Mechanisms of void formation during tensile test in a commercial, dual phase steel

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Abstract:
A detailed analysis of microstructure and failure mechanism of a dual phase steel material as a function of strain was conducted. Accordingly, three tensile tests were performed and interrupted at different strain levels in order to investigate the void nucleation, void growth and void coalescence. SEM analysis revealed that voids nucleation occurs by ferrite grain-boundary decohesion in the neighborhood of martensite grains. Further, void initiation could be observed between closely sided martensite grains. Martensite morphology and distribution has a significant impact on the accumulation of damage. The mechanism of failure was found to be influenced by deformation localization due to microstructural inhomogeneity. Based on the experimental observations and simulation results, A model describing the failure mechanism is proposed for dual phase steel material.

Keywords: Dual Phase (DP) Steel, Failure Mechanism, Tensile Test, Void initiation, Martensite shape effect.

1. Introduction

Motivation toward DP steel
In modern transportation engineering, the application of lightweight components is a central challenge. Due to economical and ecological reasons as well as improvement of product properties, a mass reduction is desired. This involves approaches from different engineering disciplines. Thus, lightweight construction can be considered as an integrative construction
technique using all available means from the field of design, material science, and manufacturing in a combined way to reduce the mass of a whole structure and its single elements while at the same time, the functional quality is increased [1].

Evaluating the material strength without decreasing the fracture strain is a design goal in modifying the microstructure of material. Considering steel materials, the class of dual phase steels is very interesting for light weight constructions because it combines a high ultimate strength with a high fracture strain. On the market, DP 800 steels with an ultimate strength of 800 MPa and a nominal fracture strain of approximately 20% are available. Other advantages of this material are its low yield strength, high hardening ratio and absence of discontinuous yielding. Therefore, dual phase steel sheets are well suited for forming and deep drawing processes.

The aforementioned utile material properties of dual phase steels are based on the typical microstructure. It consists of a ferritic matrix with a second phase of martensite which is arranged in between the ferrite grains. There may also exist a small amount of bainite and retained austenite in the microstructure [2]. Depending on the production process, the volume fraction, size and shape of the martensite phase can vary, but the basic principle that a hard and more brittle martensite phase is arranged within a soft and more ductile ferrite phase, is typical for dual phase steels.

In literature two main approaches can be identified to account for the damage behavior of dual phase steels. The first one [3-8] considers the classical ductile damage mechanism of failure (void initiation, void growth and void coalescence) while the second one discusses local shear banding and localization on micro scale as dominant failure mechanism [9-17].

**Void nucleation due to ductile damage**

In many studies the ductile failure model is used to explain the experimentally observed damage behavior of dual phase steels. Here, especially the void initiation mechanism is a
matter of interest. Depending on the particular dual phase steel material being analyzed, different mechanisms for void initiation were observed:

Some researchers consider the martensite grains as the site of the void initiation. The brittleness of the martensite phase in the microstructure of dual phase steels is likely to promote damage. Many investigators [18-26] have observed that void formation arises from both martensite particle fracture and interface decohesion. Kang [27] studied the fracture behavior of intercritically treated structure in medium carbon steels and observed that the ferrite–martensite interface decohesion was the predominant mode of void nucleation and growth, where martensite structure was the lath type. Others [28-32] have reported that void formation occurs only due to martensite–ferrite interface decohesion. Szewczyk[31] have not observed any particle cracking for $V_m$ in the range 15–20%.

Ahmed [33] reported three modes of void nucleation, namely martensite cracking, ferrite–martensite interface decohesion, and decohesion at the ferrite-ferrite interfaces with minimum plastic deformation, which has been uniquely identified by them. They reported that at low to intermediate $V_m$ the void formation was due to ferrite–martensite interface decohesion, while the other two mechanisms also occurred at high $V_m$ (above 32%).

