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Effect of temperature and heat generation on martensitic phase transformation in DH steels

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ABSTRACT

Transformation induced plasticity (TRIP) phenomenon and mechanical behavior of TRIP-aided steels are influenced by the ambient as well as local temperature which raises due to plastic deformation and latent heat of transformation. In order to study this effect, tensile tests with samples made of DH800 steel were carried out at various temperatures, while the temperature rise was monitored during the test. In parallel, a physically-based model was applied to predict the change in material behavior as a consequence of the TRIP effect. In this model, martensitic transformation is mainly stress-driven, and the self-consistent scheme was adopted to achieve a homogenized material behavior. In addition, temperature rise as a result of plastic energy dissipation and latent heat of mechanically-induced retained austenite-martensite transformation was taken into account. Empirical flow curves at various temperatures indicated that at sub-zero values, the hardening of material due to the TRIP effect is much more pronounced. It was observed that a temperature change between +80 °C and -40 °C increases the flow stress by 200 MPa. Measured temperature rise, especially at sub-zero levels, showed an initial sharp increase at low strain stages due to phase transformation that flattens afterward. The numerical results are in agreement with experimental measurements of material flow curves and retained austenite fraction.

1. Introduction

Enhanced ductility dual-phase steels (DH grades) are a class of advanced high strength steels that obtain their enhanced formability over regular dual-phase (DP) steels of the same high strength by transformation induced plasticity (TRIP) effect. TRIP is defined as a mechanically induced transformation of retained austenite (RA) to martensite that leads to the enhancement of the strength and ductility of material. The DH steels retain the relatively good weldability of DP steels due to lower alloying content compared to TRIP steels. The microstructure of dual phase steels with improved formability consists mainly of ferrite, martensite and small amounts of retained austenite. The transformation kinetics is highly influenced by temperature; heating effects associated with mechanically-induced transformation can lead to a severe reduction in ductility [1].

There exist two outstanding theories which describe the mechanism of martensitic transformation. According to one theory, proposed by Olson and Cohen [2], martensite nucleates at the intersection of shear bands that are generated during plastic deformation. While the other approach by Patel and Cohen [3] and Tamura [4] correlates the transformation to the concept of driving force. Based upon this theory, transformation initiates as soon as a potential variant in one or some austenite grains receives the necessary energy to overcome the critical energy barrier and form

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a new phase. Retained austenite stability is a function of several parameters such as temperature, carbon content and alloying elements [5, 6], austenite grain size and morphology [7–10], its orientation as well as the distribution within the microstructure and hydrostatic pressure. Among the mentioned parameters, the temperature is a significant factor that shifts the onset of phase transformation. Thereby, the material’s eventual strength and hardening curve is highly affected. The effect of temperature and strain rate on dynamic properties of low silicon TRIP steels was reported in the study by Tian et al [11]. In their paper, it is stated that with raising test temperature, tensile strength reduces. In the investigation by Kim et al [12], they considered a temperature range (room temperature to 100 °C) at different stress states including uniaxial tension, uniaxial compression, equi-biaxial tension, plane strain tension and simple shear. Based on empirical observations, they presented a stress state and temperature dependent transformation kinetics law that consists of a non-linear function of the stress triaxiality, Lode angle parameter and temperature. Their model captures the measured martensitic phase transformation kinetics of TRIP780 steel reasonably well. Van der Zwaag et al. [13] investigated the transformation behavior of retained austenite in TRIP steels, in the temperature range of 5 K-300 K under a constant magnetic field, by means of in-situ X-ray measurements and thermo-magnetization. The outcomes of this study show that thermal and mechanical stabilities are correlated to the fluctuations of carbon concentration over various austenite grains. Nanga et al. [14] conducted uniaxial tension, shear and deep drawing tests on two grades of austenitic stainless steels, AISI 301LN and AISI 201, over a temperature range of −150°C to +150°C and strain rate range of $3 \times 10^{-4}$ s$^{-1}$ and 200 s$^{-1}$. The experimental results confirm a decrease in martensitic transformation with temperature rise. In the study by Ennis et al. [15], the impact of the mechanically-induced transformation of the metastable austenite phase in a TRIP-assisted dual phase steel on the material work hardening behavior was investigated. An in-situ monitoring of transformation using high energy synchrotron diffraction under uniaxial loading was employed. Kozlowska et al. [16] assessed the mechanical and thermal stability of retained austenite in bainite-based TRIP-aided medium-Mn steels using tensile tests over a wide temperature range from 20°C to 200°C. Dilatometry was employed to investigate the thermal stability of retained austenite in parallel with SEM, EBSD, X-ray diffraction and TEM techniques to study the phase composition and microstructure evolution. The experimental observations display the significance of temperature rise in martensitic phase transformation. Rana et al. [17] captured the influence of adiabatic heating during deformation of a medium-manganese TRIP steel by means of tensile tests over the temperature range from −60°C to 100°C at strain rates from 0.002 s$^{-1}$ to 0.2 s$^{-1}$. They found that increasing temperature in the range of −60°C to 100°C, increases the stability of the austenite in 10MnAl steel, causing a decrease in tensile strength due to less transformation of retained austenite to martensite during deformation. Poling et al. [18] studied the impact of temperature on retained austenite stability and tensile mechanical properties of two various grades of steel, a quenched and partitioned (Q&P) austenitic-martensitic steel, and a medium manganese ferritic-austenitic TRIP steel. Quasi-static tensile tests were conducted at various temperatures between −10°C and 85°C. The experimental outputs
of this study showed that in the medium manganese TRIP steel, austenite stability, tensile strength and elongation are highly sensitive to increasing temperature. For the Q&P steel, flow stress was almost insensitive to temperature, but austenite stability increased with temperature rise. Among the recent studies on transformation induced plasticity-aided materials, Torganchuk et al. [19] assessed the mechanical response of a medium-manganese TRIP-assisted Fe-0.2%C-10%Mn steel. The empirical results on this ferritic-austenitic material exhibit high yield and ultimate strength as well as a pronounced elongation due to the TRIP phenomenon. Hojo et al. [20] conducted a comparison of stress and strain partitioning behaviors between medium-manganese and transformation-induced plasticity-assisted bainitic ferrite (TBF) steels via synchrotron X-ray diffraction measurements. It is inferred that the high work hardening and uniform elongation in medium-manganese steel is caused by the high rates of martensitic transformation whilst the same properties in TBF steel are attributed to the sustained ability to transform up to a large strain and stress partitioning between FCC and BCC phases. Khorrami et al. [21] investigated the influence of manganese content on mechanical properties of Ni-free high N TRIP steels. At a low manganese content of 4%, high strain hardening rate and low ductility are caused by martensitic transformation occurring before plastic deformation. In contrast, the tensile properties of the high manganese content of 8% with higher austenite fraction and stability is mainly governed by twin-aided transformation. The temperature during the test is influenced by the start temperature, the plastic dissipation, the latent heat of transformation and by conduction and convection to the environment. While deforming a metal, most of the energy is dissipated in the form of heat and merely a small amount is stored [22]. The temperature rise due to dissipation is not negligible and since the TRIP effect is extremely sensitive to temperature, there is a need for a model to predict this behavior well. There are a considerable number of studies regarding the TRIP phenomenon in metastable austenitic steels. However, the suppressive impact of temperature rise during deformation on TRIP-assisted steels response is scarcely investigated; proposed models so far miss a coherent constitutive model that predicts mechanically induced martensitic phase transformation in a multi-phase microstructure at various temperatures.

