Microstructure-Mechanical Property Relationship and Austenite Stability in Medium-Mn TRIP steels: The Effect of Austenite-Reverted Transformation and Quenching-Tempering Treatments

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Abstract

In the present study, we fundamentally explore the reasons underlying differences in mechanical properties in hot-rolled 0.2C-1.6Al-6.1Mn-Fe TRIP steels subjected to different heat treatments. Comparing with austenite reverted transformation annealing (ART) process, quenching and tempering (Q&T) process was effective in obtain excellent mechanical properties. [Q&T (UTS: 902-1235 MPa, TEL: 18-42%]; [ART (UTS: 885-945 MPa, TEL: 13-28%)]. In the ART process, long time annealing led to excessive C and Mn enrichment in austenite, which rendered austenite too stable and deteriorated TRIP effect. Furthermore, long time annealing reduced dislocation density and led to low work-hardening rate. The Q&T process enabled appropriate enrichment of elements and hence desired stability for significant TRIP effect to be observed. Thus, the steel quenched from 625°C exhibited best combination of mechanical properties (UTS: 1038 MPa, TEL: 42%, UTS×TEL: 43.6 GPa%) because of significant contribution of TRIP effect and high dislocation density in austenite.

Keywords: ART; Q&T; mechanical properties; austenite stability; TRIP effect

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1. Introduction

Transformation-induced plasticity (TRIP) light-weight steels are considered as a new class of high strength steels with exceptional formability [1-3]. Cold-rolled TRIP steels are potential candidates for automotive applications. Recently, medium Mn steels (3-12%) have received significant attention as third-generation high-strength steels [4-6] because of their outstanding combination of strength and ductility and high strain hardening ability. The addition of 1-3 wt.% Ni and Al promotes precipitation hardening in the martensite matrix [7-8]. The minor addition of C (up to ~0.2 wt.%) enhances the kinetics of austenite reversion and austenite stability, and increases the strength of the alloy [9].

A number of studies have been carried out on Fe-Mn-Al-C (wt.%) alloy system subjected to different heat treatments. For example, in Fe-7Mn (wt.%) cold-rolled sheet, tensile strength of 800-1000 MPa and total elongation of 19-28% was obtained using the austenite-reverted transformation annealing (ART) heat treatment [10]. While Fe-9Mn (wt.%) cold-rolled sheet exhibited tensile strength of 950-1200 MPa and ductility of 20-25% using ART [11-13].

Retained austenite stability plays a critical role in influencing superior mechanical properties of advanced high-strength steels involving transformation-induced plasticity (TRIP) effect. The present study is aimed at fundamentally exploring the microstructure, retained austenite stability and mechanical properties using different heat treatments, in the endeavor to understand the effect of heat treatment on the microstructural evolution and mechanical properties

of medium-Mn TRIP steels.

2. Experimental

The nominal chemical composition of experimental steel was Fe-6.1Mn-1.6Al-0.2C (wt. %). The ingots were cast using a vacuum furnace and heated at 1200°C for 2 h and hot forged to rods of section size 100 mm×30 mm, followed by air cooling to room temperature. Subsequently, the rods were soaked at 1200°C for 5 h, and hot rolled to 4 mm thickness via 5 passes in the temperature range of 1150-850°C, and finally air cooled to ambient temperature (Fig. 1).

The experimental steels were subjected to austenite-reverted transformation annealing (ART) process and quenching and tempering (Q&T) process, respectively. Tensile specimens of width ~12.5 mm and a gage length of 50 mm were machined from the heat-treated sheets with the tensile axis parallel to the rolling direction. Tensile tests were carried out at room temperature using a universal testing machine (SANSCMT5000) at a constant cross-head speed of 3 mm·min⁻¹. The specification for the tensile specimen used was GB/T 228-2002, which is equivalent to ISO 6892:1998. Microhardness measurement was carried out using Vickers hardness tester using 50 gf and hold time of 10 sec. The samples were etched with 25% sodium bisulfite solution. Microstructural studies were carried out using a combination of scanning electron microscope (SEM), electron microprobe analyzer (EMPA) and transmission electron microscope (TEM). TEM and EDS using a nanoscale electron beam were used to investigate the partitioning of C and Mn in ferrite and retained austenite. The volume fraction of austenite was determined by X-ray diffraction

(XRD) with CuK_α radiation using direct comparison method [14].

