

Deformation-induced damage in finitely-strained metal sheets: Large-area damage quantification and mechanical property assessment

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Abstract

Reliable identification of damage-induced property changes in formed metal sheets remains challenging because plastic deformation alters the mechanical response through several coupled mechanisms. Moreover, pronounced ductile damage evolution occurs predominantly under inhomogeneous states of deformation, e.g., strain localization, which compromises the characterization of potential effects of damage on macroscopic mechanical properties, e.g., elastic stiffness. In this study, damage in dual-phase steel DP800 sheets is isolated from other concurrent mechanisms using a novel “sheet extrusion” process. The flat, finitely-strained sheets exhibit large regions with homogeneous properties, enabling subsequent characterization of damage-induced property changes. Damage in the form of microscopic voids is then characterized by large-area scanning electron microscopy (SEM) measurements, and damage-induced property changes are determined by subsequent characterization tests on specimens extracted from the extruded sheets. It is found that deformation-induced damage does not measurably affect the apparent elastic stiffness or the pre-necking tensile response. In contrast, differences in fracture-related performance quantities, i.e., the remaining ductility and the tensile impact energy, indicate that higher deformation-induced damage promotes earlier macroscopic failure by increasing the likelihood of local void coalescence during subsequent loading.

Keywords: DP800, sheet metal forming, damage-induced property changes, large-area SEM scans, voids

1. Introduction

In order to fully exploit the resource-efficient and lightweight potential of formed components, current component design approaches consider not only the local work hardening and residual stresses after forming but also microscopic ductile damage (Tekkaya et al., 2020). Damage is understood in this work as the nucleation, growth, and coalescence of microscopic voids in the microstructure. Even before leading to premature failure, damage in successfully formed parts is associated with a degradation of product performance, e.g., stiffness, impact toughness, and fatigue life (Tekkaya et al., 2020; Gitschel et al., 2023). Therefore, accurate characterization and prediction of damage, as well as an understanding of the link between damage and product performance, are important to exploit a component’s full potential.

Damage in dual-phase steels is commonly characterized through inspection of micrographs acquired by scanning electron microscopy (SEM) (Tasan et al., 2014; Hoefnagels et al., 2015; Saeidi et al., 2016). Its spatial resolution is substantially higher than that of X-ray computed tomography, and void areas as small as $0.02 \mu\text{m}^2$ can be detected (Kusche et al., 2021). However, SEM-based measurements are often restricted to relatively small observation areas due to time-consuming scanning procedures. Kusche et al. (2019) addressed this limitation by combining panoramic imaging with automated machine learning-based classification of damage mechanisms. More recently, Reclik et al. (2026) improved the detection efficiency of rare microstructural features, such as void areas, by employing a strategy based on initial fast scans followed by selective rescanning of regions of interest.

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Quantifying the contribution of void evolution to property changes remains difficult because plastic deformation affects the material response through several concurrent mechanisms. For instance, numerous studies indicate that the elastic stiffness degradation originates from microplasticity rather than from damage, i.e., dislocation motions below the macroscopic yield stress (Arechabaleta et al., 2016; Chen et al., 2016; Torkabadi et al., 2018; Li and Wagoner, 2021; Chamakura et al., 2022). In contrast to irreversible void evolution, this contribution is partly reversible and typically saturates with increasing plastic deformation.

This distinction is particularly relevant for continuum damage mechanics, where elastic stiffness degradation is often attributed to a reduction of the effective load-bearing area caused by void evolution (Lemaitre, 1996; Murakami, 2012). The model may reproduce the observed softening while assigning it to the wrong internal mechanism. Consequently, internal damage variables cannot be interpreted as direct measures of void-related property changes unless the corresponding damage-induced effects have been isolated experimentally.

Experimentally-oriented studies have addressed this issue by manufacturing components that differ only in their damage state and subsequently characterizing specimens extracted from these components. Hering and Tekkaya (2020) reduced damage in cold extruded parts by controlling the hydrostatic stress state, resulting in increased high-cycle fatigue life and improved crash absorption in Charpy impact tests. Similar performance improvements were reported for bent sheet parts when a superposed radial stress was applied during forming (Meya et al., 2019b). In contrast, Gitschel et al. (2025) reduced damage by controlling the deviatoric stress state and showed that a more shear-dominated loading path induces directionally dependent damage, leading to a slight anisotropy in impact toughness.

In sheet metal forming, however, the isolation of damage from other concurrent mechanisms has so far been demonstrated only for three-point bending (Meya et al., 2019b). Although damage is successfully separated in the bent sheets, the sheets exhibit spatially inhomogeneous mechanical properties, preventing the determination of global properties, e.g., elastic stiffness, in subsequent characterization tests. A methodology that isolates damage at large plastic strains and, at the same time, enables reliable quantification of damage-induced changes in mechanical properties is still missing. To this end, we employ a so-called sheet extrusion process (Kolpak et al., 2021) to manufacture finitely-strained sheets with different damage states (Fig. 1). The damage state is controlled by applying different levels of superposed counterpressure during extrusion, while the imposed extrusion strain is kept constant. Finite element simulations with a fully coupled, gradient-enhanced ductile damage model developed by Sprave and Menzel (2020) are used to evaluate the strain and residual stress states after extrusion and specimen extraction. The damage variable of the ductile damage model implicitly represents the degradation of the elastic stiffness tensor. Consequently, it cannot be directly related to any microscale quantity. Hence, the predicted damage distributions are compared qualitatively with microscale damage metrics determined from the SEM measurements, which are acquired by applying the previously introduced state of the art efficient high-resolution analysis across the large samples (Reclik et al., 2026). Subsequently, tensile and notched-tensile impact tests are performed on specimens extracted from the extruded sheets. Based on this experimental separation, mechanical properties that are affected by deformation-induced damage are identified.

2. Materials & methods

2.1. Materials

The investigated sheet material is a dual-phase steel DP800 with a sheet thickness of $t_0 = 1.5$ mm. Its chemical composition is listed in Tab. 1 (Pütz et al., 2020). The microstructure comprises a ferritic matrix with homogeneously distributed martensite islands in the e_2 - e_3 plane (RD-TD plane) (Fig. 2). A case-hardened steel 16MnCrS5 is selected for the half-cylindrical supporting billets, because its hardening behavior is comparable to that of the DP800, preventing relative motion between the sheet and the billet halves during extrusion. The assembled workpiece is shown in Fig. 1; the flow curves of both materials are shown in the appendix in Fig. A1.

Table 1: Chemical composition of the dual-phase steel DP800 in mass-%. Data taken from Pütz et al. (2020).

C	Si	Mn	Cr	Mo	Cu
0.15	0.21	1.67	0.73	0.01	0.044

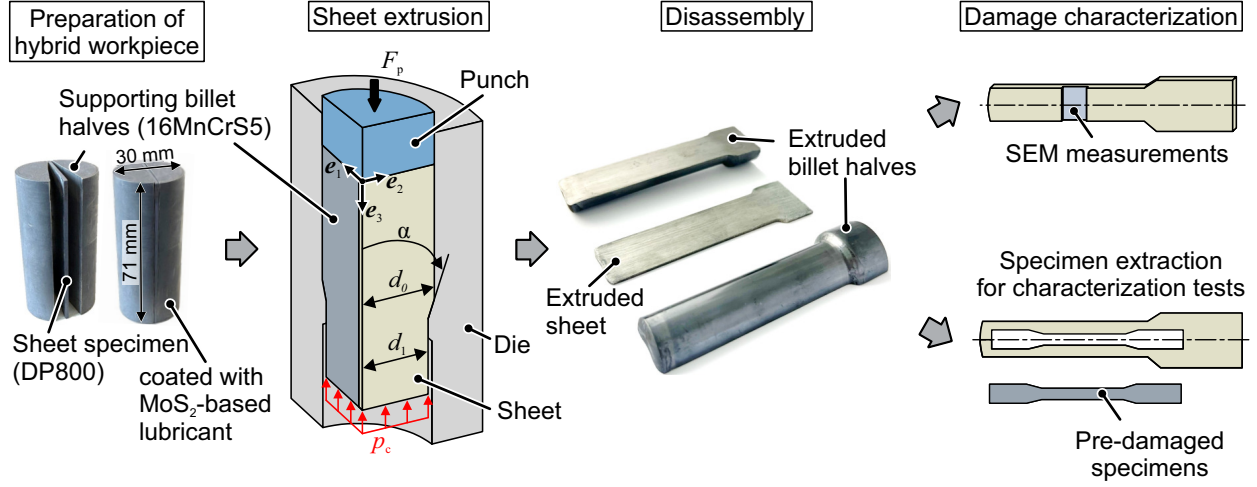


Figure 1: Methodology to investigate the isolated influence of ductile damage on mechanical properties in sheet metal forming. The stress state is controlled by applying a superposed counterpressure p_c during extrusion. Damage in the extruded sheets is characterized by means of large, high-resolution panoramic SEM images, and damage-induced changes in mechanical properties are assessed based on subsequent characterization tests.

2.2. Sheet extrusion

In this work, we employ the sheet extrusion process to manufacture components with different levels of damage. The initial sheet specimen is positioned between two half-cylindrical billets and the assembled workpiece is cold-extruded. To control the stress state during plastic deformation, different levels of constant axial counterpressure p_c are applied via an ejector during the extrusion step. A constant extrusion strain

$$\varepsilon_{\text{ex}} = 2 \ln \left(\frac{d_0}{d_1} \right), \quad (1)$$

where d_0 is the initial diameter of the assembled workpiece and d_1 is the reduced diameter after extrusion, is chosen. Consider a Cartesian basis $\{e_1, e_2, e_3\}$ with e_3 aligned with the extrusion axis. Assume process symmetry with respect to the e_1 - e_3 and e_2 - e_3 planes (cf. Fig. 1), a perfect billet-sheet bond, i.e., that no relative slip or separation occurs, and an isotropic, rate-independent von Mises plasticity material with flow stress σ_f and associated flow. An applied counterpressure affects only the hydrostatic part σ_h and increases the wall contact pressure additively, i.e., $p_w = p_w^{p_c=0} + p_c$, provided that the contact conditions are uniform in the axial (extrusion) direction. The deviatoric stress state, and thus the von Mises equivalent stress $\sigma_{\text{eq}} = \sigma_f$, remains unchanged. Hence, the stress triaxiality, a commonly used stress measure in damage mechanics to describe the rate of void growth, can be formulated based on the superposed counterpressure as

$$\eta = \frac{\sigma_h}{\sigma_{\text{eq}}} = \frac{1}{3} - \frac{p_w^{p_c=0} + p_c}{\sigma_f}. \quad (2)$$

The expression shows that the stress triaxiality decreases with increasing counterpressure. Because a decrease in triaxiality reduces void nucleation and growth (Tekkaya et al., 2020; Gitschel et al., 2023), applying different levels of counterpressure enables control of damage evolution at equal work hardening, without changing the deviatoric stress state.

2.2.1. Experiments

The sheet extrusion experiments shown in Fig. 1 are conducted on a triple-acting hydraulic press (SMG HZPUJ 260/160) with a maximum force of 2600 kN. The setup includes an additional counterpunch to apply a counterpressure during extrusion and subsequently eject the extruded part. For the preparation of the hybrid workpiece, the billet is cut in half by a saw blade with the same thickness as the sheet, $t_0 = 1.5$ mm. The hybrid workpiece has an initial

length of $l_{wp} = 71$ mm and an initial diameter of $d_0 = 30$ mm. A shoulder opening angle of $2\alpha = 90^\circ$ and an extrusion strain of $\varepsilon_{ex} = 0.5$ (i.e., a final diameter of $d_1 = 23.36$ mm) are chosen. The outer surface of the assembled hybrid workpiece is coated with the MoS₂-based lubricant Beruforge 191 by the manufacturer Carl Bechem GmbH. The experiments are conducted at room temperature with a constant punch velocity of 10 mm/s. Extrusion experiments without counterpressure and with superposed counterpressures of 35% and 70% of the initial flow stress of DP800, with $\sigma_{f,0} = 485$ MPa, are investigated. Note that because the counterpressure only alters the hydrostatic component of the stress state and does not lead to additional plastic deformation, the final rod diameter, and hence Eq. 1, remains the same in all experiments.

