Comparative Study on Damage Evolution during Sheet Metal Forming of Steels DP600 and DP1000

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Abstract. Two different dualphase steels of significantly different strength properties are compared with respect to microstructural configuration, chemical composition, mechanical properties, and ductile damage mechanisms. The investigated steel grades DP600 and DP1000 show remarkable differences in terms of martensite phase fractions, yield and ultimate tensile strength, and ductile damage evolution behaviour. In particular it turns out that steel DP600 experiences a relatively early ductile damage initiation, but only a moderate rate of damage evolution, whereas for steel DP1000, contrary behaviour is observed. This steel shows a comparably late ductile damage initiation with rapid subsequent damage evolution. This behaviour is expressed by the ratio of fracture strain over strain at damage initiation. Its origin lies in both the martensite phase fraction and the individual strength properties of the ferritic and the martensitic phases, which will provoke inhomogeneous plastic strain distributions in the materials’ microstructures. The more pronounced these differences are, the higher the local plastic strain peaks become, and the earlier ductile damage initiation happens. Nevertheless, with decreasing strength of the ferritic phase, its ability to withstand even large plastic strains without fracturing is strongly promoted, so that the fracture strain in steel DP600 can be shifted to higher values even though the material has undergone early damage initiation.

1. Introduction
Dualphase (DP) steels belong to the first generation of advanced high strength steels. They form a class of low carbon steel which is thermomechanically processed into a multiphase microstructure with a ferritic matrix and dispersed martensitic islands. The comparably soft ferritic matrix mainly contributes to the material’s formability, whereas martensite mainly delivers the strength. DP steels show an increased formability compared to ferritic-pearlitic steels of similar strength. Tasan et al. reported on the deformation behaviour of DP steels, highlighting that under uniaxial tension, the hard martensitic phase mainly constrains the plastic flow of ferrite, but martensite itself remains under elastic deformation mode [1]. These results were to a certain extent confirmed by Shen et al., who additionally found out that for DP steels with comparably high martensite volume fractions, even some plastic reactions can be observed for martensite due to shearing along the martensite-ferrite phase boundary [2]. Further studies in this field applied digital image correlation (DIC) in order to quantify the plastic deformation behaviour of martensite in DP steels. These studies revealed that the strain ratio between ferrite and martensite mainly depends on the individual phase fractions and the carbon partitioning resulting from the DP annealing cycle [3-5].

Noteworthy not only the plastic deformation behaviour, but also the mechanisms of ductile damage show several unique features in DP steels. In addition to the void nucleation at non-metallic inclusions...
as the very conventional damage initiation mechanism, mainly the localization of plastic deformation in the ferrite phase or at the ferrite-martensite interface is reported to be the origin of ductile damage in DP steels. These phenomena were thus categorized into the following modes of void nucleation: martensite cracking, ferrite-martensite interface decohesion or ferrite-ferrite grain boundary decohesion, all of them observed in DP steels of different microstructural configuration in terms of phase fractions and individual grain size distributions [5-10]. Obviously, the inhomogeneity of the plastic deformation controls the mechanisms of void nucleation in DP steels, and this feature is governed not only by the DP steel’s chemical composition (in particular the carbon content), but also the phase fractions of ferrite and martensite, the yield stress ratio of the ferritic over the martensitic phase, and the size, shape and phase fraction of martensite.

Even though there is a huge impact of microstructure configuration on the resulting macroscopic mechanical properties, DP steels are attractive material solutions especially for the automotive industry. High ultimate tensile stress, pronounced hardening, excellent ductility as well as low yield-to-tensile ratio are only some of the advantageous features of this class of modern multiphase sheet materials. As explained, this mechanical property profile mainly results from the distinctive deviation of individual mechanical properties of ferrite and martensite, respectively, so that tailoring both the microstructural configuration and the individual strength properties is the most important measure to adjust the effective properties with respect to the specific requirements identified for the full component. During the last years, these measures were applied to increase the strength of DP steel, so that nowadays, DP steels with guaranteed ultimate tensile strength values between 600 MPa and 1000 MPa are found in automotive components. In this paper, we therefore compare the microstructures, failure mechanisms and damage evolutions in two DP steels of significantly different strength. For this purpose, macroscopic mechanical properties are investigated for steels of the grades DP600 and DP1000, before a series of tensile experiments on notched dog-bone samples, plane strain samples and central hole samples is performed to investigate the stress-state dependent damage initiation and evolution behaviours of the different steels. As a quantitative measure for the damage tolerance, the ratio between the fracture strain and the damage initiation strain are proposed. These parameters result from the application of the modified Bai-Wierzbicki (MBW) model, a phenomenological description of ductile failure for the macroscopic scale.