In this paper, firstly, tensile tests on samples made of DH800 grade steels were conducted to inspect the material flow curves at various temperatures. A numerical model was used to estimate retained austenite stability with respect to temperature. In the computational procedure, based on stress-assisted phase transformation theory, a mean-field method is exploited to obtain the average macroscopic stress and strains as well as their partitioning among various phases. In the theory used in this paper, the transformation is modeled based upon the driving force resolved on martensitic variants residing in austenite grains with different orientations. The driving force is only correlated to the stress concentrated in austenite. As a consequence, once the stress and strain which is carried by austenite grains are determined via the mean-field model, the amount of transformed austenite is computed provided that the driving force supplied by mechanical loading is sufficient to surpass the Gibbs energy barrier. A complementary section shows the computational process by which the temperature increase caused by plastic energy dissipation and phase transformation
latent heat is computed. In this way, the initial temperature is updated for each strain increment. Additionally, the
temperature dependence of the Gibbs energy barrier is considered in this paper. In the results section, it is shown that
temperature can alter the strength of this material noticeably. It was observed that temperature change from 80 °C to
-40 °C increases the strength of this material up to 200 MPa.

In this paper, 1st order tensors and vectors are shown in bold face lowercase letters (\(a\)), 2nd order tensors in bold
face uppercase letters (\(A\)) and 4th order tensors with blackboard bold face uppercase letters (\(A\)).

2. Phase transformation

In this paper, stress-driven theory for mechanically induced martensitic transformation [3, 4] is used. In this theory,
the diffusionless transformation of austenite to martensite [23] is related to the concept of the mechanical driving force.
It is postulated in [24, 25] that the transformation includes deformation of the lattice that leaves an invariant plane
(habit plane) acting as an interface between parent austenite and martensite. The martensitic transformation consists
of a volume change and shear. Therefore, the deformation gradient of martensitic variants based on identity tensor \(I\),
shear direction \(s\) and normal to habit plane \(n\) are written as [26]

\[ F = I + s \otimes n \] (1)

When a mechanical load is applied, austenite may transform to martensite provided that the supplied driving force
is sufficient to exceed the barriers opposing the formation of a new phase. The maximum amount of mechanical driving
force resolves on an austenite grain favorably oriented with respect to the applied stress direction and can be found
using the following equation [26]

\[ U_{\text{max}} = \sum_{i=1,2,3} \sigma_i^* \lambda_i \] (2)