2.3. Numerical investigation of sheet extrusion

For the characterization of damage-induced property changes, specimens are required that have homogeneous, known properties and differ only in their state of damage. Therefore, the mechanical properties of the extruded sheets resulting from the sheet extrusion process are assessed using finite element damage simulations. Based on the predicted strain and residual stress distribution in the extruded sheets, the extraction geometry for the specimens used for the characterization of damage-induced property changes is defined.

2.3.1. Constitutive model

A large strain gradient-enhanced ductile damage model with a full coupling between isotropic damage and isotropic plasticity is employed. The detailed finite element implementation and the algorithmic solution scheme can be found in Sprave and Menzel (2020). In the following, a brief description of the constitutive model is presented.

Let us assume a multiplicative decomposition of the deformation gradient $\mathbf{F}$ into an elastic part $\mathbf{F}_e$ and a plastic part $\mathbf{F}_p$, so that $\mathbf{F} = \mathbf{F}_e \cdot \mathbf{F}_p$ with $J_e = \det(\mathbf{F}_e) > 0$. The elastic Finger tensor $\mathbf{b}_e := \mathbf{F}_e \cdot \mathbf{F}_e^T$ defines the deformation measure, where $\bullet^T$ denotes the transpose of a second-order tensor. Subsequently, the elastic logarithmic strain tensor can be defined in spatial form with

$$\boldsymbol{\varepsilon}_e = \frac{1}{2} \ln(\mathbf{b}_e) = \boldsymbol{\varepsilon}_e^{\text{vol}} + \boldsymbol{\varepsilon}_e^{\text{iso}}, \quad (3)$$

and can be decomposed additively into a volumetric part $\boldsymbol{\varepsilon}_e^{\text{vol}} = 1/3 \text{tr}(\boldsymbol{\varepsilon}_e) \mathbf{I}$ with $\mathbf{I}$ the second-order identity tensor and an isochoric part $\boldsymbol{\varepsilon}_e^{\text{iso}}$.

The model is available as an Abaqus user subroutine UMAT and consists of a von Mises yield surface with isotropic hardening coupled to an isotropic Lemaitre-type damage formulation. In order to regularize the damage model and to obtain mesh-objective finite element results, a micromorphic approach according to Forest (2009) is employed. This approach introduces gradient contributions of a nonlocal damage field ϕ , which is coupled to the local damage field D and thus enables the evolution equation for damage to be solved on a local level. The damage degradation functions

$$f^\bullet : \mathbb{R}_0^+ \rightarrow]0, 1], \quad f^\bullet(D) = \exp(-\gamma_r \xi_\bullet D), \quad \bullet \in \{\text{vol}, \text{iso}\}, \quad (4)$$

degrade the effective elastic properties, where γ_r (denoted as η_r in Sprave and Menzel (2020)) controls the rate of degradation. The index $\bullet$ specifies the volumetric and isochoric contributions, i.e., f^{vol} and f^{iso} , respectively, while the corresponding material parameter $\xi_\bullet$ weights the relative degradation magnitude.

The Helmholtz free energy potential is formulated as an additive split $\Psi = \Psi^{\text{loc}} + \Psi^{\text{nloc}}$ with the local part Ψ^{loc} and the nonlocal part Ψ^{nloc} . The local part is expressed by

$$\Psi^{\text{loc}}(\boldsymbol{\varepsilon}_e, D, \bar{\varepsilon}^p) = \frac{K}{2} f^{\text{vol}}(D) [\text{tr}(\boldsymbol{\varepsilon}_e)]^2 + G f^{\text{iso}}(D) \boldsymbol{\varepsilon}_e^{\text{iso}} : \boldsymbol{\varepsilon}_e^{\text{iso}} + \frac{H}{n_p + 1} (\bar{\varepsilon}^p + \varepsilon_0)^{n_p + 1} \quad (5)$$

with the equivalent plastic strain $\bar{\varepsilon}^p$, the bulk modulus K , the shear modulus G and the hardening parameters H , ε_0 and n_p . Regularization of the local damage field is introduced by the nonlocal energy contribution

$$\Psi^{\text{nloc}}(\phi, \nabla_X \phi, D) = \frac{c_d}{2} \|\nabla_X \phi\|^2 + \frac{\beta_d}{2} (\phi - D)^2, \quad (6)$$

where $\nabla_X \phi$ denotes the spatial gradient of the nonlocal damage variable. The parameters c_d and β_d represent the regularization and penalty parameters, respectively. As driving forces, the Mandel stress $\mathbf{m} := 2 \partial_{\mathbf{b}_e} \Psi \cdot \mathbf{b}_e$ and the

damage driving force $q := -\partial_D \Psi$ are employed. Additionally, an isotropic hardening stress $\beta := -\partial_{\bar{\varepsilon}^p} \Psi$ is introduced. The plastic dissipation potential Φ^p is formulated as

$$\Phi^p(\mathbf{m}_{\text{eff}}^T, \beta) = \|\text{dev}(\mathbf{m}_{\text{eff}}^T)\| - \sqrt{\frac{2}{3}} [p_g - \beta], \quad (7)$$

where p_g defines a stress-offset. The damage dissipation potential Φ^D is formulated as

$$\Phi^D(q, D) = q_{\text{eff}} - q_{\min} - (q_{\max} - q_{\min}) [1 - f^q(D)]^{n_d}, \quad (8)$$

where $q_{\min}$ and $q_{\max}$ denote a minimum and a maximum damage driving force, respectively. Both dissipation potentials are based on the effective driving forces $\mathbf{m}_{\text{eff}}^T := \mathbf{m}^T / f^m(D)$ and $q_{\text{eff}} := q / f^q(\bar{\varepsilon}^p)$. According to Eq. 4, $f^\bullet$ with $\bullet \in \{\mathbf{m}, \bar{\varepsilon}^p, q\}$ defines the respective damage function. The associative evolution equations are obtained as

$$\dot{\mathbf{r}}^p = \lambda^p \frac{\partial \Phi^p}{\partial \mathbf{m}^T}, \quad \dot{\bar{\varepsilon}}^p = \lambda^p \frac{\partial \Phi^p}{\partial \beta}, \quad \dot{D} = \lambda^D \frac{\partial \Phi^D}{\partial q} \quad (9)$$

and the two dissipation potentials are combined with classical Karush-Kuhn-Tucker conditions with the Lagrange multipliers λ^p and λ^D .

In metal forming, the Cauchy stress tensor $\boldsymbol{\sigma}$ is commonly used as a stress measure. From the relation between the spatial Mandel stress $\mathbf{m}^T$ and the Kirchhoff stress $\boldsymbol{\tau}$ via $\mathbf{m}^T = \mathbf{g} \cdot \boldsymbol{\tau}$, with $\mathbf{g}$ denoting the spatial covariant metric tensor, the Cauchy stress follows as $\boldsymbol{\sigma} = 1/J \mathbf{g}^{-1} \cdot \mathbf{m}^T$ with $J = \det(\mathbf{F})$. Subsequently, the hydrostatic stress quantities commonly used to evaluate the damage evolution in metal forming can be expressed with respect to the Cauchy stress as

$$\sigma_h = \frac{1}{3} \text{tr}(\boldsymbol{\sigma}), \quad \sigma_{\text{eq}} = \sqrt{\frac{3}{2}} \|\text{dev}(\boldsymbol{\sigma})\|, \quad \eta = \frac{\sigma_h}{\sigma_{\text{eq}}}. \quad (10)$$

The deviatoric stress state is depicted by the normalized Lode angle parameter

$$\bar{\theta} = 1 - \frac{2}{\pi} \arccos(\xi) \quad \text{with} \quad \xi = \frac{27}{2} \frac{\det(\text{dev}(\boldsymbol{\sigma}))}{\sigma_{\text{eq}}^3}. \quad (11)$$

2.3.2. Finite element model

The sheet extrusion process is modeled in Abaqus/Standard using implicit time integration. Owing to two symmetry planes, a 3D quarter model is employed (Fig. B1). The sheet and the billet are discretized with eight-node linear rectangular elements with reduced integration (Abaqus element type "C3D8RT"). These elements include a temperature degree of freedom, which is repurposed as an auxiliary field variable for the computation of the nonlocal damage field (Ostwald et al., 2019). The sheet mesh comprises 82769 elements, with five elements through half of the thickness, resulting in a minimum element length of $l_{\min} = 0.15$ mm. The sheet material is modeled by using the gradient-enhanced ductile damage model (cf. Section 2.3.1), whereas the billet is modeled as an elasto-plastic material with isotropic hardening and the die as a linear elastic body. The simulation consists of an extrusion and an ejection step. During extrusion, the punch is displacement-controlled and the counterpressure is applied to the ejector as a Neumann boundary condition. Symmetry constraints are enforced on the corresponding planes. Surface-to-surface contact is defined between the billet halves and the sheet, as well as between the combined workpiece and all tooling components. For all contact interfaces, Coulomb friction with a friction coefficient of $\mu_c = 0.04$ is employed. The process is assumed to be isothermal and rate-independent.

Model parameters of the gradient-enhanced damage model are calibrated by inverse identification using measured force and full-field displacement data from a notched tensile test specimen loaded up to fracture. The identified parameter values are listed in Tab. 2. The isotropic hardening behavior of the billet material is prescribed directly from the experimental flow curve of the 16MnCrS5 (cf. Fig. A1 in the appendix). The Young's modulus is chosen as $E = 207$ GPa for both the billet and the die material.

Table 2: Material parameters and corresponding values of the gradient-enhanced ductile damage model (cf. Section 2.3.1).

Symbol	Description	Value	Unit
E	Young's modulus	206.78	GPa
ν	Poisson's ratio	0.30	–
H	Strength parameter	1272.50	MPa
ε_0	Strain offset	$3.69 \cdot 10^{-5}$	–
n_p	Hardening exponent	0.15	–
p_g	Stress offset	0.00	MPa
$q_{\min}$	Initial damage threshold	0.50	MPa
$q_{\max}$	Damage threshold	1.48	MPa
n_d	Damage exponent	0.67	–
γ_r	Damage rate factor	0.54	–
ξ_{vol}	Volumetric damage factor	1.00	–
ξ_{iso}	Isochoric damage factor	0.53	–
ξ_q	Threshold factor	0.99	–
ξ_m	Effective stress factor	0.30	–
$\xi_{\bar{\varepsilon}^p}$	Coupling factor	0.00	–
c_d	Regularization parameter	349.00	N
β_d	Penalty parameter	178.00	MPa

2.4. Damage characterization

To determine the damage distribution as a function of the counterpressure, rectangular specimens are extracted from the sheet-extruded samples using an electrical discharge machine (EDM) according to Fig. 2. The RD-TD surface, which is perpendicular to the normal direction, is metallographically prepared to ensure a smooth, planar surface suitable for SEM analysis. To this end, all specimens were ground with SiC grinding paper up to a grit size of 4000 and then polished on DAC polishing cloths using water-based diamond suspensions with particle sizes of 3 μm and subsequently 1 μm . A final polishing step with an oxide-particle suspension (OPS) containing 0.25 μm particles was performed to achieve the required surface finish. Finally, the samples were etched in 2 % nital for 6 s, producing a topography that allows the martensitic and ferritic phases to be distinguished.

