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Microstructure evolution and mechanical properties during industrial intercritical quenching and partitioning (IQ&P) processing of a low alloy steel

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Abstract

The evolution of the microstructure, texture and mechanical properties of a low alloy steel containing 0.09C-2Mn-0.4Si (wt.%) was investigated during the industrial intercritical quenching & partitioning (IQ&P) process. The steel sheet was thermally treated and characterized by scanning electron microscopy, electron backscatter diffraction, x-ray diffraction, etc. Low alloy steel treated with different processes can reach a tensile strength above 900 MPa. A multiphase structure composed of lath martensite, fine ferrite and retained austenite was obtained after annealing at 770–870 °C, and the retained austenite produced a discontinuous transformation induced plasticity (TRIP) effect and coordinated deformation in the tensile strain. The fractions of the textures {110} ⟨112⟩ and {111} ⟨110⟩ were found to gradually abate as the annealing temperature increased, while the textures {100} ⟨001⟩ and {001} ⟨110⟩ continuously expanded. The recrystallized texture gradually disappears with decreasing quenching temperature, although the fraction of the texture caused by the martensite transformation increases, and the texture gradually changes from {113} ⟨110⟩ to {111} ⟨110⟩. The effect of texture and microstructure evolution on mechanical properties was discussed in terms of character and morphology.

1. Introduction

The low alloy steel is an economical engineering material, which contains a small amount of carbon (less than 0.2%), and also small amounts of alloying elements. The desired mechanical properties of low alloy steel are achieved by the evolution of microstructures through controlled thermomechanical processing (TMCP). Since Speer et al [1] first proposed the quenching and partitioning (Q&P) process, which is based on the distribution law of carbon from supersaturated martensite to retained austenite, Q&P steels have been significantly developed and widely applied in automobile manufacturing [2–4]. The early Q&P process included two steps: (1) a quenching step, which involved heating the cold rolled plate to the full austenitizing temperature, and then quenching between the martensite start temperature and the martensite completion temperature; (2) a partitioning step, which directly involved isothermal solute atom partitioning at or above the quenching temperature, and then finally cooling to room temperature, so that the final microstructure was composed of martensite + retained austenite [5]. In recent years, the multiphase structure (ferrite + martensite + retained austenite) obtained by intercritical annealed quenching and partitioning (IQ&P) has shown a good combination of strength and plasticity [6, 7]. During the deformation process, the retained austenite of low alloy steel transformed into high-strength martensite accompanied by volume expansion, i. e. transformation induced...
plasticity (TRIP) effect, which inhibited the plastic instability and improved the strength and plasticity at the same time [8].

The microstructure of IQ&P steel during industrial continuous annealing is a relatively complex process, including ferrite recovery and recrystallization, transformation of pearlite and ferrite into austenite and formation of martensite after quenching. The volume fraction of each phase in the final structure is closely related to the critical annealing temperature and time [9, 10]. At present, the influence of intercritical annealing temperature on the TRIP effect is not unified. Sakuma et al [11] thought through experiments that the optimum retained austenite content and mechanical properties could be obtained when the intercritical annealing temperature was slightly higher than Ac1. Chang Gil Lee et al [12, 13] chose the mean temperature of Ac1 and Ac3 to obtain a higher retained austenite fraction. Some researchers also choose that the intercritical annealing temperature is between Ac1 and Ac3, it is cooled near Ac1 at a low cooling rate after soaking around Ac3 temperature, and then quickly cooled to bainite region to obtain the best strong plastic product [14–16]. For the cold rolling and industrial continuous process of IQ&P steel, the selection of annealing temperature in the two-phase area and isothermal temperature and time in the martensitic transformation area are the main factors affecting the formation and morphology of retained austenite, and the thermodynamic calculation of equilibrium transformation point and experimental calculation of Ac1 and Ac3 are also very important in guiding the formulation of IQ&P process [17, 18]. However, in previous studies, full austenitization above Ac3 was usually performed prior to Q&P study, but the effect of the industrial IQ&P process between Ac1 and Ac3 on texture evolution and misorientation has not been quantitatively described. The microstructure and texture evolution rule and mechanical properties of the industrial IQ&P process are very important to the practical continuous annealing of low alloy steel.