In Eq. (2), \(\sigma_i^*\) are principal stresses and \(\lambda_i\) are eigenvalues of the deformation tensor \(T^i\) defined by [26]

\[ T^i = \frac{1}{2}(s^i \otimes n^i + n^i \otimes s^i) \] (3)

In which \(i\) is the variant number. The values of \(\lambda_i\) are calculated using transformation shear \(\gamma\) and dilatation \(\delta\).
Accordingly, we have [26]

\[ \lambda_{1,3} = \frac{1}{2} \left( \delta \mp \sqrt{\gamma^2 + \delta^2} \right), \lambda_2 = 0 \] (4)
The amount of shear and dilatation are derived with FCC and BCC lattice parameters measured from XRD measurements [27]. In the next step, the computation of the energy barrier is elaborated according to Gibbs free energy of the phases. At a specific value called equilibrium temperature $\theta_0$, austenite and martensite have the same Gibbs energy. However, transformation does not occur at this temperature due to the obstacles opposing phase transformation (surface energy, dislocations,...). Transformation is first observed when cooling down to the martensite start temperature $M_s$. At this temperature, the difference between Gibbs energy of austenite and martensite phase, displayed as $\Delta G^{\gamma\rightarrow\alpha'}|_{M_s}$, is released during phase transformation. This amount of energy is necessary to initiate phase transformation at any temperature in the interval of $\theta_0 < \theta < M_s$. At a higher temperature $\theta > M_s$, Gibbs energy difference between austenite and martensite ($\Delta G^{\gamma\rightarrow\alpha'}|_{\theta}$) is less than $\Delta G^{\gamma\rightarrow\alpha'}|_{M_s}$. The extra amount of mechanical energy that is necessary for phase transformation and must be supplied to the material to start transformation, is computed as

$$\Delta G_{cr} = \Delta G^{\gamma\rightarrow\alpha'}|_{M_s} - \Delta G^{\gamma\rightarrow\alpha'}|_{\theta}$$  \hspace{1cm} (5)

which is called energy barrier for phase transformation. The energy barrier designated by $\Delta G_{cr}$ is generally assumed to vary linearly with temperature [27] in the interval of $\theta_0 < \theta < M_s$. Once the maximum driving force during deformation $U_{max}$ exceeds the energy barrier $\Delta G_{cr}$, phase transformation initiates. The start of transformation is expected as $g(\sigma)$ (Eq. (6)), which is defined as the yield surface for martensitic transformation, is equal to zero [28]

$$g(\sigma) = U_{max} - \Delta G_{cr} = 0$$  \hspace{1cm} (6)

It has been shown experimentally for various uniaxial and biaxial loadings in [28] that transformed martensite evolution versus mechanical driving force follows the same trend. Hence, the amount of martensite formed is expressed as a function of $U_{max}$ and $\Delta G_{cr}$. An equation for transformed martensite fraction in TRIP steels was utilized in [29] as

$$f_{\alpha'} = 1 - \left[1 + (p - 1) \left( \frac{U_{max} - \Delta G_{cr}}{q \Delta G_{cr}} \right) \right]^{\frac{1}{1-p}}$$  \hspace{1cm} (7)

In this equation, $p$, $q$ and $r$ are constants.

### 2.1. Transformation plasticity

Enhanced formability in TRIP-assisted steels owes to the extra strain which takes place during martensitic transformation entitled "transformation strain" or "free strain" [30]. According to [30], the inelastic deformation during phase transformation is primarily due to the selective transformation of the variants exposed to stress. Under stress, the net
strain happens in the direction of austenite deviatoric stress. The kinematics of phase transformation is on the base of the formulation presented in [30]. The following equation exhibits the transformation deformation rate $D_{tr}^{f}$ [27] as

$$D_{tr}^{f} = \dot{f} \left( TN' + \frac{\delta}{3} I \right) = \dot{f} \tau$$

(8)

In which $T$ is a scalar which defines the amount of shape change, $\delta$ is the volume change, $N'$ is the direction of deviatoric stress in austenite and $\tau$ is a tensor to simplify the notation.

Based upon Magee’s theory [30], the first transforming variants are the ones aligned most favorably with respect to the external stress and their contribution to transformation strain is much more noticeable than less favorably oriented ones. The latter transform as the transformation proceeds. Indeed, during transformation, the transforming variants must have a driving force equal to the critical energy barrier, which can be reformulated as

$$\Delta G_{cr} = \sigma : \tau$$

(9)

Thereby, $T$ is resolved from Eq. (8)

$$T = \frac{1}{\sigma_{eq}} \left( \Delta G_{cr} - \sigma_{h} \delta \right)$$

(10)

where $\sigma_{h}$ is hydrostatic stress and $\sigma_{eq}$ is Von Mises equivalent stress.

