Stability and Constitutive Modelling in Multiphase TRIP Steels

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Multiphase TRIP steels are a relatively new class of steels exhibiting excellent combinations of strength and cold formability, a fact that renders them particularly attractive for automotive applications. The present work reports models regarding the prediction of the stability of retained austenite, the optimisation of the heat–treatment stages necessary for austenite stabilization in the microstructure, as well as the mechanical behaviour of these steels under deformation. Austenite stability against mechanically–induced transformation to martensite depends on chemical composition, austenite particle size, strength of the matrix and stress state. The stability of retained austenite is characterized by the $M^*_s$ temperature, which can be expressed as a function of the aforementioned parameters by an appropriate model presented in this work. Besides stability, the mechanical behaviour of TRIP steels also depends on the amount of retained austenite present in the microstructure. This amount is determined by the combinations of temperature and temporal duration of the heat–treatment stages undergone by the steel. Maximum amounts of retained austenite require optimisation of the heat–treatment conditions. A physical model is presented in this work, which is based on the interactions between bainite and austenite during the heat–treatment of multiphase TRIP steels, and which allows for the selection of treatment conditions leading to the maximization of retained austenite in the final microstructure. Finally, a constitutive micromechanical model is presented, which describes the mechanical behaviour of multiphase TRIP steels under deformation, taking into account the different plastic behaviour of the individual phases, as well as the evolution of the microstructure itself during plastic deformation. This constitutive micromechanical model is subsequently used for the calculation of forming limit diagrams (FLD) for these complex steels, an issue of great practical importance for the optimisation of stretch–forming and deep–drawing operations.

Keywords: multiphase TRIP steels, modelling, retained austenite, stability, heat–treatment, forming limit diagrams

Introduction

Low-alloy TRIP steels are a relatively new class of steels that exhibit excellent combinations of strength and cold formability, making them particularly suitable for sheet-forming applications in the automotive industry. These steels possess a multiphase microstructure containing ferrite, bainite and retained austenite. During cold-forming operations, such as stretch-forming and deep-drawing, the retained austenite transforms to martensite under the action of the applied stresses and strains. This mechanically-induced martensitic transformation of the retained austenite is responsible for the TRansformation-Induced Plasticity (TRIP) effects found in these materials. These effects include significant improvements in ductility and formability.

Austenite stability against mechanically–induced transformation to martensite is known to depend on chemical composition, austenite particle size, strength of the matrix and stress state. The stability of retained austenite can be characterized by the $M^*_s$ temperature, which can be expressed as a function of the aforementioned parameters by an appropriate model presented in this work.

Besides stability, the mechanical behaviour of multiphase TRIP steels also depends on the amount of retained austenite present in the microstructure. This amount is determined by the combinations of temperature and temporal duration of the heat–treatment stages undergone by the steel. To achieve a maximum amount of retained austenite, the optimisation of the heat–treatment conditions is required. A physical model is presented in this work, which is based on the interactions between bainite and austenite during the heat–treatment of multiphase TRIP steels, and which allows for the selection of treatment conditions leading to the maximization of retained austenite in the final microstructure.

Finally, a constitutive micromechanical model is presented, which describes the mechanical behaviour of multiphase TRIP steels under deformation, taking into account the different plastic behaviour of the individual phases, as well as the evolution of the microstructure itself during plastic deformation. This constitutive micromechanical model is subsequently used for the calculation of forming limit diagrams (FLD) for these complex steels, an issue of great practical importance for the optimisation of stretch–forming and deep–drawing operations.

Modelling of the Retained Austenite Stability

The first step towards developing a model for the stability of retained austenite against mechanically-induced transformation is to identify a single parameter to characterize stability. In this work, the $M^*_s$ temperature has been chosen for this task, in the same way that the $M_s$ temperature is used to characterize the stability of austenite against transformation on cooling. The $M^*_s$ temperature can be defined by considering the mechanically-induced martensitic transformation of austenite in Figure 1. Spontaneous transformation occurs on cooling the austenite to the $M_s$ temperature. This transformation is triggered at pre-existing nucleation sites in the austenite. The same sites can operate also above the $M_s$ temperature under the action of an externally applied stress. The higher the temperature is, the higher the required stress. This stress-assisted transformation is denoted by the line AC.

