the top three boxes on the left.

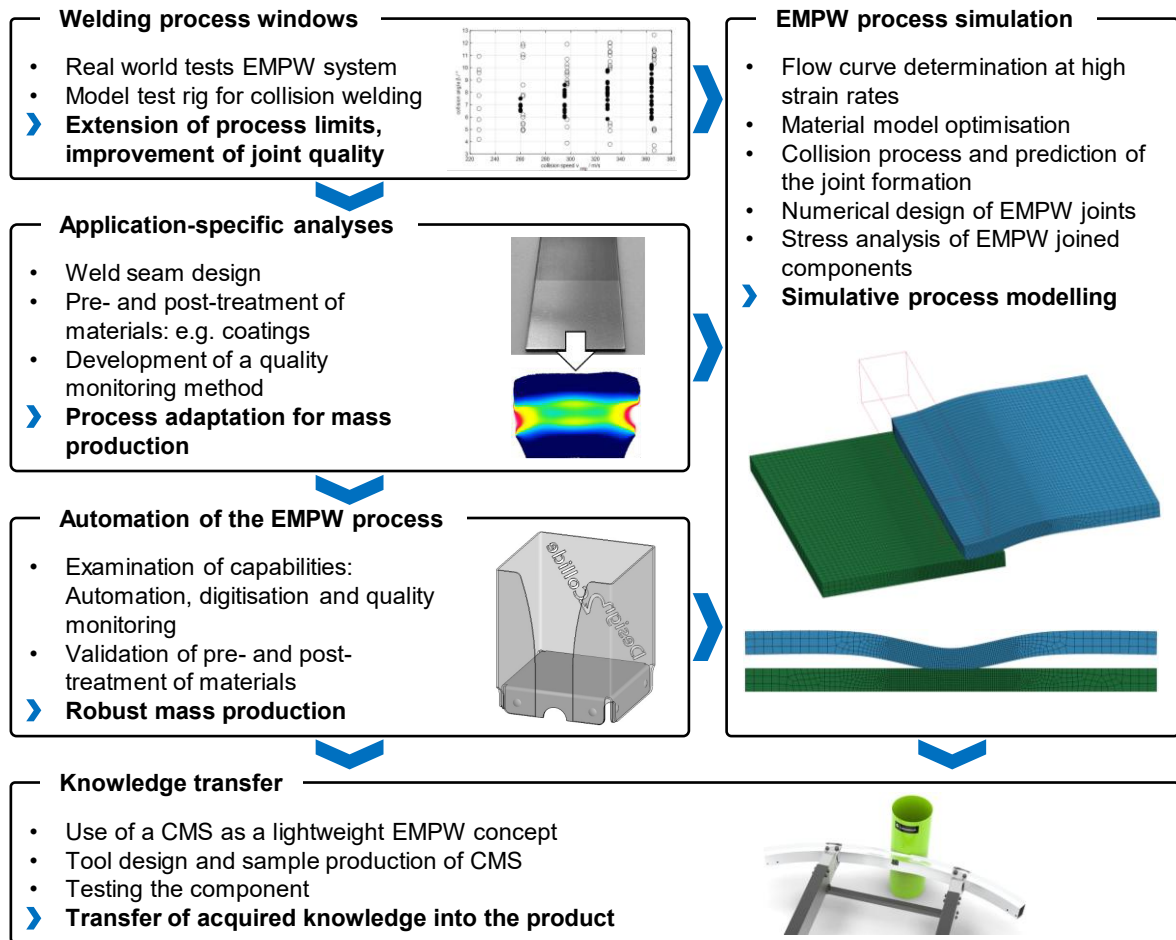


Figure 1: Schematic project flow of the Design2Collide research project

2.1 EMPW-compliant component design

The investigation of EMPW in an industrial environment is based on an established assembly consisting of two elements to be joined together. This assembly is undergoing adaptation for the EMPW. This adaptation includes the material selection, the design consideration of the acceleration gap and ensuring the damage-free formability of the component during the EMPW. A note box consisting of a base and a wall plate is selected as a representative component. The existing note box is manufactured from aluminium using semi-tubular rivets.

2.2 Investigation of collision welding with AA6016 and DC04

Welding tests are carried out on the material combination of AA6016 and DC04 using a model test rig for collision welding and an EMPW system to determine the EMPW parameters and to ensure the best possible joint design for the note box. The model test rig for collision welding, developed at the Institute for Production Engineering and Forming Machines (PtU) at the Technical University of Darmstadt, allows the collision welding

process to be investigated over the time profile of the collision with constant and precisely adjustable collision parameters thanks to its fully mechanical design. It is being used to determine the influences of the collision angle and the collision velocity on the joint formation and the weld seam quality. A detailed description of the test rig is given in (Kraus et al., 2024). The specimens used are taken from DC04 sheets (thickness t : 2.0 mm, tensile strength R_m : 310 N/mm²) and AA6016-T4 sheets (thickness t : 2.0 mm, tensile strength R_m : 240 N/mm²) by shear cutting. The tensile strength values mentioned are determined by performing three tensile tests on a *Zwick Roell 100* combined tensile and compression testing machine in accordance with DIN 50125. The process window is determined by varying collision velocity v_{imp} and collision angle β . The collision velocity is varied in five steps of 227 m/s, 262 m/s, 297 m/s, 331 m/s and 366 m/s by rotating the rotors at a specified frequency. The collision angle is varied between 3° and 13°. The specimens are cleaned with acetone immediately prior to testing to eliminate any influence on the test results from lubricant or corrosion inhibitor residues on the specimen surfaces. High-speed observation of the collision is performed using a *PCO hsfc pro* image intensifier camera, which acquires up to eight images per collision with an exposure time of 20 ns.

The EMPW process parameters are determined by experiments using a *PS48-16* EMPW system from the manufacturer *PSTproducts GmbH* and a coil with an effective area of 80 mm x 10 mm, with particular focus on the acceleration gap before welding, the discharge current and the weld seam design. The gap is analysed in two increments of 1 mm and 1.5 mm. The discharge current is analysed in three steps within a range of 424 kA to 483 kA. In this study, two weld seam designs are analysed, resulting from either a symmetric or an asymmetric collision. In an asymmetric collision, the flyer is only supported on one side. This results in a one-directional collision when it collides with the target. This creates a semi-elliptical weld seam. However, in the case of a symmetrical collision, the flyer is supported on both sides and moves in a rolling motion in both directions from the point of impact, creating an elliptical weld seam. The sheets used are 2.0 mm thick for DC04 (tensile strength R_m : 310 N/mm²) and 1.5 mm thick for AA6016-T4 (tensile strength R_m : 235 N/mm²).

2.3 Conversion of the automation system of the robotic production cell

An existing automated modular robotic production cell is modified to be used with a sheet welding tool coil. Irrespective of the actual joining method, the sequence is as follows:

- The two separate parts are manually placed in a workpiece carrier on a rotary table and fed into the production cell via a 180° rotation. Two workpiece carriers are installed on either side of the rotary table.
- The robot simultaneously picks up the note box with a suction pad (base plate) and a pneumatic gripper (wall plate). A sensor for each component is used to check that it is in place. Mechanical provisions ensure precise alignment of the two parts respective to each other.
- The robot moves the box to the joining tool and passes through the three joining positions successively.

- After the last joining operation, the robot places the finished box in the workpiece carrier and picks up a new set of parts.

Two components of the production cell need to be adapted for electromagnetic pulse welding: The riveting station is replaced by the welding unit, which mainly consists of the tool coil. The handling unit on the robot, consisting mainly of the gripper, must be modified so that the electrical components can withstand the electromagnetic field generated by the tool coil. In addition, they must be electrically isolated from the note box in case the insulation of the tool coil is damaged. The core components of the production cell are shown in Fig. 2.

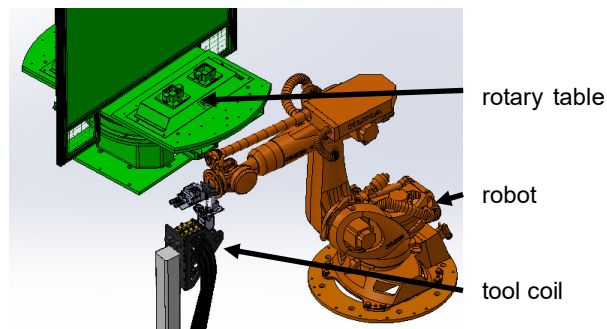


Figure 2: Core components of the production cell (the safety fence is not shown).

2.4 Design of the welding unit

The welding unit consists of the tool coil and the support (or backer) for the target (the base plate of the note box). The effective part of the tool coil, which induces the eddy currents into the workpiece, is the length of the flange of the base plate (83 mm), see Fig. 3 c). With the information from the EMPW examinations (section 3.2), the required peak discharge current is known. The cross-section of the effective part of the tool coil and the number of channels of the pulse generator can be calculated accordingly.

The backer must be able to move by a few millimetres to open the space for the note box to be inserted by the robot. During the welding impact it must be locked in place rigidly. In this closed and locked position, it must be able to withstand the impact of the flyer on the target.

2.5 Post-process quality monitoring

To ensure optimal joint quality, active thermography is utilised as a post-process quality assurance method. The inspection is carried out in a transmission arrangement, with excitation by means of induction. The thermographic system used is a *FLIR SC5600-M*. The technical data of the system are summarized in (Graß et al., 2024). The acquisition frequency is 300 Hz, the integration time is 2 ms and the evaluation time is 0.2 - 5 s after the pulse of the inductor. The evaluation in the frequency domain was carried out at 1.2 Hz. The inductor is cuboidal (80 mm x 70 mm x 65 mm) - the induction parameters are 0.1 s pulse duration, 8 kHz excitation frequency and a pulse width modulation of 470%. The intention is to generate and detect defective EMPW joints. For this purpose, welding is locally prevented

by the application of an anti-corrosion oil ($6 \mu\text{g}/\text{mm}^2$). Defective joining partners should also be taken into account. For this purpose, a joining partner with grooves is prepared before the welding process. To ensure comparability, a reference joint is thermographically analysed. The evaluation is based on the phase images, as these provide the most detailed information on heat transfer.

3 Results

3.1 EMPW-compliant component design

To investigate the automated operation of EMPW and the processes of the associated value chain, an established component, a note box consisting of a wall plate and a base plate, is used and adapted for EMPW. The base plate of the note box will be made of steel (DC04, t : 1.5 mm) and the wall plate of aluminium (AA6016, t : 1.5 mm). These material adaptations reveal the potential of the EMPW for hybrid aluminium-steel joints. In addition, the choice of aluminium for the wall plate ensures that it can be used as flyer. Fig. 3 shows the note box joined by conventional self-pierce riveting, the note box optimised for the EMPW and a detailed illustration of the joined area. The wall plate of the EMPW note box is welded to the base plate on three sides. The front side of the note box with the opening is left unwelded, as the available areas to be joined are insufficiently large. The EMPW-compliant design of the note box consists mainly of the implementation of eight embossings, ensuring an acceleration gap of 1 mm between the wall and the base plate of the note box. Furthermore, notches are included. These notches ensure the controlled deformation of the wall plate during the acceleration caused by the EMPW.

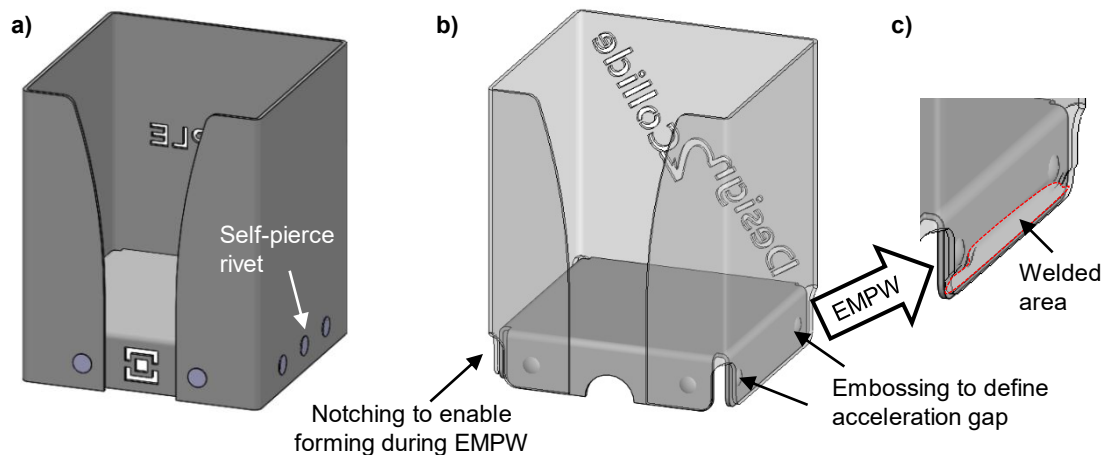


Figure 3: a) Conventional note box, joined using self-pierce rivets, b) note box adapted for EMPW c) detail of welded area of note box (flange length 83 mm)

3.2 Investigation of collision welding with AA6016 and DC04

The welding process window, as determined by using the model test rig for the material combination of aluminium AA6016-T4 (t : 2.0 mm) and DC04 steel (t : 2.0 mm), is illustrated in Fig. 4. The collision welding tests are performed at collision velocities v_{imp} of 227 m/s, 262 m/s, 297 m/s, 331 m/s and 366 m/s. Within the defined process window, a collision weld test is considered successful if the two specimens cannot be separated by manual force after the test (black dots). To improve the clarity of the diagram, the successful tests are shifted 2 m/s to the left on the collision velocity axis compared to the unsuccessful tests (white dots). The formation of the first bonds between the two materials occurs at a collision velocity of 262 m/s. The lower and upper limit angles at this velocity are 6.5° and 7.5°, respectively. At a collision velocity of 297 m/s, an increased upper limit angle of 8.6° is determined, while the lower limit angle decreases to 6.0°. Furthermore, at collision velocities of 331 m/s and 366 m/s, there is an even greater increase in the upper limit angle to 9.8° and 10.2° respectively, while the lower limit angle decreases marginally to 5.9° for both velocities. More detailed investigations into the relationship between tensile shear strength values and process parameters can be found in (Kraus et al., 2024).

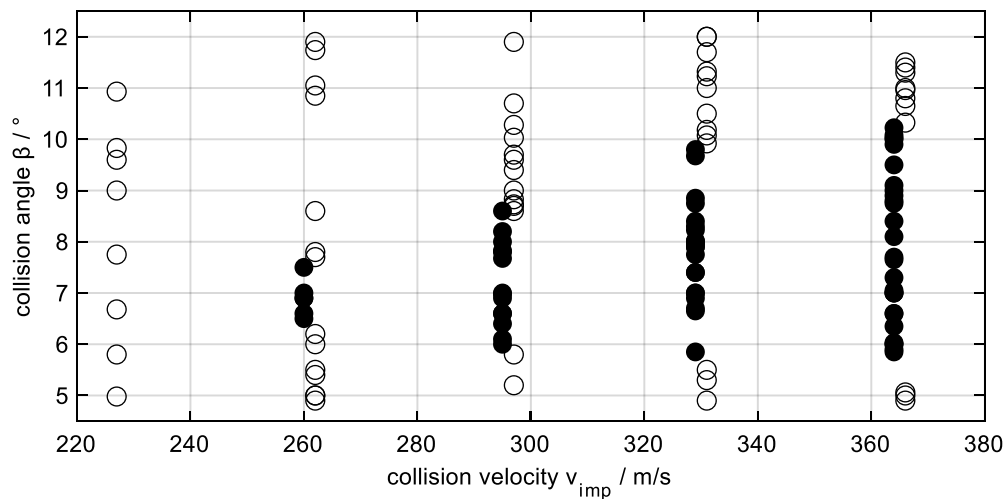


Figure 4: Welding process window of AA6016-T4 (flyer) and DC04 (target).

The investigation of the EMPW of AA6016-T4 (t : 1.5 mm) and DC04 (t : 2.0 mm) is carried out by examining two different weld designs and varying relevant process parameters. The weld seam designs considered are produced by using a symmetric collision, which results in an elliptical weld seam, see Fig. 5 a), and by forcing an asymmetric collision, which generates a crescent-shaped weld seam (Fig. 5 b). The parameters varied are the acceleration distance and the discharge current. Good joint strengths are obtained using symmetric collision and, in the best case, joints manufactured in this way fail under quasi-static tensile shear stress in the base material. With asymmetric collision, lower joint strengths are achieved in the overall discharge current range under consideration. Nevertheless, it is shown that 1 mm of acceleration gap generates significantly better joint strength than 1.5 mm (Fig. 5 c). The advantage of the symmetrical collision is presumably

due to the larger weld seam. The positive influence of the 1 mm gap is due to its associated effect on the jet. At small collision angles, the jet interacts more strongly with the surfaces of the joining partners and activates them thermally.

The metallographic analysis of the weld seam interface of a joint with symmetrical collision and fabricated with 1 mm acceleration gap shows typical EMPW characteristics. Cracks and pores can be seen at the beginning of the weld seam, compare Fig. 5 d). An EDS (Energy Dispersive X-ray Spectroscopy) line scan (Fig. 5 f) performed through a melt pocket near the beginning of the weld seam shows that oxygen is present. This indicates that parts of the aluminium oxide layer, which is usually ejected as a jet during the plasticisation process during the collision, remained in the interface. The interface at the end of the weld seam also contains localised melt pockets, but these are less prominent than at the beginning of the weld seam (Fig. 5 e). A line scan carried out at the end of the weld seam indicates the absence of oxygen. The chemical composition of the weld seam interface is also influenced by the position of the analysis. In comparison to the EDS line scan of the beginning of the weld seam, lower amounts of iron are detected in the interface, see Fig. 5 g). The differences in the chemical composition are due to the respective collision conditions (collision angle and collision point velocity). Furthermore, the analysed microstructure corroborates the findings regarding the interaction between the mechanical joint strength and the adjusted acceleration distance. The jet interacts with the boundary layers of the joining partners across the entire weld seam area, causing localised melting and resulting in sufficient joint strength.

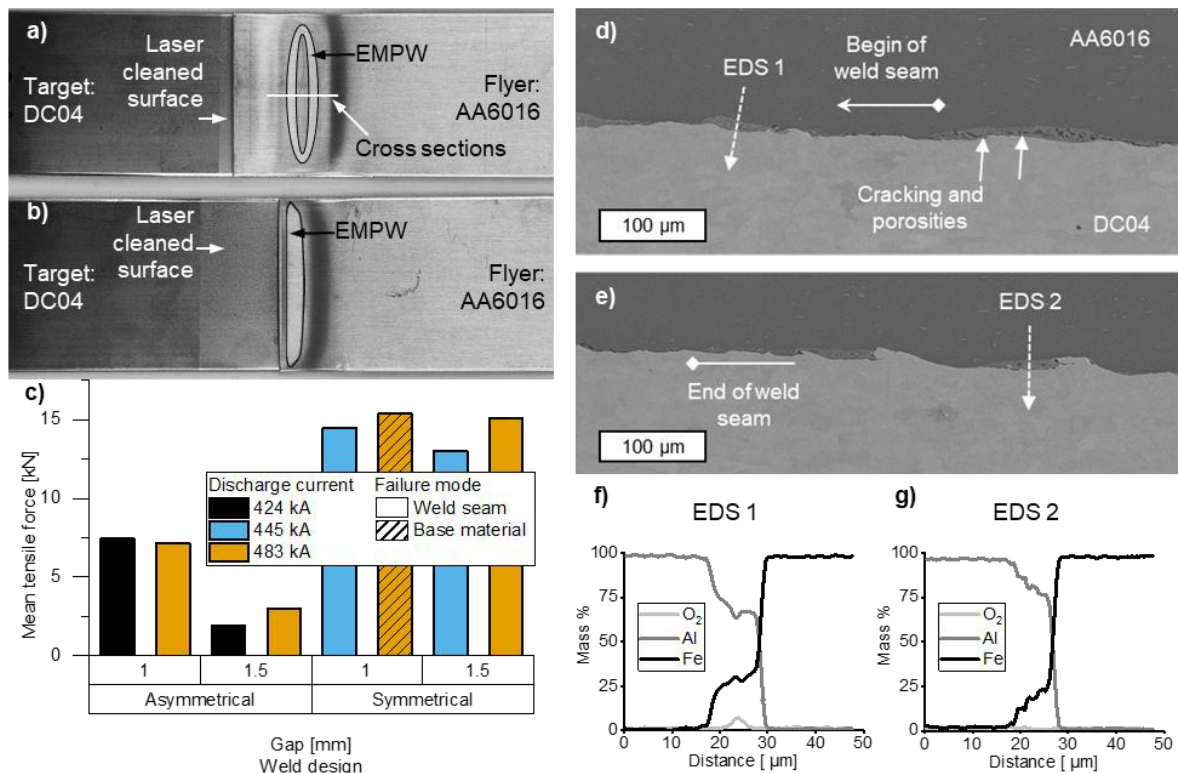


Figure 5: AA6016 / DC04 joints: a) symmetrical collision, b) asymmetric collision, c) quasi-static shear tensile strength, d) weld seam (445 kA and 1 mm gap) begin and pos. of EDS scan 1, e) weld seam end and pos. of EDS scan 2, f) EDS 1 line graph, g) EDS 2 line graph

For the experiments on the note box, the asymmetric collision is chosen despite the inferior mechanical joint properties. The reason for this choice is the accessibility required for the joining and clamping process and the subsequent non-destructive testing.

3.3 Conversion of the automation system of the robotic production cell

The gripper tool and its auxiliary components on the robot are adapted as follows:

- The inductive sensors used for detecting the presence of base and wall plate of the note box are too sensitive to be used near the tool coil due to the electromagnetic fields. They are replaced by optical sensors which effectively measure the same parameter: The distance between the sensor and the workpiece.
- All other electronic components (e.g. the pneumatic valves or the field bus coupler) are moved as far away as possible from the tool coil and the note box to reduce electromagnetic interference.
- If the insulation of the tool coil is damaged, the high voltage can cause a high current to flow into the grounded robot and all attached components. To protect the robot and the electrical components at the hand of the robot, the note box is insulated from all other components by fibre-reinforced plastic parts. An air gap of about 20 mm is sufficient for the maximum charging voltage of 16 kV.
- Eddy currents are minimised by additional layers of insulating material between conductive components. They cut off the path of the eddy currents. These insulations only need to be a few millimetres thick, as the voltage potential of the eddy currents is far lower than the charging voltage.

As a result of these adaptations, almost all other components are modified as a direct consequence.

3.4 Design of the welding unit

The welding unit consists of the tool coil, the backer and its mechanism, and a frame with a mounting flange, see Fig. 6. The assembly will be mounted on a post for ergonomic reasons. The design of the tool coil can be broken down into the effective part, which induces the eddy currents into the workpiece, its support, which absorbs the reaction force, and the clamping for the pulse cables. The length of the effective part of the tool coil is 83 mm, which is given by the dimensions of the note box. Its cross-section is defined as 10 mm x 7 mm according to the estimated maximal peak discharge current of 500 kA. This discharge current value also determines the number of capacitors. In order to have some reserve, a pulse generator with six capacitors is used.

The key requirement for the backer is that it can be moved by a few millimetres and locked in the pulse position. This is fulfilled by a toggle lever operated by a compact pneumatic cylinder. The toggle lever moves a class I lever, multiplying the force by about seven. To withstand the impact forces and absorb as much of the shock load as possible, this lever is designed with as much steel as possible. The area which is in direct contact with the note box is a separate component with a simple geometry, making it cheap and easy to

replace. This also facilitates experimental testing of different materials for this task. The adjustment of the lever is accomplished by shims and by the pneumatic cylinder.

The frame holds both the tool coil and the lever of the backer. It is strong enough to withstand the shock loads on the tool coil and the backer. The position of the tool coil within the frame can be modified for fine tuning. The mounting flange is part of the frame. For electrical safety reasons, the frame is electrically insulated from the post onto which it is mounted.

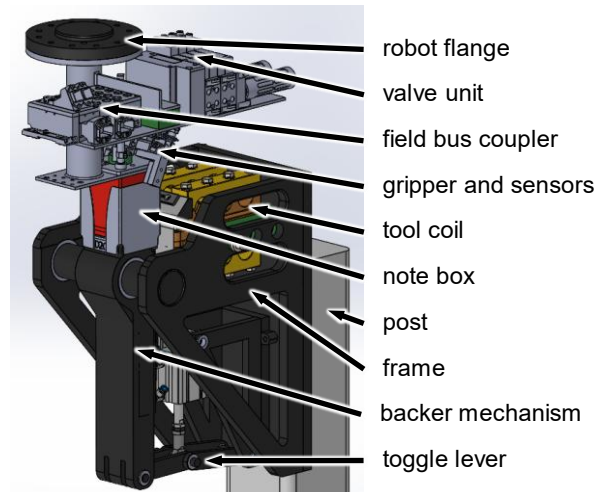


Figure 6: *Welding unit (tool coil, backer, frame) and robot tool (gripper, valves, field bus coupler)*

3.5 Post-process quality monitoring

EMPW joined specimens are analysed using active thermography. The excitation is carried out inductively and the test setup is a transmission arrangement. Both aspects ensure transferability to investigations on the note box. In addition to a defect-free reference specimen, three specimens with defects are analysed, see Fig. 7. Welding is prevented in the centre and on one side by the selective application of an anti-corrosion oil. Furthermore, three grooves are milled into a DC04 target, which are intended to negatively affect the weld seam formation. All specimens are analysed using phase images. The phase images provide information on the thermal wave characteristics at the specimen surface in relation to time aspects of the excitation.

The phase image of the reference specimen reveals a widespread heat transfer without a large time delay, indicating a consistent weld. The application of an anti-corrosion oil results in incomplete welding and inhibits heat transfer from the target to the flyer. This is shown in Fig. 7 b) and c). It is clearly recognisable that heat transfer occurs only off-centre or on one side of the specimen. The influence of grooves on the heat transfer from the target to the flyer is also visible. This results in an inconsistent time delay of the heat wave relative to the excitation. This can be perceived as hotspots whose positions correspond to those of the grooves, compare Fig. 7 d).

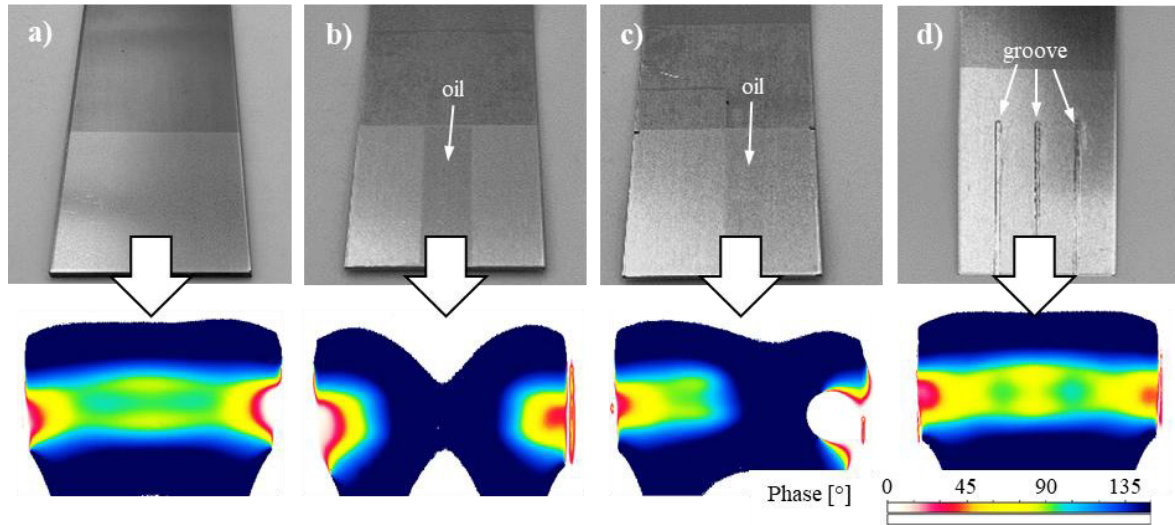


Figure 7: Images of samples and corresponding phase images of a) a reference joint, b) a joint with a central defect, c) a joint with a off-centre defect, d) a joint with three grooves

4 Summary

In the present study, EMPW was examined in the broader context of industrial application, with particular emphasis on the fundamental steps in the value chain. A suitable existing demonstration component, namely a note box, is used to show how a redesign according to EMPW can be achieved. The selection of materials is an integral part of this process. The materials selected for this project are aluminium AA6016-T4 and steel DC04, which are successfully joined after preliminary investigations using a model test rig and an EMPW system. Interactions between process and collision parameters and the joint properties are also demonstrated. Additionally, a tool coil and handling concept are developed to enable the automated production of the demonstration component. A quality monitoring process based on active thermography is investigated to ensure joint quality. Using joints fabricated by EMPW, it is demonstrated that the presence of defective welds can be detected through the analysis of phase images.

5 Outlook

In the further course of the research project, the results and measures presented in this paper will be transferred into a functional process for the automated production of the note box. The process will then be evaluated for robustness and reproducibility with regard to mass-production of EMPW joined multi-material aluminium and steel components. The effects of selected upstream and downstream material treatments, such as corrosion prevention and passivation systems, on the welding result are also being investigated using the EMPW system and the model test rig. These material treatments represent, for example, alternatives to current corrosion prevention systems such as zinc plating or anti-corrosion oils, which prevent welding with EMPW and require additional costly process steps (e.g. surface

cleaning, removal of the zinc layer) in an industrial environment. The results will also be validated using the automated process.

In the area of process simulation, material models are being optimised with already determined flow curves for various aluminium and steel materials at high strain rates. These models will be used to more accurately simulate the EMPW collision process to enable predictions of joint formation in relation to the process windows determined by the model test rig. This will enable the simulation-based design of EMPW welds and, based on this, the numerical stress analysis of EMPW welded components. The physical and virtual design results are then combined and used for the targeted design and manufacture of a product. A lightweight Crash Management System (CMS) for light vehicles has already been selected as the product to demonstrate the performance of the EMPW joints. The CMS is required to withstand a wide range of different load collectives, both in normal operation and in the event of a crash. The successful implementation of the product design will be validated by load tests of physically manufactured CMS.

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Investigation and Improvement of dissimilar Aluminum-Copper Joining using Magnetic Pulse Welding combined with local Inert Gas Application

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Abstract

Well-performing electrical conductors and contacting processes are crucial for a variety of industries and applications, including flat conductors such as busbars. The challenge is to manufacture relevant systems with excellent properties in a cost-efficient and ecological way. From a materials science perspective, one approach could be the utilization of aluminum in electrical systems. This is because aluminum is a better conductor than copper with regard to its weight. However, the exclusive use of aluminum is limited by the constraints of electrical systems that require the partial use of copper. This can be due to space limitations or connections to the power electronics, which are predominantly copper components. Hybrid systems consisting of aluminum, copper and dissimilar joints between both materials have been identified as a viable solution. However, the fabrication of these joints is challenging for conventional fusion welding processes due to chemical and thermophysical incompatibilities. Magnetic Pulse Welding (MPW) is a solid-state welding process which, due to its characteristic low energy input, allows a resource efficient fabrication of dissimilar joints. However, the limitations of MPW and their expandability, its potential for future applications, and the interactions of the process parameters with observed joint properties have not been sufficiently researched. Consequently, this study addresses MPW in the field of joining flat conductors by investigating the correlations between process parameters and mechanical, microstructural, and electrical joint properties. Furthermore, an inert gas supply during joining is investigated in order to determine the potential for improvement of the weldability and the resulting joint properties.

Keywords

Magnetic Pulse Welding, Dissimilar Materials, local Inert Gas Application, Electrical Joint Characterization

1 Introduction and state-of-the-art

Electrical contacts are becoming increasingly important. The possibility of using aluminum instead of copper as a conductor increases economic efficiency through cost and weight savings (Kaspar et al., 2014). Nevertheless, systems consisting only of aluminum are currently still not conceivable. For example, one problem is that components of power electronics are made of copper or that busbars consist of both aluminium and copper to be suitable to handle Li-ion battery packs (Schmidt et al., 2012). Therefore, a contacting process for aluminum-copper joints must be established. The contacting process should ideally be material-locking in order to ensure long-term reliable joints (Schmidt et al., 2012). This is where magnetic pulse welding (MPW) as a solid state welding process provides a solution. Many studies have shown that aluminum-copper-joints can be realized using the collision based process MPW. The MPW immanent collision is initiated by a current flow in a coil, which is generated by a controlled capacitor discharge. This current flow generates a magnetic field that induces eddy currents in a well-conducting adjacent joining partner (flyer). The eddy currents interact with the primary magnetic field, resulting in a force that accelerates the flyer. The flyer is purposefully brought into collision with the secondary joining partner, the target, after a few millimetres. The collision occurs as a rolling movement of the flyer on the target. The collision angle and collision speed vary continuously. If the collision angle and speed are within specific ranges, a mass flow, the jet, consisting of the boundary layers of both joining partners, atmospheric components and surface impurities, emerges from the collision point. This results in a welded joint between the bare metal surfaces of the two joining partners. (Graß et al., 2024) In the field of MPW of aluminum and copper this applies both to sheet-to-sheet joints (Psyk et al., 2017) and to tubes that are welded with other tubes (Wu and Shang, 2014), rods (Hokari et al., 1998; Marya et al., 2005; Oliveira et al., 2017; Raoelison et al., 2014) or cables (Graß et al., 2024).

The electrical properties of these joints were investigated for sheet-to-sheet and tube-to-cable joints (Graß et al., 2024; Psik et al., 2017). It was found that joints with good mechanical and electrical properties could be produced. In the studies by Pabst and Groche, the potential for MPW process improvement by influencing the ambient medium was investigated (Pabst and Groche, 2018). It was demonstrated that especially helium in the joining gap has a process-improving effect.

This study aims to address these findings by investigating an easy-to-implement method of enhancing the quality of MPW for aluminium and copper. A nozzle-based inert gas application was developed that significantly improves the quality of MPW joints, particularly in the area of the lower process boundary.

2 Materials and experimental methodology

2.1 Materials

Aluminum and copper alloys with good electrical properties were selected for the present investigation. The materials are specified in **Table 1**. The utilized sheets were 100 mm long

and 40 mm wide. The sheet thickness was 1 mm for copper and 1.5 mm for aluminum. These sheet thicknesses were used to partially compensate the different electrical conductivity coefficients. Before welding, all joining partners were cleaned using isopropyl.

Alloy	Tensile strength [MPa]	Electrical conductivity [MS/m]
AA1050A	106	34,15
Cu-ETP	269	59,24

Table 1: *Materials used for the investigation*

2.2 Experimental setup and welding parameters

All fabricated joints were welded with a BlueWave PS48-16 system from PSTproducts GmbH. The utilized single turn sheet welding coil features an active area of 80 mm x 10 mm and a thickness of 4 mm. The positioning of the joining partners above the active area of the coil, the GFRP spacers used to adjust the acceleration distance and to define the overlap of 30 mm and the positioning of the inert gas nozzle are shown in **Fig. 1**. The hold-down-device is not illustrated. It is fastened prior welding using the four visible bolts. Helium is used as the inert gas, the flow rate is 8 l/min. The experiments were conducted with and without an inert gas flow. The acceleration distances were 1.5 mm and 2.5 mm. The discharge current was varied in three increments: 232 kA, 321 kA and 392 kA. The system frequency for all experiments was 19,9 kHz. AA1050A was always used as the flyer, as researchers have demonstrated that better electrical and mechanical joint properties can be achieved in this way (Psyk et al., 2017).

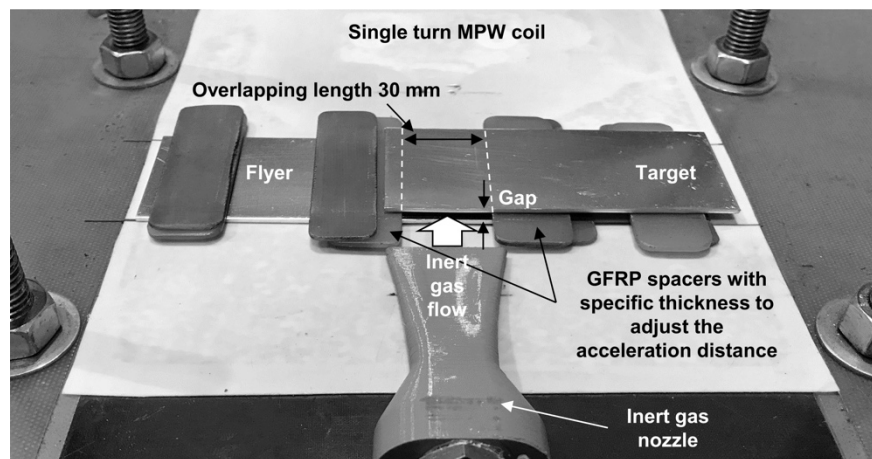


Figure 1: *Experimental setup of MPW with inert gas nozzle. Overlap area centered over the active coil area and defined using markings and GFRP spacers.*

2.3 Mechanical characterization

The welded joints were tested quasi-statically using a universal tensile testing machine (Z100, ZwickRoell GmbH & Co. KG). The test speed was 5 mm/min. No shims were used,

but the clamps were adjusted according to the acceleration distances and sheet thicknesses. After the mechanical testing, the determined maximum force was used for further evaluation. The mean, the minimal and the maximal values from three tests for each parameter variation were considered for the determination of the mechanical joint quality. The location of the failure was also documented. If all samples of a parameter variation fail in the base material, this is indicated in the diagram as a "base material" failure location.

2.4 Electrical characterization

All welded joints were examined using the four-wire Kelvin measurement method according to **Eq. (1)**. The resistance of a joint ($R_{Measurement}$) was determined based on the measured voltage difference (ΔV) and the applied constant current (I) of 50 A. For the evaluation, the measured value of the respective joint was divided by the mean value (R_{Ref}) of the resistances of the two pure joining partners, see **Eq. (2)** and **Eq. (3)**. All values lower than one are consequently better than the electrical quality of the pure sheet materials. This is possible because the MPW joint is a lap joint. Thicker conductor cross-sections can therefore be present in the overlapping and welded area. To characterize the electrical properties of the joints, five samples for each parameter variation were utilized, and the mean, minimum, and maximum values were determined.

$$R_{Measurement} = \frac{\Delta V}{I} \quad (1)$$

$$R_{Ref} = \frac{1}{2}(R_{Al} + R_{Cu}) \quad (2)$$

$$R_{Calc} = \frac{R_{Measurement}}{R_{Ref}} \quad (3)$$

An MR5-200C micro-ohmmeter (Schuetz Messtechnik GmbH) was used as the measuring device. The distance between the voltage tapping points was 50 mm for all experiments.

2.5 Analysis of weld seam geometry and microstructure

For the investigation of the microstructure after MPW, the investigated joints were cut longitudinally in the middle of the deformed area using a wet cutting grinder. The samples were embedded in cold-curing acrylic resin (ClaroCit from Struers GmbH), ground to a grain size of 4000 and polished with diamond suspension (grain size 1 μ m). The samples were examined using scanning electron microscopy (SEM, Zeiss SEM ULTRA PLUS) and energy dispersive X-ray analysis (EDS, Bruker 6|60). The SEM was operated with a voltage of 20 kV, the working distance was 10 mm and images were captured using the SE2 detector.

The weld seam geometry was analyzed by dissolving the aluminium of welded joints with a sodium hydroxide solution (10 mass% over a period of 24 hours). The remaining weld

seam was subsequently examined and quantified from a top view using a macroscope (Leica Z16).

3 Results and discussion

The fabricated joints were characterized with regard to their electrical and mechanical joint quality. **Fig. 2 a)** shows the mechanical and **Fig. 2 b)** the electrical joint quality as a function of the MPW process parameters and the ambient atmosphere. The effect of the inert gas helium in the acceleration gap on the mechanical joint quality is particularly evident in the range of low discharge currents. Joints manufactured with 232 kA without inert gas withstand hardly any forces in the quasi-static tensile shear load case. The same attempts with inert gas in the acceleration gap enable the fabrication of joints that can withstand more than 4 kN in the same load case. In both cases, the failure occurs in the weld seam. Increasing the discharge current to 321 kA or even 392 kA results in a failure in the aluminum base material regardless of the adjusted gap and the atmosphere inside. Similarly, the electrical joint quality in the range of low discharge currents (here 232 kA) is also strongly influenced. The application of an inert gas causes an improvement in this case, so that a result of almost one is obtained. In the range of larger discharge currents, the inert gas has no significant effect on the electrical joint quality. However, the MPW process parameters have an effect on the electrical joint quality. An increase in the discharge currents from 321 kA to 392 kA results in an improvement in the joint quality. The gap likewise has an impact. Joints produced with a gap of 1.5 mm and 321 kA discharge current perform electrically superior to joints produced with 321 kA and a gap of 2.5 mm.

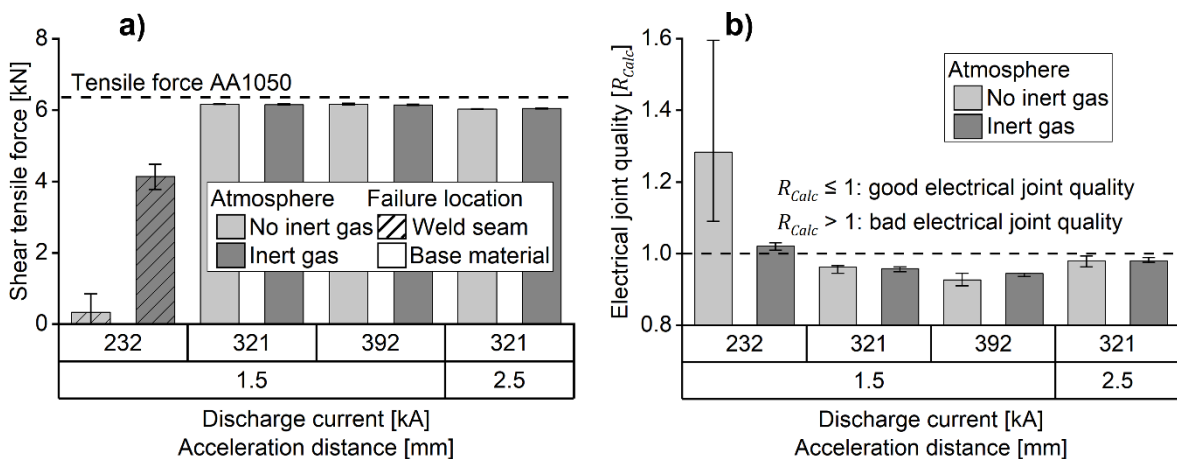


Figure 2: a) Quasi-static shear tensile strength (mean, minimum, maximum) and failure location and b) electrical quality (mean, minimum, maximum) of joints with regard to the applied discharge current (x-axis top line) and the gap prior welding (x-axis bottom line).

To investigate the influence of the inert gas on the mechanical and electrical quality of the joint, the geometry of the weld seam is analyzed. **Fig. 3** shows the top view of weld seams fabricated with different gaps and varying discharge currents with and without inert gas.

In both cases, the inert gas affects the size of the weld seam. In the case of 1.5 mm gap and 232 kA discharge current, the inert gas application results in the complete formation of the weld seam and consequently a significant increase of the weld seam size. This goes alongside the findings on the mechanical and electrical properties of these joints, compare **Fig. 2**. In the case of a 2.5 mm gap and a discharge current of 321 kA, the inert gas causes an enlargement of the weld seam. However, the electrical and mechanical joint properties are not affected, as **Fig. 2** reveals. A possible reason is that the weld seam is already adequately large without inert gas and therefore enables excellent electrical joint properties. The determined area of 67.4 mm² is already larger than the 60 mm² cross-section of the aluminum joining partner.

In order to further investigate the effects of the inert gas on the weld seam formation, the subsequent step consists of analysing cross-sections using SEM. **Fig. 4** shows images of joints fabricated using 321 kA discharge current and 2.5 mm gap with and without inert gas during MPW. **Fig. 4 a)** shows the beginning of the weld seam of a joint welded in an inert gas atmosphere. An MPW typical wavy but non-turbulent interface between the aluminum and copper joining partners is present. A material-contiguous joint was formed between aluminum and copper, as shown by an EDS line scan through a melt pocket near the beginning of the weld. Furthermore, no increase in oxygen was detected, indicating that the jet was able to be completely emitted out of the collision point.

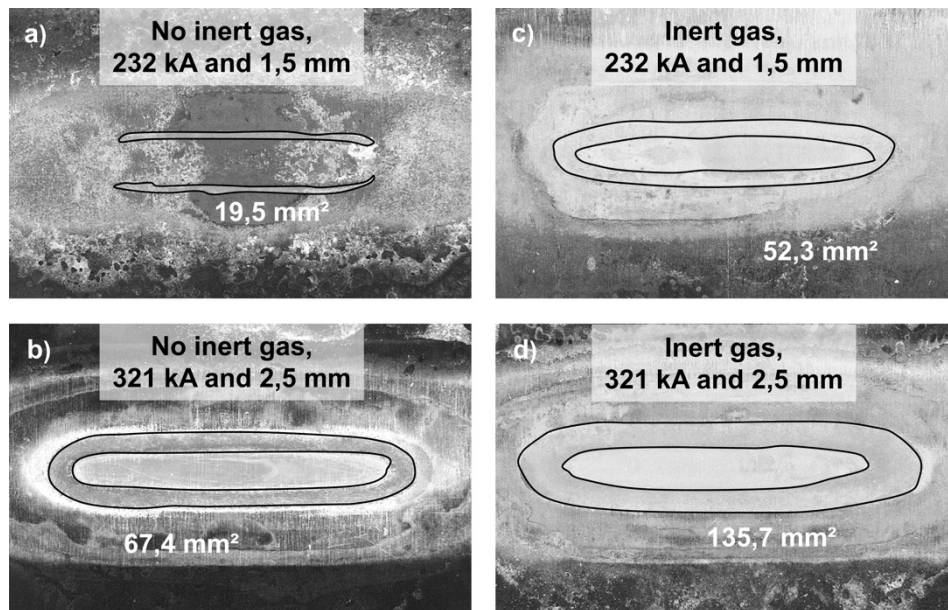


Figure 3: Top view of MPW weld seams joined using different process parameters, with and without a nozzle-based application of inert gas during welding.

A weld seam fabricated with a gap of 2.5 mm and a discharge current of 321 kA without inert gas exhibits entirely different properties at the beginning of the weld seam. A region characterized by the presence of pores and cracks is observed in proximity to the weld seam. This suggests that the joining partners have enclosed the jet through their collision or have at least interacted strongly with it, thereby preventing the complete emission of the jet from the collision point. The EDS line scan in the respective weld seam area corroborates this

finding. An increase in the oxygen content is observable. This finding suggests that incomplete emissions from the plasticized joining partner boundary layers, particularly the aluminum oxide layer, may be a contributing factor.

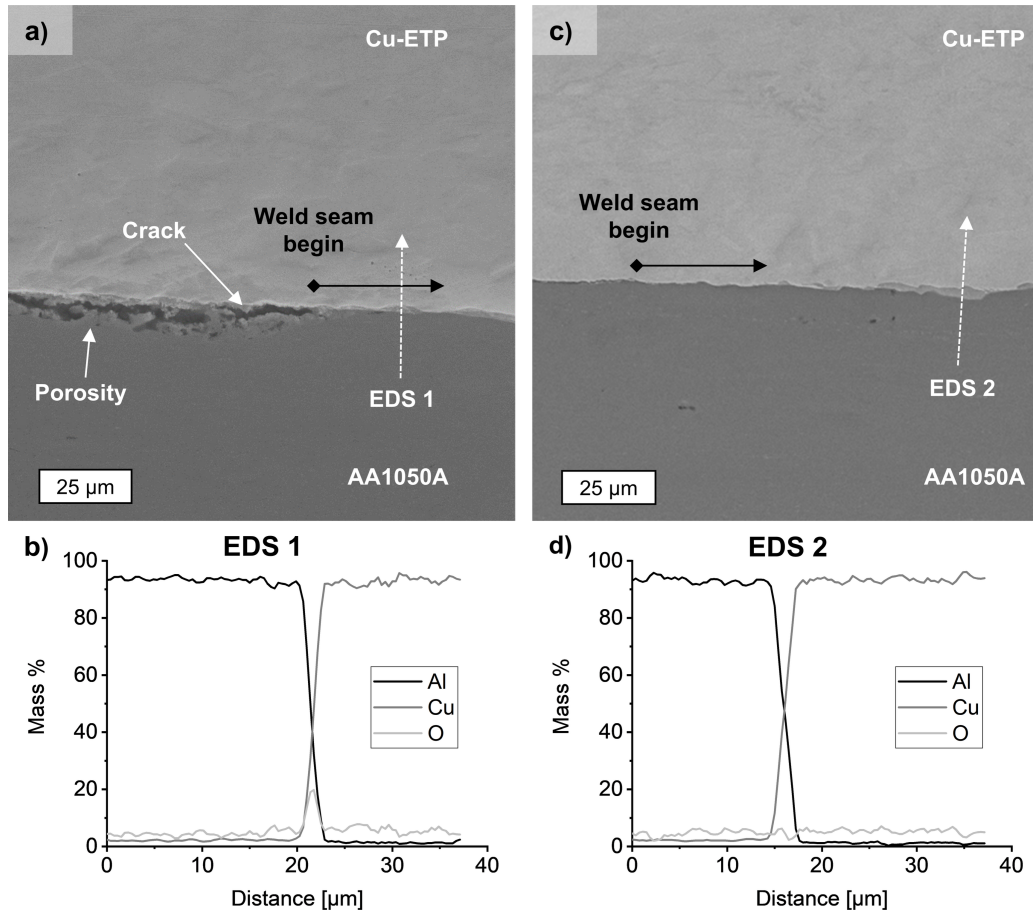


Figure 4: a) SEM image of the beginning of the weld seam and the position of EDS 1 line scan of a joint fabricated using 321 kA discharge current and a 2.5 mm gap without inert gas supply, b) result of EDS 1 line scan, c) SEM image of the beginning of the weld seam and the position of EDS 2 line scan of a joint fabricated using 321 kA discharge current and a 2.5 mm gap with inert gas supply, d) result of EDS 2 line scan.

4 Conclusion and Outlook

Hybrid aluminum-copper joints could be manufactured using MPW in a broad range of process parameters. Appropriate evaluation methods enabled the joints to be confirmed as featuring good mechanical and electrical properties. Interactions between the electrical and mechanical joint properties, the MPW process parameters and the ambient atmosphere created by a nozzle-based approach were successfully investigated. It has been proven that a nozzle-based application of the inert gas helium in the range of the lower process limit is associated with a substantial increase in joint quality. The reason is attributed to the influence of the inert gas helium on the jet. A detailed analysis of the microstructure revealed that the

presence of an inert gas within the joining gap enabled the jet to emerge from the collision point unimpeded, devoid of any interaction with the joining partners.

A thorough analysis is necessary to ascertain the precise impact of the inert gas on the processes occurring within the joining gap. A crucial aspect that necessitates investigation is the influence of the inert gas present within the joining gap on the exothermic reaction of the boundary layers of the joining partners during their collision. A further examination of the interactions between the deformation and the resulting velocity of the flyer and the atmosphere in the joining gap is also warranted. In addition, the targeted fluidic influence of the jet by an optimized inert gas nozzle should be investigated. Furthermore, a spatially resolved analysis of the joint properties might be conducted to investigate the interactions between the local microstructure, its manipulability using inert gas during welding, and the resulting global electrical, mechanical and corrosion properties of the joint.

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A Novel Coil Structure for Metal Sheet Magnetic Pulse Welding

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Abstract

Magnetic pulse welding (MPW) of metal sheets is an efficient and environment-friendly room temperature solid-state welding technology. However, the process involves significant energy dissipation. Limited by the current coil structure, high energy is required to achieve metallurgical welding of metal sheets. To address this issue, novel coil principle and structure has been proposed. The current is simultaneously concentrated into a section of conductor by a unique circuit design, achieving current summation and amplification, thereby enhancing energy efficiency. Numerical simulations were performed to explore the effects of coil parameters on current density, Lorentz force, and flyer sheet motion, revealing that the coil structure can amplify current by 1.5 to 5.3 times as coil turns increase from 2 to 7. The flyer sheet velocity exceeded the critical welding velocity. Experimental results show successful welding of 1060 (1.0 mm) / DP450 (1.0 mm) at 6.05 kJ and 5754 (1.0 mm) / DP600 (1.0 mm) at 9.8 kJ, with all interfaces displaying metallurgical characteristics. These findings confirm the excellent welding capability of the novel coil at low energy, which results from a significant current amplification effect. This reduces the stringent requirements for magnetic pulse discharge equipment, thereby facilitating the further industrial application of MPW technology.

Keywords

Magnetic pulse welding, Novel coil structure, Metal sheet

1 Introduction

Magnetic pulse welding (MPW) is an environmentally friendly and efficient room-temperature solid-state welding technology. It relies on the electromagnetic force to induce high-speed deformation and collision of the metal plates to be welded, thereby achieving the welding of dissimilar metals. Compared to traditional fusion welding, the advantages of MPW include: a higher tolerance for the differences in physical and chemical properties between dissimilar metals, and a welded interface with no heat-affected zone and minimal intermetallic compounds, among other defects (Kapil and Sharma, 2015). The MPW process for dissimilar metal plates is closely related to the matching of the inherent parameters of the magnetic pulse discharge equipment (such as low inductance, low resistance, and appropriate capacitance) and the coil tooling. However, the development of magnetic pulse discharge equipment involves high-voltage electrical systems, pulse power technology, and other advanced challenges (Liao et al., 2020; Liu et al., 2024). Therefore, to further promote the application of MPW technology in industrial settings, it is essential to conduct fundamental research on novel coil tooling structures.

Traditional E-shaped (Liu et al., 2022) and H-shaped (Li et al., 2021) coils have limitations in terms of low bonding capacity, short service life, and insufficient energy utilization, which restrict the further industrial application of MPW. Zhang et al. (Zhang et al., 2019) found that when the gap between the return arm and the work arm of the traditional E-shaped coil is approximately three times the width of the work arm, the reverse magnetic field generated by the return arm has the least effect on the deformation behavior of the workpiece. To enhance coil efficiency and bonding capacity, Khalil C et al. (Khalil et al., 2020) proposed the O-shaped coil, which exhibits a lower inductance level and allows for the production of higher-quality joints under the same discharge energy. However, much of the current research still focuses on improving traditional coil structures. The method of current flow remains largely unchanged, with large pulse currents still directly passed through a conductor, resulting in significant energy losses.

Therefore, the novel MPW coil principle proposed in this study no longer adheres to the limitations of the flat coil structure. Based on the unique configuration of a coaxial cable, the sinusoidal decaying oscillating discharge current flows through multiple paths nearly simultaneously in the current convergence part, resulting in the superposition and amplification of current intensity (amplitude). This reduces energy losses and improves energy utilization. The effects of the positioning of the coil convergence part and related structural parameters on welding performance were investigated by numerical simulations combined with experiments, ultimately achieving high-quality weld joints.

2 Coil structure and principle

The principle of the novel coil structure for sheet metal magnetic pulse welding proposed in this study is shown in **Fig. 1(a)**. Through circuit design, the energy from the multi-turn coil is concentrated into a section of conductor, thereby improving energy utilization efficiency.

To achieve this concept, as shown in **Fig. 1(b)**, a multi-turn ring circuit is designed, and the converging conductor section is connected in series with each of the ring circuits. This design ensures that the current in each turn of the ring circuit not only flows through each other but also enables superposition and amplification within the converging conductor section. Coaxial cables, with their inherent concentric conductor arrangement, are the optimal material choice to realize this design, as depicted in the schematic diagram in **Fig. 1(c)**. Two different types of novel coil structures have been designed, as shown in **Figs. 1(d) and (e)**, with the converging section placed at either the center or the side of the coil. The current first enters the core wire section of the coaxial cable and directly flows to the current transfer block at the end of the coil. The current is then transferred to the braided wire of the coaxial cable. Since each turn of the braided wire is connected in series with the converging conductor section, the current in each turn of the braided wire flows through the converging conductor section, thereby achieving current superposition and amplification. It is worth noting that due to the instantaneous nature of the current, the current in the converging section is almost simultaneously superimposed, avoiding any situation where the current flows through the converging section at different times.

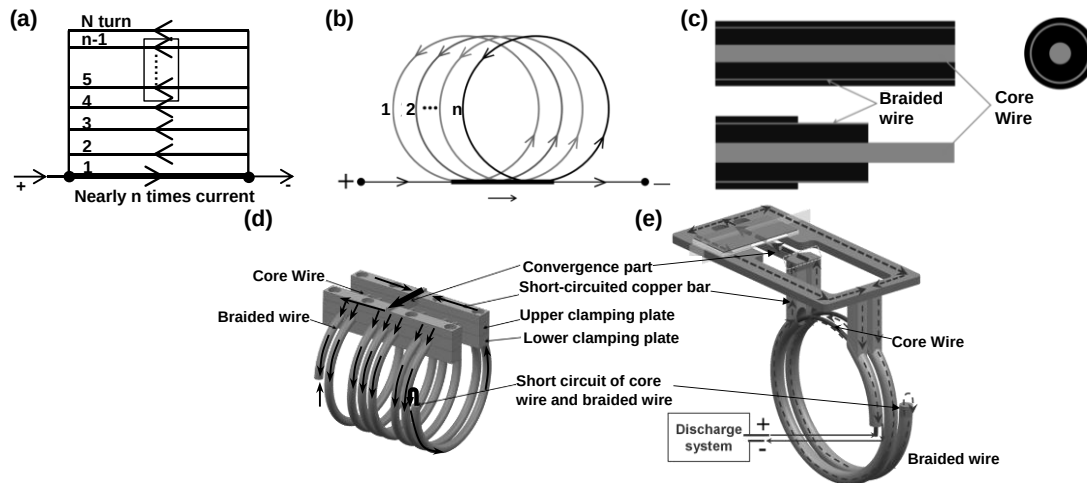


Figure 1: Schematic diagram of: (a) new coil principle, (b) coil structure, (c) coaxial cable; 3D model of (d) MP coil, and (e) SP coil

3 Experiment and numerical simulation

3.1 Material selection and equipment

In this study, the materials selected for investigating the welding capability of a novel coil structure include 1060 pure aluminum (thickness: 1 mm), 5754 aluminum alloy (thickness: 0.5 mm and 1 mm), DP450 steel (thickness: 1 mm), and DP600 steel (thickness: 1 mm). All materials have dimensions of 40 × 80 mm. Cu-OFHC was selected as the structural material for the coil, except for the coaxial cable. The material properties of all the selected materials are summarized in **Table 1**.

Material	Yield strength (MPa)	Tensile strength (MPa)	Elongation (%)
1060	31	67	54
5754	90	214	35
DP450	324	445	27
DP600	360	678	26.6
Cu-OFHC	149	261	16

Table 1: Properties of materials by tensile testing used in this study

The magnetic pulse discharge equipment used in this study is the self-developed EM20/20 and EM50/18 from Harbin Institute of Technology, which has a 100 μF capacitor and a maximum discharge energy of 20 kJ. EM20/20 and EM50/18 equipment has been successfully applied in industrial production and has high practical application reliability. Various discharge energies were used to study the coil performance, with a collision gap between flying sheet and base sheet set at 2 mm.

3.2 Numerical simulation

In this study, a multi-field coupling numerical simulation of the electromagnetic field, structural field, and force field during MPW overrun was conducted using ANSYS/LS-DYNA software. The meshing methods of MP and SP coils are shown in **Fig. 2**. The coil structure parameters used in this study are listed in **Table 2**. Due to the relatively low inductance of the 4-turn MP coil, it can be paired with EM50/18 that has a higher capacitance (equipment with higher inductance) to achieve optimal performance. In contrast, the 7-turn MP coil, with its higher inductance, is more suitable for EM20/20 with lower capacitance (equipment with lower inductance). Therefore, the MP coil was studied using both types of equipment. For the SP coil, due to the extensive research on its process parameters, only one type of equipment was used for the study. Considering the high strain rates of flyer sheets involved in MPW, the material models for the flyer sheet is based on a simplified Johnson-Cook constitutive model (**Eq. 1**). The parameters for the simplified Johnson-Cook constitutive model of each material are given in **Table 3** (Li et al., 2022; Johnson et al., 1983; Khalil et al., 2021;).

$$\sigma = (A + B\varepsilon^n) \cdot \left(1 + C \ln \frac{\dot{\varepsilon}}{\dot{\varepsilon}_0}\right) \quad (1)$$

Among them, $(A + B\varepsilon^n)$ represents strain strengthening, and $\left(1 + C \ln \frac{\dot{\varepsilon}}{\dot{\varepsilon}_0}\right)$ represents strain rate strengthening.

Coil	Turn	Diameter	Pitch
MP	4-EM50/18 and 7-EM20/20	-	-
SP	2-7	150-250	15-45

Table 2: The coil structure parameters studied in this study

Material	A	B	n	c
1060	35.5	68.7	0.14	0.015
5754	93.3	166.8	0.28	0.0281
Cu-OFHC	90	292	0.31	0.025

Table 3: The J-C model parameters of the materials used in this study

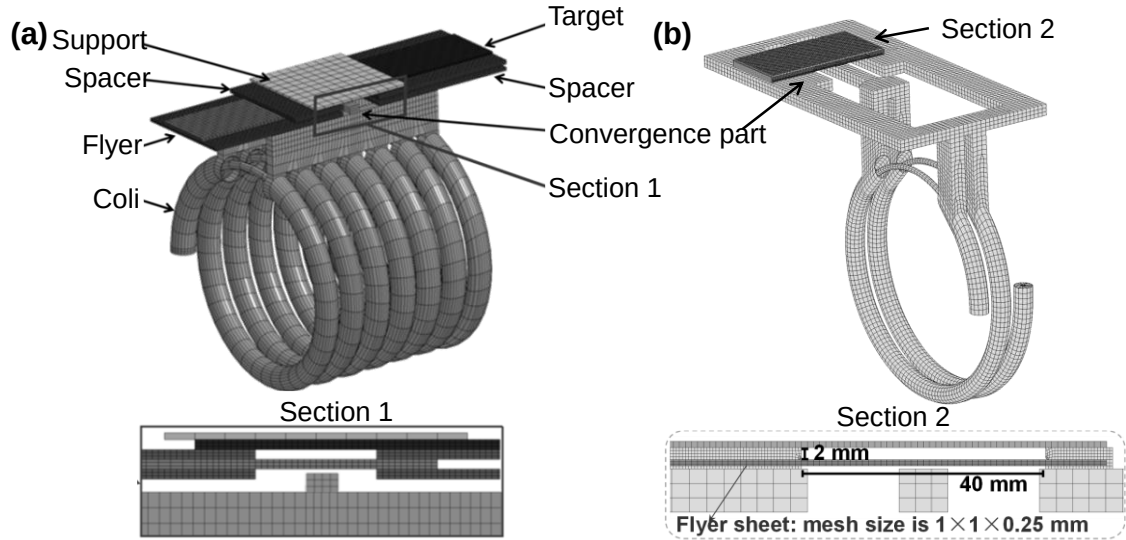


Figure 2: Grid division results of coils: (a) MP, (b) SP.

4 Results

4.1 Numerical simulation results

The numerical simulation results of MP coil were analyzed first. The discharge current curve of 7 turns MP coil is shown in **Fig. 3(a)**. The peak current at the coaxial cable part reaches 69 kA when the discharge voltage is 9 kV (4.05 kJ), while the peak current increases to 368 kA at the convergence part, achieving a 5.3-times amplification. The 1/4 discharge current period is 20.5 μ s. Within such a short period, the first current peak is reached, which imparts a very high velocity to the flyer sheet. The velocity curves of the flyer sheet under different discharge voltages were obtained, as shown in **Fig. 3(b)**. It can be observed that the collision speed of the flyer sheet is proportional to the discharge voltage (9-12 kV). At a discharge voltage of 12 kV, the maximum collision velocity of the flyer sheet reaches 265 m/s. To investigate the effect of the number of turns on the MP coil, the discharge current of the 4-turn MP coil and the speed imparted to the flyer at different voltages are shown in **Fig. 4**. From the figure, it can be observed that, at higher energy levels, the current in the convergence section of the 4-turn MP coil remains lower than that of the 7-turn MP coil. Additionally, at lower discharge energies, the 7-turn MP coil imparts a higher speed to the flyer. The higher inductance of coils with more turns is disadvantageous for welding. Therefore, further studies on coils with more turns were not conducted.