Therefore, in this work, the aim is to make a direct comparison between fully austenitized and IQ&P-treated low-alloy steels after hot rolling and cold rolling to clarify the effect of different austenitizing methods on the low-alloy steels. Based on the above mentioned, the microstructure evolution, mechanical properties, retained austenite and texture obtained by different IQ&P processes will be investigated emphatically by comparative analysis. The research results are helpful for optimizing the heat treatment technology of low alloy steel, enhancing and tailoring its mechanical properties.

### 2. Experimental material and methods

The experimental steel was melted in a vacuum induction furnace and cast into a 50 kg ingot. The chemical composition of the steel under investigation is 0.09% C, 0.4% Si, 2% Mn, 0.015% P, 0.005% S, and 0.03% Al, as shown in table 1. A Gleeble-1500D thermal simulation testing machine was heated from room temperature to 900 °C at a heating rate of 0.05 °C/s, to measure the temperature-rising expansion curve. The equilibrium transformation points of the steel measured by the tangent method Ac1 and Ac3 were 736 °C and 855 °C, respectively (figure 1(a)), the corresponding expansion curve was measured at a cooling rate of 50 °C/s after full austenitization, and the martensite start temperature (Ms) was determined to be 390 °C. The equilibrium phase fraction of steels at different annealing temperatures was calculated using Thermal-calc software and the TCFE7 database (figure 1(b)). A microstructure composed of 22% austenite and 78% ferrite was formed at 720 °C, and 85% austenite and 15% ferrite microstructure was formed at 820 °C. These calculation results have important guiding significance for the design of subsequent continuous annealing processes. The transformation point of steel always changes with different heating and cooling rates. Ac1 and Ac3 generally increase with increasing heating rate, while Ar1 and Ar3 decrease with increasing cooling rate, which is mainly because the triggering of transformation requires appropriate composition and energy gradient [19].

The cast ingot was reheated to 1250 °C and forged to a 30-mm-thick plate (finish forging temperature 950 °C). The forged plate was then heated at 1250 °C for 2 h for solution treatment. The billet was hot rolled to a thickness of 2 mm by seven passes in a pilot two-high hot rolling mill with a finish temperature of ~930 °C, followed by rapid cooling to room temperature to suppress the precipitation of carbides. The 2-mm-thick hot rolled strip was further cold rolled to 0.95 mm in a four-high cold rolling mill. The IQ&P processes performed in the industrial continuous annealing line are shown in figure 1 and table 2. The experimental annealing temperature ranges from 720 °C in the dual-phase region to 870 °C at the complete austenitization temperature. The annealing time was between 30 and 300 s. The quenching temperature and partitioning temperature of all

| Table 1. The chemical components of the experimental steel (wt. %). |
|-------------------|---|---|---|---|---|---|---|---|---|---|
| C     | Si     | Mn     | Cr     | Mo     | Al  | Nb | V | Cu | Ti | B | N |
| 0.09 | 0.40 | 2.00 | 0.56 | 0.2  | 0.03 | 0.02 | 0.01 | 0.01 | 0.01 | <0.001 | <0.001 |
processes were 720 °C and 300 °C, respectively, and the partitioning time was 300 s. The specimens with different annealing temperatures were labelled by temperature history as S720L3, S770L3, S820L3 and S870L3. The samples with annealing times of 30 s, 90 s, 180 s and 360 s were labelled S850L1, S850L2, S850L3 and S850L4, respectively.

Tensile samples with a gauge length of 14 mm and a width of 4 mm were obtained in the rolling direction, and tensile testing was performed at room temperature on a universal testing machine with a strain rate of 0.001 s⁻¹ on the AG-X tensile testing machine. The volume fractions of the individual phases were quantified by the standard procedure described in several references [20] by comparing the areas of the (111)γ, (200)γ, (311)γ, (110)α, and (200)α diffraction peaks with their calculated intensities. These diffraction peaks were calculated to quantify the amount of retained austenite by the following equation [21]:

\[ V_\gamma = \frac{1.4I_\gamma}{I_\alpha + 1.4I_\gamma} \]  

where \( V_\gamma \) is the volume fraction of retained austenite, \( I_\gamma \) and \( I_\alpha \) are the integrated intensities of the austenite and ferrite peaks respectively. The microstructures of the samples were investigated by FE-SEM (field emission scanning electron microscopy TESCAN VEGA 3) after electropolishing and then etching with 4% nitric acid for
20 s. The microstructures processed by different Q&P processes were extensively examined using electron backscatter diffraction (EBSD) performed at 20 kV, and the step size was 0.2 μm. Samples used for EBSD examination were electropolished in a solution of 10% perchloric acid and 90% ethanol at approximately −20°C. EBSD data were processed by Channel 5 and MTEX software [22].