In order to establish the appropriate heat treatment schedule, the intercritical temperature range was determined by dilatometry. The sample used for dilatometry was solid cylindrical specimen with a length of 10 mm and a diameter of 3 mm.

3 Results

3.1. Schedule of Heat Treatments

The dilatometric curve of the experimental steel is presented in Fig. 2. After thermal expansion during the heating stage (20–900°C) at the rate of 20°C/s, the sample was held at 900°C for 3 min. It is clear that no transformation took place in the steel during fast cooling, at the rate of 100°C/s, until the martensite start (M_s) temperature. As marked in Fig. 2, the intercritical temperature range of the experimental steel was 612–760°C.

Two different heat treatment processes were used to process the experimental steel (Fig. 3). For austenite reverted transformation annealing (ART) heat treatment, the steel sheet was water-quenched after austenitization in the range of 800–900°C for 1 h and then intercritically annealed at 650°C for up to 12 h and finally air cooled to room temperature (Fig. 3a). For quenching and tempering (Q&T) heat treatment (Fig. 3b), first, they were annealed in the two-phase region (600–750°C) for 1 h, followed by water quenching. Second, the quenched samples were tempered at 200°C for 20 min and air cooled to ambient temperature. Tempering effect has been previously studied by us [15]. Tempering was helpful in relieving the internal stresses, but more importantly it is expected to improve the stability of austenite through the diffusion of

C from ferrite to austenite.

3.2. Microstructural Evolution

SEM micrographs of experimental steel after ART annealing in the austenite region at different temperatures for 1 h and water quenching and then intercritically annealing at 650°C for 12 h are presented in Fig. 4. Fig. 4a describes the microstructure of the steel after austenitization at 800°C (labeled as AR-800). The microstructural constituents were primarily acicular austenite (A) and ferrite (F) together with a small amount of martensite (M). This martensite is residual martensite after water quenching and was not transformed to austenite during intercritical annealing. The combination of acicular austenite and martensite is identified with a box in Fig. 4a. The Vickers microhardness of constituent phases is presented in Table 1. The microhardness of martensite was higher than austenite and ferrite. When the steels were austenitized at 850°C and 900°C (AR-850 and AR-900), respectively, as shown in Figs. 4b and c, the microstructure comprised of austenite, ferrite and martensite. It can be seen that the volume fraction of martensite was slightly increased with increase in austenitization temperature. The quenched martensite was dense on quenching from higher austenitization temperature and more martensite was retained after annealing.

Fig. 5 shows SEM micrographs of experimental steel after different Q&T heat treatments. Figs. 5a and b are microstructures of steel quenched from 600°C and 625°C (labeled as Q-600 and Q-625), respectively, followed by tempering at 200°C. The microstructural constituents consisted of ferrite (F) and austenite (A). When the steel

was quenched from 675°C, as marked in Fig. 5c, austenite was significantly decreased because of extensive martensitic transformation.

The volume fraction of austenite was determined by X-ray diffraction (XRD) with CuK α radiation using direct comparison method. In this method, integrated intensities of (200) α and (211) α peaks and those of (220) γ and (311) γ peaks was used.

The volume fraction of austenite V_A was calculated using equation [16],

$$V_A = 1.4I_\gamma / (I_\alpha + 1.4I_\gamma) \quad (1)$$

where I_γ is the integrated intensity of austenite and I_α is the integrated intensity of α -phase. As shown in Fig. 6a, the austenite fraction of steels subjected to ART heat treatment decreased from 34.2 vol% to 25.2 vol%, with increase in austenitization temperature from 800°C to 900°C. The volume fraction of austenite was increased and the volume fraction of ferrite was decreased with increase in annealing temperature during annealing in the intercritical region. Thus, it is clear that the austenite fraction of steels subjected to Q&T heat treatment increased from 36.2 vol.% to 58.2 vol.% (Fig. 6b) with increase in quenching temperature from 600°C to 625°C, followed by a sharp decrease to 21.2 vol.%, when quenched from 675°C because of martensitic transformation (as apparent from Fig. 5c).

3.3. Mechanical Properties

The mechanical properties of ART and Q&T steels are summarized in Fig. 7. Steels subjected to ART (Fig. 7a), the ultimate tensile strength (UTS) decreased from 945 MPa to 885MPa and total elongation (TEL) decreased continuously from 28% to 13% with increase in austenitization temperature. As a result, the product of UTS and

TEL decreased from 26.5 GPa% to 11.5 GPa% (Fig. 7b).