2.2. Mean Field homogenization and macroscopic behaviour

TRIP aided steels are composed of several phases, like ferrite, martensite, retained austenite and bainite. Consequently, strain and stress partitioning occur as a result of the mechanical contrast among constituent phases. Furthermore, the phase transformation theory used in this study is based on the stresses carried by austenite. So it is crucial to take a suitable homogenization technique that is capable of evaluating the resolved strains and stresses as well as being descriptive of the overall macroscopic material response. Homogenization of the microstructure is performed with the mean-field method that is based on the average values of the field variables in subdomains. Overall strain and stresses are related to their average values for each phase as

$$\sigma = \sum_{i} f_{i} \sigma_{i} \text{ and } \varepsilon = \sum_{i} f_{i} \varepsilon_{i}$$

(11)
where $f_i$ is the volume fraction of each phase and is summed up to 1 for all phases. It is also assumed that the macroscopic stress-strain relationship holds for the average strain and stresses of each individual phase. Therefore,

$$\dot{\sigma}_i = C^\text{ep}_i : \dot{D}_i$$  \hspace{1cm} (12)

where $\dot{D}_i$ and $C^\text{ep}_i$ are the average strain rate and elastoplastic tangent in the $i$th phase, respectively. The strain concentration tensor designated with $A_i$ accounts for the correlation between average strain in each individual phase and the overall strain rate $\dot{D}$ as

$$\dot{D}_i = A_i : \dot{D}$$  \hspace{1cm} (13)

The homogenized elastoplastic tangent of the compound is obtained as

$$C^\text{ep} = \sum_i f_i C^\text{ep}_i : A_i$$  \hspace{1cm} (14)

There are some techniques for mean field homogenization among which the self-consistent scheme results in reliable predictions for polycrystalline materials response [29]. In this method, the interaction between the matrix and the individual grains is captured using Eshelby’s equivalent inclusion theory [31]. In addition, each phase is considered as an inclusion in a homogenized matrix. The strain concentration tensor for $i$th phase is then defined as [32]

$$A_i = \left[ S : \left( C^\text{ep}^{-1} : C^\text{ep}_i - \mathbb{I} \right) + \mathbb{I} \right]^{-1}$$  \hspace{1cm} (15)

where $S$ is the fourth-order Eshelby tensor which depends on the matrix stiffness and the inclusions aspect ratio and can be found in [33]. The details of the implementation of mean-field homogenization for multiphase materials are clarified in [29]. To satisfy equilibrium in the microstructure during the transformation of an austenite grain into martensite, the stress in the martensite phase must be equal to the stress that used to be in the austenite phase [27]. Austenite and martensite are presumed to have the same elastic behavior, while martensite yields later compared to austenite. Consequently, the final stress is identical to the stress before transformation occurrence which leads to

$$\dot{\sigma}_{\alpha'} = \dot{\sigma}_{\alpha} + \frac{f_{\alpha'}}{f_{\alpha'}} (\sigma_f - \sigma_{\alpha'})$$  \hspace{1cm} (16)
3. Prediction of temperature rise

It has been observed that temperature rise can suppress martensitic transformation [12] since austenite is more stable at higher temperatures. Temperature rise may happen during deformation when the material dissipates energy when deforming plastically or due to latent heat of phase transformation. Energy dissipation rate for each phase can be derived as [11]

\[
\dot{E}_{\text{disip}}^i = \sigma_i : \mathbf{D}_{\text{pl}}^i
\] (17)

where \( \dot{E}_{\text{disip}}^i \) is the dissipation energy rate, \( \sigma_i \) is stress tensor and \( \mathbf{D}_{\text{pl}}^i \) is plastic deformation rate. During phase transformation, an amount of energy, called latent heat, is released due to the difference between material properties (heat capacity and density) of the newly formed phase and parent phase. Actually, the new phase cannot store the same amount of energy in the form of enthalpy and the extra amount of energy is released as heat.

Latent heat of phase transformation in rate form is described as

\[
\dot{E}_{\text{lat}} = \dot{f}_{\alpha'} \Delta H_{\gamma \rightarrow \alpha'}
\] (18)

In which \( \dot{E}_{\text{lat}} \) is latent heat rate, \( \dot{f}_{\alpha'} \) is rate of change of martensite fraction and \( \Delta H_{\gamma \rightarrow \alpha'} \) is enthalpy difference between austenite and martensite. To calculate enthalpy difference, we refer to the definition of enthalpy change per volume for each phase \( i \) (austenite or martensite) from a reference state to a final one as

\[
\Delta H_i^\text{final} = \int_{\text{ref}}^{\text{final}} \rho c_p \, d\theta
\] (19)

In the above equation, \( \rho \), \( c_p \) and \( \theta \) are density, specific heat capacity and temperature of the phase \( i \), respectively. Thereby, the enthalpy difference between austenite and martensite phase at the final state can be defined as

\[
\Delta H_{\gamma \rightarrow \alpha'} = \Delta H_{\gamma}^\text{final} - \Delta H_{\alpha'}^\text{final}
\] (20)

