



Article Influence of Partitioning Temperature on Microstructure and Mechanical Performance of Medium Manganese Fe-C-Mn-Cu-Cr-Mo-Nb Steel

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Abstract: A hot-rolled medium manganese Q&P steel was annealed at an intercritical annealing temperature (670 °C) and quenched to four partitioning temperatures (300 °C, 280 °C, 260 °C, and 240 °C). The effect of the partitioning temperature on the microstructure and mechanical properties of the steel was studied. The microstructure evolution at different partitioning temperatures was characterized using scanning electron microscopy (SEM), a transmission electron microscope (TEM), X-ray diffraction (XRD), and electron backscatter diffraction (EBSD). The results showed that, as the partition temperature increased, the fractions and the size of the fresh massive martensite (M₂) increased, the initial lamellar martensite (M₁) was coarsened, and the amount of M₁ decreased. Moreover, the amounts of lath-like retained austenite (RA) decreased and the amounts of bulky austenite increased. The tested steel partitioned at 260 °C had relatively greater fractions of RA, in which there were sufficient contents of carbon diffused from M₁. Therefore, the test steel partitioned at 260 °C showed fine comprehensive mechanical properties. The combination of tensile strength and elongation reached 30.3 GPa.%, and its work-hardening plateau was the longest, showing excellent ductility as it had sufficient fractions of the retained austenite with good stability.

Keywords: Q&P treatment; intercritical annealing; partitioning temperatures; retained austenite; work-hardening behavior

1. Introduction

Currently, interest in environmental protection and saving energy is increasing. In the automobile industry, using high-strength steel can effectively reduce car weight, which can result in reduced fossil-fuel emissions and improved fuel efficiency. Therefore, the development of advanced high-strength steels (AHSS) has attracted significant attention. Medium manganese steel, a typical third-generation AHSS, has been widely studied due to its combination of strength and ductility [1–5].

Researchers have proposed various alloys or heat treatments to improve the mechanical performance of medium manganese steels [6–10]. In 2003, Speer first proposed a quenching and partitioning (Q&P) process to produce steels with martensite and carbonrich retained austenite [11]. The process mainly consists of two steps. First, steel is quenched to a temperature between M_s and M_f . Some of the austenite is transformed to the initial lath-like martensite (M₁), and the rest remains as untransformed austenite. Then, the steel is reheated and partitioned at a temperature no lower than the quenching temperature (one- or two-step partition treatment). In the partitioning, carbon diffusion from M₁ into the untransformed austenite is promoted. The diffusion makes the austenite stable so that more of the austenite can be retained when the steel is finally quenched to room temperature. The microstructure of the Q&P-treated steel mainly consists of the initial lath-like carbon-depleted martensite (M₁), fresh massive carbon-rich martensite (M₂), and retained austenite (RA).



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Copyright: © 2022 by the authors. Licensee MDPI, Basel, Switzerland. This article is an open access article distributed under the terms and conditions of the Creative Commons Attribution (CC BY) license (https:// creativecommons.org/licenses/by/ 4.0/). The choice of the one- or two-step Q&P treatment is crucial. For low-alloy mediumcarbon steels, higher strength and toughness were obtained by the two-step Q&P treatment [12] than by the quench-tempering (Q&T) treatment [13]. When using the one-step Q&P treatment, the work-hardening rate, strength, and ductility were also improved owing to the interaction of dislocations in martensite and the TRIP effect [14]. Teng Wu used the one-step Q&P treatment, with the partitioning temperature slightly lower than the M_s point. The results showed that the product of tensile strength and elongation of the steel was significantly improved by the treatment [15]. The martensite structures in Q&P steels were studied. The double-martensite structure (M₁ and M₂) had denser and smaller high-angle grain boundaries than the single-phase martensite (M₁) [16]. Most of the retained austenite was observed near the M₂ boundaries. The TRIP effect of RA helped release the stress concentration of hard M₂, thereby increasing the overall plasticity [17].