In the case of Q&T process (Fig. 7c), UTS continuously increased from 902 MPa to 1235 MPa with increased quenching temperature, and the TEL decreased with increased quenching temperature after the peak value of 42% at 625°C. The product of UTS and TEL approached a maximum value of 43.6 GPa% at 625°C (Fig. 7d). It is clear that steels subjected to Q&T heat treatment exhibited superior combination of mechanical properties as compared to steels subjected to ART heat treatment. Comparing with other medium Mn-content steels [17-22], the experimental steels after Q&T heat treatment have superior combination of mechanical properties. Furthermore, the Q&T heat treatment adopted in the present case is relatively more convenient than the ART heat treatment.

In TRIP steels, the mechanical properties are profoundly affected by the TRIP effect, which depends on the amount and degree of stability of retained austenite. The reason underlying the differences in mechanical properties between the steels subjected to two different heat treatments can be further elucidated by studying their deformation mechanisms and austenite stability.

4 Discussion

4.1. Analysis of Differences in Mechanical Properties between ART and Q&T Steels

The aforementioned results suggest that in the case of ART process, the ultimate tensile strength (UTS) and total elongation (TEL) decreased continuously with increase in austenitization temperature. Based on our earlier studies [23, 24], the ultimate tensile strength is determined by austenite fraction and TRIP effect. It is clear

that the volume fraction of retained austenite (measured by XRD) in steels subjected to ART process decreased with increase in austenitization temperature (Fig. 6a). Furthermore, the transformation ratio of austenite also decreased with increase in temperature (Fig. 6a), implying weakening of TRIP effect. Thus, the decreased volume fraction of retained austenite and weak TRIP effect was responsible for lower ultimate tensile strength. The small increase in volume fraction of martensite contributed to tensile strength, but the austenite fraction and TRIP effect were the main factors. The decreased total elongation is attributed to weak TRIP effect and small increase in volume fraction of martensite.

In the case of steels subjected to Q&T process, when the steels were quenched from 600-625°C, the volume fraction of retained austenite and transformation ratio of austenite was increased with increase in quenching temperature (Fig. 6b), which was responsible for the increased ultimate tensile strength and total elongation. When the steels were quenched from 650-700°C, the microstructural constituents consisted of ferrite, austenite, and martensite as the dominant phases. Thus, UTS was governed by the fraction of martensite and the TEL decreased with temperature because of increased martensite fraction and decreased transformation ratio of austenite.

It is obvious that steels subjected to Q&T (UTS: 902-1235 MPa, TEL: 18-42%, UTS×TEL: 20-44 GPa%) exhibit superior combination of mechanical properties as compared to steels subjected to ART (UTS: 885-945 MPa, TEL: 13-28%, UTS×TEL: 11-26 GPa%). As indicated by XRD data (Fig. 6), steels subjected to Q&T process had higher austenite fraction than steels subjected to ART process. Moreover, steels

subjected to Q&T, the transformation ratio of austenite was in the range of 26-76%. In contrast, the transformation ratio of austenite in steels subjected to ART was in the range of 34-42%. The austenite with low stability transformed to martensite easily during tensile deformation. Thus, it is inferred that the stability of austenite in Q&T steels is lower than that in ART steels. It is clear that the steels subjected to Q&T exhibited stronger TRIP effect which enhances the mechanical properties. Thus, it is clear that higher austenite fraction and more significant TRIP effect in steels subjected to Q&T was responsible for excellent tensile properties.

For further comparison between the effects of ART and Q&T heat treatments on the tensile properties, deformation and work-hardening behavior of AR-800 and Q-625 steels were studied.

The true strain-stress plots and work-hardening rate of AR-800 and Q-625 steels are presented in Fig. 8. The work-hardening behavior of the two steels exhibited three stages of work hardening rate (WH) evolution: the WH first decreased rapidly (stage I), and was mainly associated with the deformation of ferrite [25]. Next, the WH decreased gradually (stage II), which was attributed to the TRIP effect [26-27]. In the last stage, WH decreased again (stage III), and is presumably associated with the deformation of ferrite [28].

