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Effect of microstructural evolution on deformation behaviour of pre-strained dual-phase steel

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ABSTRACT

The influence of microstructural evolution on the strain hardening and non-uniform deformation behaviour of pre-strained ferrite-martensite DP steel was studied using a crystal plasticity finite element method (CPFEM). The material parameters in the CPFEM were identified through a systematic procedure by fitting models to the experimental results of micro-tensile tests. The CPFEM with the identified parameters were applied to the analysis of virtual models with focus on the microstructural factors such as the angle of interphase boundary, the orientation of habit plane in the martensite phase, and the intensity of texture in the ultrafine-grained ferrite phase developed in the severely-deformed sample. The analysis results suggested the following influences of microstructural factors on the deformation behaviour. The formation of deformation bands in the ferrite phase depends on the configuration of the interphase boundary, the orientation of the habit plane in neighbouring martensite phase and the intensity of texture in the ferrite phase. The suppressions of the deformation bands by controlling microstructure could contribute to later onset of necking and resultant higher strain hardening.

1. Introduction

Dual-phase (DP) steel, a mixture of a soft ferrite matrix and a dispersed, hard martensite phase, exhibits high work hardenability, low yield ratio, and a good balance between strength and ductility [1–4]. These excellent properties have increased the use of DP steel within an automotive industry that requires both fuel economy and crash safety in the vehicles it produces. In recent decades, there have been several studies on the relationships between mechanical properties and microstructural factors such as the volume fractions [5–7] and grain sizes [8–11] of both phases and the distribution [12,13] and morphology [14,15] of martensite. Lai et al. reported that tensile strength increases as the martensite precipitate fraction increases [5]. A tensile test study by Calcagnotto et al. using annealed DP steels with ferrite grain sizes of 1.2–12.4 μm revealed [8] that the yield stress and ultimate tensile strength increase as the ferrite grains are refined, without affecting ductility. In addition, Park et al. reported that a DP steel with chain-like networked martensite precipitates exhibited higher work hardenability than one with isolated martensite precipitates [13]. These reports suggest that applying the mixture rule to both phases is not enough to explain the mechanical properties of DP steels. Recent review article covers progress in microstructure evolution during processing, experimental characterization of micromechanical behaviour and numerical simulation of mechanical behaviour on DP steels [16].

DP steels are inhomogenously deformed during secondary processing because of the difference between the deformabilities of the ferrite and martensite phases. In particular, their low hole expandability is often a major drawback during punching of a thin plate. A multi-scale tensile testing study by Yokoi et al. using punched DP steel sheets revealed [17] that the locally developed microstructure at the periphery of the punched hole leads to brittle fracture. A micro-tensile testing study of cold-rolled DP steel sheets revealed [18] that in an ultrafine-grained ferrite microstructure with an average grain size of ~0.7 μm, significant texture developed locally through severe deformation and that the strain concentration in this ferrite microstructure promoted shear-type fractures causing a significant loss of ductility. Although development of an ultrafine-grained microstructure can cause a strength increase, the associated loss of work hardenability, together with the influence of strong texture on the ferrite microstructure, was suggested as causes for shear-type fractures. However, the details have not been clarified and a deeper investigation of the influence of an ultrafine-grained microstructure on work hardenability and deformation behaviour is required.

The crystal plasticity finite element method (CPFEM) has recently been widely applied to analyse the plasticity of materials [19–21]. It has also been applied to the analysis of the deformation behaviour and fracture mechanisms of DP steel [22–24]. Woo et al. investigated the

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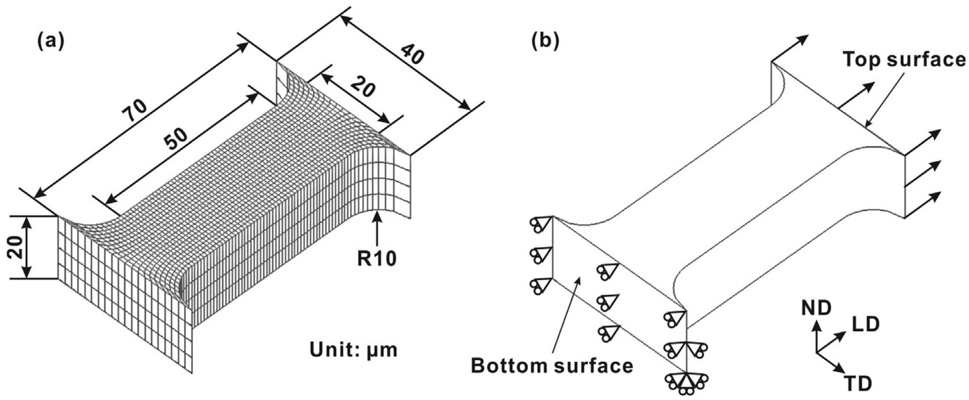

Fig. 1. (a) Finite element mesh and (b) boundary conditions for uniaxial tensile loading.

deformation behaviour of ferrite and martensite phases in DP980 steel under uniaxial tensile loading using in-situ neutron diffraction and CPFEM. This study found that the crystallographic orientation of the ferrite phase affected shear strain localisation and void formation in the ferrite region adjacent to the martensite phase [23]. Chen et al. indicated that the tensile behaviour of DP steel could be represented using the specific material parameters of each phase, by combining micro-compression testing and numerical analysis [24]. However, these studies did not discuss the influence of the microstructure, developed by the forming processes, on the deformation behaviour.