3. Results

3.1. Microstructural examination

The microstructure of the test steel after IQ&P treatment at different annealing temperatures is shown in figure 2. The microstructure of the experimental steel after hot rolling (HR) (figure 2(a)) was composed of different proportions of ferrite and martensite-austenite (MA) islands. When austenitizing at 720 °C, the microstructure of the annealed cold rolled sheet presented regular banded characteristics along the rolling direction (RD) (figure 2(b)). Due to the low annealing temperature and short annealing time, the microstructure and chemical segregation band caused by rolling deformation cannot be completely eliminated, and only partial recrystallization of deformed martensite and ferrite can occur [23]. The banded ferrite was completely recrystallized into equiaxed ferrite or transformed into austenite with an increase in annealing temperature to 770 °C (figure 2(c)). The microstructure of S770L3 sample after quenching was composed of fine martensite islands and ferrite. At this temperature, the proportion of austenite in the two-phase region increased to 40% in the equilibrium phase diagram. When the temperature rose to 820 °C, it can be clearly seen from the microstructure of figure 2(d) that the ferrite and original austenite sizes increased significantly compared with

![Figure 2. SEM micrographs of low alloy steel treated with different IQ&P processes: (a) HR, (b) S720L3, (c) S770L3, (d) S820L3, (e) S870L3, (f) S850L1, (g) S850L2, (h) S850L3, (i) S850L4.](image-url)
S720L3 and S770L3 samples, and the microstructure was composed of equiaxed grains. The ferrite in S870L3 sample almost disappeared when the temperature further increased to 870 °C, which was mainly due to the annealing temperature exceeding the complete austenitizing temperature (figure 2(e)). The morphology of martensite changed from islands to continuous blocks with increasing annealing temperature from 770 °C to 870 °C, and the morphology of internal martensite laths became more obvious. Figures 2(f)–(i) shows the change in the microstructure of the low alloy steel after soaking at 850 °C for different annealing times. All the samples were completely recrystallized when they homogenized at a temperature close to Ac3 (855 °C), and the recrystallized grain size of cold-rolled steel continued to increase with the extension of annealing time. Under the condition of a short annealing time (30 s), the original austenite grain size was mostly less than 5 μm, and the length of primary martensite after IQ&P processing was less than 5 μm. When the annealing time was extended to 300 s, some original austenite grains after quenching and distribution even reached approximately 20 μm, which is four times that obtained with S850L1.

EBSD technology was used to observe the microstructure of the samples at different annealing temperatures. Figures 3 and 4 show the IPF diagrams and grain size distribution diagram of the low alloy steel. The results of figure 3 show that after annealing at 720 °C, austenite transformation did not occur because the annealing temperature was lower than Ac1, and a heterogeneous microstructure composed of amorphous banded ferrite and fine recrystallized equiaxed ferrite was formed after cooling, the average aspect ratio of banded ferrite was 3.7. The diameter of recrystallized grains was mostly less than 2 μm (figure 4(b)). When the annealing temperature increased (770–870 °C), as shown in figures 3(b)–(d), the cold rolled ferrite was completely recrystallized and partially or completely transformed into austenite. Most of the deformed ferrite was transformed into austenite and martensite during subsequent quenching during IQ&P processing at 770 °C and 820 °C, the rest formed retained austenite, and the rest recrystallized into a large amount of fine equiaxed ferrite distributed at the martensite grain boundaries. Figure 4 shows the grain size distribution of the experimental steel in the experimental temperature range. The grain size after annealing and recrystallization was mainly concentrated in the range of 0 ~ 2 μm, and the grain diameter (d) larger than 5 μm presented a small proportion. The average grain sizes vary with increasing annealing temperature. When the temperature increased from 720 °C to 870 °C, the fraction of small grains (d < 2 μm) decreased gradually, and the average grain diameter changed from 1.47 to 1.84 μm, the statistical number of effective grains of all samples was more than 900. This variation mainly considers the increase in the activation energy of recrystallization and the shortening of the recrystallization time as the temperature increases.
Analysis of samples with an annealing temperature of 770 °C under a continuous annealing process (figure 5) shows that there were a large number of 60° orientation relationships in the microstructure. Through the analysis of grain boundary misorientation in the micro region of figure 5(a), a misorientation of approximately 60° is measured at the four grain boundaries with twin relationships. The grain boundary of figure 5(c) shows that there are many similar large angle orientation relationships in the structure. The misorientation relationship between grains is analyzed, and the appearance of this high-angle misorientation relationship is caused by martensitic transformation in the structure.[24] EBSD was used to analyze the misorientation distribution when annealed at 720 °C, 770 °C, 820 °C and 870 °C, and the obtained misorientation diagram is shown in figure 6.