2. Materials
Two DP steels of significantly different strength properties were selected for the investigations: steel DP600 and steel DP1000. Both were delivered in a thickness of 1.5 mm. Their chemical compositions are listed in table 1. Obviously, the alloying concepts of the two grades are similar, even though steel DP1000 contains increased amounts of C, Mn, Cr, and Ni. Especially the substitutional elements Mn, Cr, and Ni will increase the individual strength of the ferritic phase, likewise contributing to the strength of the multiphase material.

| Table 1. Chemical compositions of steels DP600 and DP1000, mass content in %. |
|---|---|---|---|---|---|---|---|---|---|
| C  | Si  | Mn  | P   | Cr  | Mo  | Ni  | Cu  | N_{SS} |
| DP600 | 0.11 | 0.39 | 1.38 | 0.017 | 0.18 | 0.05 | 0.020 | 0.015 | <0.0001 |
| DP1000 | 0.14 | 0.32 | 1.97 | 0.011 | 0.40 | 0.05 | 0.037 | 0.023 | <0.0001 |

For both steels, microstructures were quantitatively assessed by means of light optical metallographic investigations which were performed on samples that have undergone HNO₃ etching. A comparison of the different microstructures is given in figure 1. The investigations revealed different phase fractions of ferrite and martensite for steels DP600 and DP1000, respectively. In more detail, steel DP600 is composed of 90% of ferrite and 10% of martensite, whereas steel DP1000 is composed of 62% of ferrite and 38% of martensite. Due to the higher strength of martensite compared
to ferrite, this remarkable difference is expected to show significant impact on both yield strength and tensile strength of the two steels.

Obviously, the metallographic investigations not only reveal differences in the individual phase fractions, but they also prove that grain refinement strategies have been applied on steel DP1000 in order to increase its strength properties. Ongoing studies are actually conducted to quantitatively assess these differences. Finally, both materials show a certain tendency of creating banded microstructures, which are resulting from the cold rolling before the annealing cycles were applied.

Figure 1. Microstructures of steels DP600 and DP1000, revealed by light optical metallographic investigations on samples that have undergone HNO3 etching. Significant grain refinement is obvious for steel DP1000.

Isothermal, quasi-static room temperature tensile tests were afterwards conducted on flat samples to quantitatively assess the strength and ductility properties of the two different steels for uniaxial loading conditions. From the tensile test data, flow curves were calculated and fitted according to the Hollomon approach, which relates the stress $\sigma$ to the logarithmic plastic strain $\varphi$ in terms of an exponential expression with two fit parameters $k$ and $n_H$:

$$\sigma = k \cdot \varphi^{n_H}. \quad (1)$$

Afterwards, these flow curves were mathematically extrapolated to strains even beyond the uniform elongation. These data were used for all numerical simulations that will be presented in chapter 4. All the tensile test results are summarized in table 2, revealing significant differences between the strength, ductility and strain hardening properties of the two investigated DP steels.

Table 2. Mechanical properties of steels DP600 and DP1000 as determined in quasi-static, isothermal uniaxial tensile tests performed at 20°C.

| Steel  | $R_p^{0.2}$, MPa | $R_m$, MPa | $A_p$, % | $A_{80}$, % | $k$, MPa | $n_H$ |
|--------|-----------------|-----------|----------|------------|---------|------|
| DP600  | 390             | 704       | 16.5     | 23.0       | 1201    | 0.20 |
| DP1000 | 693             | 1039      | 8.1      | 12.0       | 1476    | 0.10 |