The polished samples were mounted in the SEM (Tescan, Clara) and imaged with an E-T secondary electron detector at a working distance of 10 mm, an accelerating voltage of 20 kV, and a beam current of 3 nA. A novel scanning-and-re-scanning approach developed in previous research (Reclik et al., 2025, 2026) was used to capture large panoramic images covering a total area of $11.91 \times 0.80 \text{ mm}^2$ per sample (shown as the orange region in Fig. 2), encompassing all possible steady-state stress states. During this process, an initial low-resolution scan (100 μm per 768 pixels) was used to quickly identify candidate damage sites via super-resolution-based up-scaling; the selected candidates were then re-scanned at high resolution (100 μm per 3072 pixel). These high-resolution images were analyzed with a machine-learning-assisted method developed and applied in previous work by Kusche et al. (2019); Medghalchi et al. (2020); Wollenweber et al. (2023), complemented by manual re-classification to remove inclusions and artefacts. The smallest detected void area is 0.012 μm^2 .

The number and area of damage sites were determined and normalized to the observed area, allowing calculation of the key damage metrics used in this work: void area fraction D^{exp} , void number density n_v , and average void area $\bar{a}_{\text{bin}}$. Leveraging the large recorded area makes it possible to show how these metrics vary with distance from the sample center and, consequently, with the different stress states encountered during the forming process.

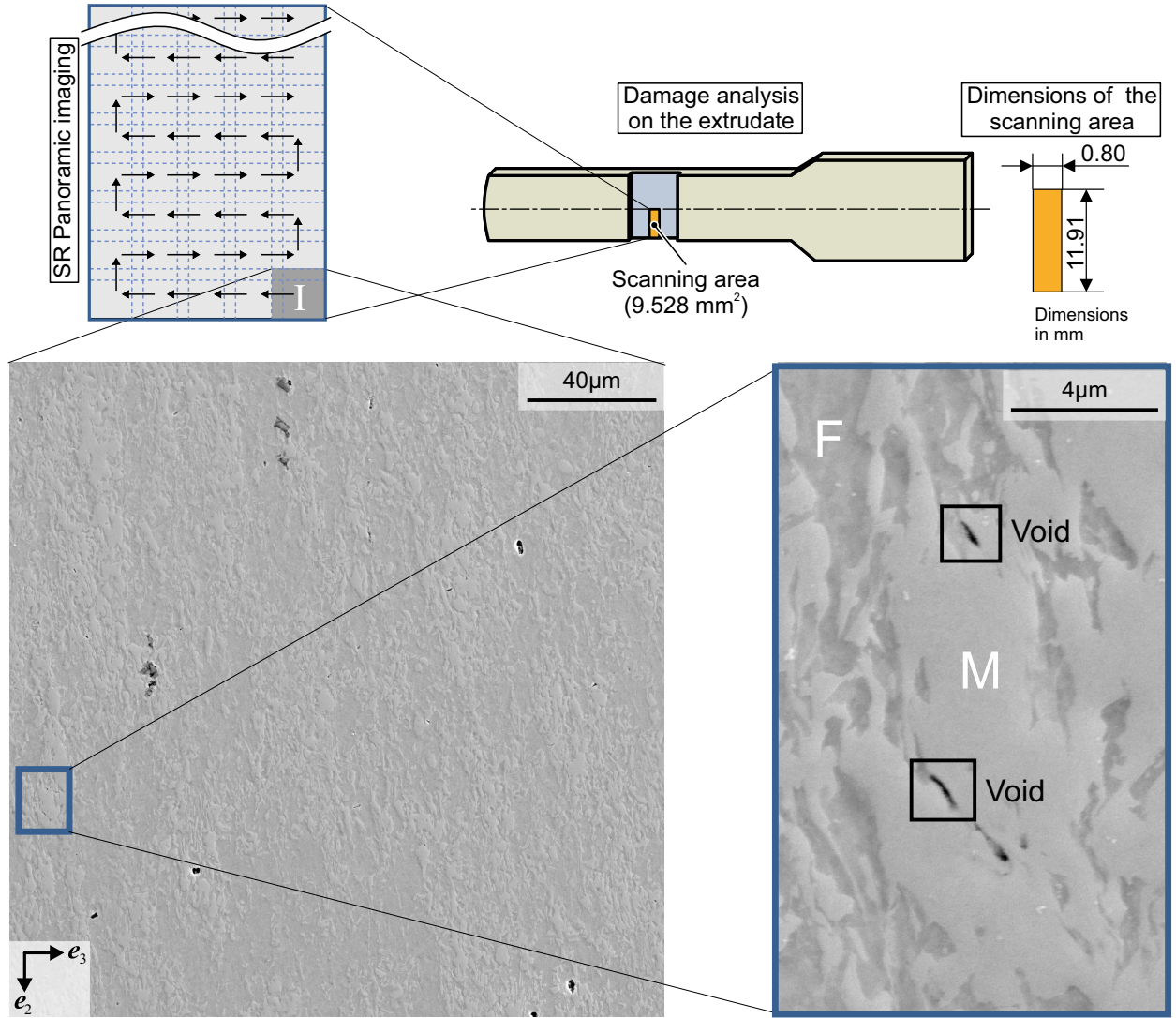


Figure 2: Overview of the automated scanning methodology for void detection. SEM micrographs show the microstructure of DP800 in the e_2 - e_3 plane, which corresponds to the RD-TD plane. Here "F" denotes the ferrite and "M" the martensite.

2.5. Tensile testing

In order to characterize the influence of damage on apparent elastic stiffness, tensile strength and remaining ductility, tensile tests on specimens extracted from the extruded sheets are conducted on a ZwickRoell Z250 universal testing machine. The specimen geometry and the extraction position are shown in Fig. 3a. The experiments are conducted displacement-controlled according to DIN EN ISO 6892-1 (DIN Deutsches Institut für Normung e. V., 2020) with a constant strain rate of 0.0067 s^{-1} . The specimen elongation is measured by using a tactile extensometer with a gauge length of $l_0 = 28 \text{ mm}$. Similar to (Vitzthum et al., 2022), an additional PT1000 temperature sensor is clipped onto the center of the specimen to measure the temperature change related to the Joule-Thomson effect (Thomson, 1857), i.e., a temperature decrease due to volume increase during elastic deformation. The temperature minimum defines the point where the temperature decrease due to elastic cooling equals the temperature increase from plastic heating. This physically motivated criterion is used to define the onset of yielding in this study. Subsequently, the slope of the linear regression between an initial stress value of 25 MPa and the stress point at the onset of yielding in the true stress-strain curve represents the apparent elastic stiffness.

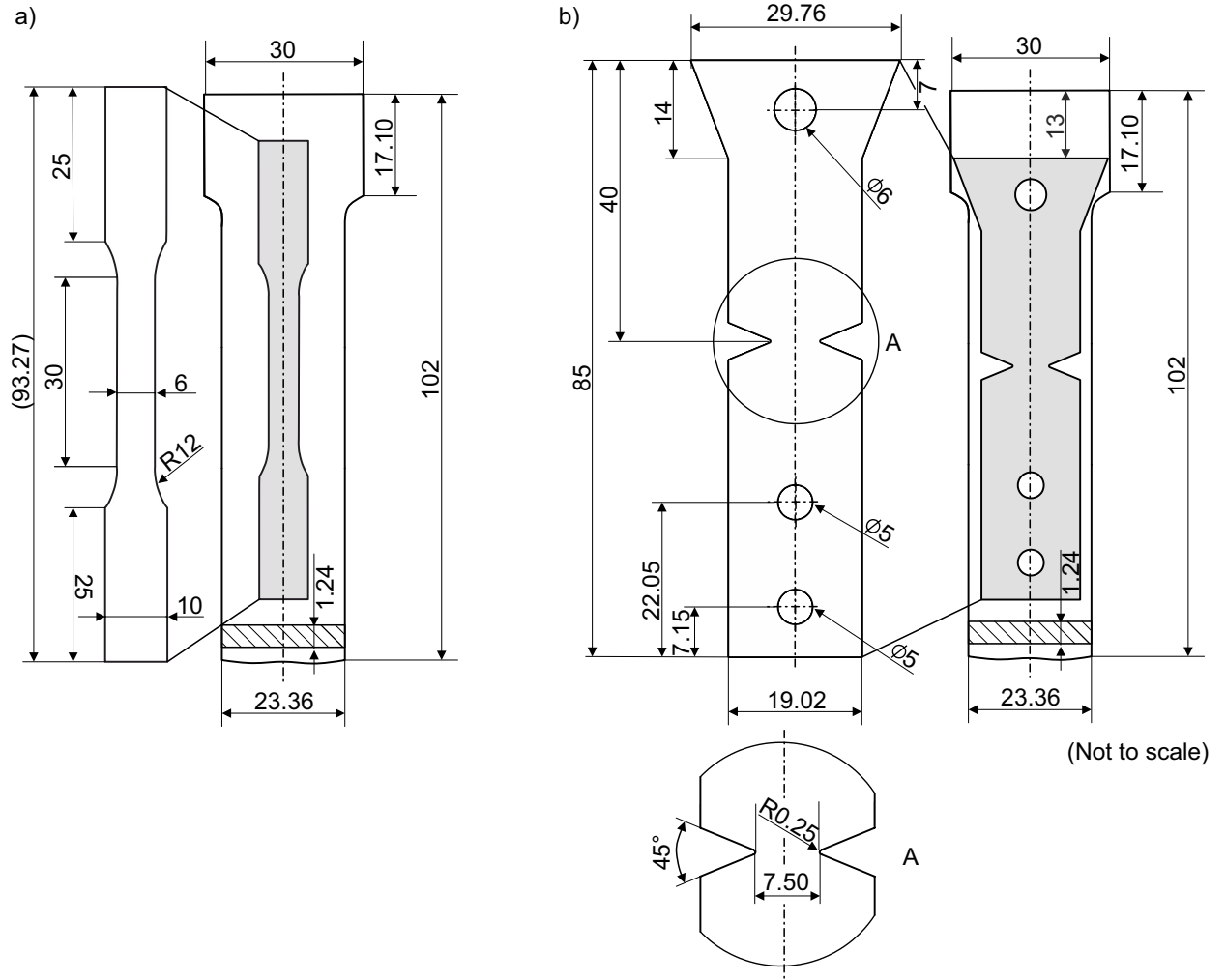


Figure 3: Technical drawings of a) tensile test specimen and b) notched-tensile impact test specimen.

2.6. Notched-tensile impact testing

The influence of damage on impact toughness is assessed using notched-tensile impact tests. The experiments are conducted by the Institute of Metal Forming at RWTH Aachen University. The corresponding test setup follows DIN EN ISO 8256 (DIN Deutsches Institut für Normung e. V., 2024) for plastics. The specimen geometry and the extraction position are shown in Fig. 3b. A 50 J pendulum hammer of the manufacturer Zwick is employed and impacts the specimen slide at 3800 mm/s. The specimen front end is clamped to the movable slide while the back end is clamped to a fixed holder. When the hammer impacts the movable slide, the slide accelerates the clamped front end in the direction of the pendulum strike, inducing tensile loading and fracture at the notch. Force is measured using an external load cell, whereas pendulum velocity and angular acceleration are obtained from the internal rotary encoder and timing system. All signals are acquired at 250 kHz. Specimen displacement during deformation is obtained by time integration of the pendulum velocity. Accordingly, the area under the force–displacement curve yields the total absorbed energy. To determine the energy attributable solely to specimen fracture, the energy required to accelerate the specimen slide is removed from the total energy. This contribution is governed mainly by slide friction and slide material. Therefore, three reference impacts are performed without a specimen, and the resulting energy is subtracted from the total energy measured in the tests with specimens to obtain the tensile impact energy.

3. Results

Prior to the numerical analysis of the sheet extrusion process, the simulation model is validated against the experimentally measured punch force-displacement data (Fig. B2 in the appendix). During the quasi-stationary extrusion stage, the simulations accurately reproduce the extrusion forces for $p_c = 0$, $p_c = 0.35\sigma_{f,0}$ and $p_c = 0.70\sigma_{f,0}$, with mean deviations from the experimentally measured forces of 4.1%, 3.9% and 2.5%, respectively. The elastic response, the maximum force and the descending part of the force-displacement responses are all captured well by the FE model for all values of p_c examined. The pronounced deviations observed during the preceding upsetting of the hybrid billet within the container and the subsequent transition to the extrusion phase (punch stroke < 20 mm) for the processes with superposed counterpressure can be attributed to the ejector force regulation of the machine. In the following, the force-validated model is used to numerically investigate the mechanical properties of the extruded sheets as a function of the applied counterpressure.

