**Quenching and partitioning (Q&P) processing of fully austenitic stainless steels**

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**A R T I C L E   I N F O**

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**A B S T R A C T**

A novel quenching and partitioning (Q&P) processing was applied to the steel Fe-15Cr-3Ni-3Mn-0.5Si-0.12N-0.16C (wt.-%) with an almost fully austenitic microstructure in the solution annealed condition. The Q&P processing consisted of a subzero cooling step and a partitioning step at 450 °C. The fraction of α′-martensite after the Q&P processing was nearly 58 vol%. During partitioning, M23C6 precipitates with an average size of 20 nm formed inside the α′-martensite. Concurrently, the interstitial content of austenite was increased by an average of almost 0.1 wt.-%. After Q&P processing, the stainless steel exhibited outstanding mechanical properties including a yield strength of 1050 MPa, an ultimate tensile strength of 1550 MPa, and a tensile elongation of 22% at room temperature. Reduction in the tensile test temperature from 100 °C to −40 °C facilitated the strain-induced α′-martensite formation and led to the simultaneous enhancement of strength and ductility. The enhancement of tensile ductility in spite of the strain-induced α′-martensite formation (α′-transformation-induced plasticity effect) was attributed to differences in the stability of austenite mainly caused by its non-uniform interstitial enrichment.

**1. Introduction**

In recent years, many researches have focused on the Quenching and Partitioning (Q&P) processing of martensitic low-alloy and stainless steels for the further enhancement of strength and ductility [1–5]. Q&P processing aims to generate the mechanical properties needed for the 3rd generation Advanced High Strength Steel (AHSS) applications. Such applications require ultimate tensile strength (UTS) levels above 1200/1300 MPa and total elongations (TE) in excess of 12% [6–8]. Q&P processing consists of an austenitization step followed by rapid cooling (quenching) to a temperature between martensite start (M_s) and martensite finish (M_f) temperatures known as the quench temperature. Low-alloy and stainless steels typically used for the Q&P processing have M_s temperatures well above room temperature (RT) so that direct quenching to RT results in only small volume fractions of austenite. Therefore, the desired martensite volume fraction has to be adjusted by quenching to a temperature above RT. After quenching, a partitioning step has to be performed in order to raise the stability of austenite. The partitioning step enables the diffusion of interstitial elements from supersaturated martensite to the untransformed austenite [9–11]. To delay or suppress the formation of transition carbides and M23C6 in the martensitic phase, Si is added to steels [12,13]. The enrichment of austenite by C and N and its increased stability inhibits the formation of as-quenched martensite in the final cooling step so that all of the austenite existing at the partitioning temperature is retained at RT. Because of its ductility-enhancing effect, the volume fraction of retained austenite plays a very important role in Q&P steels. The formability of the austenitic phase is controlled by the glide of dislocations which could be wavy or planar [14–16]. In particular, twinning-induced plasticity (TWIP) and transformation-induced plasticity (ε- and/or α′-TRIP) mechanisms as byproducts of glide planarity have profound effects on the formability [17–23]. To achieve high elongation levels in metastable high-alloy austenitic steels, stacking fault energies (SFEs) in the range of 15–40 mJm−2 are required [24–27]. The SFE is mainly controlled by temperature [28–31] and chemical composition [32–36]. Ideally, the interstitial enrichment of austenite during Q&P processing must increase the SFE to within the preceding SFE range.

The aim of the present study is to demonstrate how a fully austenitic metastable stainless steel can be made ultra-high...
strength by means of a novel Q&P processing consisting of subzero cooling in order to obtain a partially martensitic microstructure followed by partitioning. The main focus is placed on the microstructure formation processes during heat treatment steps and plasticity mechanisms during tensile tests at various temperatures.

2. Experimental

2.1. Ingot casting

The steel Cr15NC12.16 used in this study was produced in a VIM12 vacuum induction melting and casting facility from ALD Vacuum Technologies GmbH. An argon partial pressure of 350 mbar was applied in the melting step. Nitrogen gas with a partial pressure of 250 mbar was later introduced for nitriding. The steel was finally cast into a water-cooled copper mould placed in the furnace chamber. To avoid pore formation in the ingots, the nitrogen partial pressure was raised to 1500 mbar during casting. The chemical composition of the austenitic as-cast steel and its SFE at RT according to the empirical relationship proposed by Dai et al. [37] are given in Table 1.

