Recrystallization Behavior of Warm Rolling and Cold Rolling Cr-Ti-B Steel during Annealing

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1. Introduction

Low-carbon (LC) steel is often used as an automobile overlay because of its good deep drawing properties. The γ-fiber texture is beneficial, while the α-fiber texture is unfavorable, with respect to the deep drawing properties of steel. Warm rolling is an effective process to increase the sharpness of γ-fiber textures and has been used widely in interstitial-free (IF) steel [1–3]. However, warm rolling failed to match low carbon steel well, because the annealing of warm rolled low carbon steel often produced strong Goss textures, which is unfavorable to deep drawing performance. However, Gazder et al. [4] found that warm rolling followed by cold rolling could further the strengthen γ-fiber texture and weaken the Goss texture. After warm rolling and cold rolling, shear bands, deformation bands and many grain boundaries were observed in the LC steel. The evolution of these positions during the subsequent recrystallization process and their influence on texture are among the more complex and less studied aspects of steel production.

Because of the presence of interstitial carbon atoms during the annealing of LC steel, the Goss texture is enhanced, and the final γ recrystallization texture is weakened [5–7]. The addition of carbide-forming elements can reduce the influence of C atoms. Toroghinezhad et al. [8] found that adding chromium as an alloying element to LC steel enhanced the γ-fiber texture. In addition, more deformed grains containing shear bands were obtained by warm rolling LC steel to enhance the γ-fiber texture. Haldar [9] found that sharper γ-fiber textures could be formed by warm rolling than conventional cold rolling. Guo [10] performed cold rolling after warm rolling to further optimize the texture of IF steel. However, the recrystallized texture of LC steel generally cannot maintain sharp γ-fiber textures before annealing, which is related to the recrystallization mechanism during annealing. The nucleation mechanism of LC steel is complicated and involves three types of nucleation sites, shear bands, grain boundaries and deformation bands.

It is generally believed that the γ-fiber texture nucleates preferentially in shear bands, and the addition of Cr, Ti and B can increase the amount of shear band structure in
However, the Goss texture, which is not favorable for deep drawing properties, also nucleates at certain shear bands. Jonas et al. [11] observed the formation of Goss components by nucleation in the vicinity of intense shear bands in cold rolled LC steel. Grain boundaries are common recrystallization nucleation sites. Nuclei of $\gamma$-fiber have been observed frequently at grain boundaries, especially at deformed $\gamma$-fiber grains [12]. Barnett [14] expected that the presence of large misorientations in $\gamma$ grains in deformed IF steel enhanced nucleation in $\gamma$ grains during annealing. The recrystallization texture produced in LC materials was affected by specific lattice rotations near grain boundaries. This behavior was more likely to occur for non $\gamma$ deformed grains without shear bands, such as [110]-oriented grains, which led to the formation of $\alpha$-fiber during annealing [15]. In addition to shear bands and grain boundaries, recrystallization nucleation of deformation bands has been reported in LC and IF steel [16–18]. In summary, several nucleation modes are involved in the recrystallization of LC steel during annealing. Thus, the desired texture can be achieved by understanding the recrystallization mechanism.

In this paper, Cr-Ti-B LC steel underwent warm rolling followed by cold rolling. Three recrystallization nucleation points were observed at shear bands, grain boundaries and deformation bands. The effects of different deformed microstructures or nucleation positions on the final recrystallization texture were investigated. Research on the recrystallization of LC steel after warm rolling followed by cold rolling can help provide a theoretical basis for the wider application of low-carbon steels.