**Void formation due to localized deformation**

The local deformation field and its effect on the failure pattern was the subject of some other works [9-17]. Even though dual phase steels exhibits macroscopic, uniform, and homogenous deformation mode, from a micromechanical perspective, its plastic deformation is inherently inhomogeneous due to the nature of its grain level inhomogeneity. Shen et al. [17] used a scanning electron microscope (SEM) equipped with a tensile straining stage to illustrate the inhomogeneous strain distributions between the ferrite and martensite grains in DP steels. They observed that, in general, the ferrite phase was deformed immediately and at a much higher rate than the delayed deformation of the martensite phase. For DP steels with low martensite volume fraction, only the ferrite deforms and no measurable strain occurs in the
martensite particles. For dual phase steels with high martensite volume fraction, shearing of
the ferrite-martensite interface occurs extending the deformation into the martensite islands.
In situ SEM test was also carried out recently to observe the deformation field in dual phase
steels and the same result was obtained [15-16].
Tomota et. al. [36] reports pictures of deformation field in different DP steels. He reports that
dergree of plastic deformation inhomogeneity is extremely influenced by three factors: volume
fraction of martensite phase, the yield stress ratio of ferrite-martensite phases and the shape of
martensite phase.

Present work
The present work aims looking closely on the process of failure in this specific dual phase
steel and observing the process of failure in different loading stages under special
consideration of the mechanisms for void nucleation, void growth and void coalescence.
Therefore tensile testing of specimen was interrupted at specific strains: 1) when diffuse
necking happens, 2) After diffuse necking and before failure in a region which is predicted to
be localized necking and 3) after failure. It is believed that the main area of focus is the region
of necking and the process of void nucleation and growth away from this region do not give
exact information about failure mechanism. The part of the specimen was observed by SEM
and light microscopy and the results were reported. It is emphasized that for materials as
complex as the dual phase steels the interpretation of observations and discussion about the
underlying mechanisms may suffer from lack of information on local properties which may
vary by production process and/or chemical composition. Given these limitations, this work
attempts to provide an observation of the failure behavior in the present commercial DP800
steel. It is noteworthy that most previous observations were carried out on laboratory
produced dual phase steel and present work may also have some novelty in this sense. The
discussions were on the basis of present observations but might be applicable in interpretation
of failure mechanism in a larger domain of multiphase materials.
The matter of statistical representativity was considered more precisely in this study. Most of the studies on microstructural behavior suffer from a good statistical representation. This is due to the fact that they usually extract their results from a small aggregate, e.g. in situ SEM test, and then extend their anticipation for the whole structure [15-16].

In this research, in situ SEM test was not carried out because the area of observation would be very small and thus could not be considered as the representative of the whole aggregate. Statistical representativity in microstructural level was considered in the following way. The specimens were tested and then investigated thoroughly in different locations after each interruption. Around 80 SEM pictures were taken from different locations of specimens and thus the reported results could expect to reveal the general manner of the whole microstructure.
2. Experimental procedure

A commercial high-strength dual phase DP800 steel was studied in the present work. The galvannealed steel was received in the form of 1.75mm thick sheets. The galvannealing procedure has been widely used in the steel industry to promote inter-diffusion of zinc and iron, leading to an alloyed coating of better quality [37]. The chemical composition of this steel is shown in Table 1.

The tensile specimens were machined according to the dimensions in Figure 1a in such a way that the applied tensile loading axis corresponded to the rolling direction (RD) of the sheet [2]. The nominal dimension of the gauge cross section was 2 x 1.75 mm². Tensile testing was performed at a crosshead speed of 0.615μm/s on a servo-hydraulic standard testing machine.

To analyze the process of void nucleation and growth exactly, three specimens from the same part of sheet were prepared and two tensile tests were stopped in different strains (ε = 0.25 and ε = 0.31) after necking. Light microscopic pictures of the specimen right after testing are shown in Figure 1b. From the measured width and thickness of the necked region after the tensile tests the local strains could be approximated. The results are given in Table 2.

Metallographic analysis of damage accumulated along the gauge length after uniaxial tensile testing was carried out on deformed and failed samples, and cross-sections along the tensile axis. Tensile specimens were sectioned through-thickness along the mid-width in longitudinal direction. To preserve any damage during specimen preparation, wire electrical discharge machining (WEDM) was used for the cutting process. The SEM analysis of void nucleation mechanisms were carried out on the same samples. In addition, pictures of light microscopy were taken from specimens to clarify the microstructure.
3. Results

3.1 Tensile data

Typical engineering and true uniaxial tensile stress–strain curves of the analyzed DP800 steel are given in Figure 2. The load was applied in rolling direction. This steel is characterized by very uniform plastic flow until necking. Table 2 summarizes the tensile test data for the steel in terms of ultimate tensile strength (UTS), yield strength ($\sigma_y$), uniform strain ($\varepsilon_u$) and strain at fracture ($\varepsilon_f$).