3. The MBW ductile damage mechanics model

The comparison of damage evolution properties is based on a combined experimental and numerical approach. A phenomenological ductile damage mechanics model is applied to reveal the major differences between steels DP600 and DP1000 in terms of stress-state-dependent ductile damage initiation and subsequent damage evolution parameters. For these studies, the MBW model as
presented by Lian et al. [11] is employed. This macroscopic ductile damage mechanics model takes the advantages of both the uncoupled and the coupled models. It makes use of a strain-based, stress-state dependent ductile damage initiation criterion which defines the equivalent plastic strain to ductile damage initiation as a function of stress triaxiality and normalized Lode angle. In isotropic materials, these two parameters can be used to quantitatively address the local state of stress, since they are calculated from the three invariants of the stress tensor or the deviatoric stress tensor, respectively. For strains lower than the damage initiation strain, no influence of damage on the material’s plastic reaction has to be considered, so that a conventional yield potential can be used in numerical simulations of sheet metal forming operations. On the other hand, once the damage initiation criterion is fulfilled, the damage-induced softening has to be considered. For this purpose, a damage variable is coupled into the yield potential, and a corresponding damage evolution law has to be given. The MBW model has been strongly inspired by an uncoupled ductile fracture model initially proposed by Bai and Wierzbicki [12]. With the suggested modifications, especially the damage initiation criterion, the multiscale characterization of both damage and fracture can be achieved. The MBW model describes isotropic hardening whilst neglecting any kinematic hardening. With the equivalent stress denoted as \( \sigma_e \), the yield stress denoted as \( \sigma_{yld} \) and the ductile damage variable denoted as \( D \), its yield potential reads

\[
\phi = \sigma_e - (1 - D)\sigma_{yld} \leq 0. \tag{2}
\]

In agreement to the effective stress concept, the damage variable is also applied on the elastic material constants, with the initial Young’s Modulus \( E_0 \) and the effective Young’s Modulus \( E_{eff} \):

\[
E_{eff} = (1 - D)E_0 \tag{3}
\]

The ductile damage evolution law relies on the definition of an equivalent plastic strain at ductile damage initiation. This parameter however shows a pronounced sensitivity on the local state of stress. With the stress triaxiality denoted by \( \eta \) and the Lode angle denoted by \( \theta \) (note: in case the Lode angle is normalized as suggested in [12], an upper bar is added), the damage initiation criterion reads:

\[
\bar{\varepsilon}^d = \left(c_1^d \cdot \exp\left(-c_2^d \cdot \eta\right) - c_3^d \cdot \exp\left(-c_4^d \cdot \eta\right)\right)\frac{\theta}{\bar{\sigma}^2} + c_5^d \cdot \exp\left(-c_6^d \cdot \eta\right). \tag{4}
\]

Note that all “\( c \)” parameters with upper and lower indices are material parameters that have to be fitted to experimental data. With the damage initiation criterion fully calibrated, the subsequent ductile damage evolution can be expressed as:

\[
D = \begin{cases} 
0, & \bar{\varepsilon}^p \leq \bar{\varepsilon}^d \\
\frac{\sigma_{yld}}{\sigma_f} \int_{\bar{\varepsilon}^d}^{\bar{\varepsilon}^p} d\varepsilon^p, & \bar{\varepsilon}^d < \bar{\varepsilon}^p < \bar{\varepsilon}^f \\
D_{cr}, & \bar{\varepsilon}^f \leq \bar{\varepsilon}^p \end{cases} \tag{5}
\]

Herein, \( \sigma_{yld} \) denotes the equivalent stress in the material point for the instant that the damage initiation criterion has been fulfilled, so that the stress-state dependency is also considered in the subsequent damage evolution phase. Furthermore, the material parameter \( G_f \) determines the energy dissipation between damage initiation and complete fracture of the material point. \( D_{cr} \) is the critical value of the damage variable resulting in complete stiffness degradation of the material point. It becomes obvious that once the ductile damage initiation criterion is fulfilled, the MBW model assumes a linear relationship between the damage variable and the equivalent plastic strain. The model is implemented as a user-defined material model in terms of a VUMAT for Abaqus/Explicit. It is embedded into the framework of the small strain concept. In case the model is applied in the finite strain plasticity, the kinematic transformations are performed first. Then, the constitutive equations governing the finite deformation are formulated using strains and stresses and their rates defined on an unrotated frame of reference. Likewise, the stress updating procedure remains as it was for the small strain formulation. Abaqus adopts this kind of treatment for finite strain plasticity, so that only the small strain theory needs to be considered when user material subroutines are created for this FE solver.
4. Comparison of damage initiation and damage evolution properties

The model parameter calibration relies on a strategy that aims to find the best agreement between experimental and numerical results for tests on samples which are loaded at a variety of different stress states. Usually, for sheet materials the model parameter calibration campaign relies on tensile tests on dog bone samples, central hole samples, plane strain samples and shear samples.