Assuming constant density and specific heat capacity with respect to temperature per each phase and using Eqs. (19) and (20), \( \Delta H_{\gamma \rightarrow \alpha'} \) is rewritten as

\[
\Delta H_{\gamma \rightarrow \alpha'} = (\rho c_p^\gamma - \rho c_p^\alpha')(\theta_{\text{final}} - \theta_{\text{ref}})
\] (21)

Considering \( \theta_{\text{ref}} \) as zero Kelvin, the enthalpy change during martensitic phase transformation can be found using
Eq. (21). The summation of energy dissipation rate over phases multiplied by $\eta$ and latent heat of phase transformation leads to the rise of temperature. In this formulation, $\eta$ is Taylor–Quinney factor which is considered to be 0.9 [34]. This factor is generally taken into account since some of the energy during plastic deformation is stored in the material and most of it is dissipated as heat. The rise of temperature can be calculated by

$$\rho c_p \theta = \eta \sum_i E_i^{\text{disip}} + E^{\text{lat}}$$  \hspace{1cm} (22)$$

In Eq. (22), $\rho$ and $c_p$ are density and specific heat capacity of the whole mixture calculated by rule of mixture [35]. The first term on the right-hand side of Eq. (22) is only activated in the plastic regime of deformation but the second term may be present in elastic or plastic regime.

A numerical algorithm is used for the solutions of the set of equations combining plasticity, homogenization and transformation [27, 29]. To solve the computational plasticity problem of each single phase, the general return mapping algorithm [36] is applied in which Newton-Raphson method [37] is employed to find the plastic multiplier. This subroutine is called at every iteration of a self-consistent scheme for homogenization. Since this scheme is nonlinear, an iterative solution is applied to find the strain concentration tensors of the phases. In addition, a phase transformation criterion is called in every iteration using the stress concentrated in the austenite phase to compute the amount of phase transformation occurred. Algorithm 1 (the box) summarizes the numerical procedure to solve the problem in steps.

**Algorithm 1:** Stress update for homogenization of a multi-phase microstructure including phase transformation

| Input: $E, \nu, b_i, c_i, \rho^i, c_p^i, f_i, \Delta G^\text{cr}, r, p, q, \lambda_1, \lambda_3$ |
| Initialize: State variables, transformed austenite fraction ($df_a^t$), total and transformation strain increment ($d\varepsilon, d\varepsilon^{tr}$) |
| while $|\text{Residual}| > \text{Tolerance}$ : do |
| Call homogenization equations Eqs. ((11)-(15)) to compute strain and stress partitioning; |
| Calculate temperature rise due to plastic dissipation (Eq. (22)); |
| if **Transformation criterion using the stress in austenite (Eq. (6))** then |
| Obtain newly formed martensite fraction (Eq. (7)); |
| Find temperature increment caused by latent heat (Eq. (22)); |
| Dilute the stress in martensite (Eq. (16)) to satisfy equilibrium after new phase appears; |
| Find transformation strain increment (Eq. (8)); |
| Calculate Residual $= d\varepsilon - d\varepsilon^{ep} - d\varepsilon^{tr}$; |
| end |
| for each phase do |
| Update state variables; |
| Update phase fractions; |
| end |
| Return total tangent and start a new increment |

**Output:** Equivalent plastic strain in each phase, resolved strains and stresses of the phases, transformed austenite fraction, final temperature;
4. Parameter identification

Hardening of each phase is modeled using a saturation type of equation

\[ \sigma_i = \sigma_{y_i} + b_i \sqrt{1 - \exp(-c_i \varepsilon)} \]  

In Eq. (23), \( \varepsilon \) is true strain, \( \sigma_{y_i} \), \( b_i \) and \( c_i \) are the yield stress and constants for each constituent phase \( i \), respectively. First, the yield stress of ferrite, martensite and austenite was considered to vary linearly with temperature with the slope of 0.67, 0.67 and 1.67, respectively. This assumption is in accordance with [38, 39]. Other hardening parameters, called \( b_i \) and \( c_i \) in Eq. (23) are extracted by fitting to experimental outputs of this study. These parameters are calibrated by means of a grid search algorithm [40]. Based on this algorithm, the optimum values of \( b_i \) and \( c_i \) are obtained based on minimizing the sum of the squares of residuals of empirical data from the numerical model outputs. In the phase transformation part, the parameters applied in the evolution of transformation, which is defined by Eq. (7), a suitable analytic approximation for martensite fraction was found [29] by fitting an Austin–Rickett type S-curve [41] to represent transformation kinetics of TRIP steel. In this equation, \( r = 3, p = 2.5, q = 0.21 \) are constant. Young’s modulus, Poisson ratio and density of the constituent phases are derived using JMatPro® software [42] for the given composition in this paper. All phase properties at room temperature are listed in Table 1. In the modeling, besides three constituent phases, ferrite, austenite and martensite, a different phase, called transformed martensite, has been considered. Transformation of austenite to martensite accompanies a dilatation, meaning that transformed martensite has a lower density than initial austenite. Dilatation happening during phase transformation is computed as 0.03 based on the density of austenite and martensite, listed in Table 1, and shear is assumed to be approximately 0.15. It should be noted that the dilatation and shear values are in a reasonable range based upon the study by Kelly [43]. Energy barrier against phase transformation \( \Delta G_{cr} \), defined in MPa, is assumed to vary linearly with temperature as

\[ \Delta G_{cr} = 0.23\theta + 43.5 \]  

In which, \( \theta \) is in degrees Celsius. In Section 6, it will be explained how this equation is found based on the experimental data.