At point C and at the $M^*_s$ temperature, the applied stress reaches the yield stress of the austenite. Above $M^*_s$ new potent nucleation sites, which are produced by the plastic deformation of austenite, trigger the strain-induced transformation. Thus the $M^*_s$ temperature defines a boundary between the temperature regimes where separate modes of transformation dominate: below $M^*_s$ the transformation is...
stress-assisted and above $M_s^p$ the transformation is strain-induced. Near the $M_s^p$ temperature both modes operate.

Due to transformation plasticity the observed yield stress follows the stress for stress-assisted transformation below the $M_s^p$. A reversal in the temperature dependence of the flow stress provides a convenient determination of the $M_s^p$ temperature. Actually this technique has been applied successfully for the determination of the $M_s^p$ temperature in steels containing austenitic dispersions, either as retained austenite in martensitic steels [1], or more recently for retained austenite in low-alloyed multiphase TRIP steels [2].

Having defined the $M_s^p$ temperature as the single parameter characterizing stability, the model objective is to develop an expression for the $M_s^p$ as a function of the four stability parameters that affect it, i.e. the chemical composition of retained austenite, the size of the retained austenite particles, the strength of the matrix and the stress state. In other words, an expression of the following form is sought:

$$M_s^p = M_s^p(X_i, V_p, \sigma_s, \sigma_h/\overline{\sigma})$$  \hspace{1cm} (1)

In Eq. (1), $X_i$ denotes the mole fraction of alloying elements (i.e. the chemical composition) in austenite, $V_p$ is the mean austenite particle volume, $\sigma_s$ is the matrix yield strength and $\sigma_h/\overline{\sigma}$ is the ratio of the hydrostatic to the von Mises equivalent stress, which characterizes the triaxiality of the stress state. The model details have been reported elsewhere [3], so only the key points will be presented here.

The model is based on the fact that, for the case of stress-assisted transformation, the applied elastic stress aids the transformation kinetics by modifying the effective potency distribution of pre-existing nucleation sites. According to Olson and Cohen [4], heterogeneous martensitic nucleation can proceed by the dissociation of an existing defect, which has the form of a dislocation array on an existing grain or interphase boundary. The dissociation of such a defect produces a fault structure or martensitic embryo with a thickness of $n$ crystal planes, which possesses a fault energy $\gamma_f(n)$ per unit area:

$$\gamma_f(n) = n \rho \left( \Delta G_{ch} + E_{str} + W_f \right) + 2 \gamma_s$$  \hspace{1cm} (2)

In Eq. (2), $\Delta G_{ch}$ is the chemical driving force for martensitic transformation per unit volume, $\gamma_s$ the specific fault/matrix interfacial energy, $\rho$ the density of atoms in the fault plane, $E_{str}$ the elastic strain energy per unit volume associated with distortions in the fault interface plane and $W_f$ the frictional work of interfacial motion, which occurs during the dissociation process. Spontaneous martensitic nucleation occurs when $\gamma_f(n) \leq 0$. In this case, the dissociation is barrierless and occurs at a critical value of the driving force. Based on the above, the potency of a nucleation site can be defined by the thickness ($n$) of the nucleus that can be produced from the defect by barrierless dissociation. The critical $n$ for nucleation follows from Eq. (2) as:

$$n = -\frac{2 \gamma_s}{\rho \left( \Delta G_{ch} + E_{str} + W_f \right)}$$  \hspace{1cm} (3)

Based on the above, Cohen and Olson [5] derived the cumulative defect-potency distribution from the Cech and Turnbull small-particle experiments in Fe–30%Ni alloys [6] as:

$$N_s(n) = N_{w0} \exp(-a n)$$  \hspace{1cm} (4)