With increasing annealing temperature, the volume fraction of martensite increases, and the morphology of martensite gradually changes from martensite islands to martensite sheets. The peak value at a 60° angle gradually becomes sharp according to the distribution of misorientation, which may be related to the formation of lamellar martensite. The distribution of misorientation tends to be concentrated at low angles for the steel treated with different IQ&P processes, which indicates that the steel has a high dislocation density and substructure after annealing. The axis of rotation is \(\langle 111 \rangle\) when the preferential distribution of misorientation is 60°. The grain boundary misorientation of 54.74° also has a preferred distribution, and its rotation axis is \(\langle 112 \rangle\).

### 3.2. Mechanical properties

The tensile properties and strain hardening behavior of experimental steel subjected to different industrial IQ&P processes are shown in figure 7. The plasticity of the low alloy steel gradually deteriorates with increasing annealing temperature. When annealing at 720 °C, the total elongation (TE) reaches the maximum value of 13.4% as shown in figure 7(a), which is mainly due to the heterogeneous microstructure after incomplete recrystallization and a large proportion of ferrite. However, the strength of the sample does not change monotonically as TE. The yield strength (YS) and ultimate tensile strength (UTS) of S820L3 were higher than other samples, 894.4 MPa and 1076.2 MPa respectively, the strength of S870L3 unexpectedly decreases slightly when the annealing temperature (870 °C) exceeds Ac3, which may be due to the different volume fractions and
stabilities of retained austenite. Figure 7(b) reveals the complex changes in the mechanical properties after annealing at a temperature slightly lower than Ac3 (850 °C). The strength of the samples first increased and then decreased with the extension of annealing time, while the variation of TE was the opposite. When the annealing time was 90 s, S850L2 achieved the highest strength in all industrial continuous IQ&P samples, with YS and UTS of 931.5 MPa and 1096.4 MPa respectively. The strain hardening exponents of S720L3, S770L3, S820L3 and S870L3 calculated by curve the fitting method were 0.126, 0.108, 0.096 and 0.079, and S850L1, S850L2, S850L3 and S850L4 were 0.093, 0.097, 0.099 and 0.101, respectively.

To better analyze the work hardening behavior of the test steel at different annealing temperatures, the strain hardening rate (SHR) curve of the test steel is shown in figure 7(c). The curves at different annealing temperatures showed a monotonic downward trend. At the initial stage of strain, obvious work hardening
curves with small serrations were observed in the S770L3, S820L3 and S870L3 samples, and the amplitude of serration decreased gradually with increasing true strain. These small serrations were mainly caused by the TRIP effect of a small amount of retained austenite in the sample with the change in true strain, which will be discussed in detail in the discussion section.

Figure 8 shows the x-ray diffraction pattern and the volume fraction of retained austenite of samples with different annealing processes. There was almost no retained austenite in S720L3 samples. The variation in retained austenite was more complex with increasing annealing temperature, for S770 L3, S820 L3 and S870 L3 samples, it was 4.9%, 3.7% and 5.6% respectively. The microstructure in the equilibrium phase diagram was composed of 85% austenite and 15% ferrite when the IQ&P treated at 820 °C. The development of the austenite fraction led to a decrease in the average carbon content in austenite, which led to the reduction of the stability of austenite and the transformation into martensite during subsequent quenching, resulting in a lower retained austenite content than that of S720L3 sample. After IQ&P treatment at the same annealing temperature (850 °C), the minimum value of retained austenite was found to be 4.8% in S850L2 sample. The best plasticity (EL = 7.6%) and the maximum fraction of retained austenite (Vγ = 8.3%) appeared at the same point when annealing for 300 s. Figures 7(b) and 8(d) show that the plasticity of low alloy steel was positively correlated with the volume fraction of retained austenite. A higher volume fraction of retained austenite can provide better coordination of plastic deformation than ferrite and optimize the microstructure, which has been confirmed by many references on TRIP steel [25–27].

