The critical factors governing mechanical properties in hot-rolled

Fe-0.2C-11Mn-(0-6)Al TRIP steels and fracture behavior

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Abstract

In the present study, we fundamentally explore the reasons underlying mechanical property variation in hot-rolled 11Mn transformation-induced plasticity (TRIP) steels with varying Al-content. With increase in Al-content, the ultimate tensile strength (UTS) was decreased, while total elongation (TE) was increased. The variation in UTS was mainly associated with the fraction of constituent phases and stability of austenite, while the increase in total elongation (TE) with increase in Al-content was related to the increase in the fraction of ferrite. Moreover, the Al-free steel, quenched from intercritical region experienced brittle fracture. When the Al-content was increased to 2 wt.%, the total elongation increased to ~22-32% and is also attributed to TRIP effect. In 4Al steel, the tensile elongation was in the range of ~19-40%, and was associated with TRIP effect and cooperative dislocation slip. In 6Al steel, the cumulative contribution of transformation-induced plasticity (TRIP), twinning-induced plasticity (TWIP), and cooperative deformation of δ -ferrite led to high elongation of ~37-65%. The increase in the contribution of TRIP and TWIP with increase in Al-content led to dramatic change in the mode and characteristics of fracture.

Keywords: ultimate tensile strength; total elongation; fracture mode; dislocation slip; TWIP effect

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1. Introduction

The increased demand for energy conservation, environmental protection, and reduced weight of automobiles has led to significant interest in the development of new advanced high strength steels. Transformation induced plasticity (TRIP) steels with excellent combination of specific strength and ductility are considered as potential candidates for automotive applications [1, 2]. Light elements such as aluminum and magnesium are being added to TRIP steels to reduce the density of advanced high strength steels. According to the rough estimates, every 1% addition of Al contributes to 1.25% decrease in steel density [3, 4].

A number of studies have been carried out on Fe-(5-10)Mn-(0.1-0.2)C (wt%) ternary alloy system [5-8]. It was suggested that superior mechanical properties can be obtained with increase in C and Mn-content, which increases the volume fraction of retained austenite. For example, Fe-5Mn-0.2C steel exhibited tensile strength of 850-950 MPa and ductility of 20-30% [5]. While in 7Mn-steel, tensile strength of 1018 MPa and large total elongation of 32% was obtained [6]. Similarly, 1200 MPa tensile strength with 25% total elongation was obtained in Fe-9Mn-0.05C steel [9].

More recently, Al was added in medium-Mn TRIP steels to optimize austenite stability by suppressing cementite formation. In our previous study [10], Fe-0.2C-6Mn-1.6Al steel was characterized by tensile strength of 1040 MPa and ductility of 41%. In another instance, tensile strength of 1095 MPa and a total elongation of 42% was obtained in Fe-10Mn-0.14C-1.5Al steel [11]. Thus, there is evidence that Al can enhance the ductility of TRIP steels.

The objective of the study described here is to fundamentally explore the impact of Al in governing mechanical properties and fracture behavior in hot-rolled Fe-0.2C-11Mn-(0-6)Al TRIP steels.

2. Experimental

The nominal chemical composition of experimental steels (in wt. %) was Fe-10.80Mn-0.19C (0Al steel), Fe-11.20Mn-0.22C-1.95Al (2Al steel), Fe-11.02Mn-0.18C-3.81Al (4Al steel) and Fe-10.75Mn-0.21C-6.08Al (6Al steel). 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 2 h, and hot rolled to 4 mm thickness in the temperature range of 1150-850°C, and finally air cooled to ambient temperature.

The experimental steels were subjected to quenching and tempering (Q&T) heat treatment, which was envisioned by us as an alternative and effective heat treatment [12-14]. First, they were annealed in the two-phase region or austenite region 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 was helpful in relieving the internal stress, but more importantly it improved the stability of austenite through the diffusion of C from δ -ferrite to austenite [14].

Tensile specimens of width ~12.5 mm and a gauge 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 crosshead speed of $3 \text{ mm} \cdot \text{min}^{-1}$. The specification for the tensile specimen used was GB/T 228-2002, which is equivalent to ISO 6892:1998. Microhardness measurement was carried out by Vicker hardness tester using 50 gf, and the holding time was 10 sec. The samples were etched with 25% sodium bisulfite solution. Microstructural examination was carried out using optical microscope (OM), scanning electron microscope (SEM), electron microprobe analysis (EMPA) and transmission electron microscope (TEM). The volume fraction of austenite was determined by X-ray diffraction (XRD) with $\text{CuK}\alpha$ radiation. The fracture surface was studied by SEM and the region close to the tip of the fracture surface was elucidated by TEM.