In Eq. (4), $N_s(n)$ is the number density of sites of sufficient potency to nucleate martensite, $N_{w0}$ the total number of nucleation sites of all potencies, and $a$ is a constant. The effect of stress on the potency distribution can be found by adding a mechanical contribution term, $\Delta G_\sigma$, to the chemical driving force of Eq. (3) to obtain:

$$N_s(n) = N_{w0} \exp \left[ \frac{2 a \gamma_s}{\rho \left( \Delta G_{ch} + \Delta G_\sigma + E_{str} + W_f \right)} \right]$$  \hspace{1cm} (5)

The term $\Delta G_\sigma$ of Eq. (5) is equal to:

$$\Delta G_\sigma = \sigma \frac{\partial (\Delta G)}{\partial \sigma}$$  \hspace{1cm} (6)

In Eq. (6), $\Delta G$ is the overall Gibbs free energy change for the austenite to martensite transformation and $\sigma$ the applied stress.

The stress-assisted transformation of a dispersion of austenite particles of average particle volume $V_p$ is controlled by the potency distribution of Eq. (5). The fraction of particles, $f$, to transform is equal to the probability of finding at least one nucleation site in the particle, assuming

![Figure 1. Temperature for the start of martensitic transformation as a function of applied stress.](image-url)
that a single nucleation event transforms the entire austenite particle to martensite. This probability is:

$$f = 1 - \exp (-N_e \cdot V_p)$$

(7)

The transformation stress $\sigma = \sigma_t$, at which the martensitic transformation is triggered, can be found by combining Eqs. (5), (6) and (7) as:

$$\sigma_t = \frac{1}{\bar{\sigma}} \frac{\partial (\Delta G)}{\partial \sigma}$$

$$= \frac{2 a \cdot \gamma_s}{\rho \cdot \ln \left( -\frac{\ln(1-f)}{N_e \cdot V_p} \right)} - \frac{\Delta G_{ch} - E_{str} - W_f}{\bar{\sigma}}$$

(8)

In the above equation the chemical driving force term is temperature and composition dependent. The frictional work term is also a function of chemical composition, whereas the mechanical driving force contribution term is a function of stress-state. Detailed analytical expressions for these functions are given in reference [3]. The $M_s^\alpha$ temperature can be calculated by letting the transformation stress be equal to the yield stress ($\sigma_y$) in Eq. (8) and solve for the temperature, since the term $\Delta G_{ch}$ is temperature dependent. This operation yields a general expression for the $M_s^\alpha$ temperature as:

$$M_s^\alpha = (a_1 + a_2 X_C)^{-1} \left\{ b_1 + b_2 X_C + b_3 X_Mn + b_4 X_C X_Mn + \ln \left( \frac{c_1}{c_2 / V_p} \right) + \sigma_y (d_1 + d_2 (\sigma_y / \bar{\sigma})) \right\}$$

(9)

Constants ($a_1$, $b_2$, $c_1$, $d_1$) in Eq. (9) depend on the specific steel composition and, as stated earlier, are given in [3]. The key point is, however, that Eq. (9) provides an analytical expression for the $M_s^\alpha$ temperature as a function of the chemical composition of austenite, the austenite particle size, the yield strength and the stress state.

**Modelling of Heat-treatment for Austenite Stabilization**

As stated in the introduction, TRIP effects and the associated formability enhancement depend directly on the amount and stability of retained austenite. A large research effort over the last years has focused on establishing the suitable heat-treatments for obtaining stable retained austenite in the microstructure. A typical heat-treatment for the production of a multiphase TRIP steel is shown in **Figure 2**. It involves intercritical annealing to produce a ferrite-austenite mixture, followed by cooling to an intermediate temperature and holding for the isothermal transformation of austenite to bainite. During the bainitic transformation, carbon is rejected from the forming bainite to the austenite. This carbon stabilizes the remaining austenite against martensitic transformation on cooling to room temperature. The microstructure of a multiphase TRIP steel typically contains 50% ferrite, 40% bainite and 10% retained austenite. A research effort is underway to model the heat-treatment of Figure 2, in order to calculate the amount and carbon enrichment of retained austenite and preliminary results are presented here.

