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Revealing the mechanical and microstructural performance of multiphase steels during tensile, forming and flanging operations

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ABSTRACT

The mechanical performance of Dual Phase (DP) and Complex Phase (CP) steels was investigated by SEM analysis, tensile testing, Forming Limit Curve investigation and flange formability testing. The alloys of interest were Dual Phase (DP) untempered, Dual Phase (DP) tempered and Complex Phase (CP) steels. Phase content analysis showed that the distribution of the ferrite and martensite phases was the same for the two DP alloys, but the grain size and condition (tempered/untempered) for the martensite islands was much different in the two alloys. In the tempered DP steel, the smaller grain size for the martensite and the tempering process resulted in increased elongation, more formability and ability to form a flange (flangeability). In CP steels the soft ferrite phase is replaced by harder bainite, yielding a bainitic-martensitic microstructure. Bainite reduced the total elongation of the alloy during tensile testing, reduced the formability (especially under plane strain conditions) of the alloy but improved the flangeability of the alloy. Under flanging conditions, CP steels deformed to higher strains, at tighter radii with minimum springback. Microstructural inspections at the outer radius of the flanged specimens revealed that in CP steels bainite deforms similarly to martensite, therefore the strain partitioning is smaller in CP steels in comparison to DP steels. Plastic deformation in CP steels upon flanging occurs with the formation of strong slip bands in both martensite and bainite. In contrast, the martensite and ferrite grains in DP steels deform quite differently leading to strong strain localisations. Void nucleation and cracking occurred at the martensite islands or within the soft ferrite phase next to the martensite islands. In CP steels no voids or damage was observed within the matrix. A special case study was done with a thicker and stronger alloy, a Martensitic 1400 steel to reveal the flangeability limits for advanced high strength steels. Neither cracks nor damage were observed visually on the flanged specimens. However SEM observations at the outer radius of the flanged samples revealed significant void growth at inclusion sites and cracks nucleating within the matrix adjacent to the inclusions.

1. Introduction

Ultra-High Strength Steels (UHSSs) have emerged in the automotive market due to their superior strength and performance in crash in comparison to low strength steels and other metallic alloys [1–5]. Utilising higher strength alloys allows the design of structural automotive components with decreased thickness, reducing the overall vehicle weight [3–6]. A combination of different phases, such as retained austenite, martensite, bainite and ferrite are developed in these steels giving structural components a broad range of strength and formability. Increasing the percentage of the soft ferrite phase increases the deformability but on the other hand reduces the strength of the alloy with the same chemistry (but different heat treatment). In contrast increasing the percentage of martensite or bainite increases the strength of the alloy but decreases the deformation capabilities of the material [3–5]. Retaining austenite at room temperature can increase both the strength and total elongation of the alloy in Transformation-Induced Plasticity (TRIP) steels. Yet TRIP steels have been found to have weldability issues which restricted their commercial use for car-body chassis [3].

A series of studies exist, where the formability of various engineering alloys is evaluated both experimentally and numerically. The final failure in formed parts is estimated by predicting void nucleation and growth, yield surface curvature and material rate sensitivity in the following studies [7–10]. These studies were limited to isotropic materials [7–10]. However, sheet metals display highly anisotropic behaviour that is introduced when a cast billet is hot or cold rolled into sheet material. Anisotropy is frequently modelled by adding a 'correction' to

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existing isotropic models. Liao et. al in his study introduced an anisotropic criterion for the failure prediction of sheet metals under plane stress conditions, based on a modified Gurson yield criterion [8].

Similarly Chien et. al incorporated three additional fitting parameters to the modified Gurson anisotropic yield criterion to account for both the presence of anisotropy and the influence of voids [11]. Chien et al. in a latter study performed a thorough investigation on the failure of Aluminium sheets in a Limiting Dome Height (LDH) test and attributed final failure to the effect of shear localisation; both in cases where the sample showed significant or negligible amount of necking [12]. Chien et al. considered geometric imperfections within the material (such as pores, inclusions, intermetallics etc.) to be responsible for the onset of necking in sheet metals under plane stress conditions. While it was stated that it is material imperfections that result in the shear localisation or the final failure of sheet metals under generalized plane strain conditions. - However it is known that it is not material imperfections but crystal structure, crystallographic orientation and texture that governs the formation of shear bands [13]. - In their FE model, the necking component was excluded from the analysis and only the shear band localisation mode was taken into account. A plane strain formulation was followed where the shear bands form in the through thickness direction. The failure strains on the stretching sides of the bended sheets were estimated by using the failure strains for the LDH samples.

A series of mechanical tests were performed by Bai et al. to obtain the deformability and fracture locus for a wide range of stress triaxialities and sample geometries [14]. The alloys investigated were an Al2024 T351 and a TRIP RA-K40/70 alloy. A modified Mohr-Coulomb (M-C) criterion was employed and showed very good agreement between the experiment and the model for a wide range of shear dominate fractures. Only for round bars under tension the correlation did not match between the experiment and the model. It should be noted though that the specific M-C criterion assumed only material isotropy, but predicted well both the location for crack initiation and the crack directions.

