Experimental analysis of strain path dependent ductile damage mechanics and forming limits

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1. Introduction

The last decades were marked by the introduction of advanced high-strength steels (AHSS) in the sheet metal forming (SMF) industry, e.g. the dual-phase (DP) and transformation-induced plasticity (TRIP) steels. Driven by strong weight reduction demands, there are also ongoing efforts to replace steel in automotive applications with low-weight aluminium alloys, especially the Al5xxx and Al6xxx series. These developments introduced new challenges to the industry (Sadagopan and Urban, 2003; Golovashchenko, 2005; Metzger et al., 2006).

The first challenge concerns the mechanical performance of these new alloys during forming and/or service. Although many AHSS grades are reported to have superior strength-to-ductility ratios, it is also reported that the improved strength-to-weight ratio of these steels may be accompanied by a reduction in ductility. Reduced ductility possibly results in failure without significant energy absorption, which is unacceptable for car components that are designed to absorb maximum energy upon crash. Similar problems are also encountered with aluminium alloys. Metzger et al. (2006), for example, reported significant differences in fracture behavior of direct chill cast and continuous cast Al–Mg alloys with identical necking limits. In addition, severe roughening and cracking are reported to be critical limitations for the hemming of some Al6xxx alloys (Golovashchenko, 2005). It therefore remains a challenge for sheet metal manufacturers to optimize the microstructure of these alloys using the feedback from the forming industry and, vice versa, for the forming industry to design the forming operations to yield maximum service performance for a given material. This challenge...
requires an in-depth understanding of the evolution of the microstructure upon deformation leading to the failure of the material.

The second challenge relates to the experimental and numerical tools that are currently employed to predict the limits of forming. The reduced ductility and the increasing risk of ‘neckless’ fracture hampers the predictive capabilities of the conventional tools, which rely on localized necking as the limiting criteria for SMF processes. Unpredicted fractures are attributed to the evolution of damage inside the material, i.e. microvoids (Marciniak and Duncan, 1992). To address this problem and incorporate damage and fracture into these tools, several empirical and microstructure-based fracture criteria have been proposed, many of which are compared by Jain et al. (1999). Furthermore, ductile damage models which aim to describe the interaction of damage growth and plastic flow have also been developed, such as the Gurson–Tvergaard–Needleman family of models (Gurson, 1977; Chu and Needleman, 1980; Tvergaard and Needleman, 1984; Tvergaard, 1981). Using such models instead of or complementary to the experimental data may result in considerably improved predictions of failure in critical applications. However, these models are still not used extensively in the industry in their present form due to the large number of material parameters involved, for which no established experimental identification method exists. Improving on this requires a better understanding of the interaction of material microstructure, forming operation and the possible failure mechanisms.

To address the above-mentioned two challenges, a coupling between the deformation history of a given sheet at a certain strain path and the evolution of its microstructure up to the point of failure is needed.

To track the strain path followed by the material and detect the limits of safe deformation along this path, forming limit curves (FLC’s) are frequently used, which were initiated by the work of Keeler (1961). FLC’s represent the point of necking in different strain paths, which is adopted as the failure criteria in present forming simulations. This is a reasonable assumption for the forming of conventional steels (e.g. high-strength low alloy (HSLA) steels), AHSS’s on the other hand, may fail without a neck, through premature ductile fracture. In such cases, the experimentally obtained FLC’s represent a combination of necking and (also) fracture limits. In an effort to separately probe these two criteria, many researchers have studied fracture limit curves (FrLC’s) of relevant materials (notable examples include Ghosh (1976), Bourcier and Koss (1982), Jain et al. (1999), Takuda et al. (2000), Han and Kim (2003), Lee et al. (2004), Narayanasamy and Narayan (2006), Yu et al. (2007)). However, reported FLC’s and the FrLC’s are not coupled to the evolution of the underlying microstructure towards fracture, with the exception of the works of Jain et al. (1999) and Narayanasamy and Narayan (2006). Even in the latter two cases the attention is primarily focused on the final microstructure and not on the evolution of the microstructure through the deformation process. It is therefore difficult to generalize these observations in particular for different materials with different microstructures that are to be formed in a variety of strain paths.