Calculations involve physical modelling, as well as computational alloy thermodynamics and kinetics. The first stage involves the simulation of intercritical annealing. The software Thermo-Calc [7] and DICTRA [8] have been employed to calculate the volume fraction and composition of austenite that forms during the intercritical annealing, by solving the 1-D moving boundary diffusion problem in the two phase field of ferrite-austenite [9].

The results of that simulation can then serve as the starting point for the simulation of the bainitic transformation of austenite and the associated stabilization during the second step of the heat-treatment. The model is based on the physical characteristics of the bainite transformation and on the fact that every bainitic ferrite platelet that forms contributes to austenite stabilization by carbon rejection to the remaining austenite. The model adopts a concept, originally introduced by Bhadeshia and co-workers [10], according to which each bainitic ferrite platelet (or “sub-unit”) nucleates and instantaneously grows to a finite volume, which is controlled by the accommodation of plastic deformation in the surrounding austenite. From this mechanism it follows that once a bainitic ferrite sub-unit has formed, it begins to reject the excess carbon to the adjacent austenite layers. A key assumption here is that no carbide precipitation takes place. However, this assumption is reasonable for the case of TRIP steels, which contain Si and/or Al in excess of 1% mass [11]. In this way transient carbon profiles develop in the austenite, which tend to homogenize with time (**Figure 3**). Taking into consideration that there exists a minimum carbon content, $w_{C,min}^\alpha$,
which is necessary for the stabilization and retention of austenite after quenching, it follows that each sub-unit contributes to the retention of a potential volume of austenite, characterized by the thickness $s_{\gamma R}$ of the austenite layer. This thickness varies with time, as it depends on the position of the transient carbon concentration profile in austenite. This way the total amount of retained austenite can be calculated by: (a) considering the evolution of the population of bainite sub-units with transformation time and assuming that bainitic ferrite grows according to Johnson-Mehl-Avrami-Kolmogorov (JMAK) kinetics and (b) considering the population of “effective” sub-units, i.e. those sub-units which contribute to austenite stabilization at any given point during the transformation. Details of the model are given in reference [12].

Comparisons of experimental results (symbols) with model predictions (lines) are given in Figure 4 for a Fe-0.2C-1.5Mn-1.5Si (% mass) TRIP steel. Intercritical annealing was performed at 800°C, while bainitic transformation was performed at 350, 400 and 450°C. The retained austenite was measured by the saturation magnetization technique [13]. As shown, the model provides very good results and can be used for the design of heat-treatments, aiming at obtaining the maximum amount of retained austenite in the final microstructure of multiphase TRIP steels.

Constitutive Modelling of TRIP Steels

Based on micromechanical considerations, Papatriantafilou et al. [14] developed a constitutive model for the mechanical behaviour of multiphase TRIP steels. In particular, based on a previous work by Stringfellow et al. [15] for dual-phase steels, Papatriantafilou et al. [14] developed constitutive equations for four-phase TRIP steels, which consist of a ferritic matrix with dispersed bainite and austenite, the latter transforming gradually to martensite as the material deforms plastically. In the following a brief description of the constitutive model is reported. Details of the model can be found in reference [14].