For steel DP600, the set of MBW material parameters has already been calibrated for previous scientific investigations [10,13]. For the present work, only the parameter \( G_f \) has been re-calibrated, since in the preliminary work, also the element edge length \( L \) has been considered in the damage evolution law, which is not the case anymore in the present investigation.

![Figure 2. Force-elongation curves from experiment and simulation. The set-up comprises notched dog bone samples, plane strain samples, and plane strain samples.](image)

\[
\begin{array}{cccccccc}
 & c_1 & c_2 & c_3 & c_4 & G_f & D_{cr} \\
DP600 & 0.43 & 1.14 & 0.12 & 0.98 & 4000 & 0.1 \\
DP1000 & 0.44 & 1.5 & 0.30 & 2.12 & 6500 & 0.1 \\
\end{array}
\]

Table 3. MBW parameter sets for steels DP600 and DP1000.

For steel DP1000, the material parameter set has been calibrated based on the suggested strategy. Since the MBW model parameter sets are non-unique, one can only argue that with the selected parameters, for a variety of different samples representing significantly different stress states, a sufficient agreement between experiment and simulation can be achieved. This is also the intention of figure 2, showing the force-elongation plots of notched dog bone samples, plane strain samples, and central hole samples for experiment and simulation. The notched dog bone samples were manufactured from \( A_{s0} \) tensile samples. The notch radius is 2 mm, and the width of the sample is reduced from 20 mm to 4.5 mm in the notch ground. The central hole sample is manufactured from a rectangular geometry with a width of 30 mm, but in the sample’s central position a circular hole was introduced with a diameter of 7 mm. The plane strain samples are flat grooved rectangular plates with a width of 60 mm. They contain in-plane notches from both surfaces with a radius of 10 mm. These
notches reduce the sheet thickness to 1.0 mm in the notch ground. In all the simulations, C3D8R elements were utilized, and the mesh size was kept constant at an element edge length of 0.1 mm in those regions of the samples that show a high degree of plastic straining. The material parameter sets for both steels DP600 and DP1000 are summarized in table 3.

Figure 3 gives the equivalent plastic strain at ductile damage initiation for both steels DP600 and DP1000. These characteristic strains were identified according to a procedure that evaluates the disagreement between force-elongation curves from experiment and simulation with a simple plasticity model. The first significant overestimation of the forces applied in the experiments is assumed to result from the non-consideration of damage-induced softening, so that the instant of damage initiation can be determined. Afterwards, the numerical simulations are evaluated for the positions where ductile damage is assumed to be triggered first, so that the local conditions provoking ductile damage initiation are identified. Since these strains are dependent on the state of stress, the presented 3D illustration is usually referred to as the ductile damage initiation locus (DIL).

Figure 3. Ductile damage initiation locus of steels DP600 and DP1000.

The figure reveals that even though there are significant differences in strength properties of both investigated materials, there are only negligible differences in the equivalent plastic strains to ductile damage initiation. In order to argue that this observation is important to understand the different damage evolution properties of the steels under investigation, we assume the specific case of uniaxial tensile loading condition. It is characterized by a stress triaxiality of 1/3 and a normalized Lode angle parameter of 1. From the tensile tests, we know that the uniform elongation is 16.1 % for steel DP600, but only 8.1 % for steel DP1000. In addition, the DIL shows that in steel DP600, the damage initiation strain for the uniaxial condition is 1.76 x A_u, but in steel DP1000 it is even 3.21 x A_u. This proves that a significant amount of post-uniform elongation is required to provoke damage initiation in steel DP1000 for the uniaxial state of stress. It is thus an indicator for the relatively late damage initiation in steel DP1000.