Parallel to the focus on FLC’s, there have been tremendous efforts in observing and understanding the micro-mechanisms involved in the evolution of ductile damage leading to fracture. Predicting ductile fracture requires an adequate understanding of the physical mechanisms involved in this process, which necessitates experimental analyses of each stage of the ductile fracture process, i.e. void nucleation, growth and coalescence. Upon deformation, microvoids have been reported to nucleate and grow from: second-phase particles through matrix particle decohesion and particle cracking (Puttick, 1959); triple junctions between grains of identical phases (Tan et al., 2007); interfaces between different phases (Greenfield and Margolin, 1972); slipband intersections (Gysler et al., 1974); dislocation cell boundaries (Gardner et al., 1977); etc. The anisotropic nature of these mechanisms is clearly shown (Benzerga et al., 2004). Effect of damage evolution on the global mechanical behavior and failure of industrially relevant materials such as interstitial-free (IF) or dual-phase (DP) steel has also been examined, see for example, Ganser et al. (1998) and Sarwar et al. (2007), respectively. However, similar to the literature on forming limits, papers on damage and ductile fracture mechanisms are generally also case-specific and a connection to FLC’s and forming parameters, such as the strain path, is not provided.

This brief literature overview presented highlights the gap in the literature connecting the mechanical behavior of new, advanced materials and the evolution of their microstructure, up to the point of failure. The missing link between damage and strain paths constitutes an important aspect in this sense. It is generally believed that damage evolution, for most materials, occurs only after the point of localized necking, and always faster in forming operations that involve a biaxial strain path. For advanced metals, this assumption is correctly questioned in the literature. As stated in Jain et al. (1999) “…The works in the literature emphasize on improving the ductile fracture criteria available, however, a detailed examination of the effect of strain or stress state on the occurrence of a well-defined and quantitatively characterized mode of failure is needed…”. The present work aims to close this gap and address the two challenges advocated. Two different steels with different formability are deformed along different strain paths and the resulting strains, FLC’s and FrLC’s are measured and interrelated. In addition, full characterization of microstructural failure mechanisms have been carried out using a scanning electron microscope (SEM). Through extensive fractography analyses, microstructural damage evolution mechanisms are determined. The activity of these observed damage micro-mechanisms is examined at different strain paths and at different stages of deformation. An intrinsic connection is therefore identified between the FLC and the FrLC of the tested sheet metals and the microstructural evolution and failure mechanisms.
2. Experimental methodology

In a numerical approach, a fruitful strategy to study the influence of damage evolution on formability characteristics of sheet metal is to switch on and off damage evolution and observe its effects on the global formability of the sheet. Obviously, this is not possible experimentally. Nevertheless, with a suitable choice of the materials a similar effect can be achieved qualitatively. For this purpose, two steels at different extremes of formability are tested comparatively in this work: an advanced high-strength steel (DP) and a high formability steel (IF). The initial microstructures of the DP and IF steels are presented in Figs. 1 and 2, respectively.

DP steel is composed of ferritic grains with martensite located at the grain boundaries (Fig. 1(a)). Limited amount of relatively soft, elongated manganese-sulfide particles and hard aluminium oxide inclusions (Fig. 1(b)) are also identified in the as-received microstructure. In terms of mechanical performance, these alloys are known to exhibit relatively high global formability combined with an enhanced strength due to the co-existence of ferrite and martensite phases (Honeycombe and Bhadeshia, 1995). However upon deformation, local high stresses induced by the different formabilities of ferrite and martensite is also reported to trigger damage nucleation (Ahmad et al., 2000). The tested DP sheets have a thickness of 1 mm, and a normal isotropy ratio, the so-called ‘r-value’, of approximately 1 (i.e. absence of normal anisotropy).

The IF steel, on the other hand, is a single phase material with equiaxed ferritic grains (Fig. 2(a)). It has limited number of inclusions, which are mostly precipitated in the grain boundaries. These second-phase particles are identified as aluminium oxides (Fig. 2(b)) and titanium nitrides through EDX analysis. With its low carbon content, fully ferritic microstructure and relatively low amount of inclusions, IF steel is one of the most formable steels used by the sheet metal forming industry. The thickness of the IF sheet material is 0.7 mm, and its r-value is approximately 2 (indicating an increased resistance to thinning).