3. Results

Prior to exploring the factors that influenced the mechanical properties and fracture behavior of experimental steels, we revisit the alloy design and microstructure associated with the steels [12, 15, 16].

3.1 Alloy design and microstructure

The phase fractions (ferrite, austenite, cementite) of experimental steels based on equilibrium thermodynamic analysis predicted by Thermo-calc are presented in Fig. 1 [17]. It is clear that the intercritical region is enlarged and austenite region is reduced with increase in Al-content.

OM micrograph of as-hot rolled 0Al steel is presented in Fig. 2a. According to the XRD data (Fig. 2b), the microstructural constituents consisted of α -ferrite (F), austenite (A), ϵ -martensite (ϵ -M), and α -martensite (M). SEM micrographs of as-hot

rolled 2Al, 4Al and 6Al steels are presented in Fig. 3. The microstructural constituents in 2Al steel consisted of α -martensite (M) and austenite (A). While in 4Al steel, besides α -martensite and austenite, δ -ferrite (δ -F) was observed. In 6Al steel, lamellar austenite was present between the adjacent δ -ferrite stripes.

Fig. 4 illustrates the microstructure of hot rolled 2Al steel heat treated at different temperatures. Figs. 4a and b describe the microstructure of samples quenched at 650°C and 700°C, respectively, followed by tempering at 200°C. The microstructural constituents consisted of acicular α -ferrite (α -F) and austenite (A) as the dominant phase. When the samples were quenched at 750°C and 800°C, respectively, as shown in Figs. 4c and d, the microstructure comprised of martensite (M) and austenite (A). Fig. 5 shows the microstructure of heat treated 4Al steel. Because of higher Al concentration, δ -ferrite (δ -F) was developed. As shown in Figs. 5a and b, the microstructural constituents of samples quenched at 750°C and 800°C, followed by tempering at 200°C consisted of δ -ferrite, α -ferrite and austenite. While the samples quenched at 850°C and 900°C, respectively, the microstructure consisted of δ -ferrite, austenite and martensite (Figs. 5c and d). Fig. 6 describes the microstructure of 6Al steel quenched at 650°C, 850°C, 950°C and 1000°C, respectively, followed by tempering at 200°C. In contrast, the microstructural constituents of 6Al steel consisted mainly of austenite and δ -ferrite stripes in the absence of martensite and α -ferrite.

The variation in the volume fraction of austenite obtained from XRD is summarized in Fig. 7. 2Al steel had a high austenite fraction of ~80% in the temperature range of 600-700°C, followed by dramatic decrease to ~25%, when

quenching was carried out in the temperature range of 750-850 °C because of martensitic transformation, as evidenced in Figs. 4c and d. Similar trend in the austenite fraction was observed in 4Al steel. It is obvious that 4Al steel had lower austenite content (~70%) than 2Al steel. The austenite fraction in 4Al steel was ~10%, when quenched from 850-900 °C because of martensitic transformation (Figs. 5c and d). In contrast, the austenite fraction in 6Al steel was not strongly governed by the quenching temperature and was ~50% in the range of 650-1000 °C. Schematic diagrams of the volume fraction of ferrite, austenite and martensite in 2Al steel, 4Al steel and 6Al steel heat-treated at different temperatures are presented in Fig. 8. Combining with the microstructure in Figs. 4, 5 and 6, it is obvious that the fraction of δ -ferrite increased with increase in Al content, considering that Al is ferrite stabilizer. There was no δ -ferrite in 2Al steel, and the fraction of δ -ferrite in 4Al steel estimated by Image-Pro Plus software was ~5%. It was confirmed by XRD results (Fig. 7) that the austenite fraction in 6Al steel was ~50%, thus, the fraction of δ -ferrite was ~50%.

3.2 Mechanical properties

The mechanical properties of these steels were summarized in Fig. 9 and a summary of microstructure and the range of mechanical properties is present in Table 1. It is distinct that the ultimate tensile strength (UTS) decreases with increase in Al content, while total elongation (TE) increases with increase in Al content. 0Al steel had (UTS) of 1336-1391 MPa, but very low TE of 3-5%. 2Al steel had a significantly high UTS of 1370-1450 MPa than 4Al steel (735-1200 MPa) and 6Al steel (600-650 MPa). The TE of 2Al steel was in the range of 20-32%, while in 4Al steel, TE reached

a maximum at 750°C (~40%) and then decreased with increased temperature. In 6Al steel, there was only a small variation in UTS in the range of 650-1000°C, but the corresponding TE decreased markedly from 65% to 37%.