2. Materials and Methods

The experimental steels (chemical composition provided in Table 1) were melted via vacuum induction under argon. The ingots were forged into 130 mm $\times$ 100 mm $\times$ 20 mm billets. Each billet was heated to 1200 $^\circ$C and held for 2 h. Then, each billet was hot rolled to 8 mm thickness at a finishing temperature of 950 $^\circ$C and air cooled to room temperature. The hot rolled sheets were pickled to remove the oxide coating and then protected by an oil coating. A linear cutting machine was used to divide the hot rolled plates into rectangular samples with a size of 100 mm $\times$ 50 mm $\times$ 5 mm for warm rolling. The hot rolling plate was warm rolled by two high hot rolling mills. The hot-rolled plates were heated to 450 $^\circ$C, held for 0.5 h, rolled to 1.5 mm thickness with a reduction rate of 70% and water cooled to room temperature. Finally, the warm rolled plates were descaled and rolled to 1 mm thickness at room temperature with a reduction of 33.3%. The Ac1 and Ac3 temperatures of the samples were 771.9 and 923.4 $^\circ$C, respectively, according to the phase transition point test. The annealing experiment of the rolled plates was carried out on the thermal simulator MMS-100. The rolled plates were heated to 660, 680, 700, 720 and 740 $^\circ$C at a heating rate of 10 $^\circ$C/s, held for 2 s, and then cooled with water to room temperature.

Table 1. Chemical composition of Cr-Ti-B low-carbon steel (wt.%).

<table>
<thead>
<tr>
<th></th>
<th>C</th>
<th>Mn</th>
<th>Cr</th>
<th>Ti</th>
<th>Al</th>
<th>B</th>
<th>N</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cr-Ti-B steel</td>
<td>0.037</td>
<td>0.19</td>
<td>0.480</td>
<td>0.015</td>
<td>0.012</td>
<td>40 ppm</td>
<td>24 ppm</td>
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The microstructure and texture were observed in the rolling direction-normal direction (RD-ND) plane using scanning electron microscopy (SEM; Carl Zeiss AG, Jena, Germany) coupled with electron backscatter diffraction (EBSD; OXFORD Corporation, Oxford Instrument Technology Co., Ltd., Shanghai, China), with Channel 5.0 HKL software (Tango, Mambo and Salsa model, $\phi 2 = 45^\circ$ cross-section, OXFORD). The EBSD samples were electrolyzed at 20 V and 1.5 A, and the electrolyte consisted of a 1:2:7 perchloric acid:glycerinalcohol mixture. The orientation data were analyzed using the following grain definition: misorientation higher than 5 $^\circ$, minimum 4 points per grain and points with a confidence index (CI) lower than 0.1 were not considered dubious in the data analysis. The textures are represented as orientation distribution functions (ODFs) using Bunge notation and contain approximately 3000 grains. The substructure of the LC steel was...
examined and analyzed by transmission electron microscopy (TEM, TECNAI G2 F20). The TEM samples were prepared by a focused ion beam (FIB) treatment.

3. Results and Discussion
3.1. Microstructure

The microstructure of Cr-Ti-B LC steel after cold rolling, as shown in Figure 1a, consisted of slender deformed grains that contained shear bands. The microstructures at different continuous annealing temperatures are shown in Figure 1b–f. Recrystallization was not obvious at 660 °C according to Figure 1b, and the specimen was probably in the static recovery stage. When the annealing temperature was 680 °C, many recrystallized grains were present. The shear band is located in the grain and is generally 45° from the rolling direction, so the recrystallization of nucleation in the shear band also has the same characteristics. Grain boundary nucleation can be clearly observed from the recrystallization nucleates along the grain boundaries. The nucleation of recrystallization was observed in shear bands (yellow dotted lines) and grain boundaries (white dotted lines) in Figure 1c. As shown in Figure 1d,e, the deformed grains with shear bands recrystallized completely, but a few de-formed grains without shear bands did not fully recrystallize. At 740 °C, the Cr-Ti-B LC steel is completely recrystallized, as shown in Figure 1f. Observations of the recrystallization process at different annealing temperatures showed that the shear bands and grain boundaries nucleated simultaneously at the initial stage of recrystallization. However, during annealing, the shear bands completed recrystallization more quickly than the grain boundaries; this indicated that the nucleation rate was faster at the shear band than at the grain boundaries, which was attributed to their different recrystallization nucleation mechanisms.