In order to deform these steels along different strain paths, specimens of different starting geometries (i.e. Nakazima strips) are prepared (Fig. 3). The tests are done according to EN ISO 12004-2:2008, with the only exception that the punch diameter was 75.0 mm instead of 100 mm and the die inner diameter was 84.6 mm instead of 105 mm. For the equi-biaxial strain path, full circular blanks (radius of 166 mm) are used. To achieve a wide range of strain paths, the blank width is decreased to 140, 100, 80 and 60 mm. A regular measurement grid is

![Fig. 1.](image1.png) (a) Undeformed microstructure of DP steel (transverse (TD) and normal (ND) directions are indicated with the arrows.) (b) An aluminium oxide inclusion in DP steel.

![Fig. 2.](image2.png) (a) SEM micrograph of the microstructure of undeformed IF steel, composed of equiaxed ferritic grains. (b) An aluminium oxide particle in IF steel.
applied to the surface of each undeformed blank prior to deformation. The specimens are tested up to fracture using a displacement-controlled hemispherical punch. Sufficient lubricant is used to overcome friction effects between the punch and the tested sheets. All specimens are tested such that the major strain axis coincides with the "transverse direction", as this testing orientation shows necking and fracture at lower strains. Along with the fracted specimens, four specimens from each geometry (i.e. strain path) are deformed to different levels of final strains, to determine the strain path followed by each geometry up to fracture. This is accomplished by adopting different maximum punch displacements (65%, 75%, 85% and 95% of the punch displacement at fracture). Local strains are obtained by measuring the separation of the grid points using image correlation software (PHAST™). The necking strains for each strain path are plotted and together these necking strain points form the FLC.

To plot the FRLC curves, the displacement of the grid points at the point of crack initiation need to be determined in the major and minor directions. Measurement of the minor strain is straightforward and can be easily done post-fracture. However, the determination of the major strain is more challenging, and (as explained by Bourcier and Koss (1982)) can be done either directly from the displacement of grid points, or indirectly using the thickness at the point of crack initiation and volume conservation. Although many researchers choose the latter methodology (e.g. Ghosh, 1976; Takuda et al., 2000; Jain et al., 1999), our preliminary tests showed that this approach leads to erroneous results due to the errors introduced in thickness measurement. Therefore, in this work, an improved version of the former (direct) methodology is used to determine the local major strains at the point of fracture. For this, the distance between grid points on each side of the crack is measured using the 3D position of each data point obtained from the image correlation software. Then, the crack opening width is measured directly through optical microscopy and subtracted from the calculated displacement.

To investigate microstructural changes in the IF and DP steels along certain strain paths (i.e. uniaxial tension (UAT), plane strain tension (PST) and equi-biaxial tension (BAT)), metallographic examination is carried out with a scanning electron microscope (Philips XL30 ESEM-PEG), using both secondary and backscatter electron modes. The specimens are deformed to different levels of final strains and microstructural analysis is carried out on their cross-sections. To detect relevant changes prior to localization, the specimens that are deformed to an accumulated strain just below the FLC are prepared and analyzed from the top surface. To reveal the changes after localization, a cross-section of the localized neck is analyzed in each fractured sample (direction 1 in Fig. 4(b)). In order to directly measure thickness strains at necking and at fracture, and to quantify damage evolution (i.e. void area fraction), SEM images obtained at direction 1 are analyzed using a public domain image processing software (ImageJ). Furthermore, using the same samples, the final failure mechanism and the amount of damage evolution at the points of fracture is analyzed by examining the fracture surfaces (directions 2 and 3 in Fig. 4(c)).

3. Strain path and microstructure evolution up to failure

In the Nakazima test, the loads exerted by the punch on the sheet induces the deformation leading to fracture. This process involves a number of different stages, i.e. elastic, uniform plastic and localized plastic deformation, until the point of failure. In the following subsections, the changes in the microstructure of each IF and DP steel specimen are analyzed at different levels of deformation, and subsequently coupled to the strain path experienced by each specimen.