2.2. Tensile tests

For tensile testing, round specimens with a gauge diameter of 6 mm and a gauge length of 30 mm were prepared according to the DIN 50125 standard. The heat-treated specimens were tensile tested in a Zwick 1476 universal testing machine. The crosshead velocity was set to 1 mm min⁻¹ corresponding to an initial strain rate of 4 × 10⁻⁴ s⁻¹. Strain was measured using a clip-on extensometer. The adjustment of different temperatures in the range of −40 °C to 100 °C was done with the aid of a thermal chamber which surrounded the tensile specimen and its fixtures.

2.3. Magnetic measurements

For the quantification of the ferromagnetic martensite, a Metis MSAT magnetization device equipped with a Lake Shore 480 model fluxmeter was used. The equipment measures the magnetization in an external magnetic field of over 300 kA m⁻¹. The ferromagnetic phase content is quantified based on the measured magnetization after corrections for the effect of alloying elements on the magnetization of pure iron. The relative measurement error is below 1%. The device used returns the average magnetization of bulk specimens rather than that at surface which might be higher due to the transformation to martensite in the prior thermal and mechanical treatment steps [38,39].

2.4. Microstructure analysis

In addition to light optical microscopy (LOM), the microstructure of the steels was studied using a Zeiss LEO-1530 GEMINI-type field emission scanning electron microscope (FESEM) at an acceleration voltage of 20 kV. The HKL Channel 5 software was used for the analysis of the electron backscatter diffraction (EBSD) data. The step size during EBSD measurements was set to 0.1–0.2 μm. Further examination of the microstructure and lattice defects was done by means of electron channelling contrast imaging (ECCI). The analysis of precipitates by Selected Area Diffraction (SAD) and Fast-Fourier Transformation (FFT) analyses of high-resolution images was done in a Jeol JEM-2200FS transmission electron microscope (TEM) operated at an acceleration voltage of 200 kV.

To quantify the enrichment of austenite by interstitials during partitioning, the lattice parameter of austenite was determined by X-ray diffraction (XRD). Measurements were done using CuKα radiation in a Seifert-FPM RD7 diffractometer. The sample preparation for XRD consisted of hot grinding to minimize the formation of surface martensite during sample preparation followed by electro-polishing.

3. Results and discussion

3.1. Design of Q&P processing

In austenitic stainless steels with M₈ temperatures equal to or slightly below RT, the martensite prerequisite for the application of Q&P processing can be introduced by further cooling. The Q&P processing of metastable austenitic stainless steels exhibiting athermal martensite formation below RT is shown schematically in Fig. 1. The formation of athermal martensite in metastable austenitic stainless steels can come to a standstill during cooling to cryogenic temperatures [40,41]. In such cases, the γ→α′ transformation remains incomplete. This characteristic behavior can be explained by a decrease in the chemical driving force ΔG°C⁻¹α′ originating from the paramagnetic to antiferromagnetic transition of austenite at Neél temperature (T₉) which in turn influences the physical and mechanical properties [40,42–44]. Consequently, the maximum martensite fraction is generated in the vicinity of T₉ without further α′-martensite formation at lower temperatures (see Fig. 1). T₉ depends on the chemical composition of the steel; whereas Mn increases T₉, other common alloying elements in stainless steels such as Cr, Ni, Si, C, and N decrease it [43,45–47]. Designing the alloy to ensure an M₈ in the vicinity of RT and a T₉ far lower than RT allows to generate controlled austenite and martensite fractions in the microstructure by quenching to cryogenic temperatures. These characteristics simplify the quenching step in comparison to low-alloy and conventional martensitic stainless steels since the initial quench temperature (prior to cryogenic cooling) could be set equal to RT. Moreover, the partitioning step does not have to be carried out immediately after quenching because the remaining austenite is thermally stable at RT.