![Figure 1. Microstructure of Cr-Ti-B LC steel at various continuous annealing temperatures: (a) initial microstructure, (b) 660 °C, (c) 680 °C, (d) 700 °C, (e) 720 °C and (f) 740 °C.](image)

3.2. Texture

When the microstructure of the LC steel transitioned from deformed grains to recrystallized grains, the final recrystallization texture was affected by rotation of grain orientation. The textural transition of the Cr-Ti-B LC steel during annealing is shown in Figure 2a–f. The initial texture of the LC steel had the characteristics of typical ferrite rolling textures, i.e., a combination of α-fiber and γ-fiber textures (Figure 2a). The intensity of γ texture in warm rolled and cold rolled plates was stronger than that of the warm rolled...
plate, while the $\alpha$ texture intensity of warm rolling and cold rolling plates was weaker than that of the warm rolling plate in the literature [5,8]. When annealed at 660 °C, the texture did not change significantly from the initial texture, which was attributed to the low degree of recrystallization in Figure 2b. When the temperature was increased to 680 °C, the sharpness of the $\gamma$-fiber texture decreased, and a Goss texture appeared (Figure 2c).

It is worth noting that the $\alpha$-fiber texture was enhanced at 680 °C. This is because at this temperature, the deformed $\gamma$ grains with shear bands were completely recrystallized, while the deformed $\alpha$ grains without shear bands were not completely recrystallized. The $\gamma$ deformed grains with shear bands did not completely evolve into $\gamma$ recrystallized grains, which was the reason for the weak $\gamma$-fiber and strong $\alpha$-fiber textures. The $\alpha$ texture strength decreased gradually, while the $\gamma$ texture strength remained unchanged with the increase in annealing temperature, as shown in Figure 2d–f. This result indicated that a large amount of recrystallization nucleated at the grain boundaries of deformed $\alpha$ grains, and the deformed grains did not all transform into $\alpha$ recrystallized grains, which was more likely to transform into $\gamma$ recrystallized grains. Although some Goss textures were retained in the final recrystallization texture, compared with the Goss texture after annealing of the traditional warm rolled plate, the Goss texture of the warm rolled and cold rolled plate was significantly weakened. HAVE speculated that once the shear band is formed, it will not operate continuously in the remaining deformation, but have a limited life; continuous activation and deactivation of shear bands play a key role in the oriented fragmentation of ND fiber grains in cold rolled and warm rolled steels [5,14].

![Figure 2](image_url)

**Figure 2.** Texture transition of Cr-Ti-B LC steel at various continuous annealing temperatures: (a) initial texture, (b) 660 °C, (c) 680 °C, (d) 700 °C, (e) 720 °C and (f) 740 °C.

In order to illustrate the texture transformation path at different annealing temperatures, the orientation line density distribution of $\alpha$ fiber and $\gamma$ fiber textures is shown in Figure 3. The orientational density of the $\alpha$-fiber texture during recrystallization in Figure 3a, [223]<110> was dominant at the initial texture, and sharp [223]<110> and [116]<110> peaks were observed during annealing at 660 °C. When the annealing temperature increased to 680 °C, the deformed microstructure recrystallized, and the density of the [116]<110> orientation increased, while the density of the [223]<110> orientation decreased, which indicates that recrystallization was more likely to nucleate in deformed grains with the [223]<110> orientation than in those with the [116]<110> orientation. When the temperature
reached above 700 °C, most of the deformed grains transformed into recrystallized grains, and the overall α orientational density decreased. For the γ-fiber texture, the orientational density decreased first and then remained constant with increasing annealing temperature. The orientational densities of {111}<110> and {111}<112> were consistent above 680 °C. The change in the orientational densities of α and γ was related to the recrystallization nucleation process at the grain boundaries and shear bands. Therefore, it is necessary to discuss the mechanism for the nucleation of recrystallization at different positions.

Figure 3. The orientational density of the α-fiber and γ-fiber textures at various continuous annealing temperatures: (a) α-fiber texture and (b) γ-fiber texture.