3.1. Stress-state-dependent damage evolution

The extrusion process exhibits localized plastic flow within a narrow forming zone confined to the die shoulder entry and the die exit, as indicated by the normalized plastic strain increment $\Delta\bar{\epsilon}^p/\Delta\bar{\epsilon}_{\max}^p$ (Fig. 4a). No plastic deformation develops outside of this narrow zone, irrespective of the counterpressure applied. Within this narrow zone, the hydrostatic stress state is predominantly compressive ($\eta < 0$) for all loading conditions. Only in the absence of counterpressure, around the central axis and at the shoulder exit, small regions with slightly positive triaxiality develop. Fig. 4b quantifies the loading path in terms of triaxiality and the Lode angle parameter as functions of equivalent plastic strain, for material points along the central axis ($r/r_1 = 0$), in a zone described here as the pressure-dominant zone ($r/r_1 = 0.5$), and at the surface ($r/r_1 = 1$), as identified in Fig. 4a. Overall, the counterpressure shifts the triaxiality towards more negative values across the entire forming zone, while the shape of the triaxiality evolution with respect to $\bar{\epsilon}^p$ remains unchanged. In contrast, the corresponding Lode angle parameter is not influenced either in magnitude or in its evolution with respect to $\bar{\epsilon}^p$.

At the central axis ($r/r_1 = 0$), the triaxiality evolves as the material point passes through the process zone and $\bar{\epsilon}^p$ accumulates, whereas $\bar{\theta}$ remains almost constant at values close to axisymmetric tension. Only in the absence of counterpressure does triaxiality become positive, reaching a maximum of $\eta = 0.11$. Applying counterpressures of $p_c = 0.35\sigma_{f,0}$ and $p_c = 0.70\sigma_{f,0}$ reduces the maximum triaxiality to values below zero by $\Delta\eta \approx 0.16$ and $\Delta\eta \approx 0.3$, yielding maximum values of $\eta = -0.05$ and $\eta = -0.19$, respectively. At $r/r_1 = 0.5$, the shape of the triaxiality evolution is similar to that of the central axis but shifted towards more negative triaxiality. In contrast, the Lode angle parameter is not constant and decreases towards $\bar{\theta} \approx 0.5$ at the end of plastic deformation. At the surface ($r/r_1 = 1$), the material experiences a pronounced change in triaxiality. Most plastic deformation accumulates under highly negative triaxiality ($\eta < -0.5$), followed by a sharp increase of η towards positive values at large strains, immediately before the material point exits the die shoulder. The corresponding $\bar{\theta}$ evolution exhibits pronounced fluctuations within a range of $-0.2 \leq \bar{\theta} \leq 1$.

The simulated damage distributions, obtained by monitoring the damage field D during the computation, are approximately homogeneous in the axial direction; increasing counterpressure leads to an overall reduction in damage (Fig. 5). Damage is nonzero only for $p_c = 0$ on the central axis and decreases from $D \approx 0.007$ to zero, followed by a renewed increase towards the surface, where the maximum is reached ($D \approx 0.016$). Applying counterpressure shifts the onset of damage towards the surface and reduces the damage magnitude.

3.2. Residual stress and strain distributions

The equivalent plastic strain in the extruded sheet is homogeneously distributed along the axial direction, independently of the applied counterpressure (Fig. 6). In contrast, the strain distribution over the extruded width is inhomogeneous. Close to the sheet center, the equivalent plastic strain is nearly constant along the width, but progressively increases until it sharply drops to the sheet edge.

Based on the radial distribution of the equivalent plastic strain, an extraction width of $w = 6$ mm is selected for the subsequent tensile test specimens to ensure that these specimens are homogeneously prestrained. The path-weighted average strain across the extraction width deviates only marginally from the local distribution with a normalized root-mean-square error of 0.02. To confirm the numerical results, HV30 macrohardness is measured at five discrete positions across the extraction width according to DIN EN ISO 6507-1 (Fig. C1 in the appendix). The hardness is

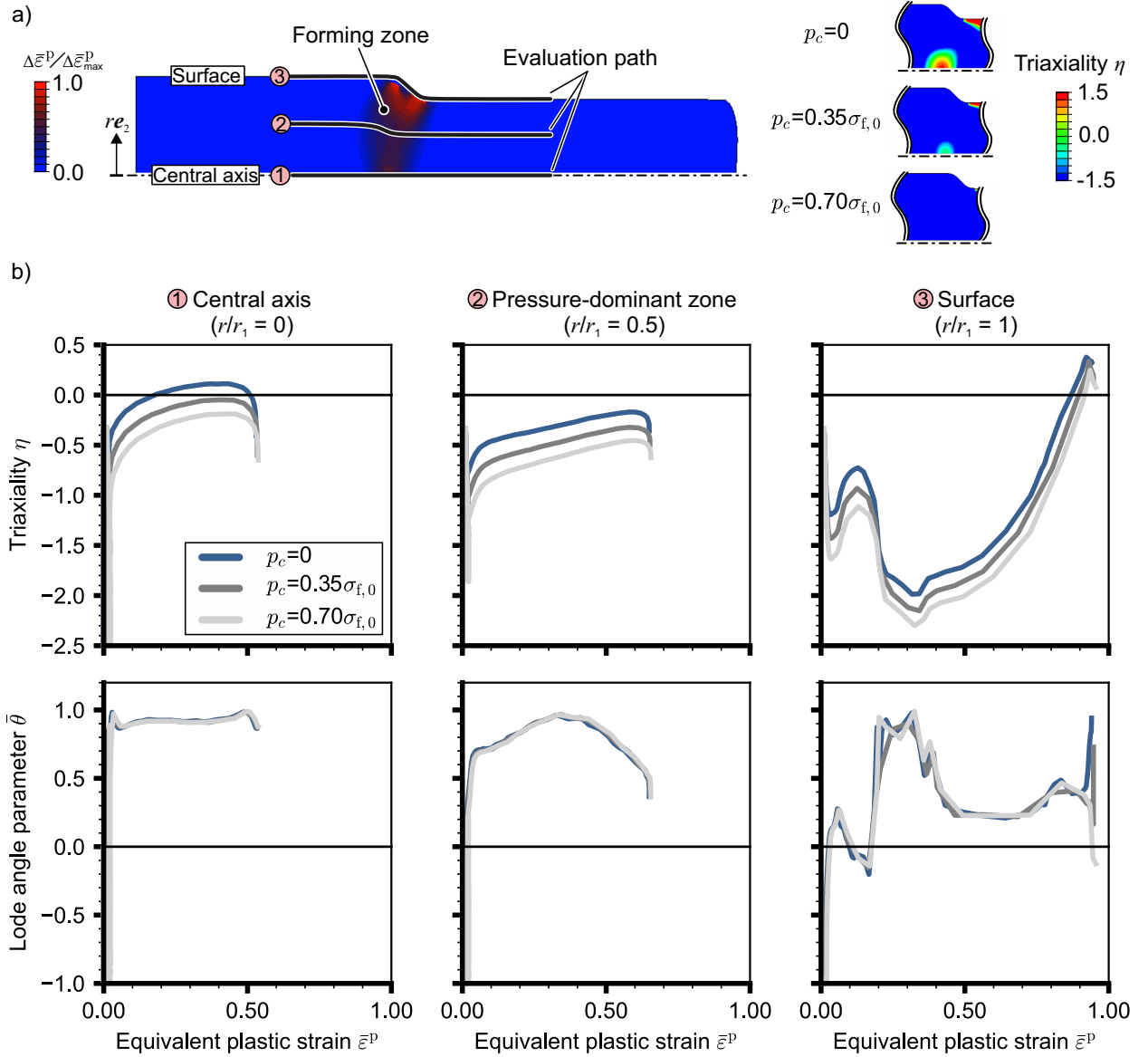


Figure 4: a) Left: Field distribution of the normalized plastic strain increment $\Delta\bar{\epsilon}^p/\Delta\bar{\epsilon}_{\max}^p$ with $\Delta\bar{\epsilon}_{\max}^p = 0.024$ in order to visualize the forming zone. The evaluation paths considered for the investigation of the triaxiality and the Lode angle parameter are represented by black lines. Right: Triaxiality distribution within the forming zone. b) Evolution of the triaxiality η and the Lode angle parameter θ as a function of the accumulated equivalent plastic strain $\bar{\epsilon}^p$ for material points on the central axis, within the so-called pressure-dominant zone, and at the surface.

uniformly distributed across the extraction width, and no systematic influence of counterpressure on the hardness is observed.

Although the extruded sheets exhibit an almost uniform strain distribution, the locally inhomogeneous plastic flow within the narrow forming zone (cf. Fig. 4) leads to substantial residual stresses that are superimposed on the applied load during subsequent testing and thereby may distort the investigation of damage-induced property changes. Therefore, the axial residual stress distribution is numerically evaluated after extrusion, after ejection and after specimen extraction, for all three sheet extrusion processes (Fig. 7). The specimen extraction is modeled by mapping the post-ejection residual stresses onto the tensile test specimen geometry and performing an implicit static equilibrium calculation to enforce equilibrium in the extracted specimen, yielding the updated residual stress state.

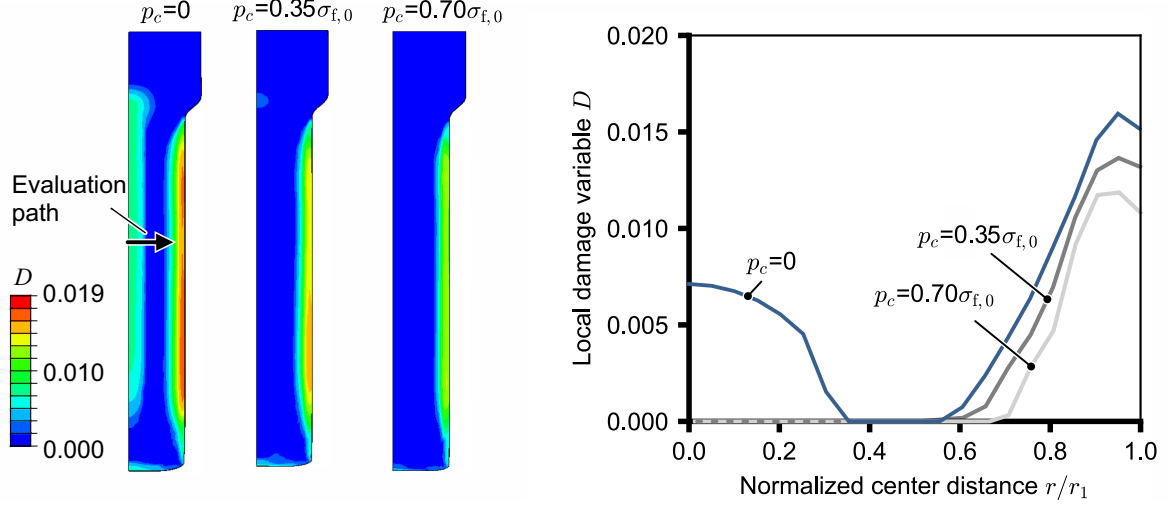


Figure 5: Simulated damage distribution for the investigated sheet extrusion processes subjected to different levels of superposed counterpressure.