In austenitic-martensitic stainless steels, an accelerated precipitation of carbidies is expected inside the α′-martensite phase. It is well known that tempering at temperatures up to about 450 °C only leads to the formation of paraequilibrium M₆C-type carbides (M denotes Fe and substitutional alloying elements) [12,48,49]. During the Q&P processing, the formation of carbidies in the partitioning step has to be prevented since it can lead to an insufficient interstitial enrichment of austenite and favor its further transformation to α′-martensite during the final cooling to RT. In view of the strength, however, the formation of nano-sized carbidies could be beneficial as it increases the yield strength [38]. To ensure a good ductility in multiphase steels consisting of austenite, α′-martensite, and carbidies, the austenite fraction should be elevated in comparison to conventional Q&P-steels. In this study, by means of a

<table>
<thead>
<tr>
<th>Alloy</th>
<th>C</th>
<th>N</th>
<th>Cr</th>
<th>Ni</th>
<th>Mn</th>
<th>Si</th>
<th>Fe + others</th>
<th>SFE at RT [mJ m⁻²]</th>
<th>Ref.</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cr15NC12.16</td>
<td>0.155</td>
<td>0.122</td>
<td>14.90</td>
<td>2.91</td>
<td>2.97</td>
<td>0.53</td>
<td>bal.</td>
<td>18</td>
<td>[37]</td>
</tr>
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</table>

Table 1 Chemical composition in wt.-% and the calculated SFE value at RT.
processing route similar to Fig. 1, a martensite fraction of approximately 58 vol% was formed by cryogenic treatment in liquid nitrogen. This fraction is almost equal to the \( \alpha' \)-martensite fraction obtained by cooling to \(-130°C\) [41]. Therefore, the spontaneous \( \alpha' \)-martensite formation, which comprises both isothermal and athermal types of martensite, comes to a standstill at temperatures below \(-130°C\) which is close to the \( T_N \) of the present alloy [41] as calculated by Thermo-Calc [41,50,51]. Therefore, transition to the antiferromagnetic state of austenite below \( T_N \) results in a decreased chemical driving force for the \( \gamma \rightarrow \alpha' \) transformation [40].

3.2. Microstructure after solution annealing

The tensile specimens were solution annealed at 1150 °C for 30 min under argon atmosphere followed by water quenching. The solution annealing step resulted in the dissolution of precipitates and reduced the chemical inhomogeneities [41]. A large amount of \( \alpha' \)-martensite was observed during optical microscopy examinations (Fig. 2). However, a ferromagnetic phase fraction of only 5.5 vol.% was quantified by bulk magnetic measurements implying that the majority of martensite in Fig. 2 has been generated during metallographic sample preparation. Moreover, the inhomogeneity of chemical composition is reflected in the non-uniform distribution of phases which highlights the as-cast dendritic microstructure. According to Fig. 2, whereas the dendritic regions have transformed to \( \alpha' \)-martensite during specimen preparation, the interdendritic regions are still austenitic.

3.3. Introduction of \( \alpha' \)-Martensite in the austenitic steel matrix

To enhance the strength, \( \alpha' \)-martensite has to be introduced in the austenitic steel matrix. To achieve this, the formation of spontaneous \( \alpha' \)-martensite after solution annealing was studied. For this purpose, solution annealed specimens were quenched in liquid nitrogen for 10 min. After this treatment, an as-quenched \( \alpha' \)-martensite fraction of 58 vol.% was quantified by magnetic measurements. Fig. 3 shows the martensitic-austenitic microstructure obtained by liquid nitrogen treatment.

![Fig. 1. Schematic representation of the thermal processing of metastable austenitic stainless steels in order to turn them into ultra-high strength steels consisting of austenite, martensite, and carbides embedded in the martensite. The abbreviation “p” stands for precipitates in the microstructure of the steel.](image)

![Fig. 2. Light optical micrograph in the solution annealed condition with an austenitic-martensitic microstructure. In this case, \( \alpha' \)-martensite is mainly induced by sample preparation.](image)