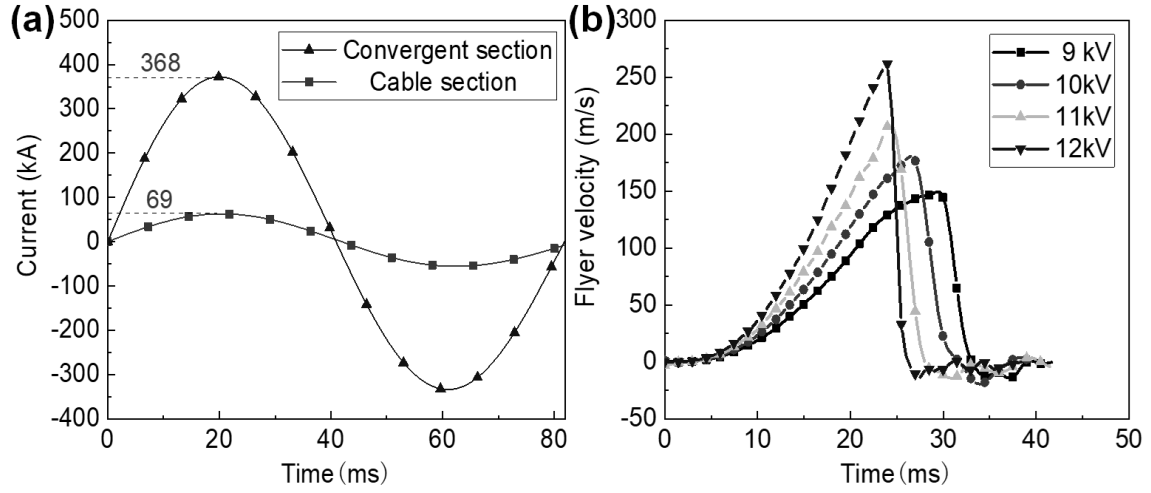


Figure 3: 7 turns MP coil (a) Current curve at 4.05 kJ, and (b) collision speeds of flying sheet at 9-12 kV

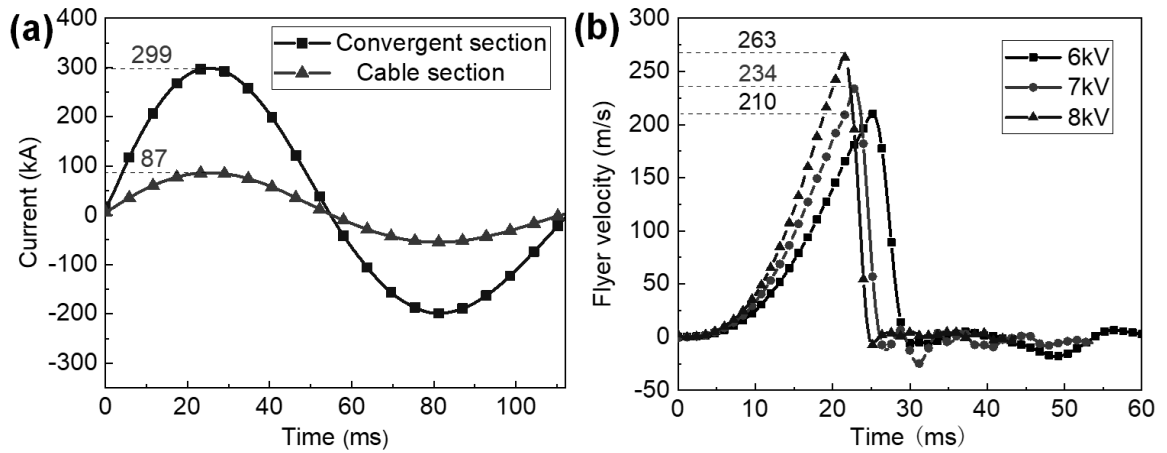


Figure 4: 4 turns MP coil (a) Current curve at 5.04 kJ, and (b) collision speeds of flying sheet at 6-8 kV

Due to the different positions of the convergence part, both the inductance and the current distribution vary. More importantly, compared to the coil structure placed within the convergence part, the SP coil allows for the welding of a wider range of plate sizes, as illustrated in the coil structure diagrams in **Figs. 1(d) and (e)**. Therefore, simulations were conducted to evaluate the discharge current and the collision velocity of the flyer sheets for SP coils with different turns, diameters, and pitches. **Fig. 5** presents the discharge current and the collision velocity of the flyer sheet for different SP coil parameters. **Fig. 5(a)** shows the discharge current with varying turns, under a discharge energy of 9.8 kJ. As turns increases from 2 to 7, the current in the coaxial cable part decreases from 126 kA to 67 kA. This reduction in current is primarily due to the increased resistance and inductance resulting from the longer length of the coaxial cable. The current in the convergence part reaches its peak at 270 kA with 6 turns, while the current in the coaxial cable part is only 74.5 kA, with a discharge current period of 108.8 μ s. However, this current begins to decrease when the number of turns reaches 7, due to the increase in coil inductance and resistance. As shown

in **Fig. 5(a)**, further increases in the number of turns have little effect on the current amplification. The main reason is that an increase in the number of turns leads to a rise in both resistance and inductance, which in turn causes a decrease in the base current in the coaxial cable. Consequently, a higher amplification factor is unable to drive the current in the convergence section beyond the current level achieved with a lower number of turns. **Fig. 5(b)** shows the flyer sheet velocity, where the 4-turn coil yields the highest velocity. The main reason is that the peak current for the 4-turn SP coil (255 kA in the convergence part and 94 kA in the coaxial cable part) is slightly lower than that of the 6-turn coil, but the discharge current period is shorter. As a result, the flyer sheet experiences higher acceleration over the same collision distance, achieving a higher collision velocity.

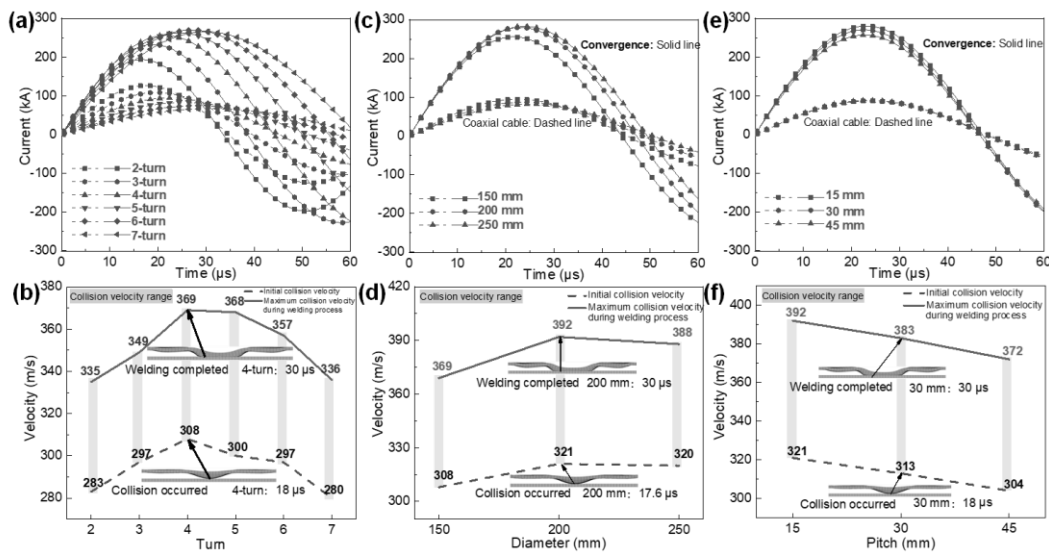


Figure 5: (a)-(c) The current curves of coils with different turns, diameters, and pitches at 9.8 kJ, (d)-(f) corresponding flying sheet speed

To further investigate the effect of coil diameter on discharge current for the 4-turn SP coil, **Fig. 5(c)** presents the results. The discharge current is found to be highest when the coil diameter reaches 200 mm. Considering both the flyer sheet velocity and the production flexibility, coils with a diameter of 250 mm are excluded from further investigation due to their large size and relatively low sheet velocity. Further research was conducted on the 4-turn SP coil with a 200 mm diameter, as shown in **Figs. 5(e) and (f)**. As the pitch increases, the coil current gradually decreases. When the pitch is 30 mm, the peak current of the coil is 270 kA (87 kA in the coaxial cable part), achieving a current amplification of 3.1 times. The flyer sheet velocity reaches a maximum of 383 m/s, which is sufficient for achieving the required collision velocity for sheet welding. Based on the overall considerations, a 4-turn SP coil with a 200 mm diameter and a 30 mm pitch is selected for production.

As shown in the **Fig. 6**, compared to the MP coil, the convergence section of the SP coil experiences a smaller Lorentz force at higher energy because it does not interact with the exposed core section of the coaxial cable. This indicates that the SP coil can withstand higher discharge energy without failure, while the convergence section of the MP coil is more prone to deformation and failure. Furthermore, the coaxial cable section of the SP coil, due to its concentric structure of the braided and core wires, results in a cancellation of the

Lorentz force. Additionally, the exposed core section of the coaxial cable is reinforced with epoxy resin in actual production, which extends the coil's lifespan. Therefore, the SP coil is stronger than the MP coil.

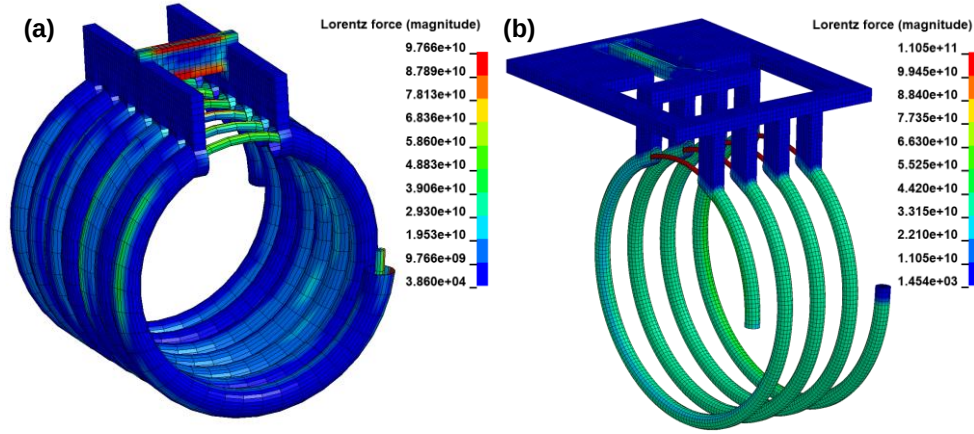


Figure 6: Lorentz force distribution map of (a)MP coil at 4.05 kJ, (b) SP coil at 9.8 kJ

4.2 Experiments

In **Fig. 7(a)**, it can be seen that the width of the flyer sheet is constrained by the length of the coil convergence part of the MP coil. Compared to the SP coil shown in **Fig. 7(b)**, although the MP coil has a higher current amplification factor, the SP coil is not limited by the size of the sheet metal processing. **Figs. 7(d) and (e)** show the actual current waveforms measured by the Rogowski coil for the both coil tools. The actual current amplification factor and cycle are similar to the simulation results, thereby confirming the reliability of the simulations results. The welded joints were observed for microstructural at the interface, as shown in **Figs. 8(a) and (d)**. All joints displayed a waveform interface, and Energy Dispersive Spectroscopy (EDS) analysis revealed significant elemental diffusion, confirming the ability of both types of coils to achieve dissimilar metal metallurgical welding.

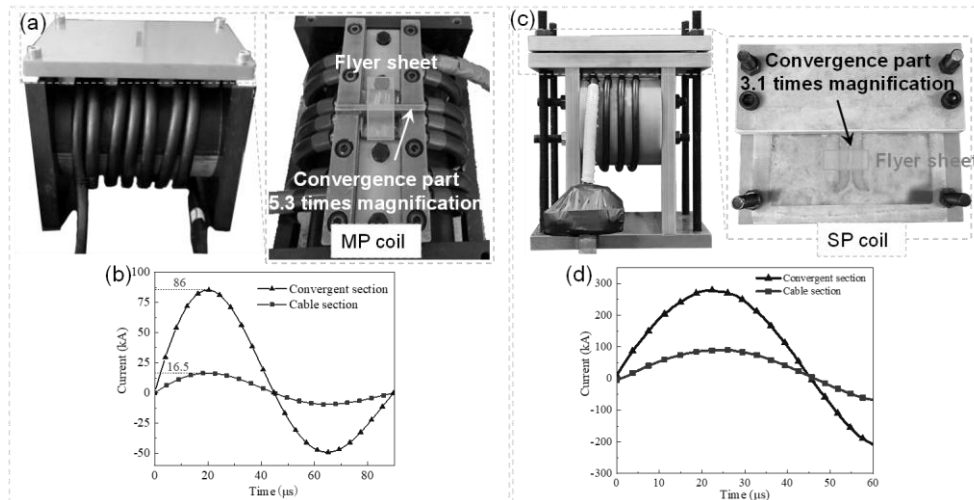


Figure 7: (a) MP coil tools, (b) measured current by the Rogowski coil at 2 kV, (c) SP coil tools, (d) measured current by the Rogowski coil at 14 kV

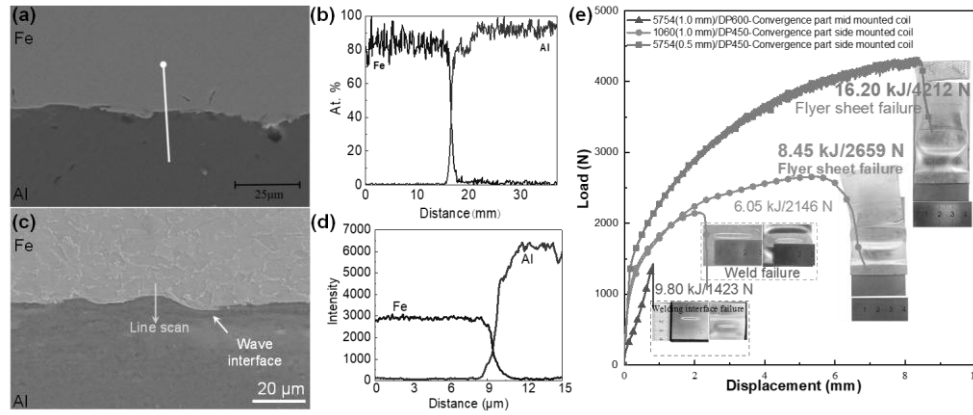


Figure 8: Microstructure and EDA line scan of weld interface based on MP and SP coil at 9.8 kJ: (a) and (b): 5754 (1.0 mm)/DP600, (c) and (d): 1060 (1.0 mm)/DP450; (e) Mechanical properties of joints

Fig. 8(e) presents the mechanical properties of the welded joints. It can be seen that the MP coil achieved a metallurgical weld between 1mm 5754 and DP600 at 9.80 kJ, and the joint's tensile load was 1423 N. Due to the interaction of magnetic fields between the convergence part and the coaxial cable of MP coil, as shown in **Fig1. (d)**, excessive discharge energy can cause deformation and failure of the convergence part. As a result, higher-energy welding experiments cannot be conducted. In contrast, the SP coil allows for a variety of experiments. From the **Fig. 8(e)**, it is evident that high-quality joints can be achieved at higher energy levels. At 8.45 kJ, the 1060 (1.0 mm) / DP450 joint and the 5754 (0.5 mm) / DP450 joint all fractured on the aluminum side. These results confirm the usability of the coil structure and its potential for further development.

5 Conclusions

- A novel coil principle and structure are proposed, utilizing the unique electrical structure of coaxial cable in combination with current circuit control to achieve current accumulation and amplification in the convergence part. Additionally, this study introduces both MP and SP coil structures.
- Numerical simulations were employed to analyze the effects of different structural parameters of coil on the current and flyer sheet velocity. The 7-turn MP coil and the 4-turn/200 mm diameter/30 mm pitch SP coil can achieve current amplifications of 5.3 and 3.1 times, respectively.

- Both coil types are capable of achieving metallurgical welding of dissimilar materials. The SP coil, due to its structural design, can withstand greater reverse Lorentz forces, enabling welding of various material combinations across a range of energy levels.

Acknowledgments

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Recent Evolutions of Electromagnetic Pulse Processes in the Industrial Mass Production

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Abstract

Electromagnetic pulse processes used to be niche production applications. Apart from research and development (R&D) use, pulse generators were found in few production facilities and low production volume applications only. During the recent years, significant advances in certain key applications have been made.

For example, the crimping of cables and the welding of bus bar connectors are both high volume applications in the automotive industry. The machines (pulse generators and tool coils) as well as the processes have undergone extensive certification routines for quality assurance and meet the requirements for mass production. The manufacturers' specifications have been developed with regard to the particularities of electromagnetic pulse processes and machines. Enabled by the continuous development, it became even possible to enter markets with high production rates and volumes.

The success for these applications does not only result in an increased number of production machines, but also act as 'door opener' for further applications in and beyond their current field of application. Next steps are for example the development of a crimping and tube welding operation which joins a copper connector via an aluminium tube (welding) to an aluminium cable (crimping).

Keywords

cable crimping, welding, bus bar, industrial production

1 Introduction

Electromagnetic pulse processes are on their way to mass production. This was started in the late last century, when numerous more or less realistic applications have been thought of.

However, the necessary equipment was not at hand yet and the applications were too ambitious in many cases. Hence, this hype turned into disillusionment. As a consequence, electromagnetic pulse processes are stigmatized as not relevant for industrial (mass) production. This is why many production companies still have pulse generators from that era, but barely use them and don't invest in process research. Pulse process equipment has been almost exclusively used for university research and few funded industrial research projects. However, since then, many aspects have been improved:

Capable pulse generators with durable components are available. Switches, capacitors and cables have been vastly improved with regard to service life and stability, and need to be replaced far less frequently. Pulse switches can reach several hundred thousand pulses to more than one million pulses before they are replaced. They don't contain mercury, which was the case with certain types. The pulse cables and capacitors are capable of far more than one million pulses. This compensates the fact that these core components are very expensive and reduces the production costs per part.

The development of the tool coils aims for three goals: A more durable design, decreasing the costs, and a broad variety of geometries to fit a wide range of applications. They are usually designed so that the wear parts are small, cheap and easy to replace, whereas the rest is robust for a long service life. Sheet metal welding is possible with workpiece sizes between several millimetres and more than a metre. The part of the tool coil that actually induces the eddy currents into the flyer is the main wear part. The diameter for tube welding and crimping is in the same range as the length for sheet metal welding. Here, the field shapers are the main wear parts. Crimping and welding of a large variety of work piece geometries is possible. The geometry of the tool coil or field shaper is adapted correspondingly. Furthermore, tool coils for the industrial mass productions usually process more than one work piece per pulse to reduce costs.

The measurement of the pulse current is the most important parameter to monitor both pulse generator and process. With an adequate design of the equipment, the repeatability of the peak discharge current and the discharge current frequency can be as precise as $\pm 0.15\%$, including measurement errors. With careful design of the workpieces, the upstream processes (such as cleaning) and the automation system, the discharge current measurement and its evaluation alone are sufficient to monitor both the pulse generator and the process. The process capability of electromagnetic pulse processes can achieve more than the required threshold value for mass production processes in multiple applications and industry branches.

The knowhow regarding the design of automation systems, which are crucial for mass production, has improved greatly. Robust sensors, actuators and bus systems enable precise movements even in close vicinity to the tool coil.

2 Market overview

The progress of pulse generators, tools and auxiliary equipment enables more and more industrial products to be manufactured by electromagnetic pulse processes. Once one

product of one manufacturer is processed using electromagnetic pulse technology (EMPT), other customers with identical or similar products follow. In many cases, the knowledge of the capabilities of electromagnetic pulse processes is spread within a company and adapted to other products. This may also involve the step from crimping a product to welding another.

This opens the markets and leads to an increasing demand for the machines. **Fig. 1** shows the delivered pulse generators within the given time periods by PSTproducts since the foundation in 2003. The number of tool coils is larger, because several customers have more than one tool coil for one pulse generator. As the value of tool coils has a very large margin and is usually lower than the value of the pulse generator (mostly about 20 %-50 %), the number of pulse generators provides a clearer picture. Furthermore, pulse generators with multiple tool coils are, without exception, mainly used for R&D purposes. It is worth mentioning that the share of production machines keeps increasing.

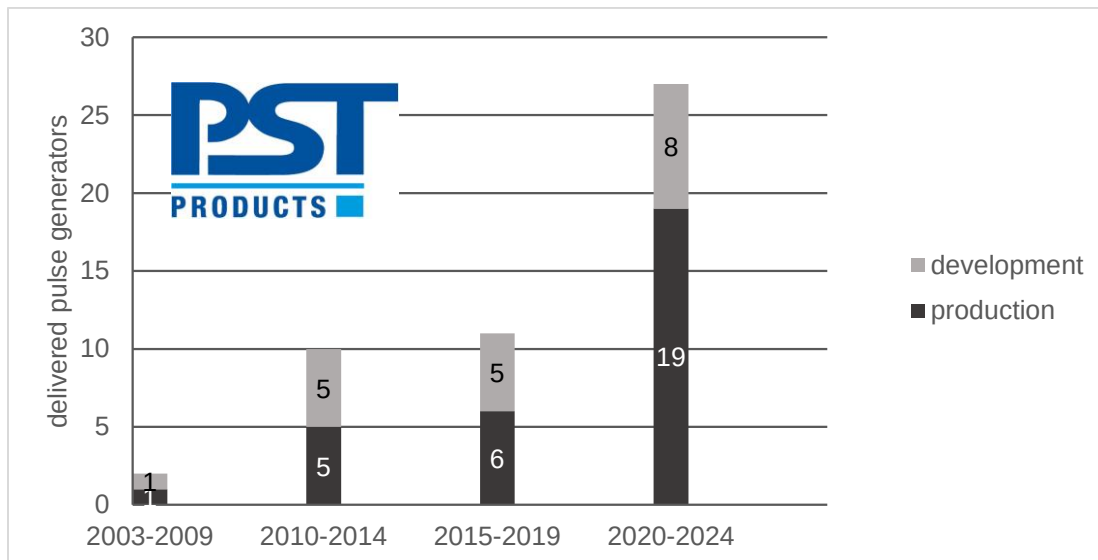


Figure 1: Number of pulse generators delivered by PSTproducts during given time periods.

The combination of a five-year time period was chosen because of its smoothening effect. Due to the low number of machines, their sometimes long production time and effects due to different accounting years, the resulting diagram on a yearly basis would require extensive explanation. Despite challenges, e.g., 2008 financial crisis, the number of delivered pulse generators keeps increasing. This is still due to a relatively small number of applications. However, this number keeps increasing, too.

After the foundation of PSTproducts in 2003, it took five years until the first pulse generator for mass production was delivered. At this early stage, most of the machines were bought by universities and labs for R&D purposes. Starting with 2009, more and more commercial enterprises became aware of and interested in the electromagnetic pulse technology. Whilst some of them started with pulse generators and tool coils for development purposes (with the option of upgrading the R&D machine to a production machine later), others started with mass production equipment right from the beginning after the trials.

The data shown here contains a delay which is worth mentioning: The time span between the first trials, either at PSTproducts' shop floor or at the customers' site with own equipment, and the delivery of the first machine ranges from two years up to nine years or more in rare cases. The reasons are manifold: economic factors, extensive tests (e.g., fatigue tests), complex development task, to name only a few.

3 Examples

Many applications and products are still under development and covered by nondisclosure agreements (NDA). In certain cases, this applies to parts in production, too. In this chapter, three undisclosed examples are described.

3.1 Pressure vessels for heat pumps

In contrast to heat pump systems with conventional refrigerants such as R-134A, new systems with CO₂ operate at far higher pressures (max. 2 MPa vs. more than 100 MPa), thus requiring all components to be far stronger, for example by increasing the wall thickness. This stands in contrast to the requirement of being lightweight in mobile applications such as cars. Electromagnetic pulse welding offers the great advantage of not weakening the workpieces, for example by heat affected zones. This allows for a decrease in wall thickness and thus weight. **Fig. 2** shows a pressure vessel which consists of a tube and two lids. The tube is welded onto the lids with two individual but sequential pulses. The diameter is usually in the range around 80 mm.



pressure vessel



start of first trials:	2014
first production machine:	2016
machines sold by mid-2025:	3

Figure 2: Pressure vessel for CO₂ heat pumps.

During the development phase, the wall thickness could be reduced due to the better dynamic strength properties of EMPT welded joints. The geometry of the joint was adapted for a further improved (fatigue) strength.

3.2 Power bus bars

Due to the electrochemical requirements, rechargeable Lithium-Ion batteries usually have a (nickel coated) copper connector and an aluminium connector. Irrespective of the design of the cell pack, these two metals need to be connected at some place. This can be realized by hybrid power bus bars, which exist in a large variety of geometries. **Fig. 3** shows an example. The thickness of the metals is typically between 1 mm and 3 mm, but may be up to more than 5 mm. Conventional joining processes, such as screws and bolts, have disadvantages regarding electrical resistance, creep strength, cycle time or joint costs. An electromagnetic pulse weld offers excellent mechanical and electrical properties.



power bus bar



start of first trials:	2013
first production machine:	2015
machines sold by mid-2025:	7

Figure 3: Power bus bar (example, not an actual production part).

The bus bars are comparably small with a typical width of some tens of millimetres. At the same time, they are mass products which need to have low manufacturing costs. The solution are sheet welding tool coils which weld more than one bus bar during one pulse. Depending on the bus bar geometry, more than ten individual welds can be produced per pulse. The capacity of the pulse generator needs to be increased only moderately, typically less than 10 % per additional welding.

3.3 Cable crimping

The mechanical crimping of cables is trivial for small cross sections of 10 mm² and less, but becomes increasingly challenging with larger cross sections. The friction between the individual wires prevents the crimping force from the outside from reaching the centre. The compression shock wave from electromagnetic crimping however reaches the whole cross section, resulting in a homogeneous and good compression of the cable. Hence both the mechanical pull-out strength as well as the electrical resistance show superior values, even after many mechanical and temperature cycles. As a result, these connections can be smaller than conventionally crimped cables.



EV cable



start of first trials:	2019
first production machine:	2021
machines sold by mid-2025:	13

Figure 4: Two different cable cross sections: 150 mm² (left, before crimping) and 70 mm² (right, after crimping).

Besides the crimping of the cable core, the shield connection is also produced by electromagnetic pulse crimping. The tool coil and the pulse generator are the same, but a different field shaper and a lower discharge current are used.

In this particular case, the electromagnetic pulse technology enables a new development: In order to save weight, aluminium cables are preferred over copper cables. Still, the connector needs to be made from copper. Crimping a copper connector to an aluminium cable is possible, but after only a few thermal cycles the joint becomes ineligible. The solution is provided by an additional aluminium-copper weld. One portion of an additional aluminium sleeve is welded to the copper connector. The other portion of the sleeve is crimped onto the aluminium cable. Depending on the diameter and geometry, both crimping and welding are possible during one pulse with one field shaper only.

4 Summary and outlook

The development of the equipment for electromagnetic pulse processes has enabled a wide spread of this manufacturing technology. Three exemplary products and processes in mass production have been described: tube welding, sheet metal welding and crimping. They prove that pulse generators and the joining processes have made their way into high volume manufacturing.

The described mass production examples have in common that there is no online process monitoring, except for the measurement of the discharge current. In some applications, even inline non-destructive testing after joining is omitted. Instead, random samples are taken for destructive testing. Hence the upstream processes (e.g., cleaning), the discharge current measurement and the stability of the tool play an important role for quality assurance. This is why the equipment has to undergo extensive certification routines to make sure that all components meet the high quality requirements for mass production.

As a conclusion it can be stated that although electromagnetic pulse processes and equipment are used in mass production, there are still open tasks left for future development, especially regarding online and inline monitoring and testing.

Design criteria for a High Power Pulsed generator for the delamination of laminate structures

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Abstract

Material assemblies combining composites, metals, and ceramics are increasingly used in structural applications due to their high strength-to-weight ratios and durability. However, disassembling such multilayered systems without damaging the constituent materials remains a challenge, particularly for recycling or repair purposes. Magnetic pulse technology offers a promising solution by generating transient mechanical loads through Lorentz forces induced in conductive layers. This paper aims to define the operating conditions of a high-power pulsed (HPP) generator capable of inducing controlled delamination in laminate structures. A simplified model of the current delivered by an HPP generator is introduced, and its temporal evolution is related to the mechanical load applied to the laminate. Previously established delamination criteria, derived from a one-dimensional wave propagation analysis using the method of characteristics, are reinterpreted to guide the design of an HPP generator. The required frequency and magnetic field amplitude are then expressed as functions of the laminate's acoustic and geometric properties. The results provide a set of design criteria for HPP generators in magnetic pulse disassembly applications, enabling effective and selective separation of bonded layers in multi-material systems.

Keywords

Magnetic pulse delamination, High-power pulsed generator, Laminate structures, Interfacial delamination

1 Introduction

Material assemblies are increasingly used in load-bearing composite structures, especially in high-value technology sectors (Ambaye et al., 2022; Clyne and Hull, 2019). Combining lightweight efficiency with durability, these assemblies face challenges in sustainably joining diverse materials like composites, ceramics, and metals, and require robust testing methods for quality assurance. Moreover, recyclability has become a critical factor for sustainable engineering.

Various tests exist to evaluate adhesion strength, primarily using acoustic waves to induce tensile stresses at material interfaces (Braccini and Dupeux, 2012; Mittal, 2014; Packham, 2006 ; Wang, 2011; Meyers, 1994; de Resseguier et al., 2009). Dynamic adhesion tests like the LASAT (Laser Shock Adhesion Test) are limited by very short pulse durations ($\sim 10^{-9}$ s), restricting their applicability to thin coatings and potentially causing surface damage (Tahan et al., 2020; Ecault et al., 2014).

Magnetic pulse technology offers a more suitable alternative, with longer pulse durations ($\sim 10^{-5}$ to 10^{-7} s), capable of evaluating thicker laminate structures without surface damage (Krongelb, 1978). The primary challenge in magnetic pulse disassembly remains accurately controlling interfacial tensile stresses. Unlike discontinuous LASAT testing, magnetic pulse tests require methods that account for continuous stress evolution to ensure reliable and predictable laminate disassembly.

Building on this context, the objective of this paper is to define the operating conditions of a high-power pulsed (HPP) generator suitable for the controlled delamination of laminated structures. The first part introduces the modelling of the mechanical loading induced by the generator. A one-dimensional wave propagation analysis using the method of characteristics, originally developed in previous studies, is reinterpreted to define the desired operating conditions for an HPP generator. From this, expressions are derived to relate the required magnetic field and pulse frequency to the laminate's properties.

2 Modelling and methodology

2.1 Loading tailoring

A high-power pulsed generator is an electrical device designed to deliver, electrical pulses of short duration that typically last from nanoseconds to microseconds as illustrated in **Fig 1**. Such generators primarily rely on high-voltage capacitors to store large amounts of electrical energy, which are then rapidly discharged through specialized high-speed switching components, including spark gaps or semiconductor switches (thyristors, IGBTs). Critical parameters of these generators include the pulse amplitude, current intensity, pulse duration, rise time, repetition frequency, and total energy delivered per pulse. These features make them particularly suited for applications requiring precise and controlled transient energy delivery, such as dynamic mechanical testing or electromagnetic processing techniques.

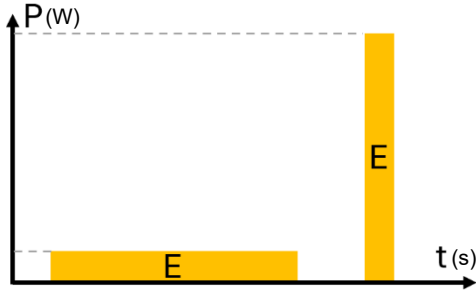


Figure 1 : Power generated with quick energy discharge for the same energy

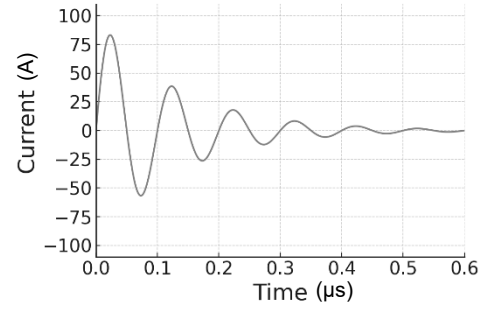


Figure 2 : Time evolution of pulse generator's discharge electrical current

The current delivered by a high-power pulsed generator typically exhibits a damped sinusoidal time evolution highlighted in **Fig 2**. This time evolution results from the interplay between the stored energy in capacitors, the inductance of the circuit (coils, connections), and resistances causing energy dissipation. An equivalent RLC representation of a HPP generator is illustrated in **Fig 3**. Consequently, the current briefly oscillates while diminishing exponentially, allowing precise and controlled electromagnetic or mechanical loading on the targeted material.

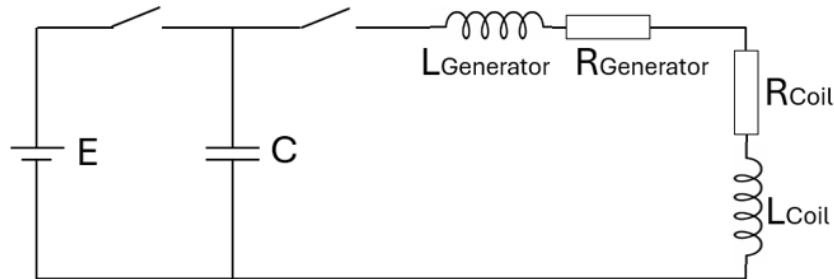


Figure 3 : Magnetic pulse welding equivalent RLC

Considering the characteristics of the generator namely capacitance C , inductance L , resistance R , and V_0 the initial voltage applied to the capacitor, the current can be expressed as :

$$I(t) = I_0 e^{\frac{-t}{T}} \sin(\omega_0 t) \quad (1)$$

$$\text{with } I_0 = V_0 \sqrt{\frac{C}{L}} ; \quad \omega_0 = \frac{1}{\sqrt{LC}} ; \quad T = \frac{2L}{R}.$$

In most applications involving high-power pulsed generators, especially magnetic pulse delamination, the main interest lies in the initial current peak, as it predominantly governs the mechanical effects induced in the target structure. To simplify analytical or numerical analyses, this initial peak can effectively be approximated by a sinusoidal function. This approximation accurately captures the rapid rise of the current to its maximum amplitude, as

well as its initial decay, facilitating the analysis of internal mechanical stresses and enabling precise predictions of dynamic behaviour, especially controlled delamination. Such a load, whose expression is given by Equation (2), is illustrated in **Fig. 4**.

$$\overline{P}(t) = \begin{cases} P_{max} \sin^2 \left(\frac{\pi t}{\tau} \right) & \forall t \in [0, \tau] \\ 0 & otherwise \end{cases} \quad (2)$$

where τ is the pseudo-period of the pulse. The pseudo period can be defined as the time between two successive maxima (or minima) in a damped oscillation.

Moreover, if the outer layer of the laminate structure is electrically conductive, the induced mechanical loading due to electromagnetic effects—specifically Lorentz forces—will follow a square sinusoidal evolution as the generated current which represents a reasonable estimate of the first half pseudo-period of the actual loading profile.

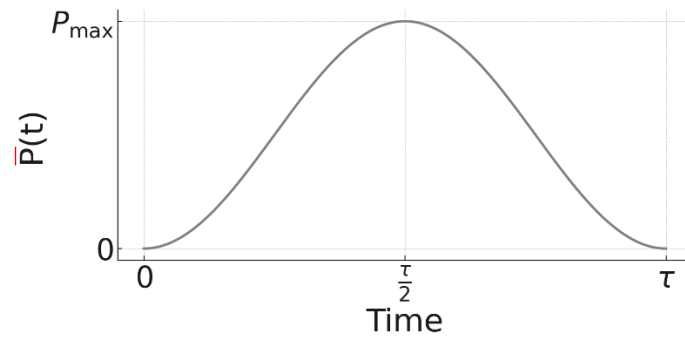


Figure 4 : Time evolution of the applied pressure

2.2 Dynamic Wave Propagation Effects

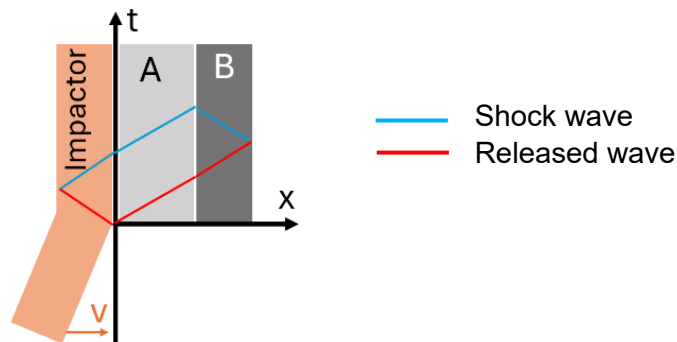


Figure 5 : Analogy between Newton pendulum and plate impact on a bi layer (A,B) laminate.

When an impactor strikes a laminate structure, stress waves are generated and propagate through both the impactor and the laminate. Then, tensile stress occurs at the crossing point of two reflected unloading waves. This behavior is depicted in **Fig. 5** through an (x,t) diagram, as typically employed in shock wave studies (Meyers, 1994). The precise location of these tensile stresses within the laminate depends on the wave round-trip time in the impactor and in each layer of the laminate. By carefully selecting the thicknesses of the impactor and back plate according to the relationship $\frac{d_{back_plate}}{d_{impactor}} = \frac{c_{back_plate}}{c_{impactor}}$ (Lagain et al., 2023), tensile stresses can be strategically localized at the desired interface, enabling controlled delamination of the targeted layer ; d_i and c_i being respectively the thickness and elastic longitudinal celerity of element i.

In plate impact scenarios, the generated mechanical load is discontinuous in time by nature. In contrast, high-power pulsed technologies provide a time-continuous and more sustained loading profile, making it possible to achieve controlled delamination of thicker laminate structures.

The magnetic pulse delamination method then aims to separate the interface between two bonded layers in a laminate structure without damaging them. It involves rapidly discharging electrical energy stored in a capacitor bank into an inductor coil placed near the laminate's conductive top layer. The resulting oscillating current induces eddy currents within the first conductive layer, generating repulsive Lorentz forces. These forces produce dynamic waves propagating primarily through the laminate's thickness.

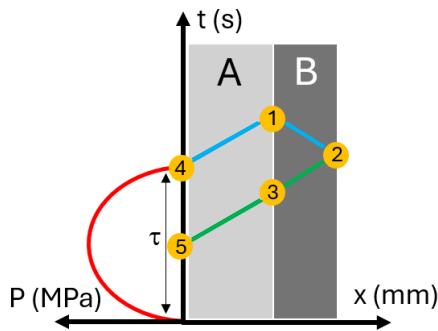


Figure 6 : Characteristic plane (x,t) plotted with two layers (A,B). Few characteristics lines are plotted from the unloaded state (in blue) and from the maximum loaded one at the left side of the laminate in the three layers (in green).

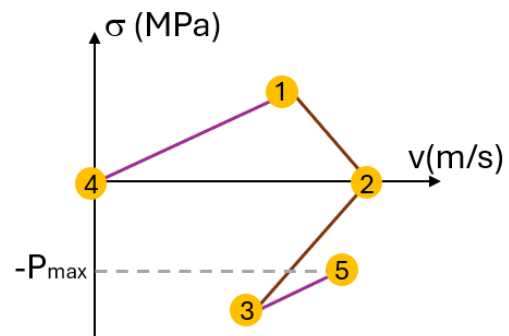


Figure 7 : Plane (v, σ) associated with the characteristic plane shown in Fig 6. Lines in purple and brown are associated with the layers A, B respectively. The plot is here made in the case $ZB > ZA$.

The Lagrange diagram shown in **Fig 6** and the associated state diagram, shown in **Fig 7** highlight this process shows a few characteristic lines along which pressure waves travel through the laminate (Meyers 1664). In this diagram, the right boundary is considered a free boundary condition. The intersection of two unloading waves, at point 1 emerging from the

intersection of characteristic lines 1-2 and 4-1, creates tensile stress as shown in Fig. 7, enabling delamination.

Based on the results developed by Lagain et al., 2023, where the method of characteristics has been used to express the interface tensile stress as a function of the properties of the laminate (Lagain et al., 2023, Fig 7), the period of the pulse should be equal to $\tau = \frac{4d_B}{c_B}$.

However, considering a damped sinusoidal waveform, this condition imposes a pseudo period of discharge of about:

$$T = \frac{8d_B}{c_B} \quad (3)$$

Therefore, based on the work of Lagain et al., 2023, to reach the interfacial rupture stress, the maximum applied pressure must satisfy:

$$P_{max} = \begin{cases} \frac{(Z_A + Z_B)^2 \sigma_R}{2Z_B \left(2Z_A + \sin^2 \left(\frac{d_A c_B \pi}{2d_B c_A} \right) (Z_B - Z_A) \right)} & \text{if } \frac{T}{8} \leq \frac{d_A}{c_A} < \frac{T}{4} \\ \frac{(Z_A + Z_B)^2 \sigma_R}{4Z_B Z_A} & \text{if } \frac{d_A}{c_A} \geq \frac{T}{4} \end{cases} \quad (4)$$

with σ_R the interfacial critical stress and Z_i the acoustic impedance of layer i.

3 HPP generator characteristics

Equation (3) provides the required period of the mechanical load for a specific structure. Assuming that the time evolution of the mechanical load matches that of the current delivered by the HPP generator, the generator must be able to produce a current with a frequency of about :

$$f = \frac{c_B}{8d_B} \quad (5)$$

If this condition is met, and considering that the magnetic pressure generated in this type of process is related to the magnetic field by $P_{mag} = \frac{B^2}{2\mu_0}$, then delaminating the last layer of the laminate requires that the magnetic field satisfies :

$$B_{max} = \sqrt{2\mu_0 P_{max}} \quad (6)$$

with μ_0 the vacuum permeability.

The generated magnetic field is influenced by multiple factors, including the impedance of the charge, the geometry and positioning of the electrodes, and the characteristics of the

surrounding environment (such as materials, distances, and boundary conditions). These elements directly affect the distribution and intensity of the magnetic field, and therefore the mechanical loading induced in the laminate. As such, a dedicated application-specific study is essential to accurately characterize the generator and its configuration. This ensures that the system can reliably produce the magnetic field required to achieve the desired delamination effect under real operating conditions.

4 Conclusion

This study set out to determine the operating conditions that high-power pulsed (HPP) generator needs to perform controlled delamination of laminated structures. The analysis focused on the generation of interfacial tensile stress through magnetic pulse loading, with the goal of reaching the interface failure without damaging the layers.

Assuming that the mechanical loading follows the time evolution of the pulsed current, this leads to a target frequency for the generator, directly linked to the thickness and acoustic properties of the laminate. From this, an expression for the required magnetic field to achieve delamination was derived, taking into account the rupture stress and the acoustic impedance of a layers. This provides a practical criterion for estimating the minimum magnetic pressure needed at the interface.

It is important to note that the magnetic field produced in practice also depends on external parameters such as the geometry and positioning of the electrodes or the surrounding environment. Therefore, while the approach presented here defines the general design rules for an HPP generator, a detailed, application-specific study remains essential to ensure that the required load conditions can be achieved.

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Conflict of interest

The authors declare they have no conflict of interests.

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Magnetic pulse welding of dissimilar aluminum-copper joints: Impact of downscaling on weldability

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Abstract

Due to the higher specific electrical conductivity (conductivity/weight) of aluminum compared to copper and the associated lightweight potential, the electrical industry is moving towards current-conducting components made of aluminum. However, the use of aluminum is associated with a necessity for hybrid aluminum-copper joints, due to space limitations as well as purchased components are often made of copper. Furthermore, it can be observed that numerous flat conductors, such as those used in DC and asynchronous motors, are only a few millimeters wide. Due to the rapidly growing market for electric vehicles, an increasing demand for aluminum-copper dissimilar joints with small dimensions is consequently to be expected.

In contrast to conventional fusion welding processes, magnetic pulse welding (MPW) is suitable for joining aluminum-copper dissimilar joints, due to the minimization of brittle intermetallic phases. This is attributed to the low energy input, which is applied as the movement of mass to fabricate a material continuous joint.

It is known that MPW reaches its process limits when the dimensions of the joining partners decreases due to the minimization of effective eddy currents. However, these limits are still insufficiently researched. Consequently, this investigation is intended to reveal the process limitations of MPW with regard to the joining partner dimensions. For this purpose, various geometries are evaluated with regard to their weldability and the resulting microstructural, mechanical and electrical properties.

Keywords

magnetic pulse welding, aluminum-copper dissimilar joints, joining partner dimension

1 Introduction

Due to the increasing electrification of vehicles and the associated high number of electrical contacts, these are becoming increasingly important. In Germany the number of newly registered electric cars increased by almost 270% between 2020 and 2023. The market only declined in 2024 due to the end of state funding for electric cars at the end of 2023. However, it can be assumed that the market will recover from this year onwards. In January 2025, an increase of 53.5% in electric car registrations was observed compared to January 2024. Strong global demand for electric cars can also be observed. Over 14.8 million new electric cars were registered worldwide in 2023. (Statista 2025a, 2025b)

Flat electrical conductors, as exemplified by those found in direct current (DC) and asynchronous motors, as well as in battery packs, are experiencing a marked increase in importance. In the context of the EU's goal of climate neutrality by 2050 (Europäische Kommission 2024), the concept of lightweight construction is also gaining in importance in the automotive sector. The potential for lightweight construction in electrical applications is given by the partial substitution of copper by aluminum due to its higher specific electrical conductivity in current-conducting components (Graß and Böhm 2023). Furthermore, the industry continues to strive the replacement of copper flat conductors with aluminum flat conductors due to the enormous cost savings of more than 80% (boerse.de 2025a, 2025b). However, complete material substitution is not possible because of the design of purchased power electronic components made of copper or installation space restrictions that require the use of copper. This results in the necessity of aluminum-copper dissimilar joints. Compared to conventional fusion welding processes, magnetic pulse welding (MPW) is suitable for joining dissimilar joints, as the joining partners are only thermally impacted locally in the joint area. As a result, intermetallic phases are significantly minimized. Since flat electrical conductors in electric motors are only a few millimetres wide due to space restrictions, it is important to investigate how reduced joining partner dimensions affect the weldability of MPW.

2 Experimental Setup

2.1 Magnetic Pulse Welding

The Blue Wave PS48-16 MPW system from the manufacturer PSTproducts GmbH was used for the experiments described in this publication. The six capacitors of the MPW system can store up to 48 kJ at a maximum charging voltage of 16 kV. The active area of the flat coil used is characterized by its coil length of 160 mm and width of 10 mm. During the welding process, a high-current pulse generated by the capacitors flows through the flat coil, creating a magnetic field surrounding the active area of the coil radially. The magnetic field in turn causes the induction of eddy currents in the first joining partner (flyer) arranged above it (shown in Figure 2(b) in Watanabe and Kumai 2009). The principle of mutual induction causes the formation of a secondary magnetic field around the flyer, which is directed in the opposite direction to the primary magnetic field of the coil. The resulting electromagnetic

force (Lorentz force) accelerates the flyer in the direction of the second joining partner (parent). The necessary acceleration distance is set up using suitable spacers. The flyer plate then collides with the parent plate at high speed and conducts a rolling movement, which is characteristic for MPW (shown in Figure 46 in Mori et al. 2013; Hahn et al. 2016)). Due to the high impact speed (several hundred m/s), or more precisely, due to the high impact pressure (up to a few thousand MPa) of the joining partners, parts of the boundary layers of the joining partners are plastified and ejected out of the closing gap between Flyer and Target (Jetting effect). Finally a metallic bond occurs (Chaturvedi and Arungalai Vendan 2021; Mori et al. 2013).

2.2 Materials, specimen dimensions and welding parameters

The aim of this publication is to determine the weldability of aluminum-copper dissimilar joints using varying joint partner widths. For this purpose, sheets made of the aluminum alloy AA 1050A (flyer plate) and the copper alloy Cu-ETP (parent plate) with the mechanical properties shown in **Table 1** were used. The sheet thickness for both alloys is 1 mm.

Alloy	Tensile strength [MPa]	Fracture strain [%]
AA 1050A	120,22	5,13
Cu-ETP	269	32

Table 1: Mechanical properties of the used alloys

To identify the lower process limit, the width of the joining partner was reduced iteratively in 1 mm steps from 15 mm on. Furthermore, the remaining framework conditions of the welding configuration were kept constant. A schematic representation of the welding configuration is shown in **Figure 1**. For the investigations carried out in this paper an acceleration distance of 1 mm, a discharge current of 482,7 kA and a constant overlap length of 20 mm was used. With the system technology used for these investigations, the used discharge current still represents the mechanical load limit of the flat coil used. A further increase in the discharge current is therefore not recommended.

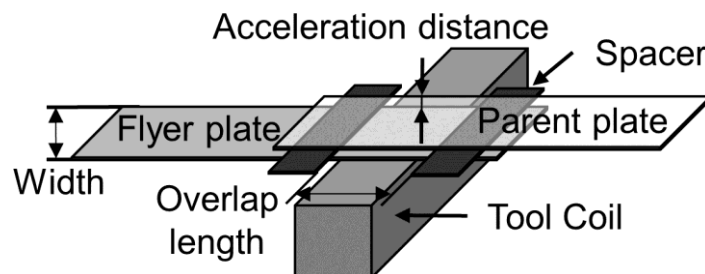


Figure 1: Schematic overview of the welding configuration

2.3 Electrical Characterization

The MR200C micro-ohmmeter from Schuetz, which uses the Kelvin four-wire resistance measurement (**Figure 2**), is used for the electrical characterization of the welded specimens.

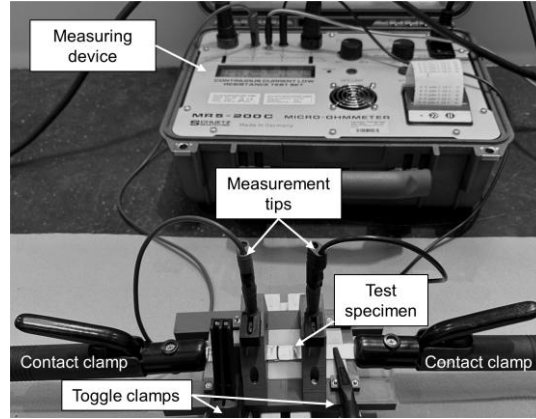


Figure 2: Resistance measurement setup

When positioning the specimen in the resistance measurement setup shown in **Figure 2**, the weld seam has to be centered between the two measurement tips. The two measurement tips are then lowered onto the specimen surface. Based on (Graß et al. 2024), a current of 50 A is then applied allowing the local voltage to be measured with high resistance via the measurement tips. The resistance of the joint is then determined from the voltage difference between the measurement tips according to **Eq. (1)**.

$$R_{MPW} = \frac{\Delta U}{I} \quad (1)$$

Reference measurements were also carried out to quantify the results of the electrical characterization. To reference the electrical resistance values, the aluminum and copper base materials were also electrically tested for each joint partner width. The reference resistance of the joint was then calculated according to **Eq.(2)**. Finally, the electrical resistance of the specimens was also normalized to its reference condition for graphical representation.

$$R_{ref} = \frac{1}{2} (R_{Al} + R_{Cu}) \quad (2)$$

2.4 Mechanical Characterization

The welded specimens were tested and evaluated mechanically in the quasi-static shear tensile test with a universal tensile testing machine (Z100, ZwickRoell GmbH & Co. KG) at 5 mm/min. A specimen that failed in the aluminum base material was assumed to be OK. In order to be able to quantify the results of the mechanical characterization adequately, reference measurements were carried out. The aluminum sheet represents the reference state

for the mechanical tests, as this is the weakest joining partner in terms of tensile strength. Based on a reference tensile specimen, the maximum tolerable force per joint partner width was calculated by multiplying it by the cross-sectional area. The maximum tolerable forces of the specimens were normalized to their respective reference condition for graphical representation.

2.5 Weld Seam Analysis

To determine the weld area on the parent plate (Cu sheet) by using optical macroscopy (Leica Z16), the flyer plate (Al sheet) was first dissolved in 10 % sodium hydroxide solution for 24 hours. The weld seam area can then be recognized as a dark ellipse on the copper sheet. For microstructural analysis of the specimens, they were analysed metallographically. For this purpose, the specimens were embedded in a resin compound, ground and polished. Scanning electron microscope (SEM, Zeiss ULTRA PLUS) and energy-dispersive X-ray spectroscopy (EDS, Bruker 6|60) examinations subsequently allow statements to be made about the weld seam characteristics in the interfaces of the joining partners.

3 Results

When looking at **Figure 3**, it can be seen that the normalized electrical resistance of 15 mm to 9 mm joint partner width is at a similarly good level. Due to the overlap joint design, the magnetic pulse welded specimens have a lower resistance than the reference sheets. As a result, the flat conductor in the weld area has a larger cross-section than the reference sheet. Just like the electrical resistance, the mechanical behavior of the samples in the widths of 15 mm to 9 mm can be evaluated as positive. These specimens completely failed in the aluminum base material in the quasi-static shear tensile test. Due to the overlap joint of the welded specimens, which results in a complex stress situation, the maximum tolerable force is lower than that of the aluminum base material (**Figure 3**).

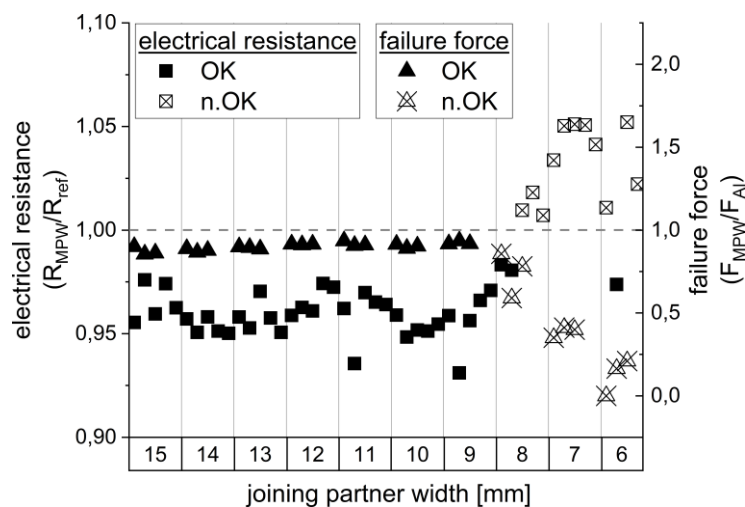


Figure 3: Standardized electrical resistance and failure force of the specimens

Joining partner widths of 8 mm and less are to be rated as not OK, as they either have bad electrical or mechanical properties, or both. An intended joint of 5 mm wide sheets is not shown in **Figure 3** as it could not be welded. Even increasing the discharge current to 497,2 kA did not result in a welded joint.

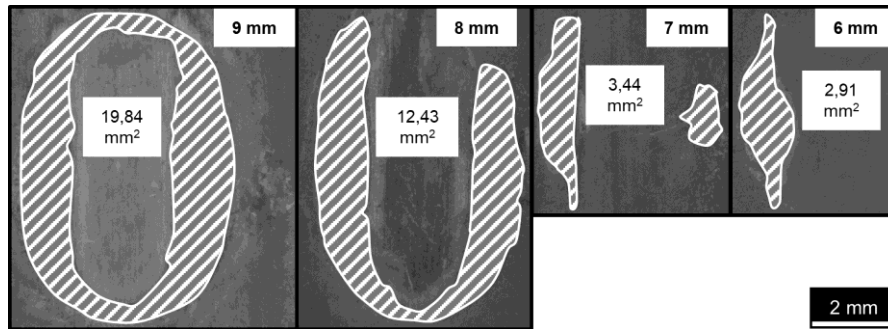


Figure 4: Macroscopic comparison of the welded areas of the samples in widths of 6 mm to 9 mm

The macroscopic images confirm a reduction in the weld joint at sheet widths of 8 mm and less. **Figure 4** shows that no elliptical weld seam typical of the MPW is formed. A further reduction in the joining partner width results even more clearly in a reduced welding area.

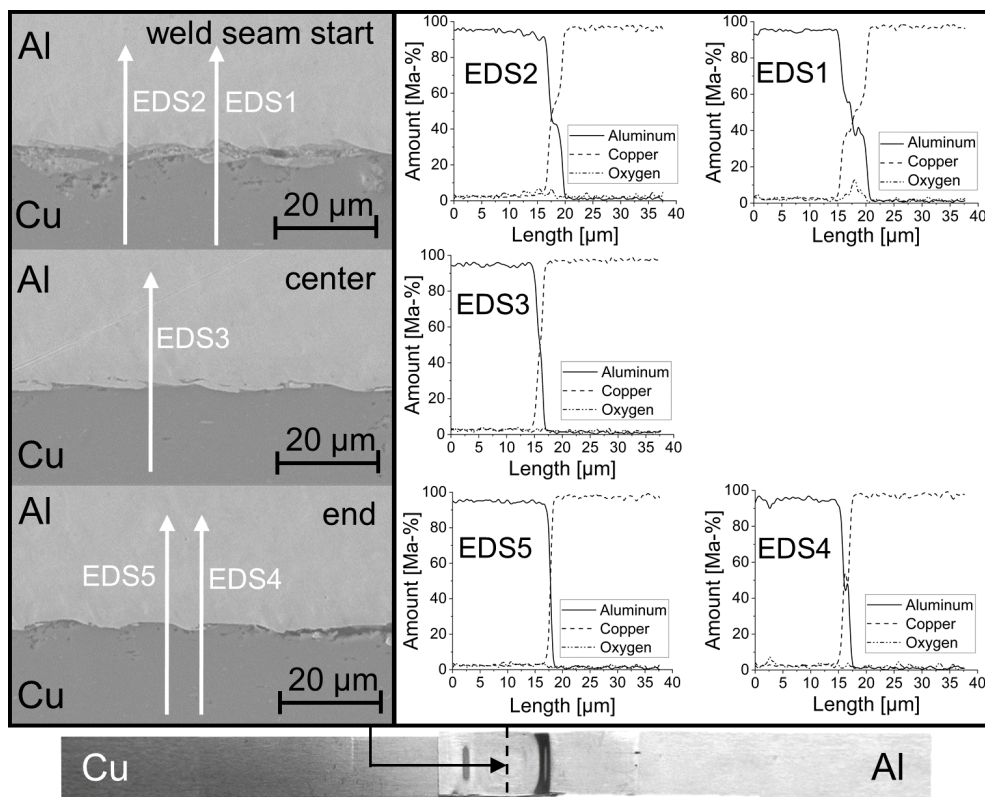


Figure 5: SE2 images and EDS analysis of the weld seam of the 15 mm wide specimen with an accelerating voltage of 16 kV and a working distance of 10 mm

The SEM images in **Figure 5** show typical interface characteristics of an MPW weld. There are clearly differentiated-looking phases before the start of the weld seam. These show an enrichment of oxygen in EDS 1. A trapped jet can be assumed in this area, which typically rolls over locally during the welding process. The weld seam center also shows typical characteristics of an MPW welded seam in the form of wave formation in the interface of the joining partners. The EDS analyses also show the typical characteristic of the interface for the MPW that the width of the interface decreases continuously towards the end of the weld seam.

4 Conclusion and Outlook

In this publication, the assumption of the process limit for MPW depending on the minimum weldable sheet width was confirmed. A substrate combination of AA 1050A and Cu-ETP with a thickness of 1 mm was used for this investigation. The MPW process limit could be determined with the used system technology at a minimum sheet width of 6 mm. However, superior electrical and mechanical properties could only be realized with 9 mm or wider joining partners. The associated welding area also confirms the determined MPW process limit. A complete ellipse of the weld seam is only achieved with a joining partner width of 9 mm or greater. In view of the rapidly decreasing welding area with a joining partner width of 8 mm and below and the small welding area with a joining partner width of 6 mm, it seems logical that a joining partner width of 5 mm no longer permits welding.

Future investigations with a more powerful system technology would have to clarify whether a significant increase in the discharge energy and, thus, a significant increase in the current flowing through the coil would allow a further minimization of the joint partner width. Furthermore, it would be of interest to investigate the influence of the surface modification of the joining partners on the possibility of further minimizing the joining partner width in relation to the material combination used in this publication. Further investigations of different welding configurations, which have a controlled influence on the rolling movement, e.g. by previously deformed joining partners or a one-sided positioning, would have to be analyzed with regard to their influence on the downscaling.

5 Acknowledgment

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Determination of Weld Interface Properties and their Effect on the Strength of Multi-Material Aluminium-Steel Joints

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Abstract

Collision welding is a promising manufacturing approach for joining similar and dissimilar metals without thermally influenced strength loss. This capability unlocks new potential for lightweight construction. Among other things, components and welded joints can be made more filigree and load-optimised. Widespread industrial use of this technology is still limited by insufficient knowledge of the underlying joining mechanisms. In this paper, collision welding is applied to an aluminium-steel material combination of steel (DC04) and aluminium (AA6016). First, the welding process window for the material combination is determined on a special model test rig by varying the collision speed and the collision angle, the two main process parameters in collision welding. Shear tensile tests are then conducted to refine the weld process window, focusing on the region exhibiting the highest shear tensile strength values. Scanning Electron Microscope (SEM) images are used to analyse the weld zone, revealing the formation of four different metallographic structures in the joint zone of the welded samples. The area fractions of these structures are quantified and correlated with collision angles and corresponding shear tensile strength values.

Keywords

Collision welding, Multi-Material Design, Lightweight Concepts, Emission Reduction

1 Introduction

Mobility is a fundamental aspect of modern society, enabling access to education, healthcare and the labour market. It includes private and public transport, logistics and infrastructure (Miraftabzadeh et al., 2024). In 2023, the transport sector accounted for 21% of global CO₂

emissions, 38% of which came from passenger cars (Crippa et al., 2024; EDGAR, 2024; IEA, 2025). In addition to CO₂ emissions, both electric and combustion vehicles emit harmful particulate matter from brake dust and tyre abrasion, known as "non-exhaust particle emissions", which exceed exhaust particulate emissions from modern combustion engines (EEA, 2023; OECD, 2020; Lopez et al., 2023; Timmers et al., 2016). Overall up to 10% of the marine microplastics released each year originate from tyre abrasion (Kole et al., 2017; Köllner, 2024a). As a result, the Euro 7 emissions standard will regulate these emissions for the first time (ADAC, 2024; EC, 2022; Visnic, 2023).

Reducing vehicle mass is a key strategy for lowering these emissions (Bandivadekar et al., 2008; Köllner, 2024b; OECD, 2017). A 100 kg reduction in the mass of a small combustion vehicle reduces CO₂ emissions by 10 g CO₂/km (OECD, 2017). Increased vehicle weight has a significant impact on non-exhaust particle emissions. For example, the additional weight of 318 kg of an electric car compared to a comparable petrol car results in up to 22% higher non-exhaust particulate emissions (Beddows and Harrison, 2021). This underlines the environmental importance of lightweight design. Multi-material construction using aluminium and steel, especially in high-strength variants, supports these goals (Tisza and Lukács, 2018). However, traditional fusion welding methods such as tungsten inert gas (TIG) and laser beam welding face challenges due to the different melting points of the materials (Atabaki et al., 2014; Carlone and Astarita, 2019). If a material bond is possible at all, the melting of the materials leads to the formation of intermetallic phases in the weld zone. The associated loss of ductility in the brittle metal joints limits the potential of the materials used (Agudo et al., 2007; Khedr et al., 2023). To realise the full potential, a joining process is required that allows strong aluminium-steel joints to be made while minimising the loss of strength.

