Crystal Plasticity Constitutive Model for Multiphase Advanced High Strength Steels to Account for Phase Transformation and Yield Point Elongation

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Abstract. A constitutive law was developed based on a rate-independent crystal plasticity to account for the mechanical behavior of multiphase advanced high strength steels. Martensitic phase transformation induced by the plastic deformation of the retained austenite was represented by considering the lattice invariant shear deformation and the orientation relationship between parent austenite and transformed martensite. The stress dependent transformation kinetics were represented by adopting the stress state dependent volume fraction evolution law. The plastic deformation of the austenite was determined to have the minimum-energy associated with the work during the phase transformation. In addition to the martensitic phase transformation, yield point elongation and subsequent hardening along with inhomogeneous plastic deformation were also represented by developing a hardening stagnation model induced by the delayed dislocation density evolution.

1. Introduction
In order to obtain the formability and strength of the sheet metals demanded for automotive industries, advanced high strength steels having multiple phases are being intensively developed. Due to the nature of the multiphase steels, it is required to develop advanced constitutive laws to account for their complex material behavior such as phase transformation and yield point elongation.

In this work, a constitutive law was developed based on a rate-independent crystal plasticity to account for the mechanical behavior of multiphase advanced high strength steels. Also, yield point elongation and subsequent hardening of the advanced high strength steels along with inhomogeneous plastic deformation were also represented by developing a hardening stagnation model induced by the delayed dislocation density evolution.

2. Development of the phase transformation model
Considering the lattice invariant shear deformation and the orientation relationship between parent austenite and transformed martensite, the martensitic phase transformation induced by the plastic deformation of the retained austenite was represented. In order to accommodate the orientation relationship, in which the corresponding closed-packed planes and directions in the parent austenite and product martensite lattices are roughly parallel to each other, the rigid body rotation was introduced to have one or more undistorted and unrotated line (invariant-line). Due to the cooperative atomic
movement during the martensitic transformation, the interface between parent lattice and product lattice should be an undistorted and unrotated plane (invariant-plane). Since the combination of the rigid body rotation and Bain distortion is an invariant–line deformation, the lattice-invariant shear deformation was additionally considered for the phenomenological description of the martensitic phase transformation [1]. Based on the lattice-invariant shear deformation on planes, the macroscopic deformation gradients having 24 Kurdjumov–Sachs (KS) orientation variants were generated by decomposing the rigid body rotation and Bain distortion into the macroscopic deformation and complementary shear deformation [2, 3].

In order to consider the stress dependency of the transformation kinetics, a stress state dependent volume fraction evolution law [4] was adopted as shown in Figure 1.

![Figure 1. Measured and calibrated transformation kinetics dependent on stress state.](image)

For the retained austenite lattice, the following additive decomposition of the strain increment was assumed,

\[ \delta \epsilon = \delta \epsilon^e + \delta \epsilon^{\text{slip}} + \delta \epsilon^{\text{mart}} \] (1)

Here, \( \delta \epsilon^e \), \( \delta \epsilon^{\text{slip}} \) and \( \delta \epsilon^{\text{mart}} \) are the elastic strain increment, plastic strain increment by slip deformations and plastic strain increment by the martensitic transformation, respectively. Among the various available plastic deformations for the martensitic transformation, the optimized plastic deformation was decided to have minimum-energy associated with the work during the phase transformation:

\[ \delta \epsilon^{\text{mart}} = \sum_{k=1}^{N} d f_{k}^{\alpha^*} \delta \epsilon_{k}^{\text{mart}} \] (2)

where \( \sum_{k=1}^{N} d f_{k}^{\alpha^*} = d f^{\alpha^*} \) and \( \delta \epsilon_{k}^{\text{mart}} \) is the strain increment of each variant, \( k \), obtained from the macroscopic deformation gradient during the martensitic transformation.

As for the product martensite, the strain increment was assumed to be decomposed into the elastic strain increment and plastic strain increment:

\[ \delta \epsilon = \delta \epsilon^e + \delta \epsilon^{\text{slip}} \] (3)
For a unit cell, which is a combination of parent austenite and product martensite as schematically shown in Figure 2, the Taylor type homogenization scheme was adopted. In the Taylor type scheme, which is an upper bound scheme, the parent austenite and product martensite share the same strain increment. Then, the homogenized stress can be obtained as,

$$\sigma = (1 - f^\alpha)\sigma^\gamma + f^\alpha\sigma^\alpha$$  \hspace{1cm} (4)$$

where $\sigma^\gamma$ and $\sigma^\alpha$ are stresses of the retained austenite and the transformed martensite, respectively.

Figure 2. Schematic view of a parent austenite unit cell.

The developed constitutive law was implemented into the commercial FE code ABAQUS/Explicit with the aid of user-defined material subroutine. As a preliminary application, stress-strain response of the retained austenite, product martensite and homogenized unit cell for simple tension was calculated and the comparison results are shown in Figure 3.

Figure 3. Comparison of the measured and calculated uniaxial tension simulation results.

3. Development of the yield point elongation model

The yield point elongation and subsequent hardening of the advanced high strength steels along with inhomogeneous plastic deformation were also represented by developing a hardening stagnation model induced by the delayed dislocation density evolution. The constitutive model was designed to account for the upper yield point due to the strong elastic interactions between the solute atoms and dislocations as well as the lower yield point due to the annihilation of the pinned dislocations. In addition to the effect of the edge dislocations on the critical resolved shear stress of the crystal slip systems, the pinned dislocation effect was added to the dislocation density:
\[ \rho^\alpha = \rho_\text{Edge}^\alpha + \rho_\text{Pin}^\alpha \] (5)

Here, \( \rho_\text{Edge}^\alpha \) is the edge dislocation density and \( \rho_\text{Pin}^\alpha \) is the pinned dislocation density. The edge dislocation density and the pinned dislocation density were designed to be delayed and annihilated by the interactions between the solute atoms and the dislocations.

Considering the impractical huge size of the representative volume element to represent the inhomogeneous material property distribution of the multiphase advanced high strength steels, the developed constitutive model was implemented into the phenomenological model by assigning different volume fraction for each phase within a dogbone specimen. Figures 4 and 5 show the simulated equivalent strain distribution and stress-strain curve during the uniaxial tensile test, respectively. As shown in the comparison in Figure 5, the serrated stress-strain curves as well as the yield point elongation were successfully predicted by the developed model.

**Figure 4.** Equivalent plastic strain distribution with a dogbone specimen during the uniaxial tension test.

**Figure 5.** Comparison of the simulated and measured stress-strain curves during the uniaxial tension test.

References

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