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Deformation behaviour of lath martensite in multi-phase steels

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ABSTRACT

This paper analyses the role of interlath retained austenite films within the martensite islands on the local strain partitioning and overall mechanical response of dual phase steels. A crystal plasticity-based modelling approach is thereby adopted. It is shown that the presence of a negligible (<2%) volume fraction of interlath austenite amplifies the strain heterogeneity in both the martensite islands and the surrounding ferrite matrix. This qualitatively matches experimental observations from literature, showing a highly anisotropic, orientation dependent local mechanical response of martensite. The results emphasise the potential key role played by interlath austenite films in microstructures involving lath martensite.

1. Introduction

One of the most widely used microstructures in low-alloyed high strength steels is lath martensite. This polycrystalline aggregate, which typically appears when the carbon content does not exceed 0.6 wt%, is characterised by subgrains (blocks and crystallographic packets [1,2]) that are in general alternate layers of two phases: martensite (the laths) and very thin films of austenite, which can be retained at lath boundaries. TEM observations of this microstructure have been reported not only for low carbon martensitic steels [4,3,6,5], but also for martensite islands in dual phase steels [8,9,7,10] as well as low alloyed TRIP steels [11,12]. Recent experimental–numerical investigations [13,14] have shown that, despite the small dimensions and volume fraction, the austenite films may play a crucial role in the deformation and (possibly) fracture behaviour of lath martensite. Indeed, these austenite layers enhance the ductility of the lath martensite microstructure [13] and can play a key role in the orientation-dependent deformation response [14].

Similar microstructures based on alternate, sub-micrometre sized layers of hard martensite and softer austenite have been processed within the past 10 years, with the purpose of improving material ductility and preserving a high strength. Among those, there are the quenched and partitioned (Q&P) steels [15], TRIP maraging steels [16] and new generations of maraging steels with reverted austenite films [17]. Furthermore, the recently developed nanobainitic microstructures [18] also reveal alternate layers of ferritic bainite and austenite films.

For all these cases, an orientation relationship exists between the BCC martensitic/bainitic phase and the FCC austenite films, which ensures that the interface between the two phases is approximately parallel to three slip systems of the $\{1\,1\,1\}_\gamma$ family. It has recently been shown [13] that this crystalline configuration makes shearing along the bicrystal interface a preferential loading condition, since the slip systems in the austenite are in the most favourable orientation to carry plastic deformation. Therefore, the austenite phase acts as a “greasy” layer on which the stiff, poorly deforming martensite laths slide. This is consistent with experimental observations [19,20], which show that slip parallel to the lath habit plane is preferential in lath martensite. Moreover, in [14] it has been shown that if the presence of the austenite films is neglected, the measured stress–strain response of martensitic samples [19] cannot be reproduced by means of a conventional crystal plasticity model.

This paper investigates the impact of the martensite substructure and interlath austenite layers on the local strain partitioning and the global response of dual phase ferritic-martensitic microstructures. To this end, the modelling approach used in [13,14,21], based on the representation of the martensite crystallographic packet with austenite layers by a (locally) infinite laminate, is adopted. A heterogenous deformation response of the martensite–austenite aggregate is observed, matching experimental observations [22,23]. This is the result of the pronounced orientation-dependent response of the martensite, which is due

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to the presence of the thin interlath austenite films. Unless the martensite volume fraction is too low (below 10%), the orientation-dependent response of martensite can also remarkably influence the local strain of the ferrite matrix, the global strain partitioning of the dual phase ferrite-martensite microstructures and the macroscopic stress–strain response. This may have implications not only for the deformation behaviour, but also for damage and fracture in dual phase steels.

2. Modelling approach

Lath martensite is modelled as an infinite laminate of a hard phase (the laths) and a soft phase (the austenite). As shown in [13,14,21], the deformation behaviour of the aggregate is dominated by slip activity in the austenite on the 3 slip systems parallel to the interface. Therefore, to reduce computational costs, for the austenite, the reduced model proposed in [21] is used, in which only these 3 slip systems are explicitly incorporated in the crystal plasticity model, while plasticity in other directions is accounted for via an isotropic flow rule (see [21] for details). The martensite laths are modelled through isotropic (visco)-plasticity.

The fine scale mechanical response of laths with interlath austenite is upscaled to the scale of martensite islands through homogenisation, using an underlying lamella model. The lamella model represents the effective behaviour of laths with embedded interlath retained austenite.