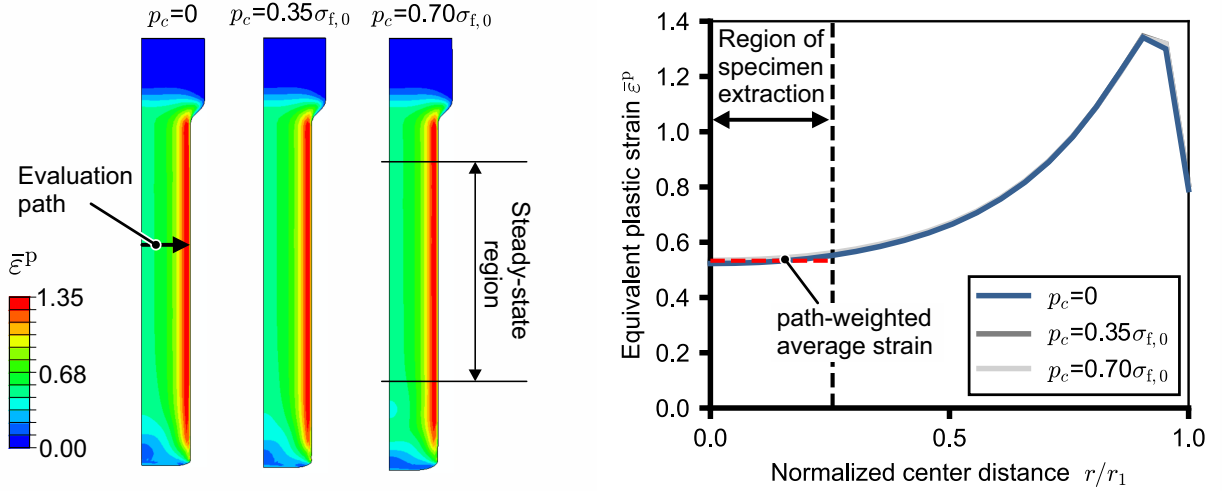


Figure 6: Influence of counterpressure on equivalent plastic strain distribution in extruded sheets. The tensile-test specimen extraction width, $w = 6$ mm, is defined according to the strain distribution.

This procedure assumes that the extraction, i.e., the machining operation, does not cause additional residual stresses.

Small localized plastic deformation at the sheet edge, induced during ejection by the springback of the elastic die after unloading, causes a release of the residual stresses (Tekkaya et al., 1985). The remaining residual stresses close to the center are still in the range of 40% of the local flow stress $\sigma_f(\bar{\epsilon}^p = 0.52) \approx 1100$ MPa for the extrusion processes with $p_c = 0$ and $p_c = 0.35\sigma_{f,0}$. The subsequent specimen extraction leads to an additional decrease in residual stresses. The remaining residual stress values are then comparable across all sheet extrusion processes and, with ± 60 MPa ($\pm 5\%$ of $\sigma_f(\bar{\epsilon}^p = 0.52)$), are negligibly small.

3.3. Damage characterization

The automated scanning method for damage characterization provides the advantage of capturing individual void data throughout the entire panoramic images (area of 11.91×0.80 mm²). In order to quantify the spatial damage distribution, the large panoramic images are subdivided into bins of $1191 \mu\text{m} \times 800 \mu\text{m}$ in the e_2 direction, shown

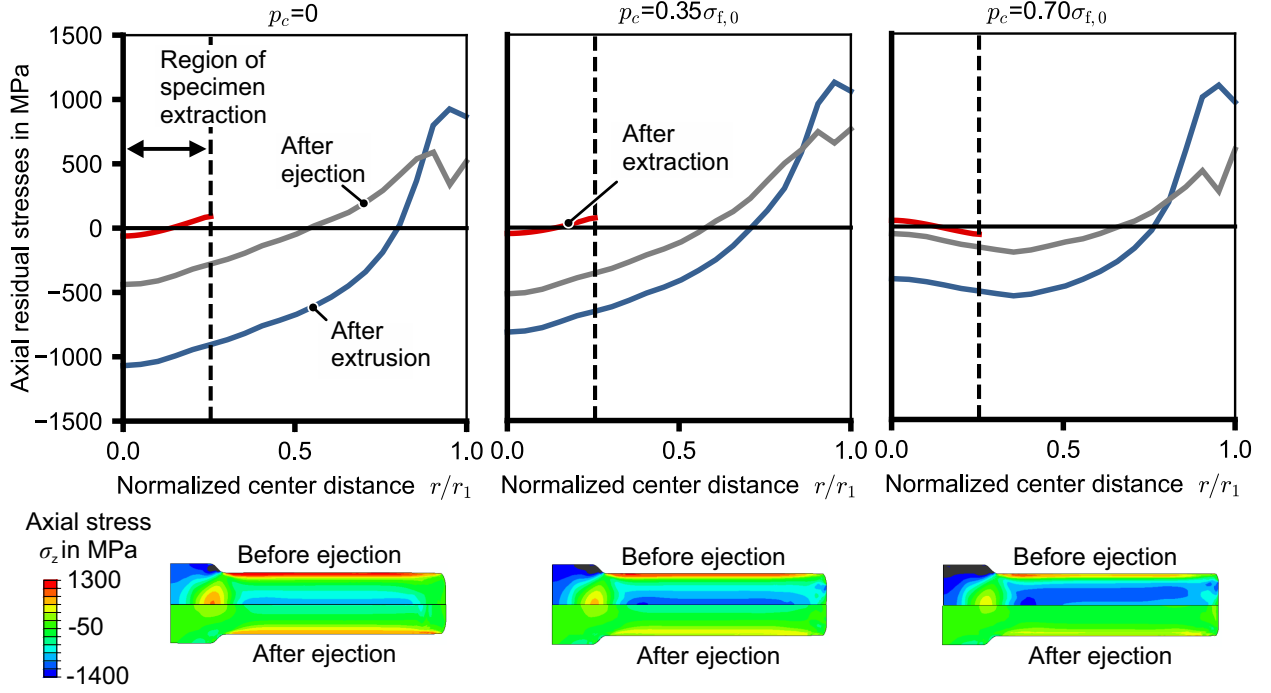


Figure 7: Influence of counterpressure on the axial residual stress across the sheet width. The stress state after specimen extraction is determined by an implicit finite element simulation of the equilibrium state of the tensile test specimen with the mapped stress state distribution after ejection.

in Fig. 2, leading to a reference area of $A_0 = 0.953 \text{ mm}^2$ for each bin. A spacing equal to the bin width of $1191 \text{ }\mu\text{m}$ is considered so that the bins do not overlap. A visual representation of the bin dimensions and arrangement is given in Fig. 8c. Experimental damage prevalence is then quantified based on mesoscopic parameters, i.e., the void area fraction $D^{\text{exp}} = 1/A_0 \sum_{i=1}^{N_v} A_{v,i}$ and the void number density $n_v = N_v/A_0$. Here, $A_{v,i}$ denotes the area of the i -th void and N_v the total number of voids within the reference area A_0 . The void area fraction and void number density distributions across the sheet width are shown in Fig. 8a.

First and foremost, across the complete sheet width, a decrease in void area fraction and void number density is observed with increasing counterpressure. D^{exp} is on average $\approx 54\%$ higher for $p_c = 0$ than for $p_c = 0.35\sigma_{f,0}$. The difference between the two processes subjected to counterpressure is minor in the range of $0 < r/r_1 < 0.65$. For $p_c = 0$, D^{exp} is almost constant up to $r/r_1 = 0.35$, followed by a decrease of 31% and a renewed increase to its maximum of $D^{\text{exp}} \approx 0.0048\%$ at $r/r_1 = 0.95$. The corresponding void number density n_v follows the same qualitative trend, except near the outer surface where the increase is steeper than that of D^{exp} . The qualitative shape for the parts subjected to $p_c = 0.35\sigma_{f,0}$ and $p_c = 0.70\sigma_{f,0}$ is different compared to $p_c = 0$. Starting from a relatively low value close to the initial void area fraction D_0^{exp} of the as-received material, a continuous increase from $r/r_1 = 0.05$ to $r/r_1 = 0.65$ is observed. The D^{exp} value for $p_c = 0.35\sigma_{f,0}$ is consistently slightly higher than for $p_c = 0.70\sigma_{f,0}$. The peaks in D^{exp} occur near the outer surface for all experiments and highlight the influence of the counterpressure: relative to $p_c = 0$, the maximum D^{exp} is reduced by $\approx 35\%$ and $\approx 80\%$ for $p_c = 0.35\sigma_{f,0}$ and $p_c = 0.70\sigma_{f,0}$, respectively.

In order to assess whether changes in D^{exp} are primarily caused by variations in the number of voids or in their average size, the average area of voids within each evaluation bin, defined as $\bar{a}_{\text{bin}} = D^{\text{exp}}/n_v$, is evaluated. The corresponding distribution of $\bar{a}_{\text{bin}}$ along the sheet width is shown in Fig. 8b. The average of the bin-wise values across the evaluated width $\bar{a}$ is $\approx 64\%$ higher for the parts subjected to counterpressure than for the parts produced with $p_c = 0$ ($\bar{a} = 0.050 \text{ }\mu\text{m}^2$). Across all experiments, the qualitative trend is comparable: For $r/r_1 < 0.65$, all $\bar{a}_{\text{bin}}$ values are below the corresponding $\bar{a}$ value. Towards the outer surface, all experiments approach values above the corresponding $\bar{a}$ value.

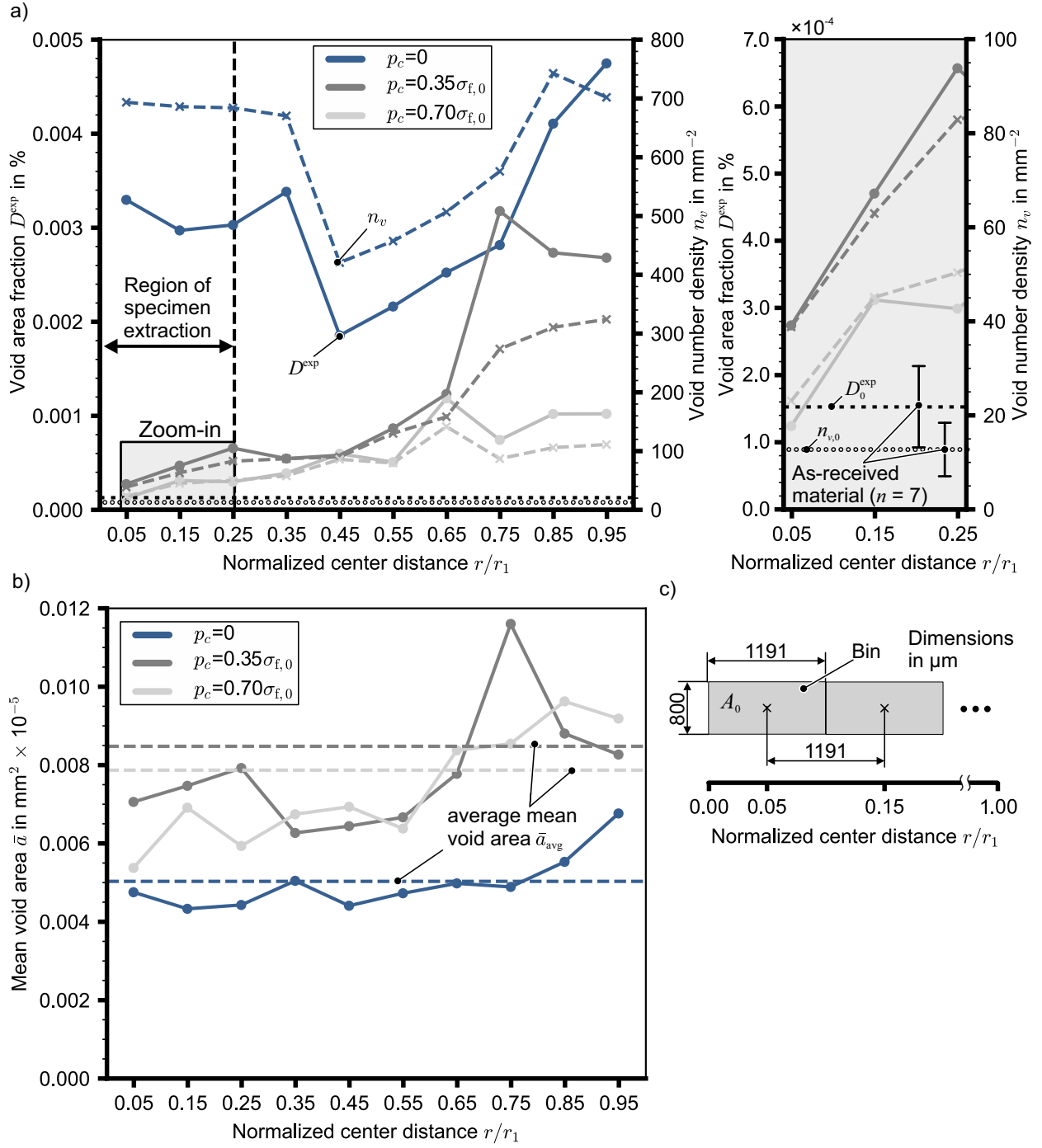


Figure 8: a) Distribution of void area fraction D^{exp} and void number density n_v along the sheet width. b) Distribution of the average void area $\bar{a}_{\text{bin}}$ along the sheet width. The dashed lines represent the averages of the bin-wise values across the evaluated width. c) Schematic representation of the subdivision of the panoramic images into bins with a reference area of $A_0 = 0.953 \text{ mm}^2$.