Some studies showed that, by intercritical annealing (IA), high strength and ductility were obtained for high or medium manganese IA-Q&P steels. A high work-hardening rate and 65% total elongation were obtained for high manganese steel (Fe-10Mn-0.3C-3Al-2Si steel) [18]. Yang studied the effect of the critical annealing temperature and duration on the microstructure and mechanical properties of cheaper medium manganese (7 Mn) steel [19]. The product of the tensile strength and elongation of the steel reached 66 GPa.% after IA-Q&P treatment. Cell- and lath-like ferrites were observed. The ferrite transformed into granular and layered austenite after the Q&P treatment. The microstructure of the IA-Q&P-treated medium manganese steel consisted of initial martensite, fresh secondary martensite, and retained austenite. Feng performed the Q&P process on cold-rolled medium manganese steel at the intercritical annealing temperature. A high fraction of stable retained austenite was obtained using the treatment. The product of tensile strength and elongation of the steel reached 43.5 GPa.% [20].

To date, few studies have been conducted on the influence of initial martensite and fresh martensite in IA-Q&P-treated medium manganese steel. In this work, a designed medium manganese steel was subjected to an IA-Q&P treatment. The effect of the partitioning temperature on the microstructure and mechanical properties of the steel was studied. The interactions between the initial martensite, fresh martensite, and retained austenite were analyzed.

2. Experimental Materials and Methods

The material used was hot-rolled medium manganese steel with a thickness of 3 mm. Its chemical composition is listed in Table 1. The tested steel underwent smelting, casting, forging, rolling, and other processes. First, iron, carbon rod, electrolytic manganese, chromium, and other alloy raw materials were added to a vacuum induction furnace. The alloy composition was adjusted to meet the design requirements. By doing so, a 50 kg ingot was obtained. The ingot was kept in a high-temperature furnace at 1200 °C for 2 h, forged into a 28 mm × 120 mm × 350 mm slab, and air-cooled to room temperature. Subsequently, the slab was heated again to 1200 °C for 30 min to homogenize the alloy composition, then hot-rolled. The initial rolling temperature was 1150 °C, and the final rolling temperature was no lower than 825 °C. After four passes, the final hot-rolled sheet was 3 mm thick. Using JMatPro simulation software (7.0, Sente Software, Guildford, UK), the transformation point temperatures of the test steel were $A_{C1} = 599$ °C, $A_{C3} = 728$ °C and $M_s = 354$ °C, $M_f = 166$ °C.

Table 1. Chemical composition of the test steel.

Element	С	Mn	Cu	Cr	Мо	Nb	В	Fe
Content/wt.%	0.11	4.96	0.53	0.35	0.18	0.10	0.001	Bal.

The hot-rolled sheets were treated with IA-Q&P, based on the previous work [20–22]. A flowchart of the IA-Q&P treatment process is as follows (shown in Figure 1):

- (1) First, the sheets were heated at 670 °C between A_{C1} and A_{C3} . They consisted of austenite (A) and ferrite (F);
- (2) Then, they were first quenched to temperatures between M_s and M_f (300 °C, 280 °C, 260 °C, 240 °C). The samples cut from the sheets were named P300, P280, P260, and P240, respectively. Some of the austenite was transformed into martensite (M_1), and the rest was not transformed (the untransformed austenite). The sheets were kept in a salt bath furnace at the prescribed temperature for 150 s for partitioning;
- (3) Finally, the sheets were quenched to room temperature. Some stable carbon-rich untransformed austenite could be retained (RA), but some was not stable and was quenched to massive fresh martensite (M₂).



Figure 1. Heat treatment process flow.

The IA-Q&P-treated steels were machined into standard tensile samples with a gauge length of 25 mm, a width of 6 mm, and a thickness of 2.5 mm. The samples were stretched at a speed of 1 mm/min using a WDW-100E (Jinan Times Metallurgical Testing Machine, Jinan, China) microcomputer-electronic universal testing machine to obtain the stress–strain curves.

For microstructure observation, the specimens were cut on a QG-1 metallographic specimen-cutting machine (Tianxing Instrument, Shanghai, China). The size of the specimens was 10 mm \times 10 mm. They were rough polished using 200–1000# grit sandpapers and then polished using metallographic specimen grinding and polishing equipment (Tianxing Instrument, Shanghai, China). After polishing, the specimens were surface etched with a 4% nitric acid alcohol solution, followed by a pure water rinse, an experimental pure anhydrous ethanol rinse, and cold air drying. For further TEM observation, 3 mm foils were obtained and mechanically polished down to a 50 μ m thickness and twin-jet polished to 10–30 μ m thickness with a 92% CH₃COOH + 8% HClO₄ solution.