3.1. Deformation up to localization

The first stage in the deformation of the specimens can be referred to as the uniform deformation stage, as the strain profile remains approximately constant over the length of the sheets with an increase towards the center (see e.g. 65%, 75% and 85% profiles in Fig. 5 for the UAT specimens of (a) IF and (b) DP steels, and also the corresponding local strain distributions in the same samples (c)). Due to the different Nakazima strip starting geometries, each specimen experiences a different strain path, as shown in Fig. 6(a) for IF steel and (b) for DP steel. Note that for all geometries the deformation is initiated in a linear path towards the point of necking (FLC), however, the slope of these linear paths (i.e. strain ratios (β) up to FLC in Table 1) is different for the same geometries of the two steels due to the difference in normal anisotropy.

![Fig. 3. Nakazima strips of different starting geometries are deformed in a hemispherical punch test to generate the different strain paths in the center of the test specimens, as schematically indicated on the major strain versus minor strain graph.](image)

![Fig. 4. Viewing directions 1, 2 and 3 for fractography analysis: (a) full FLC specimen from the top, (b) thickness cross-section perpendicular to the crack and (c) fracture surface.](image)
The higher $r$-value of IF ($r = 2$) compared to DP steel ($r = 1$) indicates a greater resistance to thinning for IF steel, which explains the observed lower minor strains (i.e. lower values of $\beta$) for all strain paths except (of course) biaxial strain, as induced by the volume conservation under plastic flow.

The SEM micrographs at this early stage reveal the elongation of the originally equiaxed ferritic grains in IF steel. The evolution of the major strain along the longitudinal loading direction of (a) the IF specimen and (b) the DP specimen, both deformed in UAT. The major strain distribution plots of the entire samples are also given in (c).

Fig. 5. The evolution of the major strain along the longitudinal loading direction of (a) the IF specimen and (b) the DP specimen, both deformed in UAT. The major strain distribution plots of the entire samples are also given in (c).

Fig. 6. Strain paths leading towards the FLC (dashed line) of (a) IF steel and (b) DP steel specimens.
specimens (e.g. UAT specimen in Fig. 7(a)). Difference in formability between the soft ferritic matrix and the embedded hard inclusions induces early nucleation of microvoids upon deformation (Fig. 7(b) and (c)). However, due to the low number of “damage sources” (i.e. inclusions), the total void fraction remains low in all strain paths and its effect on the microstructure evolution in IF steel is therefore limited. Accordingly, comparing the damage accumulation along different strain paths, no significant difference is observed between the different IF steel specimens, see Table 1. There is, however, a difference in the morphology of many of the analyzed microvoids: the voids nucleated in the uniaxial strain path (e.g. Fig. 7(b)) have less sharp tips compared to those nucleated in biaxial strain path (e.g. Fig. 7(c)). This is attributed to the positive minor strain in the BAT path, which drives the enlargement of the nucleated voids in the plane, forming penny-shaped voids, which are “thinner” than those nucleated in the uniaxial regime.

In the DP sheets, deformation is governed by the flow of the ferritic grains, similar to IF steel. The hard martensite phase covering the ferritic grains only constrains the ferritic deformation, while its own deformation remains primarily elastic (Fig. 8). The difference in flow characteristics of these two phases, along with the higher inclusion density, activates some fundamentally different damage mechanisms, resulting in significantly more damage evolution compared to IF, as tabulated in Table 1 and shown in Fig. 9(a). Based on extensive fractographic analysis these damage mechanisms were identified as particle–matrix decohesion, particle cracking, martensite fracture, and grain boundary decohesion, see Fig. 9(b)–(f). For these samples, microvoid accumulation is observed throughout the entire cross-sections, however, damage accumulation is slightly higher towards the center of each sample, probably due to the slightly higher stress triaxiality in the center. Also, there is a significant difference between the

<table>
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<th>Material</th>
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<th>Strain ratio, $\beta$</th>
<th>Void area fraction (%)</th>
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<td>PST</td>
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<td></td>
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<tr>
<td></td>
<td>From FLC to FrLC</td>
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Table 1
Strain ratios (i.e. ratio of minor strain to major strain) and the accumulated damage (i.e. void area fraction) of IF and DP steel samples at different strain paths. The standard deviation in the damage measurement is ~0.02%.