3.4. Characterization of damage-induced changes in mechanical properties

3.4.1. Tensile testing

After the microscale damage evaluation, the macroscale mechanical behavior of the tensile specimens extracted from the extruded sheets (cf. Fig. 3a) is discussed. In the subsequent tensile tests on extruded sheets, the onset of yielding is defined by the minimum temperature $T_{\min}$, as shown by three representative stress and temperature curves with respect to time (Fig. 9a). The corresponding true stress-strain curves are depicted in Fig. 9b. The initial strain corresponds to the path-weighted average strain across the specimen extraction width of the extruded sheets (cf. Fig. 6).

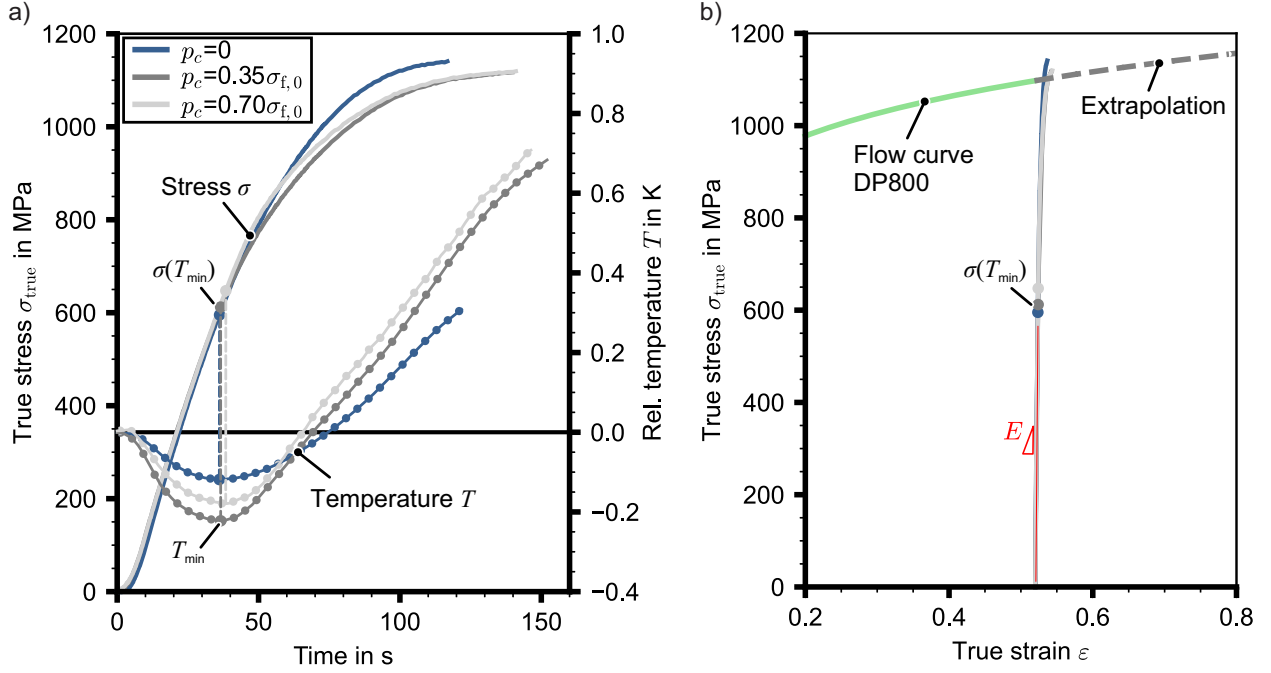


Figure 9: a) Procedure for determining the apparent elastic stiffness of extruded sheets. The onset of yielding is defined by the minimum temperature $T_{\min}$ during tensile loading. b) True stress-strain curves of extruded sheets. The segment of each true stress-strain curve between 25 MPa and $\sigma(T_{\min})$ is used for the calculation of apparent elastic stiffness by means of a linear regression with $0.997 \leq R^2 \leq 0.998$ across all experiments.

At the onset of yielding, the true stress-strain response exhibits a steep, slightly nonlinear increase that nearly follows the preceding linear elastic response. As a consequence of the pronounced work hardening induced by sheet extrusion, the maximum stress is reached after only ≈ 0.015 true axial strain in all tests. The maximum stress is similar for all tests and exceeds the value of the reference flow curve at an equivalent plastic strain of $\bar{\epsilon}^p = 0.535$ marginally. Note that the experimental measurement of the reference flow curves up to such large plastic strains is enabled by in-plane torsion testing (details of the test procedure are given in Traphöner et al. (2021)).

The apparent elastic stiffness is determined by linear regression of the stress-strain response within the strain range defined up to the individual minimum temperature of each specimen. The mean apparent elastic stiffness and mean ultimate tensile strength (UTS) for different counterpressure levels are shown in Fig. 10a. These results are further compared against the reference state (initial Young's modulus and flow stress at $\bar{\epsilon}^p = 0.535$).

No systematic differences in apparent elastic stiffness are observed between specimens extruded under different counterpressures. Their mean values range between ≈ 176 and 181 GPa, corresponding to an average reduction of $\approx 14\%$ relative to the initial Young's modulus of 207 GPa. The UTS remains close to the reference flow stress and likewise shows no systematic dependence on p_c within the experimental scatter. In contrast, counterpressure has a pronounced effect on the remaining ductility (Fig. 10b). It increases by $\approx 18\%$ and $\approx 40\%$ for $p_c = 0.35\sigma_{f,0}$ and $p_c = 0.70\sigma_{f,0}$, while the overall shape of the engineering stress-strain curves remains essentially unchanged.

These results of the tensile tests will be discussed in light of the microscale damage shown in Fig. 8. In the region of the extracted tensile specimen, the use of counterpressure leads to a lower void area fraction but a higher average

void size than in the conventional (no-counterpressure) case. These differences are associated with changes in the remaining ductility of these, otherwise geometrical-, residual stress- and prestrain-wise identical, specimens. This behavior, along with those reported in the following section, will be discussed in a holistic way in Section 4.

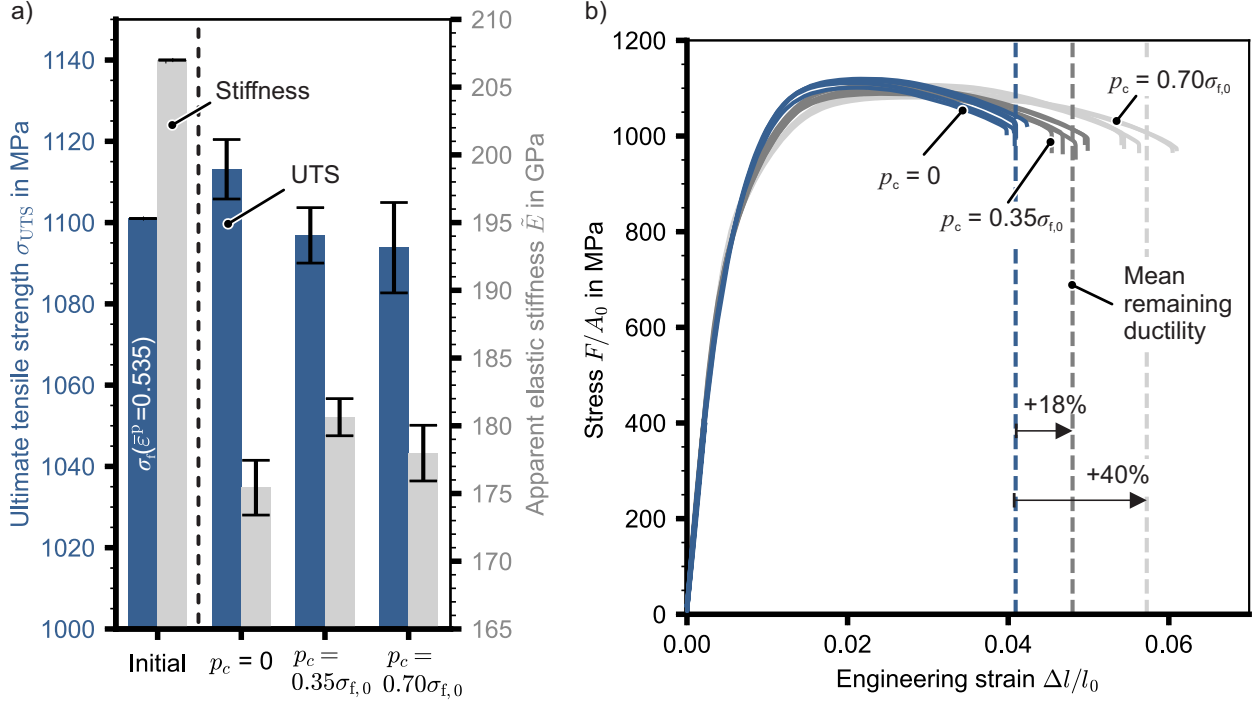


Figure 10: Influence of counterpressure p_c on a) ultimate tensile strength and apparent elastic stiffness and b) remaining ductility. The data referring to the "Initial" label represent the properties of the initial material, i.e., initial Young's modulus and flow stress at $\bar{\epsilon}^p = 0.535$ (cf. Fig. 9). The mechanical properties of the extruded sheets are determined by subsequent tensile tests of extracted specimens.

3.4.2. Tensile impact testing

The force-displacement response of the tensile impact tests on notched specimens extracted from the extruded sheets show similar trajectories for all counterpressure levels (Fig. 11a).

The displacement-to-fracture increases slightly when counterpressure is applied. However, no measurable difference is observed between the two different counterpressure levels. The tensile impact energy is obtained by numerical integration of the force-displacement curves and is normalized by the minimum cross-sectional area, $A_{T1} = 8.925 \text{ mm}^2$, for different values of p_c (Fig. 11b). Specimens extruded without counterpressure exhibit a tensile impact energy $\approx 7\%$ lower than that of specimens extruded with counterpressure, whereas no difference is resolved between $p_c = 0.35\sigma_{f,0}$ and $p_c = 0.70\sigma_{f,0}$ ($\Delta W \approx 0.17\%$).

In summary, it is seen in both Figs. 10b and 11 that failure-related quantities are influenced by the deformation-induced damage state.