Solid-state or collision welding processes offer a solution by applying sufficiently high pressure to join similar and dissimilar materials without active heat input, avoiding the formation of intermetallic phases (Agudo et al., 2007; Kapil and Sharma, 2015; Khedr et al., 2023). Joining feasibility is defined by collision speed and angle, with the process window categorised into solid-phase, liquid-phase and hybrid joining regions (Niessen, 2022). Electromagnetic pulse welding (EMPW) facilitates collision welding in industrial settings by accelerating a joining partner via a strong electromagnetic field, creating a high-speed impact weld (Atabaki et al., 2014; Golovashchenko, 2006; Kapil and Sharma, 2015; Weddeling et al., 2010). Due to transient phenomena, the collision speed and angle dynamically vary during the course of the EMPW collision, affecting the weld properties (Stankevic et al., 2020; Wang et al., 2023). Currently, EMPW process design relies on iterative methods, which are costly and introduce uncertainties in mass production (Marschner et al., 2021). For future optimised design of EMPW processes, it is therefore necessary to know the critical angle profile for the joint design during the collision in order to best influence the achievable strength properties of the joint in a robust manner.

This study uses a mechanical collision welding test rig to analyse joint formation for the material combination DC04 with AA6016-T4 at varying collision angles and speeds. The aim is to determine the optimum angle range for joining and to correlate strength values with collision parameters. Unlike previous studies which have assessed joint strength and

microstructure at discrete points, this research uses SEM imaging to map joint structures across the weld zone and relate these findings to tensile shear strength. This should provide the basis for a future analytical model that predicts EMPW joint strength based on simulated angular profiles, thereby aiding future process optimisation.

2 Materials and Methods

The tests in this work are carried out on a model test rig for collision welding, see the brief functional description below. A detailed description of the test rig is given in (Kraus et al., 2024). The specimens used are taken from aluminium sheets (AA6016-T4, thickness: 1.5 mm, tensile strength R_m : 235 N/mm²) and steel sheets (DC04, thickness: 2.0 mm, tensile strength R_m : 310 N/mm²) by shear cutting. For each material, three tensile tests are performed on a *Zwick Roell 100* combined tensile and compression testing machine to determine tensile strength in accordance with DIN 50125.

The model test rig for collision welding, developed at the Institute for Production Engineering and Forming Machines (PtU) at the Technical University of Darmstadt, allows the collision welding process to be investigated over the time profile of the collision with constant and precisely adjustable collision parameters thanks to its fully mechanical design. At the same time, it offers good observability of the process and the process phenomena. Fig. 1 (a) shows the construction of the model test rig. The specimens to be welded (collision area: length $\times$ width: 12.5 mm $\times$ 12 mm) are mounted at the ends of the two rotors rotating in the same direction. The AA6016-T4 flyer specimen is pre-bent to set the collision angle β . The DC04 target is flat and is additionally supported by an anvil, see Fig. 1 (b). The collision is triggered by an engagement mechanism when the desired speed is reached, causing the specimens to collide at the position shown. The electromagnetic pulse welding (EMPW) process described in the previous section is shown in Fig. 1 (c). In contrast to the model test rig, in EMPW the flyer is accelerated towards the target by an electromagnetic field.

On the model test rig, the collision speed v_{imp} is varied in five steps of 262 m/s, 279 m/s, 305 m/s, 331 m/s and 349 m/s by rotating the rotors at a specified speed to study the transient EMPW process at discrete points. The collision angle β is varied between 3.5° and 12.5°. The specimens are cleaned with acetone immediately prior to testing to eliminate any influence on the test results from lubricant or corrosion inhibitor residues on the specimen surfaces. High-speed observation of the collision is performed using a *PCO hsfc pro* image intensifier camera, which acquires up to eight images per collision.

As shown in Fig. 1 (d), the mounting face and side edge areas of the collision welded specimens are first cut off using a wet cut-off grinder in order to remove any effects of specimen fabrication in the edge areas of the weld and to shorten the overall length (14 ± 1 mm) and width (8 ± 1 mm) of the specimen for microscopic examination in order to analyse the weld zone. The cut surfaces of the specimens are then ground on grinding and polishing machines with SiC paper (grit size P320, P400, P600, P800, P1200, P2500 and P4000) and polished with diamond suspension (3 μ m and 1 μ m). The weld zone is examined microscopically using a *Phenom ProX* scanning electron microscope (SEM), manufactured by *Phenom-World*. The lengths of different joint structures along the length of the weld zone

are measured with the microscope to analyse correlations between the joint structures that occur and the shear tensile strength of the welded joint. The tensile shear strength of the welded joint is determined on a *Zwick Roell 100* combined tensile and compression testing machine. For more detailed information about the tensile shear tests, see (Kraus et al., 2024).

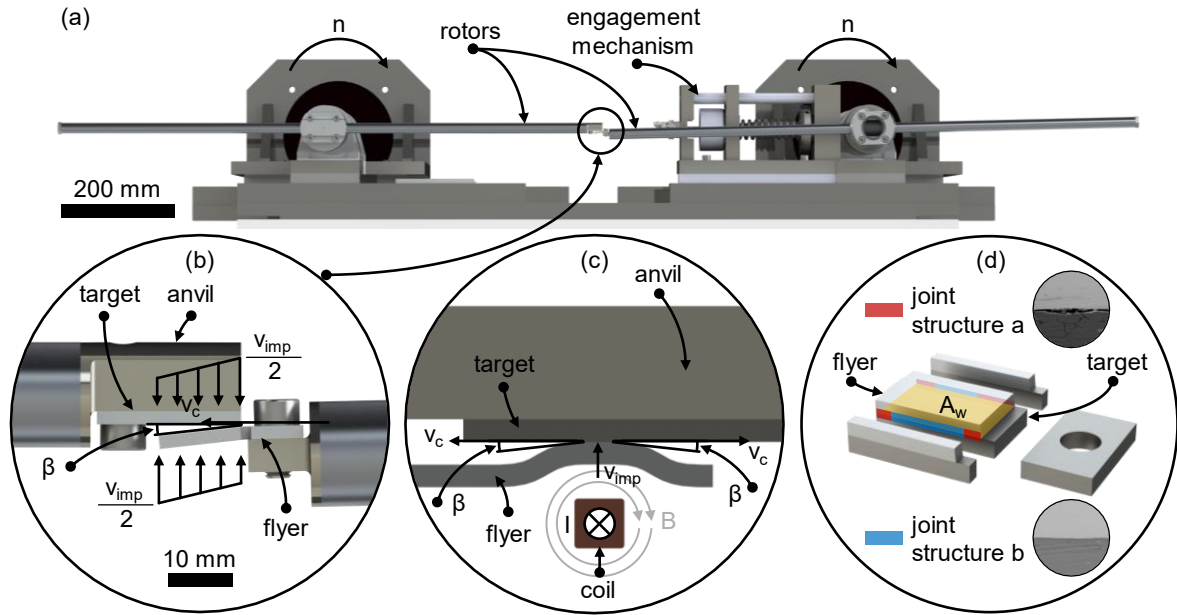


Figure 1: (a) The model test rig consists of two synchronised rotating rotors and an engaging mechanism; (b) The specimens are mounted at the end of each rotor (target backside supported by an anvil, flyer pre-bent to set the collision angle β); (c) Process setup for EMPW according to Böhme et al. (2019). The initially flat flyer is accelerated by the electric field. Collision speed v_{imp} and angle β change during the collision; (d) Graphic visualisation of the specimen cutting process and exemplary representation of different joint structures along the length of the weld zone.

3 Results

3.1 Welding Process Window

The welding process window, as determined by the model test rig for the material combination of aluminium AA6016-T4 (sheet thickness 1.5 mm) and DC04 steel (sheet thickness 2.0 mm), is illustrated in Fig. 2. The collision welding tests are performed at collision speeds v_{imp} of 262 m/s, 279 m/s, 305 m/s, 331 m/s and 349 m/s. Within the defined process window, a collision weld test is considered successful if the two specimens cannot be separated by manual force after the test (black dots). For the purpose of enhancing the clarity of the diagram, the successful tests are shifted 2 m/s to the left on the collision velocity axis in comparison to the unsuccessful tests (white dots). The formation of the first bonds between the two materials occurs at a collision velocity of 305 m/s. The lower and upper limit angles at that speed are 7.3° and 8.3°, respectively. At a collision speed of 331 m/s, an

increased upper limit angle of 9.8° is determined, while the lower limit angle decreases marginally to 6.8°. Furthermore, at a collision speed of 349 m/s, an even greater increase in the upper limit angle to 11.2° is observed, whilst the lower limit angle decreases further to 6.5°.

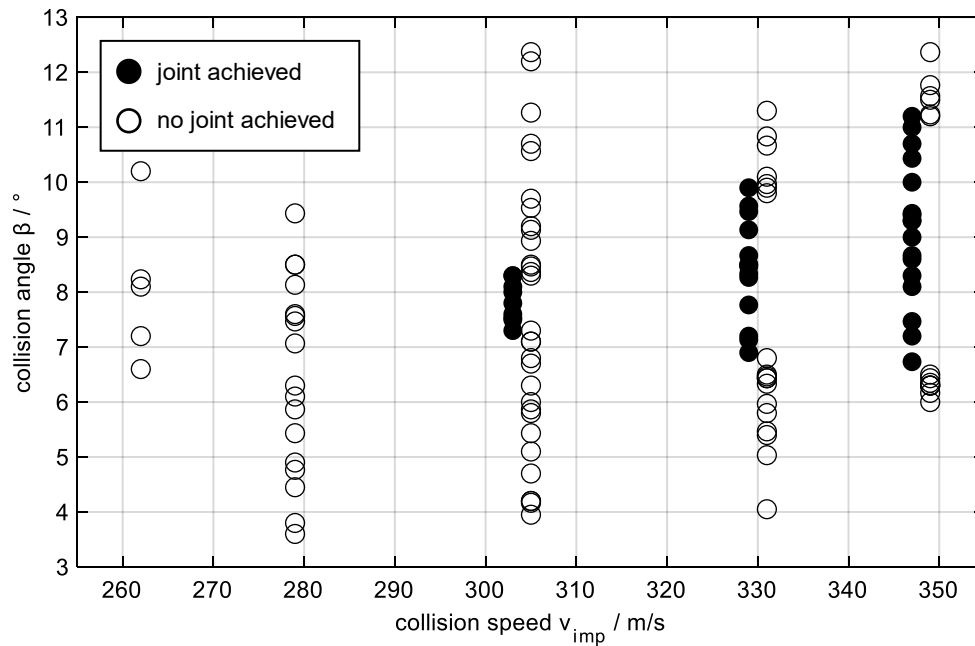


Figure 2: Welding process window of the material combination AA6016-T4 (flyer, thickness: 1.5 mm) and DC04 (target, thickness: 2.0 mm).

3.2 Tensile Shear Strength of Welded Joints in the Centre of the Welding Process Window at 305 m/s, 331 m/s and 349 m/s

Following the determination of the welding process window, an investigation is conducted into whether the tensile shear strength of the collision-welded joints is dependent on the respective collision speed or whether it is independent of the collision speed. The tensile shear strength is only compared for specimens located in the centre of the process window at each of the three collision speeds (305 m/s, 331 m/s, 349 m/s). However, it should be noted that the collision angle can only be preset in 0.5° increments by pre-bending the flyer samples. Furthermore, due to the dynamics inherent in the model test rig, the rotors undergo elastic deformations, which can be attributed to various factors such as the engagement process, air resistance, and aerodynamic turbulence within the rotor protection housing. Consequently, this results in a deviation of the actual measured collision angle from the pre-set collision angle. To address this, a greater number of samples are welded at each of the three collision speeds, and then three samples are selected for each collision speed that are within a range of $\pm 0.3^\circ$ around the mean collision angle of the process window. For the process window illustrated in Fig. 2, the mean collision angle at a collision speed of 305 m/s is 7.8°. At 331 m/s, the angle is 8.3°, and at 349 m/s, 8.8°.

The tensile shear strength values obtained demonstrate an increase with increasing collision speed. Specifically, at collision speeds of 305 m/s and 331 m/s, the tensile shear strength was recorded as 22 N/mm² and 55 N/mm², respectively. For the specimens that were welded at 349 m/s, a strength of 130 N/mm² was determined. Furthermore, the specimens demonstrate divergent failure patterns during tensile shear testing. At both lower collision speeds, the welded joints shear or slide against each other in the weld zone, see Fig. 3 (a). The specimens exhibiting the highest strength demonstrate a combined failure pattern, characterised by the presence of areas along the length of the weld zone where specimens are separated by shearing or sliding against each other, and areas where shearing occurs within the aluminium material, which possesses a lower tensile strength compared to steel. Consequently, the strength of the welded joint in these areas exceeds the tensile shear strength of the aluminium material. When viewed over the length of the weld zone, shear failure in the aluminium material occurs predominantly in the middle, whereas shear or sliding in the weld zone occurs predominantly at the beginning and end, see Fig. 3 (b), (c).

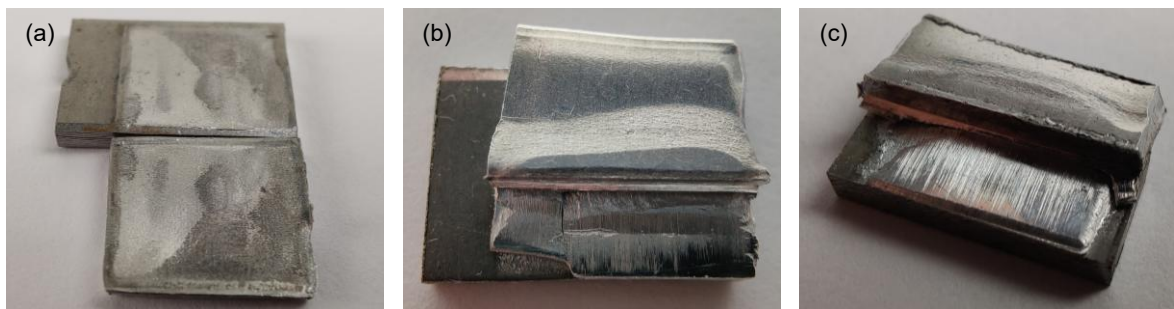


Figure 3: Welded specimens after tensile shear testing (welding direction: right to left): (a) Shear or sliding failure (top: DC04, bottom: AA6016-T4); (b), (c) Shear failure in the aluminium material (top: AA6016-T4, bottom: DC04).

3.3 Joint Structures in the Weld Zone

The preceding section's investigations indicate that the welded joints produced at a collision speed of 349 m/s exhibit the highest tensile shear strength values. Further investigations are being carried out to achieve a more precise statement about the tensile shear strength over the entire angular range of the welding process window at 349 m/s. Metallographically prepared specimens (procedure described in section 2) from the entire angular range are examined in their weld zone by SEM. In instances where differing joint structures are observed along the entire length of the weld zone of the specimens, the position and length of the corresponding areas are measured, see Fig. 1 (d). These values are then used to extrapolate the area ratios of the different joint structures to the total area A_w of the welded joint. In cases where the length ratios of the structures on both sides of the specimen are unequal, the area is calculated using non-square geometries such as a trapezoid. The SEM-examinations of the weld zones reveal four distinct joint structures, as shown in Fig. 4 and briefly explained below.

The characteristic joint structure of a weld without an interlayer is shown in Fig.4 (a). The aluminium alloy AA6016-T4 (light grey) and the steel DC04 (dark grey) are located directly on top of each other. It is notable that in this structure, there is a very smooth and nearly wave-free transition zone between the two materials, with no visible voids. The structure in which an interlayer (medium grey) is present between the aluminium alloy and the steel is illustrated in the SEM image of Fig. 4 (b), (c). It is evident that the interlayer exhibits distinct variations in both colour and structure compared to the base materials. The surface of the interlayer facing the steel (the bottom of the image) is visibly smoother in comparison to the aluminium facing side, which has a much more uneven or wavy surface. In accordance with the structure depicted in Fig.4 (a), no voids were observed in the SEM images. The interlayer thickness varies between 5 μm and 30 μm , with a predominance of 10 μm . In the interlayer, lighter streaks or more sharply defined lighter areas can be seen, the colour of which is similar to that of the aluminium material.

As illustrated in Fig. 4 (d) - (f), the structures are defined by voids of differing shapes. In the following, the structure of Fig. 4 (d), (e) will be referred to as a weld zone with a cracked interlayer. It is noteworthy that the cracks manifest exclusively within the interlayer, with no evidence of such fractures in either of the two base materials. The dimensions and optical composition of the interlayer correspond to the crack-free state shown in Fig. 4 (b), (c), so that an aluminium solid solution is also assumed here only in the cracked state. In addition, the examination reveals areas where there is no bond or direct contact between the two base materials, as illustrated in Figure Fig. 4 (f). EDS (Energy Dispersive X-ray Spectroscopy) scans of a thin interlayer are shown in Fig. 4 (g).

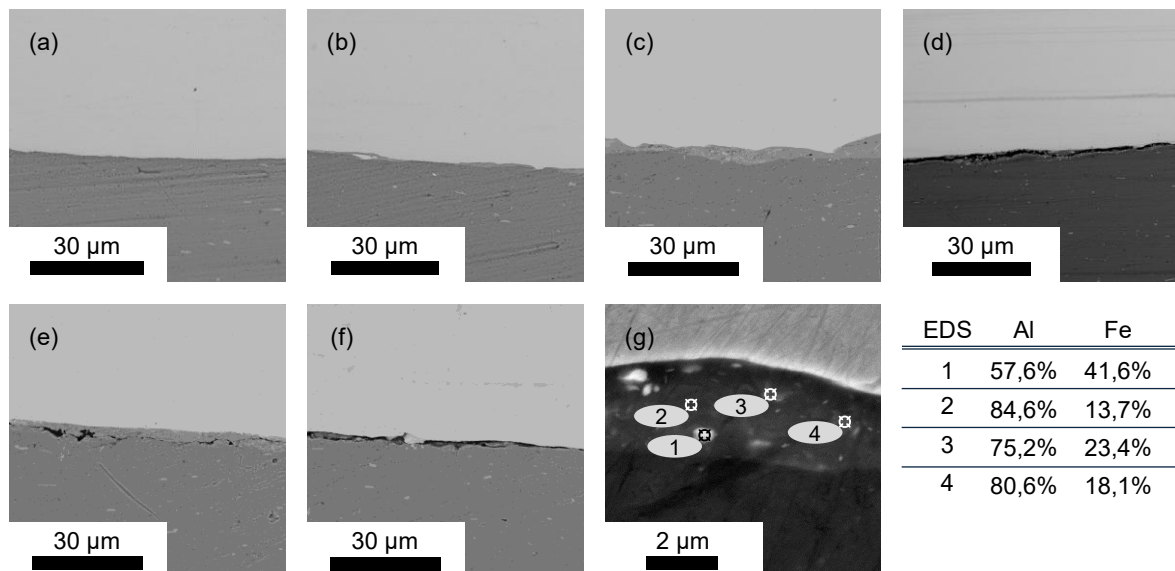


Figure 4: SEM images of the characteristic joints structures of the weld zone (top: DC04, bottom: AA6016-T4): (a) Without interlayer; (b), (c) With interlayer (medium grey, (b) thickness $\leq 3 \mu\text{m}$, (c) thickness $\leq 10 \mu\text{m}$); (d), (e) With a cracked interlayer; (f) No contact between joining parts in the weld zone; (g) EDS scans of a thin interlayer.

3.4 Tensile Shear Strength as a Function of Collision Angle

As outlined in section 3.3, the tests for the collision speed of 349 m/s are conducted on a substantially larger number of specimens, covering the full angular range of the process window. The diagram in Fig. 5 shows the curve of tensile force and crosshead travel determined with the combined tensile and compression testing machine for three specimens welded at different collision angles β at a collision speed v_{imp} of 349 m/s. The specimens with a collision angle of 7.3° and 11.2° fail after a short travel distance of the testing machine due to slipping in the welding zone, see Fig. 3 (a). The specimen with a collision angle of 9.2° , i.e., in the centre of the process window, can withstand a significantly higher maximum force. This results in failure of the welded joint due to shearing in the weaker aluminium material, see Fig. 3 (b), (c).

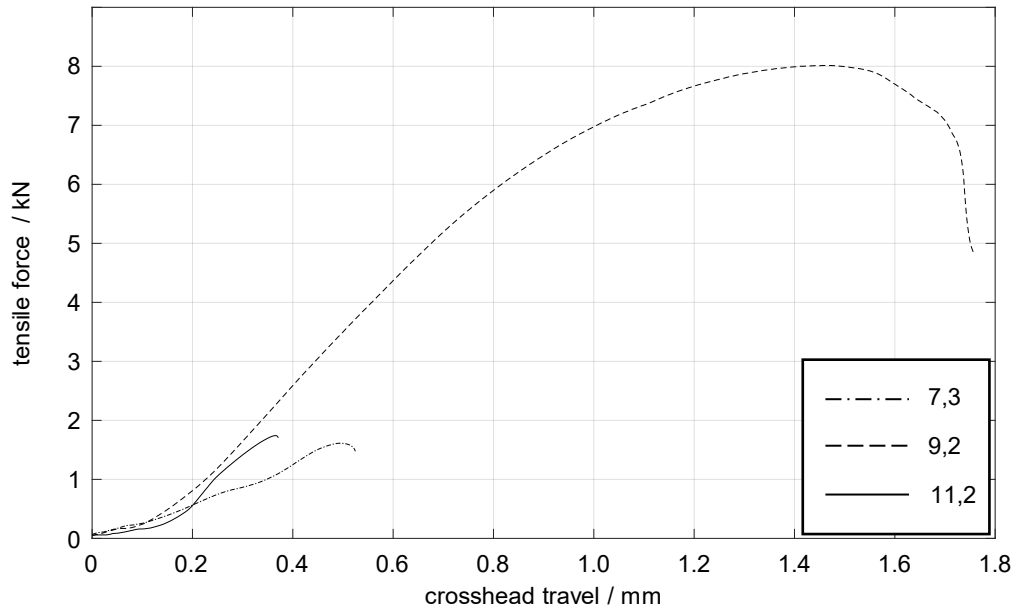


Figure 5: Force-travel diagram showing crosshead travel and tensile force in the tensile shear test for a collision speed of 349 m/s in dependence of the collision angle.

A total number of 27 specimens are investigated to determine the tensile shear strength and area ratios of the joint structures (see next section 3.5). The area between the upper limit angle (11.2°) and the lower limit angle (6.5°) is divided by these 27 specimens into three angular areas. The first area I covers the angle range $6.5^\circ - 8^\circ$, the second area II covers the angle range $8^\circ - 10^\circ$, and the third area III covers the angle range $10^\circ - 11.2^\circ$. The mean values for the tensile shear strength are shown in Fig. 6. The highest mean value of 131 N/mm^2 is observed in area II. In comparison, the strength values for area I and area III, respectively, exhibit a sharp decline towards the upper and lower limits of the process window. The lowest mean value of 30 N/mm^2 is measured in area I, while area III has a slightly higher mean tensile shear strength of 40 N/mm^2 .

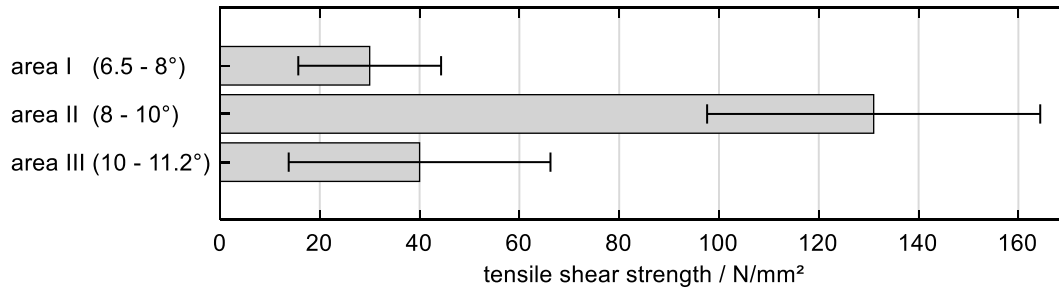


Figure 6: Tensile shear strength in the angular areas for a collision speed of 349 m/s.

3.5 Area Ratios of the Joint Structures as a Function of Collision Angle

By means of the analysis of the 27 specimens using SEM, it was possible to determine the area fractions of the structures shown and described in section 3.3, utilising the procedure outlined in that same section. The results of the analysis are presented in the following diagram (Fig. 7). The diagram contains the three joint structures *weld zone without interlayer* (Fig. 4 (a)), *weld zone with interlayer* (Fig. 4 (b), (c)) and *no contact or cracked*. The latter category comprises the *weld zone with a cracked interlayer* (Fig. 4 (d), (e)) and the *no contact between joining parts in the weld zone* (Fig. 4 (f)), since neither of these structures can transfer forces in the weld zone between the aluminium and steel material. Depending on the angular area, different joint structures dominate. In area I, the *no contact or cracked* joint structure accumulates for the largest area ratio of 73%. In contrast, in areas II and III, the area ratio is comparatively reduced to 10% and 13%, respectively. The *weld zone with interlayer* dominates the joint structure in area II with a ratio of 73%. In contrast, its proportion in area I is significantly lower, at 23%, and it is absent in area III, with only 2% observed. The *weld zone without interlayer* is the dominant joint structure in area III. The area ratio of the latter increases from 4% in area I to 17% in area II and 85% in area III.

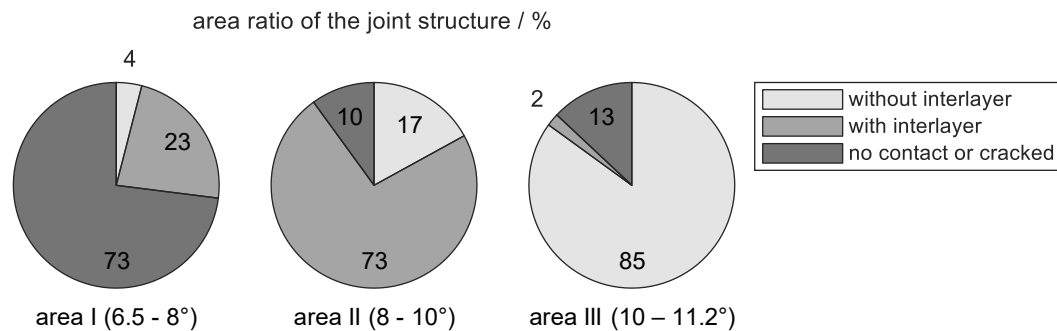


Figure 7: Determined area ratios of the joint structures for a collision speed of 349 m/s.

4 Discussion

The results of this study show, at a collision speed v_{imp} of 349 m/s, a correlation between the collision angle β , the shear strength determined in tensile shear tests and the joint structures

observed in the weld zone by SEM analysis. In the lower angle range (6.5 - 8°), the weld zone is dominated by structures classified as *no contact or cracked* (Fig. 4 (d) - (f)). This can be attributed to the highly dynamic behaviour of the closing collision gap. Small angles result in high collision point velocities v_c (Fig. 1 (b)), causing the gap to close rapidly. Surface irregularities generate strong turbulence, hindering the ejection of the jet and particles and leading to gas entrapment in the weld zone (Niessen, 2022). These entrapments appear in SEM images as cracks or areas of no contact.

In the medium angle range (8 - 10°), lower collision point velocities v_c allow the jet and particles to exit the collision gap more effectively. At the same time, the conditions in the collision gap (pressure, compression of the medium in the collision gap, gas friction due to the escaping jet, etc.) still raise the temperatures high enough to allow localised melting of the base materials, resulting in a weld zone with an interlayer (Fig. 4 (b), (c)). In the upper angle range (10 - 11.2°), the temperatures are no longer sufficient to induce melting, resulting in a smooth, interlayer-free joint (Fig. 4 (a)).

The highest shear strengths were recorded in the medium angle range, despite the dominance of the interlayer in the joints of this angle range. It is assumed that the material properties of the interlayer are inferior to those of the two base materials. The high shear strength values are explained on the one hand by the low thickness of the interlayer (average thickness 10 μm), and on the other hand by the fact that the interlayer is not continuous over the length of the joint zone. However it is always present in small sections, so the interlayer does not fully express the influence of its material properties. It is also assumed that the high shear strength of the joints mentioned is due to an increase in surface area caused by the wavy interlayer, which locally interlocks the two base materials. The presence of a wavy microstructure is also associated with favourable joint properties in terms of strength in (Carpenter and Wittmann, 1975; Wittmann, 1973), among others. However, when correlating the shear strength values with the surface areas of the joint structures, it is important to note that the determined area fractions are extrapolated from only two SEM cross sections per sample. Defects located between these sections, such as central voids, may not be detected and may lead to inaccurate strength assessments.

A comparison of the observed joint structures with the characteristic joint structure zones described in (Lysak and Kuzmin, 2003) shows that the smooth interlayer-free joint typically forms in the upper angular region of the process window and shifts to larger collision angles β with increasing collision point velocity v_c . At the investigated collision speed v_{imp} of 349 m/s, this region lies at the lower threshold, which may explain the relatively low shear strength values observed (on average 40 N/mm², see Section 3.4). As shown in (Lysak and Kuzmin, 2003), higher collision velocities shift this region to larger angles, suggesting that smoother joints could achieve higher strength at higher velocities. It is expected that an increase in collision speed will increase the welding process window for the material combination of AA6016-T4 and DC04, particularly in the direction of the upper limit angle. Based on the statements in (Lysak and Kuzmin, 2003), it is also expected that at higher collision speeds for the upper angle range, the specimens with a smooth joint structure without an interlayer will have higher tensile shear strengths.

At present, these assumptions cannot be verified due to the speed limitations of the current model test rig for collision welding. A planned upgrade to a maximum speed of 500 m/s will allow more comprehensive investigations. The correlation approach introduced in this work between shear strength and area fractions of different joint structures provides a promising basis. However, future investigations will require larger sample sizes and finer angular increments to minimise statistical scatter. In the context of electromagnetic pulse welding (EMPW), such an approach can be used to simulate the transient behaviour of the collision parameters and predict the resulting joint strength.

5 Conclusions

This study comprehensively investigates the collision welding process for the material combination DC04 and AA6016-T4. Using a mechanical collision welding test rig, different collision angles are analysed to determine their effect on the formation of the joint structure of the weld zone and the tensile shear strength of the welded joints. The results show that medium collision angles (8 - 10°) at a collision speed of 349 m/s produce the most favourable welding conditions, characterised by minimal cracking and unconnected structures, resulting in higher tensile shear strength. In future research, the planned extension of the model test rig to collision speeds of 500 m/s will allow the potential further increase in weld strength at higher collision speeds to be investigated. The use of larger sample sizes and the subdivision of the angular areas into finer increments as well as a complete analysis of the welded specimens with EDX will reduce the statistical scatter and provide more detailed insights into the properties of the weld zone.

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Grain Boundary Motion in Magnetic-Pulse-Welded Al-Fe Bimetal Systems: An Atomistic Simulation Study

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Abstract

Magnetic pulse welding (MPW) is an effective solid-state welding method of joining dissimilar metals such as Al/Steel, Al/Ti, Al/Cu et al. In order to understand fully such important phenomena as atomic diffusion, grain boundary motion, interfacial non-equilibrium-phase nucleation and growth in MPW, a detailed microscopic description of the MPW interface is necessary. This study is an extension of our previous work. In the present work, we extend the simulation investigation to polycrystal aluminium and iron systems using Molecular dynamic (MD) method. The polycrystal systems allow for the study of interfacial segregation. Our simulations present structural information on the GBs in nanocrystalline microstructures. Flat GBs can move when subjected dynamic load resulting from the high-velocity impact. Plots of the ratio of GB atoms versus time show a distinctly different GB migration behaviors between loading and unloading conditions. By contrast with our earlier simulations, it was observed that crystal order and stability are highly preserved in the loading stage. The transformation of grain boundary structural change is due to stress-driven GB migration and temperature dependent as well. Grain rotation mechanism was identified. This work could provide atomistic insights into the grain refinement during MPW process.

Keywords

Magnetic pulse welding, Molecular dynamics, Grain boundaries, Grain rotation

1 Introduction

Magnetic pulse welding (MPW) is a solid state impact welding process, which produces metallurgical bond by high-speed impact between two dissimilar materials (Psyk et al., 2011, Su et al., 2025). The ultrafine-grained structure on the MPW interface was commonly observed, which is regarded as the most essential interfacial property for high-strength joints (Zhang et al, 2022). Zhang et al. (2010) studied the Al-6061/Al-6061 and Cu-110 /Cu-110 interfaces formed by MPW. They reported a grain size of about 50 nm at the interface in the case of Al-6061/Al-6061 and nano-scale lamellar structures in the case of Cu-110/Cu-110. The authors concluded that the refinement of the interfacial structure is due to the rapid and severe plastic deformation at high strain rates. Zhu et al. (2023) observed that grain refinement occurs at the interface of 1050 Al and AZ31 Mg, which is regarded as a main reason for the excellent performance of the Al/Mg MPW joint. Raelison et al. (2024) reported a hierarchical nano-featured structure is confined within a very short total distance of about 200 nm at the Al/Cu interface during magnetic pulse impact welding. Outside these confined nanostructures, the interface exhibits a crystalline structure. Although a large volume of experimental data on the grain refinement generated by the magnetic pulse welding (MPW) process has been accumulated over the recent years, the above experimental research is not sufficient to understand its micro mechanism systematically.

Molecular dynamic (MD) simulations, on the other hand, can give real-time microstructure evolution with atomic-level details (Rapaport et al., 1996). Such atomistic simulations generate temporal-spatial information on three-dimensional atomic configurations and atom trajectories, which can be analysed to gain scientific insights into a physical system.

In this paper, we present the details of atomic motion and structural evolution across the interfacial zone for a Fe-Al system in the MPW process. Our goal is to provided illuminating insight into the atomic scale processes that occur during the MPW process.

2 Model and simulation methods

The open-source code, namely Large-scale Atomic/Molecular Massively Parallel Simulator (LAMMPS) package, was used to perform the large-scale molecular dynamics(MD) simulations. Three dimensional nanocrystalline Al-Fe samples with mean grain size of 5.5 nm have been constructed by a computer code. The selection of a mean grain size of 5.5 nm was primarily driven by its relevance to experimentally observed nanocrystalline structures and the critical need to capture essential grain boundary (GB) phenomena within computationally feasible limits. This specific size represents a deliberate balance between achieving physical representativeness and managing the substantial computational demands inherent in large-scale MD simulations. Grain orientations were randomly arranged in the samples. Fig. 1 presents the atomic configuration of the initialisation of the developed MD models, in which two metal bulks,

Al (84530 atoms) and Fe (116602 atoms) were put inside a simulation box with dimensions of 10 (X) $\times$ 10 (Y) $\times$ 48 (Z) nm³. Between them, vacuum is used to isolate the bulks' initial free surfaces. Periodic boundary conditions have been applied in the two transverse directions (x and y directions).

A constant integration time step of 1 fs was used for all simulations in this study. Simulation procedure is as follows. In the beginning process, Nose-Hoover thermostat was adopted and the whole system was relaxed at corresponding temperatures and 1 standard atmospheric pressure as an ambient pressure under the constant-pressure-temperature (NPT) ensemble to finish the initialisation. The initial thermal velocities of atoms were assumed to follow a Gaussian distribution. Subsequently, the Al bulk was given a velocity to simulate the collision. The system was relaxed for 1000 ps under the microcanonical (NVE) ensemble. In the end, with the final equilibrium temperature of the NVE simulation kept, the system was relaxed for another 1000 ps still at 1 standard atmospheric pressure (NPT ensemble). The 1000 ps relaxation period was explicitly chosen to achieve a sufficient duration for the system to establish thermodynamic equilibrium and dissipate any residual artificial stresses or energetic instabilities introduced during the initial setup. This deliberate equilibration step ensures that subsequent simulation data is collected from a stable, representative state of the system.

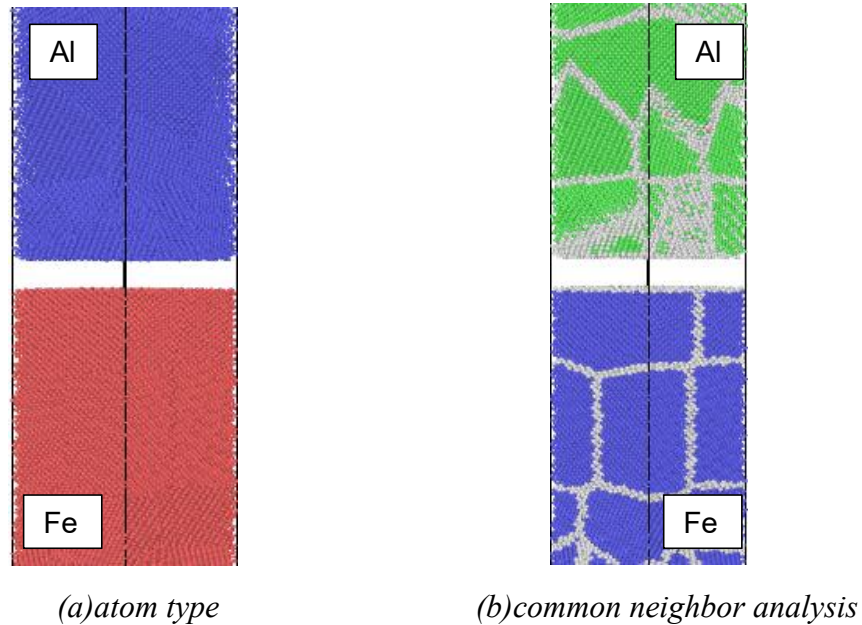


Figure 1: Initial configuration of the polycrystalline Al-Fe bimetal systems. (a) atom type identity and (b) common neighbor analysis showing grain boundaries.

The collisions with transverse velocity were simulated, which is consistent with the experimental study focusing on magnetic pulse welding. The axial velocity u_z was set at 450 m/s and the transverse velocity u_x was set at 250 m/s. The impact velocity applied in the Molecular Dynamics (MD) model is derived from the impact velocity information obtained through Finite Element Analysis (FEA) (Fan et al., 2016). Furthermore, the FEA

model itself is constructed under conditions identical to those of the Magnetic Pulse Welding (MPW) experiments. This consistency ensures that the velocity boundary conditions implemented in the MD simulations accurately reflect the dynamic loading boundary inherent in the MPW process.

An embedded-atom method (EAM) potential for Al-Fe bimetal systems was used in the simulations. The visualization of the atomic configurations was performed by the Open Visualization Tool (OVITO) software (Stukowski, 2009).

3 Results and discussion

To provide a more intuitive explanation of grain boundary (GB) deformation during the collision process, several grains in the Fe-side matrix from Figure 1(b) were numbered.

Figures 2(a)-(f) display the grain configurations of the Fe-side matrix at different time during the NVE ensemble simulation phase. From (a) to (c), the GB between G1 and G2 gradually thinned. The G1 grew larger and larger by incorporating surrounding disordered GB atoms, while the G2 shrank and became more prone to merging with adjacent larger grains. At $t=400$ ps, the shrinkage of G6 could be clearly seen through GB migration as G7, G8, and G9 absorbed nearby atoms. Concurrently, G12 and G13 gradually disappeared and merged with GBs of G10 and smaller grains below, ultimately forming a new grain G15. By $t=600$ ps (Figure 2(d)), G1 further expanded, and the GBs among G7, G8, and G9 merged to form a larger grain G16, while G11 was fully consumed by its adjacent larger grain G15. Due to the higher kinetic energy of atoms at GBs between small and large grains (Zhou, et al., 2014), these GBs exhibit enhanced atomic mobility and are more prone to migration.

Figures 2(d)-(f) reveal that disordered atoms gradually occupied lattice positions, causing GBs between grains to thin. For instance, the GB between G1 and G4 thinned and vanished, forming a new grain G18. Similarly, the GB between G16 and G18 disappeared, creating an even larger grain G19. GB migration arises from atomic movements perpendicular to the GB interface, where atoms diffuse from one grain to another, driving grain growth or shrinkage. This process ultimately leads to coarsening of Fe-side matrix grains during the loading stage. Grain boundary migration under shock loading leads to an alignment of grain boundaries perpendicular to the loading direction.

Figures 3 illustrate atomic configurations of a specific layer in the Fe-side matrix at 600 ps, 800 ps, and 1100 ps. Between 600 ps and 800 ps, G1 and G2 rotated clockwise. Under high-speed impact loading during magnetic pulse welding, GB atoms and grain atoms moved in mismatched directions, inducing mutual sliding that drove grain rotation under complex stress conditions.

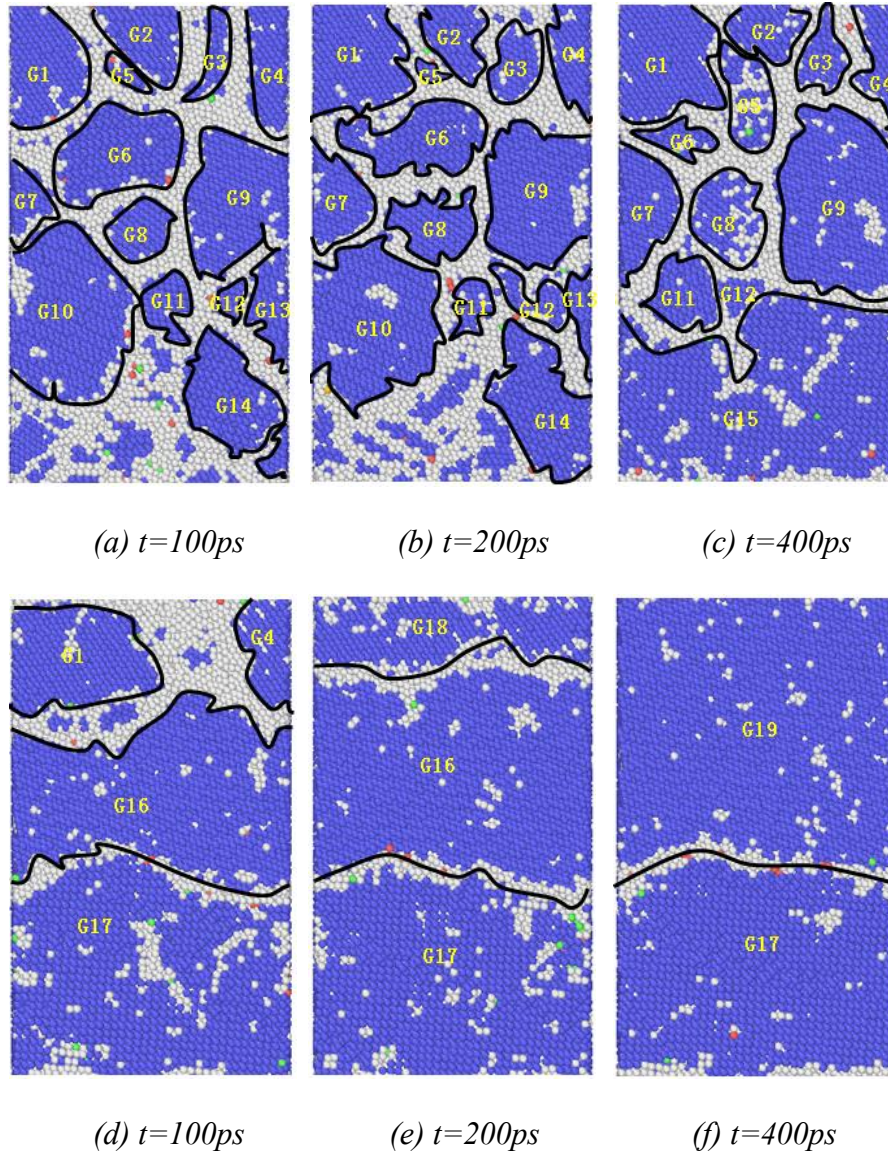


Figure 2: Snapshots of grain configurations of the Fe-side matrix at different time during the NVE ensemble simulation phase.

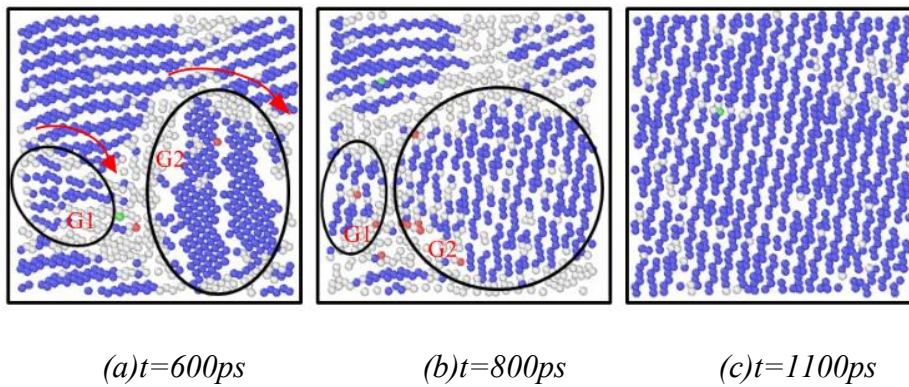


Figure 3: Atomic configuration of a layer of atoms on the Fe-side matrix at different times during the NVE ensemble simulation phase. The red arrow indicates the clockwise direction of rotation of the grain.

To further clarify GB evolution during Al/Fe magnetic pulse welding, statistical analyses of GB atom ratios in the matrix-side regions were performed. Figure 4 shows temporal variations in GB atom proportions across different zones. During the NVE ensemble mechanical relaxation phase, GB atom ratios in both Al- and Fe-side matrices progressively decreased, indicating that under high-pressure impact, fragmented grains reformed into coarser grains. Conversely, during the NPT ensemble thermal relaxation phase, GB atom ratios gradually increased and stabilized. This result shows that as pressure releases, the atomic activities around the GB were more intense at the high temperature, which led to an increase of average thickness of GB. Therefore, the movement of GB became the main factor that affected the microstructure of the Fe-side matrix.

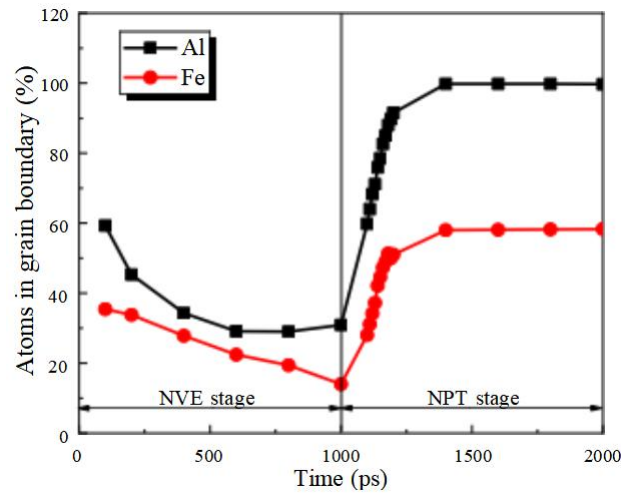


Figure 4: The fraction of atoms in grain boundary as a function of time.

Figure 5 shows bright-field (BF) and dark-field (DF) transmission electron microscopy (TEM) images of the Fe-side matrix microstructure adjacent to the Al/Fe welding interface. The BF image in Figure 5(a) reveals a lamellar grain structure in the Fe matrix. Fe grains directly adjacent to the Al/Fe interface exhibit nanoscale elongated morphologies, while slightly farther from the interface, the Fe grains on the steel side display stacked elongated configurations. These TEM experimental observations of grain morphology are consistent with the MD simulation results.

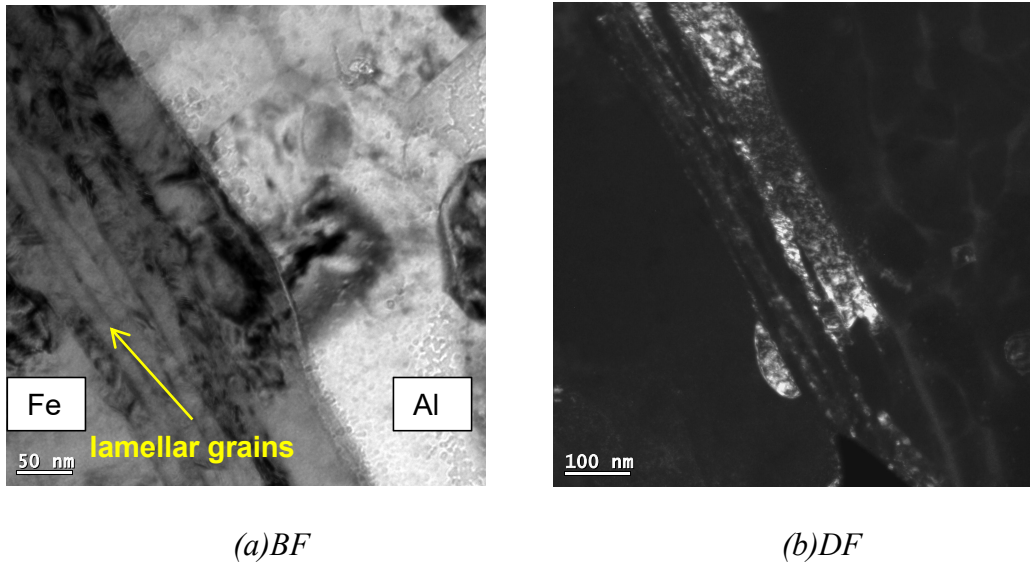


Figure 5: Transmission electron microscopy (TEM) images of the Al/Fe welding interface. (a) bright-field (BF); (b) dark-field (DF).

4 Conclusion

Due to the extreme difficulty of constructing and interpreting experiments on atomic-scale level, computer simulation has become an important tool, not only in its usual role of aiding in the development and evaluation of interface theories, but also in determining the basic microscopic phenomenology of crystal/ crystal interfaces. In the current work, the dynamic microstructure evolution and atomic motion in a magnetic-pulse-welded Al-Fe bimetal system have been investigated using the MD method. The key findings are summarized as follows:

- (1) Under impact loading, grain boundary (GB) atoms exhibit significant diffusion across adjacent grains, resulting in either grain growth or shrinkage. The deformation process is primarily governed by GB sliding accompanied by rotation of smaller grains.
- (2) Impact-induced grain boundary migration causes preferential alignment of GBs perpendicular to the loading direction, suggesting a stress-driven reorganization of the microstructure.
- (3) The MD-predicted lamellar nanocrystalline structure in the iron matrix shows good agreement with transmission electron microscopy (TEM) observations, validating the simulation approach.

These simulations provide atomic-level insights into the structural transformations and unique mechanical properties of MPW joints, bridging the gap between theoretical predictions and experimental observations. The findings contribute to a fundamental understanding of interface formation mechanisms in high-speed solid-state welding processes.

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Chapter 3

Welding other than electromagnetic

Combined use of High-Speed Videography and PDV Profilometry for Measuring High Speeds Flyers and Impact Shock Dynamics

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Abstract

Challenges remain in understanding the dynamic forces that occur with small scale explosive detonations and the subsequent impact those forces have on flyer velocity and deformation profiles. This work presents two data acquisition techniques for probing the behavior of various materials disturbed by laser induced detonations. The physical reactions of metallic flyers and transparent acrylic are cataloged and measured. The data acquisition techniques described here are high speed videography, Photon Doppler Velocimetry Profilometry (usually referred to as PDV Profilometry). High-speed videography provides a detailed temporal view of shock wave propagation, impact pressure distribution, and plasma plume dynamics. A Shimadzu Hypervision HPV-X2 camera was used at up to 10 million frames per second. For nanosecond scale velocity measurements PDV Profilometry was used to capture flyer velocity, acceleration, and deformation. PDV profilometry is a relatively new technique, here its ability to collect 8 or more velocity profiles at up to 750 km/s from a single PDV probe is demonstrated. Measurements of a shock front induced by the collision of a super sonic flyer against the surface of an acrylic cube is observed using the high speed camera. In this acrylic collision experiment impact and shock speeds up to 10km/s are recorded. Combining these techniques, a wide variety of experiments may be undertaken and observed with remarkable clarity in the nanosecond to microsecond time domain.

Keywords

Transient impact pressure, Pressure attenuation, Shock wave, Augmented laser impact Process, Photon Doppler Velocimetry (PDV), PDV Profilometry, High Speed Imaging

1 Introduction

Measurement techniques, for high-speed phenomena often suffer from a tradeoff between spatial and temporal resolution and often do not provide direct visualization of shock induced phenomena (Watts and Tassel, 1989), (Elkarous et al., 2016). In this study, two techniques for measuring highly transient shock induced phenomena with both good spatial and temporal resolution, high-speed videography and Photon Doppler Velocimetry Profilometry (PDV Profilometry) are presented. Although the tradeoff between spatial and temporal resolution still exists with these techniques, their respective strengths and ability to be used concurrently makes them highly useful and a complement to one another.

To show the capabilities of these two measurement techniques two experiments will be performed. Both experiments will utilize a 0.5mm thick 2024-T3 aluminum flyer plate, which will be driven to about twice the speed of sound by a chemically augmented laser impulse (Thurston et al., 2023). This flyer plate was 25mm square. In **Experiment 1** the aluminum flyer plate will be used to induce an impact shock wave in a transparent acrylic cube (25mm on each side). The shock wave will be recorded by the high-speed camera at 10 million frames per second (Shimadzu Hypervision HPV-X2). The properties of this shockwave will then be analyzed. In **Experiment 2** both PDV Profilometry and the high-speed camera will be used simultaneously to capture the changing velocity and profile of the aluminum flyer plate in flight. The high temporal resolution of the PDV Profilometry technique will be shown to complement the comparatively slow high-speed camera by capturing inter-frame flyer profile velocities with a much higher temporal resolution.

To drive the flyer plate used in both experiments an Augmented Laser Impact (ALI) process was used (Thurston et al., 2023). Previously, high intensity pressure pulses have been generated by laser shock for use in laser impact welding and laser shot peening (Thurston et al., 2023). While powerful, laser pulses have limitations when it comes to driving metallic flyer sheets thicker than about 0.25mm (Thurston et al., 2023). Energy may be added to these laser pulses via chemical augmentation. The chemical augmentation used in this work is provided by a mixture known as Gunpowder Nitromethane (GPN) (Abdelmaola et al., 2025). The GPN mixture used here consisted of 10g of 95% purity nitromethane and 1g of smokeless pistol powder. The chemical energy contained in the GPN mixture provides the majority of the energy needed to drive the 0.5mm thick flyer plate used in this work to useful speeds. A commercial laser system with a 3J 8ns pulse (Powerlite TM Precision II Scientific System from Continuum Lasers, Milpitas, CA, USA) was utilized to uniformly detonate the GPN mixture with a laser spot size of 5.22mm. The parameters of the laser system can be found in Table 1.

Power Density (GW/cm ²)	Energy (J)	Wavelength (nm)	Pulse Width (ns)	Spot Size (mm)
1.34	3.14	1064	8	5.22

Table 1: The parameters of the Laser system

To make efficient use of the pressure generated by the GPN detonation it is necessary to have the liquid GPN confined between the flyer plate and a transparent tamping medium. The tamping medium's mass and inertia direct the pressure pulse into the flyer plate, thereby making efficient use of the detonation pressure. When laser impulses are used to detonate the GPN the tamping medium must be transparent, therefore in **Experiments 1 & 2** a borosilicate glass tamping layer is used. For this work the borosilicate glass tamper was 9mm thick and was roughly 25x25mm square. This tamper can be seen in **Figure 1A**. Because GPN is a liquid, it must be contained by some method that prevents it from leaking out of the experimental setup and which sets the layer thickness of the GPN. In this work layers of double-sided tape are used at the periphery of the GPN layer to both confine it and to set the layer thickness. Here 4 layers of Scotch brand double sided tape were used. The GPN layer thickness defined by these tape layers was 0.37mm, a layer thickness found to offer the greatest performance (Thurston et al., 2023). The positions of the tamping layer, flyer, and GPN layer are depicted in **Figure 1A**.

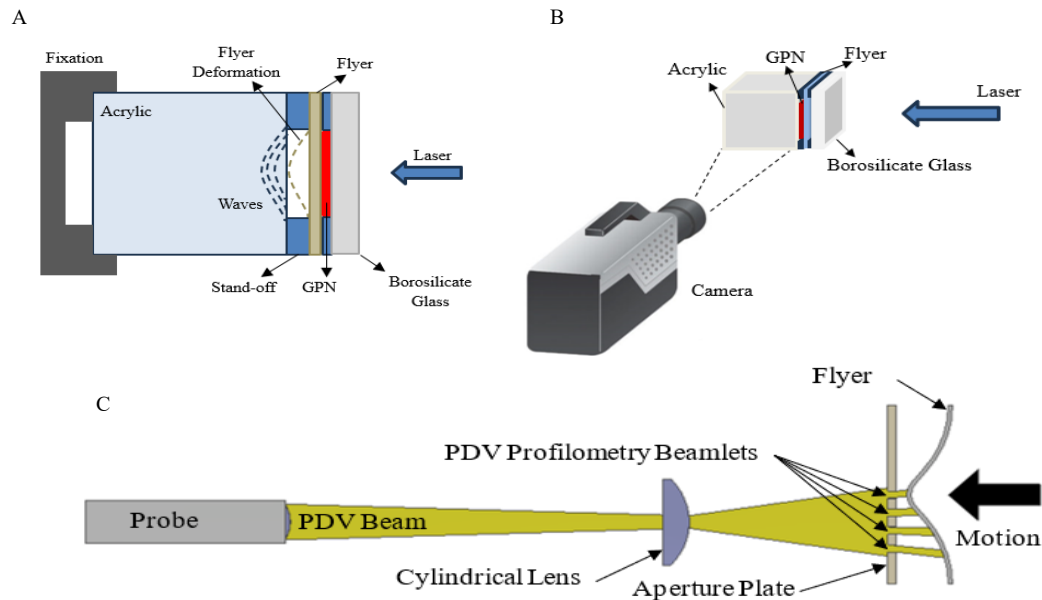


Figure 1: The setups for both **Experiment 1** and **Experiment 2** are shown. **A:** The schematic of the setup for **Experiment 1** showing the deformation of the flyer and the shock wave propagation after the flyer collides with the acrylic cube. The location of the GPN augmentation layer and the borosilicate glass tamping layer are shown. The direction of the laser pulse is also indicated. **B:** The High-Speed Camera configuration used to record **Experiments 1 & 2**. **C:** The setup for PDV Profilometry in **Experiment 2**. This setup for **Experiment 2** is identical to the setup for **Experiment 1**, with the exception that the acrylic is replaced by the cylindrical lens and aperture plate.

The setup for **Experiment 1** is most clearly depicted in **Figure 1A**. As stated previously, this experiment will capture shock propagation in an acrylic cube. As depicted in **Figure 1A**, the 25mm square, 0.5mm thick, 2024-T3 aluminum flyer will be driven toward the acrylic cube by a GPN augmented laser pulse. The aluminum flyer will travel through a standoff gap of 2mm before colliding with the surface of the acrylic cube. At high enough impact speeds a shock wave will be induced in the acrylic. Since the acrylic is transparent it should be possible to observe shock wave propagation with the use of the high-speed camera. As depicted in **Figure 1B**, the high-speed camera will be placed perpendicular

to the direction of flyer motion where it will have a clear view of both flyer motion and shock propagation in the acrylic cube.

Experiment 2 will involve almost the same setup as **Experiment 1**, with the exception that the acrylic cube will be replaced by hardware needed for PDV Profilometry. Typically PDV makes use of a single probe to measure the velocity of a surface by measuring the frequency of a beat frequency created by mixing of a source laser beam and light reflected from the moving surface in question. This beat frequency can be recorded by an oscilloscope and is directly proportional to the velocity of the surface under investigation (Strand et al., 2006). PDV Profilometry takes advantage of the fact that modern signal processing techniques (such as the moving window Fourier transform) can be used to extract multiple beat frequencies from a single PDV signal. As depicted in **Figure 1C**, a single PDV beam may be spread out and divided into individual beamlets by the combination of a cylindrical lens and an aperture plate. So long as the velocity of the flyer under each beamlet is different, the signal recorded on the oscilloscope will contain different beat frequencies for each beamlet. Similarly, because each beamlet has a width, a band of beat frequencies will be generated if a difference in flyer velocity exists across the beamlet. In **Figure 1C** the aperture plate is shown exposing points along the radius of the flyer, therefore ensuring that beamlet should cover a portion of the flyer with a different velocity. In **Experiment 2** the same flyer from Experiment 1 will be recorded moving through free space by both the high-speed camera and PDV Profilometry. The velocity signals from PDV Profilometry will be integrated over time to provide displacement curves that can be compared against measurements from the high-speed camera.

2 Experiment 1 Results and Analysis

This work for Experiment 1 is expected to appear in an upcoming publication by M. Abdelmaola. **Figure 2** provides a sample of frames taken by the high-speed camera of the impact between the aluminum flyer and acrylic, as well as of the shock wave induced in the acrylic.

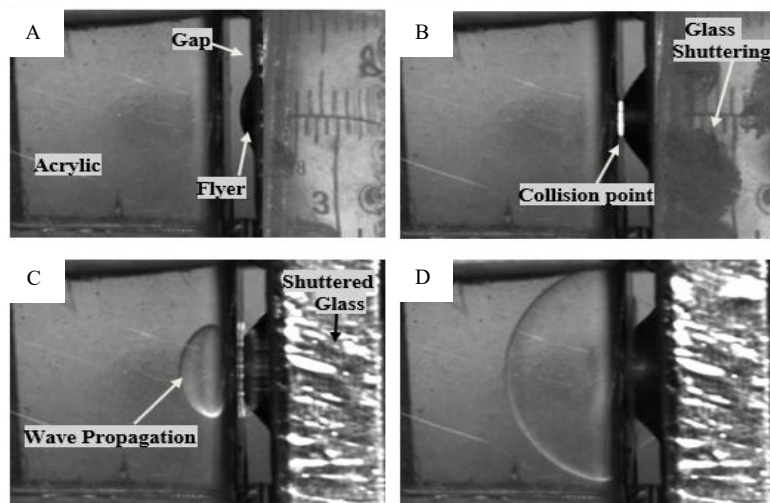


Figure 2: A: Flyer deformation after the detonation of GPN. B: Collision with target. Here the flyer was traveling 740m/s at impact. C & D: Wave propagation through the acrylic.

A direct measurement of the shock wave velocity was taken by measuring the change in position of the shock wave between frames as it traveled outward. These measurements

were taken at the central, left most point, of the shock wave. The velocity of the shockwave over time is shown in Figure 3. The peak shock speed velocity shown in Figure 3 is 3428m/s. As should be expected, the shock speed appears to asymptotically fall toward the speed of sound in acrylic, 2,750m/s (Hydrosight GmbH). While the high-speed camera can capture an impressive 10 million frames per second, the velocity of the shock wave when measured by the high speed camera shows unrealistic deviations in velocity. These deviations are due to the relatively course resolution of this camera at 10 million frames per second (200x125 pixels).

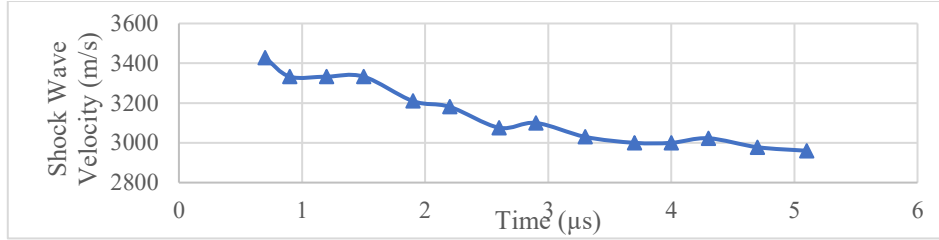


Figure 3: Shock Wave velocity plotted over time for the impact between the aluminum flyer and the acrylic cube. Here time is measured from the moment of contact. The shock wave does not come into view until 700ns after impact.

