Effects of recrystallization annealing on microstructure and mechanical properties of low-carbon air-hardening steel LH800

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Abstract
The process of recrystallization annealing affects grain size and morphology, and thus, plays a vital role in tailoring the mechanical properties of the air-hardening steel LH800. In this paper, we investigated the recrystallization behavior, microstructure evolution, and mechanical properties of cold-rolled LH800 steel during the batch annealing process. We also studied the relationship between the recrystallization behavior and the microstructure as well as the mechanical properties, in obtaining air-hardening steel with good cold formability. The results show that when annealing at 700 °C with extended holding time, the cold-rolled deformed structure recrystallizes gradually, a large number of nanoscale carbides are formed and dispersed in the matrix, and the bimodal microstructure of the coarse and fine grain ferrite is distributed in lamellae along the rolling direction. With an increase in holding time, the volume fraction of nano carbide decreases, bimodal structure disappears gradually, and the ferrite grain size tends to be uniform. The air-hardening steel LH800 has lower yield strength, highest elongation, and best cold forming performance when annealed at 700 °C for 4 h. The essence of the yield point elongation of the LH800 during tensile testing is revealed, based on the evolution of recrystallized microstructure and the change of nano-scale carbides in the ferrite matrix during the annealing process.

1. Introduction

In recent years, development trends in the automobile industry have shifted towards achieving the twin goals of energy conservation and environmental protection, and toward this direction, lightweight automobiles have emerged as one of the most effective measures to reduce pollution and save energy. The automotive industry has been increasingly utilizing advanced high strength steels, which not only meet the requirements of body safety parameters, but also reduce the body weight of the automobile [1–5]. Among the types of high-strength steel, low-alloy high-strength steel (HSLA) is widely used due to its low cost, high strength, high work hardening rate, and good formability [6–11]. The high strength of HSLA is attributed to features of its microstructure, such as fine grain strengthening [12], second phase strengthening [13], and inclusion shape control [14]. Air-hardening steel LH800 is a kind of low-alloy high-strength steel with good cold-forming properties in the annealed state after undergoing the cold rolling process. After the cold-forming process, the sheet is subjected to heat treatment with austenitization and quenching in air, to enhance its final strength. The final tensile strength of the air-hardening steel is greater than 1000 MPa and elongation is greater than 13%, which meets the design requirements of new-generation automotive steel. The production process of air-hardening steel is simple, highly efficient, low cost, and has high market application value [15–18].

In fabricating all kinds of final product parts, LH800 is first cold rolled into sheets. However, the cold-rolling process can exert a detrimental influence on the microstructure and mechanical properties of steels. The cold-rolling process of LH800 results in undesirable hardening, and this increases the difficulty in further processing. To eliminate the residual stress of the cold-rolled sheet, it is usually subjected to subsequent annealing treatment.
However, low carbon steel is prone to yield phenomenon during annealing. Yield point elongation is a discontinuous plastic deformation that has a serious detrimental effect on the surface quality and properties of steel plates [19]. Li et al [20] studied the annealing softening behavior of low-carbon steel with a dual-phase structure and proposed the mechanism of bimodal structure in the annealing process. In a study of microstructure and properties of bimodal ultra-fine grain ferritic steel, Niu [21] suggested that the bimodal structure is evolved from deformed martensite and ferrite during the annealing process. The authors opined that the appearance of the yield platform during the annealing process was mainly due to a large number of nano-scale carbides.

Cold-rolled air-hardening steel sheets need to be appropriately annealed to obtain good cold formability and to ensure complex auto parts can be produced at room temperature. However, the recrystallization and microstructure of the LH800 during batch annealing have not been systematically investigated. We studied the recrystallization behavior of cold-rolled LH800 during the batch annealing process and its effects on the microstructure and mechanical properties, and further explored the mechanism of the yield point elongation of LH800, based on the microstructure evolution.

2. Materials and experimental methods

The experimental steel used was a 25 kg steel ingot melted in a vacuum induction melting furnace, its chemical composition is shown in table 1. The ingot was heated and forged into a billet with a size of 30 mm × 80 mm × 150 mm. The steel was then reheated to 1200 °C and held for 2 h to ensure homogenization. This was followed by 7 passes of hot rolling with a final rolling temperature of 900 °C. The total reduction ratio of hot rolling was 83.3% and the final thickness of the hot-rolled plate was 5 mm. After the hot rolling process, the steel was subjected to a simulated coiling process with a holding time of 2 h at 700 °C and cooled to room temperature in the furnace. The cold rolling process, with a total reduction of 70%, was conducted using a straight-drawn four-high reversible cold-rolling mill to acquire a sheet with the final thickness of 1.5 mm.

To design an appropriate heat treatment process, the critical temperatures of Ac1, Ac3, and Mf of the experimental steel were first measured using a dilatometer. The thermal expansion curve is shown in figure 1. The measured temperatures of Ac1, Ac3, and Mf were 725 °C, 850 °C, and 450 °C respectively. The recrystallization annealing, schematized in figure 2, was then designed with the annealing temperature set at below the temperature of Ac1, to study the recrystallization kinetics of the cold rolled sheet. The recrystallization volume fraction (X) after annealing was determined by the hardness method [20]:

\[ X = \frac{H_0 - H}{H_0 - H_{\text{Rex}}} \]

Where: X is the recrystallization volume fraction of ferrite, H0 is the initial microhardness of ferrite in cold-rolled steel, H is the microhardness of the sample after the given annealing conditions, and HRe is the microhardness corresponding to the completely recrystallized sample. The average value of 5 hardness values on the cross section of the sample was used in the equation, to determine the recrystallization volume fraction.

After the recrystallization annealing treatment, tensile test samples (gauge length: 25 mm; width: 6 mm; thickness: 1.2 mm) were prepared along the rolling direction using a wire cutting machine. Uniaxial tensile tests were conducted with a deformation rate of 1 mm min⁻¹ on an electronic universal testing machine (CMT 5105, SANS, Shenzhen, China). The microstructure of the samples was observed using a laser confocal microscope (LOM, LEXT OLS4100, Olympus, Tokyo, Japan), a field emission environmental scanning electron microscope (SEM, Quanta FEG 450, FEI Company, Hillsboro, USA) together with an electron backscatter diffraction (EBSD) device, and a field emission transmission electron microscope (FEI Tecnai F20, Hillsboro, USA).

Samples for LOM and SEM were corroded with 4 vol% nitric acid alcohol after grinding and polishing. Samples for EBSD were prepared using electropolishing at a voltage of 15 V for 15 seconds, the electrolyte contained 20% perchloric acid and 80% ethanol. Samples for TEM were initially ground with SiC sandpaper to a thickness of about 50 μm, and then a disc was punched out with a diameter of 8 mm. Finally, electrolytic double spray was done at −22 °C using a solution containing 85 vol% C₂H₅OH and 15 vol% HClO₄.
3. Results and discussion

3.1. Initial microstructure of LH800

Figure 3 shows the initial microstructure of the experimental steel in the hot rolled and cold rolled state. As can be seen from figure 3(a), the hot-rolled microstructure of the LH800 consists of ferrite (F) and pearlite (P), with lamellar distribution along the rolling direction. As shown in the dotted frame in the picture, the dark area is pearlite and the white area is ferrite, with no fixed width and spacing between the pearlite and ferrite regions. The cold rolled structure of the experimental steel is shown in figure 3(b). A deformed fibrous structure is visible and the grains are elongated along the rolling direction.

3.2. Recrystallization behavior of LH800

Figure 4(a) shows the hardness-time curve of cold-rolled LH800 after annealing at different temperatures. At a given annealing temperature, the hardness gradually decreases with the increase in annealing time. For each given annealing temperature, the overall trend of decrease in hardness was similar. There was a sharp early decline followed by a gentle decline. When the annealing time was 5 min, the microhardness values of LH800 after annealing at 600 °C, 650 °C, and 700 °C were 282.13, 255.33, and 227.33 (HV0.5kg), respectively. As
expected, in the beginning, the higher the annealing temperature, the greater was the decrease in hardness of the cold-rolled LH800. Figure 4(b) shows the relationship between LH800 ferrite recrystallization volume fraction and annealing time. The variation trend of recrystallization volume fraction is consistent at different annealing temperatures, and all are typical 'S' type. There was an incubation time at the beginning of the recrystallization annealing and then the recrystallization rate increased gradually till it reached a certain approximate constant speed, and finally the rate decreased gradually. With the increase in annealing temperature, the incubation period of recrystallization became shorter. The recrystallization volume fraction increased rapidly and reached a stable state faster. That is, it took a shorter time to reach complete recrystallization.

From the Avrami equation: [22]

$$X = 1 - e^{-Kt^n}$$  \hspace{1cm} (2)

where $K$ and $n$ are constants. The following equation can be determined:

$$\ln\ln\left(\frac{1}{1-X}\right) = \ln K + n\ln t$$  \hspace{1cm} (3)

When we draw a fitting curve with $\ln\ln[1/(1-X)]$ and $\ln t$ on the coordinate graph, we get a straight line, its slope is $n$, intercept is $\ln K$. The fitting curve is shown in figure 5, where we can observe that with the increase in annealing temperature, the slope $n$ increases gradually.

The kinetics equation of ferrite recrystallization of LH800 during annealing was calculated based on the JMAK model. The kinetic equations of ferrite recrystallization for the cold-rolled LH800 annealed at 600 °C, 650
$600^\circ C$ and $700^\circ C$, are shown in equations (4)–(6), respectively:

$$X = 1 - e^{-0.17t^{0.206}}$$  \hspace{1cm} (4)  

$$X = 1 - e^{-0.23t^{0.223}}$$  \hspace{1cm} (5)  

$$X = 1 - e^{-0.23t^{0.291}}$$  \hspace{1cm} (6)  

Where: $X$ is the recrystallization volume fraction of ferrite; $t$ is the holding time;

According to the recrystallization kinetic curve, as the annealing temperature increased below the austenite transition temperature, the incubation period of recrystallization shortened. The recrystallized volume fraction increased rapidly and entered the completely recrystallized state earlier than expected. In other words, the higher the temperature, the more sufficient is the recrystallization under the same holding time. When annealed above $\text{Ac}_1$, austenite transformation occurs. In the subsequent cooling process, a martensite structure with high hardness is obtained, which is not conducive to subsequent further processing [23, 24]. We conclude that annealing at $700^\circ C$ has the best recrystallization effect during the annealing process.

### 3.3. Microstructure evolution characteristics during annealing of LH800

Based on the above results, LH800 exhibits good recrystallization effect when the annealing temperature is $700^\circ C$. In this section, the effect of holding time on microstructure and precipitates at this annealing temperature is investigated in depth. Figure 6 shows the microstructure evolution of the cold-rolled LH800 after annealing at $700^\circ C$ and holding for different time periods. When the holding time was 0.25 h, 0.5 h, 1 h, 2 h, 4 h, and 6 h, the corresponding average grain sizes were 5.25 $\mu m$, 5.67 $\mu m$, 6.13 $\mu m$, 6.45 $\mu m$, 7.26 $\mu m$, and 7.95 $\mu m$, respectively. The average grain size of ferrite increased slightly with the increase in holding time. When the holding time was between 0.25–2 h, a large number of nano-sized carbides were dispersed in the ferrite matrix and coarse carbides were distributed in grain boundaries. When the holding time was 0.25 h, the ferrite grain boundary was not clear, and the microstructure was clearly elongated along the rolling direction. When the holding time was 4 h and above, the inside of the ferrite grain became purer. The nano-scale carbides inside the ferrite grain decreased gradually, and the coarse carbides at the grain boundary increased slightly. The ferrite grains that were elongated along the rolling direction were gradually replaced by equiaxed or polygonal ferrite.

Figure 7 shows the EBSD-characterized grain boundaries of the cold-rolled and the representative annealed specimens. The blue lines represent high-angle grain boundaries and the red lines represent low-angle grain boundaries or sub-grain boundaries. A large number of sub-grains exist in the cold rolled specimens, as shown in figure 7(a). This indicates that the original ferrite and pearlite grains were broken into fine sub-grains during the cold rolling process. There are also large white areas distributed in the cold-rolled specimens. This could be caused by the ferrite grains that have better plasticity, being elongated during the cold rolling process. With the increase in annealing time, the sub-grain gradually disappeared in the original ferrite region, but still existed in the original pearlite region, as shown in figures 7(b) and (c). This shows that the nano-sized carbide particles had

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**Figure 5.** Avrami plots of the kinetic function $\ln [1/(1-X)]$ versus time.
a strong pinning effect on the sub-grain boundary under certain conditions, thereby delaying the coarsening of ferrite grains. The fraction of the area that was occupied by the sub-grains gradually decreased with the further increase in annealing time. Finally, the fraction of the area occupied by sub-grains was almost zero when annealed for 6 h at 700 °C.

The variation in distribution of grain boundary misorientation reflects the characteristic evolution of the microstructure of the material during the annealing process. The grain boundary misorientation distribution is shown in figure 8. The volume fraction of the low-angle grain boundary of the experimental steel is 68.05% and the average grain boundary angle is 16.77° in the cold-rolled state. The annealing time is 0.5 h, 2 h, and 4 h, for which the corresponding low-angle grain boundary volume fractions are: 26.13%, 14.35%, and 5.69%, respectively, and the corresponding average grain boundary angles are: 31.35°, 36.56°, and 40.15°, respectively. The number of small-angle grain boundaries decreases significantly, and the average grain boundary angle gradually increases with the increase in annealing time. It was also observed that the sub-grains grow during the subsequent annealing process. The sub-grains merge to form low-angle grain boundaries through the movement of dislocations, and with the migration of low-angle grain boundaries, and are eventually replaced by high-angle grain boundaries [25, 26].

Figure 9 is the inverse pole figure (IPF) of LH800 cold-rolled and annealed at 700 °C for 4 h. As can be seen from the figure, the initial microstructure of the experimental steel is distributed in lamella along the rolling direction. The sub-grain orientations distributed in lamella in large areas are basically the same. The ferrite grains are clearly visible and equiaxed or polygonal, and the grain orientations are randomly distributed. Figure 10 shows the Kernel average misorientation (KAM) of LH800 cold-rolled and annealed at 700 °C for 4 h.
Studies have shown that [27], KAM can qualitatively reflect the uniformity of plastic deformation. A higher value indicates a greater degree of plastic deformation or a higher density of defects. In this study, the color of the KAM map was relatively uniform and the KAM value was large. There were also local areas with different colors and the KAM value was particularly large in the cold rolled state. That the overall deformation of the experimental steel was relatively uniform, the degree of deformation was relatively large, and the local deformation was particularly large during the cold rolling process. The color of the KAM map of the microstructure was the same, and the KAM value was low when annealed at 700 °C for 4 h. It shows that ferrite recrystallization was sufficient and uniform during this annealing process.

Two kinds of carbides are mainly distributed in the matrix of cold-rolled LH800 annealed at 700 °C, as shown in figure 6—course carbides distributed at the grain boundaries and nano-sized carbides dispersed in the ferrite matrix. Based on the combination of the energy spectrum and diffraction spots, the course carbides are the M23C6 type carbide, with size between 0.5–2 μm, with no morphological rules. The coarse carbides increased slightly and remained at the grain boundaries with the increase in holding time. Figure 11 shows the morphology and energy spectrum of different sizes of nano-sized carbides and the corresponding diffraction spots, and accordingly, the nano-sized carbides are all M23C6 type. M mainly refers to the Cr element—some Cr elements are replaced by Fe, Mn, Mo, V, and other elements. The chemical formula of the nano-sized carbides is (Fe,Mn,Cr,Mo,V)23C6, with size between 20–300 nm, they are mainly distributed around 100 nm, which presented two kinds of morphology—spherical and ellipsoidal. The number of the nano-sized carbides gradually decreased with the increase in the holding time.

Figure 12 shows the variation of precipitates with changes in temperature in the experimental steel under equilibrium condition, calculated using Thermo-Calc thermodynamics software. Two kinds of M23C6 carbides are observed in the figure under the equilibrium condition of the experimental steel. It is confirmed that the chemical formulas of M23C6#1 and M23C6#2 are both (Fe,Mn,Cr,Mo,V)23C6, but the alloy composition is different. In addition, unknown phase FCC_A1#2 appeared in the precipitate, which is confirmed as
It can be seen from figure 12 that $\text{M}_{23}\text{C}_6$ begins to form at about 500 °C, and its content reaches the highest level at about 325 °C, after which it significantly decreases to zero in the range of $275 \sim 325$ °C. $\text{M}_{23}\text{C}_6$ began to precipitate at about 435 °C, and its content increased significantly to 1.2% in the range of $400 \sim 435$ °C, then remained unchanged in the range of $315 \sim 400$ °C. When the temperature continued to decrease, its content increased further, and reached the maximum level at about 265 °C. Therefore, the thermodynamic calculation shows that the precipitation temperature of $\text{M}_{23}\text{C}_6$ carbide is below 500 °C. The analysis of the precipitated phase in figure 11 shows that the carbides obtained are $(\text{Fe},\text{Mn},\text{Cr},\text{Mo},\text{V})_2\text{C}_6$ for different holding time at 700 °C, which is inconsistent with the thermodynamic equilibrium calculation. This is

Figure 8. EBSD grain boundaries and their misorientation distributions in the cold-rolled and representative annealed specimens.

Figure 9. Inverse pole figure (IPF) of LH800 when cold-rolled and annealed for 4 h at 700 °C.

$(\text{Fe},\text{Mn},\text{Cr},\text{Mo},\text{V})_2(\text{B},\text{C})_1$. It can be seen from figure 12 that $\text{M}_{23}\text{C}_6$ begins to form at about 500 °C, and its content reaches the highest level at about 325 °C, after which it significantly decreases to zero in the range of $275 \sim 325$ °C. $\text{M}_{23}\text{C}_6$ began to precipitate at about 435 °C, and its content increased significantly to 1.2% in the range of $400 \sim 435$ °C, then remained unchanged in the range of $315 \sim 400$ °C. When the temperature continued to decrease, its content increased further, and reached the maximum level at about 265 °C. Therefore, the thermodynamic calculation shows that the precipitation temperature of $\text{M}_{23}\text{C}_6$ carbide is below 500 °C. The analysis of the precipitated phase in figure 11 shows that the carbides obtained are $(\text{Fe},\text{Mn},\text{Cr},\text{Mo},\text{V})_2\text{C}_6$, for different holding time at 700 °C, which is inconsistent with the thermodynamic equilibrium calculation. This is
because the annealing process is completed under non-equilibrium conditions, and the thermodynamic calculations are performed under super-equilibrium conditions, which leads to deviations between the experimental results and the thermodynamic calculations.

At the beginning of the annealing process, the main driving force of recrystallization comes from the distortion energy stored during cold rolling [28–30]. During the cold rolling process, as the ferrite zone is softer, the amount of deformation produced in the ferrite zone is greater. Therefore, the distortion energy stored in the...
The ferrite region is higher and the driving force for recrystallization is greater. However, there are more grain boundaries in the pearlite region and more recrystallization nucleation points. Therefore, the ferrite and pearlite regions form fine-grained structures at the beginning of the cold-rolled LH800 annealing process.

Many experimental studies [31–34] have shown that the effect of the second phase particles on the growth rate of grains is related to the second phase particle radius (r) and the number of second phase particles per unit volume (volume fraction ϕ). When reaching equilibrium, the stable grain size d has the following relationship with r and ϕ:

\[ d = \frac{4r}{3\varphi} \]  

The grain size is directly proportional to the second phase particle radius and inversely proportional to the second phase particle volume fraction. In other words, the smaller the second phase particles and more the quantity, the stronger the ability to hinder the growth of the grains, and smaller the grains.

At the beginning of the annealing process, there are more nano-sized precipitates in the pearlite region. These precipitates prevent the migration of grain boundaries and affect the growth of recrystallized grains. Later, the pearlite region gradually transforms to the fine grain region. However, since there are fewer nano-scale precipitates in the original ferrite region, the grain growth resistance is feeble, hence, these grains grow rapidly and transform to the coarse grain region. Finally, a ferrite bimodal structure with coarse and fine grain distribution is formed. As shown in figure 7(b), the bimodal structure is visible when annealed at 700 °C for 0.5 h. At the same time, the fine-grained region and the coarse-grained region have a lamellar distribution, which is consistent with the distribution of cold-rolled pearlite and ferrite. Its width, spacing, and volume fraction are close to the original pearlite and ferrite region.

The main driving force for grain growth after recrystallization is the reduction of grain surface energy. The larger the grain size, the lower the surface energy [35–37]. Therefore, the growth of fine grains into coarse grains is a spontaneous process that reduces the free energy of metals. As the grain size in the fine grain region is smaller than that in the coarse grain region, the driving force of grain growth per unit volume in the fine grain region is larger than that in the coarse grain region. In addition, with the increase in holding time, the nano carbides in the fine grain region grow and dissolve back. This promotes the increase in the particle radius of the precipitates and a decrease in the number of particles per unit volume. As a result, the ability to hinder the migration of grain boundaries decreases. Based on the above two factors, the grain growth rate in the fine-grained region is greater than that in the coarse-grained region. In our study, with the increase in annealing time, the difference in grain size between the two regions became smaller and smaller, and the bimodal structure gradually disappeared, as shown in figure 7(c). With the further increase in annealing time, the grain size was uniform and the bimodal structure disappeared, as shown in figure 7(d).
3.4. The effect of annealing process on the mechanical properties of LH800

Figure 13(a) shows the room temperature tensile curve of the cold-rolled LH800 annealed at 700 °C for different time. The tensile curves of the annealed specimens at room temperature are similar and all have yield platforms. The yield platform length decreases with the increase in holding time. Studies have shown that the reduction of grain size is beneficial to the generation of yield plateaus [19, 21, 38, 39]. The length of Lüders strain ($\varepsilon_L$) and the grain size ($d$) satisfy the following relationship:

$$\varepsilon_L = \lambda d^{-0.55} \quad (8)$$

$\lambda$ is a material-related constant, according to the formula, the length of the Lüders strain reduced with the increase in grain size.

As shown in figure 13(a), when the annealing time is 0.5 h, the Lüders strain is 3.8%, and the average grain size is 5.67 μm. When the annealing time is 2 h, the Lüders strain is 3.4%, and the average grain size is 6.45 μm. When the annealing time is further extended to 4 h, the Lüders strain is reduced to 2.4%, and the average grain size is 7.26 μm. With the increase in annealing time, the average grain size gradually increases and the length of Lüders strain gradually decreases, and this is in perfect agreement with the above formula (8).

As shown in figure 5, when the annealing time is 0.5 h, a large number of nano-sized carbides are dispersed in the ferrite matrix, and coarse carbides are distributed in the grain boundaries. When the annealing time is extended to 2 h, the nano-sized carbides in the matrix gradually decrease, and the grain boundary carbides increase slightly. When the annealing time is further extended to 4 h, the nano-sized carbides in the matrix continue to decrease and the grain boundary carbides change a little. However, when the annealing time is between 0.5h-4h, with the increase in holding time, the nano-scale carbide decrease and there is more C solute in the crystal lattice. In this case, the pinning effect of the interstitial atoms is stronger. However, the yielding platform gradually decreases. This shows that the role of the precipitated phase is greater than that of the solid solution atoms. Thus, that the appearance of the yield platform of the annealed specimens is mainly related to the number of dispersed nano-scale carbides in the ferrite matrix.

In summary, the emergence of the annealed LH800 yield plateau was mainly caused by fine recrystallized ferrite grains and a large number of nano-sized carbides. With increasing annealing time, the Lüders strain can be reduced or even eliminated.

Figure 13(b) shows the mechanical properties of the cold-rolled LH800 annealed at 700 °C for different time. With the increase in the holding time, the yield and tensile strength of the annealed specimens decrease gradually, and the elongation increases first and then decreases. It is known that the mechanical properties of materials are closely related to their microstructure. It can be seen from figure 6 that the cold-rolled LH800 has a ferrite single-phase structure after annealing. As the holding time increases, the ferrite grain size gradually increases. According to the Hall-Petch relation, the larger the grain size, the lower is the yield strength. With the increase in holding time, the ferrite matrix nano-sized carbide particles gradually decrease and the precipitation strengthening effect produced by nano-sized carbides weakens gradually. Based on these two reasons, when the cold-rolled LH800 is annealed at 700 °C, the yield strength gradually decreases with the increase in the holding time. It can be seen from the recrystallization kinetics curve in figure 4(b) that when the holding time is between 0.25−4 h, the recrystallization of the annealed LH800 is not complete. With the increase in holding time, recrystallization is more sufficient. However, the grain size continues to gradually increase, but the recrystallization is dominant. Therefore, when the holding time is between 0.25−4 h, the elongation increases gradually. When the holding time is between 4−6 h, the recrystallization of the annealed LH800 is complete.
However, as the grain size continues to increase, the elongation begins to decrease again. In summary, when annealing at 700 °C for 4 h, LH800 has lower yield strength and highest elongation, which means it has the best forming performance.

4. Conclusion

(1) We obtained the ferrite recrystallization kinetic equation of LH800 under different annealing processes based on the JMAK model, and plotted its recrystallization kinetic curve. The value of its Avrami index n is between 0.2 and 0.3. With the increase in annealing temperature, the time required for complete recrystallization of ferrite is greatly shortened. Annealing at 700 °C had the best recrystallization effect.

(2) When annealed at 700 °C, the cold rolled deformed ferrite and pearlite of LH800 gradually recrystallized with the increase in holding time. When the holding time is 0.5 h, the matrix has a bimodal structure: fine-grain and coarse-grain regions, and the microstructure has a lamellar distribution along the rolling direction, consistent with the distribution of cold-rolled pearlite and ferrite. The width, spacing, and occupied volume fraction are close to the original pearlite and ferrite regions. The fine grain region is transformed from the original pearlite region, and the coarse grain region is transformed from the original ferrite region.

(3) Annealing at 700 °C for 0.25–2 h, a large number of nanoscale precipitates are dispersed in the LH800 matrix, and a bimodal structure made of coarse and fine grains appears. Due to the strengthening of the second phase and the fine grains, the yield strength is higher and not conducive to further forming of the material. With the increase in holding time, the nano precipitates in the matrix decrease gradually, the coarse and fine-grained bimodal structure disappears, and ferrite grain size becomes more uniform. When holding for 4 h, the yield strength is lower, the elongation is the highest, and it has the best cold forming performance.

(4) With the increase in annealing time, the average grain size of ferrite gradually increases. The nano-sized carbides in the matrix gradually decrease, and the coarse carbides at the grain boundaries remain unchanged. The yield plateau length of annealed tensile specimens decreases gradually. The appearance of the annealed LH800 yield platform is mainly caused by fine recrystallized ferrite and a large number of dispersed nano-sized carbides. We anticipate that the Lüders strain can be further reduced or even eliminated by increasing the annealing time.

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Data availability statement

The data generated and/or analysed during the current study are not publicly available for legal/ethical reasons but are available from the corresponding author on reasonable request.

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