The total deformation rate $\mathbf{D}$ can be written as the sum of an elastic, a viscoplastic and a transformational part, i.e. $\mathbf{D} = \mathbf{D}^e + \mathbf{D}^{pl} + \mathbf{D}^t$. The elastic properties of all phases can be considered as essentially the same and so the multiphase TRIP steel can be treated as a homogeneous material in the elastic region. Standard isotropic linear hypoelasticity is assumed and the constitutive equation for $\mathbf{D}^e$ is written as:

$$\mathbf{D}^e = \mathbf{M}^e: \dot{\mathbf{\sigma}},$$

$$\mathbf{M}^e = \frac{1}{2\mu} \mathbf{K} + \frac{1}{3\kappa} \mathbf{J},$$

$$\mathbf{J} = \frac{1}{3} \delta \delta,$$

$$\delta = \mathbf{1} - \mathbf{J},$$

In Eq. (10), $\dot{\mathbf{\sigma}}$ is the co-rotational stress rate, $\mu$ and $\kappa$ are the elastic shear and bulk moduli, $\delta$ and $\mathbf{I}$ are the second and symmetric fourth-order identity tensors with Cartesian components $\delta_{ij}$ (the Kronecker delta) and $I_{ijkl} = (\delta_{ik}\delta_{jl} + \delta_{il}\delta_{jk})/2$.

The steel is considered as a four-phase composite material in which the isotropic, viscoplastic phases are distributed statistically uniformly and isotropically. The constitutive equation for $\mathbf{D}^{pl}$ is then determined by using the homogenisation procedure developed by Ponte Castañeda [16], who developed a variational procedure for the determination of the effective properties of such non-linear composites. Suquet [17] showed that a certain variation of the “secant method” (the so-called “modified secant method”) is identical to the variational procedure of Ponte Castañeda. The resulting constitutive equation is of the form:

$$\mathbf{D}^{pl} = \dot{\mathbf{\varepsilon}}^{pl} (\mathbf{\bar{\sigma}}) \mathbf{N},$$

In Eq. (11), $\dot{\mathbf{\varepsilon}}^{pl}$ is the stress-rate vector and $\mathbf{N}$ is the effective stress tensor.

Figure 3. Transient C-concentration profiles in austenite adjacent to bainitic ferrite sub-units and the corresponding thickness of the stabilized austenite layer ($S_{\gamma R}$).

Figure 4. Experimental and calculated results for the evolution of retained austenite in a 0.20C-1.50Mn-1.50Si steel, as a function of temperature and holding time in the bainitic treatment stage.
The transformation part $D^{tr}$ is written in the form:

$$
D^{tr} = \hat{f} \left( AN + \frac{\Delta v}{3} \hat{d} \right)
$$

$$
A (\sigma_{eq}) = A_0 + A_1 \frac{\hat{\sigma}}{s_a}
$$

In Eq. (12), $\hat{f}$ is the volume fraction of martensite, a superscripted dot denoting time derivative, $\Delta v$ is the relative volume change associated with the transformation and takes values in the range to in austenitic steels depending upon alloy composition, $A_0, A_1$ are dimensionless constants and $s_a$ is a reference austenite stress.

The model is completed by the evolution equations of the volume fraction of the individual phases. Let $(f, c^{(a)}, c^{(3)}, c^{(4)})$ be the volume fractions of martensite, austenite, bainite, and ferrite respectively. The corresponding evolution equations are of the form:

**Martensite:**

$$
f = c^{(a)} \left( A_f \sigma^{(a)} + B_f \Sigma \right)
$$

**Ret. Austen.**

$$
c^{(a)} = \left[ 1 - \left( c^{(3)} + c^{(4)} \right) \Delta_v \right] \hat{f}
$$

**Bainite:**

$$
c^{(3)} = -c^{(3)} \Delta_v \hat{f}
$$

**Ferrite:**

$$
c^{(4)} = -c^{(4)} \Delta_v \hat{f}
$$

where $\Sigma = \sigma_b/\hat{\sigma}$ is the “triaxiality” of the stress state, $\sigma_b = \sigma_{eq}/3$ is the hydrostatic stress, and where:

$$
A_f (\sigma^{(a)}, \Sigma, T) = a C \frac{\dot{v}_m}{\dot{v}_f} (1 - f_{sb}) (f_{sb})^{-1} P
$$

$$
P (g) = \frac{1}{\sqrt{2\pi s_g}} \int_{-\infty}^{\infty} g \exp \left[ -\frac{1}{2} \left( \frac{g - \hat{g}}{s_g} \right)^2 \right] dg'
$$