Based on the shock velocity over time curve, and with the equation for shock pressure, $P_2 = P_1 + \rho U_s U_p$ (Marsh, 1980), it is possible to generate a pressure vs time plot for the shockwave traveling within the acrylic cube. Here, P_1 represents atmospheric pressure ($P_1 \approx 0$ for high-speed impact), ρ is the density of the target material ($1,190 \text{ kg/m}^3$ for acrylic (Hydrosight GmbH)), U_s is the shock wave velocity, and U_p is the particle velocity. P_2 is the pressure behind the wave front, or shock pressure. The particle velocity U_p can be derived from the following equation: $U_p = (U_s - C_0)/(1 + (U_s - C_0)/C_0)$ (Marsh, 1980), where C_0 indicates the speed of sound through the target material ($2,720 \text{ m/s}$ for acrylic). From these equations and the previously plotted shock wave velocity, the shock pressure over time was calculated and then plotted in Figure 4.

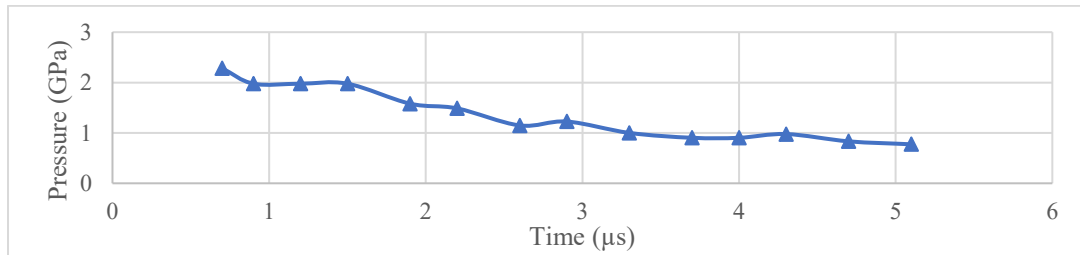


Figure 4: Pressure behind the wave front after the impact between the aluminum flyer traveling at 740m/s and the acrylic cube. Here time is measured from the moment of contact. The shock wave does not come into view until 700ns after impact.

3 Experiment 2 Results and Analysis

The work for **Experiment 2** originally appeared in a PhD dissertation by Thurston (Thurston, 2023). The PDV Profilometry data for **Experiment 2** was collected alongside use of the high-speed video camera. The combined beat frequencies from the return light from each of the beamlets was collected and passed through a moving window Fourier transform. The resulting spectrogram can be seen in Figure 5. In Figure 5, four clear signal bands can

be seen. The top signal is from the beamlet over the center of the flyer. The next three signal bands down are from the three beamlets adjacent to the centermost beamlet. A fifth partial signal band from the furthest radial hole in the aperture plate can also be seen.

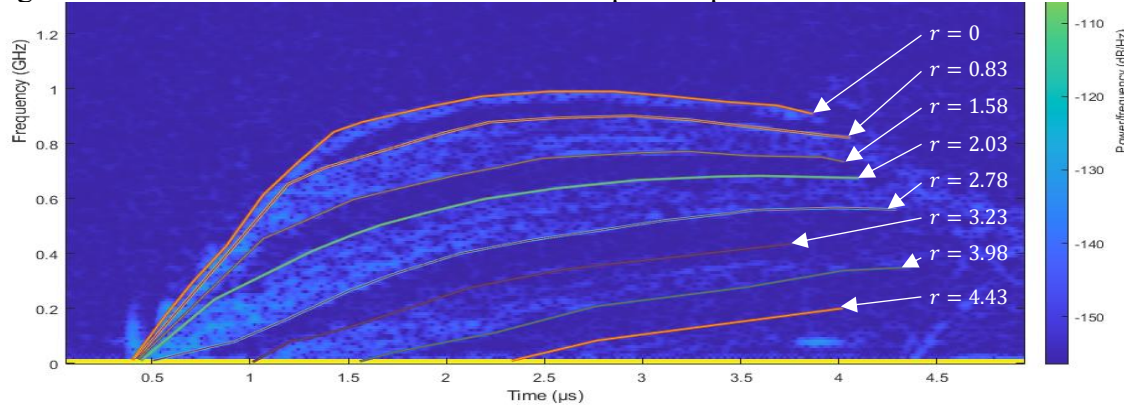


Figure 5: Spectrogram with PDV Profilometry velocity lines digitized. The radial positions of each line are indicated. The values are in mm.

The spectrogram in **Figure 5** presents both digitized flyer velocity profiles and the original frequency intensity over time. This spectrogram has a time resolution of 10ns, however the actual time resolution of the spectrogram will vary with the returned frequency, spectrogram calculation parameters, and returned signal intensity. Here the actual time resolution is clearly on the order of 100ns or better. The solid lines on the spectrogram in **Figure 5** are manually digitized representations of the tops and bottoms of the signal bands, which represent the inner and outer edges of each beamlet. This is with the exception of the central beamlet where only the top is digitized, and the fifth partial signal band, for the fifth beamlet, which is only partly present. The central beamlet, which is over the center of the flyer and therefore will have the greatest velocity, is partially digitized because the left and right edges of the beamlet are adjacent to the center of the flyer. Because the locations of the holes in the aperture plate are known it is possible to correlate each of the lines in the **Figure 5** spectrogram with a particular radial coordinate on the surface of the flyer. These radial coordinates are indicated in **Figure 5**. As previously mentioned, the frequencies seen in the spectrogram can be directly converted into velocities, and these velocities can be integrated to produce curves for displacement over time. Because multiple displacement curves are available, and because each displacement curve corresponds to a different radial position on the flyer, it is possible to reconstruct the shape evolution of the flyer over time.

The shape evolution reconstruction of the flyer is shown in **Figure 6**. The PDV Profilometry data has been reflected across the X axis for better visualization.

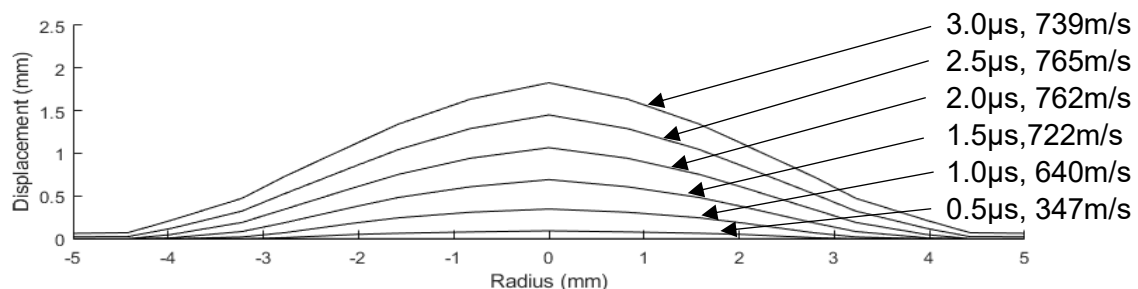


Figure 6: Reconstructed flyer displacement with flyer tip speed indicated.

While the data is represented in **Figure 6** is presented as discrete profiles in steps of 0.5μs it should be noted that the actual PDV Profilometry data has a time resolution in the

tens of nanoseconds or less. The last profile given in **Figure 6**, at $3.0\mu\text{s}$, has a displacement approaching 2mm, the same as the standoff distance between the flyer and the acrylic cube in **Experiment 1**.

In an effort to compare the strengths and weaknesses of the high-speed camera and PDV Profilometry an unconstrained flyer was filmed under the same conditions as the flyer used to collect the PDV Profilometry data, with the exception that the aperture plate was removed. In **Figure 7** several frames collected by the high-speed camera from this flyer are presented at intervals of $0.5\mu\text{s}$. The shape evolution of the flyer shown in these images superficially resembles the shape evolution reconstructed from the PDV Profilometry data presented in **Figure 6**. In order to make a more accurate comparison between the PDV Profilometry data and the high-speed camera data a MATLAB script was written to extract the leading edge of the flyer profiles from the high-speed camera images.

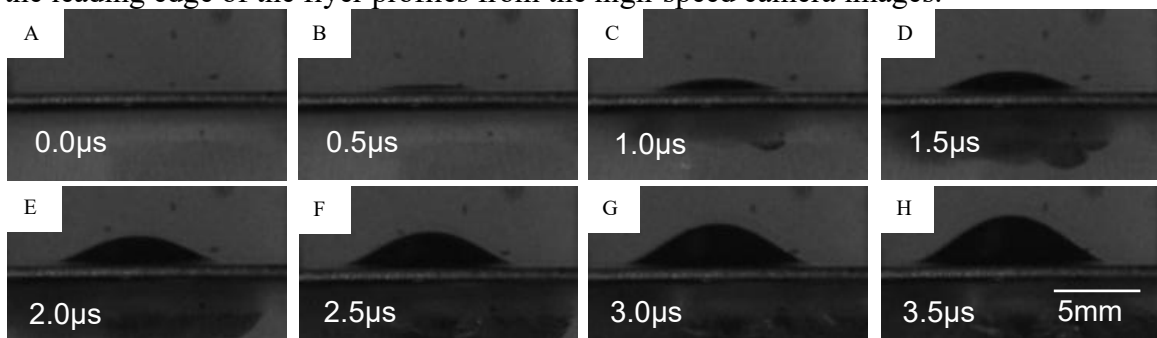


Figure 7: Frames collected by the high-speed camera.

As can be seen in **Figure 8**, the PDV Profilometry data compares well to the flyer profiles generated from the high-speed camera footage. In **Figure 8** the high-speed camera flyer profiles are shown overtop of the PDV Profilometry profiles for better comparison. The regions of the flyer away from the center point show greater displacement in the high-speed camera profiles as compared to the PDV Profilometry derived profiles. This may be due to shot-to-shot variability in the Augmented Laser Impact process, or calibration issues with the scaling of the high-speed images. For better comparison PDV Profilometry and high-speed photography should be done on the same experiment. This was not done here due to time constraints with the high-speed camera.

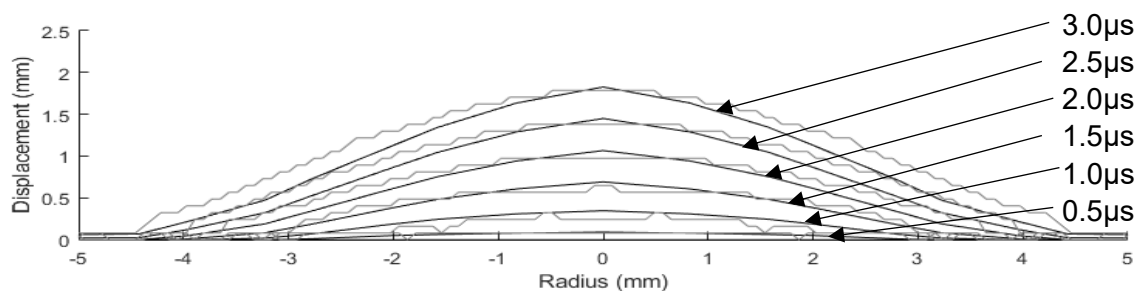


Figure 8: High-speed camera flyer profiles overlaid on PDV Profilometry data.

4 Conclusions

In this work two techniques of data acquisition at high speeds were presented. The high-speed camera was shown to offer great spatial resolution, along with a time resolution of 10 million frames per second. It was shown that for transparent media the high-speed

camera can be used to measure shock wave velocities. From those shock wave velocities, it is possible to calculate shock pressures in the transparent media as a function of time. PDV Profilometry showed that adaptations to the standard PDV technique can allow for easy extraction of flyer velocity profiles, and by integration, flyer displacement profiles over time can also be gathered. PDV Profilometry velocity and displacement data can have a tremendously high time resolution as compared to even the best high-speed cameras. Here the time resolution of the PDV Profilometry data was better than 100ns, ten times faster than the high-speed camera. Although the spatial resolution of PDV Profilometry is somewhat lacking, PDV and PDV Profilometry serve as a complement to the high-speed camera both as a means of cross validation, and as a means of measuring velocity data directly at a much higher time resolution. It is hoped that future researchers will see the utility in these data acquisition techniques for short duration and high velocity experiments.

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Feasibility and Mechanistic Analysis of 3-Stack Aluminium-Steel Spot Joints Using Vaporizing Foil Actuator Welding with Partial Pre-Forming

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Abstract

The advancement of spot joining techniques is crucial for automotive manufacturing, yet integrating aluminum-steel (Al-Fe), combinations presents inherent challenges. Traditional methods, such as resistance spot welding (RSW), often lead to the formation of brittle intermetallic compounds (IMCs) and residual stresses, compromising joint integrity. Innovative strategies that enhance mechanical performance while optimizing weight are essential for multi-material assemblies. Advanced techniques like vaporizing foil actuator welding (VFAW) offer a promising solution, enabling metallurgical bonding with minimal heat-induced damage, thereby reducing residual stresses and improving structural reliability. This study investigates the feasibility of fabricating a 3-stack (Al-Fe-Fe) spot joint using VFAW, employing partial pre-forming of intermediate plates to achieve single-step joining. Smoothed particle hydrodynamics (SPH) modeling is applied to analyze transient stress wave propagation and interfacial jetting dynamics, elucidating their mechanistic influence on interfacial bond formation during high-speed multi-plate interactions. The findings underscore the critical role of partial pre-forming in enabling successful 3-stack VFAW joints.

Keywords

Vaporizing foil actuator welding (VFAW), Al/Steel, 3-stack, Process-property relationship, SPH

1 Introduction

The integration of aluminum and steel in automotive manufacturing offers significant advantages in achieving lightweight, high-strength structures (Yang et al., 2024). Dominantly, joining techniques such as resistance spot welding (RSW) are used in the industry; however, faces significant challenges, primarily due to the formation of brittle intermetallic compounds (IMCs) at the joint interface, which compromises mechanical performance (Shi et al., 2025, Kumar et al., 2024). Additionally, disparities in melting points and thermal conductivities between aluminum and steel exacerbate welding difficulties (Kulkarni et al., 2022).

Three-stack or multi-stack joints involving layered combinations of aluminum and steel are particularly relevant for automotive applications due to their ability to integrate lightweight and high-strength materials efficiently. However, achieving reliable multi-stack joints remains challenging due to uneven stress distribution and potential defects such as cracks or delamination (Shi et al., 2023). Traditional methods like RSW often struggle with such configurations, leading to inconsistent joint quality (Achira et al., 2022, Brechelt et al., 2020). To address these challenges, advanced joining techniques like VFAW are being explored for their potential to minimize thermal effects while maintaining joint integrity.

Vaporizing Foil Actuator Welding (VFAW) emerges as a compelling alternative, utilizing high-voltage (current) pulses (100–150 kA) to sublimate an aluminum foil, creating high-pressure plasma pulse to propel flyer metal plates into collision with target metal, at high velocities (300–1500 m/s) and oblique angles (5–20 degrees), facilitating solid-state bonding without significant thermal input (Vivek et al., 2013, Kapil et al., 2025). This process effectively minimizes IMC formation, thereby enhancing joint integrity (Kumar et al., 2025).

To elucidate the underlying mechanisms of VFAW and optimize the process parameters for specific applications, Smoothed Particle Hydrodynamics (SPH) modeling serves as a robust analytical tool. Its mesh-free nature adeptly captures the high strain rates and complex interfacial phenomena inherent in VFAW, providing critical insights into stress wave propagation and jetting dynamics during the welding process.

Previously, some researchers (Mao et al., 2021, Kapil et al., 2020 and Li et al., 2021) have already utilized VFAW for a multi-stack configuration. As Mao et al., 2021, demonstrated that using a thin Al 3003-H14 interlayer between thick aluminum alloy 5754 and HSLA 340 steel achieved robust joints with reduced IMC formation due to controlled energy dissipation. Similar studies by Li et al., 2021 on heat treatment of interlayers revealed that treating the interlayer material can stabilize the microstructure, thereby improving mechanical properties and fatigue resistance. Furthermore, Kapil et al., 2020 highlighted that employing interlayers in VFAW not only aids in energy absorption during high-velocity impacts but also helps prevent localized melting, thereby maintaining joint quality and structural integrity. However, in all these studies, the middle layer was significantly thinner than both the flyer and target plates, primarily serving as an interlayer rather than a structural component.

Therefore, this study aims to investigate the feasibility of fabricating multi-stack aluminum-steel-steel spot joints using VFAW, with an emphasis on partial pre-forming technique where all three plates have comparable thickness and the middle layer functions as a structural component rather than merely an interlayer. This approach addresses the challenges associated with achieving consistent bonding and mechanical integrity in multi-layer configurations. Through comprehensive SPH modeling, the study will analyze stress distributions, interfacial dynamics and the influence of joint geometry on joint quality. The insights gained from the analytical model combined with SPH simulations are expected to guide process optimization, advancing the application of VFAW in producing robust, lightweight automotive structures, particularly where multi-material integration is critical.

2 Methodology

2.1 Materials and methods

The selected materials include an A6451P-T4 aluminum alloy flyer plate with a thickness of 1.5 mm, chosen for its lightweight properties, formability, and structural performance, making it suitable for automotive applications (Kumar et al., 2025). The intermediate plate, as well as the target plate, consists of SPCC (General purpose cold-rolled carbon steel) steel with a thickness of 0.8 mm, where the intermediate plate acts as a structural component to enhance joint strength and stability (Liu et al., 2025). The mechanical properties and chemical composition of the materials are shown in Table 1. An aluminium foil actuator with a thickness of 0.0762 mm serves as the vaporizing medium, positioned parallel to the weld axis to ensure consistent acceleration during the welding process.

The VFAW process was conducted using the APPLIED IMPULSE INC.® CIS-6 system with a maximum capacity of 6 kJ, as shown in Fig. 1 (a). The experimental setup involved arranging the three plates in a pre-formed configuration to accommodate rapid flyer acceleration and minimize initial resistance (Fig. 1(b)).

Partial pre-forming (i.e. forming to a depth lower than the pre-forming of the target plate) of the intermediate steel plate was employed to create a controlled gap, enabling efficient collision dynamics and uniform stress distribution during welding. This setup aimed to address the challenges of achieving consistent bonding when the intermediate layer functions as a structural component rather than just an interlayer (Fig. 1 (c-d)).

During the welding process, the aluminum actuator foil rapidly vaporizes, generating a high-pressure plasma pulse that propels the A6451P-T4 flyer towards the pre-formed SPCC intermediate plate. The initial impact occurs between the flyer and the intermediate layer, followed by a secondary impact between the intermediate and target layers. The use of a structurally comparable intermediate plate rather than a typical interlayer ensures that the joint can withstand higher mechanical loads while maintaining robust bonding and load transfer.

SPH modeling was conducted to model the dynamic interaction between the flyer, intermediate, and target plates. The model analyzed stress wave propagation, and interfacial

jetting phenomena, focusing on how the comparable thickness of the intermediate layer influences the bond quality compared to configurations with thinner interlayers (Mao et al., 2021). The simulations aimed to elucidate the mechanistic role of partial pre-forming in achieving effective metallurgical bonding in multi-stack joints.

Table 1: Mechanical properties and chemical composition of the workpiece material

Material	Mechanical Properties			Chemical Compositions (At%)					
	YP (MPa)	TS (MPa)	EL (%)	Fe	C	Mn	Si	Mg	Al
SPCC	176	298	50	Bal.	0.003	0.08	0.002	-	-
A6451P-T4	136.6	266.1	28.7	0.17	-	0.072	1.1	0.52	Bal.

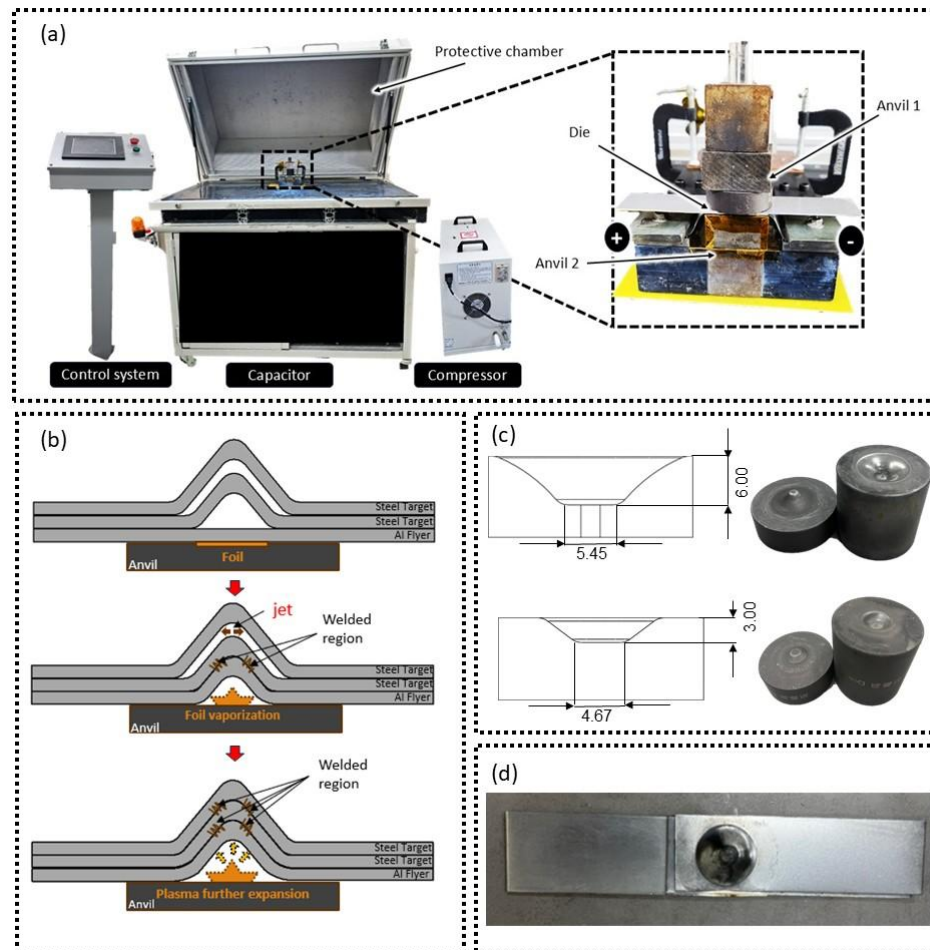


Figure 1: Pictorial representation of (a) working setup of VFAW, (b) welding process, (c) die design used during preforming and partial preforming and (d) experimentally obtained Al-Fe-Fe VFAW welded sample

2.2 Analytical modeling

In this one-dimensional (1-D) analytical approach, Plate 1 (flyer) is accelerated and impacts Plate 2 (intermediate), which is deformable, of comparable thickness, and initially unsupported, consistent with the experimental configuration. Following this, the intermediate plate deforms and travels across a finite standoff distance to impact Plate 3 (target), which is rigidly backed. This configuration divides the modelling into two sequential impact events synonymous with the experiments.

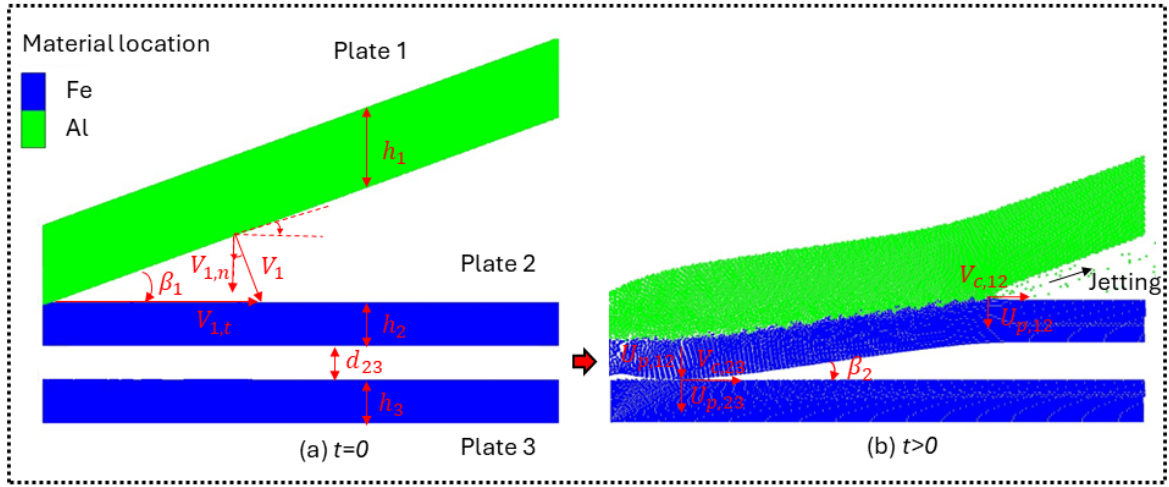


Figure 2: Pictorial representation of oblique impact at (a) interface 1 and (b) interface 2

2.2.1 Interface 1: flyer (plate 1) → intermediate (plate 2)

When Plate 1 impacts Plate 2 obliquely at an angle β_1 , shock waves are generated on both plates. The velocity components of the flyer in the normal and tangential directions are given by,

$$V_{1,n} = V_1 \sin \beta_1, V_{1,t} = V_1 \cos \beta_1, V_{2,n} = 0 \quad (1)$$

Applying the two-sided Rankine-Hugoniot relations equating pressure and particle velocities,

$$P_1 = P_2 = P_{int1}, U_{p,1} = U_{p,2} = U_{p,int1}. \quad (2)$$

$$P_1 = \rho_1(C_{01} + S_1 U_{p,1})U_{p,1}, P_2 = \rho_2(C_{02} + S_2 U_{p,2})U_{p,2}. \quad (3)$$

At interface 1, $P_1 = P_2$ and $U_{p,1} = U_{p,2} = U_{p,12}$ (shared particle velocity). As they are no longer symmetric, and both plates move post-impact. Applying conservation of momentum and mass,

$$\rho_1(C_{01} + S_1 U_P) U_{P,12} = \rho_2(C_{02} + S_2 U_{P,12}) U_{P,12} \quad (4)$$

$$\rho_1(C_{01} + S_1 U_{P,12}) = \rho_2(C_{02} + S_2 U_{P,12}) \quad (5)$$

$$U_{P,12} = \frac{\rho_2 C_{02} - \rho_1 C_{01}}{(\rho_1 S_1 - \rho_2 S_2)} \quad (6)$$

$$\text{Or using impedance, } U_{P,12} = V_i \frac{Z_2 - Z_1}{Z_1 + Z_2}, Z_i = \rho_i C_{0i} \quad (7)$$

This provides the impact-driven interface velocity in terms of material impedances. The resulting shock velocities in the flyer and intermediate plate are given by,

$$U_{s1} = C_{01} + S_1 U_P = C_{01} + S_1 \frac{\rho_2 C_{02} - \rho_1 C_{01}}{\rho_1 S_1 - \rho_2 S_2} = \frac{\rho_2 (S_1 C_{02} - S_2 C_{01})}{\rho_1 S_1 - \rho_2 S_2} \quad (8)$$

$$U_{s2} = C_{02} + S_2 \frac{\rho_2 C_{02} - \rho_1 C_{01}}{\rho_1 S_1 - \rho_2 S_2} = \frac{\rho_1 (S_1 C_{02} - S_2 C_{01})}{\rho_1 S_1 - \rho_2 S_2} \quad (9)$$

This captures the oblique shock propagation in both plates. However, as both plates are deformable with a comparable thickness, the shock reflects from the rear faces of both plates and returns towards interface 1. Their respective arrival times at the interface 1 are,

$$t_1^* = \frac{2h_1}{U_{s1} \cos \beta_1} = \frac{2h_1(\rho_1 S_1 - \rho_2 S_2)}{\rho_2 (S_1 C_{02} - S_2 C_{01}) \cos \beta_1} \quad (10)$$

$$t_2^* = \frac{2h_2}{U_{s2} \cos \beta_1} = \frac{2h_2(\rho_1 S_1 - \rho_2 S_2)}{\rho_1 (S_1 C_{02} - S_2 C_{01}) \cos \beta_1} \quad (11)$$

The earlier of the two times will dictate the potential interference with melt solidification, i.e., weld failure by dynamic solidification cracking which occurs when a reflected shock wave reaches the molten/hot zone before it solidifies.

$$t^* = \min(t_1^*, t_2^*) \quad (12)$$

$$\text{Trailing edge of the hot zone at that time, } x_1^* = V_{c,12} t_1^* - b = \frac{2V_{c,12} h_1 (\rho_1 S_1 - \rho_2 S_2)}{\rho_2 (S_1 C_{02} - S_2 C_{01}) \cos \beta_1} - b \quad (13)$$

$$\text{Here, } V_{c,12} = \frac{U_{P,12}}{2 \sin\left(\frac{\beta_2}{2}\right)} \quad (14)$$

For shock reflection to be critical, $x^* = 2h_1 \tan \beta_1$

$$\frac{2V_{c,12} h_1 (\rho_1 S_1 - \rho_2 S_2)}{\rho_2 (S_1 C_{02} - S_2 C_{01}) \cos \beta_1} - b \geq 2h_1 \tan \beta_1 \quad (15)$$

$$\text{Similarly, for the intermediate plate, } \frac{2V_{c,12}h_2(\rho_1S_1-\rho_2S_2)}{\rho_1(S_1C_{02}-S_2C_{01})\cos\beta_1} - b \geq 2h_2 \tan\beta_1 \quad (16)$$

Eq. 15 and Eq. 16 define when shock-induced tensile stress overlaps the hot zone, and the first among them to satisfy this will define the actual upper limit.

The maximum effective hot zone half-width $b_{1,max}$ is estimated from energy balance and thermal diffusion considerations as follows,

$$b_{1,max} = V_{c,12} \cdot \min(t_1^*, t_2^*) - 2h_1 \tan\beta_1 \quad (17)$$

Assuming an ideal conversion, where heat is required to reach the melting temperature of one of the materials at interface 1 is equal to the transferred kinetic energy converted to thermal energy over the hot zone. Therefore, the thermal energy input is defined as,

$$Q = \frac{1}{2}f\rho_1h_1V_1^2 \cdot V_{c,12} \quad (18)$$

$$\text{Effective thermal properties are } k' = \frac{2k_1k_2}{k_1+k_2}, \alpha' = \frac{2\alpha_1\alpha_2}{\alpha_1+\alpha_2} \quad (19)$$

$$\text{The temperature rise in hot zone with width } 2b \text{ becomes, } T_m = \frac{Qb}{k'\sqrt{\pi B}} \quad (20)$$

Stress-wave arrival time = hot zone trailing edge position. f is the kinetic energy conversion factor, i.e. the fraction of the kinetic energy of the flyer being converted into thermal energy at the weld interface. Typical value 0.9 (i.e. 90% of energy).

$$B = \frac{h_1V_{c,12}}{\alpha' \cos\beta \left(\frac{V_c}{U_{s1}} - \sin\beta_1 \right)} \quad (21)$$

B is Dimensionless Thermal Strip Parameter, A thermal modeling parameter based on the moving strip heat source model. V_c Weld Velocity (Collision Front Speed)

$$V_c = \frac{V_i}{2 \sin\left(\frac{\beta_1}{2}\right)} \quad (22)$$

Solve for V_i such that $T_m = \min(T_{m1}, T_{m2})$, V_i is the initial velocity of the flyer plate in its direction of travel at the moment of impact. Heat required to reach melt temp = Kinetic energy converted to thermal energy over a hot zone. Stress-wave arrival time = hot zone trailing edge position. Final symbolic expression

$$V_{i,max} = \sqrt{\frac{\min(T_{m1}, T_{m2}) \cdot k'}{f' h_1^2 V_c} \left(\frac{V_c}{\min(U_{s1}, U_{s2}) \cos \beta_1} - \tan \beta_1 \right)} \quad (23)$$

The shock velocity is now whichever arrives first. $U_s = \min(U_{s1}, U_{s2})$

$$V_{i,max} = 2.239 \frac{\sqrt{k_1} \sqrt{k_2} \cdot (C_{01} + S_1 U_{p1})^{\frac{1}{4}} \cdot \sqrt{\min(T_{m1}, T_{m2})} \cdot \rho_2^{\frac{1}{4}} \cdot (\alpha_1 + \alpha_2)^{1/4}}{V_c^{\frac{1}{4}} \cdot \sqrt{b} \cdot h_1^{\frac{1}{4}} \cdot \sqrt{\rho_1} \cdot \sqrt{k_1 + k_2} \cdot \sqrt{f} \cdot (\alpha_1 \alpha_2)^{\frac{1}{4}}} \cdot \left(\frac{S_1 C_{02} - S_2 C_{01}}{(C_{01} \rho_2 S_2 \sin \beta_1 - C_{02} S_1 \rho_2 \sin \beta_1 + S_1 V_c \rho_1 - S_2 V_c \rho_2) \cos \beta_1} \right)^{\frac{1}{4}} \quad (24)$$

2.2.2 Interface 2: intermediate (plate 2) → target (plate 3)

Intermediate is now the new flyer, impacting the target at velocity $V_{impact,2 \rightarrow 3} \approx U_{p,12}$. Target is not massless and final interface has its own impedance, so shared velocity changes, So we must re-solve the Rankine-Hugoniot equations. Intermediate flyer velocity $= U_{p,12}$ and new shared particle velocity $U_{p,23}$

$$U_{p,23} = \frac{\rho_3 C_{03} - \rho_2 C_{02}}{(\rho_2 S_2 - \rho_3 S_3)} \quad (25)$$

$$t_{delay} = \frac{d_{23}}{U_{p,12}} = \frac{d_{23}(\rho_1 S_1 - \rho_2 S_2)}{\rho_2 C_{02} - \rho_1 C_{01}} \quad (26)$$

$$\text{Shock Velocity Into Final Target (Plate 3), } U_{s3} = C_{03} + S_3 U_{p,23} = C_{03} + S_3 \frac{\rho_3 C_{03} - \rho_2 C_{02}}{(\rho_2 S_2 - \rho_3 S_3)} = \frac{\rho_2(S_2 C_{03} - S_3 C_{02})}{(\rho_2 S_2 - \rho_3 S_3)} \quad (27)$$

U_{s3} propagate into the target, reflect off the rigid base and return to interface 2. Another shock wave reflects back into plate 2, and a new wave re-enters plate 2. At interface 2, the intermediate plate (2) impacts the final target (3) after travelling a distance d_{23} at velocity $U_{p,12}$. Interact with the remaining molten zones between plates 1 and 2 at the interface. Possibly reflect off the flyer (plate 1) again, especially if it is still moving or elastic, which possibly adds to existing stress fields and causes secondary failure or delayed cracking at interface 1. We now compute whether the shock reflected from the back of the final target (plate 3) returns before solidification, causing dynamic cracking at the second interface.

$$t_3^* = \frac{d_{23}}{U_{p,12}} + \frac{2h_3}{U_{s3} \cos \beta_2}, \text{ here } V_{i,2} \approx U_{p,12} \quad (28)$$

$$V_{c,23} = \frac{U_{p,12}}{2 \sin\left(\frac{\beta_2}{2}\right)} \quad (29)$$

Trailing edge of hot zone at that time,

$$x_3(t) = V_{c,23} t_3^* - b_2 = V_{c,23} \left(\frac{2h_3}{U_{s3} \cos \beta_2} + \frac{d_{23}(\rho_2 S_2 - \rho_3 S_3)}{\rho_3 C_{03} - \rho_2 C_{02}} \right) - b_2 \quad (30)$$

Similarly, for target shock reflection (critical melt crack distance) $x_3^* = 2h_2 \tan \beta_2$

$$V_{c,23} \left(\frac{2h_3}{U_{s3} \cos \beta_2} + \frac{d_{23}(\rho_2 S_2 - \rho_3 S_3)}{\rho_3 C_{03} - \rho_2 C_{02}} \right) - b_2 \geq x_3^* = 2h_2 \tan \beta_2 \quad (31)$$

As interface 2 involves a delayed collision between the deformable intermediate plate (Plate 2) and a rigidly backed target (Plate 3), $b_{2,max}$ can not be estimated using the same simplified geometric form as used for $b_{1,max}$,

$$b_{2,max} = V_{c,23} \cdot t_3^* - 2h_2 \tan \beta_2 \quad (32)$$

The maximum allowable impact velocity $V_{i,2}$ at interface 2 based on thermal energy input and melting threshold - analogous to what was done for interface 1.

$$V_{i,2,max} = 2.239 \frac{\sqrt{k_2} \sqrt{k_3} \cdot \sqrt{\min(T_{m1}, T_{m2})} \cdot (\alpha_2 + \alpha_3)^{1/4}}{V_{c,23}^{1/4} \cdot \sqrt{b_2} \cdot h_2^{1/4} \cdot \sqrt{\rho_2} \cdot \sqrt{k_2 + k_3} \cdot \sqrt{f} \cdot (\alpha_2 \alpha_3)^{1/4}} \left(\frac{S_2 C_{03} - S_3 C_{02}}{(C_{02} \rho_2 S_3 \tan \beta_2 - C_{03} S_2 \rho_2 \tan \beta_2 + S_2 V_{c,23} \rho_2 - S_3 V_{c,23} \rho_3) \cos \beta_2} \right)^{1/4} \quad (33)$$

When plate 2 impacts plate 3, a shock reflects back into plate 2, and may return to interface 1.

2.2.3 Shock return time (plate 2 → interface 1)

$$t_{return} = \left(\frac{2h_2}{U_{s2} \cos \beta_1} \right) \quad (34)$$

Where $U_{s2} = C_{02} + S_2 U_{P,12}$. Now hot zone position and crack formation condition at interface 1 can be said as,

$$x(t_{return}) = V_{c,12} \cdot t_{return} - b_3 \quad (35)$$

$$V_{c,12} \left(\frac{2h_2}{U_{s2} \cos \beta_1} \right) - b_3 \geq 2h_1 \tan \beta_1 \quad (36)$$

2.3 Smoothed particle hydrodynamics

SPH simulations were conducted using ANSYS Autodyn with a particle spacing of 5 μm to accurately resolve localized deformation and wave propagation in the multi-plate VFAW configuration. All plates were modeled using the Johnson-Cook plasticity model combined with a Grüneisen equation of state (EOS) to capture high-strain-rate plastic deformation and shock-induced thermal effects. Their constant values were taken from the literature. For validation, the flyer (Plate 1) was assigned an initial velocity of ~ 1400 m/s, estimated from the input energy (4 kJ) and foil mass ($\sim 8.1 \times 10^{-5}$ kg). Boundary conditions included a fully clamped Plate 3 and a symmetric upper flyer. Thermo-mechanical coupling enabled the extraction of interfacial temperature fields and equivalent plastic strain and reflected shock interactions critical to weld evaluation and a qualitative validation of the analytical model.

3 Results and Discussion

The SPH simulations visually captured material flow and jetting behavior under oblique collision conditions. The initial impact at interface 1 produced localized plastic flow and jetting, as shown in Fig. 3 and as discussed through possible hypothesis in section 2.2. At interface 2 in Fig. 4, the SPH results effectively predicted critical shock return times and jetting criteria based on back-reflected stress waves, and the values are comparable with the analytical model, as also discussed through the possible analytical model. It was also interesting to observe that, contrary to available literature, minimal jetting was observed at interface 2.

The SPH results shown in Fig. 3 and Fig. 4 visualize the sequential shock dynamics within the three-plate assembly. The initial flyer impact creates a compressive wave at Interface 1, which then propagates through the intermediate plate (Plate 2) and drives a secondary collision with Plate 3. The resulting reflected wave from the rigidly backed target returns through Plate 2 and ultimately re-enters Plate 1. This behavior, captured in the time-resolved fringe patterns, confirms the mechanism hypothesized in our analytical model: that back-reflected shock waves may arrive at Interface 1 post-impact, potentially interfering with solidification and promoting late-stage cracking. The clear arrival timing agreement between SPH and predicted shock trajectories (Eq. 32–36) supports the validity of the proposed failure model. Notably, peak interfacial temperatures at Interface 1 exceeded 1200 K, and equivalent plastic strain localized near the jetting front surpassed 1.0, indicating severe plastic flow.

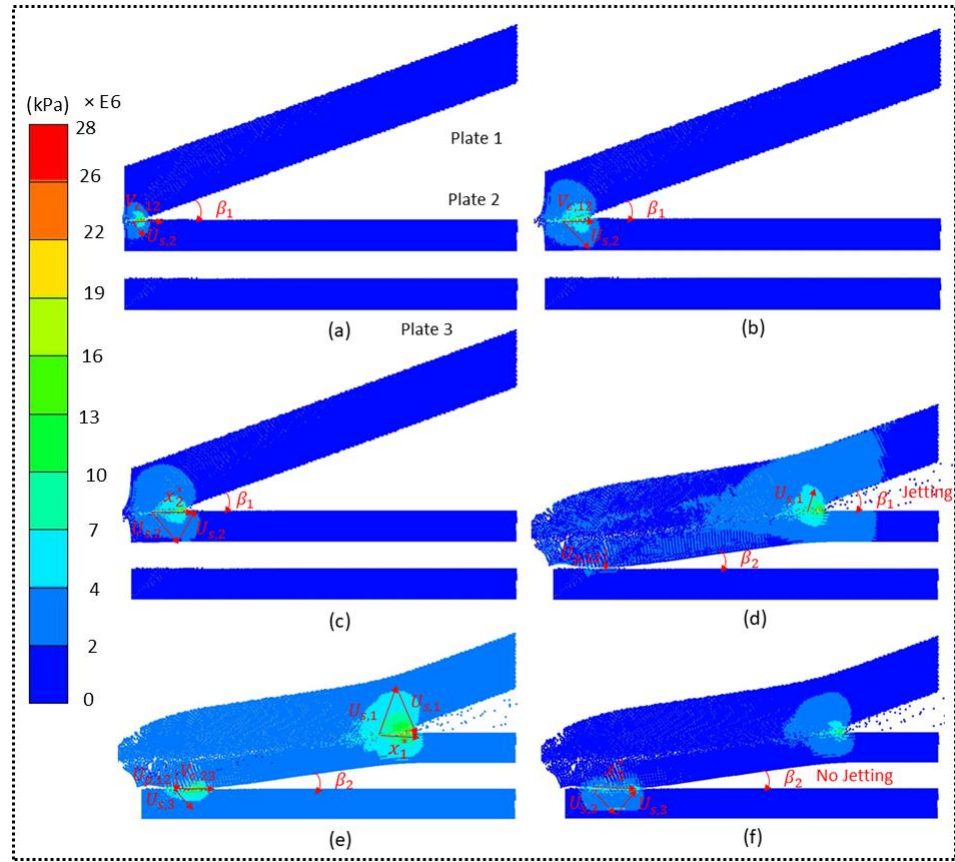


Figure 3: SPH modeling fringe pattern of shock wave showing oblique impact and material flow behavior during three-stack VFAW welding: (a–c) initial impact at interface 1 and (d–f) progressive impact at interface 2 (Eq. 1–14)

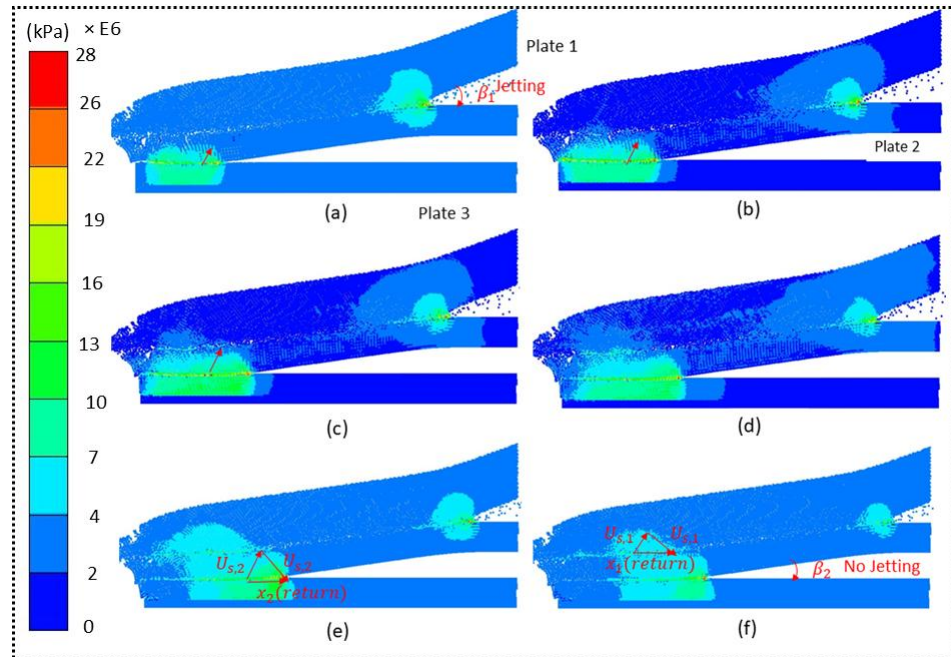


Figure 4: SPH modeling fringe pattern of shock wave reflections and return paths during three-stack VFAW welding: (a–e) shock generated at interface 2 propagates through plate 2 and returns to interface 1 and reflects back to interface 2 (as per Eq. 26–28); (d–f) extended shock wave traverses plate 1, partially reflects, and returns to interface 1 (Eq. 29–31)

4 Conclusion

This work presents an integrated analytical and simulation-based study of 3-stack VFAW joints, emphasizing partial pre-forming and material flow dynamics. The proposed analytical framework captures critical physical mechanisms, including shock propagation and back-reflected wave effect, that govern weld quality and failure. SPH simulations complement the model by providing visual confirmation of interface behaviour of weldability as well as quantitative values of shock arrival time. Together, these approaches establish a predictive methodology for understanding and optimizing high-speed, multi-plate welding systems. In future work, we will enhance the analytical framework with machine learning-based surrogate modeling, train on SPH and experimental data, and elaborate on the multiphysics nature of the process. The hybrid analytical framework will be extended to experimentally calibrate process parameters and validate results through microstructural and mechanical testing of actual welds.

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Development of Fill Impact Welding technique for metal additive manufacturing

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Abstract

Metal additive manufacturing typically utilizes powder as feed. Fill impact welding, uses dime-sized pieces of sheet metal launched at high speeds to impact and weld to the metallic substrate. The driving force can be generated by several means, including vaporizing foil actuators used in this work. A custom impact welding fixture was built to launch 0.5 mm thick and 6.35 mm diameter flat circular metallic disks (flyers) to speeds greater than 500 m/s over a travel distance of 5 millimeters. The flyers impacted target plates placed at fixed relative angles. Fully welded interfaces were observed via cross sectional metallography as well as non-destructive evaluation methods of ultrasonic testing. It was demonstrated that titanium flyers can be reproducibly welded to copper for a certain set of parameters. Furthermore, the impact velocity was found to be controllable and repeatable for various input energies from the capacitor bank power system. For example, for 2.0kJ input energy shots, the average peak velocities of the flyer were measured to be 800 m/s with a variation of less than 3%.

Keywords

Additive, welding, repair, NDE

Introduction

Additive manufacturing (AM) has rapidly expanded across industries such as automotive, aerospace, defense, medical, and consumer goods (Gebhardt, 2011; Guo and Leu, 2013). Fusion-based AM consolidates material through melting via laser, electron beam, or electric arc in inert atmospheres or shielding gas; however, it is often limited by vaporization loss of alloying elements, solidification defects (e.g., porosity, lack of fusion), elemental segregation, and hot cracking (Sames et al., 2016). Thermal cycling can further introduce microstructural anisotropy, residual stress, and distortion (Nycz et al., 2021; Liu et al., 2021).

Solid-state AM technologies reduce or eliminate many fusion-associated defects by minimizing heat input and relying on plastic deformation (Khodabakhshi and Gerlich, 2018). These techniques fall into two main categories: sintering-based (e.g., binder jetting) and deformation-based (e.g., ultrasonic, friction stir, and cold spray) (Tuncer and Bose, 2020). In deformation-based AM, intimate interatomic contact and bonding are achieved through severe plastic deformation that removes surface oxides. Processes like additive friction stir and cold spray are used in large-scale fabrication, cladding, and repair (Khodabakhshi and Gerlich, 2018; Griffiths et al., 2019). MELD, a commercial variant of additive friction stir deposition, enables bulk solid-state metal printing with tailored alloys and composites (Yu et al., 2018), though it requires a large gantry system and faces limitations in overhang angles, tool wear, and resolution.

Cold spray AM uses high-velocity impact to deposit powdered feedstock, removing oxide layers and forming local metallurgical bonds through plastic deformation at particle interfaces (Assadi et al., 2016). This method enables processing of traditionally “unweldable” metals such as 2XXX and 7XXX series aluminum (Mahoney et al., 1998; Brewer et al., 2018; Olea et al., 2013), preserves fine-scale features in metal matrix composites and oxide dispersion-strengthened alloys (Behera et al., 2019; Story et al., 2018), and enables graded compositions (Khodabakhshi and Gerlich, 2018). It also mitigates intermetallic formation at interfaces (Kuo et al., 2019) and operates effectively in ambient air, unlike fusion and sinter-based techniques.

Vaporizing Foil Actuator Welding (VFAW), an impact welding technique similar to explosive welding, uses rapid foil vaporization via capacitive discharge to generate high-pressure plasma that accelerates a flyer sheet into a target at controlled speeds (typically 300–700 m/s) and shallow angles (5°–20°), enabling oxide removal and bonding (Vivek et al., 2013). VFAW has been used to join a wide variety of dissimilar weld pairs, including Cu–Fe, Al–Fe, Cu–Ti, Ni–Ti, and Ni-based superalloys (Prasad et al., 2020; Liu et al., 2019; Vivek et al., 2014; Gupta et al., 2019; Nirudhoddi et al., 2021). It has also been explored for dynamic impact testing and powder compaction (Prasad et al., 2020).

Building on these capabilities, this work introduces Fill Impact Welding (FIW) as a new solid-state AM process. FIW involves the incremental addition of material to “fill” a three-dimensional space and create a solid object. FIW uses vaporized foil actuators to accelerate sheet or foil feedstocks, eliminating the need for large clamping loads. It operates using only electricity, metal sheets, and sacrificial foils. Compared to powder-based feedstocks, FIW’s sheet inputs are lower in cost and more readily available. Its deposition mechanism preserves more of the original feedstock microstructure than cold spray due to the larger deposited volumes, while

avoiding the extreme cold work characteristic of cold spray, resulting in as-consolidated materials with improved ductility. Though cold spray ductility can be enhanced via heat treatment (Rokni et al., 2018; Story and Brewer, 2018), FIW offers a pathway toward structurally sound builds without additional processing.

This paper presents the fundamental concepts of FIW, system characterization, and a proof-of-concept demonstration using inclined impact filler welding. Non-destructive evaluation (NDE) via ultrasonic imaging was employed to assess weld quality.

Methods

A FIW system was developed, consisting of three parts: a target holding platform, a FIW gun, and FIW cartridges. The target holding platform is a stage isolated from filler launching. It enables accurate control of the impact angle of the target plate. FIW cartridge was developed to launch different sizes of the filler disks as flyers; in this work, filler disk with a diameter of 6.35 mm was equipped in the FIW cartridge. A schematic is shown in **Fig. 1a and 1b**. The FIW gun is connected to a capacitor bank that discharges a high voltage through the FIW cartridge to activate the launch process. With no direct contact between the target holding platform and the FIW gun, those two can be moved independently, making the system suitable for applications involving large and complex components.

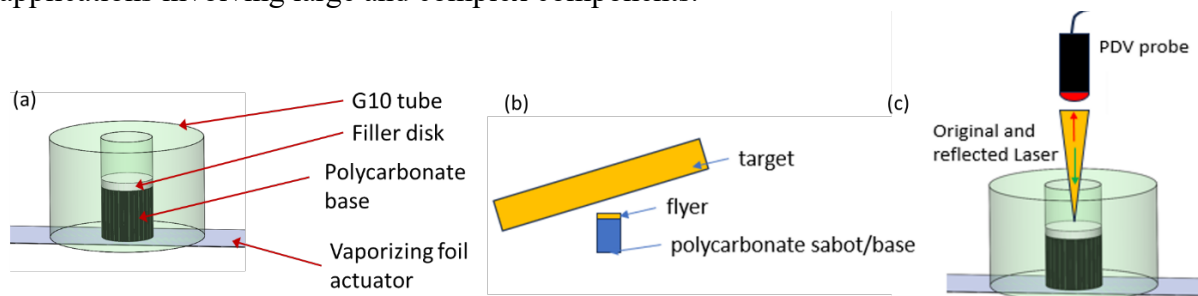


Figure 1: FIW cartridge consisting of actuator and filler flyer (a), schematic of inclined impact between the cartridge and target (b), and velocity measurement of the FIW cartridge and flyer using photonic Doppler velocimetry (c).

By changing the input energy of the capacitor, the impact speed of the filler disk is controlled. System characterization was performed with the speed of filler disk measured with photonic Doppler velocimetry (PDV) and discharging current recording with a Rogowski coil. The PDV experiment set up is shown in Fig. 1c. To test the controllability of the impact speed, filler disks of 0.5 mm thick aluminum sheet of diameter of 6.35 mm were launched at the input energies of 1.0 kJ, 1.5 kJ, and 2.0 kJ. Three samples were launched at 2.0 kJ to evaluate the repeatability of the system. The conditions for welding are mentioned in **Table 1**.

Filler material	Target material	Impact angle	Input energy	Characterization method
0.5 mm Aluminum alloy	N/A	N/A	1.0, 1.5, 2.0 kJ	PDV measurement
0.5 mm CP-Ti	4.76 mm Cu 110	7°, 14°, 21°	2.0 kJ	Ultrasonic imaging, cross-sectional optical microscopy

Table 1: Various materials welded using FIW system.

Welding experiments were conducted using CP-titanium (Grade 2) filler and Cu 110 target with various impact angles and input energy of 2.0 kJ. Ultrasonic imaging as non-destructive testing (NDT) and optical microscopy as destructive were employed testing methods to examine the welding interface and assess the welded area.

Results and discussion

Results of all the PDV tests are shown in **Fig. 2a**, where the 1.0 kJ, 1.5 kJ, and 2.0 kJ input energies generated peak speeds of the disk fillers around 550 m/s, 680 m/s, and 820 m/s, respectively. The velocity-displacement curves estimated by time integration of the velocity-time curve is shown in Fig. 2b. The variation range of the peak velocities of the three repeating experiments at 2.0 kJ was 22.6 m/s, which is less than 3% of the average peak velocity, 821.29 m/s, indicating good repeatability of the FIW system. The recorded current traces are shown in Fig. 2c, which also proves the consistency of the system during the discharging process.

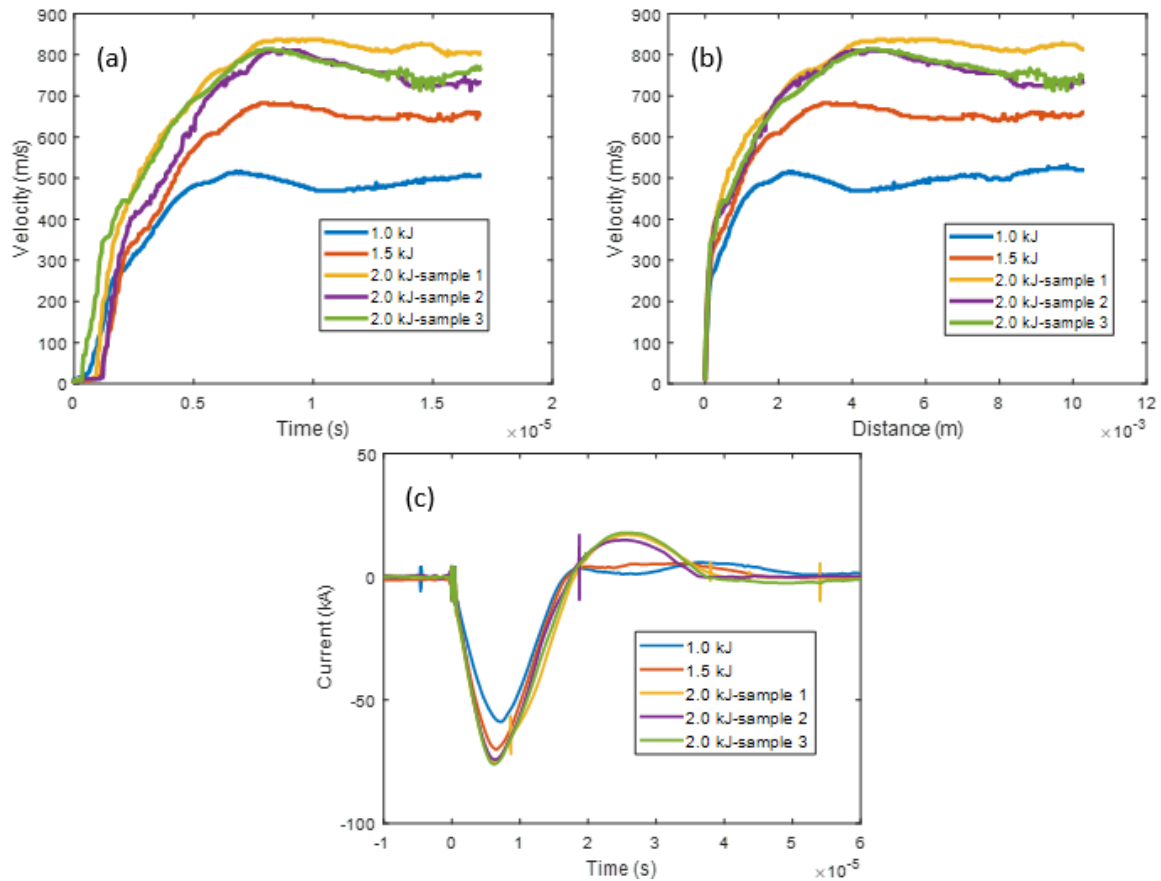


Figure 2: Diagnostics during FIW: (a) velocity-time, (b) velocity-displacement and (c) current-time profiles of the filler disks launched by FIW system.

Dissimilar welds of CP Ti-Cu110 were successfully created at three different impact angles: 7°, 14°, and 21°, shown in **Fig. 3**. Ultrasonic imaging of the CP Ti-Cu110 samples (**Fig. 4**) revealed the following: (1) At the edges of all samples, there were indentations on the target made by the flyer's impact. For the 7° and 14° samples, these regions were located on one side of the weld at the initial impact area, aligning with previously identified impact welding mechanisms. For the 21° sample, unwelded zones appeared around the weld. These unwelded areas can be machined off between welding operations. (2) In the central regions away from the edges, large, 100% welded areas without gaps were achieved for at least one of the impact angles (14°). (3) A wavy interface morphology was observed in all three samples, which are consistent with what has been found for impact welding of Cu-Ti (Vivek et al., 2013).

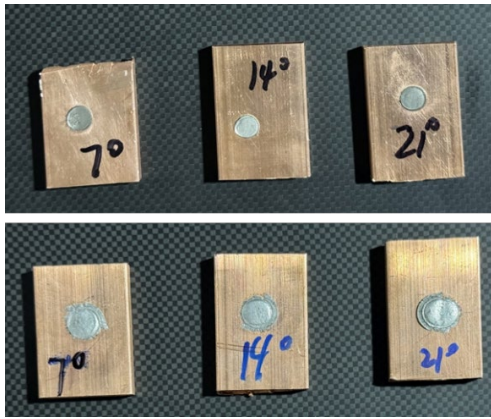


Figure 3: Impact welded CP Ti-Cu110 (upper) and Al3003-Cu110 (lower) at impact angles of 7°, 14°, and 21°.

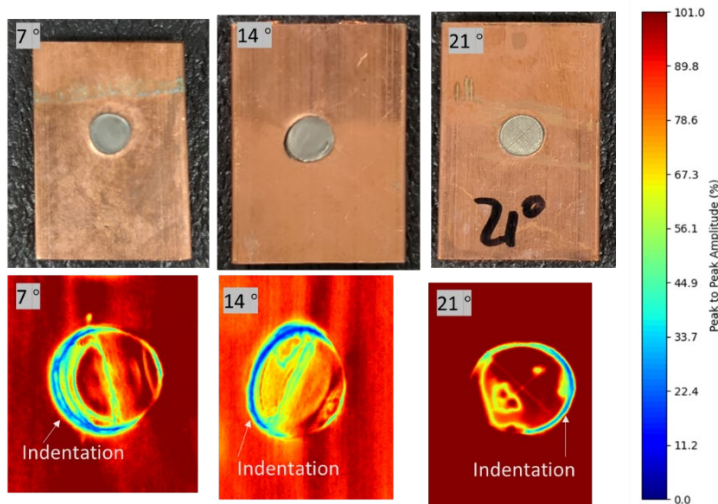


Figure 4: Impact welded CP Ti-Cu110 at impact angles of 7°, 14°, and 21° and ultrasonic imaging results. Red signifies a solid or fully bonded interface, whereas blue represents a gap or unbonded area.

The interface imaged from the diameter of the weld nugget shows near-complete welding in cross-sectional metallography as shown in **Fig. 5**. It revealed a continuous, gap-free weld interface along the entire welding length. The prominent and uniform interfacial waves across the interface indicate a consistently maintained impact angle during the welding process, which is critical for processing parameter control. This is a breakthrough for VFAW, as most previous studies showed inevitable center unwelded zones caused by insufficient impact angles due to welding configurations (Li et al., 2020). The main fully-welded, void-free area, along with the unbonded initial impact region revealed by optical microscopy, aligns with observations from ultrasonic testing, thereby validating the NDE approach.

Experimental measurements of aluminum flyers launched by the FIW system confirmed that the vaporizing foil actuator method can launch flat flyers at speeds adjustable by varying the input electrical energy. The velocity-time curves for the same energy levels nearly overlapped with each other and the flyer left the gun barrel at very similar speeds (within 5%) as shown in Fig. 2. The structure of the weld interface was characterized through ultrasonic testing and cross-sectional metallography (Fig. 4 and 5). All the methods revealed fully or nearly fully welded interfaces when an initial impact angle was set between the target and the flyer. Similar idea was previously demonstrated (Prasad et al., 2020), but this systematic FIW study enables significantly greater control and repeatability, offering scalability for various applications. Testing of the Cu-Ti weld samples revealed a higher amount of waviness. The ability to test weld structure non-destructively was shown for the first time for impact welds made at this size scale.



Figure 5: Cross sectional images of the Cu-Ti weld interface made with a 14° impact angle.

Therefore, it is proved that 1) the process, structure and performance of impact welding are reproducible and 2) that impact welds can be made with minimal to no voids. These two breakthroughs are key to transitioning the technology to a commercial process while opening weld configuration avenues that have not previously been achieved with impact welding.

There are certain aspects of the FIW process and system that need to be improved upon. While void-free welds have been successfully created by inclining the target and flyer, adding the next layer remains a challenge. The previous layer needs to be machined to provide an impact angle within the welding window of that material being welded. More flexible and

precise relative positioning methods such as use of robotic arms should be employed for better aiming of the filler disks.

Conclusions

1) For the first time, VFA was used to launch horizontal flyers in a way that allows for independent control over the positioning of the launch mechanism and the target holder.

2) The process is stable and yields a consistent impact velocity, which is needed for a reproducible welded structure and performance.

3) By controlling the process conditions, this impact welding method can be used as an additive technique where nearly all of the consumable metal (flyer) is added to the target plate.

4) Non-destructive evaluation methods such as ultrasonic testing can be utilized for assessing the impact welded region.

Acknowledgements

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Enhancement of Laser Impact Welding for Joining Thicker Aluminium Sheet

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Abstract

This study critically examines the efficacy of Augmented Laser Impact Welding (ALIW) in achieving high-quality welds by leveraging the effects of laser-induced plasma and controlled impact forces. Therefore, various Gunpowder Nitromethane (GPN) compositions were examined. The results demonstrated that the incorporation of 0.1 g of micro-air bubbles into a mixture of 1 g smokeless powder and 10 g of 95% nitromethane significantly enhances the flyer velocity. Moreover, the usage of Sapphire instead of borosilicate glass as a confinement layer can be a great alternative due to its damage resistance after each laser shot. Furthermore, the refined GPN composition was employed in experiments with a 1 mm Al 6061-T6 flyer, where different spot sizes and varying GPN thicknesses were tested to determine the optimal conditions for this specific flyer thickness. The findings revealed that the maximum flyer velocity was achieved when using 0.54 mm of GPN-3 in conjunction with a 5.22 mm laser spot size.