4. Discussion

4.1. Quantitative dependence of damage evolution on stress state

In order to investigate the stress-state-dependent damage evolution during sheet extrusion, the local loading path is related to damage patterns observed in simulation (Fig. 5) and experiment (Fig. 8). As a reminder, and as mentioned in Section 1, the damage variable D is an internal variable of the modeling framework and does not correspond to a physical quantity on the microscale, e.g., void area fraction. Furthermore, it is dimensionless. Hence, the predicted damage variable cannot be directly related to a specific microscale quantity, i.e., $D \neq D^{\text{exp}} \wedge D \neq n_v$, because it

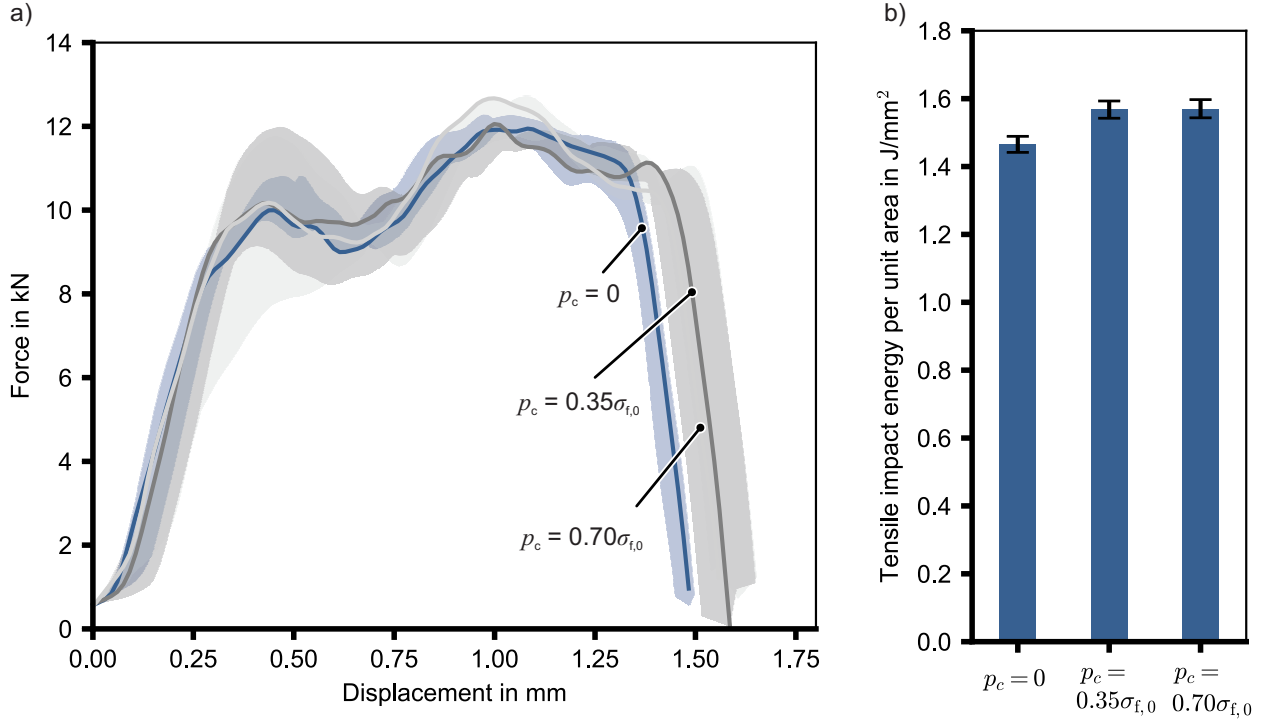


Figure 11: a) Force-displacement responses of the tensile impact tests. The envelope curve together with the experimental curve closest to the mean response is shown for each process. b) Tensile impact energy with respect to the minimum cross section $A_{T1} = 8.925 \text{ mm}^2$.

represents the implicit degradation of effective material properties through coupling of plasticity and damage driving forces (Sprave and Menzel, 2020). Therefore, the simulated damage variable D is first qualitatively compared with the experimental damage metrics D^{exp} and n_v .

In the constitutive model, damage can initiate only for $\eta > 0$. Accordingly, around the central axis, the model predicts damage only for the process without superposed counterpressure, which exhibits positive triaxiality values (Fig. 5). The predictions reflect the clear differences in damage observed in the experimental data between the parts produced with $p_c = 0$ and those subjected to counterpressure, i.e., an increase in the void area fraction by a factor of 25 relative to D_0 for $p_c = 0$, compared to an increase by a factor of 1.9 and 4.2 for $p_c = 0.70\sigma_{f,0}$ and $p_c = 0.35\sigma_{f,0}$, respectively (Fig. 8a). However, although the latter two values are of the same order of magnitude as the initial damage state D_0 , the model does not capture this minor damage evolution observed in the experiments subjected to counterpressure.

In the range of $0.25 < r/r_1 < 0.65$, even though the triaxiality for the process without counterpressure is lower than the triaxiality for $p_c = 0.35\sigma_{f,0}$ at the central axis ($\eta_{\text{max}} = -0.15$ vs $\eta_{\text{max}} = -0.05$, see Fig. 4b), a more pronounced damage evolution is observed (Fig. 8a), indicating a damage sensitivity to more shear-dominated deformation as the Lode angle parameter reduces to a minimum of $\bar{\theta} \approx 0.5$. This is consistent with the observations of Kweon (2012): at small negative triaxialities $-1/3 < \eta < 0$, damage arises from additional shear deformation components causing tensile hydrostatic stresses at grain interfaces at the microscale even though the macroscopic stress state is compressive. The constitutive model used in the present work does not include Lode-angle-dependent damage initiation. Accordingly, contrary to the experimental measurements, no damage is predicted.

Towards the surface, the simulation model qualitatively depicts the damage increase and the trends for different counterpressure levels. Both the experiment (Fig. 8a) and the simulation (Fig. 5) show that maximum damage occurs close to the surface, although most of the plastic deformation took place at pronounced negative triaxiality and only the last $\approx 7\%$ of plastic deformation occurred at positive triaxialities. This is in line with the expectation that, even after extensive forming at negative triaxialities, damage can nucleate and grow during subsequent phases of positive triaxialities. However, in contrast to the literature (Wollenweber et al., 2023), where damage prevalence is

decreased by applying prior compression, in this work the near-surface loading path still leads to increased damage prevalence despite the preceding pronounced compression. This is apparent in the $p_c = 0$ sample, where a higher damage prevalence is observed in the near-surface area than along the central axis (Fig. 8a), even though more plastic deformation took place at positive triaxialities (Fig. 5). The stronger increase in damage prevalence in the near-surface area can be attributed to the lower Lode parameter, the higher maximum triaxiality level, and the higher total strain in this region.

In the range of $r/r_1 < 0.55$, the average void area $\bar{a}_{\text{bin}}$ remains approximately constant for all counterpressures (Fig. 8b). This indicates that, within this region, the void number density n_v is approximately proportional to the void area fraction D^{exp} . This suggests that the growth of existing voids is not the dominant damage mechanism in this region. Toward the outer surface region ($0.75 < r/r_1 < 0.95$), the equivalent plastic strain increases to approximately $\bar{\epsilon}^p \approx 0.9$, while the material exhibits large positive hydrostatic stresses at the end of plastic deformation. As expected, these stress conditions promote void growth, as evidenced by the average void area values $\bar{a}_{\text{bin}}$ increasing beyond the corresponding $\bar{a}$ value in all experiments. The only aspect that does not meet expectations is that $\bar{a}$ for $p_c = 0$ is $\approx 39\%$ smaller than the corresponding values observed in experiments subjected to counterpressure. This may be attributed to extensive void nucleation, which reduces the average void area.

4.2. Damage-induced changes in mechanical properties

The previous section shows that the use of counterpressure leads to a lower void area fraction and a lower void density than are observed in the process without superposed counterpressure (Fig. 8a). In contrast, the hardening state and the residual stress state in the region from which the specimens are extracted are not influenced by the applied counterpressure. The extracted specimens exhibit identical work hardening (Fig. 6) and macroscopically negligible residual stresses (Fig. 7). Moreover, the specimens are flat and exhibit spatially homogeneous properties. Therefore, global material properties, i.e., apparent elastic stiffness and ultimate tensile strength, can be evaluated with respect to deformation-induced damage by means of characterization tests on the extracted specimens. Since the same specimen geometry can also be manufactured from the initial sheet material, the analysis is not limited to relative comparisons between different counterpressure conditions. Instead, it also enables a direct comparison with the initial material state.

In the following, a distinction is made between properties that can be evaluated with negligible additional damage accumulation and properties that additionally depend on plastic deformation during subsequent characterization. The apparent elastic stiffness and the ultimate tensile strength are considered "non-fracture-related" mechanical properties. In these tests, the subsequent loading is either predominantly elastic or involves only minor plastic deformation. Therefore, additional damage accumulation during characterization is assumed to be negligible. By contrast, properties such as the remaining ductility and the tensile impact energy are associated with plastic loading up to macroscopic failure. These properties therefore depend not only on the initial deformation-induced damage state but also on the damage evolution occurring during subsequent loading to failure. Accordingly, they are referred to as "fracture-related" mechanical properties.

4.2.1. Influence on non-fracture-related mechanical properties

The overall $\approx 14\%$ reduction in the apparent elastic stiffness of all extruded states relative to the initial state arises from the linear approximation of the inherently nonlinear elastic response associated with dislocation-driven anelastic deformation (Chen et al., 2016; Arechabaleta et al., 2018; Li and Wagoner, 2021; Chamakura et al., 2022). The nonlinearity results from reversible dislocation motion below the yield stress, e.g., short-range reversible movement of dislocation lines pinned between two obstacles. Consequently, increasing dislocation density amplifies the anelastic contribution and reduces the slope obtained from a linear approximation, i.e., the apparent elastic stiffness. In contrast, even the largest average void area fraction $D^{\text{exp}} \approx 0.0032\%$ across the specimen extraction width for $p_c = 0$, see Fig. 8a, is orders of magnitude too small to lead to the corresponding loss of load-bearing cross-section and hence cannot be the cause of the observed stiffness decrease. Additionally, the minor variations in apparent elastic stiffness among the extruded sheets indicate that differences in void area fraction do not measurably affect the elastic response.

According to Kolpak et al. (2021), the ultimate tensile strength in the subsequent tensile test of the extruded sheets corresponds to the flow stress of the material at the extrusion strain (Fig. 9b). Although an increase in damage is expected to reduce the effective load-bearing capacity and, consequently, the yield stress, no statistically significant

differences in ultimate tensile strength are found between the investigated damage states. The specimen with the highest damage exhibits an average stress that is $\approx 1.10\%$ higher than the reference, whereas the specimen with the lowest damage exhibits an average stress that is $\approx 0.90\%$ lower than the reference. However, considering the experimental scatter of $\pm 0.75\%$, these deviations do not indicate a systematic influence of damage on the ultimate tensile strength. This is consistent with the findings of Gitschel et al. (2023). They found that deformation-induced damage in the form of voids does not affect the ultimate strength of specimens made from case-hardened steel. In general, the shape of the stress-strain curves until the point of maximum stress is comparable across the different specimens. This indicates that the subsequent damage accumulation up to the onset of necking is negligible since the corresponding increase in plastic strain is minor ($\Delta \bar{\epsilon}^p < 0.03$).

In summary, for the investigated DP800 sheet extrusion states and the measured void area fractions below 0.005 %, deformation-induced damage does not measurably affect the apparent elastic stiffness or the pre-necking tensile response. A critical void fraction at which a damage effect on non-fracture-related mechanical properties becomes relevant cannot be unambiguously identified. According to the literature, measurable effects of damage on mechanical properties become evident primarily at void area fractions exceeding 10%, for example in cellular solids and foams (Gibson, 1997; De Carolis et al., 2024), as well as in composites and cast irons (Dorhmi et al., 2020). Therefore, deformation-induced damage in the investigated low-porosity DP800 states is not expected to directly influence the evaluated non-fracture-related properties.

4.2.2. Influence on fracture-related mechanical properties

In contrast to the non-fracture-related properties discussed above, the increase in remaining ductility by 18% and 40% for $p_c = 0.35\sigma_{f,0}$ and $p_c = 0.70\sigma_{f,0}$ indicates that the deformation-induced damage state affects fracture-related mechanical properties. Although the overall void area fraction remains very low, the specimens extruded without counterpressure exhibit a void number density that is higher by a factor of 10 than that of the specimens subjected to counterpressure, see Fig. 8a. The reduced remaining ductility of the specimens extruded without counterpressure is therefore interpreted as being associated with their larger population of pre-existing voids and the resulting higher probability of locally small intervoid ligaments.

