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Effect of heat treatment on microstructure and mechanical properties of low-alloy wear-resistant steel NM450

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Abstract

In this study, we investigated the effects of the quenching and tempering temperatures on the microstructure and mechanical properties of low-alloy wear-resistant steel NM450. The quenching temperatures were 870 °C–1200 °C, and held for 36 min. After quenching with 910 °C, the tempering temperatures were 200 °C–600 °C for 60 min. The results showed that as the quenching temperature increasing, the original austenite grain size increased and the proportion of high-angle grain boundaries gradually decreased, which was detrimental to the toughness of the steel. The inter-lamellar-retained austenite remaining in the martensite lath during quenching satisfied the Kurdjumov-Saches relationship with martensite. At the tempering temperatures lower than 250 °C, carbon atoms precipitated in the form of fine carbides, which improved the yield strength of the steel. With the increasing of the tempering temperature, the activity of the carbon atoms increased, the size of the carbides increased, the effects of solid solution strengthening and precipitation strengthening weakened, so the tensile and yield strengths of the steel decreased. The low-alloy wear-resistant steel tempering at 250 °C exhibited the satisfactory combination of properties. The quasi-cleavage fracture occurred mainly over the tempering temperature range of 300 °C–400 °C, and the fracture mechanism turned into ductile fracture when the tempering temperature was higher than 500 °C.

1. Introduction

Low-alloy high-strength wear-resistant steel is widely used in the production of mining machinery, engineering machinery, and cement equipment because of its low alloy content, simple processing, good comprehensive performance, and flexible production [1–3]. Currently, the most common method of producing low-alloy high-strength wear-resistant steel plates is to add more alloying elements (such as Cr, Mo, Ni, V, etc.) and employ appropriate quenching and tempering heat treatment processes [4–7]. Mo element significantly increase the hardenability, with the Mo increasing from 0.01 wt.% to 0.31 wt.%, the hardness variations along the thickness direction from surface to the middle decreased from 16.4% to 2.6% [8]. Incorporating more alloying elements not only improves the hardenability of steel, but also contributes to fine grain and precipitation strengthening. Improving the strength and hardness of such steels while maintaining their ductility is an important research topic [9–11].

Double austenitization and quenching has been used to simultaneously improve both the strength and ductility of the steel with homogenous lath martensite, but due to a large number of precipitates dissolution, the prior austenite grain size increased [12]. Muhammad et al studied the effect of the quenching medium on the microstructural and mechanical properties [13]. The tensile strength improved 47% after brine quenching compared to the non-heat treated sample, and the Rockwell hardness also improved, but the elongation and impact toughness decreased. Tempering after quenching significantly improved the elongation and impact toughness with a little expense of strength and hardness. The effect of tempering temperature and time on microstructure and mechanical properties also has been studied [14–16]. With the tempering temperature
increasing, the hardness and strength decreased, but the elongation and impact toughness increased. With the time of tempering increasing, the carbide precipitation increased. But the systematic research on the effects of heat treatment on the microstructure and mechanical properties of low-alloy wear-resistant steel NM450 has received little attention.

In this study, the microstructure and mechanical properties of a low-alloy wear-resistant steel during heat treatment were investigated. Moreover, the effect of the original austenite grain size on the martensite substructure and high-angle grain boundary ratio during the quenching process of the wear-resistant steel were studied. The change of the microstructure morphological characteristics during the tempering process were investigated. The findings of this study will be useful for the production of wear-resistant steel with good ductility.

