Identification of the continuum damage parameter: An experimental challenge in modeling damage evolution

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Abstract

To accurately predict ductile failures of new advanced metals, continuum damage models (CDMs) require experimental determination of material-specific damage parameter(s). While various experimental techniques are being used to determine these damage parameter(s), possible systematic errors due to methodological differences in damage definition have yet to be fully revealed. With the aim of finding the most reliable ductile damage quantification strategy for CDMs, this work provides an in-depth comparison of six theoretically equivalent methodologies by considering measurement accuracy, precision, damage spectrum, spatial resolution and practicality. It is found that the methodologies that quantify ductile damage from its geometry introduce significant systematic errors as they probe a very limited damage spectrum, whereas the methodologies that probe the degradation of a mechanical property suffer from low precision and high complexity, especially for high strains and material anisotropy.

Keywords: Damage; Continuum models; Deformation; Localization; Fracture

1. Introduction

In the boom days of railway construction a century ago, cracks in heavy parts were found by tedious “oil and whitening”, consisting of cleaning the part with boiling caustic soda, drying, day-long oil submersion, recleaning, applying a chalk solution, and finally “hammering heavily” to “squeeze” the oil out of the cracks, allowing visual crack identification [1].

Such early damage detection methods were limited to rough lifetime assessments of engineering components, whereas today quantitative prediction of deformation limits has become a realistic possibility through the use of continuum damage models (CDMs). However, CDMs require experimental determination of a material-specific damage parameter, \( D \), which, for new complex-microstructured metals (e.g. dual-phase or TRIP steel, ultrafine-grained metals, thin metallic films), calls for highly accurate damage measurement methodologies that preferably avoid steps such as “heavy hammering”.

Traditionally, CDMs are based on two main assumptions: (i) the reduction in the effective load-bearing cross-sectional area resulting from material damage (voids) is assumed to cause a proportional increase in the applied stress, \( \sigma_{\text{appl}} \), to yield an effective stress, \( \sigma_{\text{eff}} \) [2] (and similarly a proportional reduction in the elastic modulus); this can be expressed in its most elementary form as \( \sigma_{\text{eff}} = \sigma_{\text{appl}} (1 - D) \), where \( D = A_D / A_0 \) with \( A_D \) and \( A_0 \) the damaged and undamaged cross-sectional area. (ii) The constitutive relations for an undamaged material are assumed to remain valid for the effective stress and assumed-unaltered strain of the damaged material (the so-called “strain equivalence principle”). In their pioneering work, Lemaitre and...
Dufailly used CDM assumptions to propose different experimental techniques to measure the deformation-induced evolution of $D$ [3]. First, $D$ can be measured from the ratio of the damaged to undamaged cross-sectional area, $DA = AD/AD_0$, or from a linearly proportional quantity comprising the ratio of the damaged to undamaged volume, $DV = VD/V0$, or the decrease in the material density, $DP = 1 - DP/D0$. In addition to these geometry-based damage quantification methodologies, mechanical property-based methodologies were also proposed that either measure $D$ from the degradation of the yield or flow stress, which can be determined from local hardness measurements assuming linear scaling to flow stress, $DH = 1 - H0/H$, or from the degradation of the elastic modulus, $DE = 1 - E0/E$. Moreover, a number of indirect damage quantification methodologies were proposed that, for instance, probe the ultrasonic wave attenuation, increase in electrical resistance, or increase in tertiary creep [3]. Note that, within the CDM assumptions, only the three geometry-based and two mechanical property-based damage quantification methodologies should theoretically yield the same damage value.

Since then, these methodologies have been employed frequently in academia and industry for damage parameter identification; however, to the best of our knowledge, all publications (except Ref. [4]) have used a single methodology to measure the damage value without verifying this by other means. This general confidence in the accuracy of a single chosen damage methodology is intriguing, considering the strong assumptions underlying CDMs and the large effect of the value of $D$ on the predicted material behavior. Perhaps most researchers have ignored this issue due to the extremely challenging nature of damage measurement. Damage effects are typically very small and sparsely distributed, causing limited reproducibility due to statistical variations even within a single measurement methodology [5]. Comparisons between methodologies are significantly more challenging, as the methodological differences in damage definition also introduce different intrinsic systematic errors. Perhaps another reason is that this is an experimental challenge typically faced by the numerically oriented researchers that develop or use CDMs.

