Materials Science & Engineering A 682 (2017) 126–138

Contents lists available at ScienceDirect

Materials Science & Engineering A

journal homepage: www.elsevier.com/locate/msea

Mechanism based mean-field modeling of the work-hardening behavior of dual-phase steels

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ARTICLE INFO

Keywords:
Mean-field modeling
Dual-phase steel
Dislocation-density based modeling
Long-range stresses

ABSTRACT

Low-alloyed dual-phase (DP) steels exhibit good mechanical properties due to their composite-like microstructure of strong martensitic inclusions embedded in a ductile ferritic-matrix. This work presents an efficient two-scale approach to model the work-hardening of DP steels based on physically-motivated work-hardening contributions from geometrically necessary dislocations. The resulting mean-field model of Hashin-Shtrikman type comprises a minimal amount of free parameters and incorporates two main physical aspects motivated from presented experimental data: First, the constitutive equations for ferrite incorporate the averaged microstructural morphology (obtained by electronic backscatter diffraction) in terms of grain size and martensite coverage. Second, a direct interaction between ferrite and martensite phases is achieved via kinematic-hardening which models the long-range stresses experimentally observed in tensile tests of the bulk material. The presented approach is able to both reproduce the measured DP600 tensile curve and predict the distinctly different work-hardening rates observed for high-strength DP steels.

1. Introduction

Due to their weight reduction potential and good formability, dual-phase steels (DP) are nowadays widely used for automotive applications. Crash-relevant structural vehicle-parts like the A-pillar are, for example, deep-drawn from dual-phase sheet steels. The composite like microstructure of DP steels consists of a low-carbon and ductile ferritic matrix and 10–60 vol% strong martensitic inclusions [1].

DP steels exhibit a low initial yield-strength with continuous yielding. The high initial work-hardening rate drops of at high tensile strengths. Uniform and total elongation are high compared to other steels with similar strength [2]. The good mechanical properties for cold-sheet-forming can be adjusted within a wide range despite the low amount of alloys.

With an increasing martensite volume-fraction c_M, DP steels exhibit higher tensile strengths while the formability (uniform and total elongation) is often reduced. Additionally, the microstructure of DP steel strongly influences its behavior [3]. Finely dispersed and regular martensite-islands lead to an increased work-hardening rate and result in good usage of the ferrite work-hardening capacity [4,5]. The initial dislocation-density gradient near martensite islands spreads into the initially-uninfluenced ferrite grains [6,7]. Deformation is generally higher near martensite islands during the complete loading process [8]. A higher geometrically-necessary dislocation content near ferrite-martensite grain boundaries is confirmed by, e.g., [9]. The initial yield point is defined by ferrite hardness and initial dislocation density, while the ultimate tensile strength is governed by the martensite hardness [8,10].

Due to the different constituent yield-strengths (with a factor of 2–4), stress and strain partitioning is pronounced in DP steels. The ductile ferrite-matrix carries the majority of deformation while the stronger martensite inclusions exhibit much higher average stresses. Strain partitioning is built up during the initial deformation-phase. The strain-partitioning stagnates when martensite undergoes plastic deformation [11–13]. In comparison to ferrite-ferrite orientation incompatibilities, the morphology of the microstructure dominates the strain localization mechanism [14].

DP steels exhibit a distinct Bauschinger-effect. After pre-deformation, the yield strength for reversed loading is significantly lower than at the end of the forward loading. Thus, two hardening components can

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http://dx.doi.org/10.1016/j.msea.2016.11.005
Received 1 June 2016; Received in revised form 31 October 2016; Accepted 2 November 2016
Available online 10 November 2016
0921-5093/

Nomenclature

DP	dual-phase steels
MP	martensite particles
FF	ferrite-ferrite grain boundaries
FM	ferrite-martensite grain boundaries
XRD	x-ray Diffraction

be identified in DP steels [15]: isotropic forest hardening, and hardening from dislocation pile-ups [16]. The differentiation between unstable, ordered-dislocation patterns and more stable forest-dislocation patterns becomes a necessity for changing load paths [17,18].

The distinct Bauschinger-effect is caused by long-range stresses from dislocation arrangements with a dipole structure [19,20]. These long-range stresses lower the effective applied load in ferrite and increase the load in martensite. This mechanism induces a load-transfer from ferrite to martensite. The martensite grain-morphology influences the efficiency of the load-transfer, while the ferrite grain-size and martensite distribution influence the strain incompatibility [21]. It is noted, that the strain-gradient effect is stronger than the effects of initial anisotropy due to crystallographic texture in DP steels [7].

Different microstructure-based models from the literature highlight a characteristic rapid long-range stress evolution and a high initial dislocation-density production in DP steels. In order to model back-stresses in three-dimensional representative volume-element simulations, e.g., Kim et al. [22] propose different pile-up resistances for ferrite-martensite interfaces (6 GPa) and for ferrite-ferrite interfaces (1.2 GPa).

The well-known kinematic-hardening model of Chaboche [23] has been applied to DP steels with two spectral contributions [24]. There, the spectral contribution with a fast evolution rate represents the fast built-up of long-range stresses, i.e., when phases deform highly incompatible and dislocation pile-ups develop rapidly. The second spectral contribution describes the increase of the saturation value for long-range stresses.

Different versions of the model from [25] are used in [26] to predict an exponentially saturating kinematic saturation-stress for a DP steel. Additionally, a similar long-range stress behavior is predicted by the micromechanical model in [27,28]. The model by [27] is based on the work of Ashby [29], and utilizes an average evaluation of dislocation pile-ups at ferrite-ferrite and ferrite-martensite boundaries.