In another study, Hudgins et al. tried to reveal the mechanisms of failure during forming operations in Advanced High Strength Steels (AHSSs) [15]. Hudgins et. al suggested that fractures occurred due to shear dominated mechanisms and led to limited localised necking. Fracture was observed on alternating 45^0 planes, in the through-thickness direction. The strain paths and the fractures occurring during roller hemming operations in AHSSs are difficult to predict using Forming Limit Curves (FLC). The presented model incorporated two components, one to predict instability at die radii, and another to predict tensile failure in the free ligament. An approach was followed in order to predict shear-dominated instabilities at die radii, represented by maximum applied tensile force as a function of die radius. Fig. 1 shows the presence of a transition zone from die instability to tensile instability with increasing the die radius. Material tensile strength is the most important parameter at large radii in contrast to the die geometry for small radii as shown in Fig. 1. A critical radius is introduced (normalized by sheet thickness: R/t_{crit}), above which materials will fail in tension. The critical R/t value is dependent on the slope of instability curve at small radii and the material tensile strength at large radii (Fig. 1). The materials examined showed similar slopes at lower radii, and thus formability for these steels depends primarily on tensile strength.

The formability of a 1050 aluminium alloy was investigated under roller hemming operations in [16]. Muderrisoglu et al. suggested that the roller hemming deformation mechanisms are more complex with respect to simple bending and flanging tests as the stress fields become three dimensional [16]. It was concluded in this study that the flange height (length of flanged sheet in the direction perpendicular to the width of the sheet) has a greater effect with respect to the bend radius on the load necessary to form the part. Pradeau et al. in a latter study developed a model that predicts final failure of aluminium alloy sheets during hemming operations based on the Hosford-Coulomb rupture criterion but gave no explanation as to the absence of necking during such operations [17]. The model included an anisotropic yield function, a mixed (isotropic and kinematic) hardening law and a fracture criterion. The comparisons were performed for the load evolution, strain field and prediction of rupture. A very good correlation was obtained over all the tests. The prediction of the developed fracture criterion proved to be successful independent of the amount of pre-strain.

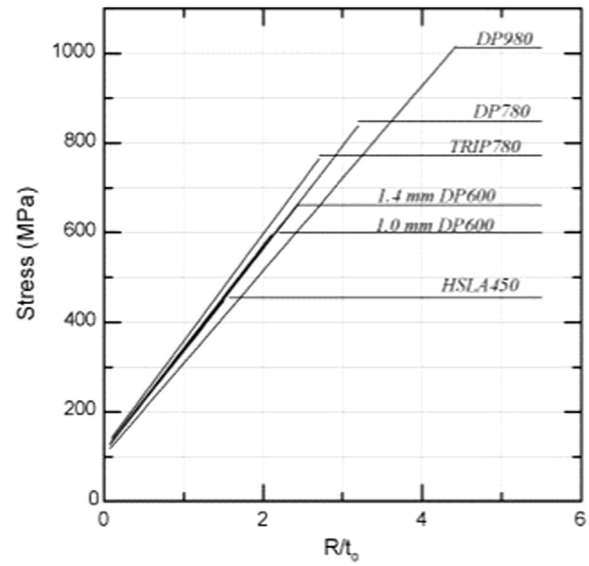

Fig. 1. Theoretical instability curves based on a strain hardening analysis for all studied materials. Similar slopes are observed in the shear fracture region at low radii, forcing the critical R/t ratio to be heavily dependent on tensile strength (after [15]).

Maout et al. used the Gurson-Tvergaard-Needleman (GTN) model to predict damage nucleation during a hemming process of a 6000 aluminium alloy [18]. Tensile testing of flat and notched samples together with biaxial testing was initially done to optimise the material parameters and allow for the consideration of the different strain paths to damage nucleation and evolution. An FE model of the hemming process was build incorporating the anisotropy of the material. The hemming limit criterion was based on the critical void volume fraction. Criteria of acceptance for the hemming-formed sheets were revealed based on visual observations of the outer radius. The decision to accept or reject components was based on crack lengths and densities.

However, Miller has pursued a thorough analysis on the behaviour of cracks and the safety of engineering components [19,20]. Miller stated that the physics of crack propagation are governed largely on its size and on the mechanical conditions applied [19]. Three different zones can be found in the lifetime of a crack in Fig. 2: the so-called microstructural short crack growth regime (A-B zone), the physically small crack growth regime where Elasto-Plastic Fracture Mechanics concepts can be applied (B-C zone) and the zone where the crack is large enough for Linear Elastic Fracture Mechanics to be applied (C-D zone). In zone A-B crack initiation and growth depends on material microstructure at the grain level. In zone A-B, precipitates or inclusions act as stress concentration points and therefore facilitate crack initiation. The A, B and C barriers are of increasing strength opposing early crack growth. These three barriers can be grain boundaries for example. The microstructure effects have an important role also in the physically short crack region. Yet Fig. 2 shows that the effect of the grains and grain boundaries becomes smaller and smaller as the crack grows. The important zone in Fig. 2 is the microstructurally short crack growth regime where crack growth speeds are very low, governing the overall fatigue life. Fig. 2 also shows why surface finish is important: if surface scratches (or in the case of a formed part surface ripples) are deeper than C, then the structural integrity of the engineering component is

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