Furthermore, with the set of MBW material parameters, also the instant of material point fracture can be assessed with respect to the corresponding equivalent plastic strain. Noteworthy, the term “fracture” indicates the instant where D_cr is reached for the first time in the simulations. Here, a contrary observation is made, since the fracture strain is significantly larger than the damage initiation strain in steel DP600, whereas the fracture event seems to follow the damage initiation event immediately in steel DP1000. It can thus be concluded that in steel DP600, damage initiation happens relatively early, but the subsequent damage evolution is slow, whereas the opposite behaviour with late damage initiation and rapid subsequent damage evolution characterizes the behaviour of steel DP1000. These findings are also visualized in figure 4. It gives the uniaxial flow curves for both materials according to the extrapolation approach presented together with the tensile test results. In both curves, the uniform elongation and the ductile damage initiation event are highlighted by dots.
Based on the presented findings, we suggest to evaluate the ratio of fracture strain over damage initiation strain to characterize the ductile damage evolution behaviour of steels. With a fully calibrated set of MBW material parameters, this ratio can be determined for any given state of stress without the need to perform additional experimental investigations. The procedure is as follows:

- Determine the equivalent strain to ductile damage initiation \( \varepsilon^d \) by solving the DIL function for the selected state of stress (eq. 4).
- Calculate the equivalent stress \( \sigma_{\nu,i} \) at the previously identified equivalent plastic strain at ductile damage initiation from the material’s flow curve (eq. 1).
- Derive the fracture strain \( \varepsilon^f \) from the damage evolution law (eq. 5). Obviously, the MBW model assumes a linear relation between the damage variable \( D \) and the equivalent plastic strain once the damage initiation criterion is fulfilled. The zero of this damage evolution function is the strain to ductile damage initiation, whereas its slope is defined by the ratio of \( \sigma_{\nu,i} \) over \( G_f \). Consequently, the fracture strain that corresponds to the critical damage \( D_{cr} \) can be easily computed:

\[
\varepsilon^f = \varepsilon^d + D_{cr} \frac{\sigma_{\nu,i}}{G_f}.
\]  

Figure 5 gives ratios of fracture strain over damage initiation strain for both steels for different selected stress states, namely uniaxial tension (stress triaxiality 1/3, normalized Lode angle 1), plane strain tension (stress triaxiality \( \sqrt{3}/2 \), normalized Lode angle 0), and biaxial tension (stress triaxiality 2/3, normalized Lode angle -1). All the presented values were computed according to the proposed procedure. Obviously, the huge damage tolerance of steel DP600 can be most easily seen for the plane strain tension condition.

5. Conclusions
The present study reveals remarkable differences between the two steels DP600 and DP1000:

- It has been shown that the differences between the two steels do not only apply on the macroscopic mechanical properties, but also on the ductile damage evolution. Steel DP600 can be categorized as a material with relatively early damage initiation, but slow subsequent damage evolution, resulting in high ratios between fracture strain and damage initiation strain. In contrast, steel DP1000 can be categorized as a material with relatively high resistance against damage initiation, but comparably rapid ductile damage evolution.
• The ratio between fracture strain and damage initiation strain is suggested as a suitable measure of the damage tolerance of sheet materials. However, the application of such parameter could result in a component design situation where a certain amount of damage is present after a metal forming operation. Accepting this status is however connected to the knowledge that the fracture strain has not yet been in reach. The term “damage tolerance” should therefore not be confused with the damage-free condition of a cold-formed part.

• With respect to the underlying mechanisms of ductile damage, we argue that with increasing mismatch of the strength properties of the martensitic over the ferritic phase, and with increasing average ferrite grain diameter, the heterogeneity of plastic strain distribution in the DP steel’s microstructure is increased. However, with increasing heterogeneity of plastic strain distribution, peak strains can result in ductile damage initiation, since this event is controlled by plastic strain. This chain of arguments explains the relatively early damage initiation in steel DP600 compared to steel DP1000.

• Furthermore, we argue that for a softer ferritic phase, a higher ductility can be expected. Therefore, despite the fact that damage has been initiated early, rather high values of equivalent plastic strain can be achieved in steel DP600, so that the ductile fracture event is significantly delayed compared to steel DP1000.

• Finally, the density of void nucleating microstructural constituents in steel DP1000 is relatively high, because ductile damage is strongly promoted by phase boundaries between ferrite and martensite. This implies that once a ductile damage initiation criterion has been fulfilled, rapid damage evolution can be expected due to the high spatial density of pores and cavities.

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