Keywords: Augmented Laser Impact Welding (ALIW), Photonic Doppler Velocimetry (PDV), Impact Welding, and Detonation Energy

1. Introduction

Laser impact welding (LIW) has emerged as a promising technique, distinguished by its superior mechanical strength, precision, and operational efficiency (Thurston et al., 2023). This process leverages high-energy laser pulses to induce shock waves at material interfaces, facilitating bonding through plastic deformation (Daehn & Lippold, 2011). This method is particularly advantageous for welding varying thicknesses of dissimilar materials, a domain where conventional welding techniques often struggle due to the formation of brittle intermetallic compounds (Sen & Puri, 2022).

In LIW technique, the flyer material gains momentum due to the optical vaporization of a thin surface layer. This vaporization generates plasma, which in turn exerts pressure on the flyer. To

enhance propulsion efficiency, a transparent confinement layer is incorporated, which confines the plasma and optimizes the energy transfer (Swift et al., 2005). Typically, LIW utilizes laser pulses with energy levels of a few Joules, directed onto surface areas measuring a few millimeters in diameter (H. Wang et al., 2015). However, the process faces inherent limitations in laser energy transmission. In particular, when optical energy densities surpass approximately 10 GW/cm², air ionization occurs, impeding further energy delivery. This phenomenon poses a significant challenge to the efficiency and scalability of LIW, thereby restricting its wider industrial application (Thurston et al., 2023).

Therefore, many studies were conducted to enhance the energy released by utilizing various ablative layers. For example, Picatinny Liquid Explosive (PLX), a mixture of 95% nitromethane combined with 5% ethylene diamine, has been widely utilized in impact welding due to its high detonation velocity of 6200 m/s and explosive yield exceeding TNT by 134% (Yoshida et al., 1985). However, regardless of these advantages, PLX presents significant safety concerns, including high sensitivity to impact, as demonstrated by its detonation risk in drop tests. Additionally, its short shelf life—spanning merely seven days—in a sealed glass bottle—further restricts its practical application. Given these limitations, there is a pressing need to develop a safer, more stable explosive suitable for industrial and laboratory use without compromising safety and reliability. Consequently, Thurston et al. (2023) introduced Gunpowder Nitromethane (GPN) as a novel chemical explosive, formulated by blending 10 g of 95% nitromethane with 1 g smokeless pistol powder, which contains various energetic compounds such as nitroglycerin and nitrocellulose. Given its high nitromethane content, GPN is expected to exhibit characteristics similar to Picatinny Liquid Explosive (PLX). While its precise oxygen balance could not be determined due to the unfamiliar composition of the smokeless powder, GPN demonstrates notable advantages over PLX, including a significantly extended shelf life of up to 1.5 years and improved impact resistance (Thurston et al., 2023). These attributes suggest its potential as a more stable alternative; however, further research is necessary to optimize its composition and enhance plasma energy output. Therefore, investigations should focus on refining the formulation through controlled variations in concentration and compounds, ensuring improved sensitivity and greater detonation efficiency.

On the other side, the confinement layer is a critical component in laser impact welding, as it directs and concentrates detonation energy onto the materials being joined while also serving as a medium for laser energy transmission. Several materials, involving glass, water, polycarbonate, and quartz, have been employed as confinement layers, each offering distinct advantages and limitations (Berthe et al., 1997). Glass, for instance, is highly effective due to its optical transparency and minimal laser beam distortion, but it fractures upon repeated laser exposure (H. Wang et al., 2015). Polycarbonate, though resistant to fracture, suffers from optical damage, limiting its reuse. Water, commonly utilized in laser shock peening, has shown promise in LIW applications, while quartz has demonstrated superior pressure generation at equivalent laser power densities (Q. Wang et al., 2023). Despite its absence in previous LIW studies, sapphire presents a compelling alternative due to its high optical transmittance, exceptional mechanical durability, and resistance to deformation under extreme thermal and pressure conditions. However, its feasibility as a confinement layer requires further evaluation,

particularly concerning surface reflection and energy absorption characteristics, which could impact welding efficiency.

Finally, this study seeks to advance the foundational work of Thurston et al. (2023), who investigated the use of chemical augmentation to enhance the LIW process. Building on their research, this investigation critically examines the effects of diverse augmentation chemistries and introduces the use of three distinct confinement layers to systematically evaluate their impact on system performance. In addition, an evaluation was performed on a case study involving 1 mm Al 6061-T6 to demonstrate the effect of the improved GPN on thicker flyers.

2. Materials and Methods

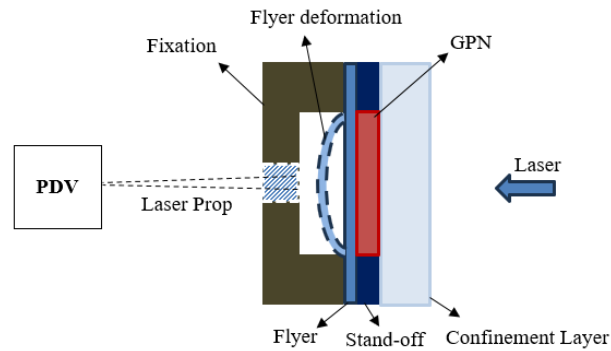
In this study, an Nd:YAG laser system (Powerlite™ Precision II Scientific System from Continuum Lasers, Milpitas, CA, USA) was utilized, operating at an energy of 3 J, a wavelength of 1064 nm, a pulse width of 8 ns, and a spot diameter of 5.22 mm. To comprehensively analyze this study, two aluminum alloys, Al 2024-T3 and Al 6061-T6, both with a thickness of 0.5 mm, were selected as flyer materials. The samples were precisely cut into 25 × 25 mm dimensions.

A critical aspect of this research involved investigating the role of gunpowder nitromethane (GPN) as an ablative layer in augmented laser impact welding. This remains an underexplored area, as the optimal compositions for maximizing detonation efficiency are not yet well-defined. Consequently, an empirical approach based on trial and error was adopted, involving modifications to the original GPN composition which consisted of 1 g of smokeless gunpowder mixed with 10 g of 95% nitromethane. The study explored the effects of introducing aluminum powder and micro-air bubbles into the GPN mixture to assess their influence on detonation energy and, subsequently, on the flyer velocity. A detailed breakdown of the different GPN compositions is presented in the corresponding **Table 1**. To ensure the precise evaluation of GPN performance, a controlled layer of 0.37 mm GPN was applied to the sample surface, as shown in **Figure 1**. This was secured using multiple layers of clear tape to maintain a consistent stand-off gap. To quantify detonation energy and measure the flyer velocity, Photonic Doppler Velocimetry (PDV) was employed, with each scenario being tested three times to obtain an averaged dataset on the flyer deformation.

On the other side, an investigation was conducted into the effect of different confinement layers on flyer velocity, aiming to identify a viable alternative to borosilicate glass, which exhibited a tendency to fracture following each laser pulse. Three confinement materials were tested: **(i)** borosilicate glass with a thickness of 9 mm, **(ii)** a sapphire cube with a thickness of 1 inch, and **(iii)** a thin layer of running water approximately 1 mm thick. The primary factor influencing the selection of these materials was their optical transmission properties, which directly affect the proportion of laser energy reaching the flyer. Hoon et al. (2009) reported that for an 8 mm thick borosilicate glass at a wavelength of 1064 nm, the transmission rate was 87%. In contrast, previous studies indicated that sapphire exhibited a slightly lower transmission rate of 83% under the same conditions (Esco Optics Inc.). These findings suggest that, at a wavelength of 1064 nm, borosilicate glass and sapphire share comparable optical transmission characteristics, making both viable candidates for confinement in augmented laser impact welding.

Table 1 Various compositions of ablative layers

Ablative Layer	Compositions
GPN-1	1 g Gunpowder + 10 g Nitromethane (Original GPN)
GPN-2	2 g Gunpowder + 10 g Nitro.
GPN-3	1 g Gunpowder + 10 g Nitro. + 0.1 g Micro-air Bubbles
GPN-4	1 g Gunpowder + 10 g Nitro. + 0.2 g Micro-air Bubbles
GPN-5	1 g Gunpowder + 10 g Nitro. + 0.05 g Micro-air Bubbles + 0.05 g Aluminum Powder
GPN-6	1 g Gunpowder + 10 g Nitro. + 0.1 g Micro-air Bubbles + 0.05 g Aluminum Powder
GPN-7	1 g Gunpowder + 10 g Nitro. + 0.1 g Micro-air Bubbles + 0.1 g Aluminum Powder

**Figure 1** Schematic of the Augmented Laser Impact Process Setup to measure flyer velocity using PDV

3. Results and Discussion

3.1. The Influence of adding Aluminium Powder and Micro-Air Bubbles:

According to PDV measurements, the detonation energy for GPN-2 was lower than GPN-1, as discussed clearly during previous work (Abdelmaola et al., 2025). Hence, this GPN composition will be excluded from further investigation.

Figure 2-Figure 3 present the highest flyer velocities recorded across all experimental conditions. The interaction of various additives—including gunpowder, aluminum powder, micro-air bubbles, and nitromethane—produced a complex effect on detonation energy and, consequently, flyer velocity. Notably, the incorporation of micro-air bubbles (GPN-3 scenario) significantly influenced energy enhancement through multiple mechanisms (Yamada et al., 2022). Firstly, micro-air bubbles promote better mixing between the explosive materials and surrounding gases or reactants, facilitating a more uniform and complete reaction that improves energy output. Secondly, they play a role in reducing shock wave attenuation by acting as cavitation nuclei, thereby enhancing shock wave propagation and amplifying explosive energy. Additionally, under certain conditions, micro-air bubbles can trigger cavitation effects, where the rapid formation and collapse of vapor cavities generate localized high pressures and

temperatures, further supporting explosive reaction initiation (Shan et al., 2023). Finally, when a shock wave interacts with a micro-air bubble, it can focus and amplify the surrounding energy, effectively lowering the energy threshold required for detonation and improving overall sensitivity (Ishii & Watanabe, 2019). However, excessive micro-air bubble inclusion (GPN-4 scenario) may reduce available detonation energy, as the bubbles occupy a portion of the explosive's volume, leading to diminished performance.

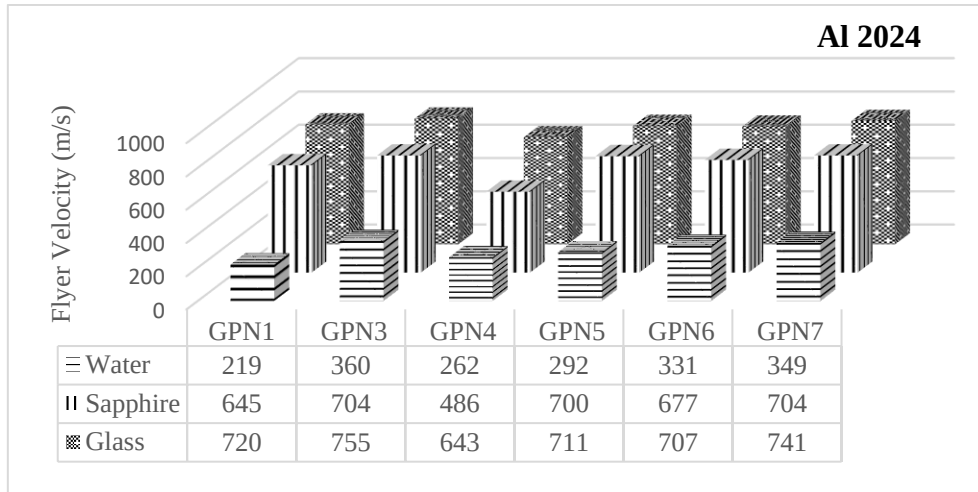


Figure 2 The highest flyer velocity with respect to various GPN compositions, using different confinement layers; water, sapphire, and borosilicate glass, for 0.5 mm thick Al 2024-T3

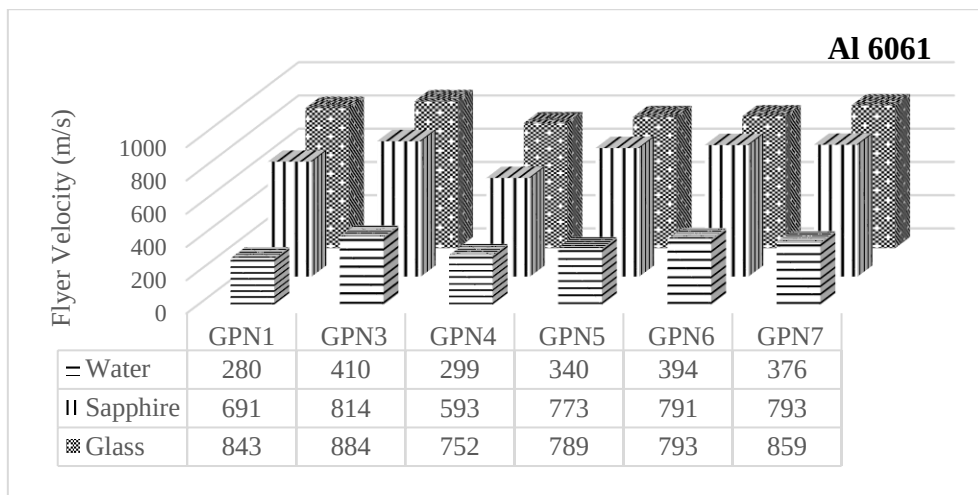


Figure 3 The highest flyer velocity with respect to various GPN compositions, using different confinement layers; water, sapphire, and borosilicate glass, for 0.5 mm thick Al 6061-T6

On the other hand, for scenarios of GPN-5,6,7 the incorporation of aluminum powder, functioning as a metal particle explosive, contributed to an increase in detonation heat due to its exothermic oxidation process. However, this modification resulted in a reduction in detonation energy and, consequently, a decrease in detonation velocity. This effect is attributed to the damping influence of Al powder, which absorbs a portion of the energy during oxidation, as previously discussed (Dong et al., 2021; Sielicki et al., 2022). Zhang et al. (2022) provided

a comprehensive analysis of the impact of Al powder on both detonation heat and velocity. Their findings indicate that with increasing content of Al powder in RDX-based explosives, there is a notable decline in detonation velocity, whereas the detonation heat exhibits a rising trend. Thus, the results presented in **Figure 2-Figure 3** confirm that the highest flyer velocities, across all experimental conditions, were accomplished with the GPN-3 composition, which contained 0.1 g of micro-air bubbles into the original GPN-1 formulation.

3.2. The Influence of Various Confinement Layers:

Figure 2-Figure 3 depict flyer velocity variations across different confinement layers, particularly borosilicate glass, sapphire, and water. Across all GPN compositions, borosilicate glass exhibits the maximum flyer velocities, with only a marginal difference compared to sapphire, followed by water. However, despite its superior performance, borosilicate glass is not a viable option for industrial applications due to its high susceptibility to fracturing and shattering, making it economically impractical. Conversely, sapphire displayed markedly superior hardness and durability compared to borosilicate glass, throughout the welding process (Chen et al., 2022). However, its high cost and susceptibility to cracking present significant challenges for implementation in industrial applications. As discussed before, both borosilicate glass and sapphire exhibit similar transmission characteristics at a 1064 nm wavelength. Considering the flyer velocity results across all cases and the minimal deviation in performance compared to borosilicate glass, sapphire emerged as a highly effective confinement layer. It achieves sufficient-flyer velocities while eliminating the need for frequent replacement due to damage, making it a more practical and sustainable choice for industrial applications.

3.3. Optimization of ALIW parameters for 1 mm Al 6061-T6 as a flyer

The performance of a flyer in an augmented laser impact process is significantly influenced by the thickness of both the flyer material and the augmented layer. The interaction between the laser energy, explosive layer, and flyer material determines the maximum velocity achieved during impact. Therefore, to investigate the impact of GPN-3 on thicker flyers, experiments were conducted using a 1 mm thick Al 6061-T6 flyer. The study examined various laser beam diameters (4 mm, 5.22 mm, and 6 mm) and different GPN-3 thicknesses ranging from 0.27 mm to 0.63 mm, to identify their effects on flyer velocity. The measured velocities (average of 3 samples for each case, with scattering +/- 2%), obtained using Photonic Doppler Velocimetry (PDV), are presented in **Table 2**. Additionally, **Figure 4** illustrates the resulting flyer deformation for each experimental condition, revealing that the most pronounced dimple deformation was observed under specific parameters.

Table 2 Max. Flyer Velocity of 1 mm Al 6061 at various parameters

Spot Diameter (mm)	GPN-3 Thickness (mm)				
	0.27	0.36	0.45	0.54	0.63
	Max. flyer Velocity (m/s)				
4	333	403	415	442	429
5.22	369	438	449	451	433
6	328	358	324	276	246

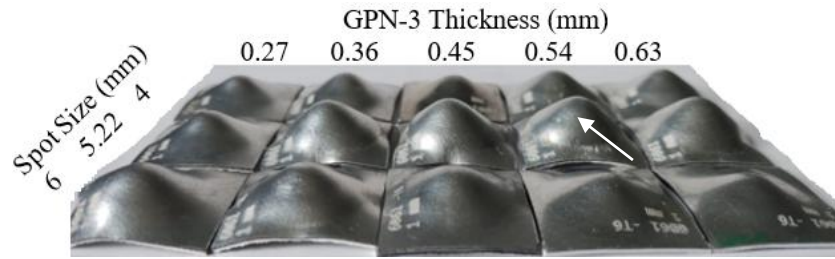


Figure 4 Flyer deformation of 1 mm Al 6061 at various GPN thicknesses and spot sizes showing the highest deformed dimple at 5.22 mm spot size and 0.54 mm GPN thickness

Experimental findings, shown in **Figure 5**, indicate that when employing a 1 mm thick Al 6061 flyer, the maximum velocity is reached at a 0.54 mm GPN-3 thickness with a 5.22 mm spot diameter, compared to previous studies (Abdelmaola et al., 2025) that found a 0.37 mm GPN thickness to be optimal when utilizing thinner flyer of a 0.5 mm thick at same spot size. This variation in optimal GPN thickness can be attributed to the interplay between various characteristics; Inertia and Momentum Considerations, Energy Absorption and Shockwave Propagation, and Stress Wave Attenuation and Dissipation. Firstly, the flyer velocity is a result of the momentum imparted by the explosive-driven shockwave. Given that momentum depends on both mass and velocity, a thicker flyer has a greater mass, requiring a proportionally higher impulse to achieve the same velocity. Since a thicker flyer has more mass, it resists acceleration more than a thinner one, necessitating a larger explosive layer to generate sufficient impulse.

Secondly, the explosive layer generates a shockwave that propagates through the flyer. Thicker flyers require a higher energy input to fully transmit this shockwave through the material and maintain an optimal pressure gradient for acceleration. A smaller explosive thickness may not provide enough sustained force for the necessary impulse transfer, leading to suboptimal flyer acceleration. Moreover, as the shockwave moves through the flyer, some of its energy is dissipated due to wave reflection and attenuation. In thicker flyers, a greater proportion of the energy is absorbed before reaching the opposite surface, reducing the efficiency of momentum transfer. Increasing the explosive thickness compensates for this energy loss, ensuring that sufficient pressure is maintained for accelerating the flyer to its maximum velocity.

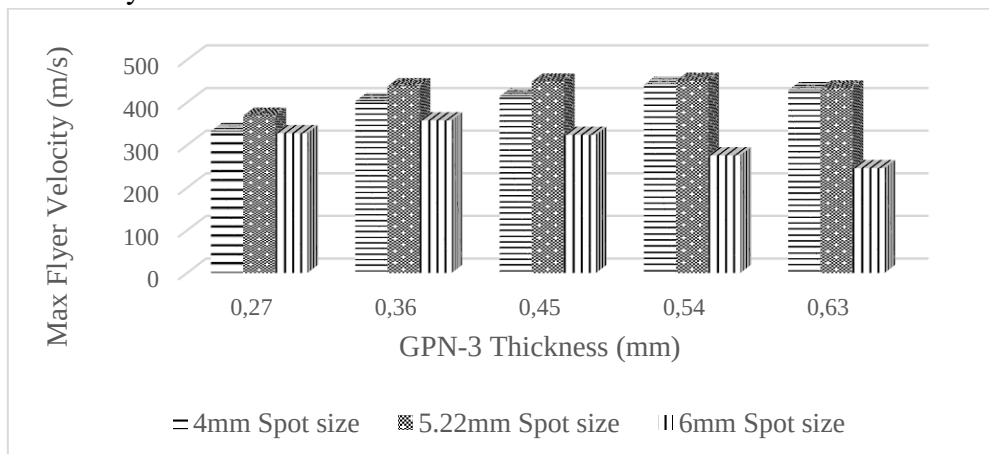
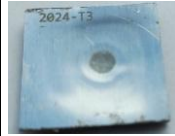


Figure 5 Maximum flyer velocity of 1 mm Al 6061 flyer at different GPN-3 thicknesses and various spot sizes, during ALIW process using borosilicate glass as confinement layer

To evaluate the effect of the optimized parameters on the Augmented Laser Impact Welding (ALIW) process, welding trials were performed using two distinct configurations: (1) a normal setup with a flat gap arrangement between the flyer (1 mm Al 6061) and target (1 mm Al 1100), and (2) an angled setup with a 15° collision angle. **Table 3** summarizes the variable parameters for each configuration. Peel testing confirmed successful bonding, revealing a fully bonded weld nugget in both cases.

Table 3 ALIW Parameters to join Al 6061 as a flyer to Al 1100 as a target with different setup

Materials (Thickness)	Ablative Layer	Confinement layer	Energy (J)	Spot Size (mm)	Setup		Peel Test
6061/1100 (1mm/1mm)	GPN-3	Glass and Sapphire	3	5.22	Stand-off (0.5mm)		
6061/1100 (1mm/1mm)					Angle 15°		

4. Conclusion

Augmented Laser Impact Welding presents a highly promising alternative to traditional joining methods, particularly due to its superior advantages, which make it a viable replacement for rivets in various industries. A key factor in the process is the augmented layer, which significantly enhances the sensitivity of detonation energy, thereby improving overall performance. The key findings of this study are as follows:

- The incorporation of 0.1 g of micro-air bubbles into the original GPN formulation enhances detonation energy, leading to an increase in flyer velocity.
- The use of borosilicate glass as a confinement layer is limited by its tendency to fracture after each laser shot. As a result, sapphire is proposed as a more durable alternative to extend the operational lifespan.
- GPN-3 has demonstrated effectiveness with thicker materials, specifically 1 mm Al 6061-T6, with optimized process parameters identified at a 5.22 mm spot size and 0.54 mm GPN-3 thickness, ensuring the highest flyer velocity.
- Sound joints with welded nuggets were created using 1 mm thick Al 6061-T6 as a flyer and 1mm thick Al 1100 as a target by utilizing the optimized parameters for ALIW found in this work, as emphasized by the peeling test shown in Table 3

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Chapter 4

Electromagnetic joining by forming

Electromagnetic incremental assembly for large thin-walled ring shells and microscopic mechanisms of dynamic deformation overlap regions

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Abstract

This paper introduced an electromagnetic incremental assembly (EMIA) method which employed large thin-walled aluminum alloy ring shells as the processing object and a weak-rigidity assembled structure with an irregular cross-section as the support object. By experiments, the effects of discharge frequency, discharge voltage, and coil position parameters on the assembly outcome were investigated. The results show that this method successfully achieved a tight fit between the skin and the assembled structure, as well as effective support. During the EMIA, there were many two-step dynamic deformation overlap regions. This paper also conducted an in-depth study on the microstructure evolution of these deformation overlap regions, revealing that dynamic pre-forming enhanced the strength and plasticity of the material. The enhanced strength was attributed to the higher dislocation density within grain boundaries. The plasticity enhancement was mainly due to two factors: firstly, the more uniform dislocation distributions which promoted homogeneous deformation; secondly, the shear stress on the second-phase particles became more concentrated, and the shear action was enhanced. As a result, a large number of original particles were sheared into multiple segments, and large-scaled particles were also fractured, leading to the accumulation of dislocations that eventually cut through the particles, releasing local shear stress.

Keywords

Large thin-walled ring shells, Electromagnetic incremental assembly, Microstructure evolution

1 Introduction

Thin-walled ring shells are a classic type of thin-walled structure characterized by a significant disparity between wall thickness and radial dimensions. In modern manufacturing for aerospace, electromechanical systems, and engineering machinery, there is a growing demand for high-precision ring shells with specific cross-sectional profile. For some structurally unstable assembly components that require support, employing thin-walled metal shells as inner lining support structures not only provides stable assembly support but also reduces the overall system weight. For instance, a sandwich tube consists of an inner tube, outer tube and core layer, and experiences minimal deformation under external forces (Huang et al., 2022). However, the thin-walled liners are typically made of lightweight, high-strength, low-formability materials, which pose challenges due to their low structural stiffness and closed processing environments, limiting the applicability of traditional processing methods.

Compared to traditional forming processes, the advantages of electromagnetic forming (EMF) in manufacturing thin-walled ring shells lie in its high strain rate characteristics, which can reduce springback and increase the material's forming limit at room temperature (Psyk et al., 2011). Additionally, as a replacement for traditional punch, the coil structure offers strong adaptability. It can be adjusted according to the part structure, thus overcoming limitations imposed by the processing environments (Cheng and Yu, 2025). Electromagnetic incremental forming (EMIF) is an extended processing form of traditional EMF. By combining multi-step discharges from coil with incremental movement, the EMIF enables the processing of large-scaled sheets and long straight tubes (Cui et al., 2014). Based on the concept of EMIF, this paper introduced a novel incremental processing method—electromagnetic incremental assembly (EMIA). For the assembled structure of rigid polyurethane foam (RPUF) blocks with large-scaled and irregular cross-sectional features, radial support assembly was achieved through the EMIF of large aluminum alloy thin-walled ring skins, allowing the skin and the RPUFs to form an integrated whole and ensuring the stability of the assembled structure. The working principle of EMIA was introduced. By conducting multi-step incremental discharge experiments, the effects of discharge voltage, discharge frequency, and coil positions on the assembly outcome were analyzed. The optimal process parameter scheme was obtained, and the feasibility of EMIA was verified.

In the EMIA process, due to the overlapping of coil incremental positions, there are significant dynamic deformation overlap regions in the skin material. Considering the processing efficiency, the coil overlap between the front and rear discharge steps should not exceed 50%. Therefore, the deformation overlap regions generally undergoes two discharges. Influenced by dynamic pre-forming, the mechanical properties and microstructural evolution in the deformation overlap regions may differ compared to fully dynamic deformation regions. This paper also investigated the effect of dynamic pre-forming on the material forming performance and microstructural evolution under two-step dynamic deformation. The mechanical properties and microscopic mechanisms of aluminum alloy under fully and two-step dynamic deformations were analyzed.

2 Electromagnetic incremental assembly

2.1 Working principle

Fig. 1 illustrates the working principle of EMIA for thin-walled ring shells, which consists of four components: thin-walled ring shell, assembly block, coil, and die. The assembly blocks are radially supported by the shell through EMIF. The process flow is as follows:

(1) The thin-walled ring shell is positioned and installed coaxially with the assembly blocks or die, ensuring that the gap between the shell and the assembly blocks is uniform.

(2) After setting the discharge voltage, perform the single-step discharge forming to achieve the local tight fitting between the shell and the assembly blocks.

(3) Keep the position of the shell fixed relative to the assembly blocks, and move the coil along the axial direction with the preset feed distance, so that the undeformed region of the shell enters the forming region of the coil. After setting the discharge voltage, perform the discharge machining again.

(4) Step (3) is repeated until the entire forming process of the shell is completed.

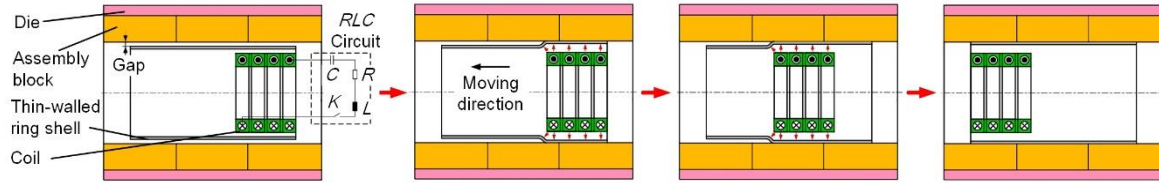


Figure 1: Working principle of EMIA

2.2 Materials and experimental setup

The thin-walled ring shell material is 5A06 annealed aluminum alloy (AA5A06) skin. The shell is made from AA5A06 rolled sheet with a thickness of 1.5 mm, with overall dimensions of 600mm × 450mm × 72mm, and the weld seam is located at the bottom center. The materials of assembly blocks are RPUFs with a density of 200 kg/m³. **Fig. 2a** shows the experimental setup, which consists of four components: thin-walled ring skin, assembled RPUFs, coil, and die. The cross-section of the assembled foams consists of six foam pieces. The coil for irregular structure uses rectangular copper wire with cross-sectional dimensions of 5mm × 7mm, wound for 5 turns. The arc and straight-wall regions of die are made from steel plate with a thickness of 10 mm, which are bent and then welded to base plate with a thickness of 30 mm to form an integrated structure. The maximum stored energy of EMF device is 30 kJ, and the energy storage capacitor has a capacitance of 150 μF. The inductance of the pulse generator in the EMF device is 14.8 μH, and the current curve under a discharge voltage of 11.5 kV is shown in **Fig. 2b**.

The initial assembly between the skin, coil, and assembled RPUFs is shown in **Fig. 2c**. The assembled RPUFs are divided into the arc region, transition region, straight-wall region, and bottom region. The regions of skin that are closely fitted with the bottom and straight-wall regions of the foams do not participate in the bulging deformation. The deformation of the skin primarily affects the transition and arc regions of foams. The complex shape of

foams results in an uneven initial gap between the skin and the foam, with the maximum and minimum initial gaps occurring between the foam transition region and the skin.

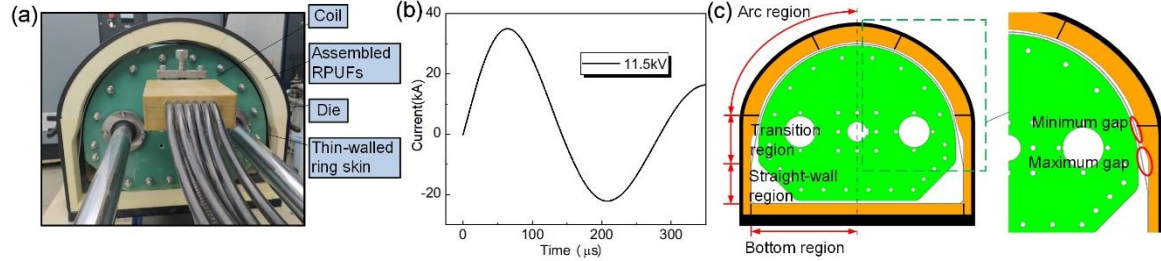


Figure 2: Experimental setup for EMIA (a) experimental equipment, (b) current curve under a discharge voltage of 11.5 kV, (c) initial assembly

In the case where the initial gap between the skin and the foam is determined, the main adjustable process parameters include discharge voltage U , discharge frequency N , and coil position c . The coil position c is represented by the center distance between the skin and the coil, as shown in **Fig. 3**. U_1 represents the first-step discharge voltage, and c_1 represents the center distance during the first discharge, with other parameters expressed similarly.

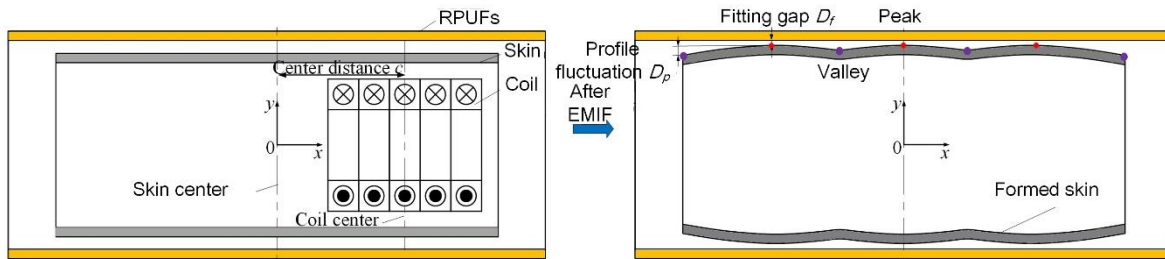


Figure 3: Assembly relationship between the coil, skin, and RPUFs

From **Fig. 3**, when the coil moves incrementally over a large distance, the deformation overlap region becomes smaller, and the skin profile after EMIF may exhibit a wavy pattern. The fitting gap D_f between the formed skin and the foams, as well as the fluctuation D_p in the profile waves, will affect the supporting performance of the formed skin. Additionally, during the EMIA, it is crucial to prevent excessive impacts from damaging the foams.

To measure the profile of formed skin and calculate the profile fluctuation D_p , a Multi-Instrument is used to measure the skin profile along axial direction. The distance data between skin profile and sensor is collected at various points along measurement trajectory. Since the contact area between the formed skin and the foams' arc region is the largest, a linear trajectory is generally chosen for measurement at this location. If the skin undergoes three discharges, the skin profile shows three peaks and four valleys. The difference between the highest peak and the lowest valley is taken as the profile fluctuation D_p of formed skin.

2.3 Experiments

The EMIA experimental schemes were conducted by adjusting different process parameters, as shown in **Table 1**. The experimental results are shown in **Fig. 4**.

Table1 Experimental schemes of EMIA

Parameters	EMIA-1	EMIA-2	EMIA-3	EMIA-4	EMIA-5	EMIA-6
U_1/kV	8.5	8.5	11	11.5	11.5	11.5
c_1/mm	+28	0	0	0	0	0
U_2/kV	8.5	9.5	12	11.5	11.5	11.5
c_2/mm	-28	+20	+20	+20	+20	+23
U_3/kV	/	10	12	11.5	11.5	11.5
c_3/mm	/	-20	-20	-20	-20	-23
U_4/kV	/	/	/	/	11.5	/
c_4/mm	/	/	/	/	+25	/
N	2	3	3	3	4	3

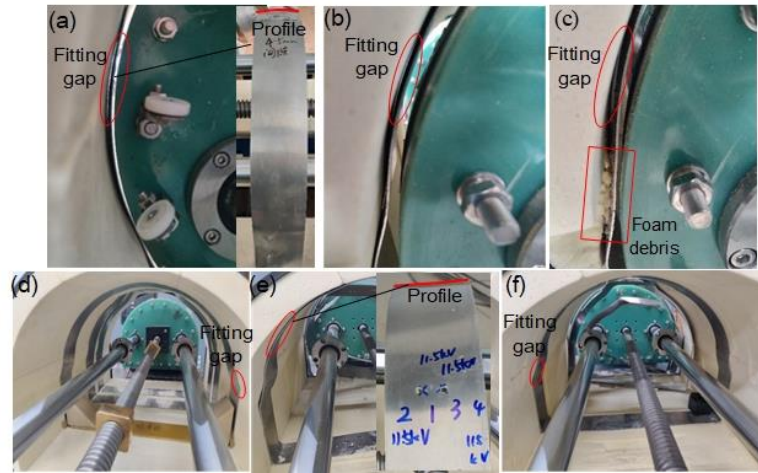


Figure 4: Results (a) EMIA-1 (b) EMIA-2 (c) EMIA-3 (d) EMIA-4 (e) EMIA-5 (f) EMIA-6

From **Figs. 4a** and **4b**, there were obvious fitting gaps between the formed skin and the foams' arc region in EMIA-1 and EMIA-2, which indicated that the discharge voltage of 8-10 kV was too low, and the discharge voltage should be increased further to reduce the fitting gap. In EMIA-1, the profile of the formed skin showed a noticeable "high in the middle and low on both edges" phenomenon, leading to worse profile fluctuation. Therefore, two discharges were not suitable. EMIA-3 attempted to set the discharge voltage to 11 kV and 12 kV. When the discharge voltage was 11 kV, the foams remained intact. However, when the discharge voltage increased to 12kV, the convex feature of foams was crushed, and foam debris appeared, as shown in **Fig. 4c**. This indicated that a discharge voltage of 12kV was too high, and the enormous impact generated by the deformation of skin material led to the damage to foams. Therefore, the discharge voltage need to be slightly reduced. EMIA-4 and EMIA-6 differ only in the coil position for each step. From **Figs. 4d** and **4f**, the fitting effects of the formed skin were similar in both experiments. The skin fitted closely to the arc region of the foams, and there was no obvious damage to the foams. However, there was about a 2.5 mm fitting gap between the skin and foams' concave feature, which resulted from that the initial gap in this region was 11.5 mm, and under the discharge voltage of 11.5

kV, the deformation of skin material was insufficient to fill the concave feature. In EMIA-5, four discharges were performed under a discharge voltage of 11.5 kV. As shown in **Fig. 4e**, there was no fitting gap between the formed skin and the foams, but the profile fluctuation was poor. Since the last two discharges occurred on the right edge of the skin, it caused accumulated deformation on the right edge of skin, resulting in a noticeable upward warping. Therefore, during the EMIA, the coil discharge positions need to be arranged symmetrically around the skin center.

Based on the analyses of fitting effects in EMIA-1~6, the fitting gaps in EMIA-4 and EMIA-6 were minimum. Next, the profiles of the formed skins in EMIA-1~6 were measured. Although the formed skins in EMIA-4 and EMIA-6 both showed good fitting with the foams, the profile fluctuation of formed skin in EMIA-4 was only 0.57 mm, which indicates that the support stability of the formed skin in EMIA-4 was better than in EMIA-6. Therefore, the process parameters in EMIA-4 represented the optimal process scheme, which included three discharge with a discharge voltage of 11.5kV for each cycle, and a coil center spacing of 20 mm during each discharge.

3 Microscopic mechanisms of deformation overlap regions

During the EMIA, the skin material undergoes multiple discharges, resulting in dynamic deformation overlap regions. These overlap regions generally experiences two-step dynamic deformation. There may be differences in mechanical properties and microstructural evolution between the dynamic deformation overlap regions and the single-deformation region. Based on existing literature (Cheng, 2025), quasi-static pre-forming can improve the plasticity of subsequent dynamic deformation. Therefore, dynamic pre-forming may also affect subsequent dynamic deformation.

To investigate it, the mechanical properties of AA5A06 sheets were tested under fully and two-step dynamic uniaxial tensions. The fully and two-step dynamic tensile tests were performed using a split Hopkinson bar. For the fully dynamic tensile tests, the strain rate for is controlled at 3000 s^{-1} . In the two-step dynamic tensile tests, the samples were first pre-formed at a strain rate of 1000 s^{-1} , and then fractured at a strain rate of 3000 s^{-1} .

From the test results, **Table 2** lists the ultimate tensile strength (UTS) and total fracture elongation (TFE) of the material under different tensile conditions. Under fully dynamic deformation, the material's UTS and TFE were 359 MPa and 45.1%, respectively. Under two-step dynamic deformation, the material's comprehensive mechanical properties were better than that under fully dynamic deformation. When the dynamic pre-strain was 16.7%, the material's UTS and TFE were 418 MPa an 54.1%, which increased by 16.4% and 20% compared to the fully dynamic deformation, respectively. The results indicated that dynamic pre-forming behavior can effectively enhance the material's strength and fracture elongation.

Table2 Mechanical properties of AA5A06 sheets under different tensile conditions

No.	Dynamic pre-strain level (%)	UTS (MPa)	TFE (%)	Tension pattern
T-1	/	359	45.1	Fully dynamic

T-2	16.7	418	54.1	Two-step dynamic
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After the tensile tests, the fractures of the samples were analyzed from the view of microscopic, with observation conducted using a transmission electron microscope (TEM). Under fully dynamic deformation, distinct and uneven dislocation cells appeared, as shown in **Fig. 5a**. During dynamic deformation, plastic deformation tended to activate multiple slip systems. The multi-slip system activation and dislocation homogenization effect in dynamic deformation can form at higher plastic strain levels and strength. **Fig. 5b** shows the dark field image of the fracture. A large number of second-phase particles were fractured into two segments, and the sizes of cut particles exceeded 500 nm. Due to the severe fully dynamic deformation, the interactions between particles and dislocations were significantly enhanced.

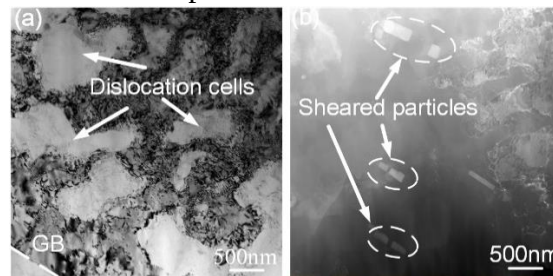


Figure 5: Fully dynamic deformation (a) bright field image (b) dark field image

As seen in **Fig. 6a**, the dislocation density within the grain boundaries (GBs) under two-step dynamic deformation was higher, and the sizes of dislocation cells were smaller. The entanglement and interaction between dislocations occurred more frequently, thus hindering dislocation movement and resulting in higher material strength. **Fig. 6b** shows a bright field image of dislocation entanglements. The dislocation distribution under two-step dynamic deformation was more uniform than that under fully dynamic deformation, indicating that two-step dynamic deformation contributed to a more homogeneous deformation, thereby exhibiting better plasticity. **Fig. 6c** shows the dark field image, a large number of fine second-phase particles were present, which may be the result of multiple cuts of the original second-phase particles due to two-step dynamic deformation. Firstly, the dynamic pre-forming caused interactions between the particles and the piled-up dislocations, leading to original particle cutting, similar to fully dynamic deformation. The second-step dynamic deformation was more intense with a wider range of dislocation pile-ups and greater dislocation multiplication, leading to a significant increase in stress concentration. As a result, the remaining particles were more easily cut in second-step dynamic deformation. The cumulative stress concentration from two steps made larger particles more susceptible to shear fracture. In **Fig. 6c**, there was sheared particle larger than 2500 nm in size.

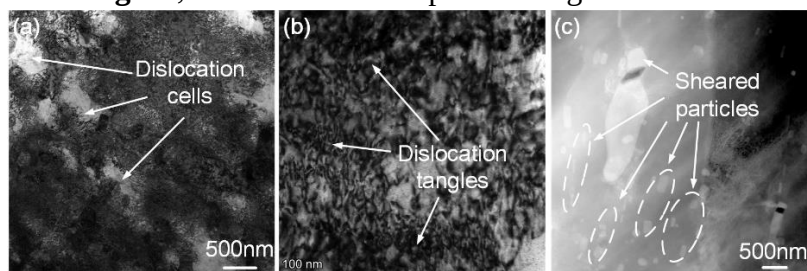


Figure 6: Two-step dynamic deformation (a), (b) bright field images (c) dark field image

4 Conclusion

The following conclusions are drawn:

1. The EMIA method was proposed, and its feasibility was experimentally validated. Moreover, the process parameters for EMIA were optimized.
2. Compared to fully dynamic deformation, the strength and fracture elongation of the material were improved during two-step dynamic deformation.
3. The high dislocation density within GBs generated by two-step dynamic deformation resulted in higher material strength.
4. The material plasticity enhancement can be attributed to two factors. First, the dislocation distribution during dynamic deformation was uniform, preventing local strain concentration. Second, the original particles were sheared into two or more segments, and larger particles were more prone to fracture. The accumulated dislocations then cut through the particles, allowing them to continue moving, thereby releasing local shear stress.

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Numerical and Experimental Studies on Electromagnetic Pulse Crimping of Joining D9 Steel Tube on SS316L (N) Plug

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Abstract

Electromagnetic Pulse crimping process is an advanced technique that utilises high-intensity electromagnetic forces to deform and join metallic components. This study investigates numerical simulations and experimental work on the EM crimping process used to crimp the D9 steel tube to SS316L (N) plug. Optimum values of tube impact velocity and displacement at the D9 steel tube with SS316L (N) plug are determined using COMSOL© Multiphysics simulations. Effects of process parameters such as current, magnetic field, plastic strain, impact velocity and voltage levels are investigated for their effect on crimping of D9 tube. In an EM Pulse crimping process experimental RLC setup was used for producing the excitation current. Experiments are performed for different discharge voltages for D9 steel tube, SS316L (N) plug, copper driver and field shaper setup. The metallurgical characterisation was performed using optical microscopy to check the weld line for bonding. The experimental and simulation data show good agreement for the deformation of the D9 tube. As voltage or energy input increases, keeping RLC constant, an increase in impact velocity, effective plastic strain, deformation, and peak coil current is observed. At the same time, the frequency of the current remains constant, which is favourable for EM Pulse crimping.

Keywords

Electromagnetic Pulse crimping, Impact Velocity, Magnetic Flux density, Voltage.

1 Introduction

EM Pulse crimping is an advanced, high strain rate metal joining technique that utilizes Lorentz forces to achieve deformation and bonding without direct mechanical contact. It can be applied to sheet metal, tube-to-tube, or tube-to-bar joining configurations. In fast breeder, D9 tubes are used for fuel cladding and SS316L (N) plugs used for sealing or capping these tubes during fabrication or assembly. (Rajak & Kore, 2017) reported that EM pulse crimping is an advantageous process for joining dissimilar metals with different mechanical properties. (Psyk et al., 2011) research focuses on working principles of EM forming, deformation pattern of workpiece and flow of energy. Kumar et al (2020) demonstrated the EM pulse welding of steel to steel in combination with microscopic examination. (Shen et al., 2022) reported that EM pulse crimping is a promising method for connecting high-voltage wiring in electric vehicles. Experiments on aluminium tubes showed that optimal crimping occurs when the size of the field-shaper is shorter than that of EM coil, and their gaps form a 90° angle.

This paper presents a numerical simulation of joining a D9 tube to an SS316L (N) plug using EM Pulse Crimping with a single-taper field shaper, compared to experimental results. The authors validated the experimental results with numerical analysis and examined the effects of process parameters, such as current, magnetic field, plastic strain, impact velocity and deformation at different voltages.

2 EM pulse crimping setup

Fig.1 illustrates the EM configuration employed in the EM crimping process. Upon closing the circuit switch, the energy accumulated in the capacitor is released. The crimping coil's fluctuating magnetic field produces magnetic pressure.

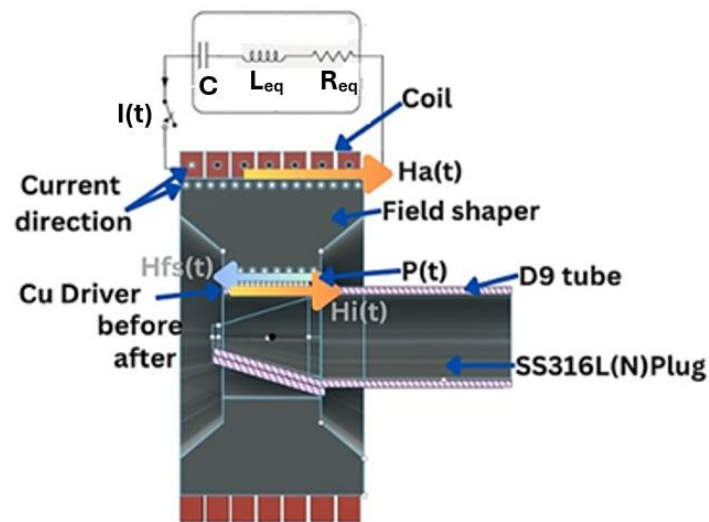


Figure 1: EM pulse crimping setup

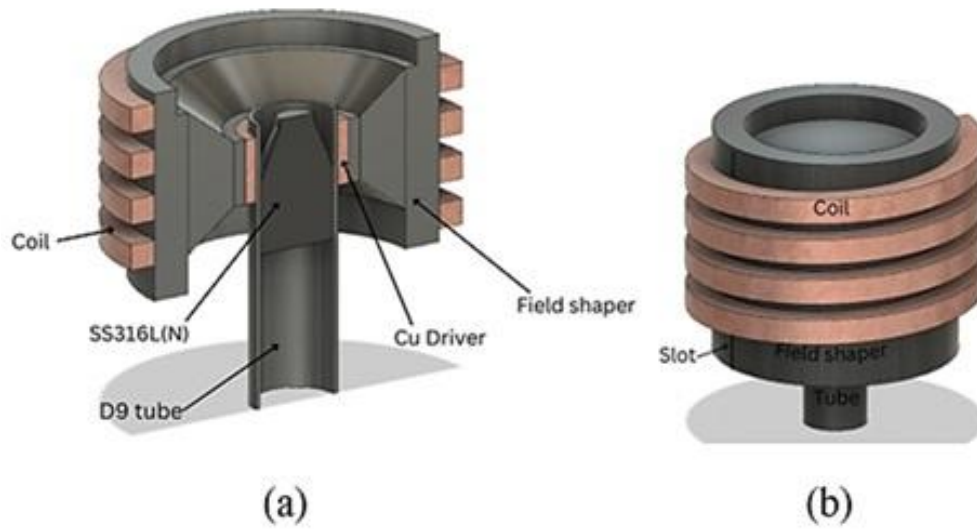


Figure 2: Components of EM crimping

The coaxially aligned D9 tube and SS316L(N) plug are crimped by a Cu driver, which is accelerated by Lorentz force generated from eddy currents induced by the coil's magnetic field, pushing the workpiece beyond its yield strength. The EM equipment, rated at 40 kJ, operates with a charging voltage of 20 kV and a capacitance of 224 μ F. It functions at a frequency of 16 kHz and produces a peak current of 450 kA at 20 kV. Fig.2 (a) shows the schematic diagram of the components of the EM crimping process.

2.1 Coil, Field-shaper and Workpiece

The 4-turn copper coil (Fig. 2 b) generates the magnetic field and pressure for EM forming. Its first and last turns measure 65×12 mm, the middle turns 65×16 mm, with an inner diameter of 90 mm and outer diameter of 220 mm. The field shaper employed is a tapered copper type with an axial slit measuring 0.8 mm. The field shaper is utilized to focus the magnetic field on the critical area of the workpiece to achieve a mechanical joint between the D9 steel tube and the SS316L(N) plug. The D9 steel tube and SS316L(N) plug combination with copper driver is used as a workpiece in this work. The coil is kept over the workpiece. The D9 steel tubular structure is crimped to the SS316L (N) plug in this study.

3 Numerical simulation using COMSOL© Multiphysics

The simulation uses the AC/DC module to solve Maxwell's equations for magnetic fields. An RLC circuit charges a capacitor, which discharges through a trigger switch to generate current in the coil. The flyer tube is placed in proximity to the EM coil, which induces current in flyer tube, **Eq. 1** gives the value of discharging energy (E), and Eq. 2 provides the value of charging voltage (V) in terms of capacitance (C) and inductance(L) and current (I).

$$E = \frac{1}{2} CV^2 \quad (1)$$

$$V = \frac{I}{\sqrt{\frac{C}{L}}} \quad (2)$$

This system uses Maxwell's equation.

$$\frac{d^2 I(t)}{dt^2} + 2\xi\omega \frac{dI(t)}{dt} + \omega^2 I(t) = 0 \quad (3)$$

By solving differential Eq. 3, the resultant coil currents I in Eq. 4 are obtained, representing a damped sinusoidal wave with initial discharge voltage V , damping factor β , and angular frequency ω , and Time t .

$$I(t) = \frac{V}{L\omega} e^{(-\beta t)} \sin(\omega t) \quad (4)$$

The magnetic pressure and the Lorentz force are generated from the discharging current. The magnetic force acts on the D9 tube and the formula for calculating the Lorentz force (f) involves current density (J) and magnetic flux density (B) as per Eq. 5. The magnetic flux density norm $B(t)$ and magnetic field strength intensity $H(t)$ are provided by Eq. 6 and 7.

$$f = J \times B = \frac{1}{\mu} (\nabla \times B) \times B \quad (5)$$

$$B(t) = \mu \frac{N}{l} I(t) \quad (6)$$

$$H(t) = \frac{NI(t)}{l} \quad (7)$$

In this context, N represents the number of turns in the EM coil, l denotes the coil's length, and μ stands for the magnetic permeability of the coil's material. This gives output as magnetic fields, meaning Lorentz forces are produced. The result of the EM module is given as input to the solid mechanics module, which produces deformation. Figure 3 shows this Multiphysics simulation process. The Cowper-Symonds constitutive model in Eq. 8, as reported by (Kapil & Sharma, 2015) It is used to find out the plastic deformation of the workpiece at each time increment.

$$\sigma = \sigma_y \left[1 + \left(\frac{\dot{\epsilon}}{p} \right)^m \right] \quad (8)$$

Here, σ_y = flow stress, p and m = specific material constants and $\dot{\epsilon}$ = plastic strain rate (s^{-1}). In COMSOL[®] Multiphysics, this Cowper-Symonds constitutive model is a swift option in the solid mechanic's module.

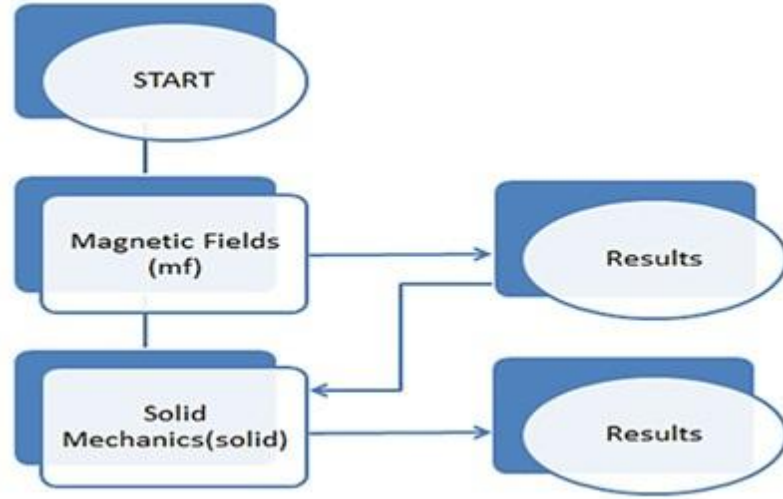


Figure 3: Flowchart of coupled simulation

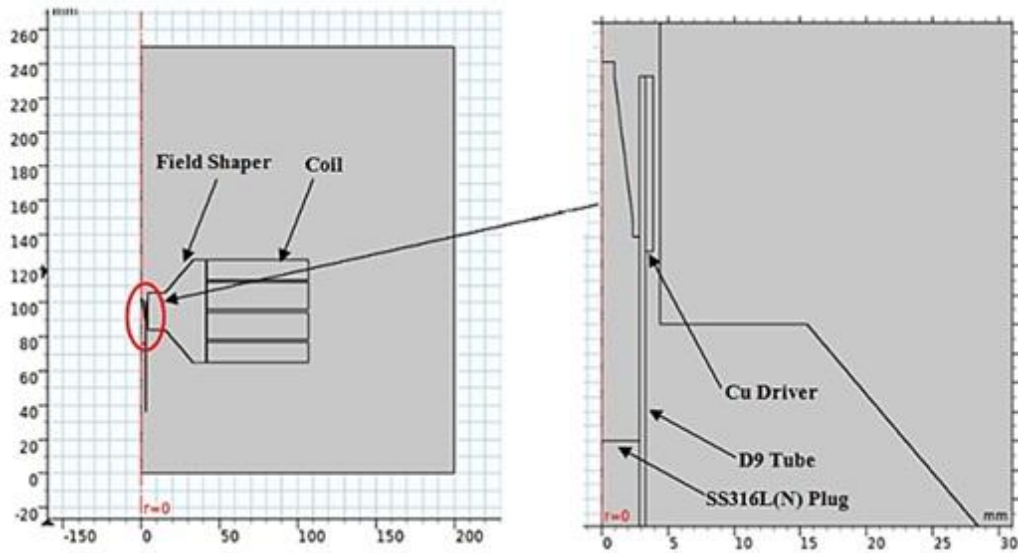


Figure 4: 2D axisymmetric model

Fig. 4 shows the 2D axisymmetric model of EM crimping, with a coaxial copper coil, D9 tube, Cu driver, and SS316L(N) plug. Only half the setup is modelled due to symmetry. Simulation inputs, including RLC values, are listed in **Table 1**.

Charging voltage (kV)	15	16	17	18
Current frequency (kHz)	16			
Inductance (μ H)	1.8			
Capacitance (μ F)	224			
Resistance (m Ω)	9.38			
Angular Frequency (rad/s)	100530.96			

Table 1: Process Parameters

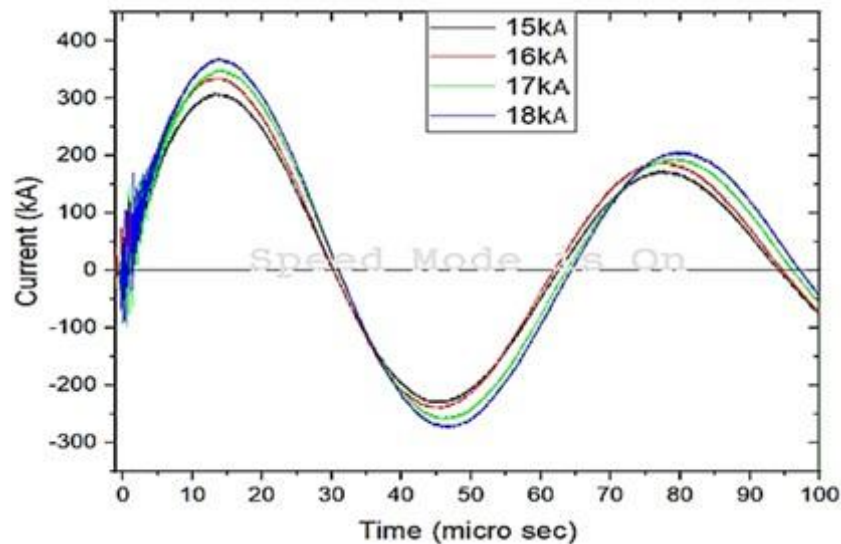
Voltage	Peak Coil Current (kA)	
	Experimental Values	Simulation Values
15kV	310	299.77
16kV	334	338.87
17kV	354	352.17
18kV	374	375.16

Table 2: Peak Coil Current (kA)

4 Results and Discussion

4.1 Current Curve

Fig. 5 and Fig. 6 illustrate the current waveform in the EM Crimping process, comparing experimental and 15kV, 16kV, 17kV, and 18kV simulation results. The current is measured using rogowski coil as reported by (Nadimetla et al., 2024) as shown in Fig. 5 as an experimental curve. Table 2 summarizes the Peak coil current values obtained in experimental and simulation during the EM Crimping process at discharge voltages.

**Figure 5** Experimental curve of Time Vs coil current for 15kV, 16kV, 17kV, and 18kV

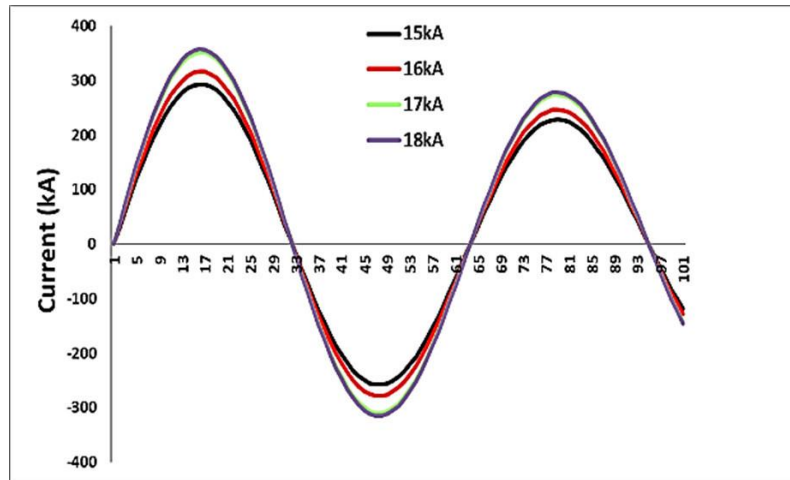
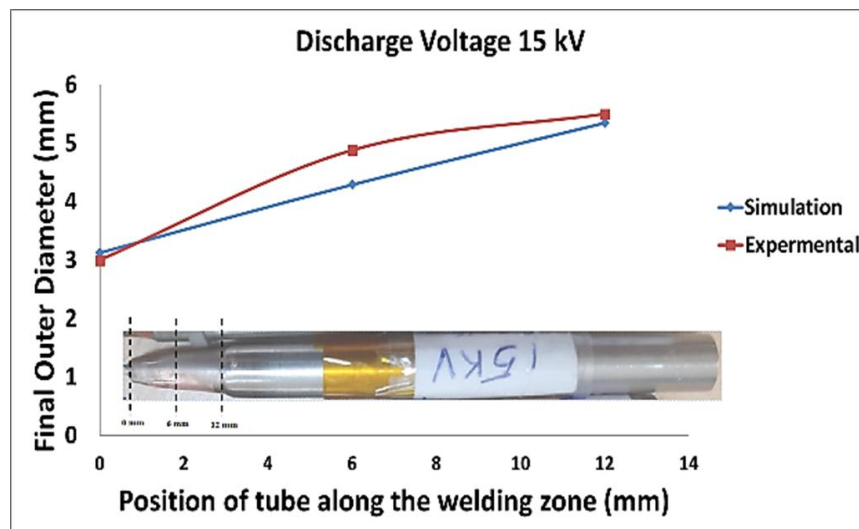


Figure 6 Simulation curve of Time Vs coil current for 15kV, 16kV, 17kV, and 18kV

4.2 Displacement

Experiments and simulations at 15–18 kV (Fig. 7 a–d) show non-uniform deformation along the D9 tube due to uneven magnetic pressure. Simulation results closely match experimental data, with 98% accuracy. The strong correlation confirms the reliability of numerical methods for predicting EM behaviour, though minor local deviations may result from material variations, mesh resolution, or experimental misalignments. Fig. 8 shows rapid, localised displacement at 15 kV, with peak deformation at all three tube locations occurring within microseconds.



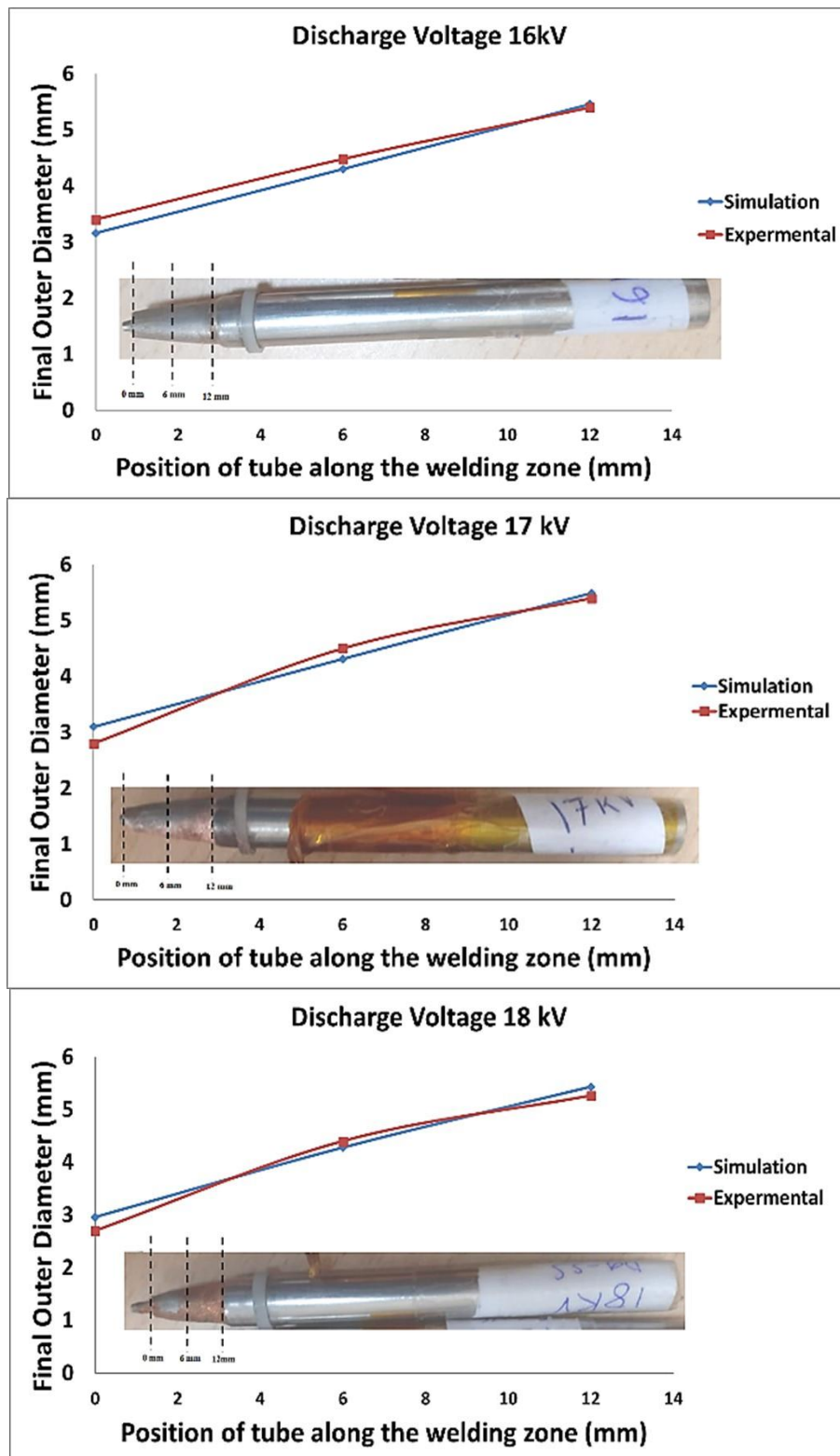


Figure 7. Outer diameter deformation of D9 Tube for (a)15 kV discharge, (b) 16 kV discharge, (c) 17 kV discharge, (d) 18 kV discharge.