In recent years, the authors carefully analyzed and extended the applicability of the five geometry-based and material property-based damage quantification methodologies, and added a sixth one (explained below) [5–12]. Based on this previous research, this current work attempts to determine the most suitable continuum damage parameter identification strategy, by comparing the six theoretically equivalent methodologies in terms of, for example, measurement accuracy, precision, damage spectrum, spatial resolution and practicality, by testing the techniques on the same pre-deformed material. It is clearly an unfeasible task to compare these techniques for all different types of materials and damage (e.g. brittle, ductile, creep, fatigue, etc.). Therefore, we have selected sheet metal, which is of evident importance, and ductile damage, which traditionally has been and still is most important for sheet metal (e.g. for advanced high-strength steels), and has therefore been the focus for many damage models, including CDMs.

The article is structured as follows. After an explanation of the preparation of the pre-deformed samples, the methodology and results of each damage quantification technique are presented. Finally, the damage evolutions of all techniques are compared in terms of accuracy, precision, etc., followed by the conclusions.

2. Preparation of pre-deformed samples

To investigate ductile damage evolution in sheet metal, uniaxial tensile deformation of an industrially relevant dual-phase 600 steel, with a microstructure of ferrite grains surrounded by martensite islands, is taken as a case study. Detailed analysis of the microstructural deformation revealed that (at least) two ductile damage mechanisms (i.e. martensite cracking and martensite–ferrite decohesion) are active [5,13]. Tensile specimens of dual-phase steel are cut (in rolling direction) by electrodischarge machining (EDM) (Fig. 1a), covered with a speckle pattern and pre-deformed using a microtensile testing stage up to the onset of fracture (Fig. 1b). Optical images, captured during pre-deformation, are post-processed by digital image correlation (DIC) analysis to yield the local equivalent strains on the pre-deformed sample (Fig. 1c).

To quantify the damage evolution as a result of the tensile deformation, the above-mentioned six methodologies are employed (Fig. 1d–h), for which at least three specimens (i.e. six specimen halves) are further processed for damage quantification, to reveal the final geometries shown in Fig. 1d–h.

3. Geometry-based methodologies

3.1. Area fraction methodology

To measure the cross-sectional area fraction, $DA = AD/AD_0$, scanning electron microscopy (SEM) is typically preferred over, for example, optical microscopy or atomic force microscopy due to its high spatial resolution and large field of view [3,4,14,15]. Hence, SEM is used here. Cross-sections of the pre-deformed sample (Fig. 1d), prepared via two optimized surface preparation protocols (mechanical polishing and electropolishing), are inspected using a FEI Quanta 600 SEM under optimized damage contrast settings (400× magnification, 30 kV acceleration.

1 Note that Lemaitre and Dufailly classified the area fraction methodology, $DA$, as “direct” and all other methodologies as “indirect”.

2 The protocols for mechanical polishing and electropolishing were meticulously optimized (through many test runs and follow-up microscopic analyses) to minimize, respectively, smearing-induced and edge rounding-induced effects in damage detection [7].
voltage, backscatter electron mode [7]). At different axial positions (i.e. strain levels) on the specimens, high-resolution cross-sectional micrographs are numerically stitched to areas of 150 × 400 μm² (Fig. 2 a), from which the damage area fraction is calculated using a greyscale threshold that separates material from air, which is visually set at the highest magnification to minimize operator-induced errors.

The results obtained with the SEM analysis reveal damage nucleation at relatively low strains of ~0.1 (Fig. 2 b). The damage increases roughly linearly up to a strain of ~0.35, beyond which it quickly rises towards the onset of fracture. Fig. 2b supports Ref. [7] in that mechanical polishing “smears out” voids, whereas electropolishing enlarges voids due to edge rounding, yielding, respectively, a significant systematic underestimation and overestimation of the damage area fraction, even when the utmost care is taken in specimen preparation. Other possible systematic errors can only be detected by comparison with other damage quantification methodologies.

3.2. Volume fraction methodology

The damage volume fraction, \( D_V = V_D / V_0 \), is ideally extracted from the exact 3-D geometry of the deformation-induced defects [16–18] making X-ray microtomography (XμT) with a spatial resolution of ~1 μm the preferred choice over alternatives such as scanning acoustic microscopy or thermal imaging. Therefore, high-resolution XμT measurements are carried out using a lab-scale Nanotom μCT scanner (Phoenix), using optimized signal-to-noise and resolution (0.5 μm voxel size) settings found in a pre-study (see Ref. [11] for details). Rods of 250 × 250 μm² cross-sectional area are electrodisharge machined from the center line of each pre-deformed sample, spanning the full deformation range (Fig. 1e).