Fig. 7. (a) The microstructure of the IF steel prior to localization (loading direction (LD) indicated with the arrow). (b) Void growth from aluminium oxide particle in IF steel in UAT strain path. (c) Void growth from titanium nitride particle in IF steel in the BAT strain path.

Fig. 8. The martensite phase surrounding the ferrite grains hardly deforms and therefore constrains the deformation of ferrite. The micrographs show snapshots of the microstructure deformation in UAT, at low (left) and high (right) strain levels. The loading direction is indicated with the white arrow.
amount of damage accumulation in different strain paths already prior to localization. Near the point of necking, the area fraction of nucleated voids is more than five times larger in the biaxial-strained specimen compared to the uniaxial-strained specimen (Table 1). Evidently, the identified damage mechanisms for DP are more active under biaxial stress condition, as a result of the higher stress triaxiality.

3.2. Localization

As the strain levels increase, the deformation starts to localize near the specimen center (see, e.g. the 95% profiles in Fig. 5). The points of localized necking, determining the FLC’s, for the tested steels are shown in Fig. 6.

The obtained FLC’s reflect their typical ‘v’-shape, i.e. the level of major strain at which localization takes place depends strongly on the strain path that is followed (Fig. 10). Since the IF steel develops only a negligible microvoid fraction prior to necking, the FLC is not affected by damage evolution. The v-shape of the FLC logically reveals easier necking in the PST strain path, since necking requires plane strain path to develop (Marciniak and Duncan, 1992). In the UAT and BAT strain paths, however, different constraints delay localization. In the UAT path, the negative minor (in-plane) strain is effectively reducing

Fig. 9. (a) Damage evolution in DP steel before strain localization: void nucleation and growth is observed through the mechanisms of (b) particle–matrix decohesion, (c and d) particle cracking, (e) martensite fracture and (f) phase boundary decohesion.
the thinning of the sheet (in thickness direction) with increasing major strain, giving rise to a FLC value with higher major strains. Along the BAT path, on the other hand, plastic incompressibility is expected to trigger early localization. However, Marciniak and Duncan (1992) argued that under biaxial straining conditions there are geometrical constraints inhibiting the localization zone to reach the plane strain path, necessary for the neck to grow. This effectively delays localization, causing extensive thinning in the equi-biaxial path compared to UAT and PST specimens. Scaled thicknesses of these samples at localization are depicted in Fig. 11.

As explained earlier, DP specimens develop significant damage evolution prior to the point of localization. It is therefore reasonable to expect that damage itself triggers local softening leading to localization in this material. Correspondingly, the FLC of DP, which has a similar v-shape, reflects lower major strains compared to the FLC of IF steel (Fig. 6(b)) for almost all strain paths. Only towards the equi-biaxial path, the FLC’s of DP and IF steels approach each other (Fig. 10). This can be explained by a more pronounced geometrical constraint to reach plane strain condition (see above) for DP compared to IF steel. That is to say that since the ability to reach a plane strain state in the necking zone depends highly on the deformability of the phases that constitute its microstructure, the
martensite surrounding the ferrite grains limits the deformation of ferrite thereby enhancing the aforementioned geometrical constraint. As a result, for DP more additional thinning is observed in the BAT strain path relative to the UAT and PST paths (Fig. 11).

3.3. Localized deformation to fracture

Further thinning following the formation of a localized neck eventually leads to fracture. In the case of the IF steel, a nearly vertical path is traced from the FLC to the FrLC for positive minor strains, as seen from the strain paths in Fig. 12(a) and tabulated strain ratios in Table 1. The neck is constrained in the minor strain direction (along its axis), yielding a plane strain state in the neck and thus a vertical strain path. For the negative minor strains, however, the strain paths between FLC and FrLC are only partially bend towards a vertical path. This seems to be caused by the decrease in constraint of the neck in the minor direction with decreasing width of the Nakazima strips. However, these paths also do not continue along the same slope as that of the pre-FLC path either, as the constraint to deformation in the (in-plane) minor direction should be higher inside the localized neck (above the FLC) than in the whole specimen during homogeneous deformation (below the FLC). The non-zero negative minor strain results in a longer path from the FLC to the FrLC for the uniaxial strain path specimens compared to the other specimens. Consecutively, examining the cross-sections of the IF specimens, a clearly longer neck is observed in the uniaxial strain path compared to other strain paths, confirming the relatively higher accumulated major strain in this path (Fig. 13(a)).

