Multi-Phase Microstructure Control and Properties of 700 MPa Grade High-Ductility Low-Carbon Low-Alloy Steel (Postprint)

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Abstract

The microstructural evolution and mechanical properties of low-carbon low-alloy steel were investigated through multi-step heat treatment processes including intercritical annealing, intercritical tempering, and tempering. The results show that after intercritical annealing, the microstructure consists of a dual-phase structure of lath-shaped intercritical ferrite and bainite/martensite. After intercritical tempering, the microstructure becomes a multi-phase structure comprising intercritical ferrite, tempered bainite/martensite, and retained austenite. The retained austenite exhibits granular and strip morphologies, distributed at the ferrite/bainite(martensite) phase interfaces and between bainite/martensite laths, with a content as high as 29%. It remains stable after tempering, primarily stabilized through the enrichment of C, Mn, Ni, and Cu in the reversed austenite. During intercritical annealing and tempering, NbC precipitates in ferrite and bainite/martensite, exhibiting spherical, elliptical, or irregular shapes with an average size of 10 nm; Cu-rich precipitates form during intercritical tempering and tempering, distributed spherically in ferrite and retained austenite with sizes ranging from 10 to 30 nm. Through the transformation-induced plasticity (TRIP) effect of retained austenite and the precipitation strengthening of nanoprecipitates, the experimental steel exhibits excellent mechanical properties: yield strength exceeding 700 MPa, tensile strength exceeding 900 MPa, uniform elongation exceeding 20%, and total elongation exceeding 30%.

Full Text

Regulation of Multi-Phase Microstructure and Mechanical Properties in a 700 MPa Grade Low Carbon Low Alloy Steel with Good Ductility

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Abstract

Low carbon and low alloy steels require a good combination of strength and ductility to ensure safety and stability of structures. Heat treatment in the intercritical region can not only produce multi-phase microstructures, but also lead to redistribution of alloying elements among different phases. Multi-step intercritical heat treatment is favorable for obtaining retained austenite stabilized by repeated enrichment of alloying elements in reversed austenite, as well as nanometer-sized precipitates that primarily form during tempering. Excellent mechanical properties are achieved through the transformation-induced plasticity (TRIP) effect of retained austenite and precipitation hardening from nanometer-sized precipitates. In this work, the microstructural evolution and corresponding mechanical properties were investigated in a low carbon low alloy steel processed by a three-step heat treatment: intercritical annealing, intercritical tempering, and tempering. After intercritical annealing, the microstructure was a typical dual-phase microstructure consisting of intercritical ferrite and bainite/martensite. After intercritical tempering, it primarily comprised intercritical ferrite, tempered bainite/martensite, and retained austenite. The retained austenite, with a volume fraction of 29%, distributed at ferrite/bainite (martensite) boundaries and between bainitic/martensitic laths. The retained austenite was stabilized by enrichment of C, Mn, Ni, and Cu in reversed austenite during the reversion transformation process. NbC precipitates with an average size of 10 nm formed in the ferrite matrix and bainite/martensite, while Cu-containing particles in the size range of 10-30 nm precipitated in ferrite and retained austenite during intercritical tempering and tempering processes. The morphology of NbC precipitates was spherical, elliptical, and irregular, while copper precipitates were spherical. With the combination of transformation-induced plasticity (TRIP) effect of retained austenite and precipitation hardening, the steel possessed outstanding mechanical properties: yield strength $>$ 700 MPa, tensile strength $>$ 900 MPa, uniform elongation $>$ 20%, and total elongation $>$ 30%.

KEY WORDS high performance, intercritical heat treatment, multi-phase microstructure, retained austenite, nanometer-sized precipitate

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Low-alloy steels are widely used in construction facilities and energy industries. To ensure structural reliability and safety, these materials require high strength and toughness/plasticity. With rapid industrial development and increasing environmental awareness, demands for lightweight engineering structures and equipment are growing, creating an urgent need for higher-strength steels. Single strengthening mechanisms and methods are well understood, and through thermomechanical processing and heat treatment, steel strength can easily reach the 10^9 Pa level. However, for steel materials, such simple strength enhancement typically leads to reduced toughness and plasticity. Only when energy is continuously absorbed during deformation and fracture does increased strength become meaningful [1].