Following the XμT acquisition, an enclosed volume of 50 × 50 × 400 μm³ is reconstructed at different axial positions (i.e. strain levels) in the rods. The damage volume fraction is calculated using two different strategies: (i) as common in the literature, the damage volume fraction is calculated by visually setting a material–air threshold in the greyscale histogram of the 3-D reconstruction (Fig. 3a1). (ii) The damage volume fraction is taken as

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D_{\text{V, E}} = \frac{V_{D, \text{E}}}{V_0}
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D_{\text{V, Mean}} = \frac{V_{D, \text{Mean}}}{V_0}
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the relative area underneath two Gaussian peaks that are peak-fitted to the solid and air-in-solid transmission peaks in the greyscale histogram (Fig. 3a2).\textsuperscript{4} Results obtained with both X\textsubscript{l}T strategies (Fig. 3b2) show that the damage values remain below the experimental accuracy of \(\sim 0.2\%\) for strains up to \(\sim 45\%\). Beyond \(\sim 45\%\) both strategies show a damage increase, though with significantly different damage values, indicating systematic error(s). In the commonly used “thresholding” approach, the damage level strongly depends on the subjective operator threshold setting through visual interpretation, which constitutes a significant source of inaccuracy, as clearly demonstrated by the three equally plausible reconstructed volumes in Fig. 3b1. This probably also explains the large difference between the two data points at 62\% strain (far exceeding the individual error bars). “Peak fitting” should provide a more scientific approach, as the fitting algorithm minimizes the residual for the complete two-peak profile, thereby eliminating operator-induced errors.\textsuperscript{5} However, when comparing these results with the damage area fraction results (Fig. 2b), it is clear that X-ray micromotography fails to sense any damage increase below \(\sim 55\%\), which is attributed to its limited spatial resolution, making it insensitive to (very) small voids. This shows the importance of damage techniques to be sensitive to all voids, for instance, by probing a material property such as the density, hardness or elastic modulus.

### 3.3. Density methodology

For the density methodology, \(D_\rho = 1 - \rho_d/\rho_0\), the experimental sensitivity of the Archimedes’ principle [19–21] is clearly insufficient for the tiny (\(\sim 1\mathrm{mm}^3\)) rectangular prisms (cuboids) needed for sufficient spatial resolution (Fig. 1f). Therefore, a new highly sensitive volume measurement technique is introduced and combined with (standard) high-sensitivity mass measurements. Because errors due to sensitivity in the volume measurement and precise cutting of cuboids exceed that of accurate mass measurement, two strategies are tested [7]. (i) Avoiding volume measurements, by cutting (with EDM) cuboids of equal volume (a constant width along sample axis, length and thickness of \(1 \times 2 \times 0.6 \mathrm{mm}^3\)) from the pre-deformed samples, after which the damage-induced density decrease can directly be measured with a mass balance (Mettler Toledo...
The results obtained with the first strategy (set 1 in Fig. 4b) show an unacceptably high experimental error (standard deviation of ~0.15%), preventing identification of any density trend. Analysis of the EDM processed surfaces [22] revealed a first possible cause of this large scatter: even for optimal EDM settings, the cutting process creates a relatively high surface roughness (i.e. due to the discontinuous formation of the so-called “white layer” reaching 5 μm in thickness), which was found to drift in time, resulting in different roughness values for different specimens.

On the other hand, the results obtained with the second strategy do reveal a significant decrease in density (and thus increase in damage) with increasing strain (set 2 in Fig. 4b).

This demonstrates that high-sensitivity volume measurements are necessary to correct for EDM-induced errors to yield a combination of sufficiently precise density and small spatial resolution to measure ductile damage evolution. Interestingly, the obtained damage values are significantly higher than those obtained from the SEM-based area fraction methodology (Fig. 2b). The correctness of both should be confirmed by comparison with the mechanical damage methodologies.