3.2 Steel Microstructure

Figure 3 shows the microstructure of the analyzed DP800 material, which was provided by the steel company. The micrograph shows a through-thickness cross-section of the central part of the sheet comprising a ferrite matrix and martensite second phase. It can be seen that the spatial distribution of martensite is not uniform: the microstructure exhibits martensite banding in lines parallel to the rolling plane. These lines originate from the former pearlite bands of the cold rolled sheet. The volume fraction of the martensite phase was found to be 23%. An average grain size of approximately $6\mu$m was obtained for the ferrite phase.

3.3 Observation of microstructure in tensile specimens

As explained in Section 2, the tests were carried out on three specimens. For the first specimen the tensile test was interrupted after diffuse necking (Figure 4a). The second one was interrupted right before failure (Figure 5a), and finally the third specimen was tested till failure (Figure 6a). Light microscopy and SEM analyses were carried out for all three specimens.

Fig. 4 gives the details of microstructure and void formation of the specimen which was tested up to diffuse necking at first. While the macroscopic shape can be observed in Fig 4a, pictures of higher magnification from the center area of the specimen can be observed in Figures 4b-4d. As expected, the void nucleation in the boundary region is much smaller than in the centre...
region. This inhomogeneous distribution of void formation in the specimen has also been reported by [19]. It is noteworthy that the pattern of voids in fig 4b is not random. One can observe an aligned pattern for the voids, which are usually located between two martensite particles.

Figure 4c and 4d show some detailed inspection of voids. Figure 4c shows that some of the voids are nucleated in the boundary of two ferrite grains. This type of void initiation seems to always take place in the direct neighborhood of a martensite particle. Thus, stress concentration or deformation mismatch might be a reason. Another type of void formation mechanism can be observed in Figure 4d. Here, a separation of ferrite/martensite grain boundary takes place. While the first type leads to more elongated voids, the second one leads to more spherical voids.

The second test was interrupted after diffuse necking and right before failure. Figure 5 gives the details of microstructure and void formation. Once again, the shape of the deformed part can be observed in Figure 5a while the microstructure in the central region can be seen in Figure 5b. Figure 5c shows two voids which are created around inclusions. Only a small number of inclusions were found in the complete analysis. Therefore, it is unlikely that they play an important role for the main failure mechanism. In the observed central region (Figure 5d and 5e), most of the voids extend in tensile direction and in the boundary of ferrite grains. Voids will also grow if they are situated between two closely spaced martensite particles (Figure 5d - 5f). Fracture of martensite particles may contribute to void initiation but this could not often be observed.

The third specimen was loaded up to fracture. The pictures and pattern of voids for this specimen are shown in Figure 6. Figure 6a show the broken specimen. A typical cup and cone fracture can be observed. In Figure 6b, the separation parallel to the tensile direction is shown. As explained previously, this line of separation is mainly due to decohesion of ferrite in grain boundary. Right under the fracture surface in Figure 6d some big voids (>10μm) exist. These
voids are located at the center of specimen. The Figures 6b and 6c illustrate the two previously mentioned types of void initiation. Depending on the initiation mechanism, it can be distinguished between more elongated and spherical voids. In Figure 6d, some voids just before coalescence can be observed. Further, this Figure shows that the local void growth also depends on the shape and arrangement of the martensite particles. Finally, Figures 6e and 6f, taken with light microscopy, show the change of failure mechanism from ductile failure in center to a shear failure in the region away from the center.

This specimen was then cut in a plane of approximately 100µm away from the center where the first cut was done. SEM pictures of this section are shown in Fig 7. It can be observed in Figures 7b and 7c that the large longitudinal voids (as seen in Figure 6) in the central region do not exist in this region. It can be seen in Figure 7c and also in Figure 6b that the fracture surface contains several cup-like sections (marked with arrows) which are assumed to be cuts through dimples which are typical for ductile fracture. These cup-like shapes are connected with more or less straight lines (marked with dotted arrows) which may be separated along ferrite/ferrite grain boundaries or even broken ferrite grains.

In the outer areas of the cross section, the appearance of the fracture surface becomes shear dominant, (Figures 7d - 7f). At higher magnification, the fracture of ferrite grains can be observed frequently (Figure 7f).

3.4. Fracture morphology

The morphology of the fracture surface of DP800 is illustrated in Fig 8. A wide view of the fracture surface can be observed in Fig 8a. Figures 8c and 8d show the shape of dimples in central region in different magnifications. Except some small area in these Figures, the voids show a symmetric pattern of size and range from 1 to 4 μm, with some of them being relatively deep. These voids seem to cause failure by impinging each other when the process of growth comes to a certain stage. Between these voids, some areas can be observed with a very fine void structure or nearly no voids. These areas may be interpreted as martensite
particles which are sited directly at the fracture surface. This assumption would also be in accordance with Figure 6b or 7c where martensite particles at the surface are visible, too. The fracture surface in the outer region of the cross section is illustrated in Figures 8e and 8f. The voids in this region change to a more flat and non-symmetric pattern and loose the deep symmetric appearance. A closer look at these voids in Figure 8f shows that the width of voids ranges from 2-3 μm but they are up to 6 μm long. These voids can be due to the shear localization.

3.5. Quantitative analysis of microstructure

The behavior of microstructure during loading was measured quantitatively in some aspects. The first analysis measured the martensite deformation during the loading. The thickness of martensite in each specimen was measured through several lines perpendicular to the tensile axis. Lines were selected to be around the necking our failure areas as shown in Figure 9. The thickness of grains in the areas of interest for each specimen was measured. The frequency of grains for a specific thickness was then calculated. A histogram showing the frequency of appearance versus thickness of martensite islands for three different specimens are shown in Fig. 9. The graphs indicate that the specimens with higher deformation contain higher number of small size grains. In other words, as the specimen deforms, the number of martensite grains that have lower thickness will be increased. This is due to the fact that by increasing the load martensite grains elongate and cause the thickness of islands to be reduced.

The result for the measurement of the fraction of porosity in each specimen along the lines perpendicular to the tensile axis of specimen is shown in Fig. 10b. One can see that a remarkable increase in the local fraction of cavities when the specimens deform from necking to failure. The distribution of void volume fraction becomes very heterogeneous along the length of the sample. Fig. 10d shows the fraction of porosity along the lines parallel to tensile axis of specimen. It can be observed the void volume fraction has a logarithmic decrease to zero in a distance of 1mm away from central point.
4. Discussion:

4.1 Void nucleation, growth and coalescence dual phase steels

Considering the microstructure of dual phase steels, the failure mechanism can be influenced by three kinds of particles: (1) large (>4 μm) martensite phase, (2) medium (1 μm < <2 μm) ceramic inclusions of aluminum-oxide or manganese-sulfide and (3) small carbide particles. While carbide particles contribute to the strength of material by impeding the motion of dislocations, the ceramic inclusions exist in the material as undesirable relics from production process or from raw materials. The martensite phase is an essential component of DP steels to increase the strength of material. As already mentioned, it is assumed that the small number of inclusions does not influence the principal failure mechanism. Further, the role of the carbide particles could not be analyzed in the experimental procedure. Therefore, we mainly concentrate on the effect of the martensite phase in the failure mechanism.

Void Nucleation

Void nucleation at second phase particles can occur by particle fracture or by decohesion of the particle–matrix interface. In ferritic structural steels, void nucleation has been identified with two types of second phase particles—non-metallic inclusions [38-40] and carbides [41-42]. The strain of void nucleation and the interface strength for these two second phase particles are reported in literature [40-43]. Poruks [44] uses the same method to measure the value of nucleation strain for martensite particles. It follows from the results of these works, that the experimentally determined void nucleation strain and the calculated interface strength are increased in the following order: non-metallic inclusion, carbide particles, and martensite grains. This is consistent with what is known about the structure of the interfaces for these various dispersed phases. The low void-nucleation strains for non-metallic inclusions arise from pre-existing cracks and weakly bonded interfaces [44]. Fe₃C particles formed by solid-state transformation often have facetedted interfaces, which is indicative of semi-coherent interfaces. The high strength ferrite/martensite interface is explained by [45] in the following
way. The carbon atoms are diffused from martensite to ferrite when the martensite phase is decomposed during tempering process. This carbon diffusion reinforces the ferritic matrix near the ferrite/martensite interface and increases the interfacial strength between ferrite and martensite. It should be noted that these analyses were not carried out for the DP800 material of the present study and certainly the measured values of nucleation strain interface strength for martensite would be different if such an analysis were done because both volume fraction of martensite and the grain size are different in present study. But, it is believed that the general trend of changes will remain the same. The fracture of martensite particles as a substantial mechanism of void initiation could not be observed in the present analysis.

It has to be mentioned that the estimation of ferrite/martensite grain boundary strength is carried out on the basis of void volume fraction in some works [44]. In this method the strength would be an average of ferrite/ferrite and ferrite/martensite boundary strength. This source of error has to be considered in the interpretation of results.

**Void Growth and coalescence**

According to the present observations, the main part of voids which grows around or in the direct neighborhood of martensite particles is caused by delamination of ferrite grains in their boundary (Figures 5d and 5e). This can be observed more vividly in the failed specimen, where the delaminated ferrite grains can be observed in the center of specimen (Figures 6b and 6c). This may be interpreted by the model of Zok and Embury [46]. They consider a higher density of carbide particles in the grain boundaries than the grain interior. Therefore, during the loading the voids initiate in the grain boundaries make planar arrays of voids which permit delamination to occur. In [46], it is mentioned that in steels with more uniform distribution of carbides, delamination is not observed, whereas it is frequently observed in material containing sheets of particles at which damage can initiate.

In dual phase steels, the strength mismatch between ferrite and martensite causes local stresses perpendicular to the loading direction. Therefore, the delamination of ferrite grains
will occur even at early stages of deformation (Figure 4c). An exact inspection of the Figures show that the delamination of ferrite grains mainly occurs in the regions where two ferrite grains have a long contact surface with the martensite. This can be due to the fact that while ferrite is deformed to a large amount, the martensite phase remains undeformed and due to the deformation mismatch, the two ferrite grains delaminate in the neighborhood of martensite. Considering the model of Zok and Embury [46], delamination of ferrite/martensite interfaces might be conceivable. However, from previous considerations it should be noted, that ferrite/martensite grain boundary is assumed to be stronger than ferrite/ferrite grain boundary. In this model, high density of carbide particles is the cause of separation or delamination in grain boundary. The carbide particles travel from the ferrite grains to the grain boundary and it is well known that martensite is free from carbide particles. Then it can be concluded that the density of carbide particles in ferrite/ferrite grain boundary is approximately twice the density in ferrite/martensite grain boundary. As a result, the grain boundary of ferrite/martensite is stronger than grain boundary of ferrite/ferrite. Direct observation of the carbide particles in the specimen would have required TEM testing which was not carried out in this study.

Considering the dual phase steel microstructure, the influence of the martensite phase on the void growth and void coalescence mechanism has to be clarified. Especially important for understanding of the failure mechanism is, how the voids are linked to each other and what the dominant kind of fracture mechanism is. The questions will be addressed to some extend in the next session.

4.2 Explanation of failure mechanism in dual phase steel through simulation and Experimental analysis

In order to explain the failure mechanisms in dual phase steels and draw conclusions, micromechanical simulation was carried out. To generate the model of the DP 800 microstructure shown in Fig. 11(a), the image is first automatically segmented into two
different phases using photo-processing software Corel Draw. The segmentation is done by adjusting contrast and colors such that all martensite grains end up black while all ferrite grains end up white. It can be observed in Fig. 11(b). In Fig 11(c), the ferrite matrix is divided into 250 grains of different orientations. The orientation of grains is assigned in a way that random texture would be considered for the material. It has to be mentioned that for assigning orientation to the grains the meshing of microstructure was carried out simultaneously with the division of ferrite matrix into separate grains. On the other hand, comparing Fig 11(a) and 9(c) shows that the exact grain shape of ferrite matrix was not considered in this study but it was divided in a way that the approximate grain size in model and real microstructure would be the same. The idea behind modeling of material was that the inhomogeneity due to microstructure and grain inhomogeneity could be modeled exactly.

Fig 12 shows the simulation results. While Fig 12a shows the deformed shape, Fig 12b shows the mises stress and Fig 12c shows the shear strain. It can be observed that the points of localization for mises stress and shear strain are the same. The sharp ends of martensite grains and more severely between two sharp ends of martensite grains are points of stress/shear strain localization in the simulation. These points are shown by circles in Fig 12c and also Fig 12d. Fig 12d shows the internal pressure in the microstructure. It can be observed that at the points of stress/shear strain localization (shown by circles in Fig 12c) the internal pressure is negative. Negative internal pressure in tensile loading is an indicator of void initiation in microstructure. The SEM pictures from microstructure of tensile specimen in Fig. 13 also show the same system.

The presented model tries to predict the damage nucleation in an undamaged microstructure of ferrite and martensite. A literature survey reveals that for such a complex microstructure (with different phases and complicated morphology), use of more sophisticated continuum based models may not give us a reasonable and accurate results while it is a difficult task too.
Continuum fracture mechanics has provided a wealth of methodologies for modeling the evolution of damage, but these methods all depend on knowing where the damage nucleated; hence a pre-existing void or crack is normally introduced arbitrarily. The process by which undamaged material develops damage (here defined as the generation of a new free surface where there was none before) is not very well understood.

Nucleation and growth of microcracks also depends strongly on microstructure evolution during prior forming history. Thus, a paradigm is needed to understand how the process of plastic deformation interacting with microstructural features leads to the development of subcritical cracks or voids.

Therefore, the presented model is practical and not very complicated; however, it can properly anticipate the mechanism of void and crack nucleation in a real microstructure.

To sum up the discussion, we propose a model for void initiation mechanism of the present dual phase steel considering the simulation results and experimental data. This model is able to explain the main features of the observed material behavior at very early steps of deformation and even before necking. Some voids initiate around martensite particles. It was discussed earlier that the distribution of martensite has an important effect on the initiation of voids and it is assumed that most of these voids are due to the following two mechanisms:

- **Type 1**: Delamination of ferrite-ferrite grain boundaries in the direct neighborhood of martensite particles initiated by local stress concentrations due to strain incompatibilities. This mechanism takes place in regions with long ferrite-ferrite grain boundaries. The initial shape of these voids is more elongated. They are shown by green color in Fig. 14.

- **Type 2**: Delamination of ferrite-martensite interfaces forming spherical voids. This type is preferred when martensite particles are sited closely together with only small ferrite grains in between. Due to the spherical shape of the voids, it is assumed that a
locally high hydrostatic stress between the martensite particles triggers the void initiation. They are shown by red color in Fig. 14.

Since the shape of the specimen’s cross sections is nearly quadratic (1.75 mm x 2.00 mm), a typical cup and cone fracture surface is obtained (Figure 6a). For this reason, the center zone and the boundary zone of the specimen must be treated separately for further discussion.

In the center zone, where a more hydrostatic stress state dominates, the voids will grow and form preferably spherical voids. Depending on the local conditions, the voids initiated at a ferrite-ferrite grain boundary may also grow further along the ferrite-ferrite grains boundaries. The direction of growth is also influenced by the “applied stress state” and the local shape and also by the distribution of the martensite grains. The voids will grow more during deformation but the growth is - in a nonlocal view - parallel to the loading axis and usually does extend normal to the tensile direction. In addition, a weak strain localization field will be generated in ferrite matrix due to the strength difference between ferrite and martensite. This corresponds to the localized field caused by the internal stress state in material. Some researchers [47] believe that this internally induced localized field (which will become stronger at larger strains) causes a more uniform distribution of strain in DP steels and it is the main reason for higher formability of these materials compared to HSLA steels.

The internally induced localized field and externally applied deformation field interact with each other to trigger localized necking of material. In this step, shear bands in the ferrite phase will be formed between the martensite and lead to final failure. What can not be proved with the available results is that voids initiated by carbide particles (in the ferrite phase) will grow and join in this shear band. In combination with the classical void coalescence mechanism, one can explain the shape of the fracture surface shown in Figures 7b and 7c as follows: The cup like shape can be interpreted as cuts through grown voids. In between, linear sections that start and end at martensite particles are visible. These sections may be interpreted as a result of the described localized necking between neighbored martensite particles.
The boundary zone is dominated by shear stresses. As reported in the previous section, the number and size of voids are significantly smaller than those of the central zone. In this part of the specimen, the classical void sheeting mechanism is of relevance for failure. Considering the present microstructure, it is assumed, that local shear bands in the ferrite grains will occur, while these shear bands will start and end at neighboring martensite grains. Figure 7f shows that at the fracture surface martensite particles are connected with straight lines. Several ferrite grains sited between the martensite are broken. A possible explanation might be the previously discussed the void sheeting mechanism.