3.3. Texture analysis

Microtexture measurements were also performed to further confirm the microstructural changes in the low alloy steel after industrial continuous hot rolling and annealing processes. Figures 9(a)–(e) show the body centered cubic (bcc) crystals of hot rolled conditions and quenched partition samples in a φ2 = 45° section of the orientation distribution function (ODF), figure 9(f) schematically illustrates the typical texture components in low alloy steel, including γ-fiber (⟨111⟩//ND) and α-fiber (⟨110⟩//RD). The intensity distribution of texture components of α-fiber and γ-fiber of all samples is shown in figure 10. The microtexture characteristics of the original hot rolled (HR) sample were mainly concentrated in the components of rotating cube {001} ⟨110⟩ and {112} ⟨110⟩ as shown in figure 9(a). When annealed at a lower temperature (720 °C), some of the cold-rolled deformed grains recrystallize into new equiaxed grains, while the other part retains the strip shape deformed along the rolling direction (figure 3). In this case, a maximum intensity of {111} ⟨011⟩ and {111}
components was formed, while the strength of the rotational cube texture inherited from hot rolling deformation weakened, resulting in an increase in \( \gamma \) texture but a decrease in \( \alpha \) texture, which was also observed in reference \[28\]. After IQ&P treatment at 770 °C and 820 °C, the overall strength of \( \alpha \)-fiber and \( \gamma \)-fiber became weaker, and the texture components of the \( \gamma \)-fiber were more evenly distributed. A strong \{332\} component unexpectedly appeared in S770L3 and S820L3 samples, but it did not exist in the hot rolled samples. The intensity of \{223\}(110) and \{113\}(110) components was slightly stronger than that annealed at 720 °C, and the rotating cubic texture had little change. In the fully austenitized S870L3 sample, the strength of the \( \alpha \)-fiber and \( \gamma \)-fiber increased significantly, and the components \{110\}(110), \{111\}(110) and \{111\}(121) also became stronger. In all samples, almost no ND fiber (\((110)/ND\)) was found.

**Figure 9.** \( \phi_2 = 45^\circ \) section of the ODF representing the main orientations commonly found in low alloy steels: (a) HR, (b) S720L3, (c) S770L3, (d) S820L3, (e) S870L3, and (f) ideal orientations.

**Figure 10.** Texture components intensity distribution of the (a) \( \alpha \) texture and (b) \( \gamma \) texture in the low alloy steel.
4. Discussion

4.1. Relationship between microstructure evolution and tensile deformation

The transformation induced plasticity (TRIP) effect in Q&P and IQ&P steel is an important plastic deformation mechanism. An indirect manifestation of the TRIP effect during deformation is the serration fluctuation on the strain hardening curve. By comparing the work hardening rate curve in figure 7(c), we found that the work hardening rate of S720L3 sample was the highest among the four samples, and decreased smoothly during plastic deformation. It should be noted that the contents of ferrite and martensite in S720L3 were 68.8% and 31.2%, respectively, and there was almost no retained austenite in the microstructure. On the one hand, the extinction rate of movable dislocations in bcc martensite is much higher than that in ferrite; on the other hand, the high proportion of ferrite martensite phase interfaces makes dislocation movement more difficult in the transformation process [29–31]. Therefore, the S720L3 sample shows the highest work hardening rate. There were some small amplitudes and sparse serrations in the strain hardening curves of S770L3, S820L3 and S870L3 samples, which were mainly attributed to the discontinuous TRIP effect. This phenomenon is widely reported in typical Q&P steel and medium manganese steel [32–35]. By locally amplifying the obvious serration in figure 7(c), we found that the fluctuation process was accompanied by a sudden drop and rapid rise of strain hardening rate. The peak stress from A1 to B2 can be divided into two stages (Fig. 6.7(b)), namely, A1 to B1 and B1 to A2. Stage A1 to B1 belong to the strain induced transformation from retained austenite to martensite, and the corresponding work hardening rate curve of this stage rises rapidly (figure 7(c)), which can be explained as follows: when the true stress reached the critical value A1 for the phase transformation of the corresponding stability level retained austenite, the martensitic transformation will be activated, resulting in a TRIP effect, and the TRIP enhancement offsets part of the true stress, resulting in stress relaxation; at the same time, the stress was transferred to the surrounding ferrite and retained austenite, resulting in the continuous phase transformation of the surrounding ferrite and retained austenite with the same stability level, in which increasing external force competes with the stress relaxation caused by the enhancement effect produced by the TRIP effect, which eventually leads to the temporary sudden drop of true stress and the rapid rise of work hardening rate curve. Stage B1 to A2 belongs to the stage of increasing and accumulating external stress, which was accompanied by a rapid decline in the work hardening rate. When the increase and accumulation of external stress reach critical point B2 for the transformation of retained austenite in the next batch with a higher stability, a new round of strain will induce retained austenite to martensite and produce TRIP effect. S770L3 and S820L3 samples showed large serrations in the range of true strain 0.01 to 0.015, which is due to the strong TRIP effect in the early stage of strain, while S870L3 samples had a mild TRIP effect in the whole strain process. Discontinuous