4. Mechanical property-based methodologies

Different strategies are employed to probe damage levels based on the measurement of a mechanical property. In earlier works, the degradation of elastic modulus was measured from stress–strain unloading slopes, yielding damage values averaged over the sample at different strains [23,24]. A more suitable, local approach utilizes an indentation apparatus to combine high spatial resolution with accurate displacement control and force measurement. Here, three such approaches are investigated that measure indentation hardness, indentation modulus [3,9,10,25] and micropillar-compression modulus [12].

4.1. Indentation methodologies

The indentation methodology, originally proposed for damage quantification from hardness degradation [3] and later extended to indentation modulus degradation [26], has attracted considerable research interest for two decades. Recently, however, the authors showed that plasticity-induced microstructural effects (strain hardening, texture evolution, grain shape change, etc.) either mask or significantly affect the damage-induced drop in hardness and modulus, which unfortunately render the damage values unreliable [9]. A remedy was also proposed [10]: removal of all non-homogeneous deformation-induced microstructural changes from the pre-deformed material.

The cubes’ thickness is limited by the sheet metal, while increasing the cuboids’ in-plane dimensions reduces the volume error due to the EDM positioning error to a theoretical maximum of only 66%, while considerably sacrificing spatial resolution.

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Fig. 4. (a) 3-D view of a typical cuboid volume measurement, showing the difference between two liquid surface height profiles measured before and after cuboid submersion. Inserts show side and top views of the polymer reservoir and its dimensions (in mm). (b) The two density measurements strategies: only mass measurements (set 1, red unfilled data points) shows no clear trend, whereas mass measurements combined with volume measurements (set 2, blue data points) captures the degradation of density and allows the calculation of damage (green data points shown below). (For color interpretation in this figure legend the reader is referred to see the web version of this article.)

XP2U with a tested reproducibility of ~0.25 μg). (ii) Avoiding EDM cutting errors, by measuring both mass and volume of the cuboids, with the added advantage of only cutting the four sides of the cuboids (variation in cuboid thickness due to sample pre-deformation is roughly compensated by decreasing the cuboid length from 2 mm at the neck to 1.2 mm at the sample end (Fig. 1f)). In this second strategy, the cuboid’s volume is cleverly measured from the displacement of a liquid surface induced by cuboid submersion (Fig. 4a), i.e. the volume equals the difference in surface height profile measured before and after cuboid submersion using high-resolution surface profilometry (Sensofar PLµ2300). To minimize random errors such as air entrapment, the choice of liquid, reservoir size and reservoir material were carefully optimized in a separate study [7].

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6 The cubes’ thickness is limited by the sheet metal, while increasing the cuboids’ in-plane dimensions reduces the volume error due to the EDM positioning error to a theoretical maximum of only 66%, while considerably sacrificing spatial resolution.
prior to indentation experiments, through a dedicated partial-homogenization heat treatment that yields a spatially homogeneous matrix but leaves the voids unaffected (Fig. 5). Damage can then be calculated using $D_H = 1 - H_D/H_0$ or $D_E = 1 - E_D/E_0$, as in the original indentation methodology [3,26].

Following this approach (Fig. 1g), cross-sectional slices are cut with EDM along the pre-deformed sample length and subjected to the partial-homogenization heat treatment (Fig. 5), followed by load-controlled microindentation experiments to a load of 4 N (CSM Instruments Micro-indenter with Berkovich tip), while local hardness and elastic modulus are obtained using the Oliver–Pharr methodology [25]. To minimize statistical uncertainty due to preparation-induced surface hardening, the cross-sectional slices are precisely prepared to a final RMS roughness of $0.1\mu m$ using successive grinding and polishing, for which all slices are mounted collectively.

Fig. 5 shows resulting hardness and modulus vs. local strain data, which is the average of six specimen slices to assess intersample reproducibility. As expected, these indentation experiments on partially homogenized samples show much higher reproducibility in both hardness and modulus than the original indentation methodology [9,10]. Both hardness and modulus values stay constant at low strain levels and show a sudden drop at a strain of $0.2$. Calculated damage parameters $D_H$ and $D_E$ are also shown in Fig. 5. Apart from the same onset of damage accumulation at $0.2$ strain, $D_H$ and $D_E$ also reveal approximately equal absolute values (within statistical uncertainties), increasing confidence in the partial-homogenization heat treatment approach. However, the damage values are interestingly even higher than for the density methodology (Fig. 4), calling for further comparison with another mechanical approach, namely the micropillar-compression methodology.