The microstructures of the treated samples were observed using field-emission scanning electron microscopy (SEM, SUPRATM 55, ZEISS, Oberkochen, Germany) and a transmission electron microscope (TEM, FEI Talos F200X, FEI, Hillsborough, OR, USA). The TEM acceleration voltage is 200 KV, the camera length is 520 mm, and the electron wavelength is 0.0251 Å. The TEM spot resolution is 0.25 nm, and the information resolution is 0.12 nm. The morphology and orientation of the electropolished samples were observed using electron backscatter diffraction (EBSD, SUPRATM 55, ZEISS, Oberkochen, Germany) with a 0.1 µm step size. An area of $24 \times 18 \ \mu\text{m}^2$ was chosen for the microstructure analysis.

The volume fraction of retained austenite was determined using an X-ray diffractometer (XRD, Smart Lab, Beijing, China) equipped with a Cu target with a scan speed of 2°/min and a scan range of 40–100°. Then, the volume fraction of retained austenite was calculated by Jade software (6.0, Materials Data (MDI), Livermore, CA, USA). The selected diffraction peaks of the BCC phase were (200) α and (211) α . The selected diffraction peaks of the FCC phase were (200) γ , (220) γ , and (311) γ . The volume fraction of the retained austenite was calculated using Equation (1) by precisely determining the corresponding diffraction angle 2 θ and integrated intensity I [23]:

$$V_{\gamma} = \frac{1}{1 + \frac{I_{\alpha} K_{\gamma}}{I_{\gamma} K_{\alpha}}} \times 100\%$$
⁽¹⁾

where V_{γ} is the volume fraction of retained austenite; I_{γ} is the integrated intensity of the austenite (200) γ , (220) γ , and (311) γ crystalline diffraction peaks; I_{α} is the integrated intensity of the martensite (200) α and (211) α crystalline diffraction peaks; and K_{γ} and K_{α} are the reflection coefficients of the austenite and martensite phases, respectively.

3. Results and Discussion

As shown in Figure 2, the microstructure of all treated samples was composed of martensite, ferrite (F), and retained austenite. Martensite was further divided into lath-like initial martensite (M_1) and massive fresh martensite (M_2) [24]. The retained austenite was mainly thin and lath-like between the initial lath martensite lamellae or near the massive fresh martensite boundaries and the grain boundaries of prior austenite.



Figure 2. SEM images at different partition temperatures: (a) P300; (b) P280; (c) P260; (d) P240.

According to the Koistinen–Marburger relationship [25], the volume fraction of the initial martensite (M_1) can be calculated. The formula is shown in Equation (2):

$$\mathbf{f}_{\mathrm{M}} = 1 - \exp(-\alpha(M_s - M_q)) \tag{2}$$

where f_M is the fraction of the initial martensite, M_s is the starting temperature of the martensite transformation, M_q is the quenching temperature, and the constant α is 1×1.0^{-2} .

Combined with Image-ProPlus 6.0 (IPP) software, the fraction of each phase (martensite, ferrite, and retained austenite) was estimated (shown in Figure 3 and Table 2). In all samples, the fraction of ferrite was close to 50%, whereas the fraction and morphology of martensite (M_1, M_2) and retained austenite were different, leading to the difference in the mechanical properties.



Figure 3. SEM images of IPP treatment at different partition temperatures: (**a**) P300; (**b**) P280; (**c**) P260; (**d**) P240.

Table 2. Areas	corresponding	to different col	or regions.

Area %	P300	P280	P260	P240
Red/α	50.0	55.3	45.1	48.2
Yellow/M(M ₁ , M ₂)	42.2(6.0, 36.2)	38.5(11.2, 27.3)	47.4(27.3, 20.1)	46.1(41.4, 4.7)
Green/γ	7.8	6.2	7.5	5.7

When the partition temperature decreased from 300 °C to 240 °C, the fraction of the massive fresh martensite (M₂) was gradually reduced from ~36.2% to ~4.7%. Meanwhile, the fraction of the initial lath-like martensite (M₁) increased from ~6.0% to ~41.4%, and the M₁ size was refined. At a partitioning temperature of 300 °C, a higher fraction of M₂ was observed, with a size of ~1.4 μ m, and the M₁ lath size was ~0.55 μ m. When the partitioning temperatures were 280 °C, 260 °C, and 240 °C, the size of M₂ decreased to ~0.7 μ m, 0.6 μ m, and 0.4 μ m, respectively. Meanwhile, the size of M₁ was gradually refined to ~0.43 μ m, 0.39 μ m, and 0.22 μ m. In the P240 sample, the fraction of initial martensite (M₁) was the largest, and its size was the finest. As the partition temperature decreased, the initial lath-like martensite and ferrite with increasing fractions continuously cut the untransformed austenite grains, resulting in M₂ refinement. Finally, the average diagonal size of M₂ showed a decreasing trend.

The martensite transformation is a non-diffusion type transformation. The amount of martensite formed was closely related to the partition temperature and austenite stability. With the decrease in the partition temperature, the fraction of the lath-like martensite (M_1) gradually increased, its average size became finer, and the fraction of the bulky fresh martensite (M_2) and the M/A islands composed of M_2 and retained austenite decreased [22]. This indicated that at a low partition temperature, the initial martensite hindered the growth of subsequent martensite, thereby refining the size of M_1 [26]. Alternatively, the

fine austenite grain and high dislocation density in the austenite also inhibited the growth of M_1 [27].

When the partition temperature decreased, an increasing amount of retained austenite was retained in the gaps between M_1 lamellae [28]. However, the XRD patterns of the samples showed that with the increase in the partition temperature, the fraction of the retained austenite first increased, then decreased, and finally increased again (Figure 4). Therefore, a dynamic relationship existed between the martensite and retained austenite. At a low partition temperature of 240 °C (the P240 sample), more fractions of initial martensite (M1, 41.4%) were formed, so there were low amounts of the untransformed austenite, leading to few fractions of the retained austenite (RA, 5.7%) and the massive fresh martensite (M_2 , 4.7%). At a high partition temperature of 280 °C (the P280 sample), the fractions of M_1 decreased (11.2%), and the fractions of the untransformed austenite increased. There was little M1 to supply carbon so the austenite became unstable, the fractions of RA became fewer (6.2%), and more fractions of M_2 were formed (27.3%). At a "moderate" partition temperature of 260 $^{\circ}$ C (the P260 sample), enough M₁ (27.3%) was formed and supplied carbon so that more fractions of RA were retained (7.5%). At a partition temperature of 300 °C near the M_s point (the P300 sample), a few fractions of M_1 were formed (6%). In contrast, the fractions of RA were relatively high (7.8%). The possible reasons were that, first, the high partition temperature was beneficial for carbon diffusion, and more fractions or regions of the untransformed austenite had carbon. Second, the M_s point of the carbon-rich austenite was reduced so that further martensite (M_2) transformation would be hindered to some extent, and the fractions of the retained austenite increased.



Figure 4. Experimental steels at different partition temperatures: (**a**) XRD pattern; (**b**) volume fraction of retained austenite.

Figure 5 shows the average carbon content in the retained austenite and martensite in the samples treated at different partitioning temperatures. The carbon content in retained austenite can be calculated with Equation (3) for the relationship between the austenite lattice parameters and carbon content [29]:

$$C_{\gamma} = (\alpha_{\gamma} - 3.547) / 0.0467 \tag{3}$$

where C_{γ} is the mass fraction of carbon in the retained austenite (%), and α_{γ} is the lattice constant of the retained austenite (220) crystal surface.

When the partitioning temperature increased, the carbon content in the retained austenite decreased, then increased, and then decreased again, whereas the carbon content in martensite increased below 280 °C, then decreased. The change of the carbon content in the retained austenite (RA) showed the opposite trend to that of the fraction of RA. An increase in RA would result in an increasing need for carbon, leading to a decrease in the average carbon content in RA. Quenched and partitioned at a low temperature of

240 °C, few fractions of RA were retained, with a relatively high C content because a greater amount of M_1 supplied carbon. At 260 °C, more fractions of RA were retained in the P260 sample. Moderate fractions of M_1 supplied carbon shared by more fractions of RA, so the average C content decreased. At 280 °C and 300 °C, few fractions of M_1 were formed and supplied carbon, while a high partition temperature was beneficial for C atom diffusion, so more fractions or regions of the austenite had few carbon elements.



Figure 5. Average carbon contents in retained austenite and martensite for the samples treated at different partitioning temperatures.

The carbon content in the martensite showed an increasing trend below 280 °C. As the partition temperature increased, the fractions of initial martensite (M_1) decreased, and more fractions of the unstable austenite transformed into fresh massive martensite (M_2). The carbon content in M_2 was higher than that in M_1 , resulting in an increase in the average carbon content in martensite. The retained austenite contents in the P300 sample were the highest. More carbon elements diffused into the retained austenite. As a result, the carbon content in the martensite decreased.

Figure 6 shows the mechanical properties of the test steel. The ultimate tensile strength (UTS), total elongation (TE), and product of tensile strength and elongation (PSE) of P260 were the maximum. In contrast, the properties of P280 were minimal. It was generally thought that more fractions of martensite were good for the tensile strength and yield strength obtained. However, the P260 sample exhibited an excellent combination of high strength and plasticity. This might be due to the TRIP effect of the retained austenite in the sample. The TRIP effect not only resulted in a substantial increase in elongation but also an increase in strength due to martensite transformation.

As the partition temperature increased, the fractions of the initial martensite (M_1) decreased, and the amounts of fresh martensite (M_2) increased. M_2 had a higher carbon content so its hardness was higher than that of M_1 . Consequently, the M_2 deformation coordination ability deteriorated during the tensile process, resulting in stress concentration and higher yield strength.

As shown in Figure 7, all the samples exhibited three-stage hardening behavior. The work-hardening rate in the first stage increased rapidly with the increase in strain, quickly decreased in the second stage, and decreased slowly in the third stage. The difference between the treated samples was in their second and third stages. In the second stage (~0.02 < $\varepsilon \le$ ~0.05), the P280 and P300 samples both had higher maximum values of work-hardening rates, owing to the stress concentration caused by the increase in fresh martensite (M₂). The third stage (~0.05 < ε) was mainly related to the TRIP effect of the retained austenite. In the enlarged figure, it can be seen that the P260 sample had a higher hardening rate plateau (average ~5000 MPa), whereas the P280 sample had the lowest, which corresponded to the trend in the elongation of the four samples. Moreover, the *n*

value provided favorable evidence of the TRIP effect [30]. Figure 7b shows that the P260 sample had the highest average *n* value, followed by the P240 and P300 samples, and the P280 sample had the lowest average *n* value.



Figure 6. Stress-strain curves and mechanical properties.



Figure 7. Work-hardening behavior of test steels at different partition temperatures: (**a**) work-hardening rate curve; (**b**) instantaneous strain-hardening exponent (*n* value).

When the partition temperature was 280 °C, the work-hardening plateau of the P280 sample was short because of lower retained austenite fractions. At the same time, owing to the insufficient stability of the retained austenite, martensite transformation occurred only at the initial stage. There was a low maximum value of the work-hardening rate of the P280 sample, as shown in Figure 7a. This resulted in a low subsequent work-hardening rate and *n* value. In addition, the fractions of hard fresh martensite (M₂) were high in the sample so the deformation stress for M₂ was transferred to the retained austenite. This accelerated the martensite transformation of the retained austenite and resulted in low plasticity.

When the partition temperature was 260 °C, there were sufficient fractions of stable retained austenite in the P260 sample. The retained austenite offered a sustained TRIP effect so the uniform elongation increased. In addition, the "appropriate" content of hard fresh martensite (M_2) provided resistance to martensite transformation, and the occurrence of necking was delayed. Thus, the elongation of the P260 sample was the highest. The low

elongation of the P280 sample might also be due to the relatively large amount of hard fresh martensite (M_2) , which hindered the TRIP effect of the retained austenite.

Furthermore, TEM observation was used to observe the microstructure of the P300 and P260 samples (shown in Figure 8). The retained austenite in the P300 sample was mainly block-shaped, with a little coarse lath-like austenite, and the retained austenite size was about 0.15–0.28 μ m. The retained austenite in the P260 sample was mainly thin and lath-like, and its size was finer, about 0.09–0.17 μ m. The size and morphology of the retained austenite were critical to its stability, along with the carbon content and fraction of the retained austenite. The lath-like retained austenite had higher stability than the block-shaped austenite. The lath-like austenite withstood more deformation, and its compatibility of deformation was better, so it slowly transformed into martensite at the early stage of deformation [32]. Plus, the smaller the size of the retained austenite, the better the stability. Therefore, although the P260 sample had relatively lower fractions of the retained austenite with lower carbon contents, the austenite was mainly lath-like, resulting in better overall mechanical properties of the P260 sample than the P300 sample.