In conclusion, the key to the discontinuous TRIP effect of the experimental steel is as follows: first, the experimental steel needs to have a certain amount of retained austenite, which transforms into martensite and expands in volume in the tensile process, to offset part of the stress concentration and transfer the stress to the surrounding phase, resulting in synergistic deformation, accompanied by stress relaxation and transfer; second, the retained austenite in the experimental steel needs the stability of different grades, and the TRIP effect can be produced when the stress value reaches the critical value of phase transformation of retained austenite in this grade.

4.2. Texture evolution

It is well known that the microtexture components of cold rolled low alloy steel are the strong cubic textures {112}, {110}, {111}, and weak {111}, and the recrystallization textures of low alloy steel are {111} and {111} [36–38]. The large fraction of the rotated-cube component in the HR steel was attributed to the relatively low finishing rolling temperature, at which the plane strain texture could be conserved with a weak softening phenomenon [39, 40]. The reason for this difference is that recrystallization nucleation usually occurs for high-storage energy orientations, and the deformation storage energies of the main texture components from high to low are {111}, {111}, {110}, {112}, {110}, {001}, {110}. The orientation {001} is a low-energy storage orientation, and the orientation {111} is a high-energy storage orientation, which may have a larger cell size/cell dislocation distribution (and corresponding larger energy storage difference), which is more conducive to the preferred recrystallization of grains.

