Effect of niobium alloying on the austenite grain growth and mechanical properties of ultrahigh-strength stainless steel

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Abstract

In the temperature range of 1000 °C–1150 °C and the holding time range of 30–150 min, the effect of niobium (Nb) on the behavior of grain growth and the evolution pattern of the mechanical properties of a martensitic stainless steel was studied. This study found that the addition of Nb allowed a large amount of undissolved NbC phase to be present in the steel, that the dragging effect of the solute atoms such as solute Nb and Mo reduced the migration rate of the grain boundary, and the pinning effect of NbC hindered the growth of grains, and that the growth rate of grains in 0.11Nb steel was slow in the temperature range of 1000 °C–1080 °C and increased significantly at the temperature range of 1080 °C–1150 °C. Next, the kinetic equations of the grain growth of 0.002Nb steel and 0.11Nb steel were constructed. The second phase strengthening of NbC and the fine grain strengthening jointly increased the yield strength of the steel but reduced the plasticity and ultimate tensile strength (UTS) of the steel. The addition of Nb had a minor effect on the content of retained austenite in the steel, but its refining effect on the hierarchical martensite microstructure increased the number of nucleation sites of retained austenite, reduced their sizes, made their distribution more dispersed, and more effectively hindered crack propagation, thus improving the toughness of the steel.

1. Introduction

Ultrahigh-strength stainless steel has been widely used in the aerospace, the energy, and other fields due to its high strength, high toughness, and excellent corrosion resistance \cite{1, 2}. The excellent comprehensive properties of ultrahigh-strength stainless steel are due to the formation of a supersaturated solid solution by solid solution strengthening, the matrix of lath martensite structure obtained in subsequent quenching, and a large number of dispersed second phases precipitating from the subsequent aging treatment, thereby greatly improving the strength \cite{3, 4}. To improve the strength level, researchers have committed extensive effort to optimize the alloying design and heat treatment. For example, Xu \textit{et al} \cite{5} developed a type of strengthened maraging stainless steel with compound precipitation of MC carbides, Cu particles, and the Ni\textsubscript{3}Ti phase, and the material had an ultimate tensile strength (UTS) of 1600 MPa. Based on the idea of nanoparticle precipitation in the ultrahigh dislocation density martensite structure, Liu \textit{et al} \cite{6} developed a new type of ultrahigh-strength steel with a strength level of 2400 MPa using nanoscale NiAl phase as the strengthening phase, and the elongation could reach 11.4\%. The strengthening methods of these steels are mainly derived from the nanoscale precipitated phases in the steel. The volume fraction and size of the precipitated phases and the interaction between the precipitated phases and dislocations all have a substantial impact on the precipitation strengthening effect \cite{7–11}. Niobium (Nb) is a common microalloying element in low-alloy and microalloyed steel. The addition of a small amount of Nb to the steel can form a precipitated NbC phase that has a pinning effect on the grain boundary or form solid solution Nb atoms that produce a drag effect on the grain boundary migration, inhibiting the growth of grains and thus improving the strength and toughness of steel \cite{12–14}. However, so far, there are few reports on the precipitation behavior of Nb during the austenitization process of ultrahigh-
strength stainless steel and its impact on microstructure refinement and toughness. For this reason, this paper studies the influence mechanism of Nb on the grain growth pattern and mechanical properties of a type of martensitic ultrahigh-strength stainless steel and provides new data and a theoretical basis for improving the strength and toughness of martensitic ultrahigh-strength stainless steel.

