

Prediction of the mechanical behaviour of TRIP steel

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Abstract

TRIP steel contains different phases, ferrite, bainite, austenite and martensite. During deformation the metastable retained austenite tends to transform to stable martensite. The accompanying transformation strain has a beneficial effect on the ductility of the steel during forming. By changing the alloy composition, the rolling procedure and the thermal processing of the steel, a wide range of different morphologies and microstructures can be obtained. Interesting parameters are the amount of retained austenite, the stability of the austenite as well as its hardness. A constitutive model is developed for TRIP steel which contains four different phases. The transformation of the metastable austenite to martensite is taken into account. The phase transformation depends on the stress in the austenite. Due to the differences in hardness of the phases the austenite stress is not equal to the overall stress. An estimate of the local stress in the austenite is obtained by homogenization of the response of the phases using a self-consistent mean-field homogenization method. Overall stress-strain results as well as stress-strain results for individual phases are compared to measurements found in literature for some TRIP steels. The model is then used to explore the influence of some possible variations in microstructural composition on the mechanical response of the steel.

Keywords: TRIP, Martensitic transformation, Mean field homogenization

Introduction

The existence of different phases in the microstructure of TRIP steels is a consequence of its chemical composition and the heat treatment during production. Two main constituent phases are ferrite and austenite and depending on the heat treatment bainite and martensite may also form. The austenite phase (γ) is in a metastable state. It can transform into stable martensite (α') during deformation. One of the attractive features of these steels is the fact that with slight changes in the heat treatment and/or chemical composition, a material with significantly different mechanical properties can be obtained (Jacques et al. (2001)). The aim of this study therefore is to build a model that can be used to predict the final mechanical properties based on know-ledge about the constituent phases.

The model is based on the Mean-Field homogenization technique for computing the stress-strain distribution into different phases (see e.g. Doghri and Friebe (2005)). In this method the fields for the mechanical variables such as strain and stress are represented by their average values over the sub-domains. This method is well established to be used for binary mixtures of phases. Extension to mixtures of three or four phases have been presented by Lani et al. (2007) and Delannay et al. (2008). The former present a parallel Mori-Tanaka model (Mori and Tanaka (1973)) of both austenite and martensite in a ferritic matrix. The latter formulate a hierarchic Mori-Tanaka model of martensite as inclusion in an austenite matrix and the aggregate of these as an inclusion in the ferritic-bainitic matrix.

In this research application of the mean-field method for more than two phases is investigated. For the model the self-consistent scheme is used that implicitly takes into account existence of any number of phases.

The martensitic transformation is modelled as a stress-driven process similar to the model of Tamura (1982) and

theoretically and experimentally justified by Chatterjee and Bhadeshia (2007) and Das et al. (2011). This is in contrast to the model of deformation induced martensitic transformation as formulated by Olson and Cohen (1975) and extended by Stringfellow et al. (1992). The model depends on the stress resolved in the austenite phase and transformation is determined as a function of the additional mechanical driving force supplied to the material as formulated by Patel and Cohen (1954) and applied to stress induced martensitic transformations by Fischer and Reisner (1998) and Geijselaers and Perdahcioğlu (2009).

Mean-Field Homogenization

The Mean-Field homogenization method is based on the interaction and evolution of the average values of the field variables in sub domains that divide the overall structure. The overall stress σ and strain ϵ are related to the averages in the individual phases by:

$$\sigma = \sum f_i \sigma_i, \quad \epsilon = \sum f_i \epsilon_i \quad (1)$$

The f_i stands for the volume fraction of the phases. It is assumed that the macroscopic stress-strain relation that is determined for an individual phase is also valid within the compound:

$$\dot{\sigma}_i = C_i : \mathbf{d}_i \quad (2)$$

where \mathbf{d}_i is the average strain rate in the i^{th} phase and C_i is an elasto-plastic tangent. Finally the relation between average phase strain rates and the average overall strain rate has to be specified through strain concentration tensors \mathbf{A}_i :

$$\mathbf{d}_i = \mathbf{A}_i : \mathbf{d} \quad (3)$$

which by virtue of Equation (1) are subject to:

$$\sum f_i \mathbf{A}_i = \mathbf{I} \quad (4)$$

The homogenized response of the composite material is then found as:

$$\mathbf{C} = \sum f_i \mathbf{C}_i : \mathbf{A}_i \quad (5)$$

Self-consistent model. Different homogenization schemes have been formulated using specific definitions of \mathbf{A} . Schemes suitable for more than two phases are e.g. iso-strain (Voigt-Taylor) and iso-stress (Reuss-Sachs) models. Here we use the self-consistent method with strain concentration calculated according to Eshelby (1957). The self-consistent scheme has originally been developed to compute the mechanical response of polycrystals (Kröner (1958), Budiansky and Wu (1962), Hill (1965)) where the interaction of the matrix and the individual grains is taken into account using Eshelby's equivalent inclusion theory. In the self-consistent scheme each phase is considered as an inclusion in a matrix which has homogenized response. The strain concentration tensor for phase i is defined as:

$$\mathbf{A}_i = \left(\mathbf{I} - \mathbf{S} : \left(\mathbf{I} - \mathbf{C}^{-1} : \mathbf{C}_i \right) \right)^{-1} \quad (6)$$

The fourth order Eshelby tensor \mathbf{S} depends on the stiffness of the matrix and the aspect ratio of the inclusions. However the matrix stiffness \mathbf{C} depends on the strain concentration tensors for the individual phases. The calculation of the \mathbf{A} -tensors therefore requires an iterative procedure.

Eshelby's equivalent inclusion theory has been formulated for isotropic elastic matrix constitutive behaviour. For use with an elasto-plastic matrix, the anisotropic material model has to be substituted by an isotropic comparison material model through an isotropic projection as outlined by Doghri and Ouaar (2003).

$$\begin{aligned} \mathbf{C}^{\text{iso}} &= 3\tilde{\kappa} \mathbf{I}^{\text{vol}} + 2\tilde{\mu} \mathbf{I}^{\text{dev}} \\ \tilde{\kappa} &= \mathbf{C} :: \mathbf{I}^{\text{vol}}, \quad \tilde{\mu} = \frac{1}{10} \mathbf{C} :: \mathbf{I}^{\text{dev}} \\ \mathbf{I}^{\text{vol}} &= \frac{1}{3} \mathbf{1} \otimes \mathbf{1}, \quad \mathbf{I}^{\text{dev}} = \mathbf{I} - \mathbf{I}^{\text{vol}} \end{aligned} \quad (7)$$

The expression for the Eshelby tensor for spherical inclusions in terms of $\tilde{\kappa}$ and $\tilde{\mu}$ is:

$$\mathbf{S} = \frac{3\tilde{\kappa}}{3\tilde{\kappa} + 4\tilde{\mu}} \mathbf{I}^{\text{vol}} + \frac{6\tilde{\kappa} + 2\tilde{\mu}}{5\tilde{\kappa} + 4\tilde{\mu}} \mathbf{I}^{\text{dev}} \quad (8)$$

For detailed modelling of the martensitic transformation each martensite plate should be modelled as a thin ellipsoid with a specific orientation (Cherkaoui et al. (2000)). In our macroscopic model the effect of many randomly oriented martensite plates is modelled by considering the

martensite inclusions as spheres. This greatly simplifies the model.

Austenite Transformation Model

The model for transformation of the retained austenite is based on the concept of mechanical driving force for martensite transformation as posed by Patel and Cohen (1954). The martensitic transformation involves a diffusionless change of crystal structure. This was analyzed by Wechsler et al. (1953) starting from the postulate of an invariant plane (habit plane) as interface between the martensite and the parent austenite. The resulting deformation gradient can be shown to be:

$$\mathbf{F}^{\text{tr}} = \mathbf{1} + \mathbf{m} \otimes \mathbf{n} \quad (9)$$

where \mathbf{n} is the normal to the habit plane and \mathbf{m} is the shear vector. Due to lattice symmetry 24 different transformation systems (\mathbf{n}, \mathbf{m}) can be identified.

When a stress $\boldsymbol{\sigma}$ acts while the transformation evolves, this supplies additional mechanical driving force for the transformation, following Patel and Cohen (1954):

$$\begin{aligned} U &= \boldsymbol{\sigma}_\gamma : (\mathbf{m} \otimes \mathbf{n}) \\ &= \boldsymbol{\sigma}_\gamma : \frac{1}{2} (\mathbf{m} \otimes \mathbf{n} + \mathbf{n} \otimes \mathbf{m}) \end{aligned} \quad (10)$$

Here $\boldsymbol{\sigma}_\gamma$ is the Cauchy stress in the austenite phase. In a polycrystalline material there are always some grains optimally oriented with respect to the local stress to maximize the mechanical driving forces. When this maximum exceeds a critical value ΔG^{cr} , according to Tamura (1982) the transformation will start.