![Fig. 3. SEM backscattered electron micrograph after solution annealing and liquid nitrogen treatment. \( \alpha' \)-martensite and austenite are present in dendritic and interdendritic regions, respectively.](image)
In agreement with the observations made for the solution annealed condition, the as-quenched \( \alpha' \)-martensite has preferably formed in the dendrites and the interdendritic areas with a higher stability are still austenitic. The presence of austenite in interdendritic regions clearly reveals the cast structure (Fig. 3). With the aid of energy-dispersive X-ray spectroscopy (EDX) analysis in SEM, the distribution of alloying elements in a liquid nitrogen treated specimen was determined. According to the maps in Fig. 4, the interdendritic regions are enriched with Cr, Mn, and Ni. As already known for the strain-induced martensitic transformation, the nucleation sites for the \( \alpha' \)-martensite formation are the intersections of slip bands containing stacking faults [52–55]. In the case of athermal martensite, the transformation can proceed by a similar process involving the spontaneous dissociation of perfect dislocations into Shockley partials. The interactions among spontaneously formed stacking faults can in turn trigger the formation of athermal martensite \( (\gamma \rightarrow SF \rightarrow \alpha') \). This speculation is in agreement with the two-step \( \gamma \rightarrow \epsilon \rightarrow \alpha' \) models proposed for the martensitic transformation [56,57].

The EBSD analysis of the studied steel after quenching

Fig. 4. a) SEM secondary electron micrograph after solution annealing and liquid nitrogen treatment; b-d) corresponding EDX maps showing the distribution of Cr, Mn, and Ni in the dendritic regions with a martensitic microstructure and interdendritic regions with an austenitic microstructure.

Fig. 5. EBSD analysis after solution annealing and treatment in liquid nitrogen: a) phase map with austenite in red and \( \alpha' \)-martensite in blue (black: non-indexed); b) IPF map of fcc phase; c) IPF map of bcc phase showing various crystallographic variants of \( \alpha' \)-martensite; d) bcc pole figures corresponding to the region shown in c and color-coded according to c. In b and c, colors indicate crystal directions normal to the plane of view. (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)
to –196 °C is presented in Fig. 5a. Only as-quenched α'-martensite (blue) and austenite (red) were identified by EBSD. Fig. 5b and c shows the inverse pole figure (IPF) maps of austenite and α'-martensite, respectively. The IPF map of austenite confirms a bimodal size distribution for austenite, namely coarse interdendritic regions and fine regions dispersed in the martensite, namely interlath austenite. The IPF map of Fig. 5c indicates the multiplicity of crystallographic variants of α'-martensite. The associated grain refinement is one of the mechanisms by which martensite contributes to the strength. In the bcc pole figures of Fig. 5d, all crystallographic variants of Pitsch, Kurdjumov-Sachs (K-S), and Nishiyama-Wasserman (N-W) orientation relationships between austenite and martensite [58] are represented. This indicates that variant selection has not taken place during the martensitic transformation by quenching. The absence of variant selection during athermal martensite formation is in agreement with observations made for other steel systems [59–62]. During the deformation-induced martensitic transformation, on the other hand, variant selection is imposed by external stresses [63–65].

The crystallographic orientation of martensite variants in the pole figures of Fig. 5d resembles the observations made for a number of martensitic steels [56]. Such characteristic pole figures have been proposed to originate from the plastic deformation of the surrounding austenite caused by the growth of martensite originally nucleated with a K-S type orientation relationship [66]. Although martensite maintains this original orientation relationship with respect to the deformed austenite at all stages of growth, its orientation relationship with respect to the remote regions of untransformed austenite will gradually deviate towards N-W and Pitsch relationships.

3.4. Partitioning

The austenitic-martensitic microstructure obtained by liquid nitrogen treatment was subsequently heat treated at 450 °C for 3 min. The main goal was to enable the diffusion of C and N from the supersaturated α'-martensite to the austenite, thereby increasing the stability of austenite [71,10,67,68]. The associated decrease in the solute C and N concentration of martensite reduces the tetragonality and hardness of the α'-martensite [69–71]. The inevitable formation of carbides and nitrides in the α'-martensite can serve as a precipitation strengthening mechanism for the α'-martensite. In steels conventionally used for Q&P processing, the Ms temperature is much higher than RT and the partitioning step must be done immediately after the quenching step [38]. In the present case, however, quench temperature is below RT and the partitioning step can be done any time afterwards.