Retained austenite can absorb deformation energy and delay necking through the transformation-induced plasticity (TRIP) effect during deformation, thereby simultaneously improving strength and plasticity. Therefore, introducing retained austenite into steel is an effective approach to achieve a good strength-ductility balance. How to obtain sufficient and stable retained austenite has been a research focus in recent years. One method involves modifying steel chemical composition by increasing austenite-stabilizing elements to obtain austenitic structures, such as high-Ni austenitic stainless steels [2,3] and high-Mn twinning-induced plasticity (TWIP) steels [4,5]. However, due to excessively high alloy element content, this approach increases steel costs and creates difficulties in smelting. Another method involves obtaining stable retained austenite through heat treatment, commonly using TRIP [6,7] and quenching & partitioning (Q&P) processes [8,9]. During these heat treatments, C atoms diffuse from bainite or martensite into untransformed austenite, significantly increasing the stability of untransformed austenite and resulting in room-temperature stable retained austenite. Since retained austenite is obtained through C enrichment, such steels typically have C contents higher than 0.2% (mass fraction). This is undesirable for structural materials because high C content greatly reduces steel weldability. Therefore, obtaining retained austenite to improve properties under low-C and low-alloy conditions, while maintaining good weldability and low cost, is of great significance for developing high-strength steels.

In steels with precursor microstructures of lath martensite or bainite, when heated to the α+γ two-phase region, reverse-transformed austenite nucleates and grows between laths. This austenite appears needle-like or lath-shaped and is enriched with alloying elements. With increasing isothermal temperature and time in the two-phase region, austenite transforms to granular morphology and the enriched alloying elements gradually decrease [10]. At lower two-phase region temperatures, austenite has higher C content and greater stability. Therefore, stable austenite structures can be obtained through reasonable control of two-phase region temperature and time.

As early as the 1970s, Miller [11] and Niikura et al. [12] obtained retained austenite in high-Mn steels through intercritical annealing. Due to high Mn content, large amounts of retained austenite could be obtained with just one-step intercritical treatment. With increasing annealing time, higher Mn enrichment in reversed austenite led to more retained austenite at room temperature. In recent years, Luo et al. [13] and Shi et al. [14] proposed the austenite reversion transformation (ART) theory, achieving retained austenite contents as high as 40% in 0.2C-5Mn steel after intercritical annealing. Combined with ultrafine-grained ferrite matrices, excellent strength-ductility combinations were obtained. For low-Ni steels, due to insufficient Ni enrichment during one-step intercritical treatment to stabilize reversed austenite to room temperature, two-step or even multi-step continuous intercritical treatments are required. In such multi-step intercritical heat treatments, the intercritical treatment temperature gradually decreases because element enrichment from the previous step lowers the phase transformation point in that region [15]. The retained austenite obtained through this multi-step intercritical treatment has high alloy content, fine size, and high stability. This provides a new approach for obtaining retained austenite in low-alloy systems: through continuous intercritical heat treatment, alloy elements are repeatedly enriched in reversed austenite to stabilize it and obtain retained austenite.

This work investigates a low-C, Nb-Cu microalloyed low-alloy steel using multi-step intercritical treatment to study the relationship between microstructure and properties during different heat treatment processes. Microstructures were characterized using optical microscopy, scanning electron microscopy, electron backscatter diffraction, and transmission electron microscopy, focusing on revealing the evolution of retained austenite and nanometer-sized precipitates. Combined with mechanical property variations, the effects of retained austenite and precipitates on properties were investigated. This study aims to understand the microstructural evolution 规律 in low-alloy steels during continuous intercritical treatment, providing theoretical and technical foundations for microstructure control and property optimization of high-performance low-alloy steels.

Experimental

The chemical composition of the experimental 700 MPa grade high-strength high-ductility steel (mass fraction, %) was: C 0.10, Mn 2.01, Si 0.78, Al 0.78, Nb 0.078, Cu 1.0, Ni 1.0, Mo 0.26, Fe balance. To ensure weldability, low-C design was adopted. Mn and Ni are austenite-stabilizing elements, Nb, Cu, and Mo are precipitation-strengthening elements, and Si and Al prevent cementite formation, ensuring more C atoms can participate in stabilizing austenite and forming precipitates. The experimental steel was smelted in a 50 kg vacuum induction furnace, then rolled into 8 mm thick plates and air-cooled to room temperature. Tensile heat treatment specimens measuring 10 mm × 8 mm ×

90 mm were cut parallel to the rolling direction, and after heat treatment were machined into standard round-bar tensile specimens with a diameter of 5 mm (gauge length 25 mm). Phase transformation points were measured using the dilatometric method according to YB/T 5127-93 at a heating rate of 0.05 °C/s.