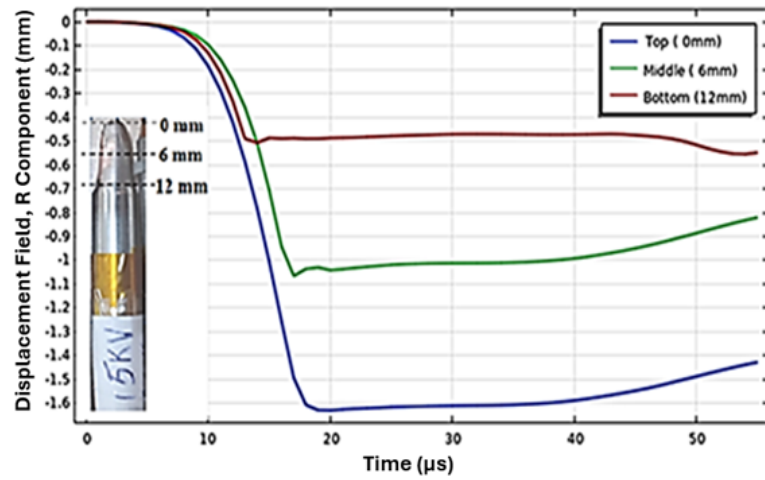


Figure 8: Time Vs the radial Displacement

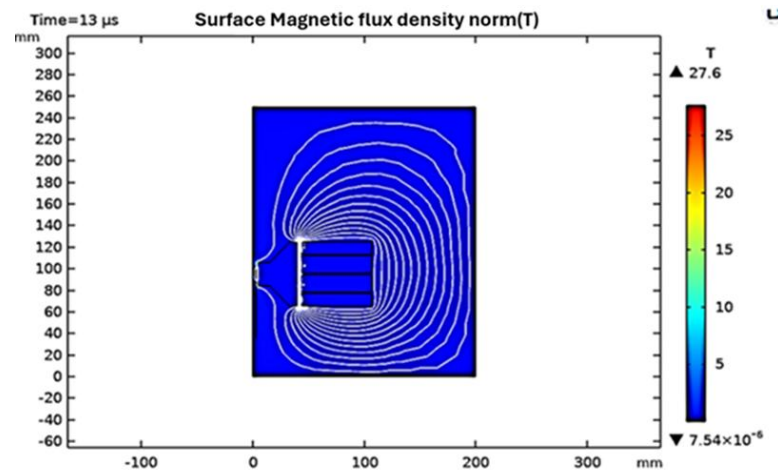


Figure 9: Magnetic field of a complete system

The findings validate the model and highlight how discharge voltage affects deformation. Peak electromagnetic force causes rapid initial acceleration and a sharp rise in radial displacement. After 18 μs , the EM force drops as the current decays. Around 40 μs , a secondary pulse induces a weaker force, causing slight oscillations in radial displacement due to inertia and internal stresses.

4.3 Magnetic Field

Each turn of the EM coil generates a localized magnetic field, which collectively results in a concentrated field at the top portion of the D9 tube, as illustrated in Fig. 9. This is due to the superimposed effect of the coil turns, leading to higher field intensity near the starting turn region. The simulation results presented in Fig. 10 show that the maximum magnetic

field varies significantly along the outer wall of the field shaper. Points A and C are positioned adjacent to coil turns, whereas Point B is located next to the gap between two coil turns: point A (top end) exhibits 23 T, point B (middle) 14 T, and point C (bottom end) 22 T. This non-uniformity confirms that the EM force acting on the tube is position-dependent, influencing the localized deformation observed experimentally. Fig. 11 further shows that the highest magnetic flux densities are achieved between the inner wall of the field shaper and the copper driver, reaching 40T at points A and C and 39T at point B. The analysis is strengthened by detailed magnetic field mapping at key points, explaining the observed asymmetrical deformation. Though based on ideal coil symmetry and uniform material, the model closely matches experimental results, indicating it effectively captures the core process physics.

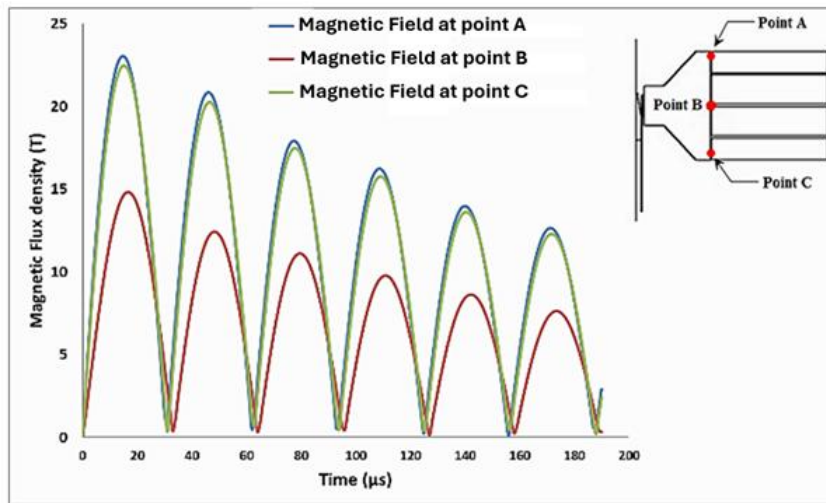


Figure 10: Time Vs. Magnetic Field at the gap between the coil & the field shaper (input voltage -15kV)

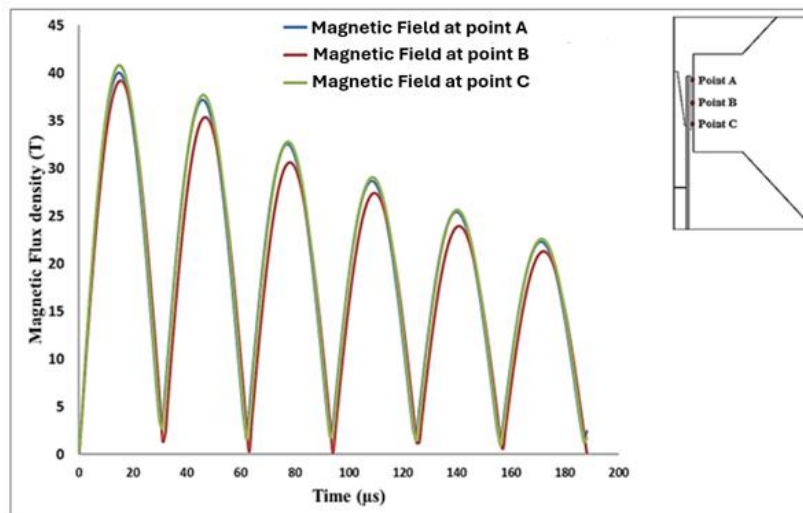


Figure 11: Time Vs Magnetic Field at the gap between Cu driver & field shaper (input voltage -15 kV)

4.4 Impact Velocity

Impact velocity plays a critical role in the effectiveness and outcome of EM, as it directly influences the strain rate and degree of plastic deformation in the workpiece. Fig. 12 presents the impact velocity at three cut sections—points A, B, and C—along the length of the D9 tube at a discharge voltage of 15 kV. At point A (top end), the peak impact velocity reaches 287.07 m/s at 15 μ s. At point B (middle), it is 203.50 m/s at 14 μ s, and at point C (bottom end), it is 147.98 m/s at 13 μ s. Impact velocity decreases from the tube top to bottom, matching the non-uniform magnetic field. Higher top velocity improves formability but risks defects, while lower bottom velocity may cause incomplete forming. Despite minor assumptions, velocity trends closely match experiments, validating the simulation.

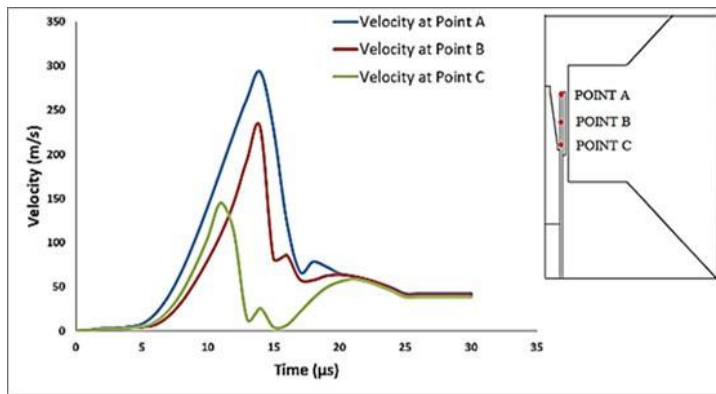


Figure 12: Time Vs velocity of the tube (input voltage- 15 kV)

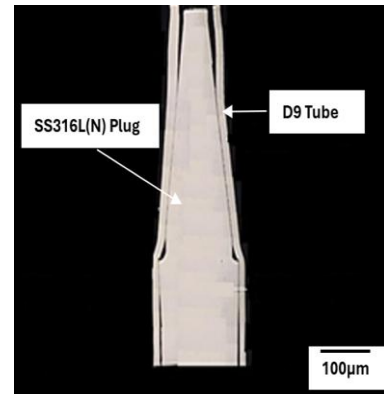


Figure 13: Optical micrograph of EM crimping sample

4.5 Characterization of the crimped samples

Joint quality is first evaluated visually for shape and geometry. Crimped microsections are prepared by abrasive wet cutting with an Al₂O₃ wheel, followed by hot mounting. Polishing is done using 200–1000 grit SiC papers and Al₂O₃ suspensions (5 μ m and 1.5 μ m), then etched for 3 minutes in a nitric-hydrofluoric acid solution.

Optical microscopy (Fig.13) shows no metallurgical bonding in the EM crimped joint between the D9 tube and SS316L(N) plug. Joints form mainly by mechanical interlocking through plastic deformation, suitable for many structural uses but less so for hermetic sealing or high loads. This analysis confirms the joining mechanism with microstructural evidence.

4.6 Effect of voltage on EM crimping

Fig. 14 to 16 collectively highlight the influence of discharge voltage on key EM parameters, namely, magnetic field strength, impact velocity and effective plastic strain of the workpiece. Fig. 14 shows that peak magnetic field strength rises with discharge voltage, driving stronger EM forces. This aligns with the initial velocity rise in Fig. 15, confirming their causal link. Fig. 15 shows workpiece velocity rising with discharge voltage, peaking at 293.59 m/s at 18 kV, due to stronger EM forces and magnetic pressure, consistent with EMF principles.

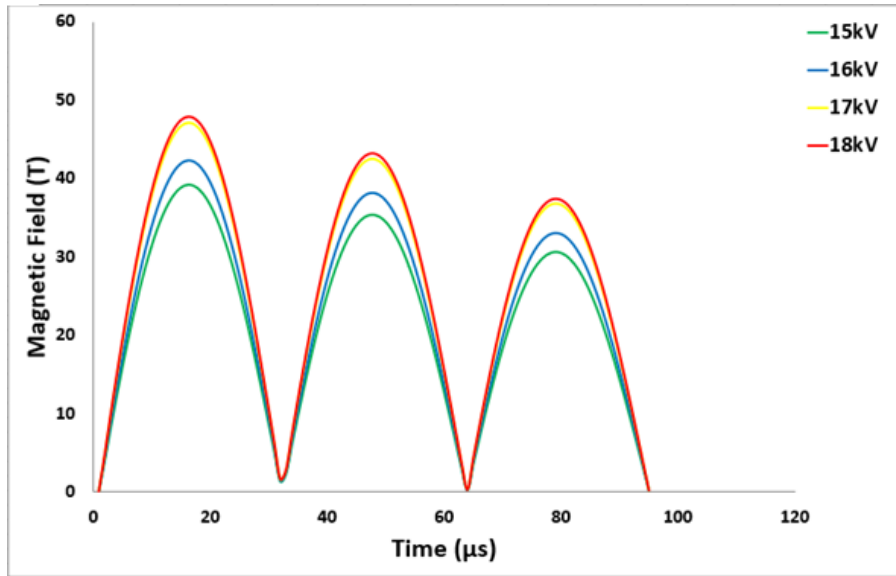


Figure 14: Variation in magnetic field density norm with time at different discharge voltages

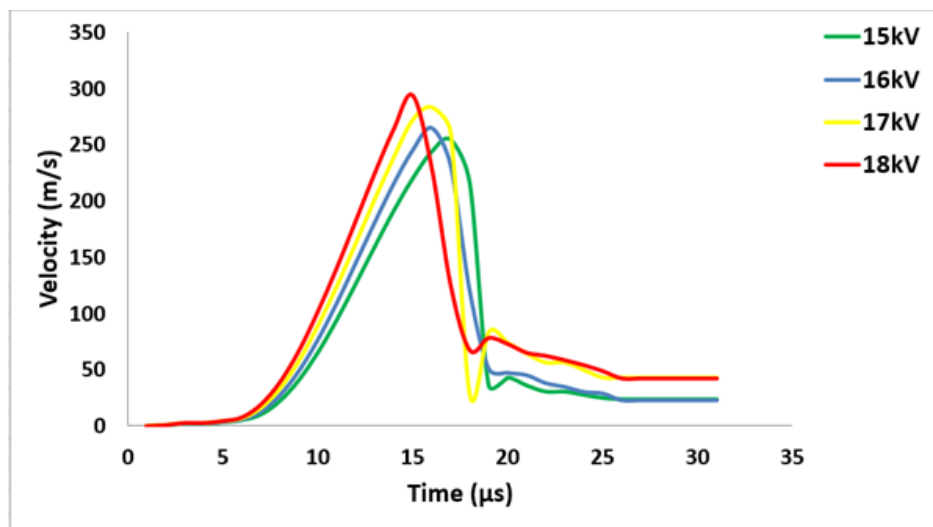


Figure 15: Variation in impact velocity with time at different discharge voltages.

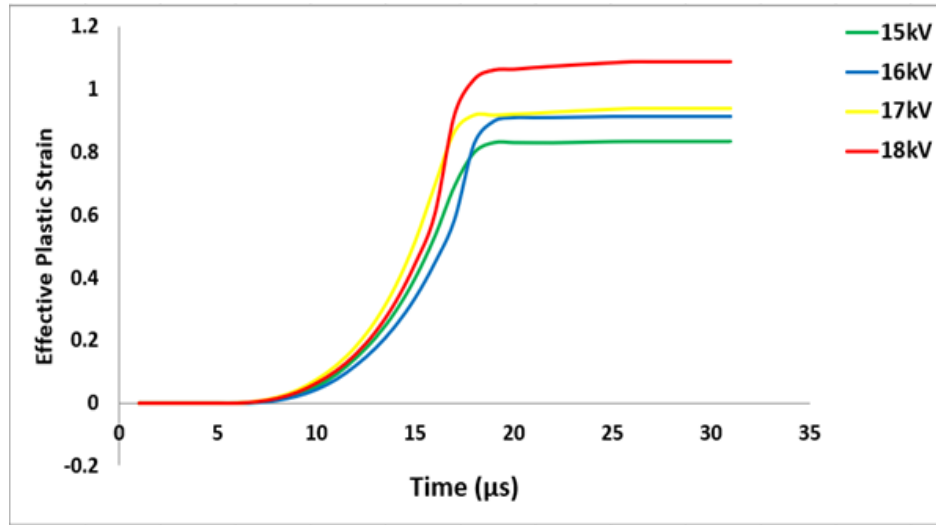


Figure 16: Variation in effective plastic strain with time at different discharge voltages

Fig. 16 illustrates how this enhanced dynamic loading influences material deformation. The effective plastic strain of the workpiece increases with discharge voltage, peaking at 1.0309 at 18 kV. Higher EM force boosts forming velocity and deformation, increasing plastic strain and energy absorption. This helps complex forming but requires careful voltage control to avoid material failure, highlighting the need to optimize discharge voltage.

5 Conclusions

In the present study, Numerical Simulation of joining of D9Tube to SS316L (N) Plug by EM Pulse Crimping with Field Shaper is carried out compared with the experimental work, and the following are the conclusions from this study:

- Though experiments used a 4-turn coil with peak displacement at the tube top and 40 T field at the field shaper center, the results apply to similar coil setups, as the field shaper critically amplifies magnetic intensity in the forming zone.
- The radial displacement graph from the 2D-axisymmetric COMSOL© Multiphysics simulation shows strong agreement (98%) with experimental results.
- The optical micrograph reveals a mechanical fit with no evidence of metallurgical bonding.
- With increased voltage or energy input while keeping resistance (R), inductance (L), and capacitance (C) constant, impact velocity, effective plastic strain, deformation, peak coil current, and stress all increase, whereas the current frequency remains unchanged. This behaviour is favourable for EM crimping.

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Magnetic Pulse Crimping of Power Cables: Process Simulation & Performance Characterization

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Abstract

Magnetic Pulse Crimping (MPC) represents an interesting alternative to mechanical crimping. This solution is a contactless process using dynamic Lorentz forces induced by high pulse power electrical discharge through a coil. MPC has proven to be effective for cables terminals with both high electrical and mechanical performances. This solution differs from conventional methods for cables with large cross-sections for high voltages especially those manufactured for electric vehicles. The challenge lies in the high thickness of the terminal and the cable to be crimped together while maintaining the performance of the connection during use and ageing, that can be hard to hold with traditional processes. The aim of this work is to characterize the electrical resistance of 120 mm² power cables terminals that are assembled by MPC, with the goal of reducing the process width and potentially reduce the lug mass. First, a numerical study of the process is performed in order to design and characterize the crimping level depending on main process parameters, in particular through the compaction level of the assembly. Then, an experimental study has been carried out by testing assemblies with different crimping widths and generator energies, in order to reduce the crimped length without degrading the performance and repeatability. For each case, the electrical resistance of the cable/lug interface is measured, before and after exposure to thermal and humidity cycling representative of real-life conditions. The results are compared with samples produced using mechanical crimping to compare the performances and repeatability of each process respectively.

Keywords

High Pulse Power (HPP), Magnetic Pulse Crimping (MPC), Power Cable, Multiphysics Simulation

1 Introduction & Stakes

Magnetic Pulse Crimping (MPC) offers a promising alternative to traditional mechanical crimping. It is a contactless process that utilizes dynamic Lorentz forces generated by a high-power electrical pulse passing through an inductor. MPC has demonstrated its effectiveness for cable terminals, delivering both excellent electrical and mechanical performance (M. Kim and J. Shim, 2023). Unlike conventional techniques, MPC is particularly well-suited for cables with large cross-sections intended for high-voltage applications (N. Ali, et al., 2023), such as those used in e-mobility—typically up to 200 mm².

The main challenge lies in the significant thickness of both the terminal and the cable, which can pose problems for traditional processes such as mechanical crimping. The pressure required to properly crimp the terminal in a quasi-static process must compress all strands uniformly across the cross-section. This generates considerable internal stresses within the cable and often requires a relatively bulky connector to maintain a reliable hold. However, current industry trends aim to reduce product weight and optimize manufacturing costs, which positions MPC as a strong candidate for such applications.

The purpose of this project is to demonstrate that MPC offers equivalent—or even superior—performance and repeatability compared to mechanical crimping, and to show that it is possible to optimize the process parameters to produce shorter crimped assemblies, thereby reducing overall assembly weight. Both numerical and experimental investigations were conducted by testing three MPC configurations, with crimping width and generator energy levels as the main varying parameters. Electrical resistance at the cable/lug interface was measured before and after subjecting the samples to thermal and humidity cycling. The performance of these samples was then compared to that of assemblies produced using mechanical crimping.

2 MPC Power Cables Simulation & Design

The study case selected for this work is a classical copper lug/cable terminal with a 120 mm² section. The main data related to this assembly are summarized in the **Table 1**.

	Lug Characteristics		120mm ² Conductor cable	
	Outer Diameter	21.8 mm	Outer diameter	15.1 mm
	Lug Thickness	2.5 mm	Single wire diam.	0.51 mm
	Working Length	26 ± 2 mm	Wires number	589
	Material	Copper-ETP	Wire material	Copper
	Finish	Tin over Nickel	Electrical res.	0.153 mΩ/m

Table 1: Assembly to be crimped main characteristics

MPC between two cylindrical pieces is performed by discharging a high current —up to 1MA over 20 μs—with a capacitive bank through a coil (see **Figure 1**). The generator used is designed by Bmax[®] and can store up to 64 kJ. The pulse induces an eddy current, first in a field shaper, then in the conductive workpiece (copper lug) located inside the

tooling. The produced strong magnetic field creates Lorentz forces accelerating the lug away from the coil (**Figure 2a**). The deformed part then impacts and crushes the cable. As minimum contact resistance is sought for such assemblies, crimping quality strongly depends on two parameters: the compaction level and the lug/strands contact surface.

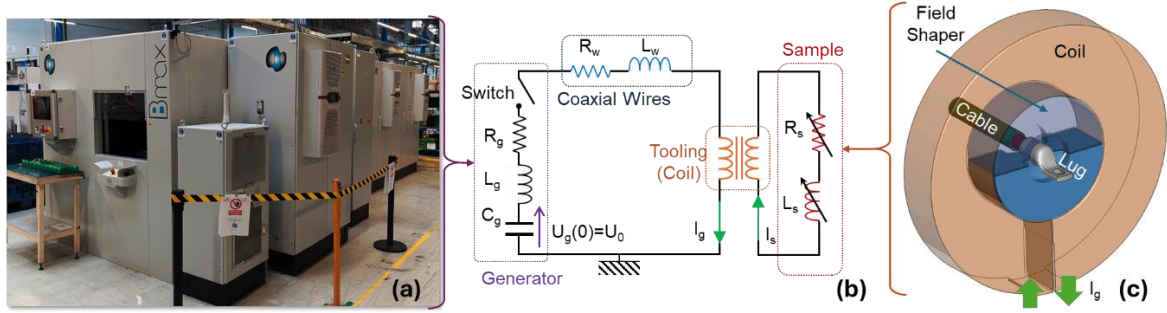


Figure 1: Magnetic Pulse Crimping setup: Bmax generator and workstation (a) ; Equivalent electrical circuit (b) ; Process tooling scheme (c)

The first step of this study is to determine the main process parameters for 120 mm² terminals. An analytical approach can be used to link the generator parameters to the mechanical loading applied on lug. The system can be modelled as a series RLC initially charged at a given energy level E_0 . By solving this circuit, the peak current pulse characteristics can be evaluated (**Eqs. 1 to 3**). Then, the magnetic pressure loading is deduced from the electromagnetic relations under a radial 1D assumption (**Eq. 4**) and non-magnetic materials (magnetic permeability $\mu = \mu_0 = 4\pi \cdot 10^{-7}$ H/m). The pressure evolves as a damped square sine wave and its magnitude P_{max} is proportional to the generator energy (**Eq. 5**) and inversely proportional to the square root of the field shaper's active zone (AZ). This non-trivial dependence includes the maximum B-field relation (B_{max}), with a shape factor (k_{sh}) accounting for inductor edge effects (H. Knoepfel, 1970), and also results from the tooling inductance. The latter influences the current/voltage relationship through the k_u coefficient. The coefficient linking the maximum magnetic pressure (P_{max}) to the stored energy can be obtained through dedicated EM 2D-simulations, yielding $k_p=1.3$ kPa. $\sqrt{m}/J$.

While the energy required to achieve a given mechanical impulse can be estimated using this approach for a fixed geometrical configuration, the corresponding assembly compaction level is not straightforward to determine. Thus, the next step aims to establish a relationship between the magnetic pressure and the cable's compacted state after the process.

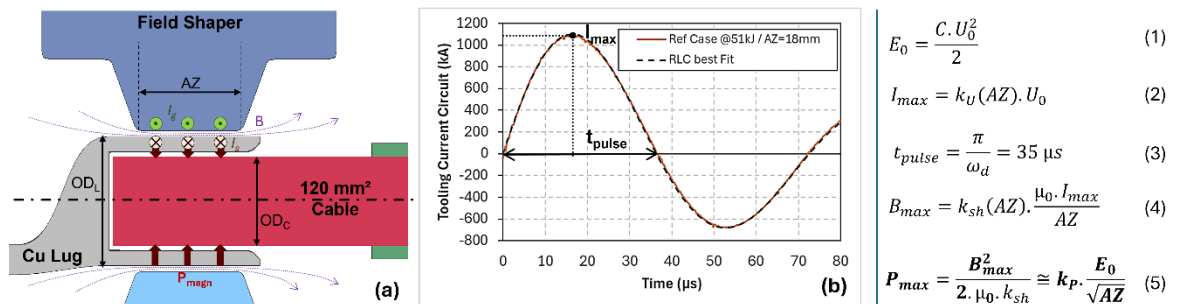


Figure 2: MPC process principle (a), Typical current pulse (b), and Basic equations

As MPC process occurs in a timescale of tens of microseconds, dynamic time-resolved simulations are useful for analysing the physical phenomena involved in compaction. MPC modelling is performed with LS-DYNA, a 3D parallel Magneto-Hydro-Dynamic (MHD) multiphysics code. The electromagnetic solver employs the eddy current approximation. Maxwell's equations are solved using a Finite Element Method (FEM) for solid conductors, coupled with a Boundary Element Method (BEM) to account for the surrounding air or insulators. To characterize the cable compaction behaviour submitted to a compressive pulse, a 2D mechanical model of the cross-section of the lug and multi-strand cable assembly is developed. The cable is represented by all its individual wires. This model enables visualization of the compaction process, as shown in **Figure 3** (A.K. Rajak and S.D. Kore, 2018R.D. Krieg, 1972). When the pressure pulse is applied to the lug outer diameter, the latter is accelerated and then impacts the outer cable strands. The wires begin to compress, and a compaction wave propagates towards the centre of the cable, where an overpressure is observed. A secondary pressure wave then expands back toward the lug. This effect can generate an undesired bounce-back of the ring, depending on the loading characteristics.

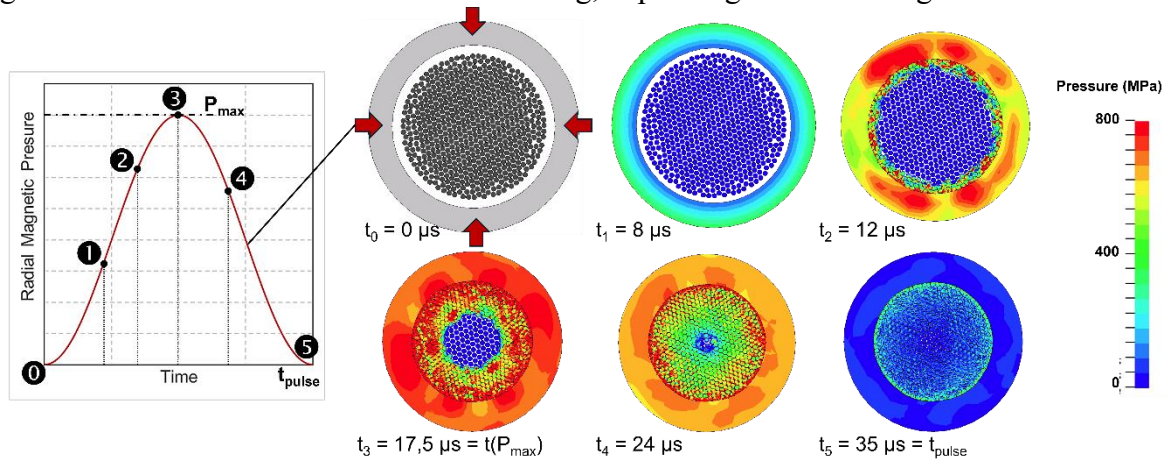


Figure 3: 2D 120mm² Cu cable compaction – $P_{max} = 500$ MPa & $t_{pulse} = 35$ μ s

The compaction level evolution is monitored via the cable's volumetric strain. The latter can be linked to the magnetic pressure peak P_{max} , leading to the **Figure 4**. A minimum pressure $P_0 = 65$ MPa is required to compress the lug to the cable outer diameter. The loading threshold to ensure a full compact cable (i.e. compacted cross-section = 120 mm²) is $P_c = 485$ MPa. Beyond this limit, the pressure increases exponentially to further compact the cable.

Thanks to this model, one can link directly the generator energy to the compaction level with **Eq. 5**. The experimental campaign to define is based on a reference configuration and two variations. The reference case uses a field shaper with an 18 mm active zone (AZ on **Figure 2**) with an energy leading at full compaction threshold: $P_{max,ref} = 500$ MPa, equivalent to $E_{0,ref} = 51$ kJ. The two other cases will employ a narrower field shaper and crimped zone (AZ=12 mm). Case #2 is designed to obtain the same compaction level ($P_{max,\#2} = P_{max,ref}$), leading to an initial energy of $E_{0,\#2} = 36$ kJ. Conversely, Case #3 has the same energy as the Ref. case, implying a higher loading of $P_{max,\#3} = 625$ MPa.

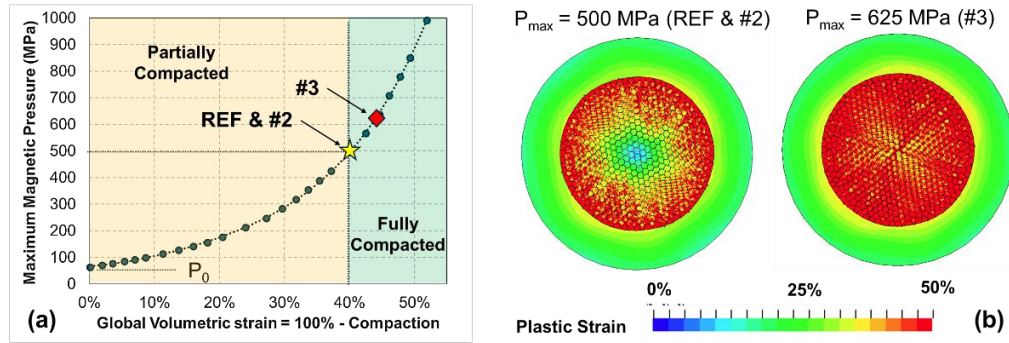


Figure 4: 120mm² Cu cable compaction level in function of the magnetic pressure peak (a) ; 2D Multi-strand final deformations for all the designed cases (b)

When considering the high number of wires that comprise the cable, this approach is not suitable for a 3D complete process analysis. A good alternative is to model it as a single part with an equivalent transversely isotropic law. The axial component of the model (parallel to the strands axis) is an elastic-plastic bilinear law with a variable bulk modulus depending on the volumetric strain. The transverse behaviour of the material is represented by a compaction model (see **Figure 5a**) similar to those used to model foam and soils (R.D. Krieg, 1972). The compression curve can be identified by measuring some key data in the multi-strand simulations. Several compression steps are visible: from the initial state (O), a zero-plasticity low pressure pre-compaction stage (I) establishes contact between neighbouring strands (II). Then the real compaction is activated, where the strands contact increases and shapes become hexagonal. Total compaction (no porosity) is reached at high pressure (IV). Beyond this limit, the material acts as bulk copper. The compaction law validity can be verified by benchmarking it to the 2D multi-strands simulations, where the cable is replaced by a single disk (see **Figure 5b**). Both methods show a good accordance in terms of compaction level. The equivalent law successfully reproduces the compressive wave and the over-pressure at the cable centre. Moreover, its calculation cost is about ten times lower than the multi-strand model.

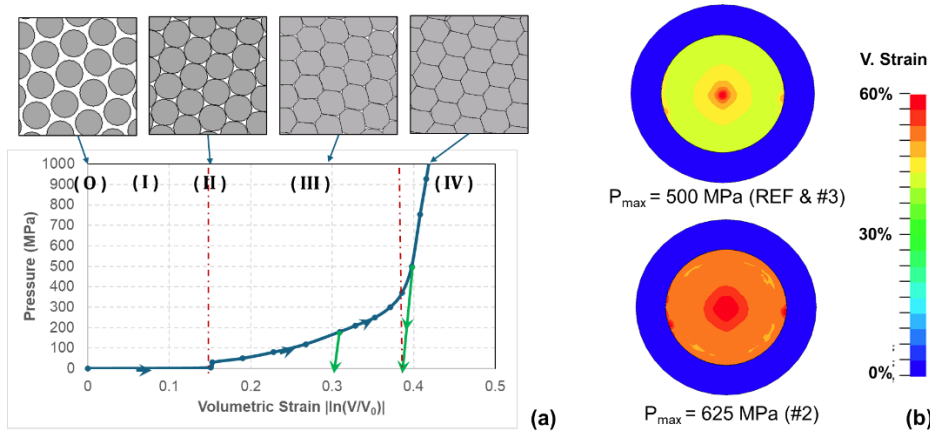


Figure 5: Transverse pressure compaction law for a 120mm² Cu cable (a); Equivalent compaction for the designed configurations (b)

3 Experimental Results & Comparison to Modelling

The three configurations determined in §2 were tested experimentally, with 20 samples per case. In addition, traditional mechanical crimping was performed on the same assembly configuration. The MPC samples can serve as a reference database to verify the validity of the simulation model.

A first point of comparison can be established through a measurement of the assembly geometry after the process. 3D scans taken for each configuration (**Figure 6a**) were used to extract the evolution of the terminal cross-sections, shown on **Figure 6b**. Between the reference case and case #2, the minimum compression ratio, obtained at the centre of the active zone, is similar. Case #3 exhibits a higher level of compaction.

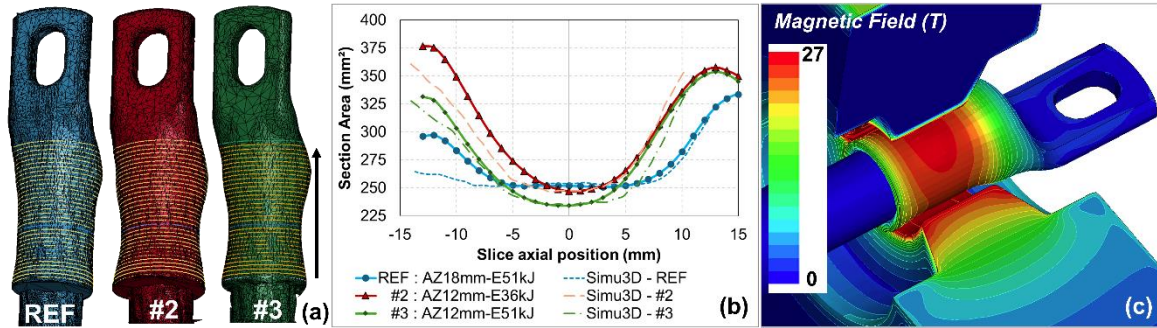


Figure 6: 3D scans of experimental MPC samples (a) – Samples assembly areas and model refits along the cable axis (b) – 3D MPC simulation at peak pressure ($t=17.5 \mu s$) for the Reference Case ($AZ=18 \text{ mm}$ $E_0=51 \text{ kJ}$) – colormap = magnetic field (c)

3D MHD simulations of the MPC process were carried out for each configuration, using the RLC circuit corresponding to the generator as input. The inductor materials, made of CuCrZr alloy, were modelled using a viscoplastic behaviour law characterised in-house (N. Mrozowski, et al., 2025). The cable is represented by the equivalent law established in §2.3. Numerical profiles show good overall agreement. However, there are discrepancies in geometry, with flatter profiles than those observed experimentally, particularly at the lug tips. This may be attributed to the constitutive behaviour, modelled with a bilinear elastic-plastic law, which may be insufficient to properly reproduce local deformations.

In addition to the 3D scans, observations of the minimum cross-section (slice at position 0 on **Figure 6b**) for each case allow the compaction of the cable to be assessed by measuring the component areas (see **Figure 7**). The reference case and case #2 show similar levels of compaction, while case #3 is more compact, with almost no porosity, reaching a compaction rate of approximately twice that of the reference cross-section (120 mm^2). All three configurations are close to the levels predicted by the 2D multi-strand model.

The same analysis was carried out on a mechanically crimped cable with a compaction rate similar to the reference case. In this case, compaction is seen to be inhomogeneous, concentrated mainly in the central region with highly deformed strands, leaving numerous inter-strand voids elsewhere, which may adversely affect joint performance.

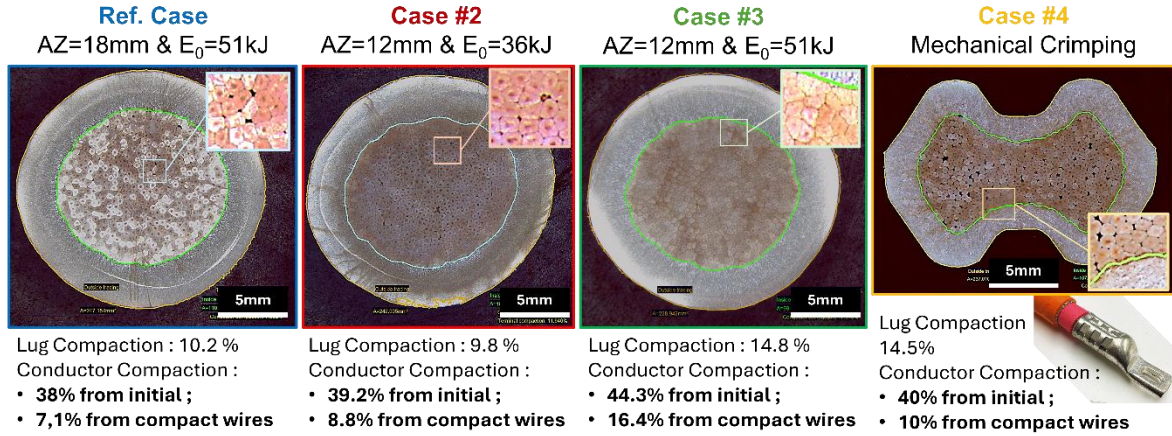


Figure 7: MPC and mechanical crimping samples optical micrographs at the minimum compacted section

4 Assemblies' Electrical performances

All crimped cables, 20 per configuration, are then tested to determine the electrical resistance of the interface. Typically, resistances of the order of 10 $\mu\Omega$ or less are sought in order to optimize link heating. Initially, the measurement, presented on the **Figure 8** is carried out for a fresh crimp. Minimal resistances with good repeatability, approx. 5 $\mu\Omega$, can be observed for the reference case maximizing the contact area, as well as the iso-energy case #3, with less crimped area but more compaction. Case #2, on the other hand, although iso-compacted with the reference case, shows a much higher (12 $\mu\Omega$) and dispersed resistance. This is probably due to a reduced contact surface compared with the reference case. The same applies to a lesser extent to mechanical crimping (7 $\mu\Omega$). In this case, the reduced effective contact surface is located in the grooves created by the crimping tool.

Next, the samples are subjected to two cycling tests, commonly used to qualify on-board electrical connections:

1. Purely thermal, with a -40/140 °C amplitude, on a 15 min cycle period, over 500 cycles.
2. Slow thermal cycling in an environment with 95% relative humidity, with a range of 25/55 °C, 10 cycles of 24 hours each.

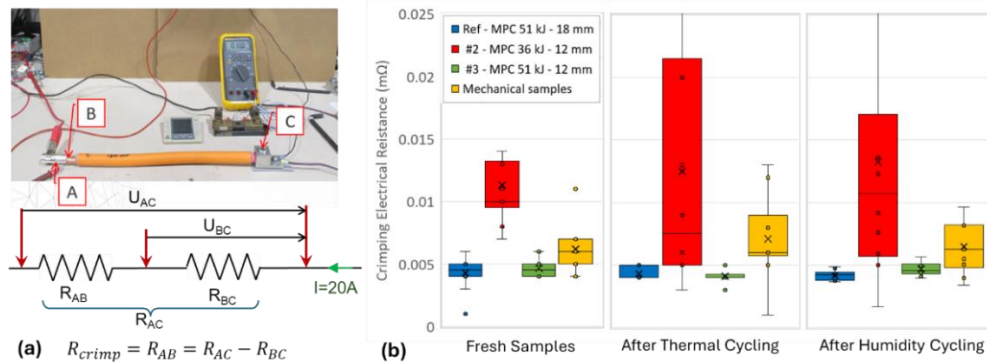


Figure 8: Interface electrical resistance measurement principle (a); Mechanical and MPC electrical resistance for fresh, and after thermal and humidity cycling (b)

The assembly resistance is measured after each test. It can be noted that the Reference and Case #3 maintain stable and excellent performance. Similar trends are observed for mechanical crimping, which retains a higher resistance and variability than the best MPC cases. MPC Case #2 resistance is significantly degraded and shows a larger variance after thermal cycling, probably caused by an insufficient contact zone versus compaction ratio. This may lead to contact loosening during thermal cycling, which exaggerates the observed dispersion. Humidity cycling does not affect the results observed. One reason that could explain this lack of variance is that these tests are not demanding enough to produce ageing that would significantly degrade assembly performance. Further investigations are required to verify whether they are sufficiently representative of actual on-board conditions.

5 Conclusions

This study investigated the crimping of power cables used for electric mobility, from both a numerical and experimental point of view. Simulations were used to better understand the MPC process and model the behaviour of a cable made up of a bundle of strands. This led to the identification of an equivalent orthotropic compaction law, which considerably reduces calculation times, making it suitable for 3D analysis. 3D MPC simulations were compared with experimental results with a good accordance overall. The performance of crimped assemblies was also characterized before and after cyclic testing. It was shown it is possible to obtain low-resistance assemblies with a good repeatability. MPC has been compared with mechanical crimping, and shows a good relative resistance decrease in the MPC best cases, by 25% on average. Moreover, it was demonstrated it was possible to reach similar performance at equal energy by reducing the crimped width from 18 to 12 mm and locally increasing the compaction state. This result can serve as basis to move towards optimised compact crimps, that can be lightened by 10% in order to reduce their cost. Some future investigations could follow the same direction, by working on the lug thickness.

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Feasibility Study on Joining Aluminum Tube with Alumina Ceramic Rod Using Magnetic Pulse Crimping Process

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Abstract

Ceramic-metal joints play a vital role in structural applications, offering benefits such as strength, lightweight, high strength-to-weight ratio, corrosion resistance, and high-temperature performance. These joints are widely used in fields ranging from electronics and biomedical devices to aerospace and automotive industries. However, joining ceramics and metals is challenging due to differences in atomic bonding and thermal expansion properties. Traditional methods, such as brazing, friction welding, and fusion welding, often induce thermal stresses and residual strain, leading to defects like cracks and intermetallic formation at the joint interface. Magnetic Pulse Crimping (MPC) is an environmentally friendly, quick, and cold solid-state material joining process that does not create heat-affected zones and mitigates these issues. This study uses the MPC process to investigate the feasibility of joining an AA 1050 aluminum alloy tube with an alumina ceramic rod. The experiments were performed at various discharge energies using an Archimedean spiral coil and a step taper field shaper to evaluate the joint's strength and metal-ceramic interface behavior. Non-sinusoidal waviness patterns have been observed at the metal-ceramic joint interface. The numerical analysis used the LS DYNA EM module to measure the current density, magnetic field, Von Mises stresses, and impact velocity on the flyer tube. The manuscript contributes to the understanding of ceramic-metal joining under electromagnetic forming conditions.

Keywords

Magnetic Pulse Crimping, Archimedean spiral coil, AA 1050, Alumina, Ceramic-metal joints.

1 Introduction

Ceramics and metal joints are becoming crucial in structural applications due to their functional requirements, such as strength, lightweight, high strength-to-weight ratio, corrosion resistance, and high-temperature performance (Lemus-Ruiz et al. 2011). These joints are used widely in vacuum tubes, electronics, insulators, sapphire-metal windows, biomedical devices, and rocket igniter bodies (Loehman and Loehman 2009). Ceramic components of an automobile engine are joined with metal using the brazing technique for strength, resistance to fatigue, creep, and oxidation. High pressure and temperature bond ceramic with metal, and a brittle intermetallic is formed at the interface due to differential cooling rate and mechanical response (Jadoon 2004; Palit and Meier 2006). Joining ceramics with metals is challenging due to differences in atomic bond configurations and properties, as ceramics have ionic and covalent bonds with stable peripheral electrons, making them resistant to fusion welding-assisted joining methods (Zhou 2008).

Diffusion bonding is one of the popular solid-state joining techniques used to join metal with ceramic by applying pressure at elevated temperatures and controlling the interface's microstructure. This method provides joint strength equal to the base metal with reduced defects, reduced cracking, and relaxed thermal residual stress, but it takes more joining time and high equipment due to vacuum conditions (Akselsen 1992). The difference in thermo-mechanical properties of aluminum alloy and ceramic due to their metal-ceramic nature makes the MPC more favorable, as it is a cold metal-forming process (Yao et al., 2023).

This manuscript presents a feasibility study on joining AA 1050 with ceramic rods using a magnetic pulse crimping process. Experiments were performed at various discharge energies using an Archimedean spiral (AS) coil and a step taper field shaper. A cross-sectional study of the sample has been conducted to understand the metal-ceramic joint interface.

2 Experimental procedure:

Joining was performed between a 1 mm-thick AA 1050 flyer tube and an 8 mm diameter alumina target rod with a 1 mm standoff distance for optimal results. To improve the bonding and joining efficacy, the contacting surface of the flyer and target was made coarse using 80-grit size polishing paper by increasing the surface energy.

Experiments were performed using a magnetic pulse forming machine with a maximum discharge energy capacity of 10 kJ, maximum voltage of 15 kV, capacitance of 100 μ F, circuit resistance of 22 m Ω , and circuit inductance of 2 μ H. In the Archimedean spiral coil, the radial distance between the consecutive turns remains constant, which provides strength and durability to the forming coil. The AA 1050 flyer tube was placed over the alumina rod in the field shaper and the primary coil vicinity. The flyer tube and target rod were fixed from the non-crimping ends to remain equidistant, maintaining a 1 mm constant standoff distance. Fig. 1 (a) and (b) show the arrangement's isometric and front-half sectional views. Fig. 1. (c) and (d) show the detailed coil and field shaper dimensions. Experiments were

Table 1. Johnson-Cook parameters for AA1050 tube and copper.

Ceramics subjected to impact load respond according to the flyer material, contact area, velocity, and mechanical characteristics (Cronin et al. 2003). Johnson-Holmquist's (JH-2) ceramic constitutive model for alumina describes brittle materials' response under large deformations, as mentioned in Eq. 3. The effective normalized strength (σ^*) combines intact (σ_i^*) and fractured (σ_f^*) material states via a damage parameter (D) :

$$\sigma^* = \sigma_i^* - D(\sigma_i^* - \sigma_f^*) \quad (3)$$

Intact strength (σ_i^*):

$$\sigma_i^* = A(P^* + T^*)^N(1 + C \ln \dot{\epsilon}^*) \quad (4)$$

Fractured strength (σ_f^*) :

$$\sigma_f^* = B(P^*)^M(1 + C \ln \dot{\epsilon}^*) \quad (5)$$

Where $P^* = P/P_{\text{HEL}}$ (pressure normalized by Hugoniot Elastic Limit pressure), and $\dot{\epsilon}^* = \dot{\epsilon}/\dot{\epsilon}_0$ (strain rate normalized by a reference rate).

Further, the equation of state (Pressure-Volume relationship):

$$P = K_1\mu + K_2\mu^2 + K_3\mu^3 + \Delta P \text{ (for compression, } \mu = \rho/\rho_0 - 1 \geq 0) \quad (6)$$

Where, K_1, K_2, K_3 is material-specific bulk modulus coefficients, and ΔP is the bulking pressure due to damage.

The Johnson-Holmquist ceramic constitutive model for alumina is mentioned in Table 2.

Strength constants	A	B	C	M	N
	0.93	0.31	0	0.60	0.60
	EPSI	Tensile strength (GPa)	SFMAX	HEL (GPa)	PHEL (GPa)
	1.00	0.2	0.0	2.79	1.46
Damage constants	D1	D2			
	0.005	1.0			
Equation of state	K1 (GPa)	K2 (GPa)	K3 (GPa)	BETA	
	130.95	0.0	0.0	1.0	

Table 2. Johnson- Holmquist ceramic constitutive model for Alumina.

Hexahedron elements provide accuracy, stability in handling extreme deformations, and better convergence behavior in nonlinear simulations. The mesh size was selected for optimal accuracy and cost of the solution. The total number of elements for the AS coil, step-

tapered FS, flyer tube, and target rod was 40950, 18876, 1392, and 1627, respectively. The mesh size of all parts and the element selected for the study is shown in Fig. 2.

The coil was selected as a rigid body, whereas J-C plasticity material properties were assigned to the FS and flyer to study the deformation, as in equation 1. Johnson-Holmquist (J-H) ceramic constitutive models have been selected to illustrate the response of brittle alumina under significant deformations.

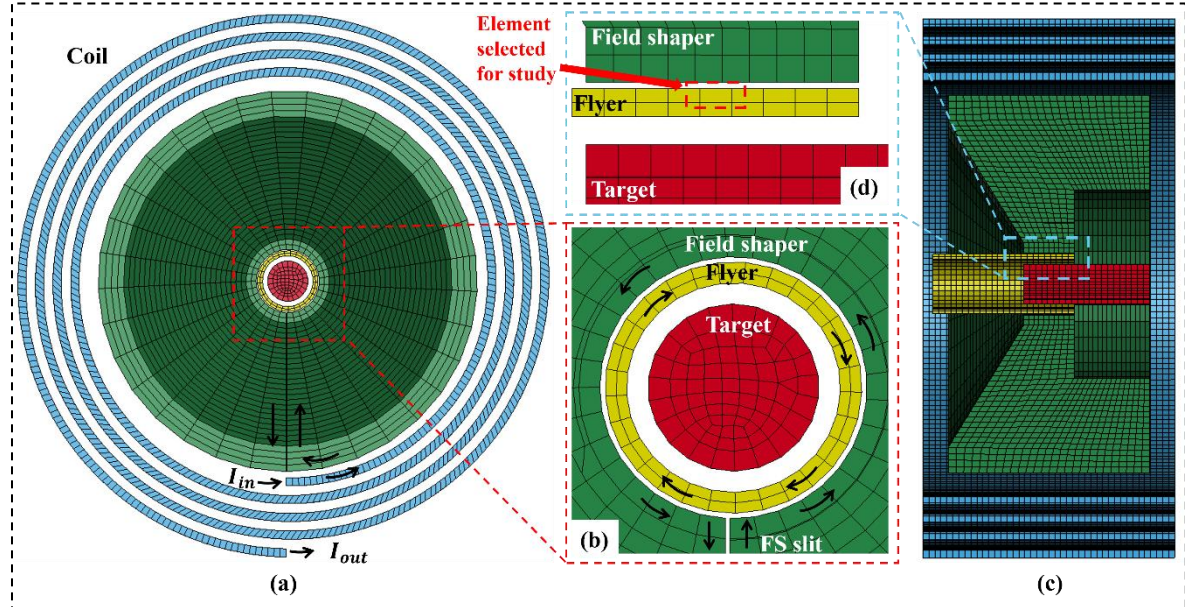


Figure 2. Meshing on the parts and element selection for the study.

4 Results and discussion:

Figure 3 (a) depicts the radial view and (b) longitudinal view along with fractured parts of the ceramic rod of the metal-ceramic samples. From the experiments, it can be concluded that the proper joining was achieved at 4 kJ energy. Below 4 kJ, the AA 1050 tube had insufficient deformation and an improper joint with the rod. For more than 5 kJ, a brittle fracture of the ceramic rod at the non-crimping zone and crushing was observed. The joint quality was assessed using visual inspection and mechanical integrity of the sample. Samples with no visible joint were considered poor, whereas samples with localized interface contact without complete continuity were considered improper, and the defect-free sample with proper contact in the crimping region was termed a proper joint. Improper joining was evident at an energy less than 4 kJ, the proper joint was achieved at 4 kJ, and the brittle fracture of the ceramic rod was observed for discharge energy of more than 5 kJ.

On the other hand, alumina crushing was prevalent at 6 kJ discharge energy. The observations are mentioned in Table 3. The deformation of the AA 1050 tube was assessed based on post-test radial and longitudinal measurements and visual inspection of the flyer tube geometry compared to its original dimensions.

S. No	Discharge energies (kJ)	Discharge voltage (kV)	Effect on the sample
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1	1.0	4.50	No joining at the interface
2	2.0	6.45	Improper joint at the interface
3	3.0	7.80	Improper joint at the interface
4	4.0	9.00	Proper joint at the interface
5	5.0	10.05	Fracture of alumina rod
6	6.0	10.95	Crushing of alumina rod

Table 3. Discharge energy optimization for joining.

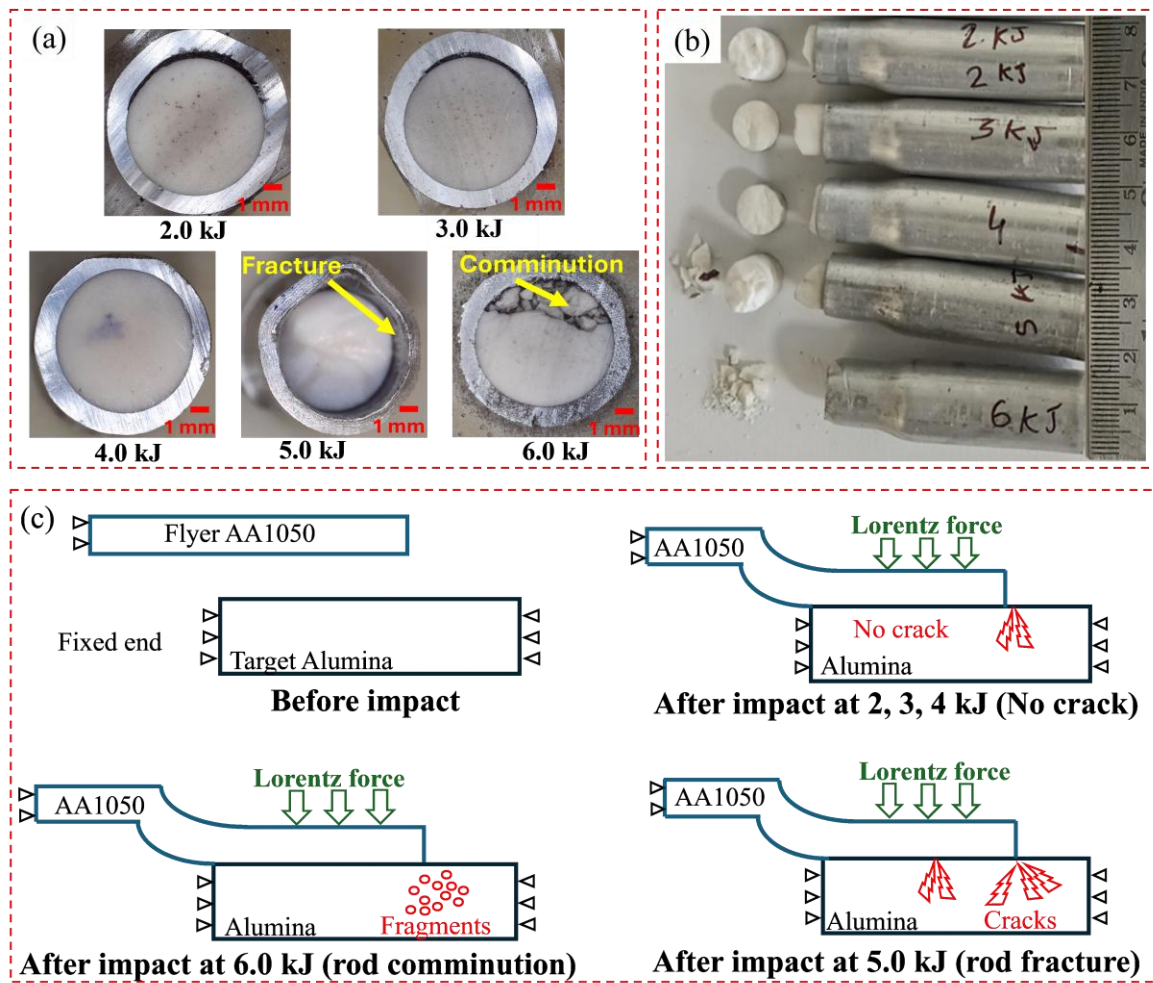


Figure 3. (a) Radial and (b) longitudinal view of joined ceramic metal samples, and (c) fracture and fragment generation in alumina rod.

During the MPC, the rapidly induced Lorentz forces cause a high-velocity impact of the flyer on the target, generating a compressive stress wave at the ductile-brittle joint interface. The geometry of the crimping joint, particularly near the crimping region, causes abrupt material transition, and edge discontinuities lead to localized stress amplification. At energy levels higher than 4.0 kJ, the compressive stress wave surpasses the local dynamic strength of the target, initiating crack formation. At higher energies of 5 and 6 kJ, the resulting fracture front propagates at the elastic wave speed of the alumina. It forms a conoid of comminuted or pulverized alumina at the impact location. When the compressive stress wave reaches the

free surface, it is reflected as a tensile wave, causing spallation or tensile cracking (Kaufmann et al. 2003).

4.1 Cross-sectional analysis:

After choosing the suitable discharge energy by trial and error, the remaining experiments were performed at 4 kJ to investigate the metal-ceramic joint interface. Fig. 4 (a) shows a cross-sectional view of a ceramic metal joint at 4 kJ, where the uniform joining of the AA1050 flyer with an alumina target is visible. Fig. 4 (b) shows the metal penetration into the ceramic with a 71-micron width, and Figs. 4 (c) and (d) show the same ceramic metal joint interface at 200X and 500X magnification, depicting irregular wavy interaction. The material hardness difference causes this irregular, non-sinusoidal wavy pattern at the interface.

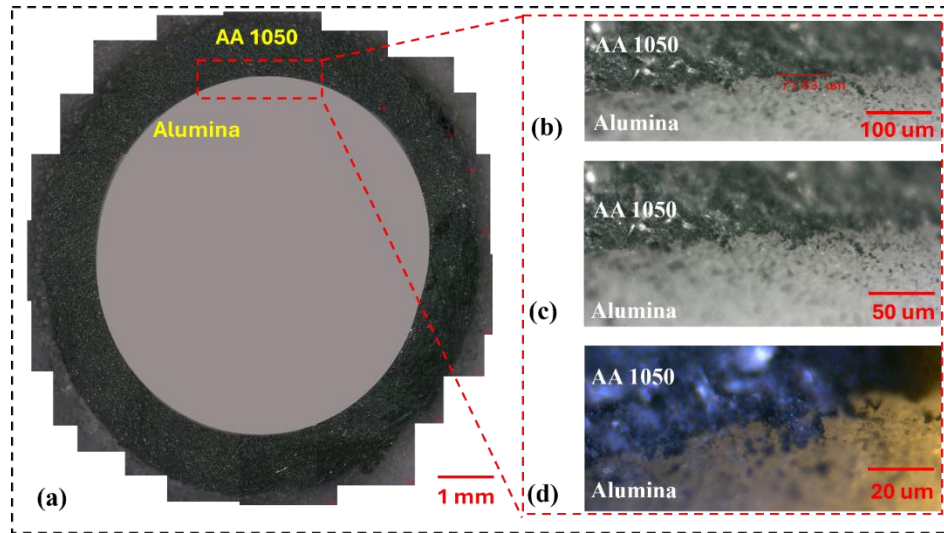


Figure 4. (a) Cross-sectional view of the ceramic metal joint at 4 kJ, (b) Interface at 100X, (c) 200X, and (d) 500X magnification.

A complex stress field develops when the aluminum flyer impacts the ceramic at high velocity. The brittle ceramic accumulates damage when local stresses exceed its dynamic strength. The compressive-induced damage and bulking (volume increase due to fracturing) cause surface distortion and irregularities near the impact zone. This transition between compressive and tensile states and evolving damage at the crimping interface generates uneven localized expansion. Additionally, fractured ceramic material tends to bulk and lose structural coherence. When confined or partially constrained by the aluminum flyer, this expansion may not occur uniformly, especially along edges and corners where geometric effects amplify stress. This asymmetry results in the wavy interface as the damaged ceramic attempts to accommodate internal stress redistribution. Therefore, irregular waviness is the physical expression of stress wave interactions, localized damage accumulation, and inhomogeneous bulking, as discussed by (Reijer 1991).

5 Simulation Results: current density, magnetic field, Von Mises stress, and Impact velocity on the flyer tube

From the experimental analysis, the samples joined at 4 kJ gave an optimal interface. Further simulations were performed at 4 kJ discharge energy to understand the current density, magnetic field, Lorentz force, and impact velocity on the flyer tube during crimping. The element on the outer flyer tube surface, located opposite the FS slit, has been selected to study the behavior of the flyer tube. Values of current density taken during the impact at 29 μs are 13.44 kA/mm², and at 41 μs corresponding to the highest value of discharge current are 9.96 kA/mm², as mentioned in Fig. 5 (a) and (b) Whereas, Fig. 5 (c) and (d) shows the magnetic field density is 10.97 Tesla and 10.81 Tesla at 29 μs and 41 μs , respectively.

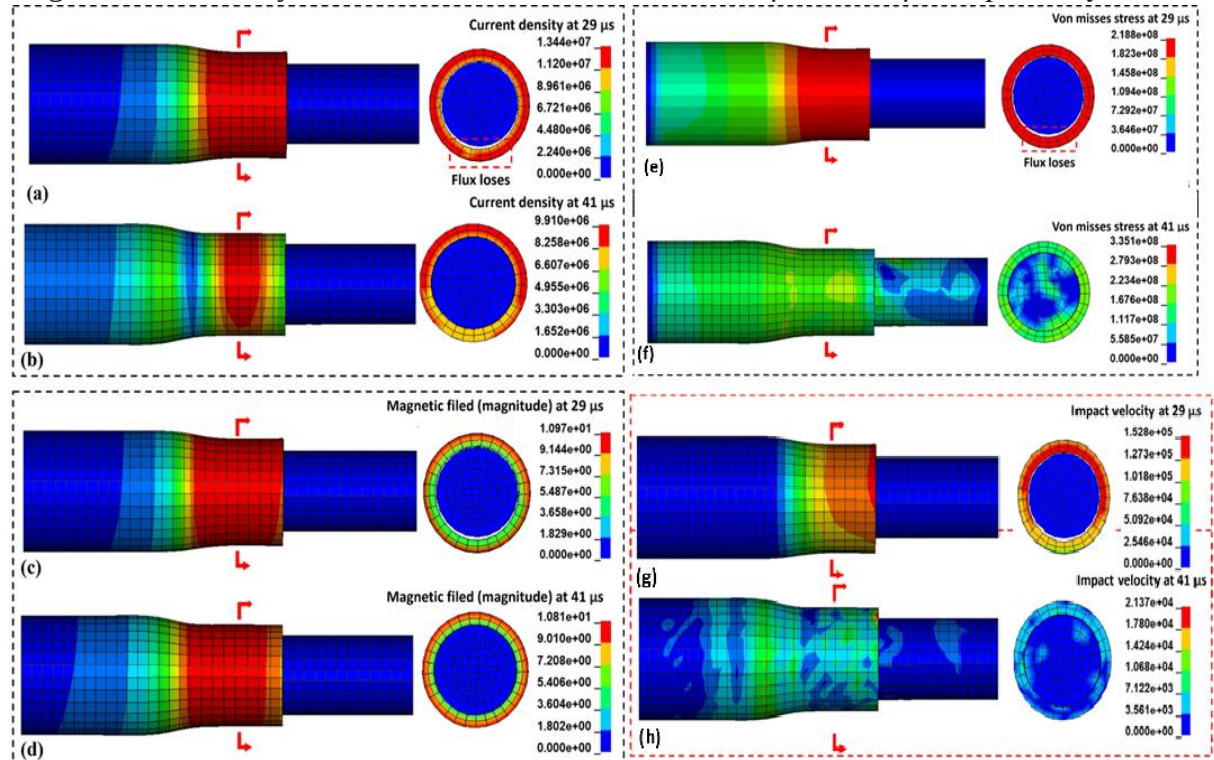


Figure 5. Contours for current density on the flyer at (a) 29 μs , (b) 41 μs , and magnetic field at (c) 29 μs , (d) 41 μs and Von Mises stresses on the flyer at (e) 29 μs , and (f) 41 μs , and Impact velocity at (g) 29 μs , and (h) 41 μs .