It was intriguing to note that 2Al steel with highest austenite content had lowest TE and 6Al steel with the lowest austenite content had highest TE among the three steels when quenched from their respective two-phase region. The underlying reasons for the variation in tensile properties can be further elucidated by studying their fraction and stability of austenite and the characteristics of the fracture surface.

4. Discussion

4.1 The critical factor governing mechanical properties

The flow stress of multiphase steel (σ_F) can be estimated using the rule of mixtures proposed by Embury and Bouaziz [10, 18-20] (Eq. (1)),

$$\sigma_F = f_\alpha \sigma_\alpha + (f_\gamma - f_{nm}) \sigma_\gamma + (f_m + f_{nm}) \sigma_m \quad (1)$$

where σ_α , σ_γ , σ_m are the flow stresses and f_α , f_γ , f_m , f_{nm} are the volume fraction of ferrite, austenite, martensite and newly generated martensite, namely, transformed martensite evolved during tensile tests respectively. The variation in UTS in 4Al steel can be explained using the rule of mixtures. For 2Al steel, there were substantial differences in the constituents of the microstructure between samples quenched from 600 to 700°C and samples quenched from 750 to 850°C, however, there was only a small variation in UTS in these samples. Thus, the rule is not applicable in the case of 2Al steel.

In the case of 2Al steel, as shown in Fig. 4, the microstructural constituents of

steel quenched from 750 °C consisted of ferrite, austenite and martensite. The martensite that was present prior to tensile deformation was original or as-quenched martensite. The microstructural constituents of 2Al steels quenched from 650 °C and 700 °C consisted of ferrite and austenite. Thus, all the martensite was newly generated strain-induced martensite in 2Al-650 and 2Al-700 fractured steels. The Vickers hardness of newly generated martensite and original martensite was measured in 2Al-650 fractured steel and 2Al-750 steel, respectively. In 2Al steels, quenched from 650 °C and 750 °C, the Vickers hardness of the newly generated martensite (654 ± 15) was more than 1.5 times than the original martensite (403 ± 25). In the case of 4Al steels, quenched from 800 °C and 900 °C, the Vickers hardness of the newly generated martensite (452 ± 10) was close to the original martensite (476 ± 20) formed during quenching. Hence, the variation in UTS in 2Al steel can be explained using a modified rule of mixtures:

$$\sigma_F = f_\alpha \sigma_\alpha + (f_\gamma - f_{nm}) \sigma_\gamma + f_{nm} \sigma_{nm} + f_m \sigma_m \quad (2)$$

where σ_{nm} is the flow stress of newly generated martensite during tensile deformation. It is known that the TRIP effect is influenced by the volume fraction of retained austenite. According to the XRD studies (Fig. 7), 2Al steel had highest volume fraction of retained austenite. In the case of 2Al steel quenched from 700 °C, the volume fraction of retained austenite was 82% and the transformation ratio of austenite during tensile deformation was ~81%, which corresponds to strong TRIP effect. Thus, the superior UTS in 2Al samples quenched from 600 to 700 °C resulted from high fraction of retained austenite and strong TRIP effect and the high fraction

of newly generated martensite had high hardness. With decrease in Al content, the ultimate tensile strength (UTS) was increased, which is a consequence of increase in the volume fraction of retained austenite before deformation and increase in fraction of transformed martensite during deformation, contributing to strong TRIP effect.

The ϵ -martensite was present in 0Al steel quenched from 650°C and 700°C based on XRD data. The XRD pattern was similar to the as-hot-rolled 0Al steel. ϵ -martensite was not observed in 2Al, 4Al and 6Al steels. Thus, the low ductility of 0Al steel is believed to be due to presence of ϵ -martensite. In 2Al steel, the newly generated martensite with high Vickers hardness played an important role in enhancing the UTS. In 6Al samples, the austenite fraction was similar (Fig. 7), but the TE decreased with increase in quenching temperature, which was attributed to the decreasing austenite stability [21]. The reasons underlying the variation in TE with increase in Al-content can be further elucidated by studying their fracture surfaces and TEM micrographs of the fractured samples.