The dilatometric method measured the experimental steel' s Ac1 (start temperature of ferrite-to-austenite transformation) and Ac3 (temperature at which ferrite completely transforms to austenite) as 660 °C and 928 °C, respectively, under equilibrium conditions. The heat treatment process was: specimens were heated to the two-phase region at 780 °C, isothermally held for 30 min, then air-cooled to room temperature; subsequently reheated to the two-phase region at 660 °C, isothermally held for 30 min, then air-cooled to room temperature; finally reheated to 500 °C, isothermally held for 30 min, then air-cooled to room temperature [Figure 1: see original paper]. The selection of heat treatment temperatures is described in references [15, 16]. In this study, samples after one-step, two-step, and three-step heat treatments were designated as samples A, B, and C, respectively. Specimens were cut from the undeformed ends of fractured tensile samples, ground and polished, then etched with LePera reagent for microstructural observation using a BX51M optical microscope (OM); after etching with 3% nitric acid alcohol, microstructures were observed using an Ultra 55 field-emission scanning electron microscope (SEM). Electron backscatter diffraction (EBSD) analysis was performed after electropolishing in a solution of $HClO_4:C_3H_8O_3:C_6H_2O = 0.5:1:8.5$ (volume ratio) at 20 kV accelerating voltage, 15 mm working distance, 70° tilt angle, and 0.08 mm scan step. Residual austenite content was measured using a DMAX-RB X-ray diffractometer (XRD) with the five-peak method according to GB8362-87. For TEM observation, 0.4 mm thick specimens were cut from tensile bar ends, ground to 100 m thickness, then punched into 3 mm diameter discs and further thinned to 50-60 m before twin-jet electropolishing in 5% perchloric acid alcohol solution at -20 °C. Microstructure and precipitate observation was conducted using a JEM-2100F transmission electron microscope (TEM) equipped with an energy dispersive spectrometer (EDS) at 200 kV operating voltage.

2.1 Microstructure and Mechanical Properties

[Figure 2: see original paper] shows the phase transformation points of the experimental steel at different heat treatment stages measured by dilatometry. In the hot-rolled condition, the Ac1 temperature for bcc-to-fcc transformation under equilibrium was 660 °C, which decreased to 607 °C after the first processing step. This indicates that the reverse transformation from bcc to fcc could proceed during the second-step heat treatment at 660 °C. The reason for Ac1 reduction is the redistribution of alloying elements during intercritical heat treatment, where alloy element enrichment in reversed austenite lowers the phase transformation temperature in that region. The decrease in Ac1 temperature not only enables the next intercritical heat treatment to be conducted at lower temperature but also facilitates formation of finer precipitates at low temperature.

[Figure 3: see original paper] and [Figure 4: see original paper] show OM and SEM images of the experimental steel after different heat treatment stages, respectively. The as-hot-rolled microstructure primarily consisted of lath bainite/martensite [FIGURE:3a and 4a]. The microstructure evolved continuously with changes in intercritical treatment temperature and steps. After intercritical annealing at 780 °C, the microstructure mainly comprised ferrite and bainite/martensite dual-phase structure, with ferrite appearing as irregular blocky or lath-shaped grains distributed between bainite/martensite laths [FIGURE:3b and 4b]. After intercritical tempering at 660 °C, ferrite showed no obvious change, but the bainite/martensite content decreased. SEM images revealed that bainite/martensite decomposed into fine granular structures dispersed around ferrite grains [Figure 4c: see original paper]. Comparing with OM images [Figure 3c: see original paper], these granular structures were primarily retained austenite, indicating that bainite/martensite underwent reverse transformation to form retained austenite during intercritical tempering. The average diameter of retained austenite particles was below 1 m. After tempering at 500 °C, the microstructure showed no significant change, still consisting mainly of ferrite and retained austenite, while bainite/martensite content decreased significantly after two tempering steps [FIGURE:3d and 4d].