In the case of AR-800 steels, the WH decreased rapidly in stage II and this region was small, which is indicative of relatively low work hardening ability. In contrast, for Q-625 steel, the WH decreased gradually with fluctuations in stage II and the corresponding strain region was large, which indicates relatively strong work

hardening ability. Thus, it is clear that the lower work hardening ability and short strain region was responsible for the low tensile strength and elongation in AR-800 steel. While the high tensile strength and excellent ductility in Q-625 steel was a consequence of strong work hardening ability with a long strain region. The fluctuations in WH are directly related to the observed serrations in the true strain-stress plot of Q-625 steel, which is attributed to the discontinuous TRIP effect involving stress relaxation and transfer during deformation with consequent enhancement in ductility [15].

Work hardening ability is expected to be influenced by the austenite stability. In the attempt to quantify the differences between AR-800 and Q-625 steels, the effect of ART and Q&T process on the stability of austenite was studied.

4.2. Effect of Different Heat Treatment on Austenite Stability

The TRIP effect which contributed to excellent mechanical properties depends on the amount and degree of stability of austenite. The stability of austenite is important and governs the mechanical property of steel. It is desirable to produce austenite with appropriate stability such that strain-induced martensitic transformation occurs gradually with increase in plastic strain [29-30]. As shown in Fig. 6, it is worth noting that the transformation ratio of austenite in AR-800 and Q-625 steels was 42% and 76%, respectively. It is clear that Q-625 steel had lower austenite stability, corresponding to higher transformation ratio of austenite and more significant TRIP effect than AR-800 steel.

The stability of austenite depends on factors including its chemical composition,

grain size and morphology. The chemical composition has a strong influence on austenite stability. An increase in C and Mn-content, which are strong austenite stabilizers, increases the stability of austenite at room temperature [1, 20]. Fig. 9 shows SEM micrograph of AR-800 steel and the corresponding carbon and manganese concentrations by EMPA. It is obvious that the austenite (marked with a rectangle in Fig. 9) had higher carbon concentration than the adjacent ferrite (Fig. 9b). Moreover, manganese was found to be enriched in austenite (Fig. 9c). Consequently, we can deduce that C and Mn diffused from ferrite to austenite during annealing.

Comparing the TEM micrographs of AR-800 and Q-625 steels and the corresponding carbon and manganese partitioning in austenite and ferrite using TEM-EDS analysis in Fig. 10, it is apparent that the average dislocation density in austenite of Q-625 steel was higher.

Figs. 10c and e show corresponding concentration of C and Mn in ferrite and austenite of AR-800 steel, respectively. The statistical results of seventeen randomly selected grain dots presented in Figs. 10a, c and e indicated that C and Mn distribution was higher in austenite grains than ferrite, which further verified that C and Mn diffused from ferrite to austenite during annealing. The variation of C and Mn distribution were similar in Q-625 steel (Figs. 10b, d and f). Comparing the C distribution between AR-800 and Q-625 steels, it is clear that the extent of C distribution in austenite of AR-800 steel was higher than Q-625 steel (Figs. 10c and d). Moreover, the austenite of AR-800 steel was enriched more in Mn than Q-625 steel. The average carbon and manganese distribution in ferrite and austenite of AR-800 and

Q-625 steels are listed in Table 2. It is clear that austenite of AR-800 steel was enriched more in C and Mn during long time annealing subjected to ART process. While the C and Mn distribution in ferrite grains of AR-800 steel was lower than Q-625 steel, which means that more C and Mn diffused from ferrite to austenite during long time annealing subjected to ART.

The effect of chemical composition, specifically carbon and manganese content can be described in terms of chemical driving force for martensitic transformation and given by the following equation [31]:

$$\Delta G^{ch} = -7381.6 + 69447X_C + 19296X_{Mn} - 38776X_CX_{Mn} + (6.7821 - 33.45X_C)T \quad (2)$$

where ΔG^{ch} is the chemical driving force of martensitic transformation, X_C and X_{Mn} are the mole fraction of carbon and manganese in austenite, respectively, and T is the temperature in Kelvin. Eq. (2) shows that both carbon and manganese are potent stabilizers of austenite. Consequently, it can be inferred that austenite in AR-800 steel had higher stability than austenite in Q-625 steel, and is attributed to the higher enrichment with C and Mn. Long time annealing made austenite extremely stable and difficult to transform during tensile deformation such that the steel subjected to ART process exhibited reduced TRIP effect and inferior mechanical properties.