The current density and magnetic field are maximum in the center of the crimping zone and decrease with distance. Uneven deformation and flux losses take place due to the FS slit area, which is shown in the inset view of Fig 5. (a). The Lorentz force induces a complex stress state in the flyer tube. Von Mises stress calculations indicate that plastic deformation initiates at 219 MPa after 29 μs , as shown in Fig. 5(e). A subsequent reduction in stress is observed during the peak magnetic field, as seen in Fig. 5(f). The highest flyer impact velocity of 152.8 m/s was observed before the impact at 29 μs . In contrast, the sudden drop in velocity occurs after the impact and becomes negligible, as mentioned in Fig. 5 (g) and (h).

6 Conclusions:

This study investigates the feasibility of joining AA 1050 aluminum with a ceramic rod using magnetic pulse crimping. The experiments were performed at various discharge energies from 1.0 kJ to 6.0 kJ, using an Archimedean spiral coil and a step taper field shaper to evaluate the joint's strength and interface behavior. The optimum cross-section was achieved at 4 kJ discharge energy. A non-sinusoidal wavy pattern with 71-micron thickness has been observed at the joint interface for the 4 kJ sample. The fracture of the alumina rod has been observed due to compressive stresses and high-velocity impacts, which have been mitigated by optimizing the discharge energy.

Numerical analysis of parameters such as current density, magnetic field, Lorentz force, and impact velocity indicates that the most significant deformation occurs at the flyer tube side because the FS concentrates and amplifies the magnetic field. The flyer tube's highest current density, magnetic field, Von Mises, and impact velocity values are 13.44 kA/mm², 10.9 T, 210 MPa, and 152 m/s, respectively. The results contribute to the understanding of ceramic-metal joining under electromagnetic forming conditions.

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Chapter 5

Adiabatic blanking

Shear Cutting of Aluminum, Zinc, and Tin at Increased Speeds

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Abstract

The temperature and temperature evaluation during shear cutting, particularly at increased speeds, is crucial for understanding the material behavior during cutting, validating thermo-mechanical material models used in the finite element method, and predicting the tool wear. In this study, we measured the punch force and temperature for EN AW 6082, zinc, and tin for punch speeds 0.12, 0.2, 0.4, and 0.6 m/s using a cutting clearance of 0.075 mm. The temperature is evaluated based on the thermoelectric voltages arising from the sheet metal and tool material based on the Seebeck effect. The results demonstrate that the sheet metal strongly influences the thermal response during shear cutting. Higher temperature maxima were generally observed to occur later than the force maximum, likely due to additional frictional heating beyond plastic deformation.

Keywords

Blanking, Temperature, Measurement

1 Introduction

Understanding the temperature during shear cutting, particularly at increased speeds, is crucial for several reasons. Firstly, it is essential for the validation of thermo-mechanical material models used in finite element method (FEM) simulations. Accurate temperature data ensures that these models can reliably predict the behavior of materials under shear cutting, leading to more precise simulations. Secondly, observing the thermal effects within the material during shear cutting provides valuable insights into how the material responds to increased temperatures. This knowledge is critical for optimizing cutting parameters and improving the quality of the cut. Additionally, assessing the thermally induced mechanical

stresses on the tool surface is vital for predicting tool wear and extending tool life. High temperatures can increase mechanical stresses, which may cause premature tool failure. Furthermore, elevated temperatures can enhance adhesion behavior (Tröber et al. 2021).

2 State of the Art

Various methodologies have been employed to determine the temperature within the contact zone between sheet metal and active elements during shear cutting. Most used are experimental and numerical methods. Experimental studies frequently utilized thermocouples embedded in punches or cutting plates (Chen et al. 2003) or connected to the sheet metal (Weiser et al. 2022) to measure temperatures. To measure the temperature right after the shearing of the part infrared (IR) cameras can be used (Schweinshaupt et al. 2024). However, this method does not give insight into the temperature evolution during the process. Furthermore, several investigations have focused on numerically modeling the temperature rise during shear cutting (Karantza et al. 2024; Galiev et al. 2023).

The thermoelectric phenomena provide the possibility of using the sheet metal and tool as a thermocouple based on different thermoelectric properties. Thermoelectric voltages are generated when two different electrical conductors are joined at one end, which has a different temperature than the other. The thermoelectric voltage is directly proportional to the temperature difference along the conductors. The proportional factor, known as the Seebeck coefficient, characterizes the thermoelectric properties of the materials (Demmel et al. 2015; Demmel 2014). This principle has been utilized to measure the temperature in forming for different materials and parameters. Demmel et al. (2015) measured a local temperature maximum of around 70 °C at material separation in shear cutting of hot rolled fine-grained steel S355MC with a thickness of 6.0 mm, a cutting clearance of 0.48 mm, and a 0.01 m/s punch speed. For the same material with a thickness of 4 mm and increased speeds up to 0.27 m/s and 0.04 mm clearance, Schrepfer, A., Welm, M. et al. (2023) measured 199 °C at the material separation. In cup drawing of the aluminum alloy EN AW 5083 with a thickness of 1 mm and a 1.3 mm drawing gap, Tröber et al. (2021) measured 38 °C at 10 mm drawing depth using the thermoelectric principle.

In this study, we measured the punch force and temperature based on thermoelectric voltages with a tool-workpiece-thermocouple for 1.5 mm thick aluminum EN AW 6082, zinc, and tin for increased punch speeds up to 0.6 m/s using an increased cutting clearance of 0.1 mm.

3 Experimental Setup

The experimental setup for the stamping experiments utilizes a Bruderer BSTA 1600-181 press (Frasnacht, Switzerland) with a stroke length of 68 mm. Stamping commenced 5 mm before the bottom dead center. Both the punch and die, as well as the sheet metal, are electrically isolated from the tool body. The stamping geometry is a coin with a 20 mm diameter, featuring a 0.075 mm cutting clearance (5 % relative clearance) and cutting edge

radii of 2 μm and 30 μm for the punch and die, respectively. The velocity was varied by adjusting the number of strokes on the press, with the punch speed determined based on the manufacturer's stroke curve at the moment of punch and sheet contact. Single stroke testing was conducted at velocities of 0.12 m/s (60 strokes/min), 0.2 m/s (100 strokes/min), 0.4 m/s (200 strokes/min), and 0.6 m/s (300 strokes/min).

The measurement technology setup for the stamping experiments involves measuring force, distance, and thermoelectric voltages. Force measurements were conducted using a Kistler (Winterthur, Switzerland) 9041C load washer with quartz sensing elements mounted in a force shunt directly above the punch. The force signals were amplified using a Kistler 5073A industrial charge amplifiers, which outputs an analog voltage signal. Distance measurements to determine the location of the bottom dead center were performed using a laser triangulation displacement sensor, optoNCDT1320-100 from Micro-Epsilon (Ortenburg, Germany) with an analog voltage output. Thermoelectric voltages were pre-amplified and filtered close to the tool using self-build precision instrumentation amplifier converting signals from the -10 to 10 mV range to a -10 to 10 V range (Demmel 2014). Data acquisition was done time-synchronously using a PCI-6280 and a PCI-6250, recording at 200 kHz with LabVIEW software, all from National Instruments (Austin, Texas, USA). Ambient temperature, used as a reference for calculating temperature differences from thermoelectric voltages, was measured with a Papouch TMU temperature sensor featuring a USB interface and a resolution of 0.1 °C by Papouch s.r.o. (Prague, Czech Republic).

4 Materials

This study utilized three materials for stamping experiments: aged aluminum EN AW 6082, 99.76 % pure zinc, and a tin alloy Sn95Sb4Cu1 with a tin content of 93.5 %. All materials are used in sheet metal form with a thickness of 1.5 mm.

The sheet metal is mechanically characterized using a quasi-static tensile test according to DIN EN ISO 6892-1. The results for the yield stress R_e and the ultimate tensile strength R_m for the materials are given in Table 1.

The material-specific Seebeck coefficient is required to convert the thermoelectric voltages measured between the tool and the sheet metal into a temperature difference. All sheet metal material and the tooling material steel 1.3343 were characterized using a special test rig developed at the Chair of Metal Forming and Casting. The sample is connected to two platinum wires, with one side maintained at 0°C while the other side is heated up to a maximum temperature (500°C for aluminum, 338°C for zinc, and 180°C for tin) and then cooled down. Platinum wires connect both sample ends to a high-precision voltmeter with a maximum deviation of 1.5 % in the temperature range. Additionally, the platinum serves as a reference material for the thermoelectric characterization and allows the calculation of the Seebeck coefficient from the measured thermoelectric voltage. Figure 1 illustrates the result of the thermoelectric characterization. The difference in relative thermoelectric voltage between the tool material 1.3343 and the sheet metal material relates to the temperature on the connecting surface. (Demmel 2014)

Material	R _e in MPa	R _m in MPa
EN AW 6082	173	335
Zinc	84	154
Tin	42	47

Table 1: Results of quasi-static tensile testing according to DIN EN ISO 6892-1

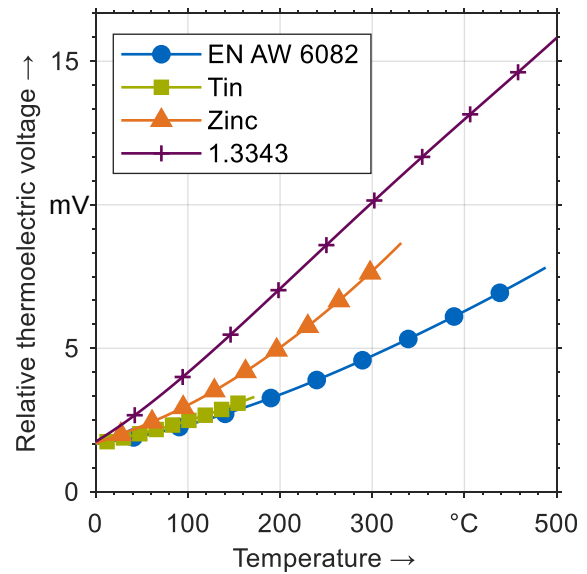


Figure 1: Results of thermoelectric characterization.

5 Results

Five valid samples were stamped per material and punch speed. The maximum force was evaluated for every specimen and averaged. The measured thermoelectric voltages were filtered with a low-pass filter of 500 Hz, converted to temperature differences based on the thermoelectric characterization, and added to the ambient temperature. The temperature maximum was calculated for all five samples and averaged. These results and the corresponding standard deviation are given in Table 2 for EN AW 6082, in Table 3 for zinc, and in Table 4 for tin.

Averaged data curves are obtained by interpolating the single stroke data sets over the longest data set and averaged for every punch travel value. The data is plotted against the manufacturer's stroke curve, with 0 mm being the bottom dead center measured by the distance sensor. Please note that these curves show different maximum values, as small distance offsets lead to different averaging than averaging over maximum values, regardless of the punch position. The temperature curve at different punch speeds is shown in Figure 2 for aluminum only.

The punch travel between the occurrence of the force maximum and temperature maximum (ΔS) and the maximum temperature measured at different punch speeds for EN AW 6082, zinc and tin is plotted in Figure 3. A positive value of ΔS indicates by how much the punch has moved from the point of maximum force to reaching the maximum temperature.

Punch speed	Max. Force in kN	Standard deviation in kN	Max. Temperature in °C	Standard deviation in °C
0.12	14.35	0.060	44.47	4.05
0.2	14.26	0.114	57.54	4.63
0.4	14.45	0.032	80.97	3.06
0.6	15.09	0.102	70.22	3.39

Table 2: Resulting maximum force and maximum temperature, averaged over five samples, of single stroke testing of EN AW 6082.

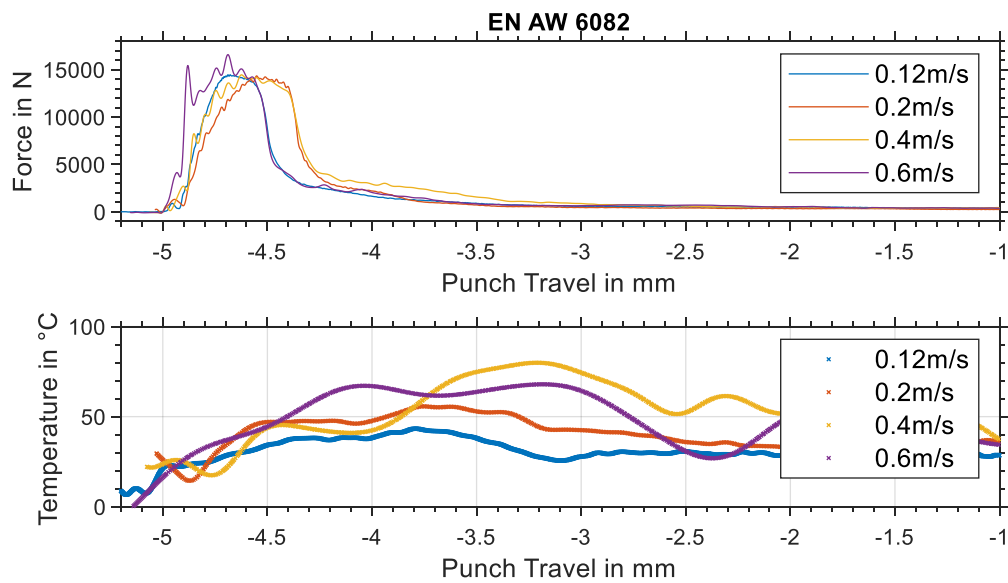


Figure 2: Averaged force and temperature curves plotted against the stroke curve at different punch speeds for EN AW 6082.

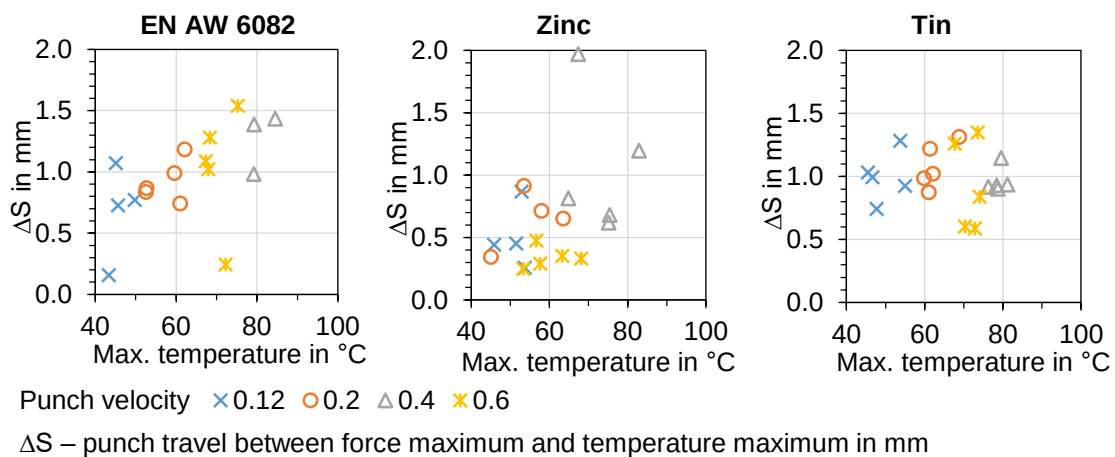


Figure 3: Punch travel between the occurrence of the force maximum and temperature maximum (ΔS) and the maximum temperature measured at different punch speeds for EN AW 6082, zinc and tin.

Punch speed	Max. Force in kN	Standard deviation in kN	Max. Temperature in °C	Standard deviation in °C
0.12	16.24	0.162	50.48	3.23
0.2	16.71	0.069	54.84	6.78
0.4	16.64	0.153	73.10	7.18
0.6	17.66	0.223	59.77	5.88

Table 3: Resulting maximum force and maximum temperature, averaged over five samples, of single stroke testing of zinc.

Punch speed	Max. Force in kN	Standard deviation in kN	Max. Temperature in °C	Standard deviation in °C
0.12	5.34	0.056	49.74	4.33
0.2	5.62	0.028	62.68	3.54
0.4	5.94	0.100	78.86	1.79
0.6	6.49	0.124	71.72	2.64

Table 4: Resulting maximum force and maximum temperature, averaged over five samples, of single stroke testing of tin.

6 Discussion

The measurement of temperature based on thermoelectric voltages shows promising results. At increased punch speeds it is not possible to differentiate blanking stages (e.g. plastic deformation) from the temperature data as shown in earlier publications with punch speeds up to 0.27 m/s (Demmel et al. 2015; Tröber et al. 2021; Schrepfer, A., Welm, M. et al. 2023). Reasons can be found in the increased punch speed, the increased cutting clearance, the reduce sheet metal thickness as well as the punch material used. To enhance thermoelectric voltage generation and thereby improve the stability of temperature measurements, the punch material CF-H40S, characterized by a negative Seebeck coefficient, could be used.

The type of sheet metal utilized in shear cutting influences the temperature data measured. The following findings were derived from the experimental data. Zinc demonstrated the highest standard deviation in temperature maxima. A reason can be the smaller difference in relative thermoelectric voltage of zinc to the punch material 1.3343. Small voltages are more prone to interference and more complex to amplify. Tin showed the most consistent temperature data, with the occurrence of the temperature maximum remaining stable across all tested punch speeds. The temperature maxima consistently appeared within a punch displacement of $\Delta S = 0.5 - 1.5$ mm following the force maximum.

Generally, higher temperature maxima were associated with a later occurrence relative to the force maximum (i.e., larger punch travel ΔS). This delay is likely attributable to increased frictional heating between the punch and the sheet metal, in addition to the heat generated by plastic deformation during shear cutting. Contrary to expectations, higher

temperatures were recorded at a punch speed of 0.4 m/s rather than at 0.6 m/s. This anomaly may be due to limitations in the amplifier's capacity to accurately process the heat signals generated at higher speeds or due process instabilities leading to varying friction between punch and sheet metal.

7 Summary and conclusion

In this study, we measured the punch force and temperature based on thermoelectric voltages for EN AW 6082, zinc, and tin for punch speeds 0.12, 0.2, 0.4, and 0.6 m/s using a cutting clearance of 0.075 mm.

The results indicate that the type of sheet metal significantly affects thermal behavior during shear cutting. Aluminum exhibited the highest average temperature maxima, while zinc showed the greatest variability, likely due to its low thermoelectric voltage differential with the punch material (1.3343). Tin demonstrated the most consistent thermal response across all punch speeds.

A general trend was observed where higher temperature maxima were associated with a delayed occurrence relative to the force maximum, attributed to frictional heating in addition to plastic deformation. Unexpectedly, the highest temperatures were recorded at 0.4 m/s rather than 0.6 m/s. These findings suggest that thermoelectric voltage-based temperature measurement is a viable method for analyzing thermal effects in high-speed shear cutting, though improvements in signal amplification and material selection (e.g., using CF-H40S with a negative Seebeck coefficient) may enhance measurement reliability at higher punch speeds.

Acknowledgments

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Investigations on the temperature-deformation behaviour of advanced high-strength steel sheets (AHSS) under shear loading from nearly isothermal to the adiabatic state

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Abstract

Steel sheet materials tend to form adiabatic shear bands at high strain rates, leading to material separation. This phenomenon is used in adiabatic blanking processes, but it should be avoided in crash safety applications. This paper deals with investigations on the main influence parameters on adiabatic heating of advanced high strength steel sheets (AHSS) under shear loading. Focussing on the effect of applied strain rate, experimental investigations were carried out from quasi-static loading up to crash-relevant strain rates. Strain and temperature fields were recorded using a high-speed video camera and a high-speed infrared camera, thus enabling a detailed analysis of the generated heat and the heat transport mechanisms in the deformed zones. The examination showed that the isothermal-adiabatic transition under shear loading ranges between strain rates of $0,01 \text{ s}^{-1}$ up to 50 s^{-1} for the investigated materials and thus at higher strain rates compared to tensile loading. This indicates an extreme strain localization in shear bands compared to necking zones at comparable strain rates. Consequently, larger temperature gradients occur in the shear zone, which significantly affect the heat transfer mechanisms. Furthermore, with increasing strain hardening, the isothermal-adiabatic transition region is also shifted to higher strain rates due to a slowdown of the strain and temperature localization process. Based on these results, local heating and the occurrence of adiabatic shear bands can be estimated without costly thermomechanical coupling finite element simulations (FE-Simulations).

Keywords

High strength steel, Metal forming, Thermal effect

1 Introduction

In highspeed forming processes and in crash scenarios highly dynamic loads occur and cause local temperature increases of several hundred Kelvin, particularly in advanced high strength steel sheets (AHSS). Under large strain rates steel materials in general show a tendency to develop adiabatic shear bands leading to material segregation, as initially demonstrated by Kravz-Tarnavskii (1928). This phenomenon is used in adiabatic blanking processes. Due to significant lower shear failure strains under high strain rates compared to quasi-static shear loading, observed e.g. by Peiers and Verleysen (2012), it is imperative to prevent it in crash safety applications. In both scenarios, it is critical to determine the range of strain rates at which adiabatic shear bands (ASB) occur, in order to induce or prevent shear failure. Additionally, estimating heat dissipation is essential for optimizing computational efficiency without resorting to thermomechanically coupled simulations.

Shear bands develop under complex thermokinetic conditions. The main results in literature are summarized by Walley (2007) and Dodd (2015). Under nearly adiabatic conditions temperature rise depends on the amount of locally dissipated plastic deformation energy converted into heat. Taylor and Quinney (1934) and Davidenkov and Mirolubov (1935) found that for metals plastic work is not completely transformed into heat. The percentage of plastic work converted into heat, characterized by the Taylor-Quinney coefficient (TQC), is assumed to be nearly constant as 90% for steels, according to Rittel (2017) under uniaxial and shear loading. The heat transfer in metals, primarily governed by thermal conduction, depends not only on thermal conductivity but particularly on the temperature gradient.

The objective of this study was to explore the factors impacting the isothermal-adiabatic transition region, especially in the presence of shear bands. Consequently, the primary influencing parameters affecting adiabatic heating in AHSS were experimentally examined, with a specific emphasis on shear loading vs. tensile loading. A comprehensive investigation into the heat generated and the heat transfer mechanisms within the intensely deformed regions was conducted through tensile and shear tests across a local strain rate spectrum ranging from approximately 0.001 s^{-1} to 50 s^{-1} .

2 Experimental investigations

2.1 Materials and test setup

The main investigations of the present study were carried out on a micro-alloyed steel HX340LAD and a dual-phase steel HCT980X. Further selective examinations were conducted on an additional dual-phase steel HCT980X+Z110MB and a complex-phase steel HCT780C, see **Table 1**. Uniaxial tensile tests and shear tensile tests were performed at test speeds between 0.02 and 25 mm/s, resulting in nominal strain rates from 0.001 s^{-1} to 1 s^{-1} . In the localized zones they induce larger strain rates, as discussed in chapter 2.3. Uniaxial tension tests utilized a flat tensile specimen geometry recommended in ISO 26203-2 (2011), which is well-suited for high strain rates. The shear tensile tests were conducted using an asymmetric shear notch tensile specimen, both geometries see **Fig. 1** left.

material	Sheet thickness in mm	$R_{p0.2}/R_{eL}$ in MPa	R_m in MPa	A_g in %	A_{20mm} in %	n_{2-Ag}	Z in %
HCT980XG	1.4	759	1049	7.1	13.4	0.08	33.5
HX340LAD	1.5	390	461	17.4	35.6	0.148	62.5
HCT780C	1.5	689	878	8.4	17.4	0.094	50.5
HCT980X+Z110MB	1.5	683	1007	10.3	18.5	0.132	35.1

Table 1: Mechanical values 90° to rolling direction

The tests were performed on a high-speed testing machine (Zwick 500/10). The local strain fields on the specimen surface were determined using a high-speed video camera of the Photron SA series and evaluation with 2D digital image correlation (DIC) using the GOM Correlate 2020 software. The temperature measurements were carried out on the back of the sample using a high-speed infrared dual-band camera (IR-camera) from the IRCAM GEMINIS 327k ML series. The force was measured using a piezo load cell. Based on the strain and temperature field measurements, the strain and temperature developments in the localized zones were evaluated. The von Mises equivalent strain was selected as the relevant strain variable, to be able to compare the temperature increase as a function of the local strain between the flat tensile and shear tensile tests, see **Fig. 1** right.

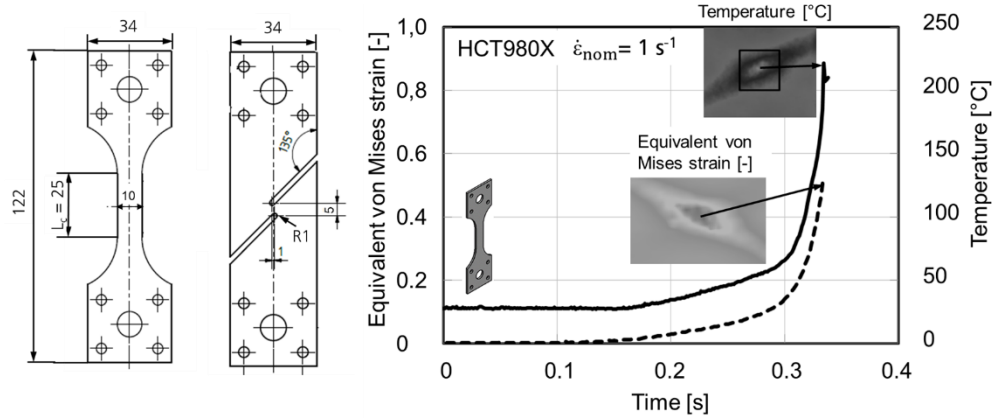


Figure 1: Flat tensile and shear tensile geometry (left) and measured maximal local temperature and strain (right) for the dual-phase steel HCT980X

2.2 Measured results

In all experiments, the local temperature rise was plotted against the local strain, as illustrated by the example of flat tensile tests on the micro-alloyed steel at various strain rates in **Fig. 2** on the left. Elevated strain rates corresponded to higher temperature increases. Nevertheless, for the comparative assessment of adiabatic temperature rise across distinct sheet materials, understanding the fraction of deformation energy retained in the form of heat is essential. This proportion is characterized by the ratio q/w of volume-specific thermal energy q to volume-specific plastic deformation work w within the localized region. The ratio q/w , defined as the "heat ratio", represents the fraction of plastic deformation work initially

transformed into heat quantified by the TQC, less the portion of energy dissipated through heat transfer.

w was determined according to equation (1) by the maximum local equivalent strain $\varepsilon_{v,max}$ in the necking area and the corresponding true equivalent stress $\sigma_{wahr,v}$. For uniaxial tensile tests $\sigma_{wahr,v}$ is calculated from the engineering stress assuming volume constancy and a homogeneous stress state over the thickness. Lateral forces in the clamping can be neglected due to the rotationally free mounting. For shear tests $\sigma_{wahr,v}$ is corrected because of the multiaxial stress state, assuming associated flow rule according to Andrade (2019).

q was calculated according to equation (2) by the measured temperature increase ΔT , a constant heat capacity c_p of 0.47 J/gK and a constant density ρ of 7.86 g/cm³. According to Richter (2010) the heat capacity of microalloyed steels can vary by 10-15% between 20°C and 200°C, hence q might be moderately underestimated for larger temperatures.

$$w = \int_0^w dw = \int_0^{\varepsilon_{v,max}} \sigma_{wahr,v} d\varepsilon_{v,max} \quad (1)$$

$$q = \rho c_p \Delta T \quad (2)$$

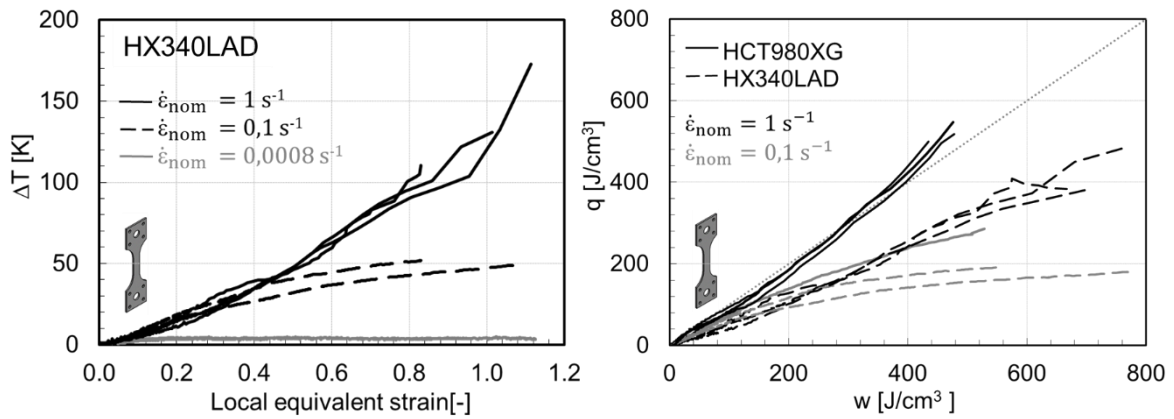


Figure 2: Measured local temperature increase dependent on the local equivalent strain at different strain rates for HX340LAD (left) and calculated local heat dependent on local plastic work at different strain rates for two materials (right) both for uniaxial tensile tests

2.3 Evaluation of isothermal-adiabatic transition region

Not only the heat ratio q/w but also the local strain rate changes over the duration of the test, see Fig. 3. In order to obtain a correlation between heat ratio and equivalent strain rate for each test, the mean value of the heat ratio $\bar{q/w}$ and the mean value of the equivalent strain rate $\bar{\dot{\varepsilon}_v}$ was calculated between a material-dependent local initial strain, located around the start of necking, and the elongation at fracture, see Fig. 3.

The strain rate dependent heat ratio $\frac{\bar{q}}{w}(\bar{\dot{\epsilon}}_v)$ differs significantly for the different steel sheet materials, see **Fig. 4** on the left. For the micro-alloyed steel HX340LAD, the values for $\frac{\bar{q}}{w}(\bar{\dot{\epsilon}}_v)$ are lower than for the dual-phase steel HCT980XG at a comparable strain rate. Assuming a nearly constant TQC of about 90%, the micro-alloyed steel therefore exhibits more heat transfer at a comparable strain rate compared to the dual-phase steel. The transition range from isothermal to adiabatic can be determined between 0.01 s^{-1} and approximately 10 s^{-1} for the dual-phase steel and up to higher strain rates for the micro-alloyed steel. This difference could be due to different temperature gradients in the localized zone, which in turn could result from different strain gradients based on localizations dependent on material properties.

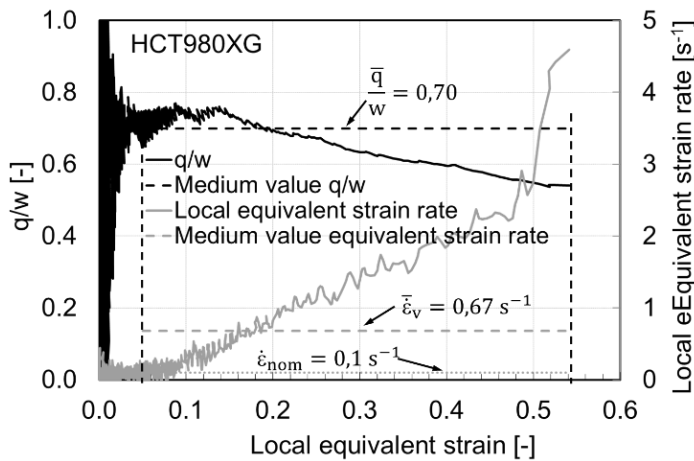


Figure 3: Mean value calculation for q/w and the equivalent strain rate in the range between 5% equivalent strain and failure strain using the example of a flat tensile test at a nominal strain rate of 0.1 s^{-1} for the sheet material HCT980XG

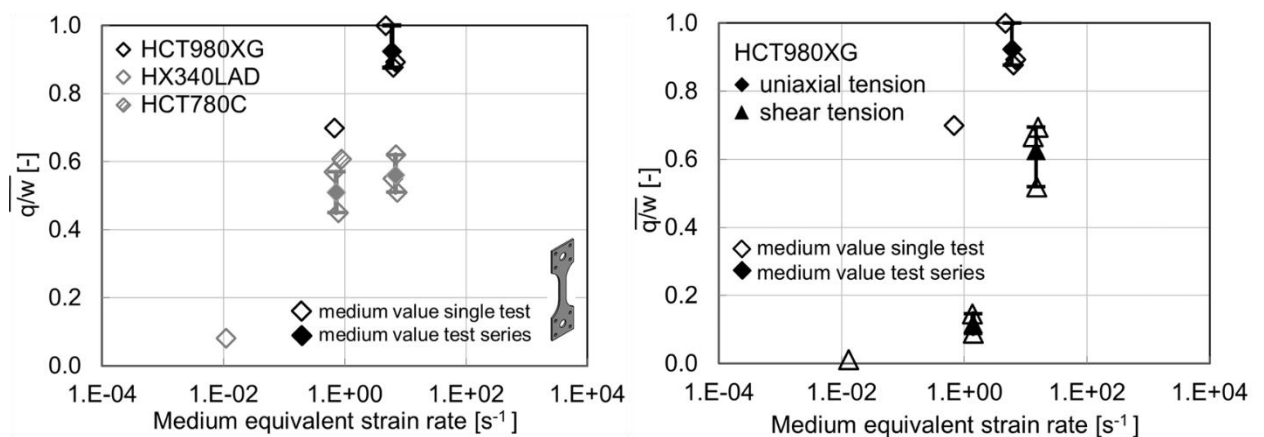


Figure 4: Average heat ratio as a function of the medium equivalent strain rate for various AHSS under tensile load (left) and for tensile and shear load for HCT980XG (right)

For the shear tensile tests, the transition range between isothermal and adiabatic is found at higher strain rates compared to the flat tensile tests, see **Fig. 4** on the right. For a

strain rate of about 10 s^{-1} for example, the flat tensile tests show a heat ratio value $\frac{\bar{q}}{w}(10 \text{ s}^{-1})$ of about 0.9 and therefore almost adiabatic conditions, while the shear tensile tests with a heat ratio value $\frac{\bar{q}}{w}(10 \text{ s}^{-1})$ of about 0.6 are found still in the transition region between isothermal and adiabatic. It can therefore be assumed that under shear loading, more heat dissipation occurs from the shear bands into the surrounding material than it is the case with the necking zones under tensile loading at comparable strain rates.

2.4 Evaluation of strain localization

The temperature gradients that occur in the localized zone are due to the localized strains and are also mutually influenced by the local strain rates, the stress state (necking or shear band) as well as by material properties like the hardening value n and the heat conductivity λ . Due to these complex relationships, extended analyses were carried out on the influence of the material properties on the localization of the strains. Therefore a key value “strain localization” is defined according to equation (3) as the height of the strain peak $\varepsilon_{loc,max}$ related to the full width at half maximum (FWHM), see also **Fig. 5**.

$$\text{Strain localization} = \frac{\varepsilon_{loc,max}}{FWHM} = \frac{\varepsilon_{v,max} - \varepsilon_w(F_{max})}{FWHM} \quad (3)$$

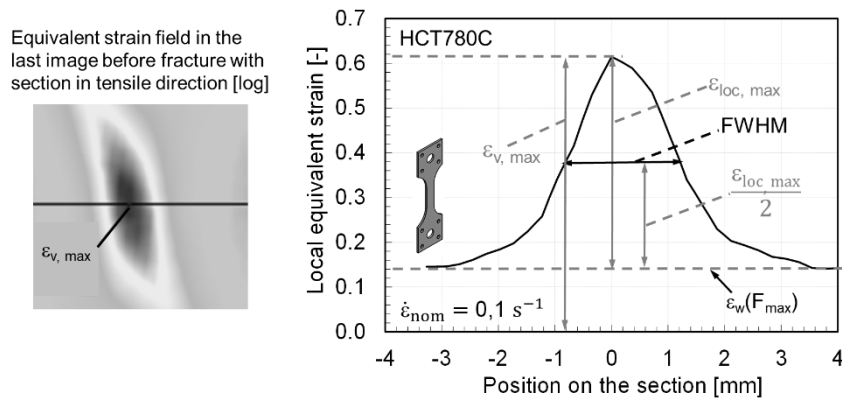


Figure 5: Definition of the relevant factors for the evaluation of the strain localization

With increasing strain localization, the local equivalent strain rate increases for the investigated sheet steel materials, see **Fig. 6** on the left. With increasing hardening value n_{2-Ag} , the strain localization also increases at a comparable strain rate. This phenomenon, initially appearing counterintuitive, is elucidated by the observation that as the n value decreases, larger strain rates occur with consistent localization. **Fig. 6** on the right shows a linear correlation of the strain localization and the equivalent strain rate and can be approximately described by the following relationship (4). Here $\dot{\varepsilon}_v$ represents the local equivalent strain rate at the localization maximum.

$$\frac{\varepsilon_{loc,max}}{FWHM} \cong n_{2-Ag} \dot{\varepsilon}_v \quad (4)$$

Hence, under similar strain rate conditions, higher temperature gradients may arise with an increasing n -value, facilitating enhanced heat transfer from the localized zone to the

surrounding material. Also Frost and Ashby (1982) showed that the transition strain rate for the isothermal-adiabatic transition shifts towards higher strain rates with increasing n -value. Therefore materials with higher n -values undergo increased heat transfer at a comparable strain rate. Under shear loading, significant larger strain localizations occur than under tensile loading, see **Fig. 7**, as a reason for the increased heat transfer in shear bands compared to necking zones under tensile loading.

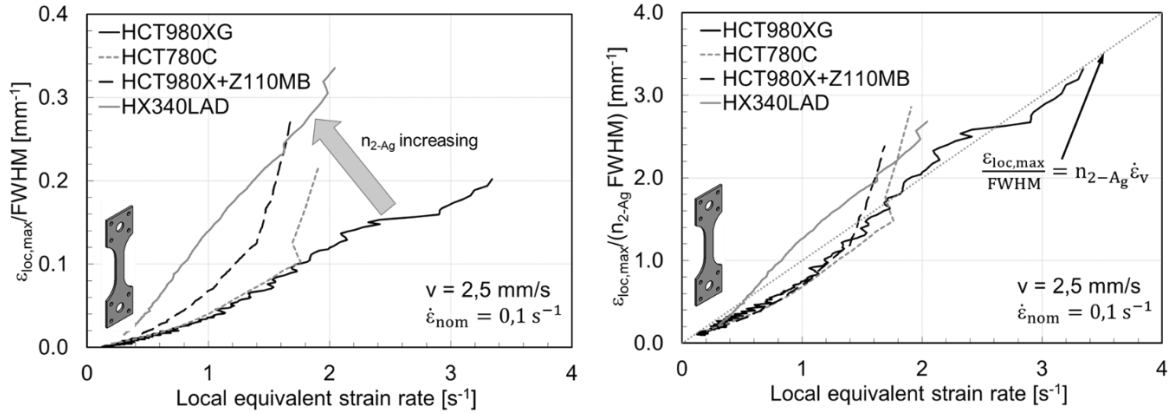


Figure 6: Strain localization (left) and n -value related strain localization (right) vs. local strain rate for tensile tests with a test speed of 2.5 mm/s for different AHSS

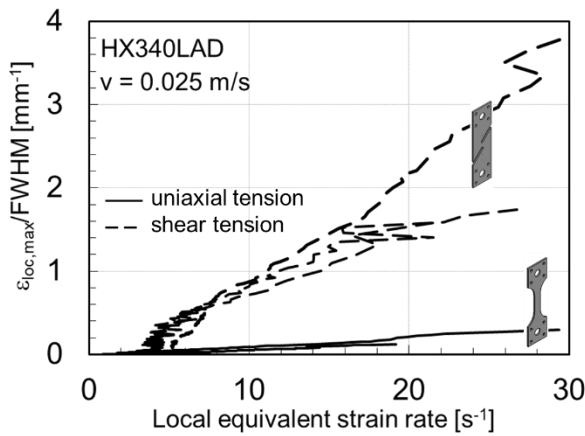


Figure 7: Strain localization dependent on the local strain rate for shear tests compared to tensile tests with a test speed of 0.025 mm/s for the micro-alloyed steel HX340LAD

3 Conclusions

The aim was to investigate the factors influencing the isothermal-adiabatic transition region, particularly when shear bands occur. The results lead to the following conclusions:

- The isothermal-adiabatic transition region shows a dependency on the strain rate, the hardening behavior and the stress state. For a HCT980X steel under tensile loading the transition range is found between strain rates around 0.01 s^{-1} and 10 s^{-1} . For a HX340LAD with a larger n -value it is shifted to higher strain rates due to a slowdown of localization.

- The correlation between strain localization and equivalent strain rate can be estimated using a linear model, with the slope being determined by the n -value.
- Under shear loading the isothermal-adiabatic transition region occurs at higher strain rates compared to tensile loading. For the investigated AHSS approximately adiabatic conditions can be assumed at local strain rates from a magnitude of about 50 s^{-1} . This is attributed to substantially larger localizations at equivalent strain rates within the peak of shear localization, resulting in enhanced heat dissipation.

The findings are pertinent for determining the potential occurrence of adiabatic shear bands. This is applicable to the desired formation of adiabatic shear bands during adiabatic blanking processes, but can also be helpful in high-speed forming using electromagnetically accelerated tools or membrane pulse techniques. It is also relevant in the undesired occurrence of such bands to avert failure in safety-critical dynamically loaded structures. The results establish an empirical foundation for assessing the fraction of heat retained within the material within the isothermal-adiabatic transition region, considering the stress state (tensile or shear) and strain rate as well as material properties. This enables the estimation of localized temperature rises without complex thermomechanical coupling FE-simulations, even in scenarios involving heat dissipation, which is shown by Klitschke (2022).

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Micro-scratch tests as a method for determining the wear resistance of high-speed blanked surfaces

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Abstract

Compared to conventional cutting processes such as normal or fine blanking, high-speed blanking (HSB) is an economically and ecologically relevant alternative to produce sheet metal parts. Depending on the material and the process parameters, the microstructure is significantly affected, and adiabatic shear bands (ASB) can form due to the localisation of the induced deformation. The resulting HSB surfaces are characterised by high hardness, small rollover depth and width, low roughness, and almost no burr. Therefore, HSB is a promising strategy for producing blanked surfaces, which can be used directly as functional surfaces without further post-processing steps. Besides the surface quality, properties like the wear resistance are relevant for technical applications. To enable the production of functional surfaces, which meet the requirements for the application, the understanding of the relationships between the blanking process parameters, the resulting microstructure, and the surface properties is essential. As standardised macroscopic wear testing methods are often not applicable to the geometrically complex shear-cut surfaces, in this study a first approach was taken utilising the micro-scratch test method for determining the scratch resistance directly measured on the blanked surface.

Keywords

Blanking, Surface Analysis, Wear

1 Introduction

High-speed blanking (HSB) is an economically and ecologically relevant alternative to produce sheet metal parts when compared to conventional cutting processes such as normal or fine blanking or laser cutting. Depending on the material and the process parameters, the microstructure is significantly affected, and adiabatic shear bands (ASB) can form due to the localisation of the induced deformation. The resulting HSB surfaces are characterized by high hardness, small rollover depth and width, low roughness, and almost no burr as found by Neugebauer et al. (2010), Schmitz et al. (2020) and Winter et al. (2021). Therefore, HSB is a promising strategy for producing blanked surfaces, which can be used directly as functional surfaces without further post-processing steps. Besides the surface quality, properties like the wear resistance are relevant for technical applications. To enable the production of functional surfaces, which meet the requirements for the application, the understanding of the relationships between the blanking process parameters, the resulting microstructure, and the surface properties is essential.

In this study, the determination and evaluation of the wear properties of HSB surfaces are in focus. A particular challenge here is that, due to the use of circular punches, the blanked surface is not a flat surface (**Figure 1**). As standardised macroscopic wear testing methods are often not applicable to the geometrically complex shear-cut surfaces, there is a need for new and adjusted experimental set-ups. A further challenge lies in realising a microstructure-sensitive testing method in order to identify correlations between the process, the microstructure, and the wear properties. Rojacz et al. (2023) showed in their study the possibility of microstructure-sensitive testing using nano-scratches to identify the wear mechanisms influenced by the microstructure. These micro- and nanoscale investigations must be carried out on metallographically prepared material sections (Gane and Skinner, 1973; Xiao et al., 2013). However, the aim of this study is to determine the property profile of the HSB surface in as-blanked state. Polishing and grinding processes as part of a metallographic preparation would remove this surface area and therefore no longer allow any reliable statement to be made about the functionality of the surface. Consequently, the scratch tests were carried out directly on the non-planar, rough HSB surface, which proved to be challenging. The extent to which this method is suitable for determining the resistance against abrasion as a function of the process parameters is examined in more detail below.

2 Experimental methods

2.1 Material and blanking process

For this study, surfaces resulting from different blanking process parameters were investigated. The non-age hardenable aluminium alloy 5754 and the press-hardened steel 22MnB5 were used as commercially available sheet material with a thickness of 2 mm. From both sheet materials, coins with a diameter of 20 mm were blanked with different process parameters listed in **Table 1** using high-speed blanking presses (MPM formerly ADIAPRESS, France). The specific machine set-up is described by Winter et al. (2021) for the ADIAflex® and by Schmitz et al. (2024) for the ADIAClip1000J®.

Material	Machine and die	Blanking parameters	Surface roughness R_a
Al5754	ADIAflex, 20.2 mm die	0.02 m/s blanking speed, 5 % clearance	$2.66 \pm 0.04 \mu\text{m}$ in clean cut area, $9.20 \pm 0.39 \mu\text{m}$ in fracture area
Al5754	ADIAflex, 20.2 mm die	10 m/s blanking speed, 5 % clearance, 1000 J blanking energy	$1.23 \pm 0.02 \mu\text{m}$ in clean cut area, $5.52 \pm 0.15 \mu\text{m}$ in fracture area
22MnB5	ADIAClip, 20.1 mm die	8 m/s blanking speed, 2.5 % clearance, 600 J blanking energy	$4.04 \pm 0.24 \mu\text{m}$ in clean cut area, $3.30 \pm 0.25 \mu\text{m}$ in fracture area
22MnB5	ADIAflex, 20.2 mm die	10 m/s blanking speed, 5 % clearance, 1000 J blanking energy	$3.67 \pm 0.06 \mu\text{m}$ in clean cut area, $2.66 \pm 0.13 \mu\text{m}$ in fracture area

Table 1: Investigated blanked conditions resulting from different process parameters.

2.2 Micro-scratch tests

For the micro-scratch tests, a universal nanoindenter UNAT (ASMEC, Germany) with a spherical diamond indenter with a 10 μm tip radius was used. The test data were analysed by the software InspectorX V3. In **Figure 1** the non-planar surface of the blanked coin and the application of the micro-scratch, perpendicular to the blanking direction, are shown schematically.

The three-step standard procedure for scratch testing consists of a pre-scan, the scratch and a post-scan. By the pre-scan, the initial surface profile is determined using a small load of 1 mN for the aluminium alloy and 2 mN for the manganese-boron steel, respectively. This profile serves as a reference for the unscratched surface with initial roughness and is later subtracted from the profile determined by the post-scan. Further, due to the radius of the blanked coins, a correction function was applied by the software to eliminate the influence of the surface curvature on the measured profile by pre- and post-scan.

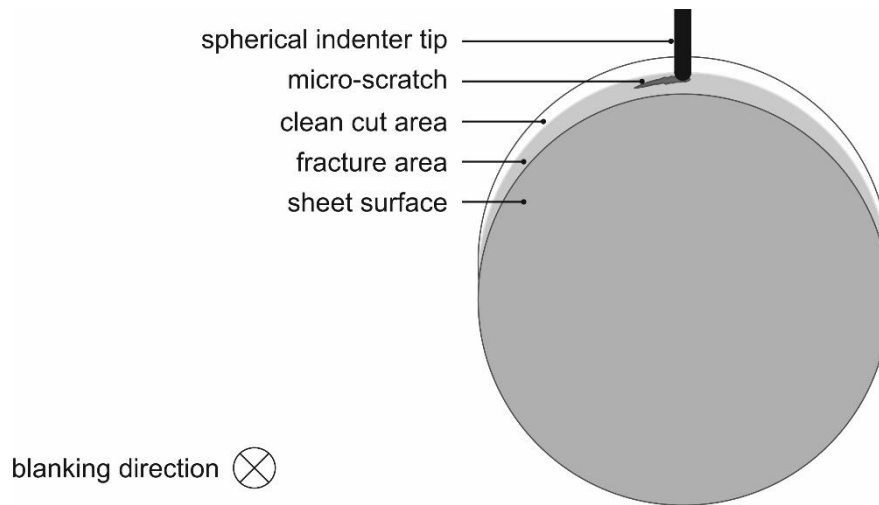


Figure 1: Schematic representation of the micro-scratch being positioned on the blanked surface perpendicular to the blanking direction.

For these first experiments, the micro-scratch test parameters listed in **Table 2** were used. These parameters ensured that no major cracking or flaking of the material occurred during the scratch, ensuring a correct testing procedure. The progressive force mode was used for determining critical loads causing material failure depending on the blanked surface and the process parameters.

Material	Scratch length	Applied normal force	Scratch speed
Al5754	200 μm	50 mN, linearly progressive	5 $\mu\text{m/s}$
22MnB5	200 μm	300 mN, linearly progressive	10 $\mu\text{m/s}$

Table 2: Parameters used for the micro-scratch testing on the blanked surfaces.

3 Results and discussion

For the aluminium alloy, the different blanking process parameters had a significant influence on the resulting surface (see **Figure 2 a, c**). The high-speed blanked surface exhibits a smaller height of the clean-cut area but a larger fracture area, and the roll-over height is significantly reduced. Further, the width of the shear zone resulting from the induced deformation by the blanking process is smaller for a higher blanking speed due to a more pronounced localisation of deformation (see **Figure 2 b, d**). The surface roughness after high-speed blanking is approximately halved when compared to the blanking speed (see **Table 1**).

For 22MnB5, there is no significant difference between both blanked conditions (see **Figure 3 a, d**). The resulting appearance and the roughness of the surfaces (see **Table 1**) are comparable, despite the different blanking speeds and clearances used. The metallographically prepared cross sections of the blanked steel show for both parameter sets pronounced adiabatic transformation shear bands with a width of approximately 17–22 μm (see **Figure 3 c, f**).

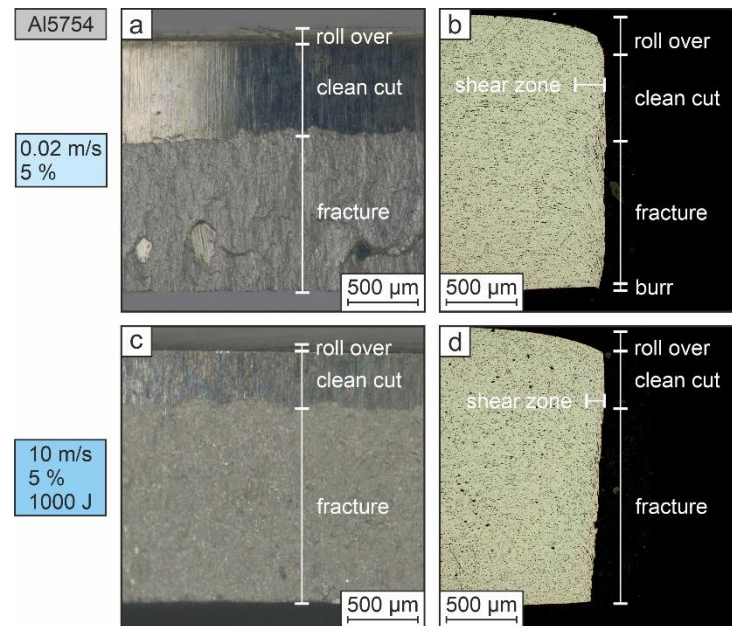


Figure 2: Optical micrographs of (a, b) conventionally and (c, d) high-speed blanked surface of the aluminium alloy 5754.

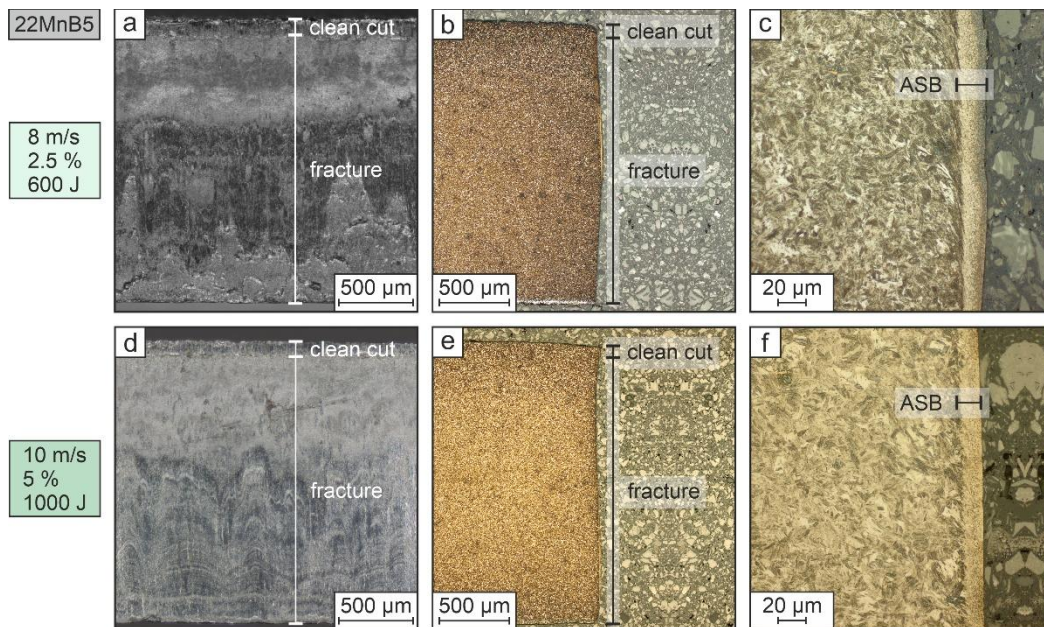


Figure 3: Optical micrographs of high-speed blanked surfaces of the used 22MnB5: (a–c) lower blanking speed and clearance, (d–f) higher blanking speed and clearance.

The surfaces of both blanked conditions for each material, respectively, were tested using micro-scratches. **Figure 4** shows exemplarily scanning electron micrographs of the scratched high-speed blanked surfaces. Secondary electron contrast was used for visualisation of the surface profile of the micro-scratches. It is noticeable that the scratch could be applied in full length without interruptions caused by surface heterogeneities. Further, no material flaking or cracking is visible in the scratched area for the used magnification.

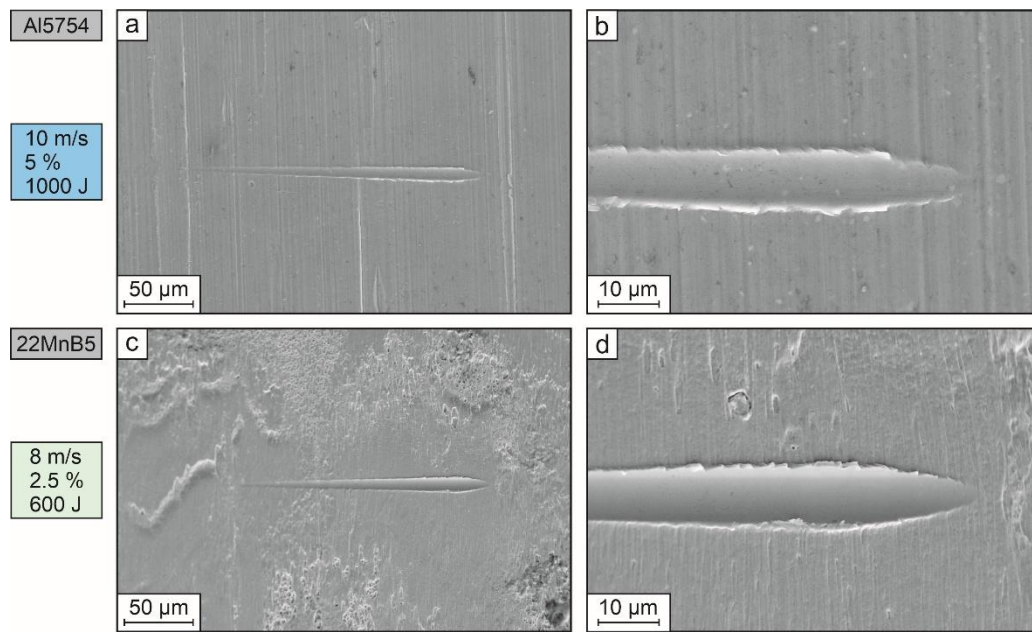


Figure 4: Secondary electron images of the scratched (a, b) aluminium alloy 5754 in the clean cut area and (c, d) steel 22MnB5 in the shear band area. a, c) full-length of micro-scratch and b, d) end of micro-scratch.

Table 3 lists selected values determined from the micro-scratch tests under progressive loading. The most important values are the critical load under which the first cracks occur (L_{c1}) and the distance that the indenter has travelled to this point (x_1). The software calculates these values from the determined post- and pre-scan detecting jumps in the profile.

These results indicate a tendency that for the aluminium alloy, the first cracks form at a higher normal force and at a higher distance for a higher blanking speed. As this corresponds directly to the wear behaviour, these first experiments indicate a higher resistance against scratches both in the clean cut and in the fracture area and, therefore, an improved abrasive wear behaviour of the high-speed blanked aluminium surface. Further, the friction coefficient, calculated by tangential force divided by normal force, is lower for the high-speed blanked condition, indicating a smaller wear rate when compared to the conventionally blanked one.

For both high-speed blanked steel surfaces processed by different blanking speeds, the micro-scratch results show no significant differences regarding the load at first crack formation L_{c1} or the distance x_1 . However, in this case, not only the blanking speed, but also

the clearance and the energy input were varied, which makes it difficult to compare the values. Therefore, further investigations are needed for 22MnB5. The micro-scratches could only be applied to the fracture area of the blanked surface, which corresponds to the shear band area. Due to the small width of the clean-cut area of all high-speed blanked 22MnB5 samples, micro-scratch testing was not possible in this area, as the indenter could not be positioned without causing material cracking and break-out at the specimen edge, which prevents a correct testing procedure.

Material	Blanking parameters	Measurement area	Micro-scratch results at first crack formation		
			distance x_1	load L_{c1}	friction coefficient μ_t
Al5754	0.02 m/s, 5 %	clean cut	23.39 μm	4.44 mN	0.083
		fracture area	15.32 μm	3.07 mN	0.095
Al5754	10 m/s, 5 %, 1000 J	clean cut	26.95 μm	9.30 mN	0.046
		fracture area	125.05 μm	30.64 mN	0.003
22MnB5	5 m/s, 2.5 %, 600 J	fracture area	14.42 μm	17.43 mN	0.005
22MnB5	10 m/s, 5 %, 1000 J	fracture area	15.95 μm	18.84 mN	0.020

Table 3: Results of the micro-scratch tests for conventionally cut (0.02 m/s) and high-speed blanked (5 and 10 m/s) Al5754 and 22MnB5 samples.

4 Conclusion and outlook

The first results presented in this study demonstrate the suitability of micro-scratch tests for determining the resistance against abrasion for non-planar blanked surfaces. However, as the aim of this method is the evaluation of the wear resistance of blanked surfaces depending on the process parameters and the resulting surface and microstructural properties, further work is necessary regarding suitable parameters for micro-scratch testing.

In the next steps, the influence of the tip radius of the spherical indenter will be considered. The used indenter tip radius of 10 μm is most suitable for a low surface roughness, which is near to a grinded condition. As the low surface roughness after high-speed blanking is still higher than after grinding, larger tip diameters will be used to minimise the influence of the surface topography. This will allow for a higher applied normal force as well, enabling a broader testing parameter range. Further, to get an understanding of the wear mechanisms influenced by the surface properties after blanking, other key values like the scratch energy density have to be determined. For analysing the scratch energy density, constant normal force has to be applied, and the volume of the scratch has to be measured using, for example, laser scanning microscopy. Besides these approaches, more tests have to be conducted to allow for a statistical evaluation, which is necessary for enabling the separation of influencing factors like surface roughness and surface microstructure. Regarding the separation of effects, micro-scratches will also be applied on metallographically polished cross sections of the blanked surfaces.

Acknowledgments

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Characterization of thermo-mechanical behavior and adiabatic shear band formation in a press-hardened 22MnB5 steel at different strain rates

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Abstract

Adiabatic Shear Band formation is observed in high-speed shear deformation of metallic materials, particularly in technologically relevant processes. It is characterized by the formation of increasingly localized areas of high shear strains. Depending on the selected process parameters, high-speed or adiabatic blanking can lead to the formation of an adiabatic shear band in the newly generated surface. These blanked surfaces have outstanding properties (e.g. high hardness, low rollover, low roughness, almost no burr) and can be used directly as functional surfaces. This reduces the need for complex and expensive, additional mechanical surface processing, significantly shortening the process chain and thus saving energy. In this contribution, we investigate the thermo-mechanical behavior and shear band formation of a press-hardened martensitic steel 22MnB5 under different strain rates. Tensile and compression tests are performed at different temperatures (in the range from 293 to 673 K) and in a wide range of nominal strain rates between 10^{-3} and $4 \cdot 10^3 \text{ s}^{-1}$. In addition, we use S-shaped samples to introduce a simple shear stress state to generate (adiabatic) shear bands, again while applying quasi-static and dynamic (nominal) strain rates, respectively. The plane surface of this sample geometry enables an in-situ observation of local strain fields by digital image correlation. We show the mechanical similarities and differences of dynamic versus quasi-static shear experiments using digital image correlation. Our experimental approach contributes to a deeper understanding of the rate and stress-state dependent formation of shear bands in press-hardened steels.

Keywords

thermo-mechanical behavior, shear band, strain rate dependency, dynamic testing

1 Introduction

The production of metallic sheets with functional surfaces and customized properties usually requires several time-consuming processing steps. High-speed blanking is a promising technique for achieving specific mechanical and microstructural properties in the blanked surfaces while at the same time significantly reducing the number of subsequent machining steps (Schmitz et al., 2020). This includes a uniform surface quality, high hardness, almost no burr as well as higher dimensional accuracy. A key phenomenon observed in this process is the formation of adiabatic shear bands (ASBs), which significantly influences the final material properties, as the cut surfaces consist of separated shear bands. ASBs commonly occur in metallic materials subjected to high shear strain rates, particularly in titanium alloys (Scholze et al., 2025a), tantalum (Nesterenko et al., 1997) as well as high-strength steels (Roux et al., 2015). The underlying mechanisms and the temporal sequence of their evolution are still not fully understood. Several, to some extent contradictory, explanations are described in the literature, including (quasi-) adiabatic heating and thereby thermal softening (Zener and Hollomon, 1944), substantial temperature increases due to frictional heating at fracture surfaces (Goviazin and Rittel, 2023), as well as the role of dynamic recrystallization (DRX) in ASB evolution (Rittel et al., 2008; Rodríguez-Martínez et al., 2015). In particular, the contribution of microstructural mechanisms in shear band formation has not yet been fully resolved.