The mechanical properties of the experimental steel at different heat treatment stages are shown in Table 1. After intercritical annealing, the steel exhibited a yield strength of 686 MPa and tensile strength as high as 1178 MPa, but with low elongation—uniform elongation was only 6.3% and total elongation was 17%. After the second-step intercritical tempering, yield strength increased to 707 MPa, tensile strength decreased to 845 MPa, but elongation increased significantly, with uniform elongation and total elongation exceeding 20% and 30%, respectively, and the strength-ductility product approaching 30 GPa • %. After the third-step tempering, yield strength increased by 17 MPa and tensile strength increased by 78 MPa, while uniform and total elongation showed no significant decrease.

2.2 Retained Austenite

[Figure 5: see original paper] shows EBSD characterization of retained austenite distribution and XRD spectra at various heat treatment stages. Black lines represent low-angle grain boundaries with misorientation less than 10°, and red regions represent retained austenite. As shown in [FIGURE:5a-c], retained austenite was primarily distributed at grain boundaries. XRD spectra calculations indicated that retained austenite content was low after intercritical annealing, only about 5%; after the second-step intercritical tempering treatment, retained austenite content reached as high as 29%; and after the third-step tempering, retained austenite content did not decrease. This demonstrates that retained austenite obtained through two-step intercritical treatment has high thermal stability. TEM images [Figure 6: see original paper] show that retained austenite was distributed as lath-shaped or granular structures at ferrite grain boundaries or ferrite/bainite phase interfaces. No retained austenite was observed within ferrite grains. EDS analysis revealed that average Mn and Ni contents in retained austenite were much higher than in the ferrite matrix [Figure 6a: see original paper], consistent with Xie et al.' s research results [17]. This suggests that alloying elements diffused and enriched from ferrite and bainite/martensite interiors into neighboring austenite during intercritical treatment. Therefore, it is necessary to investigate the partitioning behavior of alloying elements in different phases during intercritical treatment.

Thermal-Calc calculations were performed to determine alloy element partitioning between austenite and ferrite in the experimental steel after 780 °C intercritical annealing for 30 min under equilibrium conditions, as shown in Table 2. At 780 °C equilibrium, the volume fractions of ferrite and austenite were 55.0% and 45.0%, respectively. Austenite-stabilizing elements C, Mn, Ni, and Cu were all enriched to some degree in austenite. The martensite start temperature (Ms) for this austenite transforming to martensite was 333 °C, indicating that alloy element enrichment at this temperature was insufficient to stabilize austenite to room temperature. During cooling, austenite transformed to bainite/martensite structure, consistent with microstructural observations. SEM observation revealed that during the second-step intercritical treatment, primarily bainite/martensite microstructure changed. Therefore, it is necessary to investigate the phase transformation behavior of this microstructure constituent at 660 °C. Table 2 also shows alloy element partitioning in the microstructure at 660 °C for this alloy composition system. Thermal-Calc calculations indicated that Cu precipitation temperature during cooling was 680 °C, meaning Cu precipitates could form during isothermal holding at 660 °C. After reverse transformation at 660 °C, alloy elements were further enriched in the newly formed austenite, with C content of 0.5% and Mn content as high as 6.1%. The Ms temperature was calculated as 120 °C, indicating that austenite would still transform to martensite during cooling. However, in practice, after two-step processing, retained austenite content as high as 29% was obtained at room temperature. This is because, in addition to chemical composition, retained austenite size is also a critical factor affecting its stability. Takaki et al. [18] demonstrated that smaller retained austenite size leads to lower Ms and higher stability. When retained austenite grain size is less than 1 m, martensite nucleation within austenite requires extremely high chemical driving force, thus effectively preventing austenite-to-martensite transformation. Therefore, retained austenite stability results from the combined effects of alloy element enrichment and grain size.

Alloy element diffusion is determined by both temperature and diffusion time. In two-dimensional space, diffusion distance depends on the diffusion coefficient and diffusion time of alloying elements. Table 3 lists diffusion distances of C, Mn, Ni, and Cu elements at different temperatures. Calculations show that C diffusion coefficient and distance in ferrite are several orders of magnitude higher than other elements. At 660 °C, diffusion distances of Mn, Ni, and Cu were 212, 95, and 120 nm, respectively. At 500 °C, except for C which still had relatively high diffusivity, diffusion of other elements was negligible. Comparing ferrite and bainite/martensite grain sizes in SEM images, alloy element diffusion primarily occurred during the first two intercritical treatment steps. The C content of retained austenite obtained through the two-step method was 0.5%, much lower than that in conventional TRIP steel retained austenite (1.1%-1.7%) [19]. EDS measurements revealed Mn content as high as 5.0% and Ni content of 1.6% in retained austenite. This indicates that retained austenite obtained through the two-step method was primarily stabilized by Mn and Ni enrichment.