$$
g (\Theta, \Sigma) = g_0 - g_1 \Theta + g_2 \Sigma
$$

In the above equations, $\dot{V}_m$ is the average volume per martensitic unit, $\dot{V}_f$ is the average volume of a shear-band intersection, parameter $\alpha$ represents the rate of shear band formation at low strains, $C$ is a geometric constant, the exponent $\gamma$ models the orientation of shear-bands ($\gamma=2$ for random orientation, $\gamma=4$ for initially parallel shear bands), $T$ is the absolute temperature, $M_d, M_f$ are the absolute characteristic temperatures for uniaxial tension. This constitutive model can be subsequently employed in order to calculate forming limit diagrams for multiphase TRIP steels.

For the calculation of a forming limit diagram, a sheet made of multiphase TRIP steel is considered, which is deformed uniformly on its plane, in such a way that the in-plane principal strain increments increase in proportion. The possibility for the formation of an instability in the form of a narrow straight band is studied (Figure 5), and the corresponding forming limit diagram is constructed. The possibility for the formation of an instability as shown in Figure 5 is examined; both inside and outside the band a state of uniform plane stress is assumed. Since the in-plane displacements are continuous, their spatial derivatives parallel to the band remain uniform.

The only discontinuities in the displacement gradient are restricted kinematically to the form [18]:

$$
\frac{\partial u_{\alpha}}{\partial x_{\beta}} = G_{\alpha\beta} N_{\beta}
$$

In Eq. (24), $X$ is the position vector of a material point in the undeformed configuration, $\left[ \right]$ denotes the difference (jump) of the field within the band and the field outside the band, $N$ is the unit vector normal to the band in the undeformed configuration (Figure 5), and $G$ is the jump in the spatial normal derivative of the displacement $u$, i.e.,

$$
G = \left[ \frac{\partial u}{\partial X} \cdot N \right].
$$

The vector $G$ takes a constant value within the band and depends on the imposed uniform deformation gradient outside the band; a method for the determination of $G$ is discussed in the following. In view of Eq. (24), the in-plane components of the uniform deformation gradient $F$ inside the band take the form:
inhomogeneity is in the form of straight narrow band (neck) formation gradient outside the band. Using the equilibrium equations across the band together with the boundary conditions, two equations are derived of the form [14]:

\[
F_{\alpha\beta}^b = F_{\alpha\beta} + G_{\alpha} N_{\beta} \quad (25)
\]

where superscript \( b \) denotes quantities within the band, and quantities with no superscript correspond to the uniform field outside the band. Using the equilibrium equations across the band together with the boundary conditions, two equations are derived of the form [14]:

\[
A_{\alpha\beta} \dot{G}_{\beta} = b_{\alpha} \quad (26)
\]