Fig. 11 presents typical TEM micrographs of AR-800 and Q-625 steels in the vicinity of fracture. It is obvious that more austenite (marked with a rectangle) was left in AR-800 steel (Fig. 11a). While in Q-625 steel, small amount of austenite was left after tensile deformation, which corresponds to high austenite transformation ratio and strong TRIP effect. Moreover, in both steels, the dislocation density was

high.

Based on the aforementioned analysis, a schematic of the differences in microstructure, carbon concentration, manganese concentration and dislocation density in undeformed and fractured AR-800 and Q-625 steels are presented in Fig. 12. Comparing the differences between AR-800 undeformed steel and Q-625 undeformed steel (Figs. 12a and c), it is clear that the steel subjected to ART had low volume fraction of austenite. During long time annealing in ART process, more C and Mn were enriched in austenite, which made austenite highly stable and resulted in low austenite transformation ratio (comparing Figs. 12b and d) such that significantly reduced TRIP effect and inferior mechanical properties were obtained. Furthermore, low dislocation density was present in AR-800 steel on long time annealing. It is known that dislocation evolution dominates the work-hardening rate during plastic deformation [32-33]. It is also inferred that low dislocation density was responsible for the low work-hardening rate in AR-800 steel. Moreover, it is well known that strong TRIP effect enhances the work hardening rate. The reduced TRIP effect in AR-800 steel was also responsible for the low work-hardening rate. Thus, long term annealing reduced the work hardening rate. Compared the differences between AR-800 undeformed and fractured steels, high density of dislocations was present in fractured steel, which suggested that dislocation slip must have occurred during tensile deformation. Thus, the ductility in AR-800 steel was associated with TRIP effect and cooperative dislocation slip, while TRIP effect was the primary factor. In the case of the steel subjected to Q&T process, high volume

fraction of austenite and low C and Mn-concentration resulted in low austenite stability, which was responsible for high austenite transformation ratio and significant TRIP effect. High dislocation density in austenite also contributed to the work-hardening rate. Thus, it is clear that Q&T process is more effective to obtain superior combination of mechanical properties.

5 Conclusions

The microstructure, mechanical property and austenite stability of hot-rolled 0.2C-1.6Al-6.1Mn-Fe TRIP steel subjected to ART and Q&T heat treatments were studied. The major findings of the study are as follows:

(1) Steels subjected to ART process were characterized by UTS of 885-945 MPa and TEL of 13-28%. In contrast, the Q&T steels exhibited superior combination of UTS of 902-1235 MPa and TEL of 18-42%. The Q&T process was more effective to obtain superior mechanical properties as compared to ART process.

(2) In ART heat treatment, during long time annealing, C and Mn were significantly enriched in austenite, which made austenite stable, leading to reduced TRIP effect. Furthermore, long time annealing lowered the dislocation density and resulted in low work-hardening rate. Inferior TRIP effect and low work-hardening rate were responsible for inferior mechanical properties of ART steels.

(3) The steel subjected to Q&T and quenched from 625°C exhibited best combination of mechanical properties (UTS: 1038 MPa, TEL: 42%, UTS×TEL: 43.6 GPa%) because of significant contribution of TRIP effect and high dislocation density in austenite. Comparing with ART process, Q&T process was characterized by

appropriate enrichment of C and Mn in austenite and required stability to induce significant TRIP effect.

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Figures:

Fig. 1. The hot-rolled processing schedule of the experimental steel.

Fig. 2. Dilatometric curves for the experimental steel.

Fig. 3. Heat treatment schedules for steels using (a) ART (austenite reverted transformation) annealing and (b) Q&T (quenching and tempering).

Fig. 4. SEM micrographs of ART-annealed steels for 12 h at 650 °C after austenitization at (a) 800 °C, (b) 850 °C, (c) 900 °C with 1 h. (A: austenite, F: ferrite, M: martensite).

Fig. 5. SEM micrographs of steels after quenching from different temperatures. (a) 600 °C, (b) 625 °C, (c) 675 °C. (A: austenite, F: ferrite, M: martensite).

Fig. 6. Measured austenite fractions of the undeformed and fractured steels and transformation ratio of austenite for the experimental steels heat-treated at different temperatures subjected to different heat treatment processes. (a) ART process, (b) Q&T process.