4.2 Fractography

Fracture is, often, the culmination of continued deformation processes. Therefore, fracture surface may retain signature of the deformation processes that a material was subjected to. The 0Al steel, 2Al steel, 4Al steel and 6Al steel when quenched from 650°C, 650°C, 750°C and 650°C (henceforth nominated as 0Al-650, 2Al-650, 4Al-750 and 6Al-650), respectively, exhibited best total elongation, though 0Al samples had low ductility. To study the fracture mode, the fracture surface of 0Al-650, 2Al-650, 4Al-750 and 6Al-650 samples were observed using SEM (Fig. 10).

The fracture surface of 0Al-650 steel is presented in Fig. 10a. It can be seen that the fracture was predominantly intergranular and intergranular cracks were observed. It is clear that the fracture mode of 0Al-650 steel was brittle fracture, resulting in low ductility, which is most likely to be associated with ϵ -martensite. The tensile fracture surface of 2Al-650, 4Al-750 and 6Al-650 steels was ductile, such that the tensile elongation increased with increase in Al-content in steel. Ductility is related to number density, size, depth and distribution of dimples in the fracture surface [22-24]. 2Al to 6Al steels contained dimples, but the nature and size of dimples was different among the three steels. In 2Al-650 sample, the size of ductile dimples was in the range of 0.5-3 μm (Fig. 10b), with maximum elongation of 32%. In 4Al-750 sample, the size of ductile dimples was in the range of 2-5 μm (Fig. 10c), with maximum elongation of 40%. The size of ductile dimples in 6Al-650 sample was in the range of 3-7 μm (Fig. 10d), with maximum observed elongation of 65%. Moreover, comparing the ductile dimples in Figs. 10b, c and d, it is obvious that the size of ductile dimples became inhomogeneous with increase in Al-content. Thus, it is inferred that the size of ductile dimples was increased and the distribution of ductile dimples became more non-uniform with increasing Al-content, corresponding to increase in total elongation.

4.3 Electron microscopy of deformed region

TEM micrographs of 0Al-650, 2Al-650, 4Al-750 and 6Al-650 steel corresponding to the tip of the fracture surface are presented in Fig. 11. Fig. 11a shows that the microstructural constituents in 0Al-650 steel comprised of austenite and lath martensite. The microstructure was similar to the undeformed sample. In

2Al-650 steel, some lath martensite was presented in the austenite matrix (Fig. 11b), while in 4Al-750 steel, more lath martensite was present in the austenite matrix (Fig. 11c). Moreover, comparing with 2Al-650 steel, it may be seen that the average dislocation density in austenite of 4Al-750 fractured steel was higher and fine strain-induced martensite was observed. In the case of 6Al-650 fractured steel, besides strain-induced martensite and dislocations in austenite, deformed δ -ferrite (Fig. 11d) and deformation twins (Figs. 11e and f) were observed. It is known that dislocation slip can effectively improve the ductility [25, 26]. In TRIP steels, the excellent total elongation can be obtained through the contribution of TRIP effect [27] and the ductility can be further enhanced via twinning-induced plasticity [28, 29].

In 0Al-650 fractured steel, low dislocation density was present, so the dislocation slip was inhibited during tensile test in 0Al steel. It could be responsible for the low ductility in 0Al-650 sample. Fine martensite (strain-induced martensite) (Fig. 11b) was evolved in the austenite matrix of 2Al-650 fractured steel during tensile deformation. According to the XRD results (Fig. 7), the original austenite fraction in 2Al-650 sample was ~84%, and was only ~35% retained after tensile fracture. Thus, transformation ratio of austenite during tensile deformation was ~58%. The transformation from austenite to martensite led to redistribution of stresses, an aspect that can be related to high elongation. Thus, in 2Al-650 steel, the TRIP effect contributed to total elongation of 32%. In 4Al-750 fractured steel, more transformed martensite was presented in the austenite matrix (Fig. 11c). According to the XRD results, the transformation ratio of austenite was 76%. Comparing with 2Al-650

fractured steel, high density of dislocations was present in austenite in 4Al-750 fractured steel, which suggested that significant dislocation slip must have occurred during tensile deformation. It is therefore inferred that the total elongation (40%) in 4Al-750 sample was associated with TRIP effect and cooperative dislocation slip. Transformed martensite and dislocations in austenite were observed in 6Al-650 fractured sample. The transformation ratio of austenite was 53%. As shown in Fig. 8, the ferrite fraction increased with increase in Al-content and when the Al-content increased to 6%, a large amount of δ -ferrite was obtained (Fig. 6), and α -ferrite was suppressed. Deformed δ -ferrite was also present in the TEM micrographs of 6Al-650 fractured steel (Fig. 11d). The martensitic transformation leads to volume expansion, which results in plastic deformation of δ -ferrite, thereby invoking relaxation and transfer of local stress, and contributes to additional strain [30]. Moreover, a high density of deformation twins (Figs. 11e and f) were also observed. Fig. 11f shows indexed selected area electron diffraction spots corresponding to the twinned region. Deformation twins can effectively inhibit deformation and hence delay the onset of necking, eventually leading to high total elongation. It can be further deduced that stacking fault energy increased with increase in Al-content, which contributed to TWIP effect. Thus, besides TRIP effect, TWIP effect also contributed to superior total elongation (65%) in 6Al-650 steel, together with the cooperative deformation of δ -ferrite.