Let us now concentrate on damage mechanisms in IF steel samples. As mentioned earlier there is no damage evolution in IF steel prior to localization, however following necking, a limited amount of damage accumulates locally near the fracture surface in all strain paths (e.g. Fig. 13(b)). For the UAT specimen, it is mentioned above that the material is more ‘formable’ in the major strain direction, as a result of the less stringent constraints exerted on the localized neck in the minor strain direction. This results in a significantly higher degree of damage accumulation in the uniaxial strain path compared to plane strain or the biaxial strain paths (Table 1). More microvoids are nucleated and the nucleated microvoids grow to a larger size before a critical void size and density is reached at which fracture occurs. In the biaxial strain path, as shown above, the sheet goes through excessive thinning prior to the formation of the localized neck (Fig. 11), during which voids with relatively sharper morphologies are nucleated (Fig. 7(c)). As a result, beyond the point of necking, the critical size of voids at fracture is relatively smaller compared

![Fig. 13.](image-url)
to uniaxial strain path, while the spacing of the voids is relatively larger. On a more general note it should also be mentioned that the thickness at fracture is almost independent of the strain path followed, although localization occurs at significantly different thickness strains (Fig. 11).

Let us now concentrate on DP. Examining the strain paths in Fig. 12 and the tabulated strain ratios in Table 1, it is apparent that all strain paths turn to a plane strain path from FLC onwards for positive minor strains. However, again the UAT strain path does not completely turn to the plane strain stress state inside the neck. But the effect described above for UAT in IF steel is relatively more pronounced for DP steel, probably due to the constraints exerted on the ferrite grains by the martensite phase.

Unlike IF steel, fracture occurs without significant thinning beyond localization in all DP steel samples (Fig. 11). This is understandably due to the significant amount of damage accumulation prior to localization, tabulated in Table 1. However, the influence of damage is observed to be different in different strain paths. For positive minor strains and especially in the equi-biaxial tension path, the FrLC is closer to the FLC (Fig. 12). Comparing the overall damage density in the analyzed micrographs, it is observed that damage accumulation is more pronounced in biaxial deformation compared to uniaxial tension (Fig. 14(b)), although only a very small portion of this damage is accumulated after necking. Apparently, in the BAT strain path a critical damage is already reached at the point of localization, leading to failure without significant thinning or damage evolution following localization, consistent with the reduced distance between the FLC and the FrLC at that point. The relatively small length of the localization zone in the biaxial strain path specimen (Fig. 14(a)) and the homogeneous damage distribution in this local neck (compared to the damage accumulation that is increasing towards the fracture surface in UAT and PST specimens (Fig. 14(b))) both agree with this analysis.

In the uniaxial strain path, on the other hand, fracture apparently occurs only at a somewhat higher damage density, causing a build up of damage towards the crack, consistent with the larger distance between FLC and FrLC and the measured damage values in Table 1. These observations are in accordance with the effect of the negative minor strain in the uniaxial strain delaying fracture even after the formation of the localized neck.

3.4. Fracture

Evidently, localized deformation of sheet metal ends in fracture. The observation of the fracture surfaces of the tested specimens at different scales completes the analysis carried out above. It should be stated, however, that the crack initiation triggers complex strain states that cannot be correlated to the strain path before fracture, in a

![Fig. 14.](image)
Nevertheless, this analysis has significant added-value, as many researchers use fractography as their main source of information for failure analysis.

Macroscopic fracture examination of the IF and DP sheets deformed along different strain paths reveals that, apart from the IF–UAT specimens, all specimens fail with a macroscopic crack running perpendicular to the direction of major strain (Fig. 15(b)). For IF–UAT specimen, localization through in-plane shear bands is observed, which triggers the final macroscopic crack fracturing the sheet.