In TRIP or Q&P steels, C atoms enrich in neighboring austenite from bainite or martensite laths during isothermal quenching or partitioning, thereby obtaining room-temperature stable retained austenite. This type of austenite forms at medium-low temperature isothermal stages and has high C content. During high-temperature tempering, it decomposes into cementite and ferrite [20]. However, tempering stability increases with increasing Mn and Ni content in retained austenite. In multi-step intercritical heat treatment, retained austenite stability is primarily enhanced through enrichment of alloy elements such as Mn and Ni in reversed austenite, with very low C content, which greatly reduces the chemical driving force for cementite formation during tempering. Meanwhile, high Si and Al contents in the experimental steel effectively suppress cementite formation. Therefore, after tempering, no cementite structure was observed in the microstructure, and retained austenite maintained high content.

2.3 Nanometer-sized Precipitates

Forming precipitates by adding microalloying elements is a conventional method for improving steel strength. As common precipitate-forming elements, the equilibrium precipitation start temperatures for Nb and Cu are 1200 °C and 680 °C, respectively (calculated by Thermal-Calc). This indicates that Nb begins to precipitate during hot rolling, while Cu precipitates require lower temperatures to form. To investigate precipitation behavior of Nb and Cu at different heat treatment stages, carbon replica and thin foil specimens were observed by TEM, and precipitate sizes were statistically analyzed. [Figure 7: see original paper] shows TEM images of carbon replica precipitates after hot rolling, intercritical annealing, and intercritical tempering. After hot rolling, two different sized precipitates were observed: one larger than 20 nm with irregular shape, formed at relatively high temperature with little contribution to strength; the other 4-14 nm in size, spherical and distributed in the matrix [Figure 7a: see original paper], formed during rolling or cooling processes. The average precipitate size was 10.9 nm [Figure 7b: see original paper], and EDS results identified these as NbC. After intercritical annealing, most newly formed NbC precipitates were 7-15 nm in size, spherical and dispersedly distributed in the matrix [FIGURE:7c and d].

After intercritical tempering, the average NbC size decreased to about 6.0 nm because numerous fine precipitates (3-8 nm) formed at this stage [FIGURE:7e and f]. In addition to fine NbC precipitates, large amounts of Cu precipitates were observed in the ferrite matrix after intercritical tempering [Figure 8a: see original paper]. Cu precipitate particle sizes ranged from 10-30 nm [Figure 8b: see original paper], spherical or ellipsoidal and distributed in ferrite. Considering the influence of Nb precipitates, particles below 10 nm were ignored when statistically analyzing Cu precipitate sizes. These results indicate that Nb begins to precipitate during hot rolling and continues precipitating during intercritical annealing and tempering, with precipitate size decreasing as temperature decreases. Cu begins to precipitate during intercritical tempering, with relatively large precipitate particle sizes.

After the third-step tempering treatment, numerous precipitate particles appeared in retained austenite [FIGURE:8c-e]. EDS analysis identified these as Cu precipitates [Figure 8f: see original paper], with sizes of 10-20 nm. Cu precipitates were also observed in neighboring ferrite, but with much lower density than in austenite. Cu precipitate distribution in ferrite was non-uniform: higher density in ferrite adjacent to retained austenite, and lower density in large ferrite grains far from austenite [Figure 8c: see original paper]. This also indicates that Cu element redistribution occurred during the first two heat treatment steps, i.e., diffusion enrichment from ferrite into austenite. Generally, due to high Cu solubility in austenite, precipitation in austenite is difficult. Chi et al.' s [21] research on Cu-containing austenitic stainless steels showed that Cu precipitation in austenite proceeds through diffusion to form Cu-rich regions in local areas of austenite, followed by gradual precipitation during aging. This requires not only high Cu content (mass fraction > 2%) but also long isothermal time (> 5 h) in austenite. In this experiment, Cu enrichment in reversed austenite during intercritical annealing and tempering prepared conditions for forming Cu-rich regions. Moreover, Cu precipitation was active at 500 °C, so Cu precipitated in austenite after the third-step tempering treatment.