2. Experimental materials and methods

The used low-alloy wear-resistant steel with chemical composition 0.18 C, 0.33 Si, 1.50 Mn, 0.20 Ti, 0.22 Mo, 0.20 Cr, 0.22 Ni, bal. Fe, in wt.% was smelted in a 150 kg vacuum induction melting furnace. The slab was rolled to 12 mm by two-stage controlled rolling in the austenite recrystallization and non-recrystallization zones [17]. The initial rolling temperature in the recrystallization zone and the non-recrystallization zone was 1150 °C–1050 °C and 920 °C–880 °C, and the thickness reduction ratio in each pass was 15%–30%. The rolled slab was then air-cooled to room temperature. A Φ3 × 10 mm sample was cut from the hot-rolled steel plate, and the critical transformation temperature of the experimental steel was measured using a SETSYS Evolution-2400 thermal expansion analyzer. The critical transformation temperature from ferrite to austenite (Ac1) and the ending temperature (Ac3) during heating of the experimental steel were 780 °C and 856 °C, respectively. For hypoeutectoid steels, heating temperature should generally be carried out at 20 °C–50 °C above the Ac3 temperature in order to obtain uniform and fine austenite grains, so fine martensite structures can be obtained after quenching. If the heating temperature is lower than the Ac3 temperature, a part of the ferrite remains in the structure, and the austenite grains coarsen easily if the heating temperature is too high. Considering the influence of the alloying elements for low-alloy steels, appropriately increasing the austenitizing temperature can accelerate the austenitizing process. Therefore, the hot-rolled plate was heated in a box-type resistance furnace to different temperatures over the range of 870 °C–1200 °C, held for 36 min, and then water quenched to investigate the effect of the quenching temperature on the austenite grain size and martensite transformation. In the tempering experiment, after quenching with 910 °C the plate was heated in a box-type resistance furnace to 200 °C, 250 °C, 300 °C, 350 °C, 400 °C, 500 °C, and 600 °C for 60 min before air-cooling to room temperature.

A CMT7000 (SANS)-type microcomputer-controlled electronic universal testing machine was used to measure the strength and ductility of the experimental steel at room temperature. The gauge section of the tensile specimens was 5 mm diameter and the gauge length was 25 mm. The tensile test was followed to the GB228.1–2010 Chinese standard with the tensile speed of 1 mm min⁻¹. In the low-temperature impact toughness test, the Charpy V-type standard samples with the dimensions of 10 mm × 10 mm × 55 mm in accordance with GB229–1994 ‘Metal Charpy Notch Impact Test Method’ were machined, and the impact toughness test was carried out by the INSTRON 9250 HV drop-weight impact tester. Three specimens were tested for each steel. The macroscopic hardness of the experimental steel using an HV-50A Vickers hardness tester with a load of 10 kg and average value from 10 measurements was determined. The metallographic sample was corroded with the 4% nitric acid alcohol solution, and the microstructure was observed using a LEICAD MIRM multifunctional optical microscope (OM). An FEIQUANTA600 scanning electron microscope (SEM) was used to observe the surface and fracture morphologies of the sample. Thin film samples were prepared and thinned by double spraying with a 10% perchloric acid alcohol solution. The phase microstructures and substructure morphologies of the samples were further observed using a FEITecnaiG2F20 transmission electron microscope (TEM). The Oxford electron backscatter diffraction (EBSD) system installed on a ZEISSULTRA-55 field emission scanning electron microscope was used to analyse the grain size and grain boundary distribution of the experimental steel.

3. Results and discussion

3.1. The effect of original austenite grain size on the substructure of martensite

The microstructures of the quenched steel with different austenitizing temperatures are shown in figure 1. With the increasing of the austenitizing temperature, the original austenite grain size increased. When the austenitizing temperature was lower than 950 °C, the original austenite grains were relatively uniform, and the grain size increased from 9 to 22 μm with the increasing of the quenching temperature. When the heating
temperature was 1000 °C, the individual grain extraordinary growth, with an average grain size of 27 μm. At the heating temperatures higher than 1100 °C, the austenite grains tended to be uniform and coarsening severely. The average grain size reached 50 and 79 μm at 1100 °C and 1200 °C, respectively.

Figure 1. Original austenite grains of the experimental steel quenched at different temperatures: (a) 870 °C; (b) 910 °C; (c) 950 °C; (d) 1000 °C; (e) 1100 °C; (f) 1200 °C.

Figure 2. TEM images of the martensite laths quenched at different temperatures: (a) 910 °C; (b) 1000 °C; (c) 1100 °C; (d) 1200 °C.

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Figure 2 shows the TEM images of the martensite laths quenched at 910 °C, 1000 °C, 1100 °C, and 1200 °C. Martensite laths can be observed clearly from the images. Dislocations (with high density) were present in the substructure of the martensite laths. Rod-like or flake-like carbides were also observed in the laths. With the increasing of the austenitizing temperature, the average size of the original austenite grains increased (11, 27, 50, and 79 μm at 910 °C, 1000 °C, 1100 °C, and 1200 °C, respectively). It has been reported that in the lath martensite structure, the original austenite grains are divided into several bundles (Packet) [18, 19]. The bundles are further divided into blocks (Block), and the blocks are composed of several laths (Lath). With the increasing of the austenite grain size, the size of the bundle increased (see figure 2). It can be observed from figure 2 that the width of the martensite laths did not change significantly with the average size of the austenite grains. Speich [20]
reported that as the carbon content of low-carbon Fe-C alloy increasing to 0.05% (mass fraction), the martensite lath width decreased, but with the carbon content increasing continuously, the lath width little changed.