Figure 8. TEM of P300 and P260 samples: (**a**) bright-field image of P300 sample; (**b**) dark-field image of P300 sample; (**c**) bright-field image of P260 sample; (**d**) dark-field image of P260 sample.

As shown in Figures 9 and 10, the P300, P260, and P240 samples were selected for EBSD analysis. As can be seen in Figure 9, if the misorientation of grain boundaries was between 2° and 15° , the boundaries were considered the low-angle grain boundaries, indicated by

red lines, and if over 15°, the boundaries were the high-angle grain boundaries, indicated by yellow lines. In the three samples (Figure 9a,c,e), the fractions of low-angle grain boundaries were about 47.1%, 45.3%, and 41.9%, while the fractions of high-angle grain boundaries were about 52.9%, 54.3%, and 58.1%, respectively. As the partition temperature decreased, the fractions of the low-angle grain boundaries displayed a decreasing trend, while the fractions of the high-angle grain boundaries displayed the opposite trend. Therefore, the grains deflected to some extent as the partition temperature decreased. However, in the P240 sample, the size of phases and the length of the grain boundaries were larger than that in the P300 and P260 samples.



Figure 9. EBSD test: (**a**,**b**) metallographic, polar, and IPF images of the P300 sample; (**c**,**d**) metallographic, polar, and IPF images of the P260 sample; (**e**,**f**) metallographic, polar, and IPF images of the P240 sample.



Figure 10. Recrystallization analysis: (a) P300 specimen; (b) P260 specimen; (c) P240 specimen.

The polar and inverse polar plots (IPF) of the three samples are shown in Figure 9b,d,f. It was observed that both the P260 sample and the P300 sample had strong <111> and <101> textures and a weak <001> texture. The P240 sample had a strong <111> texture and weak <001> and <101> textures. The three samples had strong textures, whereas the P300 and P260 samples showed a multi-oriented fine microstructure. Such microstructures were not significant in the P240 sample. The finer multi-oriented microstructure was favorable for plasticity and strength.

The microstructures of the samples were analyzed, as shown in Figure 10. There were recrystallized structures (blue zones), substructures (yellow zones), and deformed structures (red zones) in the samples. All three kinds of structures in P300 and P260 samples were fine. The deformed structure surrounded the recrystallized structure and substructure in the P300 and P260 samples. In other words, the substructure and recrystallized structures were uniformly distributed. In the P240 sample, the proportions of the substructures and deformed structures were the same. The fractions of the deformed structures in samples

P300 and P260 were similar. However, there was a small difference in the contents of the substructure and recrystallized structure, as shown in Figure 10a,b.

The recrystallized structure helped reduce the dislocation density, eliminate or reduce the stress in the matrix, and compensate for defects to some extent. Therefore, although the P300 and P260 samples had similar textures, grain morphologies, and contents, the P260 sample had better ductility. Comparatively, the recrystallized structure content in the P240 sample was lower than that in the P260 and P300 samples. The decrease in the recrystallized structure content with the decrease in the partition temperature might provide a reason for its low plasticity.

4. Conclusions

- (1) When the partitioning temperature increased, the fractions of the fresh massive martensite (M₂) increased, the fractions of the initial lath-like martensite (M₁) decreased, and M₁ became coarser. Meanwhile, the fractions of the thin lath-like retained austenite (RA) decreased, and the fractions of the block-shaped austenite increased. The fraction of retained austenite and its carbon content exhibited the opposite trend;
- (2) The lath-like retained austenite (RA) had better stability than the block-shaped austenite. When the fractions of RA were high, the TRIP effect, good for the plasticity of the steel, would be more obvious, showing a longer hardening rate plateau and a higher average *n* value during plastic deformation;
- (3) In the sample partitioned at high temperature, the fractions of the high-angle grain boundaries were few, but the microstructure (the lengths of the low- and high-angle grain boundaries, the recrystallized structure, substructure, and deformed structure) was finer and multi-oriented, and the substructure and recrystallized structure were distributed uniformly. The finer microstructure was favorable for plasticity.

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