5. Conclusion

The reported results as well as the proposed failure model are in accordance with the results reported in literature. Some aspects already discussed in existing publications could also be found in the present analysis. On the other hand, not all aspects discussed in literature could be confirmed. One example is the relevance of breaking martensite particles for void initiation that seems to be negligible in the present study. Furthermore, the discussion of “Type 1” mechanism for void initiation was not found in the literature but seems to be important for the present material.

Therefore, it must be noted that parameters like the volume fraction of the martensite phase, the chemical composition, the yield stress ratio of ferrite and martensite as well as the size, shape and distribution of the martensite particles play an important role in dominant failure mechanism. Further, the specimen’s shape influencing stress state in the gauge should also be considered in the discussions.

To conclude, a formulation of a general failure mechanism for dual phase steels seems not to be applicable. It is suggested that further works may concentrate on the description of the leading failure mechanism under consideration of the existing relevant parameter of the dual phase steel material.

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References:


Fig 1: (a) Test specimen.

Fig. 1b: Light microscopy pictures of the three specimens after testing. From left to right: After necking, before failure and after failure. The first row shows the top view, the second row the corresponding side view.

Fig 2: Engineering and true stress strain curve of DP800
Figure 3. Microstructure of the DP800 steel in two different magnifications.

Figure 4: Void pattern after diffuse necking.
Fig 5: Void pattern before material failure
Fig 6. Void pattern and microstructure of material in mid-plane of specimen after failure
Fig 7. Void pattern and microstructure of material away from the mid-plane of specimen after failure
Figure 8: Fracture morphology
Figure 9. Histogram showing the thickness of martensite islands for specimens of diffused necking (specimens-1), localized necking (specimens-2) and failure (specimens-3)
Figure 10. (a,b) The fraction of cavities in slices perpendicular to the tensile axis for specimens of diffused necking (specimens-1), localized necking (specimens-2) and failure (specimens-3) (c,d) The fraction of cavities in slices parallel to the tensile axis for specimen of localized necking.

Fig 11: generation of micromechanical model for DP steel
Figure 12: Deformed shape of grains (a) von Mises equivalent stress, (b) accumulated shear plastic deformation, (c) and hydrostatic pressure (d).
Fig 13. Points of void initiation in SEM picture of tensile specimen.

Figure 14: Illustration of the proposed failure mechanism in the center zone and the boundary zone of the analyzed specimens.
### Table 1: Chemical composition of the investigated DP800 (in wt %)

<table>
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<tr>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Al_{tot}</th>
<th>Cr+Mo</th>
<th>Nb+Ti</th>
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<td>0.037</td>
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### Table 2: Approximated values of the local strain in the necked region of the specimen

<table>
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<th>Tested specimen</th>
<th>initial width [µm]</th>
<th>initial thickness [µm]</th>
<th>final width [µm]</th>
<th>final thickness [µm]</th>
<th>logarithmic strain [-]</th>
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</thead>
<tbody>
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<td>Case 1: after necking</td>
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<td>1747</td>
<td>1860</td>
<td>1632</td>
<td>0.77</td>
</tr>
<tr>
<td>Case 2: before fracture</td>
<td>2023</td>
<td>1755</td>
<td>1780</td>
<td>1456</td>
<td>0.86</td>
</tr>
<tr>
<td>Case 3: after fracture</td>
<td>2022</td>
<td>1745</td>
<td>1636</td>
<td>1328</td>
<td>0.97</td>
</tr>
</tbody>
</table>

### Table 3: Tensile data for DP800

<table>
<thead>
<tr>
<th></th>
<th>UTS_{av} [MPa]</th>
<th>σ_y [Mpa]</th>
<th>uniform strain [-]</th>
<th>fracture strain [-]</th>
</tr>
</thead>
<tbody>
<tr>
<td>datasheet</td>
<td>780-900</td>
<td>450-560</td>
<td>-</td>
<td>&gt; 14 %</td>
</tr>
<tr>
<td>experiment</td>
<td>795</td>
<td>480</td>
<td>14 %</td>
<td>20 %</td>
</tr>
</tbody>
</table>