In addition to the blocks and martensite laths, there were several regions with large differences in the orientation of the original austenite grain. The OM images revealed that the packets with black and white contrast showed different orientations. Because the packet and block boundaries are high-angle grain boundaries, the packet and block as the basic structure unit are considered to be effective grains that can control the strength and toughness of martensitic steel. Because the width of the martensite block cannot be easily determined using an OM and SEM, it is difficult to distinguish the block and lath boundaries using a TEM. Therefore, an EBSD test system can be used to analyze the crystallographic orientation and grain boundary misorientation of the martensite microstructure.

The quenched steel had a small amount of retained austenite in addition to the martensite matrix. Figure 3(a) shows the distribution of the retained austenite in the martensite matrix quenched at 950 °C. It can be observed from figure 3 that the retained austenite was filminly and distributed along the boundary of the martensite lath. In the martensite transformation process, the transformed martensite caused the coordinated deformation of the adjacent untransformed austenite, and the volume expansion occurred when the martensite was formed, so the untransformed austenite was in a compressed state. These compressive stresses should be eliminated in order to maintain phase transition. If the transformation is hindered, a part of the austenite will be remained. A set of typical bright-field images and dark-field images are shown in figures 3(b) and (c). Selected area electron diffraction analysis was carried out on the samples to obtain a diffraction pattern. It can be observed from the diffraction patterns that the retained austenite [110] crystal belt axis was paralleled to the martensite matrix phase [111] crystal belt axis. During the quenching process, the face-centered cubic austenite undergoes shear transformation to body-centered cubic martensite. The crystallographic orientation relationship between austenite and martensite is: \( \{111\} \gamma // \{110\} \alpha , \{100\} \gamma // \{111\} \alpha \), which satisfies the Kurdjumov-Saches (K-S) relationship.

The martensite transformation occurred in a coherent shear transformation. The new martensite phase and austenite parent phase show the following crystallographic orientation relationships: the K-S, Nishiyama-Wassermann (N-W), and Greninger-Troiano (G-T) relationships [21–25]. In the K-S relationship, the following orientation relationship exists between the martensite (\( \alpha' \)) and austenite (\( \gamma \)) phases: \( \{011\} \alpha' // \{111\} \gamma ; <111> \alpha' // <011> \gamma \). In the \( \{111\} \gamma \) family of crystal planes, the martensite on each crystal plane may have six different orientations, while the \( \{111\} \) family of crystal planes of the cubic lattice may have four crystal planes, so there may be 24 equivalents of crystal orientation (variant). Blocks are generally a group of laths with the same variant.

Figure 4 shows the EBSD analysis results of the experimental steel quenched at 870 °C with the scanning step of 0.05 \( \mu \text{m} \), including the EBSD diffraction pattern quality map of this region and the corresponding crystal
orientation map, grain boundary map, and misorientation angle distribution map of this region. Taking the crystallographic coordinate system of a grain as the reference coordinate, and the angle between the adjacent grain is the misorientation angle. The colors in the orientation map of figure 4(b) represent the crystallographic orientation. It can be seen that there were several different colors in the packet, which were the variants with different orientations. The blue lines in the grain boundary map of figure 4(c) represent the high-angle grain boundaries (HAGBs) with a misorientation large than 15°, and the red lines represent the low-angle grain boundaries with a misorientation less than 15°. Combining the diffraction pattern quality map in figure 4(a), it can be observed that there were several high-angle grain boundary structures (martensite blocks) in the martensite packet. The low-angle grain boundaries were adjacent to each other in the martensite block. The block width was 0.1–3 μm, with an average width of 0.57 μm. The misorientation distribution is shown in figure 4(d). From the misorientation distribution, it can be observed that the misorientation of the high-angle grain boundary was mainly distributed at approximately 60°. It can be inferred that the block boundary misorientation of the martensite was approximately 60°. It can also be observed from figure 5(d) that the HAGBs (orientation difference $\geq 15^\circ$) were relatively large, accounting for 63.5% of the total, which is very beneficial for improving the toughness of the steel.