5. Experimental test set up

Tensile tests were performed for samples made of DH800 steel at various temperatures: -40; -10; 24; 50 and 80 °C. The microstructure of the tested material in the undeformed state is illustrated in Fig. 1, an image taken by scanning electron microscope. The crosshead speed was 0.1 mm/s which gives 0.00083333s⁻¹ initial strain rate on the A80
standard sample (80 mm measurement length on 120 mm parallel length, but a gauge length of only 25 mm has been used). Finally, 1200 measurement points were obtained with a sampling frequency of 5 Hz. Temperature change during straining was recorded using a type K thermocouple with a wire diameter of 0.2 mm. In order to distinguish the chemical composition of the samples, chemical analyses were employed. The C-content has been calculated using the formula of Onink et al. [44]. The amount of retained austenite has been determined by XRD and subsequent Rietveld analysis. The XRD patterns were recorded in the range of 38° to 164° (2θ) in reflection mode using a fully automated Panalytical Xpert PRO diffractometer (COK radiation). Quantitative determination of phase proportions was also performed by Rietveld analysis. Unit cell parameters, background coefficients, preferred orientations, profile parameters and phase proportions were refined using the Bruker Topas software for Rietveld refinement. In order to improve the accuracy of austenite proportions, a BCC martensite structure was used to fit the asymmetry of ferrite reflections. Since the samples are highly textured, different models to correct for preferred orientations could lead to significantly different relative proportions of ferrite and martensite (austenite proportions are not affected). All XRD patterns were fitted using the same model for martensite and ferrite to get consistent results within the data set. Electron backscatter diffraction (EBSD) test was performed to get accurate measurements of martensite fraction as well. EBSD was performed with a Zeiss FEG-SEM Ultra55 microscope equipped with a Hikari EBSD camera from AMETEK operated with TSL software OIM DC vs7.2. The detector used for imaging is the in-lens secondary electron (in-lens SE) detector and standard SE detector. The etchant used was 1% Nital and the sample was etched for approximately 3s.

![Microstructure of DH800 captured with SEM](image)

**Figure 1**: Microstructure of DH800 captured with SEM. The letters on the image display various phases distributed over the microstructure, "A", "F" and "M" stand for Austenite, Ferrite and Martensite, respectively.
6. Results and discussion

True stress to true strain up to 12% strain for the samples at five different experimented temperatures are displayed in Fig. 2. Repeated tests at -40 °C, 24 °C and 80 °C showed virtually overlapping curves and only one test per temperature is included in Fig. 2. At -40 °C, an apparent softening is observed immediately after the elastic regime due to martensitic transformation. Although the yield stress of individual phases decreases with temperature rise, it is seen that at higher temperatures, the material yields later compared to -40 °C. Hence, the apparent softening at this temperature can merely be attributed to phase transformation in the elastic region of the hardening curve. This observation supports the stress-driven phase transformation theory, and it can be inferred that martensite does not necessarily have to nucleate at the intersection of shear bands generated by plastic deformation.

![Figure 2: Experimental true stress-true strain curve captured by tensile tests of DH800 samples over a temperature range of -40°C to 80°C.](image)