$$U^{\text{max}} = \sum \sigma_{\gamma_i} \lambda_i > \Delta G^{\text{cr}} \quad (11)$$

where λ_i are the eigenvalues of the transformation deformation tensor in Equation (10) and σ_{γ_i} are the eigenvalues of the local austenite stress tensor, both sorted in ascending order. In terms of the often used parameters, transformation dilatation $\delta = \mathbf{m} \cdot \mathbf{n}$ and transformation shear $\gamma = (\mathbf{1} - \mathbf{n} \otimes \mathbf{n}) \cdot \mathbf{m}$, the values of λ can easily be calculated as:

$$\lambda_{1,3} = \frac{1}{2} \left(\delta \mp \sqrt{\gamma^2 + \delta^2} \right); \quad \lambda_2 = 0 \quad (12)$$

The amount of martensite formed is expressed as a function of U^{max} :

$$f_{\alpha'} = f_{\alpha'}^0 + F(U^{\text{max}} - \Delta G^{\text{cr}}) f_{\gamma}^0 \quad (13)$$

The function F is a saturating exponential curve as in Koistinen and Marburger (1959), fitted with a smooth transition at $U^{\text{max}} - \Delta G^{\text{cr}} = 0$ as in Perdahcioğlu (2008).

Transformation Plasticity

The significant enhancement of the formability of TRIP steel is attributed to the transformation plasticity, which in martensitic transformations is explained by Magee (1966) by variant selection and preferential orientation of the martensite. The local transformation strain is given by Equation (9). Macroscopically this is modelled as a dilatation plus a deviatoric component aligned with the stress in the austenite phase:

$$\mathbf{d}^{\text{tr}} = \mathbf{T} \dot{f}_{\alpha'} = \left(\frac{3}{2} T \mathbf{s}_{\gamma} + \frac{1}{3} \delta \mathbf{1} \right) \dot{f}_{\alpha'} \quad (14)$$

The factor T can be calculated by assuming that $\sigma_{\gamma} : \mathbf{d}^{\text{tr}}$ equals ΔG^{cr} :

$$T = \frac{1}{(\sigma_{\gamma}^{\text{VM}})^2} (\Delta G^{\text{cr}} - \sigma_{\gamma}^{\text{h}} \delta) \quad (15)$$

This means that for higher stresses, when most of the austenite has already transformed, the transformation plasticity becomes less pronounced.

Constitutive model

To obtain a stress-strain relation first the deformation rate is partitioned in an elasto-plastic deformation rate \mathbf{d}^{ep} and a transformation plasticity rate \mathbf{d}^{tr} . The elasto-plastic rate is then partitioned among the phases:

$$\dot{\boldsymbol{\sigma}} = \left(\sum f_i \mathbf{C}_i : \mathbf{A}_i \right) : (\mathbf{d} - \mathbf{d}^{\text{tr}}) \quad (16)$$

The transformation plasticity depends on the transformation rate $\dot{f}_{\alpha'}$, which in turn depends on the stress rate in the austenite. It is possible to write \mathbf{d}^{tr} as a function of the overall strain rate \mathbf{d} :

$$\mathbf{d}^{\text{tr}} = \frac{\mathbf{T} \otimes \frac{d f_{\alpha'}}{d \boldsymbol{\sigma}_{\gamma}} : \mathbf{C}_{\gamma} : \mathbf{A}_{\gamma}}{1 + \frac{d f_{\alpha'}}{d \boldsymbol{\sigma}_{\gamma}} : \mathbf{C}_{\gamma} : \mathbf{A}_{\gamma} : \mathbf{T}} : \mathbf{d} \quad (17)$$

Finally the stress response for the homogenized material including transformation plasticity is obtained as:

$$\dot{\boldsymbol{\sigma}} = \mathbf{C} : \left(\mathbf{I} - \frac{\mathbf{T} \otimes \frac{d f_{\alpha'}}{d \boldsymbol{\sigma}_{\gamma}} : \mathbf{C}_{\gamma} : \mathbf{A}_{\gamma}}{1 + \frac{d f_{\alpha'}}{d \boldsymbol{\sigma}_{\gamma}} : \mathbf{C}_{\gamma} : \mathbf{A}_{\gamma} : \mathbf{T}} \right) : \mathbf{d} \quad (18)$$

It is interesting to note that this material tangent possesses minor symmetry but not major symmetry.