4.2. Micropillar-compression methodology

Plasticity-induced microstructural effects which render the conventional indentation methodology unreliable [9] can also be avoided by probing purely the elastic response (i.e. modulus), locally, by flat-punch compression tests on micropillars (see inset image in Fig. 6). As detailed in Ref. [12], a field of $0.2 \times 0.2 \text{mm}^2$ square pillars is produced at the sample width cross-section, by two perpendicular sets of 0.4 mm deep parallel EDM cuts (Fig. 1h). To minimize surface plasticity effects, the pillars are compressed (CSM Instruments Micro-Indenter with a $350\mu m$ diameter flat punch) to 2.0 N (i.e. roughly half the material yield

![Fig. 5. Above, schematic of the partial-homogenization heat treatment; below, results of indentation measurements on the partially homogenized microstructure. Hardness (blue), modulus (red) and corresponding damage values are plotted for the same strain levels and with same color coding. Error bars represent standard deviations of the mean. (For color interpretation in this figure legend the reader is referred to see the web version of this article.)](image1)

![Fig. 6. Above, top view optical and electron microscopic image of the pillars; below, compression modulus, for as-deformed and partially homogenized samples, vs. local strain (dashed lines indicate zero-strain reference moduli measured outside the clamps), and corresponding damage values. Error bars show standard deviation of the mean.](image2)
The measured modulus evolutions, for as-deformed and partially homogenized samples, and their corresponding damage evolutions are presented in Fig. 6. Statistical uncertainty is clearly large. Nevertheless, both damage evolutions show no (or very little) damage up to strains of 0.4 and 1.0 N and the damage parameter is obtained from $D_E = 1 - E_D/E_0$, where $E_0$ is measured on pillars outside of the clamps. Finally, $D_E$ is coupled to the local strains to reveal its evolution with deformation. Additionally, to confirm that modulus values are not affected by pre-deformation-induced microstructural changes (e.g. texture evolution, residual stress build-up, etc.), some compression experiments are repeated on samples that received the same partial homogenization heat treatment as discussed in Section 4.1.

Regarding statistical accuracy, the area and volume methodology, including specimen preparation, systematic error of the area fraction methodology, also found for the indentation results (Fig. 5), confirms the insensitivity of the micropillar-compression methodology to pre-deformation-induced microstructural changes. A complete comparison of the methodologies is presented next.

5. Comparative analysis of the investigated methodologies

To compare the six theoretically equivalent methodologies, their damage evolutions are replotted in Fig. 7. For visual clarity, the geometry-based and mechanical property-based methodologies are plotted separately with different y-axes, except for the density methodology shown in both. All methodologies reveal two different regimes of damage evolution: an initial regime of slow damage accumulation followed by a regime of accelerated damage accumulation. However, clear differences are also exposed.

Regarding statistical accuracy, the area and volume fraction methodologies are clearly superior to the density methodology and especially to the three mechanical methodologies. The low precision of the mechanical methods, attributed to the multitude of preparation steps, can (of course) be improved (to some extent) by measurement repetition, but remains troublesome considering that damage levels at necking for typical engineering metals are on the order of a few per cent [5]. Regarding systematic accuracy, however, the three mechanical methodologies (indentation hardness, indentation modulus and elastic compression modulus) agree within experimental uncertainty (except at very high strains, addressed below), whereas the density methodology measures (for strains above ~30%) significantly lower damage values, which are still significantly higher than those of the SEM-based area fraction methodology, which in turn are significantly higher than those of the XµT-based volume fraction methodology (except for strains above ~55%). The latter may have been expected considering the limited spatial resolution of XµT (and all 3-D imaging techniques), preventing detection of small voids. However, Fig. 7 confirms that the area fraction methodology also fails to detect a significant portion of the damage, even though high-resolution SEM images were used.7 This may be caused by nanosized defects (voids, cracks, etc.) that are effectively removed with all (standard) specimen preparation techniques. The density methodology should be sensitive to all nanosized voids, but still seems to miss a significant part of the damage captured by the mechanical techniques. We explain this as follows: whereas geometry-based methods are fundamentally limited to their resolvable geometric damage spectrum and thus incapable of capturing deformation-induced “volumeless” damage (e.g. closed microcracks), the mechanical methodologies probe the mechanical response of the investigated measurement volume, including the influence of all (closed) defects that may still have a noticeable overall softening effect on the mechanical response.8 It is exactly this effect on mechanical response that is used in

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7 Extended testing and careful investigation of all measurement uncertainties of the area fraction methodology, including specimen preparation, imaging (magnification, contrast adjustment, etc.), image analysis (e.g. thresholding), and their reproducibility, showed a maximum total systematic error of ~2.