With the increasing of the austenitizing temperature, the austenite grain size increased, and the grain boundary structure of the lath martensite also changed. Figure 5 shows the EBSD analysis results of the experimental steel quenched at 950°C. It can be observed from the quality map shown in figure 5(a) that the original austenite grain boundaries divided the microstructure into several grains with different sizes. From the orientation map in figure 5(b) and the grain boundary structure map in figure 5(c), it can be observed that the size of the blocks in the smaller austenite grains on the left was very small, while the blocks in the larger grains on the right were significantly coarser. The maximum block width reached approximately 8 μm, and the average block width was 1.53 μm. From the distribution of misorientation in figure 5(d), the proportion of the HAGBs formed during the quenching at 950°C was 59.5%, which was slightly lower than the proportion of the HAGBs formed during the quenching at 870°C.

Figure 6 shows the EBSD analysis results of the experimental steel quenched at 1000°C. As can be observed from the grain boundary map shows in figure 6(c), the low-angle grain boundaries (red lines) distributed in the martensite block increased significantly, and the average block width increased to 2.03 μm. As shows in the misorientation distribution diagram in figure 6(d), the proportion of the HAGBs formed during the quenching at 1000°C decreased to 52.2% with the increasing of the quenching temperature. In the statistics the size of the block, the martensite structures of the original austenite grain with different sizes were compared. When the grain size was small, the width of the martensite block was small, but some large blocks could also be observed (see figure 4). When the original austenite grain size was large, the width of the block increased, but the width of the block was relatively uniform (see figure 6). Martensite preferentially nucleates at the original austenite grain.
boundaries, and the blocks are the lath regions with the same orientation. When martensite nucleates at the austenite grain boundary and grows into laths, the growth of the blocks is achieved by the nucleation and growth of the other parallel laths between the original laths. In the growth process, the martensite variants coordinate with each other to reduce the strain energy \[26\]. The laths formed in the early stages of the transformation process are relatively thick, and when a lath formed, the block containing different variants grow next to the existing laths in order to reduce the strain energy. Although the growth rate of the block was very high, the size of the packet increased with the increasing of the grain size, so there was enough time and space for the remaining microstructure to fully grow without phase transition. The size of the bundle was also small in fine austenite grains, and the remaining area after the phase transition process also decreased. Therefore, it can be observed

Figure 5. EBSD analysis results of the experimental steel quenched at 950 °C: (a) Microstructure; (b) Orientation distribution diagram; (c) Grain boundary structure diagram; (d) Orientation distribution.

Figure 6. EBSD analysis results of the experimental steel quenched at 1000 °C: (a) Microstructure; (b) Orientation distribution diagram; (c) Grain boundary structure diagram; (d) Orientation distribution.
that there was a block in the packet occupying most of the area, and the other blocks could not grow and were relatively small. These results indicated that in the martensite lath structure of the quenched steel, the HAGBs mainly included the original austenite grain boundary, martensite packet boundary and martensite block boundary, while the lath boundary belonged to the low-angle grain boundary. The grain size of the original austenite during the quenching of the experimental steel changed significantly with the increasing of the austenitizing temperature. The size and width of the martensite lath also changed significantly, while the size of the lath width did not change significantly with the original austenite grain size increasing. It can also be concluded that the austenitizing temperature directly affected the austenite grain size, which indirect affected the substructure of the martensite packets and blocks that acted as effective grains, thereby affected the strength and toughness of the steel [27, 28].

Figure 7 shows the effect of the original austenite grain size on the hardness and impact toughness of the experimental steel. With the increasing of the austenitizing temperature, the grain size increased, and the hardness and impact toughness decreased. According to the fracture mechanism [29], when a crack encountered HAGBs, the crack propagation was suppressed or changed, thereby improving the fracture strength and low-temperature toughness of the material. Owing to the large number of HAGBs, the propagation direction of the cracks frequently changed. The strengthening and toughening of the HAGBs were stronger than those of low-angle grain boundaries. Therefore, a high proportion of HAGBs in the structure can improve the low-temperature toughness of steel. With the increasing of the austenitizing temperature, the HAGB ratio decreased and the effective grain size increased, thereby reduced the hardness and impact toughness of the steel.

3.2. Effect of tempering transformation on the mechanical properties of the steel

The quenched microstructure of the experimental steel and the SEM images of the tempered microstructure at different tempering temperatures (figure 8) show different microstructure features, but the basic structure was mainly composed of the laths and carbides. As shows in figure 8, the microstructure of the experimental steel was lath martensite after quenching (figure 8(a)), because of the addition of the austenite stabilizing elements such as Cr, B, and Mo, and the laths paralleling to the packets were narrow and slender. At the tempering temperature of 250 °C, clear and sharp martensite laths could be observed (figure 8(b)). Because of carbon segregation or carbide precipitation, the brightness in some areas (mainly the lath boundaries and original austenite grain boundaries) was higher, but the precipitated carbides could not be observed clearly from the SEM images. When the tempering temperature exceeded 300 °C, the lath structure was no longer as clear and slender as that obtained at the tempering temperature of 250 °C (figures 8(c) and (d)), and the martensitic lath boundary blurred gradually. The carbides precipitated from the matrix could be observed using the SEM. The number of these precipitates increased with the increasing of the tempering temperature and showed a dispersed distribution. When the tempering temperature increased to 400 °C–500 °C, some of the laths merged and became wider and even became isometric, and the lath structure deteriorated. At the same time, carbides distributed on the substrate coarsened significantly (figures 8(e) and (f)) [30]. The changing of microstructural during the tempering process indicated that after tempering the martensite recovery of the quenched experimental steel increased, and the lath boundary underwent interatomic diffusion, aggregation, merging, and reorganization. The phase interface blurred gradually and the carbides gradually precipitated from the martensite structure and coarsened.

The effect of the tempering temperature on the strength and ductility of the experimental steel was investigated by tensile experiments. Figure 9 shows the tensile stress-strain curves of the experimental steel tempered at different temperatures. When the tempering temperature was 200 °C, a smooth stress-strain curve
with no obvious yield point and yield platform was obtained. When the tempering temperature exceeded 300 °C, the stress-strain curve started showing a yield point, and as the tempering temperature increasing continuously, the yield platform became more obviously.

The shear transformation of the martensite transformation during the quenching process produced a large number of dislocation substructures. The dislocation density was still high when the tempering was carried out at the low temperature of 200 °C, and the dislocation distribution was uneven. During the tensile deformation, the dislocations entangled in the high-density dislocation regions were difficult to move, while the dislocations

Figure 8. SEM images of the tempered steel at different temperatures: (a) Room temperature; (b) 250 °C (c) 300 °C (d) 350 °C (e) 400 °C (f) 500 °C.

Figure 9. Tensile stress-strain curves of the experimental steel at different tempering temperatures.
in the low-density dislocation regions were highly mobility and can migrate as long as the stress is sufficient. Plastic deformation started from the dislocation moving and gradually multiplication to the other positions. Thus, there was no significant yield platform on the stress-strain curve, but a continuous yield phenomenon was observed. With the increasing of the tempering temperature, the high-density dislocations recovered gradually. With the decrease of the density of the dislocations, their interactions weakened, especially the number of movable dislocations decreased significantly when they were offset or entangled with the other dislocations. At the same time, the carbide precipitation and alloys effectively pinned the dislocations and inhibited their movement\[31]. Therefore, it is necessary to release the dislocation pinning to accumulate enough movable dislocations during the tensile deformation, which requires a higher stress drive to realize the 'unpinning' of dislocations. The yield point appears on the tensile stress-strain curve, and the higher the tempering temperature, the more obvious this phenomenon was.

Figure 10 shows the relationship between the tensile properties of the experimental steel and the tempering temperature. It can be seen that the yield strength increased slightly with the increasing of the tempering temperature to 200 °C–250 °C, and the yield strength gradually decreased when the tempering temperature exceeded 250 °C. The tensile strength decreased with the increasing of the tempering temperature beyond 250 °C. The elongation curve valley appeared when the tempering temperature was 300 °C–400 °C, and the elongation significantly increased at 500 °C.

Figure 11 shows the tensile fracture morphology of the experimental steel tempered at 200 °C. As shows in figure 11(a), the fracture morphology of the steel exhibits a 'cup-cone' feature, which is the characteristic of ductile fracture. The large fiber fracture area indicates that large plastic deformation occurred. No significant cracks and inclusions were observed. From the high-magnification morphology of the fracture fiber zone shows in figure 11(b), it can be observed that the steel have a typical microporous aggregate fracture morphology with dimples of different sizes distributed in the central fracture zone (figure 11(c)). Moreover, the dimples were deep and consisted of small inclusion particles (indicated by the arrows) with the maximum size of approximately 2 μm. The result of energy spectrum (figure 11(d)) shows that the inclusions consist mainly of O, Al, S, and Mn and are identified as MnS and Al₂O₃. This was because the inclusions were incorporated into the steel during the smelting process. It produced brittle phases such as sulfides, oxides, and aluminates. In addition, because a small amount of rare earth elements with strong affinity for the remaining sulfur and oxygen during the smelting process existed in the inclusions and increased the ability of the inclusions and grain boundaries to resist the formation and propagation of cracks.

Figure 12 shows the effect of the tempering temperature on the Vickers hardness and low-temperature impact toughness of the experimental steel. As can be observed from the hardness curve (figure 12(a)), with the increasing of the tempering temperature from 200 °C to 500 °C, the average Vickers hardness of the experimental steel decreases gradually from 440 to 320 HV. However, the average hardness of the steel after tempering at 200 °C was slightly higher than that in the quenched state. As can be observed from the toughness curve (figure 12(b)), when the tempering temperature increases from room temperature to 250 °C, the impact energy of the experimental steel increases slightly, while it decreases significantly with the increasing of the tempering temperature to 350 °C. The steel exhibited temper brittleness at the tempering temperature range of 300 °C–400 °C. With the increasing of the tempering temperature to 500 °C, the impact energy increased significantly.
Figure 11. SEM images and inclusion analysis of the tensile fracture of the experimental steel after tempering at 200 °C.

Figure 12. Effect of the tempering temperature on the hardness and impact toughness of the experimental steel.

Figure 13. SEM images of the impact fracture of the experimental steel after tempering at different temperatures: (a) 200 °C; (b) 300 °C; (c) 400 °C; (d) 500 °C.
The SEM images of the steel were captured to study the impact fracture morphology at different tempering temperatures. As shown in figure 13(a), at the tempering temperature of 200 °C, the morphology of impact fracture shows the river-like quasi-cleavage surface (indicated by the white arrow) and dimples (indicated by the black arrow). With the increasing of the tempering temperature to 300 °C, the proportion of the quasi-cleavage surface on the impact section increased, while the number of dimples decreased (figure 13(b)). At the tempering temperature of 400 °C, the morphology of impact fracture mainly shows quasi-cleavage fracture with only a few scattered dimples (figure 13(c)). When the tempering temperature increased to 500 °C, ductile fracture occurred, and the morphology of impact fracture shows a typical dimple structure with a more tearing ridge (figure 13(d)), indicating that high energy was absorbed during the passivation and propagation of the crack tip in the crack propagation process. This can mainly be attributed to the fact that the interlaminar carbides formed by the decomposition of retained austenite causes embrittlement and that carbides precipitate easily along the original austenite grain boundaries, causing intergranular embrittlement [32, 33]. In addition, the segregation of impurity elements such as P in the original austenite grain boundary weakens the binding energy of the grain boundary, which also causes embrittlement.

4. Conclusion

1. When the experimental steel was fully austenitized, with the increasing of the quenching temperature, the original austenite grain size increased, thereby the proportion of high-angle grain boundaries reduced, which was not conducive to the toughness of the experimental steel. At the same time, with the increasing of the quenching temperature, the block size in the martensite also increased, the strength and hardness of the experimental steel reduced. However, the changing of the original austenite grain size did not affect the width of the martensite lath significantly.

2. When the experimental steel was tempered at low temperatures, short-range diffusion of interstitial carbon atoms and internal stress release were occurred. With the increasing of the tempering temperature, the segregation of carbon atoms continued along the original austenite and lath martensite interface to form cementite. During the early stage of carbide formation, the release of the internal stress and the formation of fine carbides were conducive to improving the fracture toughness of the experimental steel while maintaining its hardness. With the increasing of the tempering temperature, the coarsening and growth of carbides and the reduction of the dislocation density reduced the hardness and toughness of the experimental steel.

3. The fracture mechanism during the low-temperature tempering was the combination of quasi-cleavage and ductile fractures. The quasi-cleavage fracture occurred mainly over the tempering temperature range of 300 °C–400 °C (tempering embrittlement temperature), and the fracture mechanism turned into ductile fracture when the tempering temperature was higher than 500 °C. This temper embrittlement can mainly be attributed to the fact that the interlaminar carbides formed by the decomposition of retained austenite cause embrittlement and that carbides easily precipitate along the original austenite grain boundaries, causing intergranular embrittlement. In addition, the segregation of impurity elements such as P in the original austenite grain boundary reduced the binding energy of the grain boundary, which also caused embrittlement.

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