The lamella model implicitly accounts for the orientation of the laminate's interface. The model equations comply with equilibrium and compatibility at the interface. For more details on this modelling approach and its validation see [14,21]. The parameter identification has been performed in [14,21] based on the experimental response of martensitic specimens [19]. The material parameters for the austenite layers and the martensite laths are those listed in [21], Tables 1 and 2. For the austenite, the initial slip resistance of the slip systems is 270 MPa, while the yield stress along the remaining directions is 550 MPa. For the martensite, the yield stress is 1.14 GPa. The austenite volume fraction within the martensite–austenite aggregate is taken equal to 5%, corresponding to austenite films of 5 nm thickness between martensite laths of 100 nm thickness.

Ferrite grains are modelled with a conventional crystal plasticity model [24]. The material parameters listed in [21], Table 3, are adopted. The initial slip resistance is 200 MPa.

To study the role of the martensite substructure, the results will be compared to simulations on the same microstructures, but with martensite modelled as an isotropic (visco)-plastic phase, by neglecting the presence of the thin austenite layers. The material parameters of the "isotropic" martensite are those listed in [21], Table 2, representing the isotropic laths in the lamella model.

3. Isotropic *vs* lamella model for martensite in DP steels

The dual phase microstructure shown in Fig. 1(a) is considered. The material is a DP800 grade [25]. It is modelled by assigning the measured orientations of the ferrite grains and the martensite subunits (variants). The orientation of the martensite/austenite interface has been identified after indexing the martensite variants with respect to the small retained austenite islands (Fig. 1(a)), the orientation of which is assumed to be the same as that of the prior austenite grains. These small retained austenite islands were not explicitly modelled. The main crystallographic packets (CP) are indicated in Fig. 1(a).

The microstructure is discretized using finite elements. Periodic boundary conditions are applied along the three spatial directions, therefore the geometry is assumed constant in the third dimension. Uniaxial tension along the X (horizontal) axis is prescribed. The martensite islands are modelled by either the lamella model (including austenite films) or the isotropic visco-plasticity (disregarding the presence of austenite films).

The computed stress–strain responses of the two models, Fig. 1(b), deviate only to a minor extent. Fig. 1(c) and (d) shows the comparison of the strain component ε_{xx} computed with the model with isotropic martensite (c) and the model with lamella martensite (d). The local deformation response of the ferrite is hardly affected by the enriched martensite model, presumably due to the small martensite content in the considered region (10% ca.). The local deformation within the martensite is also comparable for island B. However, differences can be noticed in island A, especially inside the areas belonging to CP4. The local strain response is more heterogeneous in island A, and local peaks show a significant increase, from about 8% to 12% strain. Furthermore, comparison between islands A and B, modelled as isotropic (Fig. 1(c)), reveals that the local strain values are comparable. The same does not hold when the lamella model for martensite is used, where the strain level clearly differs between the two islands.

The strain heterogeneity computed with the lamella model of the martensite is due to the orientation dependent response of different crystallographic packets, which is accounted for by the lamella model and is ruled out when adopting an isotropic visco-plastic model. In [14] it has been shown that the orientation dependent response is amplified by the presence of the austenite layers; when modelling the martensite variants as BCC and neglecting austenite films the orientation dependent response is noticeably weaker. Consistent with the present results, remarkable strain heterogeneities have also been measured experimentally [22,23], both within a single martensite island and among multiple martensite islands.

4. Dual phase microstructures with multiple martensite islands

Next, a larger dual phase ferrite-martensite microstructure with multiple martensite islands is modelled. The model geometry (Fig. 2(a)) is based on a DP morphology obtained with optical micrography. However, since no EBSD measurements were available for this microstructure, random orientations are assigned to the ferrite grains and prior austenite grains. In each prior austenite grain, the orientation of a single crystallographic packet is chosen to model martensite. The colour coding in Fig. 2(a) reflects the assigned orientations according to the inverse pole figure map. Periodic boundary conditions are applied along the three spatial directions, therefore the geometry is assumed to be constant along the thickness. Uniaxial loading in horizontal direction is applied.

Two simulations have been performed on this microstructure. In the first simulation, the lamella model of the martensite–austenite aggregate with 5% volume fraction of interlath austenite has been used to model the martensite islands. This corresponds to 1.7% austenite volume fraction in the whole material, which commonly is considered as negligible. In the second simulation, the martensite has been modelled as isotropic (by setting the austenite volume fraction to zero), while keeping all material parameters and the orientation of ferrite grains the same.

The comparison of the stress–strain responses of the two models, in Fig. 2(b), shows that a negligible amount of austenite yields a noticeable increase in global strain, for a given stress level. For example, at 700 MPa the computed overall strain increases more than twice, from 2.24% to 4.95%. A comparison with experiments is not trivial, since it is challenging to control the amount of the interlath austenite and to measure such small volume fractions. Such systematic study might be an interesting subject for future experimental investigation.