The stress-assisted phase transformation, regardless of the elastic or plastic regime of material behavior, was observed experimentally in the study by Perdahcıoğlu et al. [26] as well. At -40 °C, the stress at which transformation takes place (420 MPa based on Fig. 2) is lower than higher temperatures which indicates a lower energy barrier against phase transformation. However, with the TRIP effect, the initial softening is accompanied by a hardening which causes higher strength of the material. Based on Fig. 2, at -40°C transformation occurs early in the elastic regime. The occurrence of phase transformation leads to an extra strain, so-called "transformation strain" which adds up to the existing strains values. This phenomenon results in a shift of the flow curve at -40°C to the right. Moreover, transformation of a high fraction of retained austenite to martensite at -40°C brings about a noticeable increase in the material strength. At -10°C, almost the same behavior is observed and as the temperature increases, the flow curves settle down and it is hard to distinguish the start of martensitic transformation based on the hardening curves. Using Eq. (4) and the amount of shear and dilatation (given in Section 4), the values of $\lambda_1$ and $\lambda_3$ are found as 0.0915 and -0.0615. Moreover,
uniaxial loading condition in this paper, simplifies Eq. (2) as \( U_{\text{max}} = \sigma_1 \lambda_1 \). As it was mentioned earlier, at -40°C, the martensitic phase transformation initiates in the elastic regime of the material flow curve at a tensile stress of 420 MPa. Hence, \( U_{\text{max}} = 420 \times 0.0915 = 38\text{MPa} \). Based upon the transformation initiation criterion (Eq. (6)), once \( U_{\text{max}} = \Delta G_{\text{cr}} \), transformation starts. Therefore, \( \Delta G_{\text{cr}} \) for DH800 at -40°C is approximately 38 MPa which is very close to the value obtained by Eq. (24) by setting \( \theta \) to -40°C. The linear equation describing the dependence of energy barrier on temperature is mentioned in Section 4 (Eq. (24)). The amount of retained austenite measured experimentally at various temperatures, regarding three states: at no deformation, up to 12% strain and till fracture is depicted in Fig. 3. As Fig. 3 shows, the value of retained austenite at no deformation state is the same at -40°C, -10°C and 24°C. If \( M_s \) temperature was in this range (-40°C to 24°C), thermally activated martensitic transformation should have happened. Hence, it is concluded from Fig. 3 that \( M_s \) temperature must be lower than -40°C. Besides, the measured values of retained austenite up to 12% strain in Fig. 3, exhibit that retained austenite decreases less with increasing temperature. This observation confirms the suppressive impact of temperature rise on martensitic phase transformation. The eventual values of retained austenite at fractured state, which has been measured at -40°C, 24°C and 80°C, are lower than at 12% strain state. It is concluded that extra strain and mechanical driving force from 12% strain up to the fracture point, has led to higher mechanically induced martensitic transformation. At martensite start temperature, the energy barrier of phase transformation is zero. So using the energy barrier equation, explained in Section 4 (Eq. (24)), one can find approximate martensite start temperature, \( M_s \), as -189°C. In an experimental study by Blonde [45] for metastable austenite, no transformation was detected up to very low temperatures, which supports the validity of the aforementioned approximation for \( M_s \). As mentioned earlier, austenite stability depends on its carbon content and grain size; carbon is not distributed homogeneously over the microstructure and not all austenite grains have the same size. Thereby, remaining austenite up to room temperature is stable and it transforms provided that it is supplied with sufficient mechanical driving force.

The amount of retained austenite up to 12% strain is schematically depicted in Fig. 4. There is a good agreement between experimental and numerical results based on the comparison. \( M_d \) is defined as the temperature above which no mechanically induced martensitic phase transformation occurs [2]. This temperature is not a part of the current model or an explicit input for it. \( M_d \) temperature can be predicted as an output of the numerical model. Increasing the temperature in the model up to the point that transformed martensite is zero, \( M_d \) at 12% strain is estimated to be 100°C. This temperature is a strain and stress-state dependent value [46] and the predicted value of the current numerical model is valid for 12% strain at uniaxial stress state. Furthermore, fitting of experimentally measured RA values (Fig. 4) at 12% strain up to initial austenite fraction of 10.4% with a parabolic function \( 0.0004\theta^2 + 0.03\theta + 1.22 \) approximates \( M_d \) to be 123°C. It is concluded that the initial testing temperature range in this paper (-40°C to 80°C) is sufficiently below \( M_d \).
Figure 3: Experimentally measured retained austenite with X-ray diffraction versus initial testing temperature represented for three various cases; with no deformation, tested up to 12% strain and tested till fracture.

Figure 4: Comparison of the retained austenite values versus initial testing temperature, measured by X-ray diffraction of the samples tested up to 12% strain and computed by the numerical model.

In the next step, using parameters expressed in Section 4, stress-strain curves at measured temperatures were compared for numerical model and experimental tests in Fig. 5. Hardening parameters of constituent phases used in Eq. (23), their specific heat capacity, mass density and phase fractions at room temperature are listed in Table 1. In the modeling, besides three constituent phases, ferrite, austenite and martensite, an extra phase, called transformed martensite, has been considered. The original martensite comes from the thermo-mechanical process in steel production and has a lower carbon content than the retained austenite. Due to the assumption of diffusionless austenite-martensite transformation [23], all carbon content of austenite is transferred to newly formed martensite, which means that it is harder than initial martensite. Hence, transformed martensite was captured as a new phase which behaves differently.
Table 1
Phase properties and fractions.