The holding time of 3 min at 450 °C enables the long-range diffusion of the interstitial elements C and N in ferrite and austenite [68]. The substitutional alloying elements (Cr, Mn, and Ni), on the other hand, could be assumed immobile within the timeframe of partitioning [72]. According to the diffusivity equation proposed by Ågren [73], the diffusion distance of C in the austenite phase of the present steel (450°C/3 min) is estimated to about 693 nm. Considering the diffusion coefficients at 450 °C of $D_C^T = 1.46 \times 10^{-11}$ cm²s⁻¹ and $D_N^T = 5.21 \times 10^{-11}$ cm²s⁻¹ [74] for C and N diffusion in austenite, respectively, the stabilization of austenite might be more pronounced by N. Given the likelihood of precipitation reactions in the martensitic phase, the amount of interstitials available for the thermal stabilization of austenite is lower than that predicted by the full partitioning assumption. For instance, fine M3C-type precipitates have been reported during tempering of α'-martensite between 400 °C and 450 °C [12,38,48,49,68]. Yuan et al. [49] have identified lenticular M3C precipitates with an average length of 70 nm and an average width of 5 nm after partitioning at 400 °C of an Fe-13.6Cr-0.44C stainless steel for 1 min. In the selected area diffraction (SAD) pattern of Fig. 6a, diffraction spots pertaining to an orthorhombic M3C type carbide (space group Pnma) and the surrounding martensite are indexed. The HRTEM image of another M3C carbide with a size of nearly 20 nm and its corresponding FFT image are shown in Fig. 6b.

The enrichment of carbon and nitrogen in the remaining austenite was investigated by XRD measurements to estimate the lattice parameter of austenite before and after partitioning. For this purpose, a specimen was solution annealed at 1150 °C, water quenched to RT, and further cooled in liquid nitrogen (Q condition). A second specimen was treated to the Q condition followed by partitioning at 450 °C for 3 min (Q&P condition). In Fig. 7, the high-angle (311), diffraction peak profiles before and after Rietveld refinement by MAUD program [75] are shown for the Q and Q&P conditions. Due to the partitioning of interstitials, the (311), peak position was shifted to lower angles indicating an increase in the lattice parameter of austenite [72]. The (311), peak after partitioning was not symmetric and the raw intensity to the left of the peak maximum was clearly higher than the fitted peak. This could be related to the non-uniform interstitial enrichment of austenite arising from differences in the size and distribution of austenitic regions. In this regard, fine austenitic regions surrounded by a locally high fraction of martensite (interlath austenite) will be enriched to a larger extent than coarse interdendritic regions of austenite surrounded by small martensite fractions.

The lattice parameter of austenite increased from

![Fig. 6. TEM analysis of precipitates in the partitioned condition: a) SAD pattern of α'-martensite and an embedded M3C-type carbide with [201]C and [172]_M3C zone axes, respectively; b) HRTEM image of an M3C-type carbide and the corresponding FFT image. Zone axis for the precipitate in b is [2 1 0]_M3C.](image-url)
0.35923 ± 0.00003 nm in the Q condition to 0.36000 ± 0.00003 nm in the Q&P condition. According to [72,76], the lattice parameter of austenite in stainless steels increases by 0.00045 nm per 1 at.% solute carbon. Assuming a similar lattice dilatation effect for solute nitrogen, the observed increase in the lattice parameter of austenite can be correlated with an increase of 0.1 wt.% in the solute interstitial amount of austenite. Accordingly, the average interstitial content of austenite is estimated to have increased from 0.28 wt.% before partitioning to 0.38 wt.% after partitioning.

### 3.5. Mechanical properties under uniaxial tensile loading

In Fig. 8(a,d), stress-strain curves under uniaxial tensile loading are presented for the Q&P and solution annealed (SA) conditions. The presence of tempered α'-martensite (~58 vol.-%) in the Q&P-processed condition resulted in a pronounced increase of yield strength (YS) compared to the SA condition. Accordingly, a YS of 1050 MPa, an ultimate tensile strength (UTS) of 1550 MPa, and a total elongation (TE) of 22% were achieved in the Q&P condition at RT. According to Fig. 8a, strength and elongation enhance concurrently at the lower tensile test temperature of -40 °C. Phase fractions after tensile testing of Q&P specimens were determined by bulk magnetic measurements (Fig. 8b). At 100 °C, the formation of strain-induced α'-martensite in the austenite was almost fully suppressed. Due to the absence of α'-martensite formation in the grip section of tensile tested specimens, where the stress level might have been up to the elastic limit, the formation of stress-induced α'-martensite in Q&P specimens can be excluded.