In S770L3 and S820L3 samples, the strengths of {113}, {110} and {332} were the highest. However, compared with S720L3 samples, S770L3 and S820L3 steels fully recrystallized at higher temperatures, which seems to play a key role in this observation. To understand the reason for the highest texture strength of {113}, {110} and {332} in the S770L3 and S820L3 samples, it is necessary to understand the formation of recrystallized ferrite nuclei and the orientation of parent austenite grains. Cold rolling produces a deformation band in the microstructure, which is conducive to recrystallization and nucleation in subsequent annealing. The nucleation of recrystallization in the grain boundary region of the original austenite had no orientation.
relationship with austenite, but the nucleus in the crystal had a direct relationship [41]. The nuclei formed and grown from the grain bonding zone have random orientations, but the nuclei formed in the {110} {112} and {112} {111} grains will be oriented in {332} {113} and {113} {110} directions. For S870L3 sample annealed at a higher temperature, the nucleation rate at the prior {110} {112} austenite grain boundary was higher than that of coherent {332} {113} grains, which led to a decrease in the strength of the {332} {113} composition, which is consistent with the texture components observed in references [42] after Q&P treatment. In addition, the expected {332} {113} texture components cause less anisotropy in strength and toughness, so it is necessary to strengthen the intensity of the {332} {113} texture components from the perspective of improving strength and toughness [43]. Different annealing temperatures and annealing times are the two most important means to control the recrystallization and transformation texture of advanced high-strength steel (AHSS) and affect the mechanical properties of steel [44–48]. Therefore, the composition and intensity of the recrystallization texture are different at different annealing temperatures. As shown in table 3, with increasing annealing temperature, the recrystallization texture fraction increases and the rolling texture fraction decreases. After annealing in the two-phase region (770 °C and 820 °C), the IQ&P treatment samples had the maximum proportion of {332} {113} texture, and its volume fraction reached 13.1% and 12.3% when the maximum deviation was 15°. The {112} {110} component is very favorable in terms of rolling mechanical properties, and its texture fraction is generally the highest among the hot rolled steel plates below the critical recrystallization temperature ($T_{NR}$). In contrast, the {332} {113} texture is the most favorable component of the phase transformation texture, which increases strength and impact toughness [36, 49–51]. The texture obtained from recrystallized austenite is composed mainly of {100} {110}, and the {100} {001} texture derived from parent austenite is harmful because it will lead to delamination and crack propagation [52]. The {001} {110} component reduces the impact toughness of the steel, especially in the direction of 45° relative to the rolling direction. The loss in formability caused by one {001} orientation texture needs to be compensated by 4 or 5 {111} textures. In the transformation of texture, the {001} {110} texture is not conducive to the toughness of low alloy steel and always makes the material brittle because it easily leads to delamination and crack propagation [36, 53]. Therefore, the harmful effect of the {001} {110} components on the toughness is very obvious in hot rolled steel plates. Because of these harmful effects on toughness, it is necessary to reduce the {001} {110} texture components in most cases to obtain a good combination of strength and toughness. The {112} {110} component is a very stable orientation [36], which is conducive to the improvement of toughness in the rolling direction, but it will produce anisotropy in toughness and make the material brittle in the direction 45° relative to the rolling direction. This is also a significant reason that S770L3 has better plasticity than S820L3. The fact that S870L sample contained more coarse martensite structures and that no quasi-polygonal or polygonal ferrites were recorded are likely the reasons that this steel has rather poor toughness. It is known that martensite has a lower toughness than ferrite constituents, and the {100} {001} texture was also responsible for the anisotropy in the strength of the S870L3 specimen. On the other hand, according to figure 6, the S820L3 sample had the high-angle grain boundaries. According to Diaz Fuentes and Gutierrez [54], the toughness of the acicular ferrite microstructure was related to its density of high-angle boundaries, which forces crack cleavage to change the micropropagation plane to adapt to the new local crystallography [54], which leads to higher tensile strength of S820L3 than S770L3 and S870L3.

### 5. Conclusions

The evolution of the microstructure and mechanical properties of a low alloy steel after hot rolling and cold rolling in different industrial IQ&P processes (annealing temperatures varying from 720–880 °C) was investigated. The main conclusions are as follows:

| Samples    | {001} | {113} | {225} | {111} | {332} |
|------------|-------|-------|-------|-------|-------|
|            | (110) | (110) | (110) | (011) | (113) |
| S720L3     | 7.2   | 11.3  | 14.9  | 13.7  | 8.0   |
| S770L3     | 7.3   | 12.2  | 11.5  | 7.2   | 13.1  |
| S820L3     | 7.4   | 12.3  | 9.4   | 9.3   | 12.3  |
| S870L3     | 6.5   | 11.9  | 10.7  | 9     | 10.6  |

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Table 3. Texture components with different annealing temperatures (STs).
1. The yield strength, tensile strength and elongation of low alloy steel treated by different IQ&P processes are 602–931 MPa, 968–1096 MPa and 6.3%–13.4%, respectively. The microstructure after annealing at 720 °C presented a heterogeneous structure of banded ferrite and fine recrystallized grains.

2. The S770L3, S820L3 and S870L3 samples show multiphase structures of lath martensite, fine ferrite and retained austenite after annealing at 770 °C, 820 °C, and 870 °C, respectively. The retained austenite transformed into martensite discontinuously in the subsequent tensile deformation, i.e., the discontinuous TRIP effect.

3. The low alloy steels processed with HR and IQ&P treatment at 720 °C steel exhibited high intensity cube {001} (110) and copper {112} (111) textures. The fraction of the textures {111}{112} and {111} (110) decreased with increasing annealing temperature, while the fraction of [100] {001} and [001] (110) increased, and the texture components gradually changed from {113} {110} to {111} (110). Among the samples heated at 770 °C and 820 °C, the beneficial components {113} {110} and {332} {113} had the greatest strength, which indicates that the low alloy steel had a good combination of strength and toughness.

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Data availability statement

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

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