The long-range stress models in [30,17,31] are derived from each other. For all three works, modeling is based on the average interaction of opposing dislocation pile-ups on both sides of single-phase (ferrite-ferrite) grain-boundaries. A long-range stress evolution similar to the work at hand work is predicted but the models do not account for martensite geometry.

To the best knowledge of the authors, there exists no dislocation-density based mean-field material model that is (i) simple *and* physically-sound for different DP steels and (ii) incorporates important microstructural features besides grain-size. There are several works utilizing two- or three-dimensional representative volume-elements in conjunction with crystal-plasticity models [32]. While this approach has been shown to yield good results, the computational efforts are comparably high. The inclusion of non-local effects like long-range stresses or gradient-effects further increase the computational costs of full-field simulations.

In DP steels a scale separation is ensured by several orders of magnitude between local dislocation-interactions and structural-part dimensions. A multitude of different scale-bridging techniques exists in the literature. Originally developed for linear problems, the scale-bridging approach adopted in this work belongs to the group of analytical or semi-analytical mean-field methods. The two first order bounds – Voigt and Reuss bounds – result in a large possible effective-property spectrum if the phase properties vary significantly. In order to approximate fluctuations and interactions in more detail, refined estimation approaches rely on the analytical Eshelby solution for ellipsoidal inclusions. Classic examples are, e.g., the dilute-distribution model, the Mori-Tanaka model [17], the differential-scheme and self-consistent estimates [33]. Narrower bounds are possible by the variational approach of Hashin and Shtrikman [34]. The mentioned techniques have been applied and modified for non-linear material behavior, as done in this work.

Outline. Section 2 summarizes the nonlinear mean-field model that defines the strain localization. Section 3 introduces the proposed material-model. The experimental methods, applied to a DP600 steel for this work, are summarized in Section 4. Simulation results are discussed in Section 5, additionally, the influences of different microstructural parameters are investigated. Section 6 concludes this work's main findings.

Notation. A direct tensor notation is preferred throughout the text. For tensor components, Latin indices are used and Einstein's summation convention is applied. Vectors and second-order tensors are denoted by lowercase and uppercase bold letters. The composition of two second-order or two fourth-order tensors is formulated by $\boldsymbol{AB}$ and $\mathbb{AB}$. A linear mapping of second-order tensors by a fourth-order tensor is written as $\boldsymbol{A} = \mathbb{C}[\boldsymbol{B}]$. Scalar and dyadic products are denoted, e.g., by $\boldsymbol{A}\cdot\boldsymbol{B}$ and $\boldsymbol{A} \otimes \boldsymbol{B}$, respectively. The second-order identity is denoted by $\boldsymbol{I}$, and the identity on symmetric second-order tensors by $\mathbb{I}^S$. Completely symmetric and traceless tensors are designated by a prime while spherical tensors are denoted by a circle, e.g., $\boldsymbol{A}'$ and $\boldsymbol{A}^\circ$. The spherical projector is given by $\mathbb{P}_{sph} = \boldsymbol{I} \otimes \boldsymbol{I}/3$ and the deviatoric projector by $\mathbb{P}_{dev} = \mathbb{I}^S - \mathbb{P}_1$. DP x denotes a dual-phase steel with an ultimate tensile strength of x given in MPa.

2. Mean-field model

The different constituent or phase material-models interact through a non-linear, Hashin-Shtrikman mean-field model, as summarized below [35,36]. For a more detailed description, the reader is referred to [37,38]. The term phase is used in the sense of a domain with identical material properties.

The effective material behavior is modeled based on a linear-elastic, homogeneous comparison-material [34,39]. Stress polarizations are given relative to a comparison medium with the stiffness tensor $\mathbb{C}_0$, $\boldsymbol{p}(\boldsymbol{x}) = \boldsymbol{\sigma}(\boldsymbol{x}) - \mathbb{C}_0[\boldsymbol{\varepsilon}(\boldsymbol{x})]$. The stress tensor is then decomposed by $\boldsymbol{\sigma}(\boldsymbol{x}) = \mathbb{C}_0[\boldsymbol{\varepsilon}(\boldsymbol{x})] + \boldsymbol{p}(\boldsymbol{x})$, and a modified boundary value problem is given by

$$\mathrm{div}(\mathbb{C}_0[\boldsymbol{\varepsilon}(\boldsymbol{x})]) + \mathrm{div}(\boldsymbol{p}(\boldsymbol{x})) = \boldsymbol{0}. \tag{1}$$

In the following, piecewise-constant trial polarizations are assumed in each phase $\boldsymbol{p}(\boldsymbol{x}) = \sum_{\alpha=1}^{N} \chi_\alpha(\boldsymbol{x})\boldsymbol{p}_\alpha$, where $\chi_\alpha(\boldsymbol{x})$ denotes the indicator function of phase α. If the polarizations $\boldsymbol{p}(\boldsymbol{x})$ were known, the strains that solve the boundary-value problem in Eq. (1) can be simplified to the ensemble average

$$\boldsymbol{\varepsilon}_\alpha = \bar{\boldsymbol{\varepsilon}} + \frac{1}{c_\beta}\sum_{\beta=1}^{N} \mathbb{G}_{\alpha\beta}[\boldsymbol{p}_\beta]. \tag{2}$$

Here, $\bar{\boldsymbol{\varepsilon}}$ is the effective strain tensor, $\boldsymbol{\varepsilon}_\alpha = \langle\boldsymbol{\varepsilon}\rangle_\alpha$ are the average phase strains, the volume fractions of the respective domains are denoted by