In Eq. (26), \( A_{\alpha\beta} \) and \( b_{\alpha} \) depend on the stress-state, the constitutive parameters, the orientation \( N \), and the initial thicknesses \( H^b \) and \( H^c \) inside and outside the band, respectively. The last equation defines the evolution of \( G \) as the sheet is stretched. In a perfect sheet before necking occurs, the right hand side of Eq. (26) vanishes, and the deformation remains homogeneous \( (G = 0) \) until at some stage the determinant of the coefficient matrix \( [A] \) vanishes, and this is the condition of a local necking bifurcation. In Eq. (26), Greek indices take values in the range (1,2), where \( X_1-X_2 \) is the plane of the sheet, and the “summation convention” on repeated indices is used. The approach of Marciniak and Kuzynski [19], known as the “M-K” model, in which the sheet is assumed to contain a small initial inhomogeneity (imperfect sheet) and necking results from a gradual localization of the strains at the inhomogeneity was adopted. The inhomogeneity is in the form of straight narrow band (neck) of reduced thickness \( H^t < H \) (Fig. 5). Both inside and outside the band a state of uniform plane stress is assumed, and the analysis consists in determining the uniform state of deformation inside the band that is consistent kinematically and statically with the prescribed uniform state outside the band. In the presence of an initial thickness imperfection, the right hand side of Eq. (26) does not vanish, and these equations provide a system that defines the two unknowns \( G_1 \) and \( G_2 \). Given the initial sheet thickness inside and outside the band and the imposed uniform deformation history outside the band, Eq. (26) is solved incrementally for \( \Delta G = G \Delta t \) to obtain the deformation history inside the band. Localization is said to occur when the ratio of some scalar measure of the amount of incremental straining inside the band to the corresponding value outside the band becomes very large; in particular, the calculations are terminated when either one of the two conditions \( |\Delta G_1|/\Delta \lambda_1 > 30 \) or \( |\Delta G_2|/\Delta \lambda_1 > 30 \) is satisfied. The material constants used in the calculations are reported in reference [14]. The initial volume fractions of the four phases in the TRIP steel are assumed to be \( f_0 = 0.017, c^{(0)} = 0.103, c^{(3)} = 0.38 \) and \( c^{(4)} = 0.50 \). For comparison purposes, a separate set of calculations is carried out for a non-transforming steel that consists of the three phases, i.e., retained austenite, bainite and ferrite with constant volume fractions \( f_0 = 0, c^{(0)} = 0.12, c^{(3)} = 0.38 \) and \( c^{(4)} = 0.50 \). In all cases, a constant strain rate outside the band \( \dot{\varepsilon}_t = 10^{-4} \text{ sec}^{-1} \) is imposed. Fig. 6 shows forming limit curves obtained for imposed proportional straining \( \rho \) for two different values of the initial thickness imperfection, namely \( H^t/H=0.999 \) and \( H^t/H=0.99 \). The two solid curves correspond to the TRIP steel, whereas the dashed curves are for the non-transforming steel. The TRIP phenomenon increases the necking localization strains. In particular, for an initial thickness imperfection of \( H^t/H=0.999 \) and \( \rho=0 \) (plane strain), the critical strain \( \varepsilon_1^{cr} \) increases from \( 0.2145 \) for the non-transforming steel to \( 0.2541 \) for the TRIP steel; the corresponding values of \( \varepsilon_1^{cr} \) for \( H^t/H=0.999 \) and \( \rho=0 \) are \( 0.3179 \) for the non-transforming steel and \( 0.3567 \) for the TRIP steel. A comparison of the model predictions with available experimental data is also presented in Figure 6. An “Erichsen” universal sheet metal testing machine was employed for the experiments. A hemispherical punch with a diameter of 50 mm was used and the punch velocity was set to 1 mm/sec. The agreement between the model prediction and the experimental data is reasonable.

**Conclusions**

In the present work, models regarding the stability of retained austenite, the optimisation of heat-treatment conditions in order to maximize the amount of retained austenite,
and the mechanical behaviour of low-alloy TRIP steels were presented.

The following concluding remarks summarize the issues discussed in this work:

Enhancements in cold formability of multiphase TRIP steels necessitate the optimisation of retained austenite stability.

Stability depends on chemical composition, retained austenite particle size, strength of the matrix and stress-state.

An analytical model has been developed, by which the characteristic $M_s$ temperature, i.e. the single parameter characterizing the stability of retained austenite against mechanically-induced transformation to martensite, is expressed as a function of the aforementioned multiscale parameters.

Models have been developed, involving both physical modelling and numerical simulation, which provide the ability to optimise heat-treatment conditions for multiphase TRIP steels, in order to maximize the amount of retained austenite in their final microstructure.

A micromechanical constitutive model has been developed, which takes into account the multiphase microstructure of low-alloyed TRIP steels, as well as the evolution of this microstructure during plastic deformation. This model has been employed in order to calculate forming limit diagrams for multiphase TRIP steels, an issue of great importance for the design of deep-drawing and stretch-forming operations in these steels.

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