<table>
<thead>
<tr>
<th>Phase</th>
<th>$E$(GPa)</th>
<th>$\nu$</th>
<th>$\sigma_y$ (MPa)</th>
<th>$b_i$</th>
<th>$c_i$</th>
<th>$c_p^i$(J/kgK)</th>
<th>$\rho_i$(kg/m$^3$)</th>
<th>Initial fraction(%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ferrite</td>
<td>207</td>
<td>0.3</td>
<td>447</td>
<td>521</td>
<td>5</td>
<td>950</td>
<td>7773</td>
<td>78.1</td>
</tr>
<tr>
<td>Austenite</td>
<td>207</td>
<td>0.3</td>
<td>607</td>
<td>321</td>
<td>6.5</td>
<td>855</td>
<td>7997</td>
<td>10.4</td>
</tr>
<tr>
<td>Martensite</td>
<td>207</td>
<td>0.3</td>
<td>958</td>
<td>830</td>
<td>15</td>
<td>640</td>
<td>7754</td>
<td>11.5</td>
</tr>
<tr>
<td>Transformed Martensite</td>
<td>207</td>
<td>0.3</td>
<td>988</td>
<td>830</td>
<td>15</td>
<td>640</td>
<td>7754</td>
<td>0</td>
</tr>
</tbody>
</table>

from the initial martensite. Most of the material properties for this phase, presented in Table 1, have been assumed the same as initial martensite except for the yield stress. Fig. 6 displays the evolution of austenite fraction derived from the numerical model, taking the temperature rise influence into account, with respect to strain. As illustrated in Fig. 6, at the higher temperature, the transformation is suppressed, and the amount of decrease in austenite fraction during deformation is less.

The temperature change was measured during tensile tests. The temperature rise is partly due to plastic deformation and partly because of latent heat, which is released while austenite transforms into martensite. As it is depicted in Fig. 7, an initial sharp change is noticed for -40 °C and -10 °C. It can be attributed to the high amount of phase transformation at these temperatures. Afterward, the trend is almost linear with respect to strain. It is observed that the slope of this linear behavior is higher at low temperatures, which is correlated to higher stress levels that cause larger amounts of plastic energy dissipation.

Temperature rise during deformation at various initial testing temperatures has been estimated for adiabatic conditions by the numerical model and displayed in Fig. 8. Based on this figure, for room testing temperature, the numerical approximation is following experimental data (Fig. 7) since there is no heat leakage from/to ambient. At other temperatures, although the trends of temperature change curves extracted from the model are the same as the outputs displayed in Fig. 7, the difference in the values is a result of heat leakage to ambient at 50 °C and 80 °C and from ambient to samples at -40 °C and -10 °C. The experiments were carried out at slow strain rate to minimise dynamic effects and the fan of the temperature chamber was switched off during testing so as not to induce noise on the strain signal due to excitation of the extensometer wiring. However, this led to significant heat loss (when test temperature above ambient temperature) during testing. This is not readily visible after the plastic work dissipation is increasing the temperature but it is visible before plastic work sets in, i.e. in the beginning of the curve. The room temperature test (24°C) was taken as a reference for calibration at which initial leakage of chamber heat to the ambient does not exist. And consequently this does not show up in Fig. 8, the model results. However, it is visible that the predictions of the model under and over estimate the temperature change at high and low temperatures, respectively.
Figure 5: Comparison of the hardening curves estimated at various temperatures (-40°C to 80°C) obtained by experimental measurements and the numerical model.

7. Conclusion

In this study, the influence of temperature on transformation of retained austenite was investigated. An experimental study was conducted, which consists of tensile tests on samples made of a kind of TRIP-aided steel called DH800 at various temperatures. A numerical model was built up, which is capable of predicting strain and stress partitioning in the microstructure as well as a check-point for martensitic phase transformation. The results of this paper can be recapitulated as

1. The influence of temperature rise caused by plastic energy dissipation and latent heat of phase transformation
Figure 6: Evolution of the retained austenite fraction during deformation at various initial testing temperatures (-40°C to 80°C) predicted by the numerical model, including the influence of temperature rise caused by plastic deformation and martensitic phase transformation.

Figure 7: Measured temperature change during tensile tests at various temperatures (-40°C to 80°C). Heat leakage to ambient plays a role in these measurements due to low strain rate.

was taken into account in this model. The experimental observations of temperature rise prove this effect to be significant.

2. In the tensile test samples at various temperatures, it is observed that at sub-zero values, austenite transforms to martensite in elastic regime of deformation. Since extra strain is accompanied by earlier transformation at -40 °C and -10 °C, the eventual strength and hardening of material is higher at these temperatures.

3. Retained austenite in the microstructure of DH800 steels is highly stable and the martensite start temperature is
Figure 8: Temperature change obtained by the numerical model assuming adiabatic conditions, plastic energy dissipation and transformation latent heat.

predicted to be -189 °C.

4. Temperature rise with respect to strain, starts with a sharp slope for the samples tested at -40 °C and -10 °C due to early phase transformation. Then the curve flattens and the temperature rise happens because of plastic energy dissipation.

5. Based on the evolution of austenite fraction during deformation, at higher temperatures, less austenite transforms to martensite that supports the suppressive influence of temperature on phase transformation.

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Highlights:

- A physical model to capture suppressive effect of temperature on martensitic phase transformation
- A model to compute the influence of transformation latent heat in phase transformation progression
- Essential influence of temperature on hardening of DH800 based on experimental observation
- Promising agreement between numerical and empirical outputs
- Investigation of transformed martensite evolution during deformation
Declaration of interests

☒ The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

☐ The authors declare the following financial interests/personal relationships which may be considered as potential competing interests: