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Annealing treatment for improving the mechanical properties of 00Cr18Nb ferritic stainless steel prepared by investment casting

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Abstract

The as-cast and annealed mechanical properties of 00Cr18Nb ferritic stainless steel prepared by investment casting process were studied. Experimental results indicated that the problem of unstable plastic fluctuations in the as-cast state can be effectively solved by annealing treatment. The plasticity of the investment-cast 00Cr18Nb steel after annealing at 800 and 850 °C was more stable than that of 900 °C. It showed that the optimum annealing condition was 850 °C for 2 h with the appropriate elongation and tensile strength of 26.3 ± 0.4% and 513 ± 4 MPa, respectively. The main reason for annealing to improve plasticity is the Laves phase in the microstructure, which has better effects on mechanical properties than (Ti,Nb)(C,N) particles and grain size. With the increase of annealing temperature and time, the size of Laves phase tended to increase. The size, quantity, morphology and distribution of Laves phase after annealing are closely related to the mechanical properties. The Laves phase with a large amount of uniform distribution in the form of small particles can effectively improve the mechanical properties after annealing treatment.

1. Introduction

Ferritic stainless steel has high thermal conductivity, low expansion coefficient, good oxidation resistance, excellent stress corrosion resistance, and is relatively cheap. It is often used for heat exchangers and automobile exhaust components [1]. The grade of 00Cr18Nb, which is a plastic-deformation ferritic stainless steel (prepared by rolling or forging process), has mainly been focused on its formability properties [2], high-temperature behavior [3], interconnects material in solid oxide fuel cells [4], corrosion performance [5], and oxidation behavior [6]. However, the as-cast properties of 00Cr18Nb plastic-deformation steel, especially its investment casting properties, have not been well investigated. Furthermore, in consideration of the good weldability of 00Cr18Nb ferritic stainless steel and reducing the production cost, a domestic investment casting foundry has recently tried to produce exhaust-system castings of 00Cr18Nb steel by investment casting process.

The actual investigation of the as-cast mechanical properties of the investment-cast 00Cr18Nb steel showed that the elongations were less than 5% from time to time, indicating the highly unstable fluctuations of the plasticity. For example, the lowest elongations of the investment-cast 00Cr18Nb steel were 4% and 4.7% respectively, and the number of the lower elongations increased with increasing niobium (Nb) contents. The unstable fluctuations of the plasticity seriously affect the safety application of the investment-cast 00Cr18Nb steel. Therefore, the as-cast plasticity of the investment-cast 00Cr18Nb steel needs to be improved. In addition, while the plasticity is improved, the strength should also be enhanced. Annealing treatment is an appropriate choice for improving the mechanical properties [7] of the investment-cast 00Cr18Nb steel.

Referring to the literatures, there are relatively few researches on the mechanical properties and annealing treatment of the investment-cast 00Cr18Nb steel. Kang et al [8] investigated 00Cr18Nb (UNS S44100) steel prepared by investment casting process, but focused on normalizing treatment and the individual effects of titanium (Ti) content on the microstructure. A systematic study of the as-cast performance of 00Cr18Nb and the annealing effects are therefore needed. Investment-cast 00Cr18Nb steel contains both Ti and Nb, therefore...
studies of annealing on Ti–Nb microalloyed stainless steels are relevant. Fujita et al. [9] subjected several ferritic stainless steels (Nb: 0.0–0.6 wt%; Ti: 0.0–0.6 wt%) to hot- and cool-rolling processes and annealing at temperatures between 800 and 950 °C for 1 min. Their study showed that Ti–Nb dual stabilization after annealing was better than single stabilization of Ti or Nb. Zhang et al. [10] investigated the effects of Ti and Nb and annealing on forged hot-rolled low-carbon ferritic stainless steels containing 0.12–0.14 wt% Ti and 0.13–0.15 wt% Nb. This investigation indicated that the annealing treatment and lower Ti and Nb contents were conducive to improve the plasticity. In addition, the specific grade steels containing Ti and Nb were also investigated. Maruma et al. [11] studied the effects of annealing at 900, 950 and 1025 °C for 180 s on AISI 441 steel (Ti: 0.15 wt%; Nb: 0.44 wt%). The AISI 441 investigated in their research was prepared by rolling process, and the results showed that annealing temperature above 900 °C can have an effect on the deep-drawing capability. Similarly, the high temperature annealing (950 °C–1010 °C) on the AISI 444 samples (0.13 wt% Ti and 0.20 wt% Nb) was studied by Abreu et al. [12]. It showed that the high temperature annealing was not suitable for improving room-temperature toughness. The above studies of the effects of annealing on the microstructure and properties of Ti–Nb microalloyed stainless steels provide information relevant to the investment-cast 00Cr18Nb steel.

This study aims to investigate the investment-cast performance of 00Cr18Nb steel and discuss the effects of annealing on the microstructure and mechanical properties. Appropriate annealing condition for improving the mechanical properties is analyzed.

### 2. Experimental procedure

Low carbon ferritic stainless steel was melted as matrix in an XY-150 medium-frequency induction furnace (Xinyu Mechanical and Electrical Company, Xiamen, China). To minimize Ti oxidation during smelting process, a gas protection device was installed. The Ti content was added and effectively controlled to 0.1–0.2 wt%. The temperature of the molten steel was measured with an SWD-3 digital thermometer (Xinanjiang Nihong Casting Instrument Company, Jiande, China), and its composition was controlled by using a Metal Lab 75/80 spectrum direct-reading instrument (Zhujin Analytical Instruments Company, Shanghai, China). The Nb content of the molten steel was controlled to 0.4–0.5 wt%; the main components are listed in Table 1. The temperature of the molten steel before pouring was 1600 °C. The mold shell sample, which is shown in figure 1(a), was pre-heated for 2 h at 1200 °C. The molten steel was poured into the heated shells and then were placed vertically at approximately 90° at equal spacings for air cooling. Investment-cast samples of 00Cr18Nb were obtained by removing the mold shells, as shown in figure 1(b). Each group of casting samples was cut into round bars and were heat-treated in a KXS 2-10–16 rapid-heating furnace (Changcheng Electric Furnace Company, Changsha, China). The annealing temperatures were 800, 850 and 900 °C and the annealing times were 1, 2 and 3 h, respectively.

Considering that the heat-treated round bars will be cooled in the furnace under 340 °C–516 °C [13], which may produce a small amount of chromium-rich α’ phase precipitation, causing brittleness at 475 °C [14]. The elongation and the toughness of ferritic stainless steel are significantly reduced, which will affect the results of the experiment. Therefore, during the cooling process of the annealing treatment, these round bars should be avoided to stay for a long time in the range of 340 °C–516 °C. When the furnace cooling temperature reduced to

![Figure 1. Experiment mold and casting: (a) mold shell sample; (b) investment-cast sample.](image)

| C  | N  | Ti | Nb | Cr | Si | Mn | P  | S  | Fe |
|----|----|----|----|----|----|----|----|----|----|
| 0.025 | 0.023 | 0.16 | 0.44 | 18.6 | 0.79 | 0.58 | 0.025 | 0.006 | Bal. |

Table 1. Main compositions of the investment-cast 00Cr18Nb steel (wt%).
$500 \pm 10 \degree C$, the annealed round bars were taken out from the furnace for air cooling to increase the cooling rate, and further eliminate the effect of brittleness at $475 \degree C$.

The correlation analysis about sampling position, microstructure and mechnical performance was carried out before the experiment. Three of the four as-cast round bars in the same casting group were selected, and the upper, middle and lower parts of the three round bar were processed into tensile bars. The results showed that the mechanical properties of the tensile bars in the three regions were similar, which indicated that the sampling position of the tensile bar had little correlation with the fluctuation of the plasticity in this experiment. In addition, The metallographic samples were taken from the top, middle and bottom areas of the same round bar to observe the microstructure. The results showed that there was no significant difference in the distribution of precipitates in the three areas. Because the top of the round bar was close to the pouring cup, which was the final solidification area of the round bar. The grain size of the sample from the top area was slightly greater than that of the middle and bottom area. Considering that there may be casting defects in the top area of the round bar, therefore, the middle part of the round bar was selected to processed into tensile bar, and the bottom area of the round bar was selected as the metallographic sample, as shown in figure 1(b).

Based on the above comparative analysis before the experiment, the round bars after annealing were machined into tensile bars of diameter $12 \mathrm{~mm}$ and length $140 \mathrm{~mm}$ by using an SL 50 numerical control machine tool (Hangzhi Equipment Technology Company, Nantong, China). Three tensile bars from the same casting groups at each annealing temperature were selected for tensile testing performed with a WEW $\times 300B$ instrument (Fangyuan Testing Machine Company, Jinan, China). The metallographic specimens were sampled from the bottom of the as-cast and annealed round bars, these specimens were polished using the standard metallographic procedure and chemically etched in $50\%$ aqua regia aqueous solution. The metallographic microstructures were observed by MV 6000 microscope (Lianchuang Analysis Instrument Manufacturing Company, Nanjing, China) and OLS 4000 laser confocal microscope (Olympus, Tokyo, Japan). The morphology and composition of the precipitates were characterized using a Quanta 250 scanning electron microscope (SEM) (FEI, Hillsboro, OR, USA) equipped with an energy dispersive x-ray spectrometer (EDS). With the aid of the Image-pro Plus (IPP) software (version 6.0, Media Cybernetics, Rockville, MD, USA), the size of precipitates in the microstructure was analyzed and defined as the mean equivalent length (MEL). The measurement principle of MEL is that size is measured at 2 degree intervals and passing through object’s centroid.

3. Results and discussion

3.1. As-cast properties

A total of nine round bars were selected and processed into tensile bars according to the experimental procedure. The elongations of these tensile bars were $14.5\%, 4\%, 18\%, 4.7\%, 16\%, 21.3\%, 7.8\%, 27.6\%$ and $6.9\%$, respectively. There is a great difference between the maximum elongation of $27.6\%$ and the minimum value of $4\%$, which shows the unstable plasticity fluctuations of the investment-cast 00Cr18Nb steel. Some fluctuations in the tensile strength were also observed, and the values were 455, 415, 465, 355, 425, 415, 385, 405 and $365 \mathrm{MPa}$, respectively. The difference between the maximum and minimum tensile strength reached to $110 \mathrm{MPa}$. It can be inferred that the unstable changes of mechanical properties in the investment-cast 00Cr18Nb steel are mainly related to the as-cast microstructure.

The equilibrium phase precipitation calculated by Thermo-Calc software (Tccr-Dos, Thermo-Calc Software Company, Solna, Sweden) is shown in figure 2, which is a reference for the types of theoretical precipitates of the 00Cr18Nb steel. However, there is a difference between theoretical and actual precipitation. According to the Fe-Cr binary phase diagram, the composition range of chromium content for easy-formation of sigma phase...
was 25%–30%, but in practice, for ferritic stainless steels with a chromium content of less than 20%, the formation of the sigma phase becomes quite difficult and requires thousands of hours of aging to achieve. In addition, the general heating process of welding, casting and metallurgical production of ferritic stainless steel with a chromium content of 15%–33%, there is not enough time to form the sigma phase. The research of Sello et al. showed that for rolled AISI 441 stainless steel, the form of the sigma phase was actually not easy. A research also showed that the sigma precipitation was mainly a result of the high chromium and molybdenum contents in the ferrite. In the present study, The chromium content is 18.6%, which is lower than 20%. It appears that sigma phase is relatively difficult to form. A large amount of actual EDS analysis results in this experiment showed that the content of sigma phase in the microstructure was very low, so the precipitates were mainly (Ti,Nb)(C,N) and Laves phase.

According to the related literatures and extensive EDS analysis in the experiment, the Laves phase ([Fe, Si, Cr]$_2$Nb) [19] and (Ti,Nb)(C,N) [20] precipitates in AISI 441 stainless steel can be effectively identified in figure 3. The (Ti,Nb)(C,N) and Laves phase precipitates dispersed in grains with the size of 7.8 and 0.29 μm (statistic values of MEL), respectively, and the quantity of (Ti,Nb)(C,N) particle in the as-cast microstructure was greater than that of Laves phase, as shown in figure 3. The precipitation strengthening effect of (Ti,Nb)(C,N) particle was better than Laves phase in the as-cast microstructure. Although the size of Laves phase was small, precipitation strengthening effect can still be formed. It can be speculated that the fluctuations in the plasticity and strength of the investment-cast 00Cr18Nb steel are mainly related to precipitates in the microstructure. The changes of these precipitates after annealing will affect the mechanical properties.

3.2. Effects of annealing

3.2.1. Effect of annealing on mechanical properties

The effects of annealing temperature and time on mechanical properties are shown in figure 4. The error bar marking rules are as follows: If the values of the three tensile results are close and the standard deviation value is less than 1, the corresponding error bar will not be marked, for example, the error bar of elongation at 800 °C for 1 h in figure 4(a), as well as the tensile strength at 800 °C for 1 h and 900 °C for 3 h in figure 4(b). In general, annealing treatment improved the mechanical properties of 00Cr18Nb investment-cast bars. Figure 4(a) shows that annealing treatment can greatly stabilize the plasticity fluctuations. Compared with the elongation in the as-cast state, i.e., 13 ± 8%, the elongations at 800 and 850 °C for 1, 2 and 3 h were in the range of 22.4%–26.5% (greater than 13%), with the standard deviation mainly in the range of 0.3%–3.5% (better than 8%). However,
when the annealing temperature reached 900 °C, the elongation decreased and the fluctuations increased. For example, the elongations after annealing at 900 °C for 1 and 2 h were 24.7% and 14.9%, respectively, and the standard deviations were 4.5% and 8.5%, respectively. Although the plasticity of the sample annealed at 900 °C for 3 h reached 28.8%, its standard deviation was 6.1%, i.e., the plasticity was improved but the fluctuation value increased at 900 °C for 3 h. Therefore, annealing temperatures of 800 and 850 °C are suitable for the elongation stabilization of the investment-cast 00Cr18Nb steel.

Annealing treatment has also enhanced the tensile strength, as shown in figure 4(b). The tensile strengths after annealing were mainly in the range of 491–525 MPa; these values were greater than 409 MPa of the as-cast steel. The standard deviation of the tensile strength after annealing was in the range of 3–16 MPa, which was also better than 37 MPa (standard deviation) of the as-cast steel. This indicates that the strength stability after annealing is greater than that of the as-cast steel. Nevertheless, unstable fluctuations of the tensile strength occurred after annealing at 900 °C for 1 and 2 h. The strength decreased at 900 °C and the standard deviation increased.

Figure 5 shows the fracture surfaces at room temperature under different annealing treatments. Dimple fracture dominated at 850 °C and the dimple area was greater than that at 800 °C, as shown in figures 5(a) and (b). When the annealing temperature reached 900 °C, the dimple area greatly decreased, as shown in figure 5(c). It can be inferred that suitable plasticity can be obtained by annealing treatment at 850 °C for 2 h.

In summary, annealing treatment can stabilize plasticity fluctuations and enhance the strength. The plasticity and strength improvements achieved by annealing at 800 and 850 °C were better than those obtained by annealing at 900 °C. The better combination of plasticity and strength was achieved by annealing at 850 °C for 2 h. It gave more stable elongation fluctuation and tensile strength close to those of other annealing conditions.

### 3.2.2. Effect of annealing on grain size

Figure 6(a) shows the variations in the grain size of the investment-cast 00Cr18Nb steel after annealing treatment. The grain size of the samples after annealing was 0.67–0.71 mm, the values of the grain size after annealing increased slightly than that of as-cast samples (0.66 mm). It did not increase abnormally. Therefore, the slight increase of the grain size cannot be the main factor affecting the plasticity.
In addition to the effect on the plasticity, the variations of grain size after annealing also affect grain-boundary strengthening, as described by the Hall–Petch equation [21]:

\[
\Delta Y_S = k_Y D^{-1/2}
\]

where \( \Delta Y_S \) is the increment of the yield strength, \( k_Y \) is 17.4 MPa \( \cdot \) mm\(^{1/2}\) for steel and iron materials [22], and \( D \) is the grain diameter, which was approximated by the grain size in this work. The calculation results showed that the \( \Delta Y_S \) values were in the range of 20.7–21.3 MPa, which was close to the as-cast value, i.e., 21.4 MPa, as shown in figure 6(b). It is considered that the slight increase of the grain size after annealing has not greatly changed the mechanical properties compared with those of the as-cast steel.

3.2.3. Effect of annealing on (Ti,Nb)(C,N) precipitates

The SEM image and EDS maps of the (Ti,Nb)(C,N) precipitate are shown in figure 7. It can be seen that Nb(C,N) precipitated on the edge of a TiN particle via non-uniform nucleation. There are two reasons for this distribution. First, TiN and Nb(C,N) have similar lattice parameters and the same face-centered cubic structure. Secondly, the solubility of TiN is lower than that of Nb(C,N) and therefore precipitates first during cooling; Nb(C,N) then forms on the precipitated TiN by heterogeneous nucleation [17]. The Thermo-Calc result in figure 2 shows that the precipitation temperature of (Ti,Nb)(C,N) is 1284 °C, from which it is deduced that (Ti,Nb)(C,N) is a high-temperature precipitate. It has been reported that Nb(C,N) or TiN precipitates do not dissolve in the range 900°C–1000°C [23]. In this study, the annealing temperature ranged from 800 to 900 °C, which is lower than the dissolution temperature of carbonitrides and therefore cannot affect the (Ti,Nb)(C,N) morphology.
The distribution of (Ti,Nb)(C,N) precipitates after annealing is shown in figure 8; these precipitates were still mainly dispersed in the grains. The size of the (Ti,Nb)(C,N) precipitates was basically unchanged by the annealing treatment, as shown in figure 9(a). This is closely related to the high-temperature precipitation property of (Ti,Nb)(C,N). The size of the (Ti,Nb)(C,N) precipitates before and after annealing is basically unchanged, therefore the precipitation strength effect of (Ti,Nb)(C,N) precipitates in the microstructure after annealing can be considered to be close to that of the as-cast samples.

The effect of these (Ti,Nb)(C,N) precipitates on the strength can be determined by the Ashby–Orowan equation [24]:

\[
\Delta\sigma = \left( \frac{0.538Gb f^{1/2}}{d} \right) \ln \left( \frac{d}{2b} \right)
\]

where \(\Delta\sigma\) is the increment of the yield strength, \(f\) is the volume fraction of the second-phase particles, \(d\) is the particle size, \(b\) is Burgers vector (equal to 0.248 nm for ferrite) [25], and \(G\) is the shear modulus (82 GPa for steel) [26]. Annealing temperature of 800 °C–900 °C basically did not change the value of \(d\) for (Ti,Nb)(C,N) precipitate, therefore \(f\) can be considered unchanged after annealing. The values of \(G\) and \(b\) in equation (2) are constant, therefore the influence of (Ti,Nb)(C,N) precipitates on strength increment \(\Delta\sigma\) is almost unchanged compared with that of the as-cast steel.
3.2.4. Effect of annealing on Laves phase

In addition to (Ti,Nb)(C,N) precipitates, the Laves phase is another main precipitate in the microstructure. Even after annealing at 900 °C for 3 h, the Laves phase still existed in grains and at grain boundaries, as shown in figure 10. The Laves phase precipitates in grains were mainly in a dispersed state, and there were relatively few Laves phase in the as-cast microstructure, as shown in figure 11(a). The MEL of the Laves phase in grains increased with increasing annealing temperature and time, as shown in figure 9(b). For instance, the MEL of the Laves phase in grains increased from 0.35 μm for annealing at 800 °C for 2 h to 0.44 μm for annealing at 900 °C for 2 h, as shown in figures 11(b) and (e), both the values were greater than that of the as-cast steel, i.e., 0.29 μm. The morphology of the Laves phase in grains changed from small granular and short-strip shape to mainly large granular and long-strip shape. The amount of long-strip shape of Laves phase in grains increased with increasing annealing time (1, 2 and 3 h) at 900 °C, as shown in figures 11(d)–(f).

The annealing temperature and time also affected the distribution of Laves phase in grains, as shown in figures 11(b)–(f). The Laves phase in grains annealed at 800 and 850 °C for 2 h was small in size (MEL: 0.35 and 0.38 μm respectively) but great in quantity, whereas annealing at 900 °C for 2 h gave small quantities of the Laves phase and larger size (MEL: 0.44 μm). When the annealing time was extended to 1, 2 and 3 h at 900 °C, respectively, the Laves phase size increased (MEL: 0.42, 0.44 and 0.43 μm respectively) but the quantity decreased. The Laves phase distribution in grains was homogeneous after annealing at 800 and 850 °C, but was nonhomogeneous after annealing at 900 °C.

The effects of the size, distribution, morphology and quantity of the Laves phase after annealing on the plasticity and strength can be explained as follows. Annealing at 800 and 850 °C gave large quantities of uniformly distributed and small Laves phase precipitates. This change leads to the reduction contents of alloying elements in matrix. The plastic deformation ability of the matrix is improved, and the precipitated Laves phase contributes to precipitation strengthening effect to a certain extent. Therefore, the plasticity and strength are finally improved. It is confirmed by the elongation and tensile strength values. For example, the elongations and tensile strengths after annealing at 800 and 850 °C for 2 h were 24.8 ± 3.5% and 26.3 ± 0.4%, respectively, and 517 ± 8 MPa and 513 ± 4 MPa, respectively. These values were greater than 13 ± 8% (the as-cast elongation) and 409 ± 37 MPa (the as-cast tensile strength). However, when annealed at 900 °C for 1, 2 and 3 h, the MEL of the Laves phase furtherly increased, and the amounts decreased, the morphology mainly changed to long-strip shape, and the distribution became uneven, as shown in figures 11(e), (d) and (f). The further increasing size of Laves phase and the change of morphology are not conducive to the improvement of plasticity, because the Laves phase with long-strip shape may lead to the increase of new stress concentration areas during stretching.
deformation, resulting in a decrease in plasticity and strength. The corresponding minimum plasticity and tensile strength after annealing at 900 °C decreased to 14.9 ± 8.5% and 491 ± 16 MPa, i.e., the values decreased and the fluctuation increased. It can be seen that the size, distribution, morphology and quantity of the Laves phase in the annealed microstructures are the main factors affecting the plasticity and strength.

With the increase of annealing temperature and time, the MEL of the Laves phase at grain boundaries also shows an increasing trend, as shown in figure 9(c). The morphology of the Laves phase was mainly granular with an interval distribution state along the as-cast grain boundaries, as shown in figure 12(a). As the annealing temperature and time increased, the morphology of Laves phase at grain boundaries was still granular, but due to the increasing MEL of the Laves phase at grain boundaries, the intervals of the Laves phase along grain boundaries were reduced compared with that of the as-cast state, as shown in figures 12(b)–(f). This changes of Laves phase at grain boundaries are mainly related to the precipitation characteristics of the Laves phase and the incoherent interface between the Laves phase and the Fe matrix [27]. A transformation kinetic study of the Laves phase by Sello et al [28] indicated that Laves phase nucleation at grain boundaries was dominant above 750 °C.

Generally speaking, the increasing size of the Laves phase at grain boundaries will weaken the plasticity and strength. However, referring to the results (the plasticity and strength were improved after annealing) of the tensile bars in figure 4, the adverse effect of the Laves phase at grain boundaries is offset by the influence of the Laves phase in grains. Therefore, under 800 and 850 °C annealing conditions, the uniform distribution and small size of the Laves phase in grains have an effect on improving the plasticity and strength of the matrix, and it also effectively weakens the adverse effect caused by the increasing size of the Laves phase at grain boundaries under these annealing temperature. However, when annealing at 900 °C, due to the increasing size and uneven distribution of the Laves phase with long-strip shape in grains, the plasticity and strength of the matrix are reduced, and the adverse effects of the increasing size of the Laves phase at grain boundaries are superimposed on the samples. The plasticity and strength of the samples are adversely affected at 900 °C. Morris et al [29] also reported that the strength was decreased because of rapid coarsening of the Laves precipitates in Fe-25%Al-2% Nb material when the temperature reached up to 900 °C. In addition, the increasing size of the Laves phase at grain boundaries will furtherly increase the stress concentration areas accordingly, which leads to the initiation and aggregation of microcracks with the increase of external stress, resulting in the decrease of the plasticity and strength.

In summary, annealing treatment can affect the size of Laves phase in grains and at grain boundaries. The Laves phase size increased with increasing annealing temperature and time. The size, distribution, morphology and quantity of the Laves phase are related to the effects of annealing on the mechanical properties.
4. Conclusions

(1) Annealing treatment can solve the problem of unstable plastic fluctuations of the investment-cast 00Cr18Nb steel. The plastic stability under 800 and 850 °C was better than that of 900 °C, the main reasons are related to the changes of the Laves phase. Better plasticity and strength can be obtained at 850 °C for 2 h.

(2) The size of the Laves phase tended to increase with increasing annealing temperature and time. The Laves phase with small size, large quantity and uniformly distribution in the microstructure after annealing plays a major role in improving the mechanical properties.

(3) Based on the high temperature precipitation characteristic of the (Ti,Nb)(C,N) particles and the small fluctuation of the grain size, the improvement effect of (Ti,Nb)(C,N) particles and the grain size after annealing on mechanical properties is less than that of Laves phase.

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Data availability statement

The data that support the findings of this study are available upon reasonable request from the authors.

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