Crystal plasticity finite element analysis is an effective way to elucidate the relationship between microstructural factors and plastic behaviour of DP steels because this method can focus on arbitrary microstructural factors. In the present study, the material parameters in the constitutive model were identified based on previously reported experimental results of micro-tensile tests that used pre-strained DP steel samples with inhomogeneous microstructures [18]. The present study shows that the deformation behaviour of pre-strained DP steels can be understood by CPFEM analysis of the DP steel specimens containing a coarse-grained ferrite phase. We also performed deformation analysis of DP steel specimens containing an ultrafine-grained ferrite phase comparable to the one developed in the vicinity of the punched hole. The deformation mechanism was elucidated from the strain distributions and the slip activity.

2. Numerical procedure

2.1. Crystal plasticity finite element method

In this study, crystal plasticity finite element analysis was performed to investigate the effects of microstructure on the stress–strain behaviour, non-uniform deformation behaviour, and activities of each deformation mode. The constitutive model used in the analysis was a rate dependent large strain crystal plasticity model proposed by Peirce et al. [25] A brief summary of the present numerical method is described below:

The slip rate of the slip system i is estimated using

$$\dot{\gamma}^{(i)} = \dot{\gamma}_0 \mathrm{sgn}(\tau^{(i)}) \left| \frac{\tau^{(i)}}{g^{(i)}} \right|^{\frac{1}{m}}, \tag{1}$$

where $\dot{\gamma}_0$ and m are the reference slip rate and strain sensitivity parameter, respectively. In Eq. (1), $\tau^{(i)}$ and $g^{(i)}$ are the resolved shear stress and reference stress, respectively, for slip system i. The following evolution equation was used for $g^{(i)}$:

$$\dot{g}^{(i)} = \sum_{j=1}^{N} h|\dot{\gamma}^{(j)}|. \tag{2}$$

In Eq. (2), h is the work hardening coefficient. In this study, the following Voce-type hardening law was used:

$$h = \theta^{(i)} \exp\left(-\frac{\theta^{(i)}}{\tau_1^{(i)}}\Gamma\right). \tag{3}$$

Here, $\theta^{(i)}$ and $\tau_1^{(i)}$ are material parameters, which are identified by fitting models to experimental stress–strain curves, as discussed later. In the present study, the $\{1\bar{1}0\} <111>$ and $\{11\bar{2}\} <111>$ slip system families of a body-centred cubic structure were considered in deformation mechanisms for both ferrite and martensite. In Eq. (3), Γ is the accumulated shear strain by the activities of all slip modes. In this study, in accordance with recent experimental measurements [26], isotropic elasticity was assumed with Young's modulus, E=200 GPa, and Poisson's ratio, ν=0.3. The strain rate sensitivity parameter, m, was set to be 0.02, which is slightly different from the 0.01 to 0.05 values used in recent crystal plasticity studies on dual phase steels [22,24,27]. In the present study, focusing on plastic deformation, the effects of any damages are not considered while the deformation and fracture behaviours in DP steels could strongly depend on nano- and micro-cracking as reported in the previous researches [18,28]. To describe deformation behaviour accompanying various damages, further development of constitutive model based on systematic experimental observation is required.

2.2. Analysis models

Fig. 1 shows the analysis model used in this study. The model shape is the same as that of a micrometre-sized specimen with gauge section dimensions of ~20 μm×20 μm×50 μm. This model, including specimen shoulder parts, is divided into 5600 finite elements by 20-node solid elements, as shown in Fig. 1a. The boundary conditions for uniaxial tensile loading are applied as shown in Fig. 1b. The bottom surface is fixed in the loading direction (LD). A tensile loading is applied on the top surface by a constant increment of displacement along the LD. This displacement rate corresponds to the experimental rate of displacement, i.e. 0.1 μm s^{-1}.

Fig. 2 shows the micrometre-sized specimens, including the interphase boundary used in our previous study [18], and corresponding analysis models. CR60 and CR88 represent the 60% and 88% cold-rolled samples, respectively. The ferrite and martensite phases are arranged in series in the CR60-N specimen and parallel in the CR60-P specimen. Slip bands develop in the coarse-grained ferrite phase of CR60 (Fig. 2a and b), and an ultrafine-grained ferrite microstructure develops locally in CR88-I (Fig. 2c). The initial crystallographic orientations of each element are allocated by Euler angles obtained from electron backscatter diffraction (EBSD) analysis. For the ultrafine-grained ferrite region of the CR88-I model, the initial Euler angles