3.3. Nucleation and Recrystallization in Shear Bands

Preferential nucleation of the shear band was related to its formation and nucleation mechanism. The substructure of the rolled plate is shown in Figure 4a,b. A shear band is formed by microbands through dislocation slip and rigid body rotation [19,20]. In the rolling process, uneven deformation in the cellular substructure area of LC steel results in the formation of non-equiaxial cellular substructures that are generally termed microbands. Microbands stack one by one to replace the normal cellular substructure as shown (Figure 4a), and then, dislocation slip causes the microbands to kink and form S-bands. Rigid-body rotation in the kinking regions of microbands eventually forms shear bands (Figure 4b). The formation mechanism of shear bands determines that shear bands belong to long strip-like substructures formed by dislocation movement. Thus, the recrystallization mechanism of the shear bands was more likely to be subgrain coalescence, and the subgrains were elongated in the shear bands. Unlike equiaxial subgrains, elongated subgrains quickly accumulate large misorientations with the surrounding matrix in the process of coalescence [21,22]. In summary, recrystallization nucleation preferentially occurs at shear bands due to its high dislocation density and special nucleation mechanism in warm rolled plates.
In the process of recrystallization nucleation in a shear band, [111] and Goss recrystallization textures have been mentioned the most. Goss recrystallization was generally considered nucleation at shear band in warm rolled LC steel. In this experiment, the recrystallization process on the shear band was observed in Figure 5. Goss subgrains appeared preferentially within striated structures, such as shear bands, in Figure 5a, which resulted in the formation of Goss recrystallization textures at shear bands during annealing. The misorientation between Goss and [111]<112> recrystallized grains was <110>30°, indicating that the grain boundaries between Goss and [111]<112> recrystallized grains had the highest migration rate [23]. Goss recrystallization could increase by swallowing the surrounding [111] deformed matrix at the initial stage of recrystallization. However, with the progress of recrystallization, there was a competitive relationship between the growth of Goss and [111]<112> recrystallized grains, which was the cause of the decrease in Goss texture at the final recrystallization stage in Figure 2f. In the shear band, except for the Goss and [111] oriented grains, other oriented recrystallized grains were rare.

In addition to shear bands, grain boundaries were also the main areas of recrystallization nucleation. The recrystallization nucleation in shear bands was dominant in the warm rolled plate, while the recrystallization nucleation in grain boundary plays a major role in the warm rolled and cold rolled plates [4]. During the deformation process, larger dislocations accumulated at the grain boundaries, and the energy storage was higher. Local
misorientation indirectly reflected the dislocation density in the initial and recovery stages of Cr-Ti-B LC steel (Figure 6a,b). Local misorientation at the grain boundaries decreased in the recovery stage, indicating a reduction in the dislocation density. In the recovery stage, subgrains formed by eliminating point defects and canceling and rearranging dislocations, which did not involve the migration of high-angle boundaries (white arrow). The dislocation density at most of the grain boundaries changed obviously, while the dislocation density at shear bands changed slightly. According to the calculation, the average misorientation of the initial stage and recovery stage was 0.39 and 0.38, respectively, and the proportion of low angle grain boundaries was 66.2% and 64.8%, respectively (Figure 6c,d).

Figure 6. Local misorientation and grain boundaries at the initial and recovery stages: (a) distribution of local misorientation at the initial stage, (b) distribution of local misorientation at the recovery stage, (c) average local misorientation and (d) grain boundaries.

Subgrains and recrystallized grains at a grain boundary are shown in Figure 7. When the annealing temperature increased from 660 °C to 680 °C, recovery and recrystallization occurred. During the recovery stage, dislocations rearranged to form subgrain boundaries, as shown in Figure 7a. There were two main mechanisms of grain boundary nucleation, including strain-induced grain boundary migration and subgrain coalescence. The strain-induced migration mechanism occurred in low deformation metals, while subgrain coalescence occurred in the high deformation metals [24]. Certain misoriented subgrain boundaries had higher activity, engulfed the surrounding subgrains and gradually transformed into high-angle grain boundaries, an outcome of the direct growth of some of the subgrains. After warm rolling and cold rolling, the reduction in Cr-Ti-B low carbon steel reached 80%, which belonged to large deformation, and the dislocation density between deformed grains of warm rolling and cold rolling Cr-Ti-B steel was close (Figure 6a). Therefore, it is inferred that the main nucleation mechanism of recrystallization at the grain boundary may be subgrain coalescence.
The recrystallization of α and γ grain boundaries is shown in Figure 8. When recrystallization occurred at the boundaries of α and γ grains, a large number of [111]-oriented grains and a small number of Goss and cube-oriented grains were observed. The grains with Goss orientation at grain boundaries were few, and they were easily absorbed by [111]<112>-oriented grains; the main nucleation area of the Goss texture remained the shear bands. Although the amount of cube-oriented texture was not dominant at the early stage of grain boundary nucleation, the misorientation between cube-oriented and [111] grains was approximately <231>56°, and the grain boundaries between them did not have a high migration rate. Thus, the [111] grains had no direct competitive relationship with cube-oriented grains. The cube-oriented grains occupied a certain position in the final recrystallization texture because of their prior size advantage. It can be found from the previous experiments that the recrystallization nucleation at the grain boundary of γ grains dominates the whole recrystallization process.