Given the significantly lower strength of the austenitic constituent in the Q&P condition, it is expected to carry a noticeably higher strain than the coexisting α'-martensite. Therefore, the plastic deformation at 100 °C has been primarily accommodated by the glide of dislocations in the austenitic phase which explains the relatively low rate of strain hardening. The decrease in the SFE of...
austenite at lower temperatures facilitates the dissociation of perfect dislocations \((\mathbf{b} = \frac{a}{2}[10\bar{1}])\) into Shockley partials (leading: \(\mathbf{b}_1 = \frac{a}{8}[11\bar{2}]\); trailing: \(\mathbf{b}_2 = \frac{a}{8}[2\bar{1}T]\)) [77]. As long as the \(\alpha'\)-martensite formation is suppressed, the increase in the glide planarity of austenite at lower temperatures results in an enhanced ductility [78].

In the present case, the strain-induced \(\alpha'\)-martensite fraction after deformation until fracture increased from 1.5 vol-% at 100 °C to 25 vol-% at −40 °C. The formation of \(\alpha'\)-martensite platelets inside the austenite glide bands or at their intersections reduces the mean free path of dislocations thereby increasing the strength [52]. As a result, the difference in the strain hardening rates at various temperatures is mainly controlled by the \(\alpha'\)-martensite formation rate during tensile deformation. At −40 °C, the steel exhibited the highest TE and UTS values (Fig. 8c). The enhancement of ductility upon the formation of \(\alpha'\)-martensite implies the occurrence of the \(\alpha'\)-TRIP effect. In fully austenitic steels with a uniform distribution of alloying elements, the \(\alpha'\)-martensite formation has a detrimental effect on the tensile ductility [79–81]. The enhancement of ductility in a fully austenitic stainless steel has been attributed to the segregation of alloying elements and the presence of microstructural regions with different stabilities with respect to the strain-induced \(\alpha'\)-martensite formation [81]. In the present case too, the observation of an \(\alpha'\)-TRIP effect might be attributed to inhomogeneities in the mechanical stability of austenitic regions. Such inhomogeneities can arise from differences in the chemical composition and size of austenitic regions. The chemical inhomogeneities could originate from the solidification step as discussed in 3.3 and the partitioning step as discussed in 3.4.

The engineering stress–strain curves in the SA condition are shown in Fig. 8d. Because of the low stability of austenite in the SA condition at RT and −40 °C, the \(\alpha'\)-martensite formed at a high rate during plastic deformation. As a result, high strain hardening rates and low elongations were observed. At RT, a YS of 270 MPa, a UTS of 1300 MPa, and a TE of only 9% were obtained. The total \(\alpha'\)-martensite phase fraction after tensile testing at RT was 81 vol-%. The rate of \(\alpha'\)-martensite formation and the strain hardening rate increased with decreasing temperature. The occurrence of stress-induced \(\alpha'\)-martensite and the fast kinetics of strain-induced \(\alpha'\)-martensite formation in the SA condition are the main reasons for the poor TEs.

To evaluate the mechanical stability of austenite against the deformation-induced martensite formation, an equation proposed by Sugimoto et al. [82] was used:

\[
k = -\ln\left(\frac{f_0^G - f_0^q}{f_0^G}\right) e^{-1}
\]

In Equation (1), \(k\) is the stability parameter of austenite, \(f_0^G\) is the austenite fraction prior to deformation, \(f_0^q\) denotes the deformation-induced \(\alpha'\)-martensite fraction, and \(e\) is the applied true tensile strain. The parameter \(k\) increases with decreasing austenite stability. In Fig. 8e the temperature dependent \(k\) value is shown for the SA and Q&P conditions. Clearly, the mechanical stability of austenite decreases with decreasing temperature. Fig. 8e further shows that the deformation-induced \(\alpha'\)-martensite formation in austenite is significantly retarded by Q&P processing. Due to the enrichment of austenite by interstitial elements during partitioning, the \(k\) parameter at −40 °C in the Q&P condition is comparable to that at 100 °C in the SA condition. It must be noted that in order to calculate \(k\) parameters using Equation (1), the applied global strain was used. Due to the strain partitioning between austenite and martensite in the Q&P condition, the actual strain accommodated by the austenite must have been higher than the global strain. Accordingly, the \(k\) parameters in the Q&P condition are somewhat overestimated.