The work hardening rate (WH) obtained from the tensile tests for 2Al-650 and 6Al-650 steels are presented in Fig. 12. The WH of 2Al-650 steel first decreases

(stage 1), then increases (stage 2), and then decreases with fluctuations (stage 3) (Fig. 12a). The first stage is mainly associated with the deformation of ferrite, and the second intermediate stage characterized by increase in WH is attributed to TRIP effect [7, 31]. In our previous study, austenite with different degree of stability was responsible for the TRIP effect that occurred discontinuously, which led to fluctuation in stage 3 of work hardening rate curves [10, 27]. The 6Al-650 steel also exhibits three stages of WH evolution (Fig. 12b): (1) WH rapidly decreases; (2) WH slowly decreases; (3) WH increases and then remains nearly constant for a strain range. The first stage is mainly associated with the deformation of ferrite, while the gradient of the downtrend in stage 2 is coordinated by the TRIP effect. In stage 3, the WH increases and remained almost constant instead of decreasing with strain. Thus, besides TRIP effect, TWIP effect was dominant in stage 3.

It is inferred that the increase in ductility was dependent on increase in the fraction of ferrite with increase in Al content. Moreover, in 2Al steel, the total elongation was also a contribution of TRIP effect. In 4Al steel, the total elongation was a cumulative effect of TRIP effect and cooperative dislocation slip, while in 6Al steel, besides high fraction of δ -ferrite, the cumulative contribution of transformation-induced plasticity (TRIP), twinning-induced plasticity (TWIP) contributed to highest elongation.

5. Conclusions

(1) The microstructure constituents and tensile properties were greatly influenced by Al content. With decrease in Al content, UTS was increased, which resulted from

the increase in fraction of austenite before deformation and increase in fraction of transformed martensite during deformation due to strong TRIP effect. Excellent combinations of UTS and TE were obtained for the steels containing 2-4 wt.% Al which had high austenite fraction and good austenite stability. In contrast, the Al-free steel exhibited low ductility (3-5%), and the steel containing 6 wt.% Al demonstrated low UTS (~650MPa).

- (2) 0Al steel quenched from intercritical region was brittle fracture, while the fracture mode of 2Al, 4Al and 6Al steels was ductile fracture. Moreover, the size of ductile dimples was increased and the distribution of ductile dimples became more inhomogeneous with the increasing Al-content, corresponding to the increase in total elongation.
- (3) The total elongation (TE) increased with increase in Al-content was governed by the increase in fraction of ferrite. Moreover, TRIP effect also contributed to the total elongation (32%) in 2Al-650 steel. In 4Al-750 steel, the total elongation of 40% was associated with TRIP effect and cooperative dislocation slip. In 6Al-650 steel, the extremely high elongation (65%) was a cumulative contribution of TRIP and TWIP effect and cooperative deformation of δ -ferrite.
- (4) In summary, the critical factors that govern mechanical properties (UTS and TE) of hot-rolled Fe-0.2C-11Mn-(0-6)Al TRIP steels depend on fraction of retained austenite, stability of retained austenite (hence transformation ratio of austenite during deformation), twinning and fraction of δ -ferrite.

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Figure(s)

Fig. 1. The phase fractions based on an equilibrium thermodynamic analysis predicted by Thermo-Calc for (a) 0Al steel, (b) 2Al steel, (c) 4Al steel and (d) 6Al steel. (F: ferrite, A:austenite, C: cementite) [17].

Fig. 2. (a) Optical micrograph and (b) XRD pattern of as-hot-rolled 0-Al steel.

Fig. 3. SEM micrographs of three as-hot-rolled steels. (a) 2Al steel, (b) 4Al steel and (c) 6Al steel.

Fig. 4. SEM micrographs of hot-rolled 2Al steel quenched from different temperatures and then tempered at 200°C. (a) 650°C, (b) 700°C, (c) 750°C, (d) 800°C.