8 This “damage spectrum” difference between mechanical and geometry-based methods, here observed in samples pre-deformed under tension prone to void growth, is probably larger for pre-deformation modes prone to microcracking (e.g. shear, torsion, fatigue).
CDMs and thus should be captured accurately in the damage parameter measurement. Therefore, these results suggest that, despite their relatively limited precision, the mechanical methodologies have (much) better systematic accuracy than the geometry-based methodologies.

A general complication for all methodologies is strain localization. In most fracture modes, strong strain and damage gradients develop, which limit the number of data points measurable at the same strain, increasing statistical uncertainty. Alternatively, damage gradients within the measurement volume introduce systematic errors, whereas a decrease in measurement volume size below the minimum representative microstructural volume increases statistical uncertainty. These effects, which may explain the large differences between mechanical methodologies for strains above ~65% (Fig. 7), limit all methodologies, and thus CDMs, to relatively low damage values.

Isotropic damage models assume an equal damage effect for all directions around a material point, enabling the use of a single (scalar) damage parameter. Since this is rarely the case in engineering materials, an ideal damage quantification methodology should also be able to provide different components of what is sometimes called the damage tensor. For anisotropic materials, at least three damage parameters are needed to describe the effect of material texture on damage evolution. These parameters can be measured with any of the six methodologies, by machining cross-sections in at least three directions of the pre-deformed samples and measuring the damage values at each cross section. Even for isotropic materials, however, (two) additional damage parameters are needed to describe different damage behavior in tension vs. compression (e.g. due to closed microcracks), the so-called “unilateral effect”, and in normal vs. shear loading. To describe a unilateral effect in anisotropic materials at least six damage parameters are required, and at least nine parameters to capture also the anisotropic shear loading effects. Measurement of such loading direction effects may perhaps be achieved by extending the micropillar-compression methodology to apply tension and/or shear loading to the pillars (e.g. by producing a gripper on the pillars); however, the two indentation methodologies seem intrinsically limited to probing the damage effect only under compression, whereas the geometry-based damage definitions are fundamentally uncoupled with respect to load application direction. It is clear that for these cases it becomes incrementally more complicated to (accurately) measure the complete damage evolution.

A final discussion point is the ability of these different strategies to identify the relevant microscopic damage mechanisms. Even though the mechanical property-based methodologies measure the continuum damage parameter with better accuracy, it is clear that these methodologies do not provide information on the type of damage mechanisms. Consequently, one would always require simultaneous analysis with one of the (2-D or 3-D) microscopic techniques (e.g. SEM or XµT) to find the underlying reason behind the observed trend(s) in continuum damage parameter evolution.

A final qualitative comparison of all six methodologies is given in Table 1 (which is also a salute to the well-known damage methodology selection chart (the “wine chart”) in the pioneering work of Lemaitre and Dufailly [3]).

### Table 1

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6. Conclusions

This work presents an in-depth comparison of six theoretically equivalent damage quantification methodologies for determination of material-specific damage parameters for CDMs, through analysis of ductile damage evolution in industrially relevant dual-phase 600 steel as a case study. The results obtained show that geometry-based methodologies (i.e. density, SEM-based area fraction and XµT-based volume fraction) have higher precision than mechanical property-based techniques (i.e. indentation hardness, indentation modulus and micropillar compression modulus), but probe a more limited damage spectrum, and hence, introduce significant systematic errors into their continuum damage value. However, imaging techniques such as SEM or XµT remain valuable for identifying, for example, the exact damage nucleation strain level or the damage micromechanisms. The three mechanical property-based methodologies have higher accuracy and probe a larger damage spectrum, probably because mechanical tests more effectively capture the detrimental mechanical influence of different damage morphologies (from nanovoids to volumeless cracks), and it is exactly this detrimental influence that is introduced by damage parameter(s) in CDMs. However, the large statistical uncertainty may obscure the damage evolution at necking in many engineering materials, while extension to (very) high strains or damage levels or incorporation of loading direction effects or material anisotropy is
far from trivial. Perhaps, in this light, physical micromechanics-based (multiscale) modeling approaches may look more attractive than the “black box” continuum damage approaches, though both present significant experimental challenges.

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