3.6. Microstructural examinations after tensile straining

Due to the small size of austenitic regions between the tempered lath martensite, their C and N enrichment during partitioning is more pronounced than in the coarse interdendritic regions of austenite. Moreover, the fine size of interlath austenite adds to its stability with respect to the strain-induced \(\alpha'\)-martensite formation [83–85] although contrary results exist too [86]. Limited straining of austenite in the interlath regions arising from the shielding effect of martensite could be another mechanism for the apparently higher mechanical stability of austenite in such regions [87]. Accordingly, the interdendritic regions of austenite are more susceptible to the \(\alpha'\)-martensite formation than the small regions of interlath austenite in the dendrite cores. The microstructure of the Q&P steel strained until fracture at RT confirms the occurrence of strain-induced \(\alpha'\)-martensite within glide bands in an interdendritic region of austenite (Fig. 9a). Due to the lower driving force for the strain-induced \(\alpha'\)-martensite at higher temperatures, glide bands in the interdendritic regions of austenite in the specimen tensile tested at 60 °C contained less martensite (Fig. 9b).

EBSD phase map and IPF maps of bcc and fcc phases in the Q&P specimen tensile tested at RT are shown in Fig. 10. The regions within the marked ellipses in Fig. 10b are likely interdendritic austenite regions prior to tensile deformation which now additionally contain strain-induced \(\alpha'\)-martensite. The variants of \(\alpha'\)-martensite in these regions are almost parallel to the direction of maximum shear stress at approximately 45° to the prior tensile loading direction (vertical in the plane of view). Pronounced variant selection and directionality of martensite platelets facilitate the

![Fig. 9. ECCI images showing the microstructure of the Q&P steel after tensile deformation until fracture at a) RT and b) 60 °C. The tensile direction is horizontal in the plane of view. The black arrows mark strain-induced \(\alpha'\)-martensite platelets in the slip bands of interdendritic austenite.](image-url)
discrimination of strain-induced $\alpha'$-martensite from pre-existing as-quenched $\alpha'$-martensite. The distribution of austenite in the IPF map of fcc phase in Fig. 10c supports that the regions marked by ellipses in Fig. 10b are indeed interdendritic austenite. The coexistence of $\varepsilon$-martensite next to $\alpha'$-martensite within the slip bands of Fig. 10a suggests that the $\alpha'$-martensite has formed according to the sequence $\gamma \rightarrow \varepsilon \rightarrow \alpha'$.

4. Conclusions

A novel Q&P processing was applied to the steel Fe-15Cr-3Ni-3Mn-0.5Si-0.12N-0.16C (wt.-%) with an almost fully austenitic microstructure in the solution annealed condition. The Q&P processing consisted of subzero cooling in liquid nitrogen and partitioning at 450 °C for 3 min. The main conclusions are as follows:

1. The fraction of athermal $\alpha'$-martensite in the quenching step reached a maximum of approximately 58 vol% near the Néel temperature of −130 °C and remained constant at lower temperatures. Athermal $\alpha'$-martensite preferentially occurred in the dendritic regions of austenite which were depleted of Cr, Mn, and Ni in the solidification step.
2. During partitioning at 450 °C, M$_2$C precipitates with an average size of 20 nm formed inside the $\alpha'$-martensite. The presence of tempered martensite in the Q&P condition significantly enhanced the strength compared to the solution annealed condition.
3. The XRD-based lattice parameter of austenite increased from 0.35923 ± 0.00003 nm prior to partitioning to 0.36000 ± 0.00003 nm after partitioning. This lattice dilatation can be correlated with an average interstitial enrichment in the austenite of 0.1 wt.-%. The enrichment is expected to have been more effective for the fine interlath austenite than the coarse interdendritic austenite.
4. After Q&P processing, the steel exhibited outstanding mechanical properties including a YS of 1050 MPa, a UTS of 1550 MPa, and a TE of 22% at RT. These properties are required for the 3rd generation AHSS applications.
5. Reduction in the tensile test temperature from 100 °C to 40 °C facilitated the strain-induced $\alpha'$-martensite formation and led to the simultaneous enhancement of strength and ductility. The enhancement of tensile ductility in spite of the strain-induced $\alpha'$-martensite formation ($\alpha'$-TRIP effect) was attributed to differences in the stability of austenite mainly caused by its non-uniform interstitial enrichment.

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