Even when the in-plane fracture path is similar, through-thickness fracture may still proceed in different ways, e.g. cup-and-cone, through-thickness shear fracture, etc. For both the IF and DP sheet (Figs. 13 and 14), the final fracture predominantly occurs via a through-thickness shear fracture mechanism. However, a detailed examination of the fracture surface reveals that alternating regions of either cup-and-cone fracture or through-thickness shear fracture mechanisms can be identified, see e.g. Fig. 16 for the plane strain IF specimen. Micro-analysis of crack propagation (e.g. through the distortion of dimples) reveals that the fracture always initiates through a cup-and-cone crack which makes a transition to through-thickness shearing at some point after the crack starts propagating in the minor strain direction. In the shear fracture zone the crack starts at the top or bottom of the specimen, where damage has already nucleated and weakened the material, and the crack propagates by shearing of the two specimen sides from each other.

Fig. 16 also clearly shows microvoid accumulation in the center of the cup-and-cone structure, but not in the shear fracture zone. These void-free surfaces could be a result of the smearing effect of the two surfaces as the fracture proceeds (as also reported by Bron et al. (2004)), or of extensive shearing of the voids.²

The examination of the fracture surface of the IF steel reveals two interesting points. First, there are two groups of dimples in the fracture surface as shown in Fig. 17(a). The first group of these dimples are large (10–20 μm) with

² Note that the co-existence of different fracture-induced morphologies and the reported smearing effect complicates any analysis of the failure mechanism that leads to fracture. Our observations show that it is possible to have two contradictory SEM images from the same fracture surface, one completely filled with dimples and the other without any dimples. It is therefore more objective to also examine cross-sections in such specimens, even though many researchers prefer otherwise.
visible inclusions inside and the second group of dimples are smaller (2 μm or below) without any inclusions inside. Second, for IF steel, the BAT specimen has an almost dimple-free surface as shown in Fig. 17(b). This is consistent with the observation made from the cross-section images in Fig. 13 for IF steel, where a lower number of voids were revealed in biaxial strain path specimens.

The DP steel shows a through-thickness shear fracture in all strain paths (Fig. 18(a)). Unlike the IF steel, the fracture surface is completely filled with dimples (Fig. 18(b)), emphasizing the effect of damage in the prior deformation. However, compared to the IF specimens the average size of the voids in the fracture surface of DP specimens is smaller (~3 μm). This is essentially due to the fact that IF steel is more formable around the voids. Consequently in DP steel, once the material reaches the point of fracture, the high concentration of voids and the low formable character of the microstructure induced premature rupture without any significant smearing effects.

4. Conclusions

From the detailed analysis presented above, the following generalized conclusions can be drawn regarding the influence of ductile damage on sheet metal forming limits:

For formable, homogeneous microstructures (i.e. microstructures with few damage mechanisms):

(i) Damage does not play a role before (or on) localization, but only beyond localization.

(ii) Aforementioned influence of damage is more “critical” in strain paths with positive minor strains (e.g. equi-biaxial strain path), as geometrical constraints delaying localization causes excessive thinning and the sharp, elongated morphology of the nucleated voids cause higher stress concentration.

(iii) Along strain paths with negative strain ratio, the negative minor strain effectively retains the thinning of material (even after localization), making it more damage and fracture resistant.

On the other hand, for microstructures with phases of different deformation characteristics (i.e. microstructures with easily activated damage mechanisms):

(i) Damage plays a significant role on localization, and on fracture.

(ii) Damage accumulation takes place even before localization, and possibly leads to inhomogeneities in local softening triggering localization, causing relatively lower forming limits (e.g. either FLC or FrLC).
(iii) As significant amounts of damage accumulates prior to localization, beyond the point of localization these metals are sensitive to fracture, and fail without significant further thinning. Consequentially, fracture limit curves of these materials are very close to their forming limit curves.

(iv) The overall damage accumulation is observed to be significantly higher in the strain paths with positive minor strains, already prior to localization, causing a higher fracture sensitivity upon necking.

In light of these findings, important suggestions can be made for improving the applicability of AHSS’s in industrial applications. For example, FLC of these new steels can be increased to higher strains by somehow inhibiting the damage micro-mechanisms that are active before localization. Also, formability loss in these materials can be diminished by preventing the hard phase (e.g. martensite in DP steel) to completely cover the circumference of the soft parts (e.g. ferrite in DP steel), although this comes easily at expense of the yield strength when the martensite morphology is not controlled perfectly and becomes too open.

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