Yielding Behavior of Nb Micro-alloyed C–Mn–Si TRIP Steel Studied by In-situ Synchrotron X-ray Diffraction

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The stress and strain partitioning between the different micro-structural constituents, and the initiation of the martensitic transformation during the yielding of multi-phase transformation-induced plasticity steel were studied by in-situ synchrotron X-ray diffraction experiments under tensile tests. Position, intensity and width of retained austenite and ferrite diffraction peaks were used to determine lattice strain and phase fractions. At low tensile stress, small elastic lattice deformations were observed. Whereas the Poisson strains in the ferrite were found to be reduced at the macroscopic yield stress, the strain in the austenite increased. The results clearly show that at the elasto–plastic transition the transformation of some of the retained austenite is stress induced.

KEY WORDS: TRIP; TWIP; IF steel; synchrotron radiation; mechanical properties.

1. Introduction

Transformation-Induced Plasticity (TRIP) steel is used for safety-related structural automotive parts due to its superior high strain rate performance, resulting in a significant dynamic energy absorption. TRIP steel typically contains ferrite, a micro-structural constituent containing bainitic ferrite and retained austenite and, in some cases, athermal martensite. The retained austenite is particularly important due to its deformation-induced martensitic transformation, which contributes to an excellent combination of strength and uniform elongation. Automotive sheet steel usually undergoes forming processes, and, in the case of TRIP steel, some retained austenite is transformed to martensite during this forming, thereby affecting the amount of retained austenite available for transformation during critical in-service situations such as automotive collisions. Therefore, it is important and necessary to understand the evolution of the TRIP microstructure during this initial deformation in order to optimize the properties of TRIP steels for specific applications.

The temperature at which the retained austenite transforms to martensite on cooling is the martensite start temperature $M_s$. If the austenite is elastically strained, the $M_s$ temperature is raised and the transformation is said to be stress-induced. When the stress exceeds the yield stress of the austenite, yielding occurs prior to transformation and the transformation is strain-induced. The change of transformation mode from stress-induced to strain-induced occurs at the $M_f$ temperature. The stability of retained austenite in TRIP steel is usually adjusted so that the in-service temperature is slightly above the $M_f$ temperature.

The stability of the retained austenite in TRIP steel depends mainly on two factors\textsuperscript{1–11}: the composition, in particular the C content, and the size of retained austenite island. The stress state, morphology and distribution of the retained austenite, the presence of precipitates and the Intrinsic Stacking Fault Energy (ISFE) of the retained austenite play a secondary role.

Samek et al.\textsuperscript{12} carried out interrupted tensile tests to evaluate the transformation rates for different types of TRIP steels. They reported that the austenite in the C–Mn–Si TRIP steel, which has a composition similar to that of the material used in the present work, exhibited the fastest transformation kinetics, i.e. 35% of the retained austenite transformed to martensite at 2.5% strain. They also reported that the $M_s$ temperature of C–Mn–Si TRIP steel was 10°C.

Furnemont et al.\textsuperscript{13} have reported intergranular stress measurements using in-situ neutron diffraction during tensile testing of TRIP-assisted multi-phase steel. They reported that the ferrite yielded before the austenite, and that the austenite carried a larger elastic strain than ferrite at a given stress level. Their results indicated that the harder austenite carried more load, while the softer ferrite sustained greater plastic strain due to the stress partitioning.

Jacques et al.\textsuperscript{14,15} used neutron diffraction to study the plastic deformation of the individual phases in a TRIP steel during uni-axial tensile testing, using two minor peaks for their analysis, (211)$_g$ and (311)$_g$. They determined the yield stress of the different phases, citing a yield strength of 720 MPa and 970 MPa for retained austenite with 1.25 mass% and 1.5 mass% carbon, respectively. Tomota et al.\textsuperscript{16} have also studied the tensile behaviour of TRIP steel by means of neutron diffraction, and their results support the observations made by Jacques et al.: they reported a 1130 MPa yield strength for retained austenite with 1.57 mass% C.

The latter value is rather high, considering that the initiation of the transformation of retained austenite does generally not require high stresses because the transformation is initiated at the same time as the plastic deformation. Meanwhile, the fact that the $M_f$ temperature is slightly below room temperature implies that the yielding of the retained austenite is initiated by dislocation glide. These seemingly contradictory facts can only be reconciled if one considers...
the presence of low stability austenite, with a larger island size and/or low carbon content, which undergoes stress-induced, rather than strain-induced, transformation. The yield strength of both bainitic ferrite and inter-critical ferrite was reported to be 695 MPa.\textsuperscript{15} This latter conclusion is questionable as these constituents differ strongly in grain size, dislocation density and C content. The authors also report that the austenite transformation is discrete, and that the autocatalytic effect is limited, a conclusion also reported previously by Samek \textit{et al}.\textsuperscript{12}

Using cryogenic cooling, Jimenez-Melero \textit{et al}.\textsuperscript{17,18} made a detailed study of the athermal martensitic transformation of individual retained austenite grains in TRIP steel using high energy 80 keV synchrotron X-rays. The mechanism of martensitic transformation they observed was therefore very much below the M\textsubscript{s} temperature, whereas the actual use of TRIP steels is above the M\textsubscript{s} temperature. Their results have revealed some important features about the retained austenite. They observed that smaller austenite grains tended to have a higher C content. Very small grains with a volume smaller than 5 \(\mu\text{m}^3\), i.e. with a grain size of about 2 \(\mu\text{m}\) in diameter were found not to transform athermally. They reported the presence of two types of austenite grains, blocky austenite and film-type austenite, confirming an earlier transmission electron microscopy (TEM) observation of Samek \textit{et al}.\textsuperscript{22} In addition, they reported the presence of retained austenite with two distinct carbon contents. Size stabilization of the retained austenite was only observed when the particle volume was less than 15 \(\mu\text{m}^3\), i.e. with a grain size of 3 \(\mu\text{m}\). Otherwise the stabilization by a higher carbon content dominated. They explained the large differences in M\textsubscript{s} temperature they observed by arguing that there was a distribution in stability from austenite grain to austenite grain.

Streicher \textit{et al}.\textsuperscript{6} and Sugimoto \textit{et al}.\textsuperscript{20,21} explained the presence of internal stresses after the unloading of strained TRIP-assisted steel on the basis of strain partitioning between soft ferrite and hard austenite. Their results showed that, in the absence of external stress, ferrite was subjected to a compressive residual stress, while austenite was subjected to a tensile residual stress. This is in agreement with the observations of Jacques \textit{et al}.\textsuperscript{14,15} who reported that ferrite had a lower flow stress than austenite and that, as a result of stress partitioning, the austenite was subject to a higher tensile stress level than ferrite.

Barbe \textit{et al}.\textsuperscript{23} reported that high C retained austenite had low stacking fault energy based on the analysis of the shift of the austenite neutron diffraction peak position changes after deformation.

In these previous reports the details of the early stages of the deformation, i.e. the elasto–plastic transition, have not been analyzed. Local elastic strains and dislocation formation associated with the martensitic transformation were rarely considered in the analysis of the diffraction data. In addition, no mention was made of elastic strains resulting from inhomogeneous plastic deformations and solid state phase transformations, although inhomogeneous inter-granular elastic strains have been observed previously for a variety of materials.\textsuperscript{23–25}

The aim of the present work was therefore to clarify the austenite and ferrite deformation behavior as well as the martensitic transformation of retained austenite in a Nb micro-alloyed TRIP steel during the elasto–plastic transition by means of tensile testing and transverse lattice strain measurements by means of \textit{in-situ} synchrotron X-ray diffraction (XRD). Synchrotron radiation provides X-rays with much higher intensity and much better coherency than conventional X-ray sources. This made it possible to measure accurately the transverse Poisson response on the scale of the grain size by high resolution \textit{in-situ} XRD during tensile deformation. The \textit{in-situ} monitoring of the changes in X-ray diffraction peak intensities, position and shape during tensile testing, provided unique information about the deformation behavior of the material as well as the microstructure evolution related to the phase transformations and the internal stress state. Relative peak intensities gave information about the different phase fractions, and provided information about the transformation kinetics of the retained austenite during deformation. Stresses affecting the interplanar spacing, \(d_{hkl}\) resulted in observable peak shifts, while peak broadening, caused by inhomogeneous elastic strain or plastic strain, i.e. displacements, or the solid state transformation of the retained austenite, were also observed and analyzed quantitatively.

2. Experimental

The materials used in the present work were a Nb-micro alloyed TRIP steel, two single phase reference materials, an austenitic high Mn Al-alloyed Twinning-Induced Plasticity (TWIP) steel and a ferritic Ti-stabilized interstitial-free (Ti-IF) steel. The composition of the materials is listed in Table 1. The yielding behavior of the Ti-IF steel and the TWIP steel were used for comparison with that of the multi-phase TRIP steel. The TRIP steel was prepared in a vacuum induction furnace. The starting ingot size prior to hot rolling was 25 mm in thickness. The blocks were hot rolled to 1.8 mm in thickness. A two step thermal treatment which included intercritical annealing and austempering, was carried out on the hot rolled steel. The intercritical annealing temperature of 864°C was chosen to obtain a phase fraction of 50% ferrite. The austempering temperature was 460°C. The steel was held 3 min at the intercritical temperature. The isothermal bainitic transformation time was 48 s.

The \textit{in-situ} tensile tests were carried out using a micro-tensile machine mounted on a high precision 6-axis goniometer installed in the 8C1 beam line of the Pohang Accelerator Light Source (PAL, South Korea). The PAL synchrotron is a third generation electron synchrotron with an energy range of 3–22 keV. The beam flux was 10\textsuperscript{14} photons/s. The focused beam size was 0.5 mm by 0.7 mm. A wavelength of 0.1789 nm was selected for the \textit{in-situ} tests. The step size and the scanning time were 0.03° and 1 s, respectively. The maximum force which could be applied to the sample was 9 800 N. The micro tensile system had a crosshead speed of 1 219 \(\mu\text{m}/\text{min}\), corresponding to a strain rate of 0.001 s\textsuperscript{-1}. The gauge length of the specimen was 20.32 mm.

The \textit{in-situ} tensile tests were performed using a step-wise force sequence. Once the specimen was strained, the micro-tensile machine was stopped maintaining the applied load while recording the diffraction data. The schematic description of the experimental orientation is shown in Fig. 1. Specimen was tilted along the X-ray beam direction by 0°,
18° and 26° by adjusting the ψ, and the diffraction data was recorded. The diffraction peaks from (200)\textsubscript{\textalpha}, (211)\textsubscript{\textalpha}, (220)\textsubscript{\textalpha}, and (311)\textsubscript{\textgamma} lattice planes parallel to the sample surface were monitored during the experiments: the austenite volume fraction and carbon content were determined using methods described in the literature.\textsuperscript{26,27} The volume fraction of the retained austenite was determined by the Miller’s empirical equation, Eq. (1):

\[ V = 1.4 \left( \frac{I_{220} + I_{311}}{2} \right) \]

\[ = \frac{I_{220} + I_{311}}{I_{211} + 1.4 \left( \frac{I_{220} + I_{311}}{2} \right)} \]

where \( I_{hkl} \) is the integrated area of the \((hkl)\) ferrite peak, and \( I_{hkl} \), the integrated intensity of the \((hkl)\) austenite peak.

In order to determine the C content of the retained austenite, the following equation was used, Eq. (2):

\[ a = (0.36306 + 0.078c) \times 1 + 24.9 - 50c \times 10^{-6}(T - 1000) \]

where \( a \) is the lattice parameter, \( c \), the C content in mass% and \( T \), the temperature, in K.

The elastic micro strain were defined as \( e(hkl) = (d_{0hkl} - d_{shkl})/d_{0hkl} \), where \( d_{0hkl} \) and \( d_{shkl} \) are the \((hkl)\) interplanar spacing in the stress-free and in the stressed condition, respectively.

X-ray diffraction peaks are broadened when the crystal lattice is imperfect. Two main factors cause peak broadening: a small particle size and the presence of non-uniform strains.\textsuperscript{28,29} Two additional effects had to be considered in the present case: the martensitic transformation and the strain and stress partitioning in the multi-phase microstructure. As the TRIP steel specimen was strained, martensite peaks appeared as a result of the strain-induced martensitic transformation. These peaks overlapped with the major ferrite peaks and resulted in ferrite peak broadening. In addition, the transformation was accompanied with a significant volume change leading to non-uniform compressive strains and local plastic deformation, i.e. a local increase of the dislocation density around the transformed volume.

3. Results

3.1. Micro-structural Analysis

Figure 2 shows the micro-structure of the TRIP steel obtained by means of scanning electron microscopy (SEM) and transmission electron microscopy (TEM). The large grains in Fig. 2(a) are ferrite. Due to the presence of Si, which suppresses cementite formation, \textit{meta}-stable austenite with a high carbon content remains in the carbide-free bainite constituent. In TEM micrographs, the bainite constituent shows alternating layers of bainitic ferrite and retained austenite islands as shown in Fig. 2(b). As the specimen is strained, this retained austenite transforms to martensite. Figure 2(c) shows partially transformed retained austenite observed in a specimen strained to fracture.

3.2. Tensile Tests and \textit{In-situ} Synchrotron Radiation X-ray Diffraction

3.2.1. Mechanical Properties

Figure 3 compares the stress–strain curves of the TRIP steel, the high Mn Al-TWIP steel and the IF steel as measured on a universal tensile testing machine.
ured on a universal tensile testing machine. The curve of the TRIP steel shows the presence of a yield point elongation prior to homogeneous strain hardening. This yield plateau is characteristic of the stress–strain curve of the TRIP steels which did not go through temper rolling. In contrast, the IF steel and the TWIP steel have a continuous yielding behavior without yield point elongation. The TWIP steel has a high strain hardening rate resulting in a high ultimate tensile strength and a large elongation.

3.2.2. Peak Intensity Change and Peak Position Shift

Typical diffraction peak profiles obtained during in-situ tensile tests are shown in Fig. 4. Changes in both intensity and position of the peaks were observed. For the (211)\(_\alpha\) peak and (311)\(_\gamma\) peak, the intensity decreased gradually as the applied stress increased, as shown in Figs. 4(a) and 4(b). The decrease of the X-ray peak intensity can be ascribed to several factors: peak broadening resulting from non-uniform elastic or plastic strain and texture evolution or crystal rotation due to slip deformation. In addition, the strain-induced martensitic transformation may result in a peak shape change in the case of TRIP steels. In Figs. 4(c) and 4(d), the integrated peak area of the ferrite and austenite is plotted as a function of the applied stress. The integrated peak area is normalized relative to the initial stress-free integrated peak area. In the case of the ferrite peak, the data ranges from 0.9 to 1.5 and most of the points are above the initial value. Throughout the whole range, no significant relation between the integrated area and the stress is observed. On the other hand, the integrated peak area of the austenite decreased when the stress exceeded 450 MPa. This is due to the martensitic transformation of the retained austenite. Using Miller’s empirical equation, Eq. (1), the retained austenite volume fraction was calculated and the results are presented in Fig. 4(e). The initial volume fraction was 7.8%, and this was reduced to about 6% when the applied stress reached 450 MPa. When this stress was exceeded, the volume fraction of retained austenite started to decrease gradually, and it reached 4% at a stress of 560 MPa.

Within the elastic strain range (~450 MPa), the center position of the peaks was shifted in the direction of larger Bragg angles. The interplanar distance \(d_{hkl}\) of the diffracting \((hkl)\) planes parallel to the sample surface became smaller as a result of the Poisson effect. When the yield stress was exceeded, the shift of the diffraction peak was reduced because plastic deformation was mainly carried by dislocation movement rather than the elastic lattice deformation.

3.2.3. Stress Dependence of the Strain in (200)\(_\alpha\)/ND Oriented Grains

Two \(\psi\) angles, (0°, 18°) were selected for the X-ray diffraction from (200)\(_\alpha\)/ND oriented grains. Figure 5(a) shows that the micro strain, \(\varepsilon_{(200)\alpha}\), for \(\psi=0^\circ\) and 18° increased linearly with increasing stress in the elastic range, i.e. the interplanar distance \(d_{(200)\alpha}\) decreased. When \(\psi=18^\circ\), the diffracting planes are tilted by 18° from the tensile direction so that the interplanar distance is less compressed than when \(\psi=0^\circ\). When the stress exceeded 440 MPa, a sharp reduction of the compression was observed. The sudden decrease of the micro strain was mainly due to a stress relaxation occurring during the elasto–plastic transition. Note that the relative peak width increased rapidly above 460 MPa, implying the initiation of plastic deformation on these grains as shown in Fig. 5(c). The present results confirm the observations made using neutron diffraction which
were reported previously by Furnemont et al.\textsuperscript{13} and Tomota et al.\textsuperscript{16,30} Furnemont et al. found that the lattice strain for (211)\textsubscript{a} lattice planes decreased and that the lattice strain for (311)\textsubscript{g} planes increased rapidly directly after macroscopic yielding. Tomota et al. reported a similar decrease for the (110)\textsubscript{a} lattice strain. However, in both cases, the (200)\textsubscript{a} strain continued to increase with increasing stress. The observed relaxation is due to the stress partitioning between hard grains and soft grains caused by inhomogeneous plastic deformation.\textsuperscript{16,30} Once the plastic deformation of softer grains starts, the stress is partitioned to the undeformed hard phases or grains which are still strained elastically.\textsuperscript{16,30} In-situ experiments which were carried out using the Ti-stabilized IF steel are shown in Fig. 5(b) for comparison. The data for the IF steel showed a micro-yielding behavior (micro-plasticity) resulting from the presence of the mobile dislocations as the stress does not contain the interstitial solute C and N atoms which usually form atmospheres pinning the dislocations (Fig. 3). In the single phase IF steel, the relaxation phenomenon is observed as well in the absence of strain partitioning effect between different phases. It is likely due to the stress partitioning resulting from non-uniform yielding of ferrite grains. Note that in case of the IF steel, the peak broadening started at low stress, i.e. prior to the stress relaxation (Fig. 5(c)). Unlike the TRIP steel, the peak width continually increased even in the elastic stage. This is very likely due to the early initiation of the micro plastic deformation in the IF steel.

3.2.4. Stress Dependence of the Strain in ⟨211⟩\textsubscript{g}/ND Oriented Grains

In Fig. 6, the micro strain, ε\textsubscript{(211)g} in TRIP steel is shown. It has a similar stress dependence as ε\textsubscript{(200)a}. In this case, the X-ray elastic modulus was larger than that of (200)\textsubscript{a}/ND oriented grains. No clear stress relaxation was observed for (211)\textsubscript{g}/ND oriented grains. A similar strain behavior was observed for the (211)\textsubscript{g}/ND oriented grains in the IF steel as shown in Figs. 6(b) and 6(c). The size of the stress relaxation at Ψ=0° was smaller when it is compared to the measurements at Ψ=18° and Ψ=26°. Due to the micro-plasticity in the elastic range, the peak broadening increased continuously with increasing stress.

3.2.5. Stress Dependence of the Strain in ⟨220⟩\textsubscript{g}/ND and ⟨311⟩\textsubscript{g}/ND Oriented Grains

In contrast to the ferrite phase, micro strains for the
and (311) planes of the retained austenite increased sharply in a compressive direction when the applied stress exceeded 460 MPa as shown in Figs. 7(a) and 8(a). Note that above 480 MPa, the maximum error bar within that range is indicated on the right side to avoid making the figures in Figs. 7(a) and 8(a) too complex. It is noteworthy that the stress relaxation observed in the ferrite initiated at 440 MPa (Fig. 5(a), Fig. 6(a)). This indicates that the ferrite grains yielded first, and that the yielding of the austenite grains took place at a higher stress. The observation of the peak width change was also in agreement with the microstrain behavior in that the onset of the peak broadening was 460 MPa for the austenite grains as shown in Fig. 7(c). Furnemont et al.\textsuperscript{13} reported that ferrite yielded before austenite, and that the austenite carried a larger stress for a given applied stress level than ferrite. Tomota et al.\textsuperscript{16,30} also reported that the austenite lattice strain increased rapidly at the macroscopic yield point, while the ferrite lattice strain decreased or increased continuously depending on the grain orientation. They suggested that the stress partitioning due to the plastic misfit strain was the main reason of the characteristic behavior in “grain to grain yielding”.

The decrease of retained austenite volume fraction as shown in Fig. 4(c) is indicative of the martensitic transformation. In addition, texture evolution also needs to be considered to account for this phenomenon, as will be shown in Chap. 4.

The micro strain and the peak width change for the TWIP steel are plotted in Figs. 7(b), 7(c) and 8(b), 8(c) for comparison. One distinctive difference between the cases of TWIP steel and TRIP steel is that the TRIP steel revealed a sharp increase of the compressive strain in the elasto-plastic transition while the TWIP steel showed a gradual compression. During the elasto-plastic transition, the following three phenomena take place simultaneously: (i) yield point elongation (ii) stress partitioning and (iii) martensitic transformation. The interaction between these phenomena will be reviewed in Chap. 4.

3.2.8. XRD Elastic Modulus

The slopes of the linear component of the curves as shown in Figs. 5–8 correspond to the elastic strain increase per unit stress, $\Delta E_{ijkl}/\Delta \sigma_{11}$. Within the elastic strain range, the XRD elastic moduli, $E_{ijkl}$ were determined using Eq. (5):

$$E_{ijkl} = -v_{ijkl} \frac{\Delta \sigma_{11}}{\Delta d_{ijkl}} d_0 = -v_{ijkl} \frac{\Delta \sigma_{11}}{\Delta E_{ijkl}} \ldots \ldots (5)$$
where \( \sigma_{11} \) is the uniaxial stress, \( \nu_{hkl} \) is Poisson ratio, \( \Delta d_{hkl} \) is the change of the interplanar spacing \( d_{hkl} \), \( d_0 \) is the initial interplanar spacing, and \( E_{hkl} \) is the strain parallel to the \((hkl)\) plane normal. For this calculation, the Poisson ratio, \( \nu_{hkl} \), was obtained from the Reuss model assuming that the stress was continuous across grain boundaries. The results for the four different grains are presented in Table 2. The calculated values are based on published data for \( C_{11}, \ C_{12} \) and \( C_{44} \) in \( \alpha\text{-Fe} \) and \( \gamma\text{-Fe} \). The results from the measurement and the Reuss model are in agreement for the \((200)\), and \((311)\) peaks. The values of \( E_{(211)} \) is however slightly higher than the value predicted by the Reuss model. The measured value of \( E_{(220)} \) is slightly lower than the calculated value of \( E_{(220)} \).

### 4. Discussion

#### 4.1. Comparison of Elastic and Plastic Behavior of Ferrite and Austenite

In Fig. 9, a schematic comparison of the observed micro strain behavior for the TRIP steel, the TWIP steel, and the IF steel is shown. In the TRIP steel, the \((200) \) \( \gamma/ND \) and \((211) \) \( \alpha/ND \) planes of ferrite have a relatively large elastic strain prior to yielding, while the \((220) \gamma/ND \) and \((311) \gamma/ND \) austenite grains show a larger plastic strain. In the elasto–plastic transition range, the stress-dependence of micro strains deviate from their initial trend abruptly. The ferrite grains undergo stress relaxation while the lattice planes of the austenite grains are more compressed. The stress relaxation of the ferrite in the TRIP steel is also observed in the single ferritic IF steel (Figs. 5(b), 6(b) and 9). The lattice compression of the austenite in the TRIP steel is similar to what is observed for the single phase austenitic TWIP steel (Figs. 7(b), 8(b) and 9). Additional diffraction data including \((111)\) and \((200)\), which are not presented in the paper because of the lack of space, show the same lattice compression behavior during the elasto–plastic transition. This observation of the lattice compression is not necessarily due to the multi-phase micro-structure of TRIP steel or the occurrence of a martensitic transformation. The different behavior of the elasto–plastic transition between ferrite and austenite can be explained by the different strain hardening mechanism of the ferrite and the austenite. Due to the high stacking fault energy, screw dislocation segments can easily cross slip in ferrite, and this results in a modest strain hardening for ferritic steels compared to the austenitic steels. In the early stage of plastic strain, ferrite grains can be strained without a large increase of the applied stress due to localized plastic deformation as well as modest strain hardening rate. In a macroscopic view, this is regarded as the yield point elongation phenomenon. In addition to this, micro-strain analysis indicates that the stress relaxation in ferrite grains takes place within the yield point elongation. In contrast, the austenite has a high rate of strain hardening at the beginning of yielding because the cross slip of the screw dislocations and the recombination of partial dislocations are more difficult when the stacking fault width is large. This is likely to lead to a more pronounced stress partitioning into the austenite micro-structure resulting in the severe compressive strain.

#### 4.2. Yielding Behavior of the TRIP Steel and the TWIP Steel

Referring to Fig. 9, the TRIP steel reveals a sharp increase of the compressive strain i.e. yielding in the elasto–plastic transition, and the TWIP steel presents a gradual and smooth lattice compression. In Sec. 4.1, it was stated the austenite lattice compression is due to the high degree of strain hardening in the austenite phase. In the engineering strain–stress curve of the TRIP steel (Fig. 3), the yield point elongation is observed approximately from 0.25 to 2%. Within this range, the externally applied stress remains almost constant, and austenite lattice compression takes place. In order to have a clear idea of what is happening in the yield plateau, the micro-strain and X-ray peak width for \((311) \gamma/ND \) austenite grains are replotted as a function of the engineering strain as shown in Figs. 10(a) and 10(b). At the early stage of the yield plateau in the TRIP steel, an abrupt increase of compressive micro-strain is observed. This coincides with the X-ray peak width increase. Exceeding 0.7%, the micro-strain does not increase anymore but remains stationary. However, in case of the TWIP steel (Figs. 10(c) and 10(d)), such a sudden lattice compression is not shown in the elasto–plastic transition. This difference is mainly ascribed to the following factors: stress partitioning in multi-phase system and martensitic transformation, which are not present in the TWIP steel. Firstly, considering micro-strain behavior of the austenite and the ferrite in Figs. 10(a) and 10(b), compressive micro-strain in the austenite is exactly accompanied by lattice relaxation in the \((200) \alpha/ND \) ferrite grains. This is the obvious evidence of stress partitioning between softer ferrite and harder austenite phase. Secondly, referring to Fig. 4(e), the martensitic transformation takes place substantially fast at the early stage of yielding, and the transformation rate becomes modest. This fast transformation also coincides with the austenite lattice compression. The

| Table 2. XRD elastic moduli. |
|-----------------------------|
| Ferrite | Austenite |
| Slope, nm/MPa | \(-4 \times 10^{-7}\) | \(-4 \times 10^{-7}\) |
| Slope, nm/MPa | \(-1 \times 10^{-7}\) | \(-1 \times 10^{-7}\) |
| Slope, nm/MPa | \(-2 \times 10^{-7}\) | \(-2 \times 10^{-7}\) |
| Reuss (200) | 0.1433 | 0.1169 |
| Reuss (200) | 0.1276 | 0.1688 |
| Reuss (311) | 0.35 | 0.25 |
| Reuss (311) | 0.25 | 0.30 |
| Reuss (220) | 0.25 | 0.30 |
| Reuss (220) | 0.25 | 0.30 |
| Elastic modulus, GPa | 125 | 189 |
| Elastic modulus, GPa | 158 | 162 |
Martensitic transformation possibly contributes to the compressive micro-strain i.e. yielding in austenite grains, and the relevant mechanism will be presented in the next section.

In Fig. 11, integrated peak area changes are plotted for the TRIP steel and the TWIP steel. In the elastic range the integrated peak area of the TWIP steel decreases with stress. At the yield point it starts to increase. In case of the TRIP steel, the integrated peak area is almost constant up to the yield point, and it starts to decrease once yielding starts. The similar observations were made when $\psi = 18^\circ$ and $\psi = 26^\circ$. The change of the integrated peak area is unlikely to reflect a substantial crystallographic texture evolution within such a small strain range. The phase transformation is therefore the main cause of the peak area changes in the case of the retained austenite in TRIP steel taking place during deformation. As for the integrated peak area change, it is difficult to tell the effect of the texture evolution from the effect of the martensitic transformation in austenite grains of the TRIP steel. Considering that for the fully austenitic TWIP steel which does not undergo phase transformation the integrated peak area slightly increases when yielding is initiated, the austenite peak area decrease in the TRIP steel is obviously related to the martensitic transformation.

4.3. The Effect of Martensitic Transformation on the Yielding Behavior of TRIP Steel

During the elasto–plastic transition, martensitic transformation needs to be taken into account in addition to the stress partitioning and yield point elongation. When martensite forms, the relative increase of the volume associated with the transformation leads to a localized plastic deformation, i.e. a local increase of the dislocation density and a localized compression. Fisher et al.\textsuperscript{34} proposed an analytical solution for the strain and stress state around a spherical phase subjected to a sudden dilatation due to

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Fig. 10. (a) The strain dependence of $\epsilon_{(311)}$ in TRIP steel. (b) The strain dependence of the relative peak width of (311) in TRIP steel. (c) The strain dependence of $\epsilon_{(311)}$ in TWIP steel. (d) The strain dependence of the relative peak width of (311) in TWIP steel. (e) The strain dependence of $\epsilon_{(200)}$ in TRIP steel.

Fig. 11. (a) Integrated peak area of the (220) peak for the TRIP and TWIP steel. (b) Integrated peak area of the (311) peak for the TRIP and TWIP steel.
transformation. According to them, the ratio of the radius of the outer boundary of the plastic zone induced around this spherical phase by a 4.2% transformation volume expansion, to the radius of the original spherical phase, is estimated to be about 3. Within this region, inhomogeneous plastic deformation is expected to take place. If it is assumed that in the present case a sphere-shaped austenite grain with a radius of 1 mm and the volume is ~4 \( \mu \text{m}^3 \), transforms to martensite, the size of the inhomogeneous plastic zone is ~109 \( \mu \text{m}^3 \).

In order to have a clear understanding of the austenite lattice contraction in the elasto–plastic transition range at 450 MPa, the austenite–martensite orientation relationships must also be considered. Figure 12(a) schematically illustrates the fact that when the austenite lattice goes through the martensitic transformation via a shear-mechanism, the ‘a’ and ‘b’ axes of the martensite lattice are lengthened and the ‘c’ axis is shortened. This lengthening results in a volume expansion parallel to \((220)\), \(g\), and the newly formed martensite phase inside a parent austenite grain compresses the residual austenite as shown in Fig. 12(b).

Due to the austenite–martensite orientation relationship, selective martensitic transformation will very likely take place during straining. The stability of the retained austenite in TRIP steel depends mainly on the C content and the size of retained austenite island. Grains with a higher C content have a higher stability.\(^1\)–\(^11\) Figureres 13(a) and 13(b) schematically explain a change in X-ray peak shape of austenite phase due to martensitic transformation. Considering two types of the austenite grains with two distinct C contents, a high C and a low C content, the austenite grain with the low C content will be less stable and more likely to transform to martensite in the initial stages of yielding. This will result in a change of the X-ray peak shape and a peak center position shifting to smaller 2\(\theta\) values (Fig. 13(b)). This change will be interpreted as an increase of the interplanar distance of the austenite grains i.e. lattice relaxation. This X-ray peak shape change was however not clearly observed in the peak shape analysis.

In summary, the observed Poisson compression behavior of the \((220)\), \((311)\), austenite grains during the elasto–plastic transition results from the interaction of three phenomena. Firstly, the stress partitioning between hard and soft austenite grains contributes to the compressive micro strain in the austenite, as was observed for both TRIP and TWIP steels. Secondly, the stress partitioning between ferrite and austenite also needs to be considered to contribute to the austenite lattice compression. In the TRIP steel, it can be explained by the observation that lattice relaxation takes place in the ferrite while the austenite grains undergo the Poisson compression. In addition to this, martensitic transformation generates an inhomogeneous plastic zone in the surrounding grains, which leads to additional strain hardening resulting in more stress partitioning. Furthermore, the austenite/martensite orientation relationship is likely to accelerate the compressive strain as explained in the previous paragraph. In contrast, the feature of the selective martensitic transformation based on the stability of the retained austenite will result in the X-ray peak shape change. The austenite grain with low C content tends to transform to martensite in the initial stage of yielding, and it leads to X-ray peak center position shift to the lower 2\(\theta\) direction, i.e. lattice relaxation.

### 4.4. Strain or Stress Induced Martensitic Transformation

At temperatures higher than \(M_s\), the strain-induced martensitic transformation requires a certain amount of strain for its initiation, as martensite is nucleated at sites where mobile dislocations interact with each other. In Fig. 14, the summary of the micro-strain behavior for the two ferrite and two austenite grains which were investigated in the present work is presented together with the retained austenite volume fraction change as a function of the applied stress. It is obvious that the yielding is initiated first in the ferrite grains. This is then followed by the yielding of austenite grains. The early yielding of the ferrite is very likely to cause the yield point elongation in the TRIP steel because of its modest strain hardening with low mobile dislocation density in the initial stage of yielding.

It appears that the martensitic transformation coincides with the yielding of the austenite. The simultaneous initiation of both the martensitic transformation and the yielding of the austenite in the present work seems to be in a good agreement with Samek’s observation. The stress-induced martensitic transformation may actually cause yielding of austenite grains as it generates mobile dislocations in its surroundings. The matter of defining the transformation...
mode as the strain-induced and the stress-induced is not straightforward. Here, the definition of the yielding is important. If it means a macroscopic yielding or the yielding of the ferrite grains, the transformation mode is obviously the strain-induced. However, in a sense that it indicates the yielding of the austenite grains, it can be considered to be either the strain-induced or the stress induced because the yielding of the austenite almost coincides with the onset of the martensitic transformation.

5. Conclusions

Tensile tests and in-situ synchrotron XRD were carried out with a C–Mn–Si TRIP steel to evaluate the deformation behavior of austenite and ferrite as well as the martensitic transformation of the retained austenite during yielding. The main conclusions of the present work are as follows:

(1) During the elasto–plastic transition, three phenomena take place simultaneously: (i) yield point elongation, (ii) stress partitioning, and (iii) martensitic transformation.

(2) The ferrite grains undergo stress relaxation i.e. yielding. Within the elasto–plastic transition, the ferrite is strained without the additionally applied stress resulting in yield point elongation because of the modest strain hardening at the initial stage of yielding.

(3) The austenite grains show an increased lattice compression i.e. yielding during the elasto–plastic transition. The reason for this is that the austenite presents a high strain hardening rate at the beginning of yielding, which is likely to partition more stress in the lattice resulting in the severe compressive strain.

(4) TRIP steel shows a sharp increase in compressive strain in the elasto–plastic transition while the TWIP steel presents a gradual increase of the lattice compression. It is because of the interaction of three phenomena: stress partitioning between grains, stress partitioning between phases, and martensitic transformation.

(5) In case of TRIP steel, the integrated peak area is almost constant up to the macroscopic yield point, and then starts to decrease once yielding starts, which implies the onset of the martensitic transformation.

(6) The martensitic transformation in TRIP steels is usually considered to be strain-induced. The present micro strain analysis strongly suggests that it can be considered to be both strain-induced and stress-induced because the yielding of the austenite coincides with the onset of the martensitic transformation.

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