2. Experimental materials and methods

The vacuum induction melting and vacuum arc remelting processes were adopted to prepare different martensitic stainless steel with different Nb contents. The alloy composition is shown in Table 1. After the steel ingot was homogenized at 1150 °C for 4 h, it was forged into a round bar with a diameter of φ150 mm at the temperature range of 900 °C–1050 °C. All metallographic specimens with dimensions of 10 × 10 × 5 mm³ were cut at 1/2 R of the round bar. The specimens were heated to 1000, 1050, 1080, 1100, and 1150 °C, and the soaking times were 30, 60, 90, 120, and 150 min, followed by oil quenching to room temperature. After the metallographic specimen was ground and polished, the grain boundary was corroded with 10% potassium permanganate solution. A PHILIPS APD-10 x-ray diffractometer (XRD) was used to measure and analyze the austenite content in the steel. The morphology and distribution of austenite in the specimens were characterized by Electron back-scattered diffraction (EBSD). The specimens for tensile and impact testing were cut at 1/2 R of the φ 150-mm round bar, the specimens were heated to 1000, 1050, 1080, 1100, and 1150 °C, and the soaking times was 60 min, followed by oil quenching to room temperature. Subsequently, the specimens were stored at −73 °C for 8 h. The room–temperature tensile properties of the steel were tested on a LOS-600 electronic universal testing machine following the GB/T 228.1–2010 standard. According to the GB/T 229–2007 standard, the NI-130 impact testing machine was used to test the room–temperature impact toughness of standard Charpy U-notch (CUN) specimens. A Neophot optical microscope was used to observe the original austenite grain microstructure, and the average grain size was measured by the image analysis software Nano measure. A scanning electron microscope (SEM) (Quantax650) was used to observe the fracture morphology and the second phase morphology and distribution of the specimens, and the thermodynamic software Thermal-Calc and the TCFE9 database were used to calculate the property diagram of the steel specimens.

3. Experimental results

3.1. Thermodynamic calculation of the equilibrium microstructure of the second phase in steel

The equilibrium property diagrams of 0.002Nb steel, 0.06Nb steel, and 0.11Nb steel were calculated by ThermoCalc and the TCFE9 database, as shown in Figure 1. Below 1400 °C, the second phases in the steel were mainly M₆C and NbC, and there was no NbC precipitate in the equilibrium phase diagram of 0.002Nb steel. As shown in Figure 1(a), the complete solid solution temperature of the only precipitated phase in the steel, M₆C, was approximately 1040 °C. As shown in Figure 1(b), the second phases in 0.06Nb steel were M₆C and NbC, and the complete solid solution temperatures were approximately 1080 and 1113 °C, respectively. The complete solid solution temperatures of M₆C and NbC in 0.11Nb steel were approximately 1080 and 1220 °C, respectively. As shown in Figure 1(c), the mole fraction of NbC in 0.11Nb steel reached the maximum at approximately 1080 °C. It can be seen from the calculated results that the precipitation temperature range and mole fraction increased with increasing Nb content in the steel. In addition, after Nb is added to the steel, the complete solid solution temperature of the M₆C phase was increased.

3.2. The influence of the heating temperature on the original austenite grains

Figure 2 shows images of the original austenite grains of the steel specimens after different heating temperatures were applied for 60 min. With increasing heating temperature, the sizes of the original austenite grains of the three steel specimens all showed an increasing trend. In addition, the average grain size is larger for the martensitic stainless steel with lower amount of Nb at all temperatures. Different from those of the 0.002Nb steel and 0.06Nb steel, the grain size of 0.11Nb steel did not increase substantially in the heating temperature range of 1000 to 1080 °C, and the grain microstructure was relatively uniform, as shown in figures 2(c), (f), and (i). It may
be attributed to the pinning forces of carbide particles on austenitic grain boundaries, which will be proved by SEM observation. When the heating temperature was higher than 1100 °C, the trend of grain growth was obvious, as shown in figures 2(l) and (o).

To explore the corresponding relationships of the austenite grain growth rate of the steel specimens with the heating temperature and holding time, the original austenite grain sizes of the 0.002Nb steel and 0.11Nb steel specimens were statistically analyzed, and the results are shown in figure 3. It can be seen from figure 3(a) that within the heating temperature range of 1000 to 1150 °C, the original austenite grain size of 0.002Nb steel showed almost a linear growth trend with increasing heating temperature, and that the grain growth rate was almost unchanged. The trend of change in the original austenite grain size of 0.11Nb steel with heating temperature and holding time is shown in figure 3(b). It can be seen from the figure that the growth rate of the original austenite grain size was different in the temperature range of 1000 °C–1150 °C. In the temperature range below 1080 °C, the growth rate of the original austenite grains was slow, and when the heating temperature exceeded 1080 °C, the growth rate of the original austenite grains was significantly increased.

3.3. The influence of the heating temperature on the second phase

Figure 4 shows the microstructure morphology of the steel specimens at different heating temperatures. It can be seen from figure 4(a) that when the heating temperature was 1000 °C, partially undissolved second phases were present at the original austenite grain boundaries and the grains of 0.002Nb steel, and that their sizes were between 300–900 nm. After the energy-dispersive x-ray spectroscopy (EDS) analysis, it was determined that the undissolved second phase was the M₆C phase rich in Fe, Mo, W, and Cr (as shown in figure 4(e)). When the heating temperature was 1050 °C, the second phase was completely dissolved (as shown in figure 4(b)). When the heating temperature was 1000 °C, there were two types of second phases in the 0.11Nb steel, i.e., M₆C phase and NbC phase (as shown in figure 4(c)). The EDS results of the enlarged morphology of the M₆C and NbC phases in 0.11Nb are shown in figures 4(f) and (g), respectively. It is worth noting that when the heating temperature was increased to 1100 °C, the M₆C phase in 0.11Nb steel was essentially completely dissolved. At this time, the carbides in 0.11Nb were mainly NbC, and the NbC phase at 1100 °C are often present in the grains rather than being on the boundaries. The EDS results are shown in figure 4(h), which was generally consistent with the thermodynamic calculation result in figure 1(c).
3.4. Model of the austenite grain growth

The effect of the second phase and solute atoms such as solute Nb and Mo on the growth of austenite grains is mainly a pinning effect and the dragging effect on the austenite grain boundaries [15–18]. In the present article the effect of the second phase particles is more prominent and will be discussed. The dissolution of the second phase weakens the pinning effect of the second phase on the grain boundary, so the size of the original austenite grains increases rapidly. Based on this, the effects of heating temperature, holding time, and initial grain size on the kinetics of the original austenite grain growth of 0.002Nb steel and 0.11Nb steel were explored, and the kinetic equation of austenite grain growth [19–23] was expressed as equation (1):

Figure 2. The metallographic morphologies of the original austenite grains of the steel specimens. (a) 1000 °C (0.002Nb steel), (b) 1000 °C (0.06Nb steel), (c) 1000 °C (0.11Nb steel), (d) 1050 °C (0.002Nb steel), (e) 1050 °C (0.06Nb steel), (f) 1050 °C (0.11Nb steel), (g) 1080 °C (0.002Nb steel), (h) 1080 °C (0.06Nb steel), (i) 1080 °C (0.11Nb steel), (j) 1100 °C (0.002Nb steel), (k) 1100 °C (0.06Nb steel), (l) 1100 °C (0.11Nb steel), (m) 1150 °C (0.002Nb steel), (n) 1150 °C (0.06Nb steel), (o) 1150 °C (0.11Nb steel).
where $t$ is the holding time, min; $n$ is the grain growth exponent; $d_i$ is the average size of the austenite grains when heated to a certain temperature with a holding time of $t$, μm; $d_0$ is the average size of the original austenite grains, μm; $A$ is a constant determined by the grain boundary diffusion coefficient and isothermal time; $m$ is the time index; $Q$ is the activation energy for grain growth, J mol$^{-1}$; $R$ is the gas constant, 8.314 J mol$^{-1}$ K$^{-1}$; and $T$ is the peak heating temperature, K.

After $d_0^n$ was moved to the left side of the equation, the natural logarithm was applied to both sides of equation (1), yielding equation (2):

$$\ln (d_i^n - d_0^n) = \ln A + m \ln t - \frac{Q}{RT}$$

At a fixed heating temperature, the equation to solve $m$ could be expressed as equation (3):

$$m = \left. \frac{\partial [\ln (d_i^n - d_0^n)]}{\partial (\ln t)} \right|_{T=\text{Constant}}$$

At a fixed holding time, the equation to solve $Q$ could be expressed as equation (4):

$$Q = \left. R \frac{\partial [\ln (d_i^n - d_0^n)]}{\partial (1/T)} \right|_{t=\text{Constant}}$$

According to equations (3) and (4), $m$ and $Q$ corresponding to different values of $n$ could be obtained. With the combination of equation (2), linear regression of the experimental data was performed, and the values of $A$ under different values of $n$ could be obtained. Thus, we could obtain the function of austenite grain growth under different values of $n$. The error between the grain size $d_i$ corresponding to different $n$ values calculated according to equation (5) and the actual grain size $d$ obtained from the experiment, $f_{\text{Error}}$, was calculated.

$$f_{\text{Error}} = \sum_j (d_j - d)^2$$

The equations to calculate $f_{\text{Error}}$ for 0.002Nb steel in the temperature range of 1000 °C–1150 °C and 0.11Nb steel in the temperature ranges of 1000 °C–1080 °C and 1080 °C–1150 °C were as follows:

(0.002Nb steel (1000 °C–1150 °C))

$$f_{\text{Error}} = -22475.8418 + 234926.6939n - 434629.63116n^2 + 311022.20131n^3 - 95119.70295n^4 + 10542.30803n^5$$

(0.11Nb steel (1000 °C–1080 °C))

$$f_{\text{Error}} = 9515.3803 - 12288.51023n + 3561.53112n^2 + 1543.12843n^3 - 1005.77167n^4 + 144.86435n^5$$

(0.11Nb steel (1080 °C–1150 °C))

$$f_{\text{Error}} = 2541.1524 - 37863.03045n + 56943.36358n^2 - 33898.7995n^3 + 9117.83607n^4 - 924.1244n^5$$

Figure 3. The relationship between the austenite grain size of the steel specimens and the heating temperature; (a) 0.002Nb steel, (b) 0.11Nb steel.
It can be seen from figure 5(a) that when $f'_\text{Error}$ was 0, the $n$ of 0.002Nb steel in the temperature range of 1000 °C–1150 °C was 1.34651. Similarly, it can be seen from figures 5(b) and (c) that when $f'_\text{Error}$ was 0, the $n$ values of 0.11Nb steel in the temperature ranges of 1000 °C–1080 °C and 1080 °C–1150 °C were 2.48654 and 1.55856, respectively. By converting the proportional relationship between $\ln (d''_n - d'^{0}_n)$ and $\ln t$ in figure 6, we can conclude that the $m$ of 0.002Nb steel was 1.140204. The $m$ values of 0.11Nb steel in the temperature ranges of...
1000 °C–1080 °C and 1080 °C–1150 °C were 0.65397 and 0.76465, respectively. According to the relationship between \( \ln (d'_{f} - d''_{f}) \) and \( 1000T^{-1} \) in figure 7, we can conclude that the \( Q \) of 0.002Nb steel was 133423.338 J mol\(^{-1}\), and the \( Q \) values of 0.11Nb steel in the temperature ranges of 1000 °C–1080 °C and 1080 °C–1150 °C were 247196.205 J mol\(^{-1}\) and 215739.562 J mol\(^{-1}\), respectively.
The grain sizes of 0.002Nb steel and 0.11Nb steel at different heating temperatures and different holding times were subjected to linear regression, and the $A$ values obtained by linear fitting are shown in Figure 8. As shown in Figure 8, the $A$ of 0.002Nb steel was $3.75 \times 10^5$, and the $A$ of 0.11Nb steel was $1.77 \times 10^{12}$ in the temperature range of 1000°C–1080°C and $6.56 \times 10^9$ in the temperature range of 1080°C–1150°C.

In summary, we obtained the $A$, $Q$, and $m$ values of 0.002Nb steel and 0.11Nb steel in different temperature ranges. By substituting the obtained results of $A$, $Q$, and $m$ into equation (2), we deduced the mathematical model of austenite grain growth as follows in equations (9)–(11):

$$\ln (d_t^{1.34651} - d_0^{1.34651}) = \ln 3.75 \times 10^5 + 1.140204 \ln t - \frac{133423.338}{RT}$$ (0.002Nb steel (1000°C–1150°C))

$$\ln (d_t^{2.48654} - d_0^{2.48654}) = \ln 1.77 \times 10^{12} + 0.65397 \ln t - \frac{229593.205}{RT}$$ (0.11Nb steel (1000°C–1080°C))

$$\ln (d_t^{1.55856} - d_0^{1.55856}) = \ln 6.56 \times 10^9 + 0.76465 \ln t - \frac{215739.562}{RT}$$ (0.11Nb steel (1080°C–1150°C))

The average grain size of the original austenite predicted by the established mathematical model was generally consistent with the measured average grain size of the original austenite, as shown in Figure 9. The above result showed that the mathematical model could be used to predict the austenite grain growth behavior of ultrahigh-strength stainless steel.

### 3.5. Room-temperature mechanical properties of the steel specimens

Figure 10 shows the curve of the room-temperature mechanical properties of the steel specimens as a function of heating temperature. It can be seen from Figure 10(a) that the UTS of the steel specimens have no significant change with increasing heating temperature and that the yield strength (YS) gradually decreased with increasing heating temperature. Under the same heating temperature, the UTS of the steel decreased while YS increased with the increase of Nb content. It can be seen from Figure 10(b) that with increasing temperature, the impact energy ($A_{KU2}$) of the steel specimens gradually increased and then gradually decreased. The $A_{KU2}$ of the 0.002Nb steel reached the maximum value of 192 J at 1050°C, while the $A_{KU2}$ values of the 0.06 Nb steel and 0.11Nb steel...
Figure 8. The A values of 0.002Nb and 0.11Nb in different temperature ranges obtained by linear fitting. (a) 0.002Nb steel (1000 °C–1150 °C), (b) 0.11Nb steel (1000 °C–1080 °C), and (c) 0.11Nb steel (1080 °C–1150 °C).

Figure 9. Comparison of measured and predicted austenite grain sizes. (a) 0.002Nb steel, (b) 0.11Nb steel.

Figure 10. Tensile and impact test results of the steel specimens; (a) UTS and YS and (b) impact energy.
reached their maximum values (207 and 225 J, respectively) at 1080 °C. This is due to the further dissolution of the second phase and the further growth of the austenite grain, resulting in a lower resistance to dislocation movement. It is worth noting that the AKU2 values of the 0.002Nb steel and 0.11Nb steel were significantly higher than that of the 0.002Nb steel at the same heating temperature. For example, when the heating temperature was 1080 °C, the AKU2 values of the 0.002Nb steel, 0.06Nb steel, and 0.11Nb steel were 187.5, 207, and 224 J, respectively.

3.6. Fracture morphology

Figure 11 shows the room-temperature impact fracture morphologies of the steel specimens. It can be seen from the figure that the fracture morphologies of the three steel specimens at different heating temperatures were all ductile dimple morphologies, which were typical ductile fractures. It can be seen from figures 11(a)–(c) that when the heating temperature was 1000 °C, the size of the ductile dimples of the impact fracture was relatively small, and there were large-sized second phases in the dimples. At this time, the AKU2 values of the steel specimens were low. When the heating temperature was 1080 °C, no large-scale second phase was observed in the ductile dimples of the three steel specimens, as shown in figures 11(d), (e), and (f), which was related to a large amount of solid solution in the precipitated phase at 1080 °C (as shown in figures 4(b) and (d)). Because of the large amount of solid solution in the undissolved phase, the toughness of the steel increased [24], and the ductile dimple size was relatively large at this time. The increase of dimple size is mainly caused by the dissolution of the second phase and the growth of grain size [25].

4. Discussion

The microstructure characterization results showed that Nb microalloying could effectively inhibit the growth of austenite grains and that the effect of inhibiting grain growth was more obvious with increasing Nb content. The grain structure refinement mainly occurred for two reasons: first, the drag effect of the solute Nb reduced the migration rate of the grain boundary, and the pinning effect of the NbC precipitated phase prevented the migration of the grain boundary [26]; second, Nb microalloying increased the complete solution temperature of M6C, as shown in figure 1. When the heating temperature was lower than the complete solid solution temperature of M6C, the M6C in the 0.06Nb steel and 0.11Nb steel also played a role in pinning the grain boundaries [27], which was the main reason for the fact that the growth rate of the austenite grain size of the 0.06Nb steel was less than that of the 0.002Nb steel when the heating temperature was lower than 1080 °C. It can be seen from the transmission electron microscopy (TEM) morphology in figures 12(a) and (b) that when the heating temperature was 1080 °C, the microstructure of the 0.002Nb steel was mainly composed of lath
martensite with high dislocation density, and that no undissolved second phase was found. At this temperature, there was still a small amount of spherical second phase in the 0.06Nb steel. The diffraction spot calibration result showed that in addition to M₆C, there was NbC in the second phase of the 0.06Nb steel. The sizes of M₆C and NbC were approximately 100 and 50 nm, respectively (as shown in figures 12(c)–(f)). In contrast, the number of second phases with a nearly spherical shape in the 0.11Nb steel increased considerably. The diffraction spot calibration showed that the second phase was face-centered cubic NbC with a size of approximately 150 nm, as shown in figures 12(g) and (h).

Nb microalloying improves the YS of steel because at the same heating temperature, as the Nb content increased, the content of the undissolved second phase in the steel increased. According to the Orowan strengthening mechanism, the second phase can effectively prevent the movement of dislocations, thereby increasing the YS of steel [28]. In addition, with increasing Nb content, the microstructure of steel was significantly refined. According to the Hall-Petch equation, an increase in grain size can lead to a decrease in the YS of steel [29–32]. However, an increase in the size of the undissolved second phase causes a decrease in the plasticity of steel because the existence of large-sized second phase particles disrupts the continuity of the matrix, resulting in stress concentration near the second phase during the deformation process. Consequently, the interface between the second phase and the matrix becomes an important source of crack nucleation [33], the plasticity of the steel decreases, and the tensile strength decreases, as shown in figure 10(a). Retained austenite is also one of the main factors affecting the mechanical properties of steel [34]. To explore the effect of retained austenite content on the mechanical properties of the steel specimens, the retained austenite contents of the three steel specimens under different heat treatment conditions were measured. The results are shown in figure 13. As shown from figures 13(a)–(c), the matrix structure of the steel was composed of martensite and reversed austenite. At the same heating temperature, the austenite contents of the three steel specimens increased slightly with the increase in Nb content (13(d)). According to the research results of Wang et al [35], the reversed austenite mainly existed at the original austenite grain boundaries, martensite lath block boundaries, martensite lath bundle boundaries, and martensite lath boundaries. The refinement of the original austenite grains leads to refinements of the martensite lath blocks, martensite lath bundles, and martensite laths, thus increasing the number of nucleation sites of retained austenite. The increase in retained austenite nucleation sites results in a more uniform distribution of austenite in the matrix, which in turn leads to a stronger impediment to crack extension and better toughness of the steel. In this study, it was observed that the retained austenite was uniformly distributed in the martensitic matrix (see figure 14). Figure 14 is intended to depict the distribution of retained austenite (red area in figure 14) in 0.002Nb and 0.11Nb steels. As can be seen in figure 14, the retained austenite nucleation sites in 0.11Nb steel are slightly higher than that in 0.002Nb steel. Austenite as a ductile phase in steel can improve the toughness of the steel. Although the increase in Nb content does not have a significant impact on the retained austenite content, the refinement of the hierarchical martensite microstructure affects the size and uniformity of the retained austenite, which also improves the toughness of the steel.

Figure 12. TEM morphology of steel specimens after the temperature was held at 1080 °C for 60 min (a) TEM bright-field image of 0.002Nb steel, (b) TEM dark-field image of 0.002Nb steel, (c) TEM bright-field image of M₆C in 0.06Nb steel, (d) Selected area electron diffraction of M₆C, (e) TEM bright-field image of NbC in 0.06Nb steel, (f) Selected area electron diffraction of NbC, (g) TEM bright-field image of NbC in 0.11Nb steel, (h) Selected area electron diffraction of NbC.
5. Conclusions

In the temperature range of 1000 °C–1150 °C and the holding time range of 30–150 min, the grain growth behavior and the evolution pattern of the mechanical properties of Nb in ultrahigh-strength stainless steel were studied. The main conclusions obtained were as follows:

(1) The growth rate of the grains in the 0.002Nb steel was almost unchanged in the temperature range of 1000 °C–1150 °C. The growth rate of the grains in the 0.11Nb steel was slow in the temperature range of 1000 °C–1080 °C but increased considerably in the temperature range of 1080 °C–1150 °C. This is because in the temperature range of 1000 °C–1080 °C, the addition of Nb results in a large amount of undissolved...
 NbC phase in the steel, and the pinning effect of NbC phase and the dragging effect of the solute atoms hinders the grain growth. When the temperature is higher than 1080 °C, the solid dissolution of a large amount of NbC occurs, the pinning effect on the grain boundaries is reduced, and the grain growth rate is accelerated.

(2) With the introduction of error equations, the calculated grain growth exponent (n) of the 0.002Nb steel was 1.347, and that of the 0.11Nb steel was 2.486 at 1000 °C–1080 °C and 1.558 at 1080 °C–1150 °C, respectively, the activation energies for the grain growth of the steel specimens were found to be 133.42 KJ mol⁻¹ (0.002Nb steel), 229.59 KJ mol⁻¹ (0.11Nb steel, 1000 °C–1080 °C), and 215.74 KJ mol⁻¹ (0.11Nb steel, 1080 °C–1150 °C), respectively, and the kinetic equations for the grain growth of the steel specimens were constructed:

\[
\ln (d_t^{1.34651} - d_0^{1.34651}) = \ln 1.05 \times 10^{11} + 1.140204 \ln t - \frac{133423.338}{RT} (0.002 Nb \text{ steel, } 1000 ^\circ C - 1150 ^\circ C) \\
\ln (d_t^{2.48654} - d_0^{2.48654}) = \ln 1.77 \times 10^{12} + 0.65397 \ln t - \frac{229593.205}{RT} (0.11 Nb \text{ steel, } 1000 ^\circ C - 1080 ^\circ C) \\
\ln (d_t^{1.55856} - d_0^{1.55856}) = \ln 6.56 \times 10^{9} + 0.76465 \ln t - \frac{215739.562}{RT} (0.11 Nb \text{ steel, } 1080 ^\circ C - 1150 ^\circ C)
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(3) The pinning effect of the NbC phase and the dragging effect of the solute atoms considerably reduced the original austenite grain size of the steel. The combined effect of the second phase strengthening of the NbC and the fine grain strengthening increased the YS of the steel, and the presence of the NbC phase led to an increase in the unevenness of the plastic deformation, thus causing decreases in the plasticity and tensile strength of the steel. The addition of Nb had a minor effect on the content of retained austenite, reduced their sizes, made their distribution more dispersed, and more effectively hindered crack propagation, thus improving the toughness of steel.

Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

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