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Effects of Induction Heating in the ESP Line on Microstructural Evolution and Nb Dissolution Behavior of Nb-Ti Micro-Alloyed Steel

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Abstract: The thermo-mechanical control processing of Nb-Ti micro-alloyed steel by induction heating in the endless strip production (ESP) line was analyzed to better understand the microstructural evolution and Nb precipitation and dissolution behavior in austenite during rapid heating to high temperatures. The Nb-Ti micro-alloyed steel consisting of 0.05 wt% C and 0.05 wt% Nb was processed through simulated rough rolling at 1050 °C followed by rapid isothermal reheating at 1150 °C. The austenite coarsening behavior and the Nb dissolution behavior at different holding times were compared, and the coarsening kinetics of austenite grains and the dissolution kinetics of precipitates were investigated. It was found that during induction heating, the size of austenite grains gradually increased with the isothermal time, and the amounts of precipitates were greatly reduced. Round precipitates of (Ti, Nb) (C, N) and square precipitates of Ti (C, N) gradually dissolved into the austenite matrix with the holding time. The Nb content in the solution increased from 0.0137 to 0.0299 wt% as the holding time increased from 1 to 40 s; therefore, about 59.8% of the total Nb content dissolved into the austenite matrix during the induction heating process.

Keywords: Nb-Ti micro-alloyed steel; ESP; induction heating; precipitation; dissolution

1. Introduction

Microalloying is a valid method to enhance the mechanical properties of steel. Highstrength low-alloy (HSLA) steels micro-alloyed with Nb, Ti, and V are extensively applied to automobiles, ships, oil, and gas transmission lines [1–4]. Microalloying elements can enhance the strength and toughness of HSLA steels by grain refinement, solid solution hardening, and precipitation hardening. The study on the precipitation hardening of Ti-Nb microalloyed steels with Mo and W has shown that the steel with Nb-Mo precipitates turned out to be most effective in precipitation hardening [5]. Development of austenite grain structures has been studied in microalloyed steels (Nb-Ti and Nb-V steels) and C-Mn steel after soaking at 950–1250 °C for 1 h. The results have shown that the fine austenite grain size in microalloyed steel at the lower soaking temperature can be attributed to the pinning effect from Nb (C, N) and V (C, N) precipitates. At higher soaking temperatures, dissolution of Nb precipitates led to austenite grain growth in microalloyed steels. Compared to C-Mn steel, microalloyed steels showed a significantly refined grain size [6]. Ti has frequently been added to HSLA steels to control the grain sizes of austenite and transformed ferrite during hot deformation, as well as the subsequent heat treatment. Nb facilitates grain refinement through carbide or carbonitride precipitation in austenite, restrains the static recrystallization of austenite, and contributes to dispersion hardening through ferrite precipitation during or after austenite-to-ferrite transformation [7,8]. The study on Nb precipitation and recrystallization kinetics for a Ti-Nb steel containing 0.021 Ti and 0.064 Nb (wt%) has shown that the Nb solute atoms and Nb precipitates in the <5 nm size range were



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Copyright: © 2021 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/). effectively retarding the recrystallization of austenite [9]. The study on the precipitation behavior and mechanical properties of Ti-Nb steel has shown that the fine-scale strain-induced precipitation of carbides occurred on dislocations and in the ferrite matrix. The Ti-Nb steel exhibited a yield strength and tensile strength of 562–600 and 629–687 MPa, respectively, with an elongation of 22–27% [10].

Thermo-mechanical control processing (TMCP) including controlled hot rolling and subsequent controlled cooling is another effective route to improve the mechanical properties of HSLA steels [11,12]. Endless strip production (ESP) continuous casting-rolling line (characterized by the short processing time and fast strip speed) is the most advanced endless rolling production line in the world today. In comparison to the traditional thinslab continuous casting-rolling process, the energy consumption is reduced by 50–70% and the water consumption is reduced by 60-80% in the ESP line [13-17]. In the ESP line, the induction heating process between rough rolling and finish rolling can rapidly reheat the intermediate billet to the required finish rolling entry temperature, improve the productivity and reduce the energy consumption for hot strip production, and maximize the microalloying effects during subsequent finish rolling, laminar cooling, and coiling by deformation-induced ferrite transformation (DIFT) and strain-induced precipitation (SIP) [18]. However, the Ti helps to improve the thermodynamic stability of precipitates and suppresses the dissolution behavior of Nb and C in Nb-Ti microalloying steels at high temperatures [19,20]. This resulted in a completely different dissolution behavior of Ti and Nb in the ESP process.

Introducing heating between rough rolling and finishing rolling is a brand-new process. Few works have studied how this process affects the microalloying elements in HSLA steels. In addition, the research on microalloying of HSLA steels has been more focused on the morphology, size, and distribution of precipitates, while less work has quantitatively determined the amounts of microalloying elements (in solution or as a precipitate). The present work focused on the effects of induction heating on the microstructural evolution and Nb dissolution behavior of Nb-Ti micro-alloyed steel with an amount of 0.05 wt% Nb. The dissolution kinetics of Nb and the coarsening kinetics of austenite grains were investigated by quantitatively analyzing the amount of Nb in solution and austenite grain size as a function of isothermal time. The research results will provide a guidance for improving the mechanical properties of Nb-Ti microalloyed steel.

2. Materials and Methods

The chemical composition of the experimental steel is presented in Table 1. The Nb-Ti microalloyed steel discussed here was industrially produced on the ESP line in Rizhao steel Co., Ltd. of China.

| Elements | С | Mn | Si | Als | Ti | Nb | Р | S |
|----------|------|------|------|-------|------|------|-------|--------|
| wt% | 0.05 | 1.65 | 0.25 | 0.025 | 0.06 | 0.05 | 0.012 | 0.0008 |

Table 1. Chemical composition of the experimental steel in wt%.

Specimens of size $\Phi 10$ (diameter) mm $\times 15$ (length) mm were obtained from the experimental steel. The induction heating process of the ESP line was simulated in a Gleeble-3500 thermal simulator (DSI, Gleeble-3500). The specimens were heated at 1300 °C for 120 s and cooled to a compression temperature of 1050 °C, and then a strain of 50% was applied at a constant true strain rate of 5 s⁻¹ and then cooled to 900 °C at a rate of 2.2 °C/s. In the simulated induction heating stage, the specimens were reheated to 1150 °C at a rate of 15 °C/s, then held for different times at this temperature (1, 10, 20, 40, 100, and 500 s), and quickly cooled to room temperature at a rate of 50 °C/s to maintain the high-temperature morphology (Figure 1).



Figure 1. Process specification of the experimental steel by Gleeble-3500 thermal simulator.

The amount of niobium (Nb) in the solution was determined by inductively coupled plasma optical emission spectrometry (ICP-OES: Varian 715-ES. VARIAN, Palo Alto, CA, USA). Metal scraps were dissolved completely in a mixed solution of hydrochloric acid, stannous chloride, and distilled water (volume ratio = 46:2:52) at 70 °C for 3–4 h, and the metal scraps weighing *M* (g) were taken from the above heat-treated specimens. After filtering, the prepared solution was transferred to a flask and diluted with distilled water to a volume of *V* (mL). According to the spectral intensity of the solution, the mass concentration of Nb (*X* in µg/mL) was measured by ICP-OES with an accuracy of 0.005 µg/mL. Eventually, the mass fraction (wt%) of Nb in the solution was calculated by Equation (1) [21–23].

$$W_{Nb}(\%) = \frac{X \times V}{M \times 10^6} \times 100\% \tag{1}$$

Longitudinal samples in the direction of hot compression were prepared for microstructural observations using a standard technique. The samples were mechanically ground using silicon carbide papers (200-grit to 2000-grit, Gold Sun Co., Ltd., Dongguan, China) and polished with diamond pastes (W1.5 10000-grit, Aotai Abrasives Co., Ltd., Hangzhou, China), and then eroded in 4% nitric acid alcohol solution for 10–15 s. The microstructures of samples were observed by scanning electron microscopy (SEM: Quanta FEG450. FEI Company, Hillsboro, OR, USA), and the average grains' diameter of austenite was determined by the linear intercept method. The microstructures and dislocation morphologies were observed by transmission electron microscopy (TEM: Tecnai F20. FEI Company, Hillsboro, OR, USA). Foil samples of thickness 0.5 mm for TEM analysis were obtained from the hot compression samples and thinned down to $50 \pm 10 \,\mu\text{m}$ using silicon carbide papers, then polished by a twin-jet electro-polisher (Tenupoi-5. STRUERS Ltd., Shanghai, China) in a solution of perchloric acid and ethanol (volume ratio = 15:85) at 20 V and –20 °C for 20–25 s [24].

The morphological observation and qualitative analysis of precipitates at different holding times were conducted by TEM and energy-dispersive spectroscopy (EDS: Inca X-Max50. Oxford Instrument, Oxford, UK). Carbon-extracted replica samples were prepared for TEM analysis. Samples were mechanically ground with silicon carbide papers and polished with W1.5 diamond pastes, and then chemically etched in 4% nitric acid alcohol solution for 10 s. After spraying the carbon films on the surface of samples, the carbon films were floated to the surface by etching in 7% nitric acid alcohol solution. The carbon replica films were washed in a mixed solution of 70% ethanol (purity >99.7%) + 30% deionized water, 50% ethanol (purity >99.7%) + 50% deionized water, and 10% ethanol (purity >99.7%) + 90% deionized water successively; and finally dried for TEM observation.

The volume fractions of precipitates in the samples were measured by quantitative image analysis. The particle diameter distributions in precipitates were quantitatively determined by Image Pro Plus image processing software (Version 6.0.0.260. Media Cybernetics, Rockville, MD, USA).

The micro-hardness of the samples was tested in a micro-hardness tester (HXD-1000TM. SHANGHAI OIF, Shanghai, China). The Vickers hardness test for each sample was carried out by a diamond indenter tip under a load of 4.9 N for 20 s. Thirty microhardness measurements were carried out for each sample with a distance of 0.1 mm.

3. Results

3.1. Microstructure

The microstructure of the sample before heating is presented in Figure 2. Figure 2a displays the morphologies of pearlite and austenite grain boundaries (GBs) of the sample before heating, and the average austenite grain size was measured as $41.4 \pm 3.5 \mu m$ by the linear intercept method. Moreover, partial dislocations were distributed in parallelly-arranged slab grains (Figure 2b).



Figure 2. Microstructures of the sample before heating: (a) SEM morphology; (b) TEM morphology.

The SEM morphologies of the samples after heating to 1150 °C for different isothermal times are shown in Figure 3. Figure 3a reveals austenite grain boundaries still present in the sample after holding for 1 s followed by rapid cooling to room temperature. The average austenite grain size was measured as $86.7 \pm 4.2 \,\mu\text{m}$, which greatly coarsened in comparison to the sample before heating. It is clear from Figure 3b–f that the size of austenite grains gradually increased with the holding time. The average austenite grain size reached 92.2 \pm 3.8, 94.1 \pm 4.7, and 106.4 \pm 5.9 μm after holding for 10, 20, and 40 s, respectively. The average austenite grain size rapidly increased from 145.1 \pm 4.6 to 216.9 \pm 6.7 μm as the holding time increased from 100 to 500 s. The coarsening behavior of austenite grains is discussed in Section 4.1.

3.2. Dissolution Behavior

The TEM morphologies and the corresponding EDS spectra of strain-induced precipitates in the sample are displayed in Figure 4. The dispersion of fine precipitates was observed in the foil. It is evident from the EDS spectra that round precipitates in Figure 4a and square precipitates in Figure 4c are (Ti, Nb) (C, N) (Figure 4b) and Ti (C, N) (Figure 4d), respectively. The appearance of (Ti, Nb) (C, N) occurred due to the interchangeability of Ti and Nb in the precipitate lattice because of their similar crystal structures and lattice parameters. The amounts of round precipitates were higher than those of square precipitates in the sample.



Figure 3. SEM morphologies of the samples heated to 1150 °C for different holding times: (**a**) 1 s, (**b**) 10 s, (**c**) 20 s, (**d**) 40 s, (**e**) 100 s, and (**f**) 500 s.

The variation in morphologies and particle diameter distribution of precipitates with isothermal time after heating to 1150 °C are presented in Figure 5. Precipitates continuously re-dissolved into the austenite matrix, and the amounts of precipitates gradually decreased with the holding time. The particle size of precipitates ranged between 20 and 30 nm after holding for 1 s; however, the number of precipitates was significantly reduced after holding for 10 s. When the holding time reached 40 s, coarsening of the precipitates slightly occurred and the number of precipitates further decreased. It was found that most of the undissolved particles had a size between 50 and 70 nm. The average particle diameters of precipitates were 20 ± 0.8 , 37.8 ± 1.3 , and 53.3 ± 1.0 nm after holding for 1, 10, and 20 s, respectively, and it reached 67.5 ± 1.7 nm after holding for 40 s.

The particle size distribution statistics of precipitates at different holding times are presented in Figure 6. The number of precipitates with a diameter less than 30 nm gradually decreased with the holding time. The peak value of the distribution for precipitates moved from 25 nm after holding for 1 s to 60 nm after 40 s. In general, the amounts of precipitates continuously decreased with the holding time, indicating that fine precipitates preferentially re-dissolved during the heating process.



Figure 4. TEM morphologies of precipitates in the specimen before heating displaying (**a**) round (Ti, Nb) (C, N) and (**c**) square Ti (C, N); panels (**b**,**d**) are the EDS spectra of the precipitate shown in panels (**a**,**c**), respectively.

3.3. Dissolution of Niobium

The amounts of Nb in the solution after heating at 1150 °C for different holding times were determined by ICP-OES, and the corresponding results are listed in Table 2.

| Table 2. Amounts of Nb in the solution held at 1150 °C for different t |
|---|
|---|

| Holding Time/s | Before Heating | 1 | 10 | 20 | 40 |
|----------------------------------|-----------------------|--------|--------|--------|--------|
| The amount of Nb in solution/wt% | 0.0051 | 0.0137 | 0.0221 | 0.0282 | 0.0299 |
| Account for total additions/% | 10.2% | 27.4% | 44.2% | 56.4% | 59.8% |

The content of Nb in the solution before heating was 0.0051 wt%, which accounted for 10.2% of the total amount of added Nb. The content of Nb in the solution gradually increased with holding time. The amounts of Nb in the solution were 0.0137 wt% (accounting for 27.4% of the total amount of added Nb), 0.0221 wt%, 0.0282 wt%, and 0.0299 wt% after holding for 1, 10, 20, and 40 s, respectively. Therefore, about 59.8% of the total amount of added Nb dissolved into the austenite matrix during the induction heating process.

(a)

100 nm

100 nm

(c)



Figure 5. TEM morphologies of the precipitates in the samples heated to 1150 °C for different holding times, (**a**) 1 s, (**b**) 10 s, (**c**) 20 s, and (**d**) 40 s.

100 nm



Figure 6. Particle size distribution statistics of precipitates at different holding times.

3.4. Micro-Hardness

The micro-hardness values of the samples after heating at 1150 °C for different isothermal times are presented in Figure 7. When the isothermal time was 1 s, the micro-hardness was measured as 241.1 \pm 3.2 HV. When the isothermal time increased to 10 s, the microhardness value decreased to 237.8 \pm 1.6 HV. The micro-hardness value gradually decreased from 236.5 \pm 1.8 HV to 234.1 \pm 2.7 HV as the isothermal time increased from 20 to 40 s. The micro-hardness value further decreased from 230.3 \pm 3.5 HV to 225.4 \pm 2.9 HV as the isothermal time increased from 100 and 500 s. Therefore, the micro-hardness of experimental steel gradually decreased with the isothermal time. It indicates that the coarsening of austenite grains and the decrease in carbide amount due to the dissolution of microalloying elements during the heating process had a marked impact on the micro-hardness value.



Figure 7. Average micro-hardness values for the samples under different holding times at 1150 °C.

4. Discussion

4.1. Coarsening of Austenite Grains during the Heating Process

Figure 3 reveals that austenite grains continuously coarsened at 1150 $^{\circ}$ C with the increase in holding time. The coarsening behavior of austenite grains during isothermal heating can be better understood by the theory of Feltham [25]. The change in the austenite grain size with the holding time can be calculated by Equation (2).

$$(D^*)^2 - (D_0^*)^2 = (\lambda Va\sigma)t \exp(-H/TK) = K \times t$$
⁽²⁾

where D_0^* represents the grain size at *t* of zero, D^* represents the grain size at a specific time, λ is a constant, *V* represents the volume per atom, *a* denotes the lattice spacing, σ denotes the specific grain boundary energy, and *H* represents Planck's constant. In the isothermal process, Equation (2) is equivalent to $K \times t$, where *K* denotes a constant. Therefore, using Equation (2) and the grain size data at the holding times of 1 and 10 s, the value of *K* is equal to 97.85. The austenite grain diameters calculated by Equation (2) are presented in Figure 8. The predicted austenite grain sizes after 20, 40, 100, and 500 s were 97.3, 106.9, 131.5, and 237.7 µm, respectively. The austenite grain sizes predicted by the theory of Feltham are well consistent with the experimental data.

The Zener pinning effect indicates that the grain growth can be hindered due to the pinning of austenite grain boundaries by precipitates. The continuous dissolution of niobium and titanium weakened the pinning force of the precipitates on the austenite grain boundaries; thus, the coarsening of austenite grains occurred during the heating process. However, after isothermal holding for a certain time, the coarsening rate of austenite grains was restrained due to the dragging effect of dissolved microalloying elements.



Figure 8. Experimental data and calculated data for grain diameters of the samples under different isothermal times at 1150 °C.

4.2. Dissolution Behavior of Nb during the Heating Process

The volume fraction of precipitates measured from replica samples by TEM is presented in Table 3. It is noticeable that the volume fraction of precipitates decreased with the isothermal time at 1150 °C. The contents of Nb in the solution after heating at 1150 °C for different holding times were determined by ICP-OES (listed in Table 2), and its simulated data were calculated by Equations (3) and (4) [26].

$$f(t) = \left[\frac{\rho_r / \rho_{precipitate} \times ((Z+1)/Z)}{100}\right] \times (W_m - [Nb])$$
(3)

$$Z = \{Nb\}/\{C\}\tag{4}$$

where f(t) represents the volume fraction of precipitates that changes with the holding time, ρ_r and $\rho_{precipitate}$ are the austenite density and the precipitate density, respectively, W_m is the total amount of added Nb in the test steel, [Nb] represents the dissolved Nb content, and {Nb} and {C} represent Nb and C in the equilibrium state of NbC precipitates, respectively. According to the theory of Johnson–Mehl [27], the relationship between the volume fraction of precipitates and the isothermal time can be determined by Equation (5).

$$f(t) = 1 - \exp(-kt^n) \tag{5}$$

where f(t) represents the volume fraction of precipitates that changes with the holding time, k represents a factor associated with the nucleation rate and the growth rate, t represents the isothermal holding time, and n represents the nucleation rate factor. The volume fraction of precipitates is zero at the initial holding time, but in this experiment, undissolved precipitates still existed when the test steel was reheated to 1150 °C and held for 1 s. Therefore, the increase in volume fraction of precipitates for the holding time of 1 s can be expressed by Equation (6).

$$f(t) = 1 + f_1 - \exp(-kt^n)$$
(6)

where f_1 represents the volume fraction of precipitates at holding for 1 s. Eventually, the changes in the Nb content in the solution with isothermal time can be calculated by Equation (7).

$$1 + f_1 - exp(-kt^n) = \left[\frac{\rho_r / \rho_{precipitate} \times ((Z+1)/Z)}{100}\right] \times (W_m - [Nb]) \tag{7}$$

| Isothermal Time/s | Volume Fraction of Precipitates/ $10^{-5} \ \mu m^{-3}$ | The Content of Nb in the Solution/wt% |
|-------------------|---|--|
| 1 | 63.5 ± 1.9 | 0.0122 |
| 10 | 51.8 ± 2.8 | 0.0245 |
| 20 | 37.2 ± 3.1 | 0.0291 |
| 40 | 32.7 ± 2.3 | 0.0322 |

Table 3. Volume fraction of precipitates and the calculated values of Nb content in the solution at various isothermal times.

The values of k and n were obtained from the measured data for the holding times of 1 and 10 s. The contents of Nb in the solution at various isothermal times calculated by Equation (7) are presented in Table 3. As the holding time increased from 1 to 40 s, the mass fraction of Nb in the solution increased from 0.0122 to 0.0322 (wt%); thus, most of the added Nb dissolved into the austenite matrix during isothermal heating.

The comparison between the experimental data and the simulation data by Equation (7) is presented in Figure 9. It is evident that the dissolution model in Equation (7) provides an appropriate prediction for the Nb dissolution behavior.



Figure 9. Measured data and calculated data of Nb dissolution behavior in the samples with holding times at 1150 °C.

The precipitation and dissolution of Nb were greatly affected by different holding times. The amount of Nb in the solution predicted by Equation (7) was well consistent with the measured data when the isothermal holding time was less than 20 s; however, the measured data gradually became smaller than the calculated values with the holding time. Generally, the addition of Ti improves the high-temperature stability of precipitates and inhibits the dissolution of alloying elements (Nb and C) at high temperature. Therefore, the addition of Ti suppressed the dissolution kinetics of Nb in the test steels.

4.3. Effects of the Heating Process on DIFT

To study the influence of induction heating on the dissolution behavior of Nb, the test steel cooled to 900 °C after rough rolling was rapidly cooled to room temperature at a rate of 50 °C/s (as shown by red line in Figure 1). It is noticeable from Figures 2 and 4 that the microstructure was fine and the average grain diameters were $41.4 \pm 3.5 \mu m$ before heating; however, the austenite grain size gradually increased from 86.7 \pm 4.2 μm at 1 s to 216.9 \pm 6.7 μm at 500 s. A large amount of precipitates existed in the sample before heating, and the content of Nb in the solution accounted for only 10.2% of the total added amount. The amounts of precipitates gradually decreased with the isothermal time at

1150 °C, and about 59.8% of Nb dissolved into the austenite matrix at an isothermal time of 40 s. The induction heating process promoted the dissolution of microalloying elements, which exerted their precipitation strengthening effects during the finish rolling process.

The induction heating of the ESP line generally refines the microstructure through deformation-induced ferrite transformation. In the special thermomechanical process of induction heating, the precipitates formed in the roughing stage dissolved again by rapid reheating to 1150 °C in a period after roughing at 1050 °C, and the dissolution of a large amount of Nb and Ti into the austenite matrix increased the recrystallization temperature of austenite by solid solution dragging [28]. Hence, the incomplete recrystallization of austenite occurred during the rolling process at the high temperature, and the storage energy and accumulation of deformation defects increased the nucleation rate of ferrite during the finish rolling process, thereby accelerating the conversion rate of the γ to α phase transformation and refining the ferrite grains. Finally, the mechanical properties of test steel are improved by fine-grain strengthening.

The dissolution of precipitates during induction heating increased the supersaturation of Nb and Ti in the austenite matrix and increased the driving force for microalloying precipitation in the test steel. In addition, the volume fraction and average particle size of strain-induced precipitates are also directly affected by the deformation law [29,30]. The sufficient deformation strain and the incomplete recrystallization of austenite during finish rolling increased the deformation storage energy, increased the driving force for precipitation, and promoted the influence of supersaturation on the precipitation behavior. Therefore, microalloying elements (Nb and Ti) dissolved into the matrix during induction heating, and the finish rolling process was completed in the region where austenite grains were not completely recrystallized. This phenomenon promoted the precipitation of large amounts of fine particles by strain-induced precipitation and enhanced the mechanical properties of the Nb-Ti micro-alloyed steel by precipitation strengthening.

Based on the above study, the Nb-Ti micro-alloyed steel with an amount of 0.05 wt% Nb was industrially produced on an ESP line. The cast billets were rough-rolled with a strain of 50% in the range of 1150–1050 °C, and cooled to 900 °C at a rate of 2.2 °C/s; the intermediate billets were immediately reheated to 1150 °C at a rate of 15 °C/s, was held for 20 s, and then continuously rolled to the required thin strips through five finishing mills with an average reduction of 30% in the range 1000–850 °C. Finally, the thin strips were water-cooled to 620 °C for coiling and then air-cooled to room temperature. The microstructure of the experimental steel contains mainly polygonal ferrite and a small content of quasi-polygonal ferrite, with an average grain size of 2.97 μ m. At the same time, many precipitates with a size less than 18 nm appeared. The experimental steel displayed a yield strength and tensile strength of 684 and 745 MPa, respectively, with an elongation of 18.9% due to the combined effects of DIFT and SIP.

5. Conclusions

- 1. The average sizes of austenite grains were 92.2 ± 3.8 , 94.1 ± 4.7 , 106.4 ± 5.9 , 145.1 ± 4.6 , and $216.9 \pm 6.7 \mu m$ after holding for 10, 20, 40, 100, and 500 s, respectively, and the predicted austenite grain sizes after 20, 40, 100, and 500 s were 97.3, 106.9, 131.5, and 237.7 μm , respectively. Therefore, the coarsening behavior of austenite predicted by Feltham's theory was consistent with the obtained experimental data.
- 2. Round precipitates of (Ti, Nb) (C, N) and square precipitates of Ti (C, N) gradually dissolved into the austenite matrix with the holding time. The amounts of Nb in the solution were 0.0137, 0.0221, 0.0282, and 0.0299 wt% after holding for 1, 10, 20, and 40 s, respectively. The amounts of Nb in the solution accounting for the total amount of added Nb increased from 27.4% to 59.8% as the holding time increased from 1 to 40 s.
- 3. The contents of Nb in the solution calculated by the JMAK equation were 0.0122, 0.0245, 0.0291, and 0.0322 wt% after holding for 1, 10, 20, and 40 s, respectively. The Nb dissolution behavior with the isothermal time simulated by the Johnson–Mehl

theory was consistent with the measured data when the isothermal time was less than 20 s; however, the measured data gradually became smaller than the calculated values with the holding time; thus, the addition of Ti inhibited the dissolution of Nb.

4. Nb-Ti micro-alloyed steel with an amount of 0.05 wt% Nb was industrially produced on an ESP line. The microstructure of the test steel contains mainly polygonal ferrite and a small content of quasi-polygonal ferrite, with an average grain size of 2.97 μm. The test steel displayed a yield strength and tensile strength of 684 and 745 MPa, respectively, with an elongation of 18.9%.

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