Fig. 5. SEM micrographs of hot-rolled 4Al steel quenched from different temperatures and then tempered at 200°C. (a) 750°C, (b) 800°C, (c) 850°C, (d) 900°C.

Fig. 6. Optical micrographs of hot-rolled 6Al steel quenched from (a) 650°C, (b) 850°C, (c) 950°C, (d) 1000°C, respectively, and then tempered at 200°C.

Fig. 7. Measured austenite fraction of the undeformed and fractured samples as a function of quenching temperature.

Fig. 8. Schematic diagrams of the volume fraction of ferrite, austenite and martensite in (a) 2Al steel, (b) 4Al steel and (c) 6Al steel heat-treated at different temperatures. (F: ferrite, A: austenite, M: martensite).

Fig. 9. Tensile properties of the four experimental steels. (a) ultimate tensile strength, (b) total elongation. (UTS: ultimate tensile strength, TE: total elongation).

Fig. 10. SEM fractographs of (a) 0Al-650, (b) 2Al-650, (c) 4Al-750 and (d) 6Al-650 samples.

Fig. 12. Work hardening rate of 2Al-650 steel (a) and 6Al-650 steel (b) (S1: stage 1, S2: stage 2, S3: stage 3).

Table 1 Summary of microstructural constituents and tensile properties of experimental steels

Constituents			UTS (MPa)	TE (%)	PSE (GPa%)
0Al-11Mn	austenite, α -ferrite, ϵ -martensite, α -martensite		1336-1391	3-5	4-6.8
2Al-11Mn	austenite, α -ferrite (600-700°C)	austenite, α -ferrite, martensite (750-850°C)	1370-1450	22-32	31.1-44.8
4Al-11Mn	austenite, α -ferrite, δ -ferrite (700-800°C)	austenite, α -ferrite, δ -ferrite, martensite (850-900°C)	735-1200	19-40	22.7-37.8
6Al-11Mn	austenite, δ -ferrite and a small fraction of α -ferrite		600-650	37-65	22.1-42.3

UTS-ultimate tensile strength, TE-total elongation, PSE- product of strength and elongation

Figure captions:

Fig. 1. The phase fractions based on an equilibrium thermodynamic analysis predicted by Thermo-Calc for (a) 0Al steel, (b) 2Al steel, (c) 4Al steel and (d) 6Al steel. (F: ferrite, A: austenite, C: cementite) [17].

Fig. 2. (a) Optical micrograph and (b) XRD pattern of as-hot-rolled 0-Al steel.

Fig. 3. SEM micrographs of three as-hot-rolled steels. (a) 2Al steel, (b) 4Al steel and (c) 6Al steel.

Fig. 4. SEM micrographs of hot-rolled 2Al steel quenched from different temperatures and then tempered at 200°C. (a) 650°C, (b) 700°C, (c) 750°C, (d) 800°C.

Fig. 5. SEM micrographs of hot-rolled 4Al steel quenched from different temperatures and then tempered at 200°C. (a) 750°C, (b) 800°C, (c) 850°C, (d) 900°C.

Fig. 6. Optical micrographs of hot-rolled 6Al steel quenched from (a) 650°C, (b) 850°C, (c) 950°C, (d) 1000°C, respectively, and then tempered at 200°C.

Fig. 7. Measured austenite fraction of the undeformed and fractured samples as a function of quenching temperature.

Fig. 8. Schematic diagrams of the volume fraction of ferrite, austenite and martensite in (a) 2Al steel, (b) 4Al steel and (c) 6Al steel heat-treated at different temperatures. (F: ferrite, A: austenite, M: martensite).

Fig. 9. Tensile properties of the four experimental steels. (a) ultimate tensile strength, (b) total elongation. (UTS: ultimate tensile strength, TE: total elongation).

Fig. 10. SEM fractographs of (a) 0Al-650, (b) 2Al-650, (c) 4Al-750 and (d) 6Al-650 samples.

Fig. 11. TEM micrographs of (a) 0Al-650 fractured sample, (b) 2Al-650 fractured sample, (c) 4Al-750 fractured sample, (d) and (e) 6Al-650 fractured sample, (f) TEM micrographs of 6Al-650 fractured sample and indexing of spots of the selected area electron diffraction pattern from the twinning region.

Fig. 12. Work hardening rate of 2Al-650 steel (a) and 6Al-650 steel (b) (S1: stage 1, S2: stage 2, S3: stage 3).