Comparison with Experiments

Extensive stress and strain measurements on a specific TRIP steel are presented by Jacques et al. (2007). The strain partitioning among the phases was measured by digital image correlation on SEM micrographs acquired in situ during tensile tests. The stress partitioning between the phases was measured by neutron diffraction in situ during tensile tests. The elastic strains of the fcc phase austenite and the bcc phases ferrite and bainite could be measured. Stresses in individual bcc phases cannot be obtained since these give identical diffraction peaks.

The stress strain response of this TRIP steel has been simulated. The material data used in the simulation are given in **Table 1**.

Table 1. Material data used for simulation.

Phase	fraction	σ_0^{γ} (MPa)	K (MPa)	m	ε^0
martensite	0.00	1500	1000	0.12	0.001
austenite	0.12	1150	1500	0.21	0.010
bainite	0.33	700	1000	0.19	0.008
ferrite	0.55	600	1500	0.19	0.008

The elasticity modulus and Poisson ratio for all phases are chosen equal, $E = 210$ GPa and $\nu = 0.3$. The yield stress is described by the hardening function

$\sigma_{\gamma}^{\gamma}(\varepsilon_i^p) = \sigma_i^0 + K_i(\varepsilon_i^0 + \varepsilon_i^p)^{m_i}$. The critical energy barrier is chosen as $\Delta G^{\text{cr}} = 175$ MPa. The transformation is characterized by $\delta = 0.02$ and $\gamma = 0.23$.

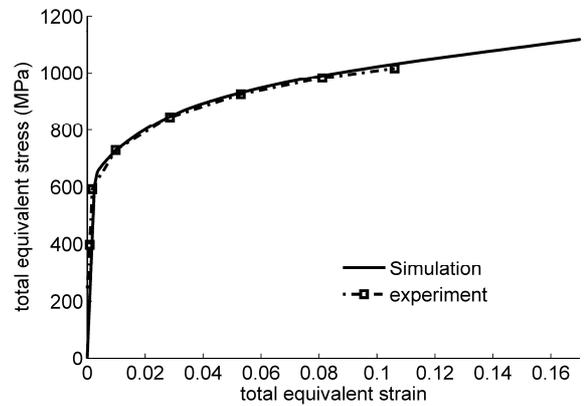


Figure 1. Stress-strain response of TRIP steel, comparison of simulation and experiment.

In **Fig. 1** the computed response of the TRIP steel under uni-axial tension is shown. The simulations agree well with the experimental data.

The partitioning of the stress among the fcc and bcc phases is shown in **Fig. 2**. The stress in the bcc phase is the average stress in ferrite and bainite. Note the apparent low yield stress in the austenite. This is an artefact of the strain concentration into the softer ferrite and bainite. The actual strain in the austenite is much lower than the overall

macroscopic strain. Yielding of the austenite only happens at a stress of approximately 1150 MPa.

In **Fig. 3** the evolution of the retained austenite fraction is shown, compared to experimental results. This result is not very convincing. In Jacques et al. (2009) it is shown that reliable quantitative measurement of retained austenite is very difficult. On the other hand the function F in Equation (13) as used in the calculations is based on measurements on austenitic stainless steel. The retained austenite in TRIP steel shows a broader spread in local Carbon content and therefore also a broader stability distribution. The function F should perhaps be less steep for TRIP steel.

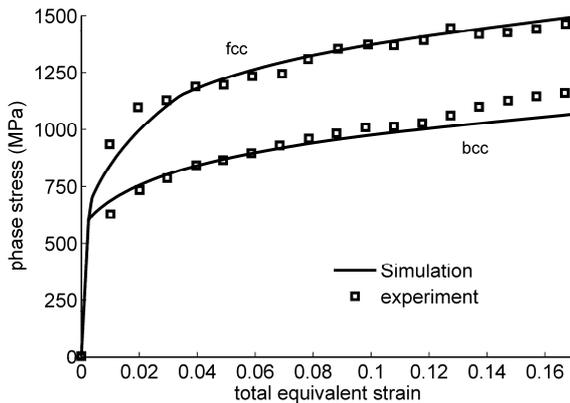


Figure 2. Stress partitioning among the phases.

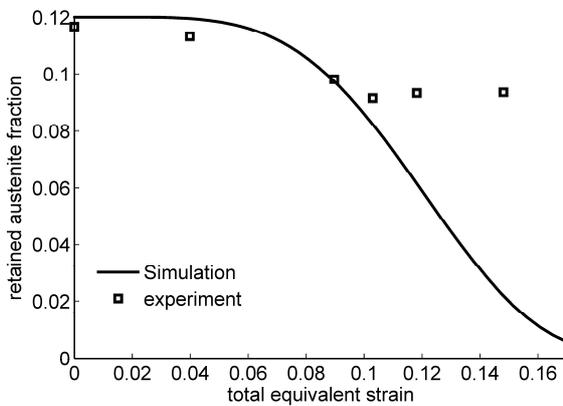


Figure 3. The evolution of retained austenite.

Material Data Variations

By applying slight changes in heat treatment or alloy composition it is possible to effect big changes in morphology of TRIP steels. Here we will use the developed model to explore the influence on the macroscopic behaviour of the steel of some of these changes.

Three material parameters are varied independently with respect to the morphology and data of the material considered here: the initial austenite fraction, the stability of the retained austenite and the yield stress of the austenite. In reality it is not possible to independently change these

parameters. All three are linked through the Carbon content of the retained austenite. However by adding different alloying elements or by changing the heat treatment or applying more or less hot rolling reduction Jacques et al. (2001) show that there is some room to play.

The effects of the variations on the stress-strain response as well as the incremental hardening exponent which is a measure of the formability are considered.

Initial austenite fraction. At low initial austenite fractions hardly any effect of transformation is visible in the stress-strain diagram of Fig. 4a. In the hardening diagram, Fig 4b, a small secondary hardening peak is visible at high strain. At high initial austenite fraction the stress strain response starts resembling the behaviour of fully austenitic steel with a pronounced secondary hardening when the austenite starts transforming. This is also apparent in the hardening diagram where the range of uniform elongation is extended from 15% to over 20%.

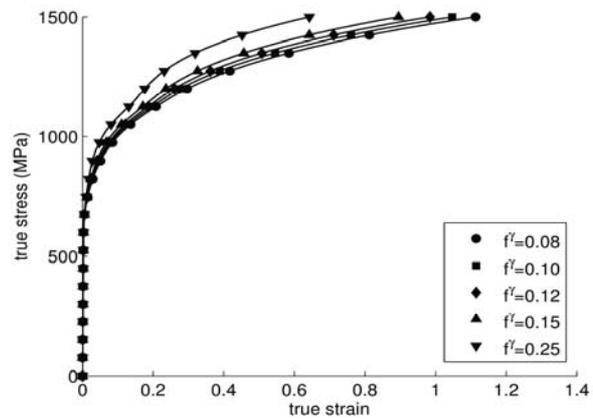


Figure 4a. Influence of initial retained austenite fraction on stress-strain response.

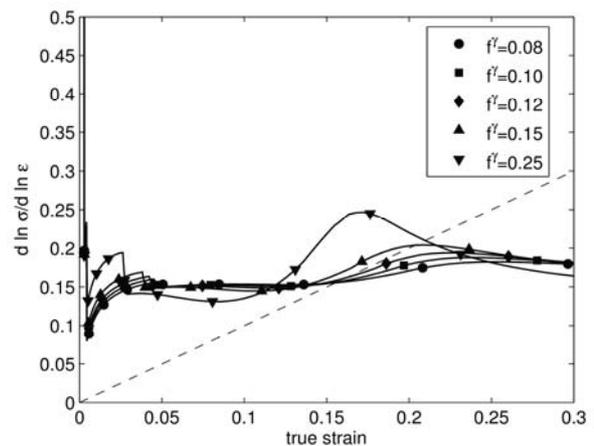


Figure 4b. Influence of initial retained austenite fraction on the formability.

Austenite stability. At first sight the effect of variation of austenite stability is negligible. This especially applies

when only the stress-strain curve of **Fig. 5a** is considered. The hardening diagram, **Fig. 5b**, shows however that the effect is more subtle. It seems possible to tune the formability of the material when the austenite stability can be controlled.

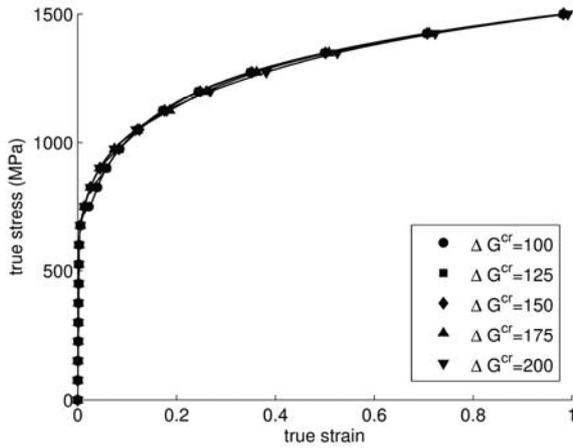


Figure 5a. Influence of austenite stability on stress-strain response.

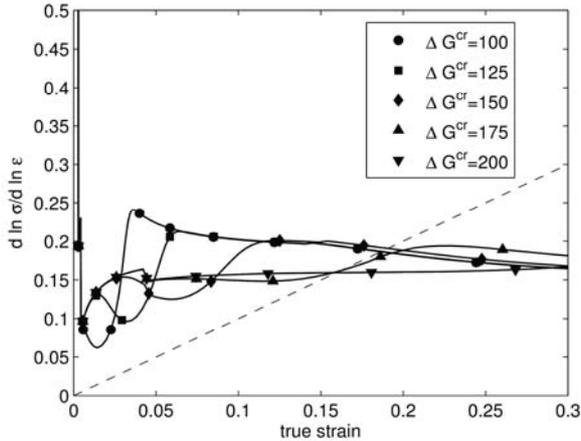


Figure 5b. Influence of austenite stability on formability.

Austenite yield stress. Also the initial austenite yield stress seems to have only little influence when only the stress-strain curve in **Fig. 6a** is considered. From the hardening diagram, **Fig. 6b**, it appears that an optimal value of austenite hardness exists, probably in a combination with a specific value of austenite stability ΔG^{Cr} .

On the basis of the limited results shown here a tentative guess for the optimal relation between austenite stability and austenite initial yield stress might be:

$$\Delta G^{Cr} > \lambda_3 \sigma_{0\gamma}^y$$

This would mean that (in a tensile test) ideally the transformation should start after the austenite has started to yield.

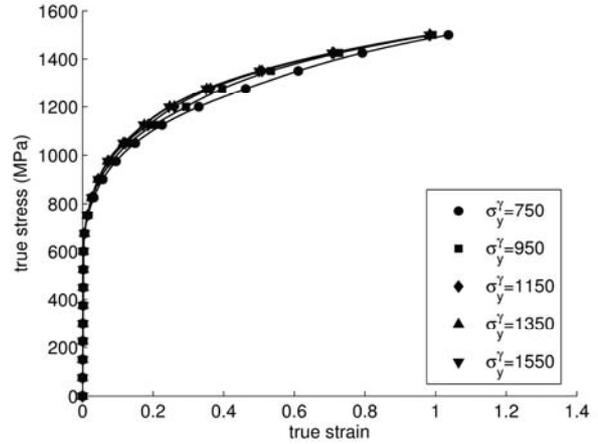


Figure 6a. Influence of austenite yield stress on stress-strain response.

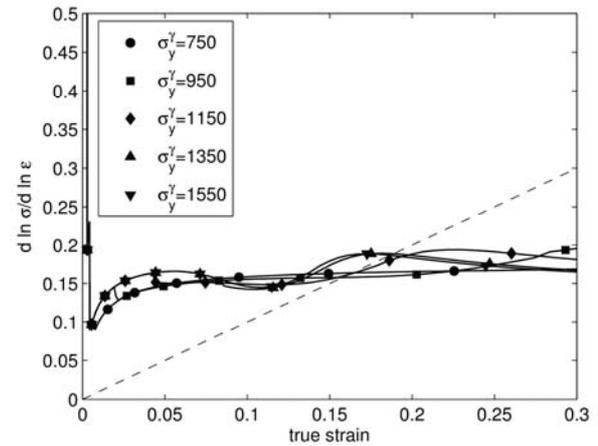


Figure 6b. Influence of austenite yield stress on formability.

Conclusion

The constitutive behaviour of a TRIP steel has been modelled using self-consistent mean-field homogenization. This allows calculating an estimate of the partitioning of the applied strain over the individual phases as well as evaluation of the average stresses in the phases.

The transformation of the metastable retained austenite to stable martensite is modelled by considering the average stress in the austenite as the driving force for the transformation.

Application of this model to results published in literature shows that it very well reproduces the overall stress-strain behaviour as well as the partitioning of the stresses

to the individual phases. The prediction of the transformed fraction of retained austenite is however less convincing.

The model can be used to explore the influence of variations in material parameters which can be influenced when designing a TRIP steel by varying its alloy contents, its heat treatment or its hot rolling program.

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