3.5. Nucleation and Recrystallization in Deformation Bands

During rolling, certain grains undergo orientation splitting where they are separated by a sharp boundary or gradually overbent from one orientation to another; this process is termed deformation banding [18]. The nucleation of two deformation bands is shown in Figure 9a. The transition from one orientation to another and recrystallization were observed in the grain and deformation bands, respectively. In the deformation band, the recrystallization mechanism was more likely to be subgrain coalescence due to the higher dislocation density. The recrystallization was found in the deformation bands at 700 °C, but recrystallization grains on the deformation band were much less than those at the grain boundary and shear band. This is because recrystallization occurred easily at the grain boundary and shear band.
boundaries due to the large distortion, and recrystallization occurred in shear bands prior to deformation bands due to the special nucleation mechanism. Therefore, recrystallization of the deformation band was not the main mechanism of recrystallization nucleation in Cr-Ti-B low-carbon steel.

![Recrystallization of a deformed band at 700 °C](image)

**Figure 9.** Recrystallization of a deformed band at 700 °C: (a) inverse pole figure (IPF), (b) DB1, (c) DB2 and (d) pole figures of DB1 and DB2.

4. Conclusions

Based on warm rolling and cold rolling Cr-Ti-B low carbon steel, three different recrystallization nucleation sites were observed. The effects of different deformed microstructures or nucleation positions on the final recrystallization texture were investigated.

1. Compared with the warm rolling process, warm rolling and cold rolling processes can obtain stronger [111] deformation textures, which are conducive to the retention of [111] texture strength during annealing.

2. Due to the high dislocation density and the nucleation mechanism of the merging of slender subgrains, recrystallization generally nucleated preferentially at the shear band. The recrystallization nucleated at the shear band was mainly Goss orientation, which was due to the existence of the Goss subgrain. There is a high grain boundary mobility between Goss texture and [111] texture, so Goss grains grow by swallowing the surrounding [111] deformed matrix at the initial stage of recrystallization.

3. The preferential recrystallization of grain boundaries was observed because of the high grain boundary energy. Although the recrystallization nucleation mechanism of the shear band and grain boundary was the subgrain combination, the recrystallization nucleation of the shear band was faster because of its slender subgrain. The texture component of recrystallization nucleation on grain boundaries was mostly [111]. Therefore, [111] texture was enhanced, and Goss texture was weakened after complete recrystallization.

4. During the annealing process, recrystallization in the deformation bands was observed at higher annealing temperatures. The recrystallization nucleation in deformation bands was much slower than that in shear bands and at grain boundaries. This is because recrystallization occurred easily at the grain boundaries due to the large distortion, and recrystallization occurred in shear bands prior to deformation bands due to the special nucleation mechanism.

**Author Contributions:** Conceptualization, methodology, Z.W.; investigation, A.H.; data curation, A.H. and Q.Y.; writing—original draft preparation, A.H.; writing—review and editing, Z.W., R.C., J.Q., Y.Z. and W.L. All authors have read and agreed to the published version of the manuscript.

**Funding:** This study was supported by the National Natural Science Foundation of China (Grant No. 51704132), the Key Research and Development Program of Jiangxi Province (Grant No. 20192ACB50010 and 20192BBEL50016), and the Program of Qingjiang Excellent Young Talents, Jiangxi University of Science and Technology (JXUSTQJBJ2020007).
Data Availability Statement: The data presented in this study are available in the article.

Conflicts of Interest: The authors declare no conflict of interest.

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