Martensitic steels, as key engineering materials, exhibit pronounced ASB formation under rapid shear loading conditions (Jo et al., 2020). While ASBs in these materials have traditionally been associated with negative effects, such as structural instabilities and uncontrolled failure (Dodd, 2012), they can also be utilized in high-speed blanking to reduce processing forces while enhancing the cutting quality of functional surfaces (Schmitz et al., 2020). A thorough understanding of the thermo-mechanical behavior associated with ASB formation plays an important role for optimizing material processing and precisely controlling the properties of the cut surfaces formed by ASBs.

In this study, we investigate the thermo-mechanical behavior of a press-hardened 22MnB5 steel over a wide range of temperatures (293 K to 673 K) and strain rates, from quasi-static to dynamic conditions. For an investigation of shear band formation, simple shear tests are performed using an S-shaped sample geometry under quasi-static and high-rate conditions, combined with digital image correlation (DIC) and subsequent microstructural analysis. The results form a basis for a deeper understanding of ASB formation, particularly in terms of correlations between uniaxial thermo-mechanical behavior, shear localization, and the resulting microstructures.

2 Materials and experiments

Sheets of commercially pure 22MnB5 (1.5528) steel with a thickness of 2 mm were obtained from Salzgitter Mannesmann Forschung GmbH (Salzgitter, Germany). The chemical composition was validated by glow discharge optical emission spectroscopy as 0.19 wt.% C, 1.04 wt.% Mn, 0.001 wt.-% B, 0.22 wt.% Si, 0.77 wt.% Cr and 0.21 wt.% Ni (with balance Fe). The material was first austenitized in a chamber furnace at 1203 K for 6 minutes and

then press-hardened using a water-cooled steel plate tool by Salzgitter Automotive Engineering GmbH & Co. KG (Osnabrück, Germany). To investigate the strain-rate and temperature dependence of the mechanical behavior of the press-hardened steel, uniaxial tensile and compression tests were conducted under various strain rates at room temperature and elevated temperatures (473 K and 673 K). Quasi-static and mid-range strain rate tests were performed at nominal strain rates of 10^{-3} s^{-1} , 1 s^{-1} , and 10 s^{-1} using tensile specimens (50 mm gauge length, 10 mm width) that were machined by laser cutting. The tests were carried out in a Schenck PTT 250 K1 (Carl Schenck AG, Darmstadt, Germany) hydraulic universal testing machine, equipped with a 250 kN load cell and a custom-built conductive heating system. To obtain thermo-mechanical data at high strain rates, dynamic compression tests of cylindrical specimens (2 mm diameter, 2 mm height, machined by electrical discharge machining) were performed with a nominal strain rate of $4 \cdot 10^3 \text{ s}^{-1}$ using a Split-Hopkinson Pressure Bar (SHPB) setup equipped with an inductive heating system. Strain measurements in all testing setups were performed using DIC. For tests at a nominal strain rate of 10^{-3} s^{-1} , an ARAMIS Adjustable 24M system from Carl Zeiss GOM Metrology GmbH (Braunschweig, Germany) was used. For mid-range to high strain rates, a high-speed camera (Fastcam SA5, Photron, Tokyo, Japan) was used to record the deformation. Conductive heating (for tensile tests) and inductive heating (for SHPB compression tests) were configured to ensure that the test temperature was reached prior to testing within a short time interval of 10 seconds in both cases, minimizing long-term thermal effects. Three samples were tested for each strain rate and temperature.

To specifically investigate ASB formation in press-hardened 22MnB5, S-shaped samples were machined from the 2 mm sheets using electrical discharge machining. The plane geometry of the shear sample, with an overlap length of 1.6 mm, was designed to result in a stress state that approaches simple shear during nominal uniaxial compressive loading of the sample while simultaneously allowing for DIC measurements (Scholze et al., 2025b). The S-shaped sample has a height of 13.4 mm and a width of 10 mm. The 1 mm notches are machined at an angle of 45° . For further details on notch root radii etc. the reader is referred to (Scholze et al., 2025b). Quasi-static shear testing (with a crosshead displacement rate of 0.0016 mm/s) was conducted in a ZwickRoell UPM1475 universal testing machine (ZwickRoell GmbH & Co. KG, Ulm, Germany), with the ARAMIS Adjustable 24M system used for in-plane deformation capturing. Similar to the compression tests, dynamic shear experiments were performed using the SHPB setup with a striker velocity of approximately 9 m/s. A Kirana ultra-high-speed camera (Specialised Imaging Ltd., Pitstone, United Kingdom) was used for in-plane recording of the deformation of the S-shaped samples during dynamic shear testing and to document ASB propagation. The DIC data was evaluated using the Zeiss Inspect Correlate 2023 software. Conventional metallographic preparation, including hot mounting, grinding, polishing, and etching (925 ml ethanol, 25 g picric acid, 50 ml hydrochloric acid), of the fractured samples was carried out. Optical microscopy was performed using an Olympus GX51 inverted reflected-light microscope (Olympus Corporation, Tokyo, Japan).

3 Results and discussion

In **Fig. 1** the engineering stress-strain curves from uniaxial mechanical testing of the 22MnB5 press-hardened steel at different strain rates and temperatures are shown. The results from testing at the highest rate, performed in the SHPB, exhibit some fluctuations that are associated with wave propagation effects. We observe that strength generally decreases with increasing temperature. This thermal softening phenomenon is more pronounced between 473 K to 673 K than from room temperature (293 K) to 473 K. Also, at elevated temperatures a decrease of ductility with increasing strain rate occurs; some experiments in the mid-range nominal strain rates at elevated temperatures were interrupted prematurely due to closed-loop feedback control issues in the challenging thermo-mechanical setup. During quasi-static testing at 673 K (Fig. 1a), an early onset of plastic instability is observed, which is directly related to thermal and microstructural softening mechanisms at elevated temperatures, as also observed in (Winter et al., 2017).

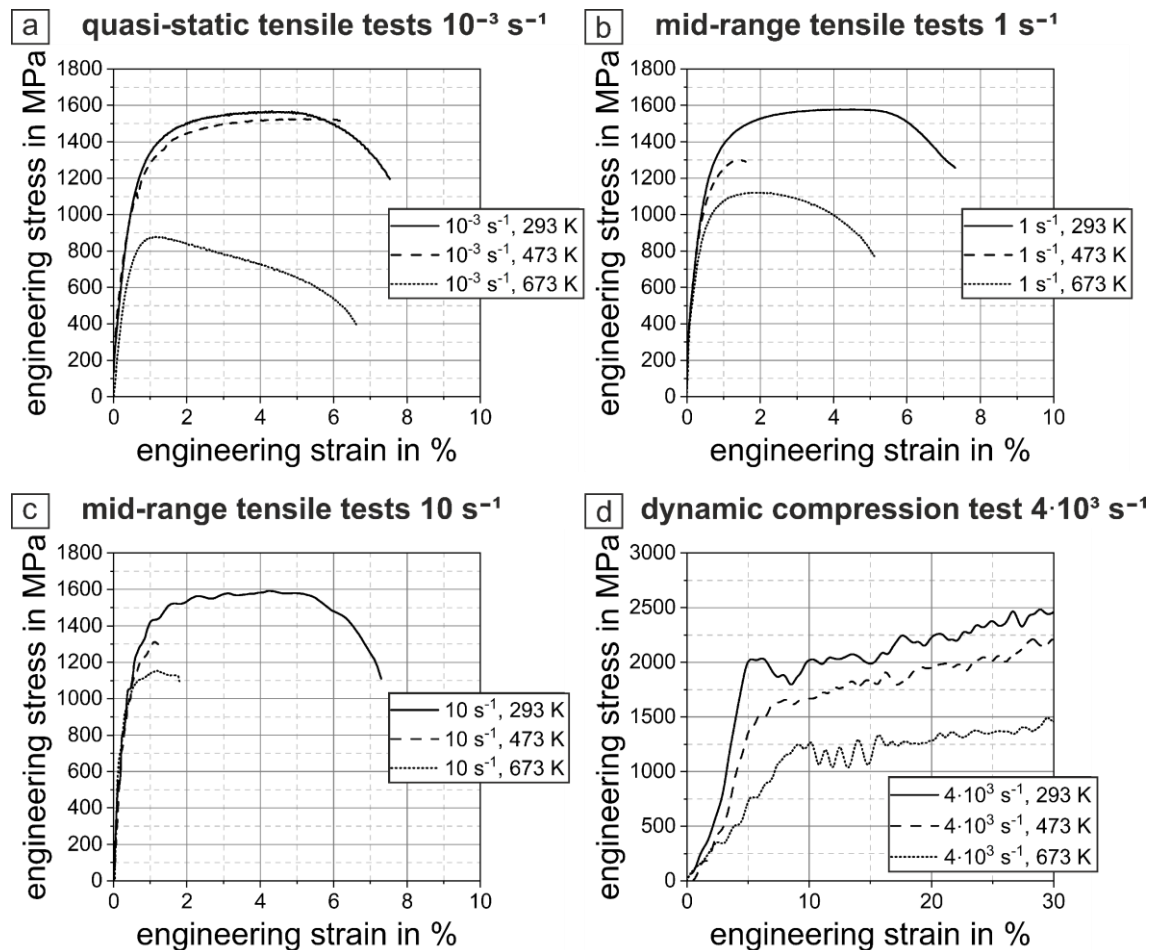


Figure 1: Engineering stress-strain curves of representative samples deformed at 293 K, 473 K and 673 K, at strain rates of (a) 10^{-3} s^{-1} , (b) 1 s^{-1} , (c) 10 s^{-1} , and (d) $4 \cdot 10^3 \text{ s}^{-1}$.

Since the stress level for all three quasi-static to mid-range strain rates is almost identical at 293 K, the strain-rate sensitivity seems to be relatively low (which agrees with results

reported in (Schmitz et al., 2020)) in contrast to that at elevated temperatures. We note that, for a well-defined determination of the strain-rate sensitivity, it would be necessary to perform jump-tests to separate rate effects from those of microstructural evolution, which is most pronounced at elevated temperatures due to thermal effects and also depends on the duration of individual tests.

Strain hardening, thermal softening and strain rate hardening directly influence the formation and growth of ASBs. When thermal (or microstructural) softening overcomes the hardening terms, adiabatic shear banding can occur (Dodd, 2012). To investigate and generate (adiabatic) shear bands in the 22MnB5 sheet, we performed simple shear testing, both at quasi-static and dynamic nominal shear rates, until fracture of the samples. In **Fig. 2**, representative DIC fields of shear strain and shear strain rate for the simple shear tests are shown. For both nominal strain rates the DIC fields have been selected to represent the deformation stage just before fracture. The shear deformation zone is somewhat inclined with respect to the global loading direction in the case of quasi-static loading. Moreover, the local deformation is illustrated by thin black lines spanning the entire sample from top to bottom. The deformation of the samples, concentrated in the well-defined shear zone, results in an S-shape of these lines. Shear strain and shear strain rate are further evaluated along the central black line in the DIC fields. The corresponding line profiles are visualized in Figs. 2e and f. Clearly, the shear strain deformation zone is wider for quasi-static testing (with a width of about 400 μm) than the corresponding zone for dynamic shear testing (approx. 200 μm). The maximum shear strain before fracture is determined from the DIC data as 2.16 for the quasi-static and 0.94 for the dynamic case.

The shear strain rate data demonstrate the localized deformation mode inherent in S-shaped sample testing: shear strain rates are concentrated in the center of the shear zones, and the maximum, local shear strain rates 0.0346 s^{-1} for quasi-static and up to 320 000 s^{-1} for dynamic testing, respectively) exceed the nominal deformation rates by several orders of magnitude. These results demonstrate the effect of thermal softening on shear localization in the S-shaped samples, and they are also in line with our observations from uniaxial testing: under quasi-static shear loading, thermal effects (self-heating) hardly affect the mechanical response and strain hardening is the dominating mechanism. In contrast, under dynamic loading conditions, thermal softening exceeds the contribution of strain-rate hardening and clearly leads to local instability, shear localization and ASB formation.

The local shear strain rate values discussed here are of course limited by the different spatial resolutions of the different camera systems used for the quasi-static and dynamic tests, respectively. For dynamic testing in particular 14 pixels per facet are required for DIC correlation, resulting in a size of 100 μm for one DIC facet. This is ten times larger than the ASB width of about 10 μm previously reported in shear compression tests of a similar press-hardened steel (Schmitz et al., 2020). Consequently, while the shear strain rate values reported here confirm the trend of increased rates associated with localized deformation in a shear band, the actual values can be expected to be even higher.

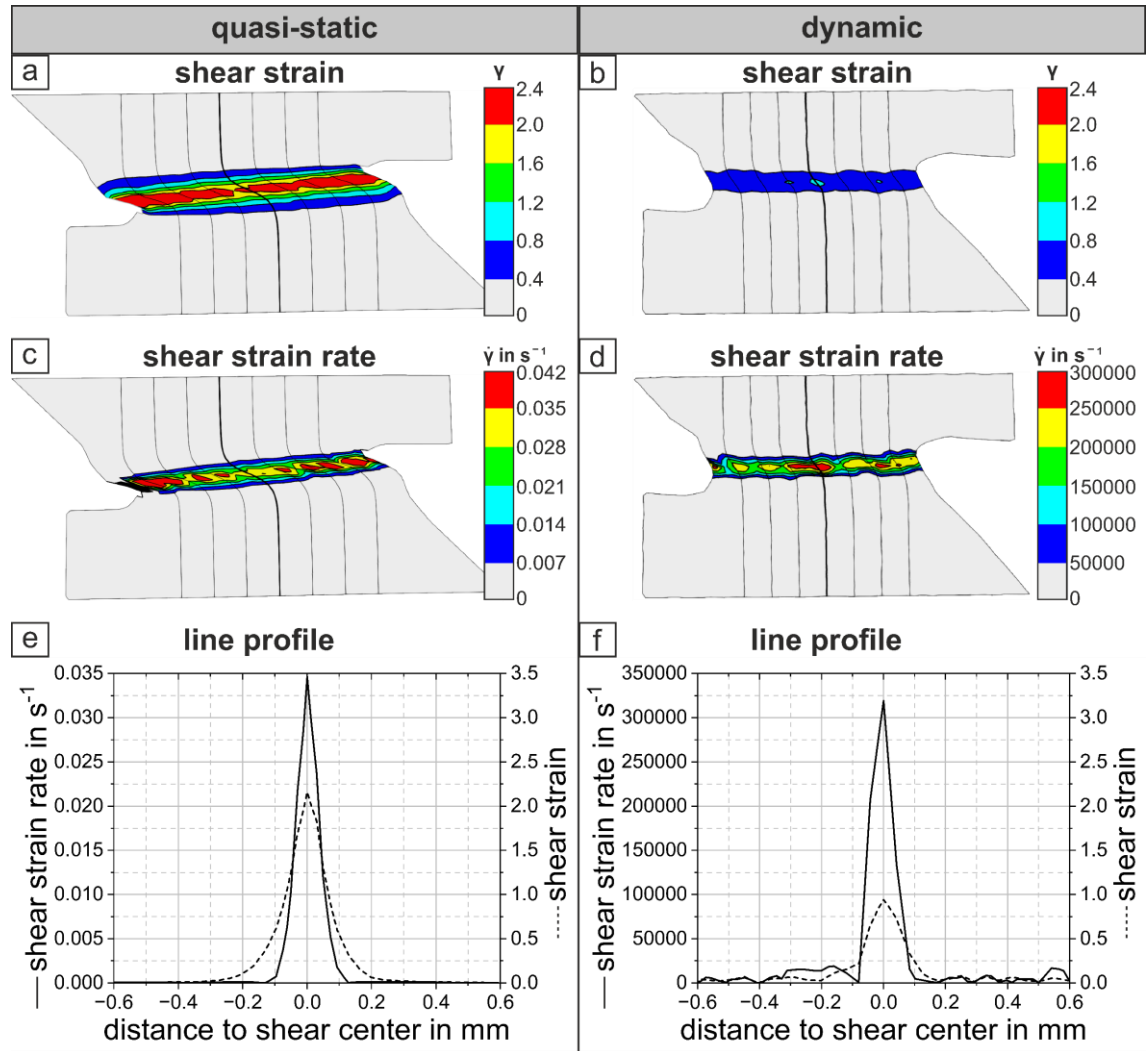


Figure 2: Results of DIC measurements during quasi-static (10^{-3} s^{-1}) and dynamic ($4 \cdot 10^3 \text{ s}^{-1}$) simple shear deformation. Distributions of (a) quasi-static shear strain, (b) dynamic shear strain, (c) quasi-static shear strain rate and (d) dynamic shear strain rate immediately before fracture. Lines from top to bottom indicate the local displacement fields. (e-f) Line profiles of shear strain and shear strain rate along the central black line.

In **Fig. 3** optical micrographs taken from S-shaped simple shear samples (quasi-static and dynamically deformed until fracture) are shown. We observe a relatively wide deformation zone with a shear band half-width of at least $40 \mu\text{m}$ in the quasi-static sample (Fig. 3a) based on the strongly sheared martensitic microstructure. In contrast, the microstructure of the S-shaped sample deformed under dynamic loading condition (Fig. 3b) is characterized by an adiabatic shear band with a half-width of only approx. $4 \mu\text{m}$ and no noticeably deformed martensite in the vicinity. This clearly confirms that deformation and microstructural changes are strongly limited to a very narrow shear band region (that could be fully resolved by our DIC measurements) under dynamic loading conditions.

We note in closing that the fracture surface of the dynamically deformed shear samples is much smoother than the quasi-statically obtained fracture surface. This morphology is

particularly promising with respect to the proposed use of ASB formation during high-speed blanking to produce functional surfaces.

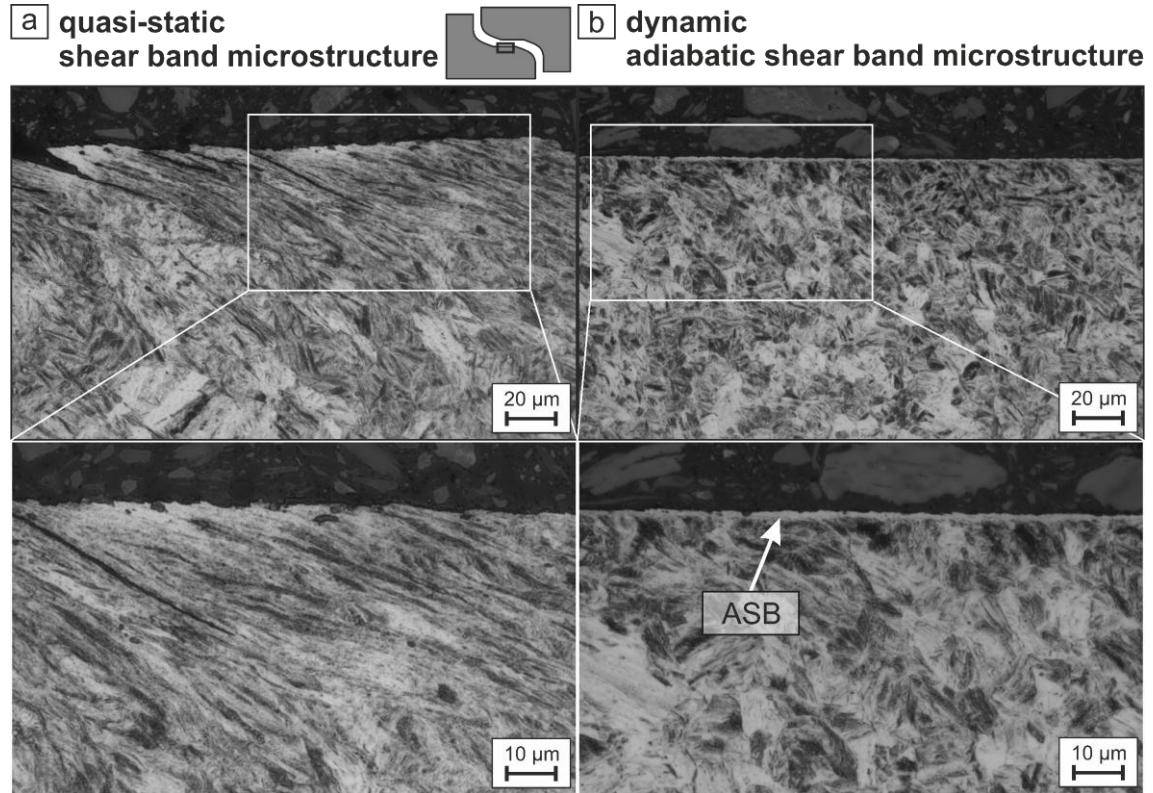


Figure 3: Optical micrographs of (a) quasi-static and (b) dynamically deformed simple shear samples.

4 Summary and conclusions

In summary, we investigated a press-hardened 22MnB5 steel by characterizing the thermo-mechanical behavior under different strain rates (10^{-4} to $4 \cdot 10^3 \text{ s}^{-1}$) and temperatures (293 to 673 K) using uniaxial tensile and compression tests. We then used S-shaped samples to introduce a simple shear stress state while applying quasi-static and dynamic (nominal) strain rates. The key results can be summarized as follows:

- Thermal softening: Strength decreases significantly between 293 K, 473 K and 673 K, with a ductility reduction at elevated strain rates.
- Strain-rate sensitivity: Low positive strain-rate sensitivity at 293 K and higher strain rate-sensitivity values at elevated temperatures.
- Quasi-static simple shear: Shear strain width of approx. 400 μm , maximum shear strain of 2.16 and maximum shear strain rate of 0.0346 s^{-1} determined by DIC. Shear band half-width: 40 μm .
- Dynamic simple shear: Shear strain width of approx. 200 μm , maximum shear strain of 0.94 and maximum shear strain rate of $320\,000 \text{ s}^{-1}$. Local shear strain rates are presumably higher than calculated from DIC data due to the spatial resolution of the high-speed camera.

- Microstructural evidence for distinct shear localization under dynamic loading conditions is given in the investigated ASB with a shear band half-width of 4 μm .

Acknowledgments

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Determination of GISSMO Damage Parameters for High-Speed Blanking of 22MnB5 and EN AW-5754

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Abstract

High-speed blanking (HSB) of high-strength materials, such as the steel 22MnB5, is an established method that delivers high-quality cut surfaces. Due to high shear rates, this process is associated with the formation of adiabatic shear bands (ASB) in the blanking zone. For an in-depth analysis of the effects arising in this process and to support the prototyping of blanking tools, a numerical simulation is required. For this purpose, material parameters, which are valid at elevated strain rates are necessary. In this study, the parameter identification for the GISSMO damage model is presented. For the parameter identification, a series of coupled experimental and numerical investigations were carried out on 22MnB5 and EN AW-5754 with a sheet thickness of 2 mm. The main objective was to determine the damage and failure behaviour of the investigated materials under different loading modes (or triaxialities). The resulting data was used as input for the subsequent optimisation of the parameters in LS-DYNA and LS-OPT.

Keywords

FE-Simulation, Damage model, high-speed, LS-DYNA

1 Introduction

The use of damage models to simulate complex forming processes has been increasing significantly in recent years (Gutknecht et al. 2016; Krinninger et al. 2017; Malakizadi et al. 2019). Damage model differs from conventional failure criteria in that it not only assesses the qualitative feasibility of the respective components, but also quantitatively evaluates the damage applied to the material during the forming process. This in turn is a significant advantage for following process steps such as joining (Otroshi et al. 2020; Rusia and Weihe

2020) and crash simulations (Valayil and Issac 2013; Praveen Kumar and Shriyaathsav 2019) which will further promote the use of these models in the future. Particularly interesting is the use of damage models in the simulation of cutting operations (Maranhão and Paulo Davim 2010; Zhang et al. 2017), especially of high-strength materials or at high speeds and strain rates as is case of high-speed blanking (HSB) (Schmitz et al. 2020; Winter et al. 2021; Galiev 2023).

The Johnson-Cook (JC) constitutive model (Johnson and Cook 1985) is a phenomenological formulation that extends conventional plasticity models by incorporating strain-rate sensitivity, thermal softening, and strain hardening. These features enable the model to describe characteristic material responses observed under dynamic loading conditions such as impact and penetration of metals. The three most important material reactions are strain hardening, strain rate effects and thermal softening. These three effects are combined multiplicatively in the constitutive Johnson-Cook model: The JC model is a most widely used failure model associated with conventional continuum mechanical models.

However, the JC model reaches its limits in more complex applications, especially in the simulation of high-speed blanking processes, as it does not include explicit mechanisms to account for damage and failure. To overcome these limitations, the Generalised Incremental Stress State dependent damage Model (GISSMO) was developed (Haufe et al.; Neukamm et al. 2009; Haufe et al. 2011) The most important aspect of GISSMO is the combination of damage models for crash simulation with localisation and instability models for forming applications. Especially for HSB processes, where a precise description of the material separation is crucial, GISSMO offers higher prediction accuracy than the classical Johnson-Cook model (Labinot Topilla). By combining crash simulation methods with localisation and instability models, GISSMO is particularly well suited for applications where both plastic deformation and failure mechanisms need to be captured realistically.

As part of this research, was parameter identification for GISSMO damage model by experimental and numerical inverse simulations for EN AW-5754 aluminium alloy and the press-hardened 22MnB5 steel (45 HRC) were investigated. The selection of these materials with completely different mechanical properties and fracture mechanisms enables a comprehensive evaluation of the GISSMO damage model. It also offers the opportunity to map the fracture behaviour for a broader range of materials in the context of the HSB process.

2 Methodology

2.1 Sample Geometry

The sample geometry was selected based on the recommendation of (Andrade et al. 2016) and then numerically tested in advance for suitable triaxiality ranges. The following sample types were then used for experimental investigations.

- **Tensile specimen:** Pure tensile load to determine the material behaviour under uniaxial tensile stress.
- **Shear tensile specimen 0°:** Shear load along the longitudinal axis to investigate the material behaviour under shear stress.
- **Shear tensile specimen 45°:** Shear loading at a 45° angle to the longitudinal axis to capture anisotropic effects and mixed loading conditions.
- **Notched tensile specimen:** Tensile specimen with a notch to analyse the behaviour under high triaxiality and the influence of notch effects on the damage.

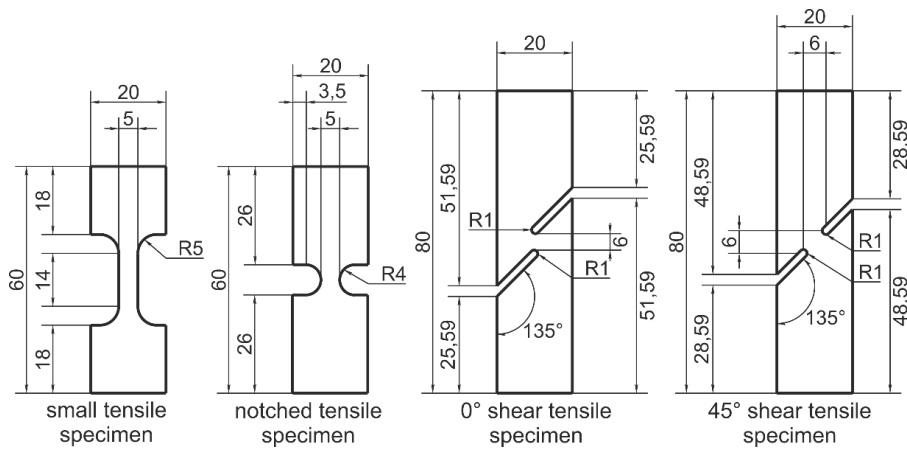


Figure 1: Specimen geometry causing different triaxialities in the material to identify parameters from GISSMO model

The tensile tests were performed using a Zwick universal testing machine ZPM 1475 in accordance with ISO 6892-1:2009 (Aegerter et al. 2011). The specimens were carefully prepared and clamped in the testing machine. For the GISSMO characterisation, tensile tests leading to different triaxialities were performed. The variation of specimen geometry (**Fig.1**) allowed a comprehensive analysis of the damage behaviour of the material under different stress conditions.

2.2 FE Simulation with LS-DYNA

As part of numerical investigations in LS-DYNA, four investigated sample geometries from the experimental tests were simulated in order to later use them in an inverse simulation using LS-Opt to identify damage characteristics. In this study, the failure strain $\epsilon_f(\eta)$ and critical strain $\epsilon_{crit}(\eta)$ were determined using LS-OPT to accurately describe material behavior within the GISSMO damage model. These parameters are essential for predicting material failure and localisation in crash and forming simulations.

In the GISSMO damage model, damage evolution is treated as a phenomenological process based on the accumulation of equivalent plastic strain in relation to stress triaxiality.

Importantly, the model does not directly simulate microstructural damage mechanisms (e.g., void nucleation or crack growth), but instead approximates their macroscopic effect through calibrated strain-based functions.

Damage (D) is computed incrementally using the equivalent plastic strain (ε_{eq}) and the failure strain $\varepsilon_f(\eta)$, which varies with the local stress triaxiality η . Once the damage parameter D reaches a value of 1.0, complete material failure is assumed, and (if activated) element erosion may occur. The mathematical formulation is given in Equation (1).

$$D = \left(\frac{\varepsilon_{eq}^p}{\varepsilon_f(\eta)} \right)^n \quad (1)$$

Critical strain $\varepsilon_{crit}(\eta)$ marks the onset of material instability or strain localization. It is used to define the instability parameter F according to equation (2). When the instability parameter reaches the value on ($F=1$) damage-stress coupling is activated, leading to progressive material degradation.

$$F = \left(\frac{\varepsilon_{eq}^p}{\varepsilon_{crit}(\eta)} \right)^n \quad (2)$$

The GISSMO damage and failure model must be implemented in LS-DYNA using the keyword `*MAT_ADD_EROSION` and activated using the setting `IDAM=1`. The functions ε_f and ε_{crit} are defined as input parameters by keyword `*DEFINE_CURVE` and the parameters n and m are defined directly on the material card. A combination with other plasticity models available in LS-DYNA is required. For example, the classic `*MAT_024_PIECEWISE_LINEAR_PLASTICITY`, `*MAT_133_BARLAT_YLD2000`, `*MAT_243_HILL_90` etc. are frequently used. In this case, `MAT_133_BARLAT_YLD2000` was used because it takes into account the direction-dependent material behavior and provides better agreement with the experimental results.

To achieve statistically validated results, five samples were tested for each geometry. The target parameter for each specimen geometry for later validation and optimisation with LS-DYNA was the force-displacement curve of each specimen geometry. The force signal was provided by the Zwick tensile testing machine and the displacement was measured using a DIC camera system, with two points defined in each necking area - as shown in **Fig 2**. The elongation of the sample was evaluated by tracking the displacement of two points in the centre of the sample using `HISTORY_NODE`.

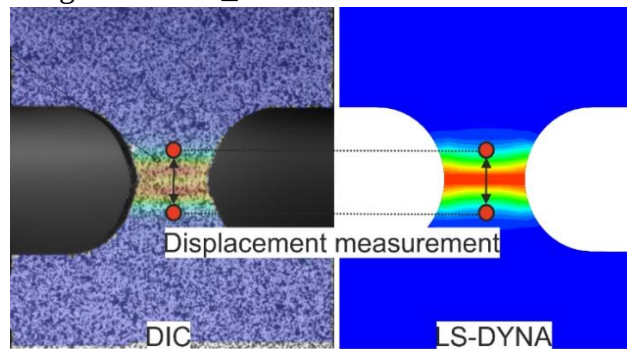


Figure 2: Length extension measurement in the experimental DIC measurements and in the simulation

As a first step, even before the inverse optimisation run, it was important to numerically test the selected tensile specimens in order to check the agreement between experiment and simulation in relation to the plasticity model without damage. For this purpose, the numerical model of the tensile specimens was set up in LS-DYNA using shells with ELFORM 1 and a mesh, as shown in **Fig 3**. Due to the large number of damage parameters investigated, the calculation was performed via implicit simulation in order to minimise the computational effort required as far as possible.

One side of the tested samples was fixed with ‘BOUNDARY SPC SET’ and the sample was pulled from the other side along the X-axis at the test speed recorded during the experimental test (2 mm/min). This corresponds to quasi-static tensile testing. In order to ensure good comparability between the measured and numerically determined data, the determination of the change of the parameter l_0 in the sample was carried out analogue to the measuring principle of DIC.

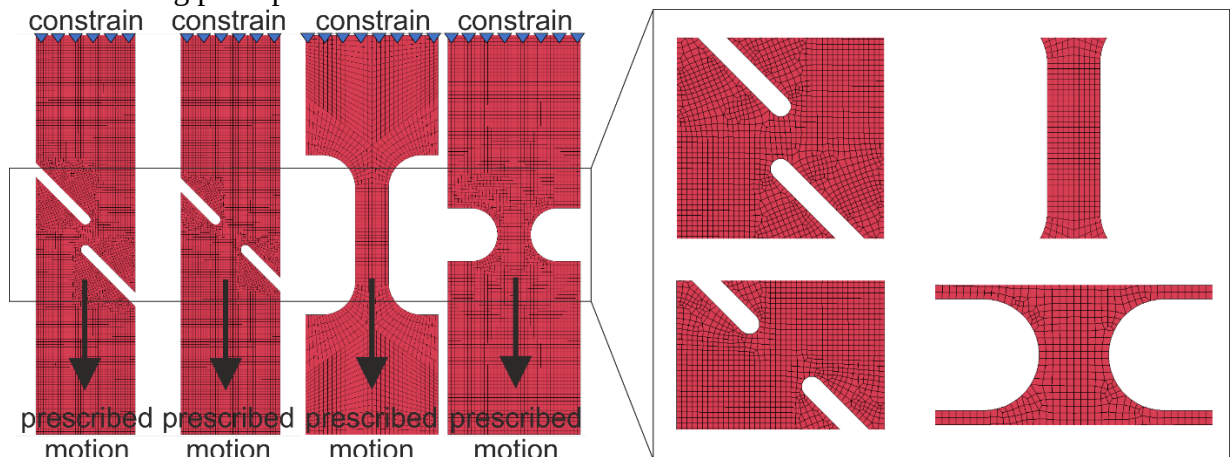


Figure 3: Mesh sample geometry for determining the GISSMO parameters

2.3 FE Optimisation with LS-OPT

Based on the experimental results of the samples analysed during the tests, a parameter identification study of the GISSMO model was carried out using LS-OPT. Specifically, the selected parameters, namely failure strain and instability strain of the GISSMO model. Figure 26 start setup of variable parameters in LS-OPT were calibrated by inverse simulation. The initial parameters were identified based on experimental failure strain values within a defined range. The force-displacement curve was defined as the target variable for a good match between experimental and numerical results. After the individual calculations of the selected tensile specimens, the expected triaxiality values were also replicated in the samples., and it was possible to start the inverse optimisation in LS-OPT. The structure of the optimisation algorithm in LS-OPT is shown in **Fig 4**.

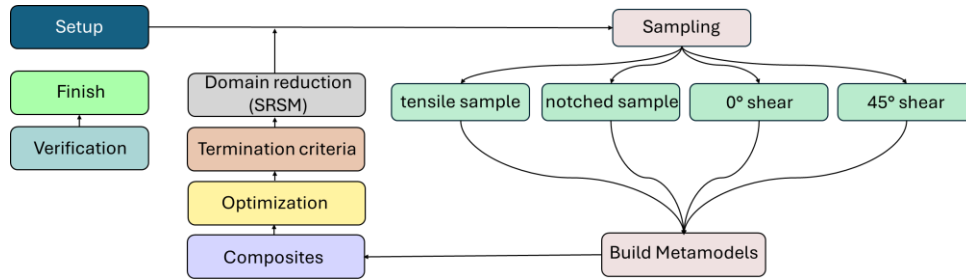


Figure 4: Schematic of the process flow for the optimisation algorithm in LS-OPT

Since the GISSMO parameters in LS-DYNA are primarily defined by the critical strain and failure strain as a function of stress triaxiality, an optimisation was performed using LS-OPT. The main objective of this optimisation was to calibrate the GISSMO parameters so that the simulated force-displacement curves match the experimental data as closely as possible. This ensures that the material behaviour under different loading conditions is accurately represented. As shown in **Fig.4**, the optimisation process follows a structured flow. After the sampling phase, four different specimen geometries—tensile, notched, 0° shear, and 45° shear—are evaluated, each corresponding to a specific triaxiality level. This variation is crucial for capturing the damage evolution across a wide stress state spectrum, which is essential for reliable GISSMO calibration.

A total of 20 optimisation iterations were performed, with 7 simulations per specimen geometry in each iteration, resulting in a total of 560 simulations. The use of metamodels in LS-OPT enabled efficient exploration of the parameter space and significantly reduced computational time. The domain reduction (SRSM) and termination criteria ensured that the optimisation converged towards physically meaningful and numerically stable parameter sets.

3 Results

As a result of the LS-OPT optimisation process, the force-displacement curves for the investigated specimen shapes of 22MnB5 and EN AW-5754 were mapped. Based on the agreement between numerical and experimental results, the resulting values for critical strain and failure strain were determined. For each tested sample, the experimental values were compared with the mean value of the numerical simulation. It is important to note that the determination of GISSMO parameters in this study is based on macroscopic force-displacement data without direct microstructural validation. While the onset of localization (e.g., necking) is used as a criterion for critical strain, it does not necessarily indicate the initiation of microstructural damage. Therefore, the identified parameters represent a phenomenological calibration. Future work will include microstructural analyses to support and refine the interpretation of damage evolution.

Validation results for 22MnB5 are shown in **Fig.5**. In general, good agreement is achieved for each specimen shape. For the uniaxial tensile specimen (triaxiality 0.33) and the 0° tensile specimen (triaxiality 0.05) there was a deviation for simulation in the range with transition between elastic and plastic range. This, in turn, could not be directly mapped

with some idealised boundary conditions such as machine stiffness or friction due to clamping of the sample. However, each of the selected specimens showed good agreement at the time of fracture with experiment, which is of higher significance with regard to damage modelling.

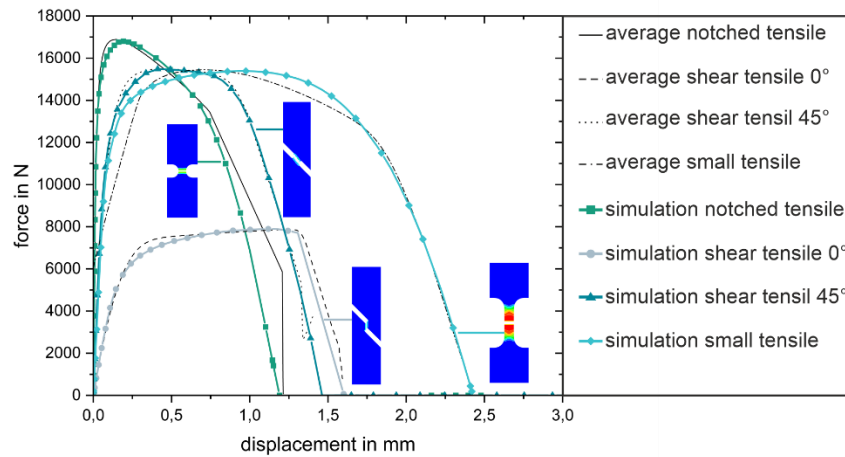


Figure 5: Comparison of the force-displacement curves from experiment and simulation for 22MnB5 (2 mm) using the LS-OPT optimisation algorithm for tensile test, notched tensile test, 0° shear test and 45° shear test

For the more ductile EN AW-5754 aluminum alloy the agreement of experimentally and numerically determined curves was better than for 22MnB5, as shown in Figure 6. Interesting in this case is that in the notch tensile test (triaxiality 0.66) where the general trend of the curve showed good agreement, a slight delay in the transition to failure occurred. This might be improved by a more precise definition of critical strain in future work. Overall, the optimisation algorithm provided good agreement for this material.

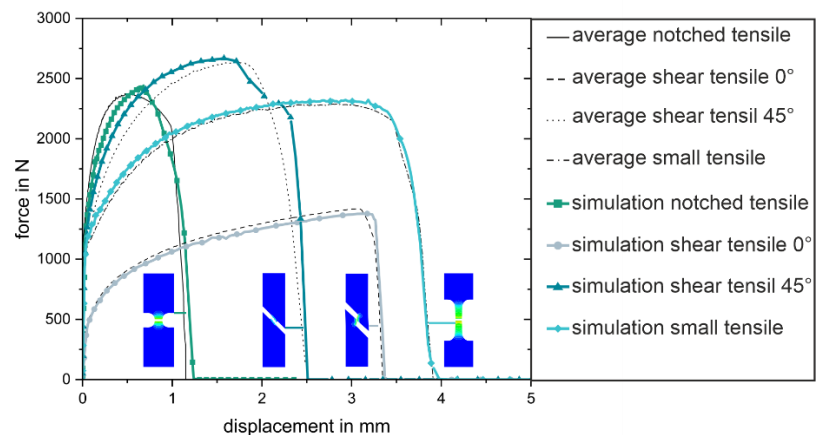


Figure 6: Comparison of the force-displacement curves from experiment and simulation for EN AW-5754 (2 mm) using the LS-OPT optimisation algorithm for tensile test, notched tensile test, 0° shear test and 45° shear test

4 Conclusion

In this study, the GISSMO damage parameters for the aluminium alloy EN AW-5754 and the steel 22MnB5 were calibrated with LS-OPT (**Fig.7**). The critical strains and failure strains were determined as functions of stress triaxiality to allow a material-dependent description of the damage and failure mechanisms. The results show that both the critical strain (KD - ECRIT) and the failure strain (VD - LCSDG) exhibit a clear dependence on the stress triaxiality. The determined strain curves show that:

- **EN AW-5754** shows a significant increase in elongation at break at higher triaxialities, which indicates ductile behaviour with a strong strain hardening effect.
- **22MnB5** shows a stronger tendency towards localisation, whereby the failure strain decreases significantly at medium triaxialities, which is typical for high strength but less ductile steels.

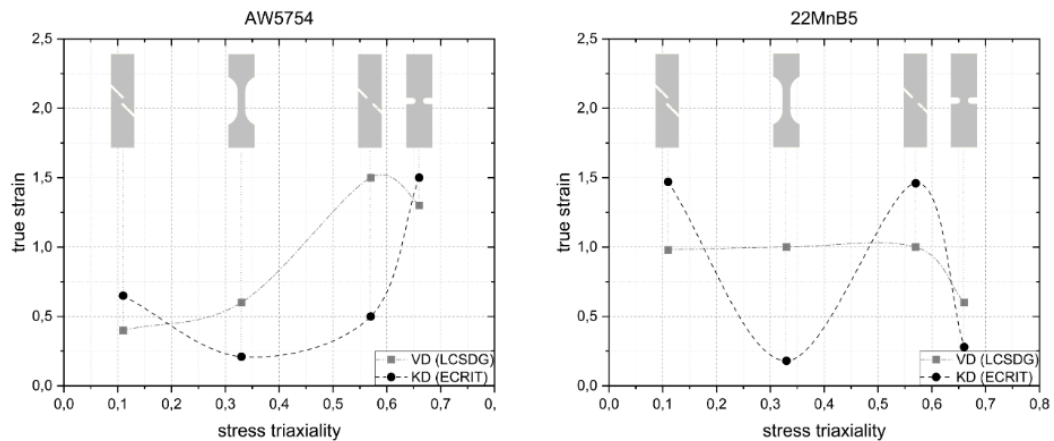


Figure 7: The resulting GISSMO parameters for the critical strain and the failure strain of the analysed materials (22MnB5 and AW5754)

As a final phase of the investigation of the damage parameters for 22MnB5 steel, a numerical 2D simulation was performed in LS-DYNA, as shown in **Fig. 8**. The element size in the cutting zone was 4 μm , and a high-speed blanking process was simulated with a punch speed of 10 m/s. The resulting fracture pattern showed the typical S-shape for this process and was in good agreement with the previously known experimental results. This indicates good agreement in the application of the GISSMO models for the simulation of high-speed blanking processes.

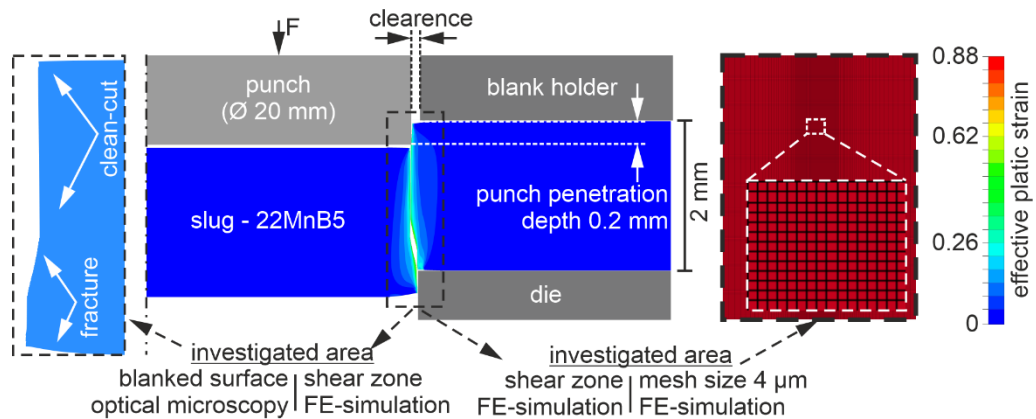


Figure 8: Numerical 2D simulation of the high-speed blanking process (HSB) of 22MnB5 using calibrated GISSMO damage parameters

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Analysis of Shear Localization in High-Speed Blanking of Press-Hardened 22MnB5 Steel

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Abstract

High-speed blanking induces a shear localization in high-strength steels, improving the quality of the blanked edge in terms of geometrical shape, separation surface roughness, and shear-influenced zone close to the cut edge. High-speed blanking experiments with a 20 mm punch diameter and 8 m/s velocity of 2 mm thick 22MnB5-sheets (ultimate tensile strength of 1550 MPa) utilizing a relative clearance of 2.5 % are carried out. Two types of shear bands are observed in the post-process: A 'deformation shear band,' characterized by elongated grains in shear direction, is distinguished from a 'transformation shear band' with a change in microstructure appearing as a white layer in optical microscopy. Previous investigations by transmission electron microscopy showed that this white layer consists of nanosized grains in 22MnB5 material. Light microscopic images show that the transformation shear band width varies over the sheet thickness. Independent of the occurrence of a transformation shear band, a localization with a width of less than 20 μm is observed along the blanked edge. Fractographic analysis by scanning electron microscopy shows no evidence of brittle fracture. Characteristic areas of the separation surfaces coincide with the observed two types of shear bands at the blanked edge.

Keywords

High-speed blanking, Shear localization, High-strength steels, Microstructure

1 Introduction

High-speed blanking with punch speeds in the range of 5 to 12 m/s, two orders of magnitude higher than conventional blanking, can achieve high-quality blanked edges even for high-strength steels. High-speed blanked edges show a high geometric straightness, low rollover, and almost no burr formation (Landgrebe et al., 2017). High-speed blanking tests with punch speeds of 5.3 m/s of the three high-strength steels 22MnB5-press hardened, C75S-hardened, and a bainitic steel HR660Y760T-CP, were performed by Schmitz et al. (2020). Analyzing the microstructure at the blanked edge for 22MnB5-press hardened and C75S-hardened, a change in microstructure occurs at the blanked edge. The so-called transformation shear bands in those two materials have nanocrystalline microstructures, wherefore the grain boundaries cannot be etched and appear white in light microscopic imaging (Schmitz et al., 2020). A nano-scaled equiaxed narrow band in certain areas of the surface is also observed at the high-speed blanked edge of the steel alloy 42SiCr after quenching and partitioning (Birnbbaum et al., 2024). In HR660Y760T-CP, no transformation shear band is observed; nonetheless, a very localized deformation with strongly elongated grains in the shear direction – a deformation shear band – occurs (Schmitz et al., 2020). Investigating two different states of the carbon spring steel C75S - reference state of 330 HV10 and in a hardened state with 450 HV10 - it is shown that for the reference state, a deformation shear band forms, whereby in the hardened state, a transformation shear band can be observed. Independent of the formation of either a deformation or transformation shear band, a high-quality blanked edge in terms of geometric accuracy is produced for both material states by high-speed blanking (Schmitz et al., 2024). An influence of punch speed, impact energy, and blanking clearance on the formation of the transformation shear band is shown in the high-speed blanking of 22MnB5-hardened by Winter et al. (2021).

The formation of a localized deformation shear band indicates ductile behavior even for high-strength steels with low formability. In this study, the separation mechanisms in high-speed blanking of 22MnB5-press hardened are analyzed by scanning electron microscopy analysis of the blanked edge surface. The characteristic areas of the separation surface, showing ductile separation mechanisms, are found to coincide with the two types of shear bands - transformation and deformation - at the blanked edge.

2 Methods

2.1 Material

The 22MnB5 sheet material was heat-treated in a manner similar to the press hardening of ultra-high-strength steel components. The 2-mm-thick sheets were austenitized in a chamber furnace at 930 °C for 6 minutes. Afterwards, the material was rapidly quenched using a water-cooled tool for flat sheets to induce a martensitic microstructure. The material's successful hardening is demonstrated by a Vickers hardness of 480 HV (HV 10) and a measured ultimate tensile strength of 1550 MPa.

2.2 High-speed blanking test setup

High-speed blanking experiments are carried out on a hydraulic impact press, ADI-AClip1000J® of MPM France (**Fig. 1**). To achieve high punch velocities in a range of 5 to 10 m/s, a hydraulic piston of the machine is accelerated by discharging a hydraulic accumulator with an initial pressure of 300 bar. After a free acceleration of 100 mm, the piston hits the tool, accelerating the punch, which is laying directly on the sheet at the beginning of the process. In this investigation a machine setting which achieves 8 m/s punch speed is used, determined by acceleration measurements (see Fig. 1). High-speed blanking tests are carried out with a punch diameter (d_{punch}) of 20 mm in combination with a die diameter (d_{die}) of 20.1 mm resulting in a relative clearance $c_{\text{rel}} = (d_{\text{die}} - d_{\text{punch}})/(2 \cdot s_0)$ of 2.5 % for the 2 mm thick material.

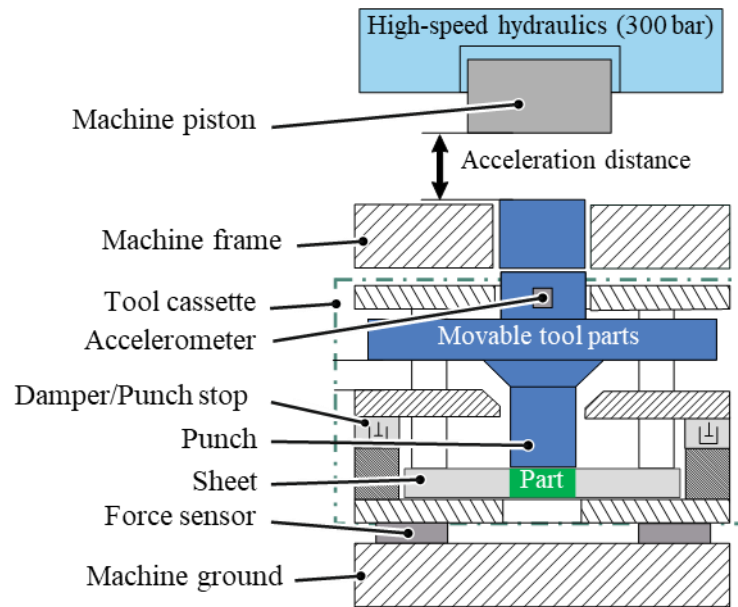


Figure 1: Working principle of the hydraulic impact press AdiaClip 1kJ, adapted from (Schmitz et al., 2024)

3 Results

3.1 Microstructure analysis at the blanked edge

After blanking, the blanked slug is molded, polished, and etched for light-microscopic imaging. **Fig. 2** shows the microstructure in the shear-affected zone in the form of a shear band along the blanked edge. Dodd and Bai (2012) defined a shear band as areas in the microstructure in which a transition from a homogeneous to a localized strain distribution has taken place. Thereby, according to Rogers (1979) two different types of shear bands: a transformation and deformation shear band can be identified at the high-speed blanking edge

(Fig. 2). The deformation shear band is characterized by strongly concentrated plastic deformation, in this case characterized by elongated and deformed grains in the shearing direction. A transformation shear band appears white in light microscopy, indicating a change of microstructure. Microstructure changes could be in the form of grain rotation, dynamic recrystallization, Amorphisation, formation of martensite, or heat-treated martensite (Yan et al., 2021). In high-speed blanking of press-hardened 22MnB5 with similar process parameters than in this investigation Schmitz et al. (2020) showed that the transformation shear band has a nanocrystalline microstructure. For the given process parameters and material of this investigation, the following types of bands can be observed at the blanked edge of the slug: A transformation shear band is induced at approximately one-third of the sheet thickness, whereby its width varies along the thickness direction. A deformation shear band can be observed at nearly two-thirds of the sheet thickness. The maximum width of the deformation shear band is 20 μm from the edge, whereas a maximum width of 9 μm for the transformation shear band was measured based on optical observation in the light-microscopic images (Fig. 2).

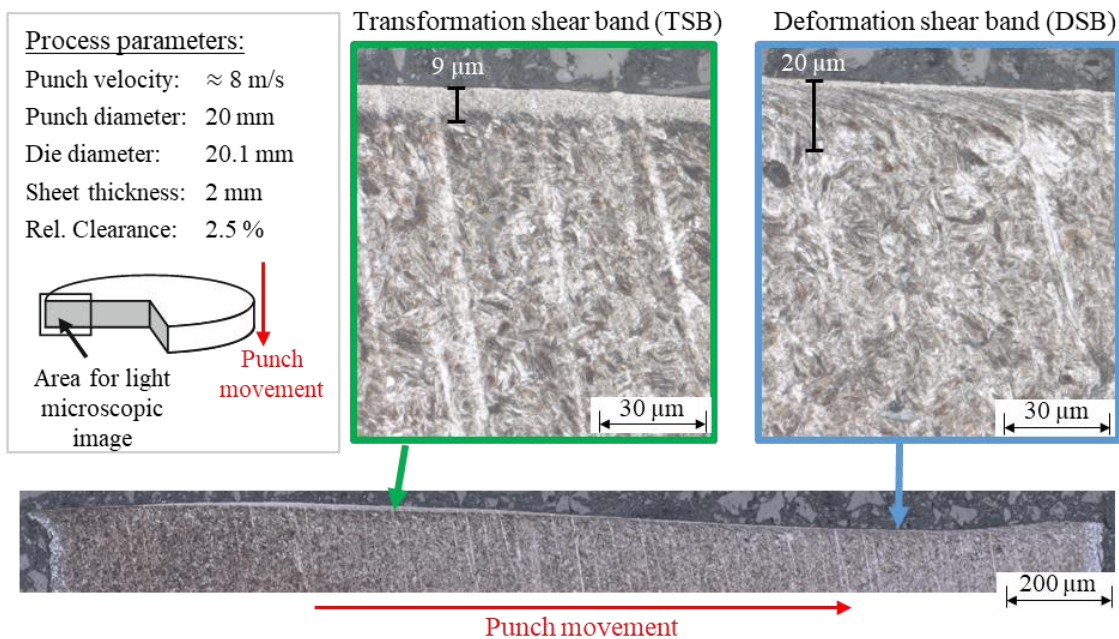


Figure 2: Microstructure analysis of the blanked edge

Looking at the blanked edge of the sheet, the deformation and transformation shear band appears point-symmetric compared to the blanked edge of the slug. This means that a deformation shear band is observed at approximately one-third of the sheet's thickness, from the top towards the middle. A transformation shear band is observed at approximately two-thirds of the sheet thickness at the bottom. Consequently, the zone affected by shear is limited to a band of deformation and transformation (respectively grain refinement) with a width of 1% of the sheet thickness. The band's width is smaller than the blanking clearance of 50 μm .

3.2 Fractographic analysis by Scanning Electron Microscopy (SEM)

Fractographic analysis by scanning electron microscopy (SEM) shows a smooth fracture surface with a homogeneous, even fracture surface without cracks or break-outs (**Fig. 3a**). Three characteristic areas can be identified with further magnification. At the top of the blanked edge a fine shear surface is observed (**Fig. 3 b**). Qualitatively, this area coincides well with the occurrence of the transformation shear band in the cross-section view. This area merges continuously into an area characterized by ductile dimples elongated in the shear direction (**Fig. 3c**). This characteristic fracture surface is observed from approximately the middle of the sheet thickness towards the bottom. It coincides with a deformation shear band in the cross-section view. At the very bottom of the sheet, a smaller area with a less smooth, rather unclear fracture surface is observed (**Fig. 3d**). This surface area has undergone a frictional contact with the die during the process, wherefore the appearance is interpreted as a ductile fracture which is then smeared by friction in the shearing direction.

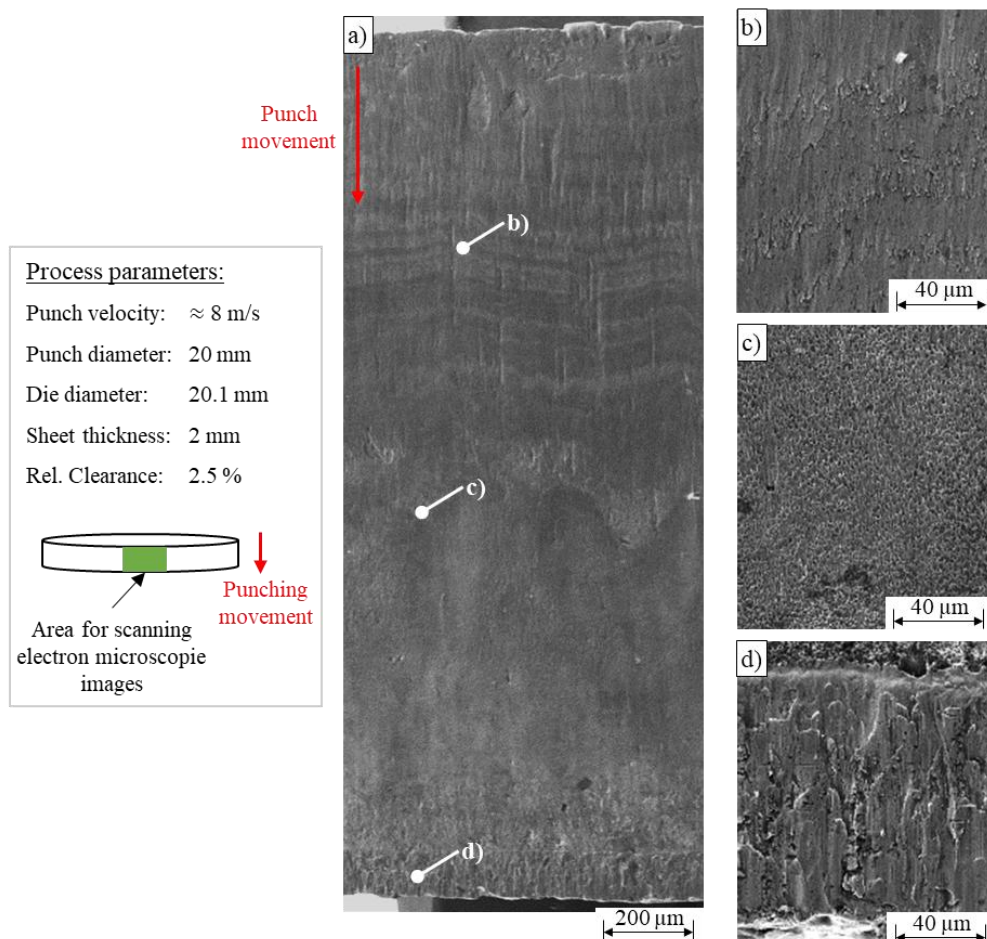


Figure 3: Scanning Electron Microscopic (SEM) analysis of fracture surface: a) Overview images of the blanked edge, and b-d) detail images of the three characteristic areas

4 Discussion

Even if a press-hardened 22MnB5 has a very low formability, the fractographic analysis shows that a ductile fracture mechanism is triggered in high-speed blanking. No evidence for brittle failure of the material could be found. Remarkable is also the concentration of the shear-affected zone to a very narrow band with a width of less than 1 % of the sheet thickness. Within the localized deformation, two types of shear bands occur: a deformation shear band and a transformation shear band. The transformation shear band, which has a nanocrystalline microstructure, coincides with a smooth shear surface at the top of the sheet. A shear-tensile separation surface, coinciding with a deformation shear band, indicates a superposition of tensile stresses to the dominant shear loads. A frictional contact with the die influences the third characteristic surface area at the bottom of the blanked edge.

Two explanations coexist for the formation of the localized shear band. On the one hand, it could be assumed that the temperature increase due to plastic work-to-heat conversion is leading to thermal softening, causing a localization (Klepaczko, 1998). The prerequisite for thermal softening is that high-speed blanking takes place under adiabatic conditions - Fourier number less than 0.01 - due to the short process time and the low thermal conductivity of steels (Schmitz, 2024). In contrast, it could be assumed that dynamic recrystallization (DRX) and its corresponding microstructural evolution lead to localized material softening prior to thermal softening (Rittel et al., 2008). New grain formation and grain rotation in DRX reduce the dislocation density, leading to a macroscopically reduced material flow stress (Yan et al., 2021). Unclear is whether one or both mechanisms are triggered, or if both are triggered in parallel or subsequently. Independent of the mechanisms for triggering the localization, the material's ductility is increased in the shear band. Therefore, both explanations could explain a ductile shear, respectively, tensile-shear separation mechanism at the whole blanked edge.

The formation of a transformation shear band could be explained by the mechanism of DRX, independent of whether DRX causes the localization or if DRX happens within a localization. DRX requires the subsequent steps of: dislocations rearrangement into elongated dislocation cells (i. e. dynamic recovery), formation of elongation subgrains, break up of subgrains into approximately equiaxed micrograins and grain rotation (Meyers et al., 2000). According to DRX mechanisms, a deformation shear band indicates a stage at which DRX has not been completed, for example, if it has been stopped during dislocation rearrangement. The steps of DRX are completed within a transformation shear band. The shear-tensile separation at the deformation shear band indicates that tensile stresses limit the plastic deformation required for completing the DRX. Nonetheless, the type of shear band coincides with characteristic surface areas of the blanked, wherefore the formation mechanisms and the corresponding separation mechanisms are most likely dependent on each other.

5 Conclusion

Fractographic analysis of the high-speed blanked edges draws a correlation between the microstructural observations of shear bands in the cross-section and the characteristics of the separation surface. A smooth shear surface matches the appearance of a transformation shear band, which is characterized by a nanocrystallized microstructure. A shear-tensile separation surface with ductile dimples elongated in the shear direction coincides with a deformation shear band characterized by elongated grains in the shearing direction. The deformation shear band could be seen as a previous stage of the transformation shear band, within the completion of the dynamic recrystallization is hindered by evolving tensile stresses in the process leading to material separation prior to the formation of the nanocrystallized microstructure. Nonetheless, for both bands, the maximum width is 20 μm , wherefore the shear-affected zone is less than 1 % of sheet thickness and also smaller than the blanking clearance. A third characteristic surface area with ductile fracture smeared into the shearing direction by the frictional contact with the die is identified at the bottom of the slug.

In order to further enhance the quality of high-speed blanked edges in terms of a smooth, homogeneous separation surface and a minimal shear-affected zone close to the blanked edge, a transformation shear band needs to be formed over the whole sheet thickness. Evolving tensile loads, judged by the separation surface, at the bottom of the sheet might be the reason for the formation of a deformation instead of a transformation shear band since material separation happens at lower strains.

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