During subsequent plastic loading, the pre-existing voids are expected to grow and interact. A larger population of pre-existing voids increases the probability that a critical local condition for rapid void linkage and the initiation of a macroscopic crack is reached at a lower additional strain, ultimately resulting in lower remaining ductility.

The differences between the two counterpressure levels further indicate a nonlinear relation between the deformation-induced damage state and remaining ductility: small differences in the resolved damage metrics across the tensile specimen width (Fig. 8) correspond to a 22% difference in remaining ductility (Fig. 10b).

As an additional fracture-related mechanical property, the tensile impact energy is evaluated. The similar shape of the force-displacement curves within the experimental scatter supports the conclusion that the deformation-induced damage state is too low to measurably affect that aspect of the mechanical response. Differences are observed only in the fracture onset; specifically, specimens extruded without counterpressure fracture measurably earlier than those produced under either counterpressure condition, resulting in a 7% lower tensile impact energy. This supports the interpretation that the higher void number density increases the probability of locally small intervoid ligaments and, consequently, of reaching the critical local condition for void coalescence at a lower additional strain during subsequent loading.

5. Conclusion

In this study, the influence of damage on the mechanical properties of formed DP800 sheet metal components was experimentally and numerically investigated. The components were manufactured using a sheet extrusion process, while the damage state was controlled by applying different levels of superposed counterpressure. Subsequently, damage in the form of voids was characterized by large-area SEM measurements, and damage-induced property changes were assessed by tensile tests and notched-tensile impact tests on specimens extracted from the extruded sheets. The main conclusions are as follows:

- Due to the pressure-dominant stress state characteristic of the sheet extrusion process, damage evolution is less governed by void growth. This is reflected by the overall low void area fraction observed in all experiments

compared with that observed in processes that experience predominantly tensile loading conditions. Instead, a pronounced sensitivity to void nucleation becomes apparent: when critical stress conditions are reached, even temporarily, the number of voids increases markedly, while the detected voids remain small and contribute only little to the overall void area fraction.

- Deformation-induced damage does not measurably degrade the apparent elastic stiffness as the void area fraction is too low to explicitly reduce the load-bearing area. The significant reduction of the initial apparent elastic stiffness compared to the apparent elastic stiffness after extrusion results from the increasing nonlinear elastic response due to mechanically reversible movements of newly generated dislocations below the macroscopic yield stress, referred to as microplasticity.
- Deformation-induced damage does not measurably affect the pre-necking tensile response. This is reflected by the very similar stress-strain curves up to the ultimate tensile strength of the subsequent tensile tests on specimens with different deformation-induced damage states.
- The observed reductions in remaining ductility and tensile impact energy at higher deformation-induced damage states indicate that a higher deformation-induced void number density increases the probability of reaching a critical local condition for void coalescence at a lower additional strain, thereby promoting earlier macroscopic fracture. Consequently, fracture-related properties, particularly the in-service performance of a formed component subjected to further damage accumulation, depend on the deformation-induced damage state.
- The proposed experimental damage separation enables future validation of continuum damage models based on isolated void-related property changes, rather than on macroscopic softening that may also arise from other plasticity-induced mechanisms.

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Author Contribution Statement

Jan Gerlach: Methodology, Formal analysis, Investigation, Writing - Original Draft **Tom Reclik:** Methodology, Investigation, Writing - Review & Editing **Maximilian A. Wollenweber:** Methodology, Investigation, Writing - Review & Editing **Robin Gitschel:** Methodology, Writing - Review & Editing **Marlon Hahn:** Conceptualisation, Writing - Review & Editing **Talal Al-Samman:** Conceptualisation, Writing - Review & Editing **Sandra Korte-Kerzel:** Conceptualisation, Writing - Review & Editing, Supervision, Funding Acquisition **Ulrich Kerzel:** Conceptualisation, Writing - Review & Editing, Supervision, Funding Acquisition **Yannis P. Korkolis:** Conceptualisation, Writing - Review & Editing, Supervision, Funding Acquisition

Declaration of generative AI and AI-assisted technologies in the manuscript preparation process

During the preparation of this work, the authors used Campus-KI provided by TU Dortmund University (providing access to OpenAI ChatGPT) in order to improve language and readability. After using this tool/service, the authors reviewed and edited the content as needed and take full responsibility for the content of the published article.

Appendix A Flow curves

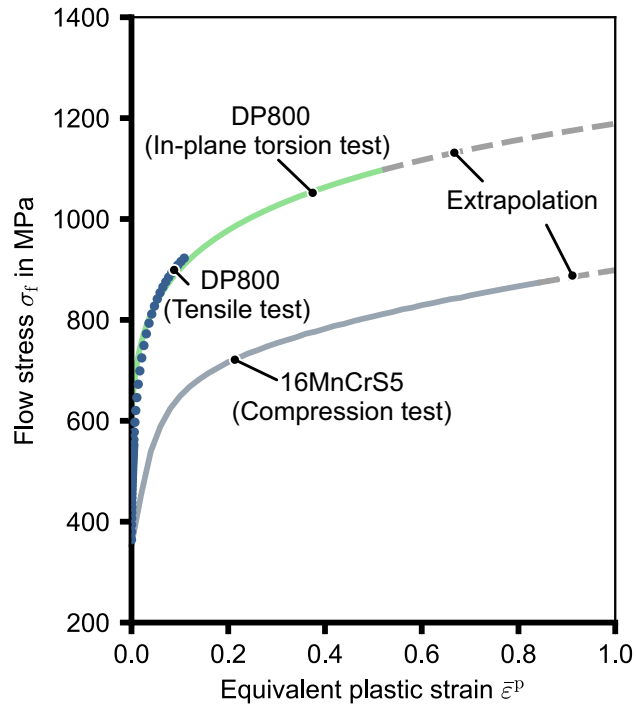


Figure A1: Flow curves of the DP800 sheet determined by a tensile test and an in-plane torsion test, as well as the flow curve of the 16MnCrS5 rod determined by a compression test.

Appendix B Finite element model for sheet extrusion

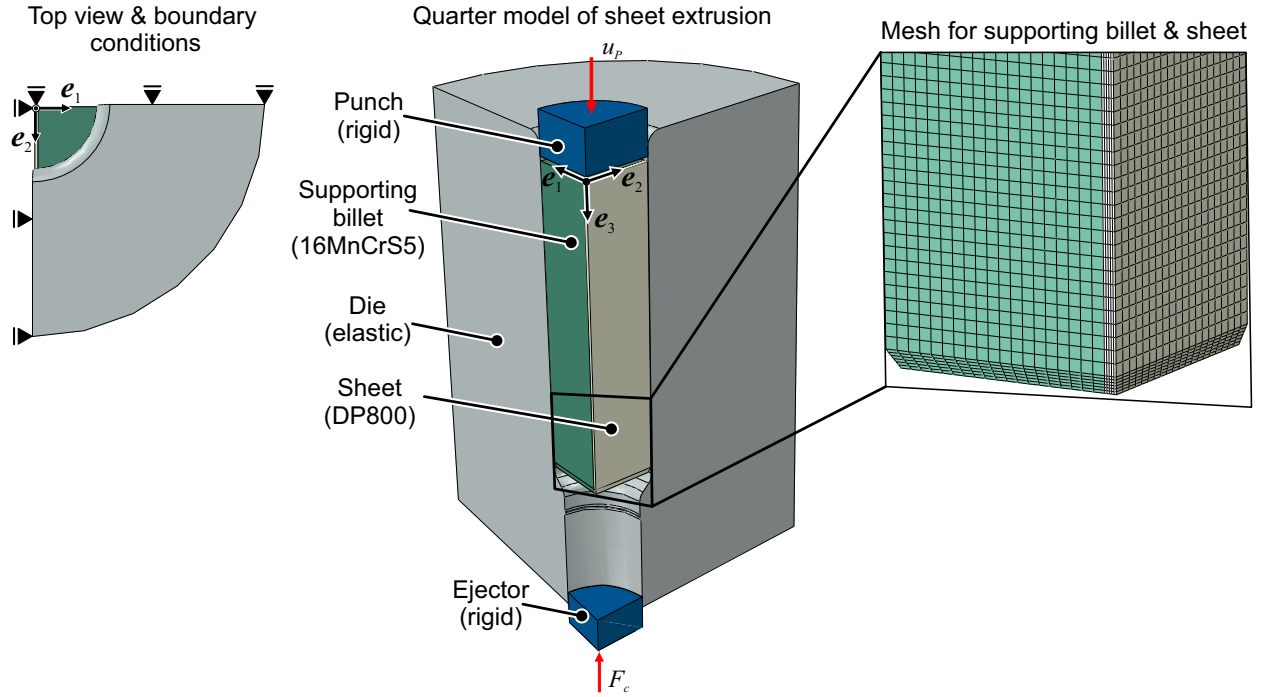


Figure B1: Finite element model for sheet extrusion.

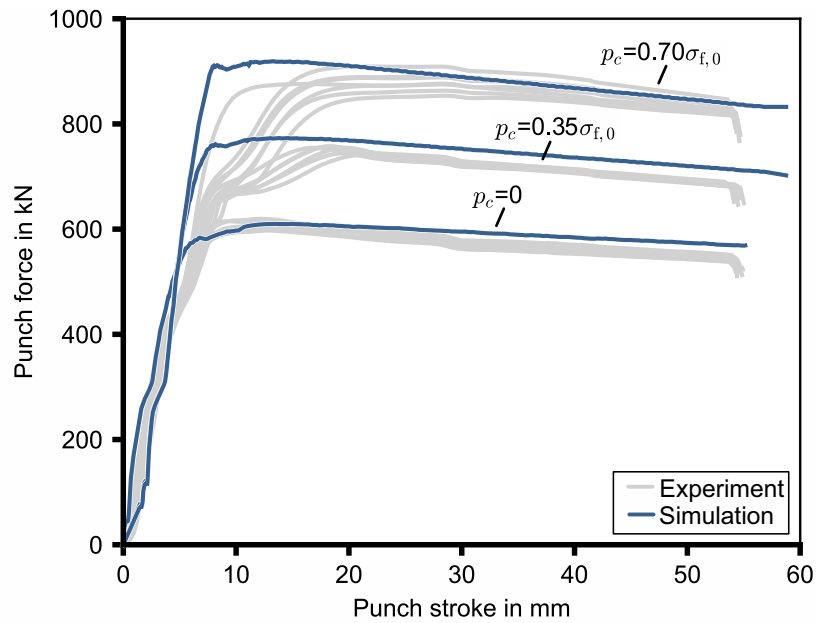


Figure B2: Punch forces during sheet extrusion with different levels of superposed counterpressure p_c . Simulation results obtained using the FE model shown in Fig. B1 are also included.

Appendix C Hardness measurements

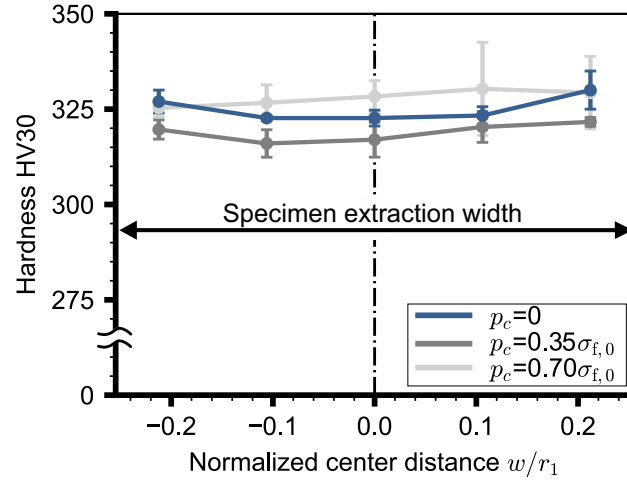


Figure C1: Hardness measurements across the tensile specimen extraction width, shown in Figs. 1 and 3a. The hardness of the as-received DP800 sheet material is 260 HV0.1 (Meya et al., 2019a).

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