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3 Analysis and Discussion

During intercritical treatment, alloy element enrichment in reversed austenite not only increases austenite stability but also lowers the equilibrium Ac1 temperature for bcc-to-fcc transformation start, as shown in [Figure 2: see original paper]. The decrease in Ac1 enables subsequent intercritical treatment at lower temperatures and facilitates formation of finer precipitates at low temperature, ensuring simultaneous acquisition of large amounts of retained austenite and Nb and Cu precipitates in the same processing step. After two-step heat treatment, ferrite content was about 50% [16], similar to conventional C-Mn-Si TRIP steel proportions [22], but with higher strength and plasticity than traditional TRIP steel (yield strength 400-600 MPa, total elongation 20%-30%) [23]. This is attributed to the ultrafine microstructure, stable retained austenite, and nanometer-sized precipitates. Unlike large polygonal ferrite in TRIP steel, ferrite obtained in the intercritical region appeared as fine lath-shaped or blocky structures distributed between bainite/martensite laths, retaining high dislocation density internally. This ultrafine ferrite microstructure also appears in Q&P steels and medium-Mn steels processed by austenite reversion transformation (ART), and is considered a factor contributing to high yield strength. Nanometer-sized precipitates formed during tempering were dispersedly distributed in ferrite, greatly strengthening the ferrite matrix. Studies by Funakawa et al. [24] and Yen et al. [25] showed that when precipitate size in ferrite is below 5 nm, the contribution to yield strength exceeds 300 MPa. Therefore, despite high ferrite content, the experimental steel still possessed high yield strength.

After the third-step tempering treatment, Cu continued to precipitate in ferrite and austenite. However, due to relatively large Cu precipitate particle sizes, the contribution to yield strength was limited. Compared with the second-step intercritical tempering, yield strength increased by only 17 MPa, but tensile strength increased substantially. This is related to Cu precipitation in retained austenite. On one hand, alloy element enrichment and precipitates in austenite increased austenite strength [26], increasing the proportion of hard phases in the microstructure. On the other hand, retained austenite absorbed dislocations from surrounding microstructure during deformation, leading to increased dislocation density within austenite [27]. Interaction between Cu precipitates and dislocations also strengthened austenite and improved its stability to some extent. Additionally, Cu precipitates strengthened martensite after austenite transformed to martensite during deformation. After tempering treatment, the work hardening rate of the experimental steel was significantly improved [28], ensuring high tensile strength while maintaining high yield strength.

Conclusions

(1) Intercritical heat treatment causes alloy element enrichment in reversed austenite, thereby lowering the phase transformation point of the reversed structure and enabling subsequent reverse transformation at lower temperatures. Continuous intercritical treatment allows this reverse transformation to proceed repeatedly, improving austenite stability through multiple enrichments of alloy elements in reversed austenite, thus obtaining room-temperature stable retained austenite. Meanwhile, the gradually decreasing intercritical treatment temperature reduces precipitate size and promotes formation of low-temperature precipitates, enabling simultaneous acquisition of retained austenite and precipitates in the same processing step and effectively combining precipitation strengthening with the TRIP effect of retained austenite. Through multi-phase microstructure, TRIP effect, and precipitation strengthening, the low-carbon low-alloy experimental steel exhibits excellent mechanical properties: yield strength >

700 MPa, tensile strength > 900 MPa, uniform elongation > 20%, and total elongation > 30%.

(2) After continuous three-step heat treatment of intercritical annealing, intercritical tempering, and tempering, a multi-phase microstructure was obtained consisting of a lath-shaped ferrite matrix, lath bainite/martensite, and retained austenite. Retained austenite had an average size below 1 m, distributed at ferrite/bainite (martensite) phase boundaries and bainite/martensite lath boundaries. Retained austenite primarily formed during intercritical tempering, stabilized mainly by enrichment of C, Mn, Ni, and Cu in reversed austenite, with content as high as 29%. High alloy element content and fine size endowed retained austenite with high stability. During the third-step tempering process, retained austenite remained stable with unchanged content.

(3) The microstructure primarily contained two types of precipitates: NbC and Cu. The former precipitated during all three heat treatment steps, with size gradually decreasing as temperature decreased, dispersedly distributed in ferrite and bainite/martensite. The latter partially precipitated in ferrite during intercritical tempering, and primarily precipitated in ferrite and retained austenite during tempering. NbC precipitation prevented strength loss during high-temperature annealing and tempering, ensuring high strength of the matrix microstructure. In addition to precipitation strengthening, Cu precipitation in retained austenite improved retained austenite stability to some extent, thereby increasing work hardening rate and tensile strength.

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