Fig. 7. Tensile properties of the experimental steels using different heat treatments. (a) UTS and TEL of steels subjected to ART, (b) product of UTS and TEL of steels subjected to ART, (c) UTS and TEL of steels subjected to Q&T and (d) product of UTS and TEL of steels subjected to Q&T. (UTS-ultimate tensile strength, TEL-total elongation).

Fig. 8. True strain–stress plots and work hardening rate of AR-800 and Q-625 steels. (AR-800: steel after austenitization at 800 °C subjected to ART process; Q-625: steel quenching from 625 °C subjected to Q&T process; S_I: stage I; S_{II}: stage II; S_{III}: stage III).

Fig. 9. SEM micrograph of AR-800 steel and the corresponding carbon and manganese concentrations by EMPA. (a) SEM micrograph, (b) C concentration and (c) Mn concentration.

Fig. 10. TEM micrograph of AR-800 and Q-625 steels and the corresponding carbon and manganese partitioned in austenite and ferrite. (a) TEM micrograph of AR-800 steel, corresponding (c) carbon and (e) manganese partitioned in austenite and ferrite

of AR-800 steel; (b) TEM micrograph of Q-625 steel, corresponding (d) carbon and (f) manganese partitioned in austenite and ferrite of Q-625 steel.

Fig. 11. TEM micrographs of the steels after tensile deformation. (a) AR-800 fractured steel and (b) Q-625 fractured steel.

Fig. 12. Schematic sketch of difference in microstructure, carbon-concentration, manganese-concentration and dislocation density in undeformed and fractured AR-800 and Q-625 steels.

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Fig. 4. SEM micrographs of ART-annealed steels for 12 h at 650°C after austenitization at (a) 800°C, (b) 850°C, (c) 900°C with 1 h. (A: austenite, F: ferrite, M: martensite).

Fig. 5. SEM micrographs of steels after quenching from different temperatures. (a) 600 °C, (b) 625 °C, (c) 675 °C. (A: austenite, F: ferrite, M: martensite).

Fig. 6. Measured austenite fractions of the undeformed and fractured steels and transformation ratio of austenite for the experimental steels heat-treated at different temperatures subjected to different heat treatment processes. (a) ART process, (b) Q&T process.

Fig. 7. Tensile properties of the experimental steels using different heat treatments. (a) UTS and TEL of steels subjected to ART, (b) product of UTS and TEL of steels subjected to ART, (c) UTS and TEL of steels subjected to Q&T and (d) product of UTS and TEL of steels subjected to Q&T. (UTS-ultimate tensile strength, TEL-total elongation).

Fig. 8. True strain–stress plots and work hardening rate of AR-800 and Q-625 steels. (AR-800: steel after austenitization at 800°C subjected to ART process; Q-625: steel quenching from 625°C subjected to Q&T process; SI: stage I; SII: stage II; SIII: stage III).

Fig. 9. SEM micrograph of AR-800 steel and the corresponding carbon and manganese concentrations by EMPA. (a) SEM micrograph, (b) C concentration and (c) Mn concentration.

Fig. 10. TEM micrograph of AR-800 and Q-625 steels and the corresponding carbon and manganese partitioned in austenite and ferrite. (a) TEM micrograph of AR-800 steel, corresponding (c) carbon and (e) manganese partitioned in austenite and ferrite of AR-800 steel; (b) TEM micrograph of Q-625 steel, corresponding (d) carbon and (f) manganese partitioned in austenite and ferrite of Q-625 steel.

Fig. 11. TEM micrographs of the steels after tensile deformation. (a) AR-800 fractured steel and (b) Q-625 fractured steel.

Fig. 12. Schematic sketch of difference in microstructure, carbon-concentration, manganese-concentration and dislocation density in undeformed and fractured AR-800 and Q-625 steels.

Table 1. Vickers microhardness of constituent phases in AR-800 steel.

Phases	Vickers hardness, HV
Austenite	247 ± 10
Ferrite	328 ± 15
Martensite	461 ± 10

Table 2. Average carbon and manganese distribution in ferrite and austenite of AR-800 and Q-625 steels.

Steels	Element	Ferrite (wt%)	Austenite (wt%)
AR-800	C	0.1	0.31
	Mn	3.1	11.4
Q-625	C	0.14	0.24
	Mn	4.5	7.6