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Microstructure and texture evolution of ultra-thin high grade non-oriented silicon steel used in new energy vehicle

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Abstract

The evolutions of microstructure and texture of ultra-thin high-grade non-oriented silicon steel for new energy vehicles were investigated in this paper, and the formation mechanism of typical recrystallized $\alpha^\ast$-fiber texture was described. The results show that: The microstructure of the hot rolled plate was inhomogeneous along the thickness direction because of the shear force and temperature gradient, resulting in the texture in each layer of the hot rolled plate appearing rotational distribution around the Goss orientation. After normalizing, the fully recrystallized microstructure was obtained, and $\alpha^\ast$-fiber texture was formed. Banded structure was obtained in the cold rolled sheet, with $\alpha$-fiber texture $\{114\}<110>$ dominated. The typical $\alpha^\ast$-fiber texture was formed after annealing recrystallization, which mainly consist of $\{114\}<481>$ and $\{113\}<361>$. The $\{114\}<481>$ oriented grains were mainly nucleated within the deformed $\{114\}<110>$ grains and at the grain boundaries of $\alpha$-fiber deformed grains, and without size, quantity and strength advantages. $\{111\}<112>$ is dominated in the early stage of recrystallization, but $\{114\}<481>$ became the main texture with size, strength and quantitative advantages in the late stage of recrystallization.

1. Introduction

Environmental protection and energy issues present severe challenges to the sustainable development of the automobile industry. With the shortage of non-renewable resources, the development of new energy vehicles can reduce the consumption of the country’s petroleum resources, and reduce the emission of atmospheric pollutants during the operation of vehicles, which is of great significance in adjusting the energy structure, improving urban air quality and protecting people’s health.

New energy vehicles put forward hardcore index technical requirements \cite{1} named H\textsuperscript{3} (High magnetic induction, High-frequency low-iron loss, High strength) for non-oriented silicon steel, and ultra-thin non-oriented silicon steel below 0.35 mm has become irreplaceable core material for driving motor of new energy vehicles. At present, a few enterprises such as Nippon Steel, JFE, POSCO, Baotou Steel Union Co., Ltd and Capital Steel Group have the capacity to supply electrical steel for new energy vehicles. Each enterprise keeps their own production technology as corporate secrets, and there is a little related research reports.

The difference in thickness is the essential difference between the non-oriented silicon steel for new energy vehicles and the traditional high-grade non-oriented silicon steel, namely, the different cold rolling reduction rates. The cold rolling reduction rate of traditional non-oriented silicon steel is 75% ~ 80%, while that of non-oriented silicon steel for new energy vehicles is above 90%. The recrystallization texture of non-oriented silicon steel is obviously affected by the over 90% reduction rate, and the breakthrough of texture control technology is the processing core of ultra-thinning.

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Zhang Y X [2], Hölscher [3], Paolinelli [4] et al found that with the continuous increase of reduction rate, the number of shear bands in cold rolled sheet decreased gradually. The reduction rate was 91.3%, with only a small amount of shear bands in the center of thickness. When the reduction rate was 93.48%, the shear bands almost disappeared. When the reduction rate reached 95.65%, the shear bands were completely replaced by fiber structures.

Quadir [5] et al found that when the reduction rate was over 85%, the α-deformed grains would undergo orientation splitting, forming the deformed matrices such as [001] <210>, [114] <481> ~ [113] <361> which were beneficial to the nucleation of α grains. Moreover, there was a large deformation energy storage accompanied by orientation splitting, which also created conditions for the nucleation of low energy storage deformed matrices such as [001] <210>, [114] <481>.

Kang [6] et al showed that when the reduction rate increased to 85% ~ 95%, [211] <111> was decomposed into {112} <110> and {100} <011> with the new slip system activated. When the reduction rate was 85%, {112} <110> was the strong point. When the reduction rate increased to 95%, the strong point became {100} <011>.

Yunbo X [7] showed that {111} <112> and {223} <110> became the main texture after annealing when the cold rolling reduction rate was 91%; it was also found that although the cold rolling reduction rate reached 91%, a small amount of {100} <0vw> texture remained, which made the existence of weak cube texture in recrystallized texture.

Yang P [8–10], Gobernado P [11], and other studies found that with the increase of cold rolled reduction rate, strong {113} <361> and {114} <481> α*-textures could form in the recrystallization texture. Part of these nucleated and grew by subcrystalline aggregation at grain boundaries of {110} <001> and {100} <011> and within {112} <110> deformed grains. The others were {110} <229> orientation rotating to {113} <361>, because the increase of shear strain led to the change of grain rotation path.

Li Xie [12] found that when annealed at 1000 °C, diffuse {100} and {110} textures were obtained at 87.5% reduction rate, and mixed textures of α, Goss and {114} <481> were obtained at 90% reduction rate.

Hommra [13] found that after more than 90% large deformation, {h,1,1} <1/h,1,2> nucleus formed at the grain boundaries of α-fiber microstructure and were difficult to form in the grain interior. However, Yasuda [14] suggested that at the early stage of recrystallization, [411] <148> oriented grains could be nucleated inside α-fiber grains. Li [15] et al believed that the inhomogeneous deformed microstructure provided the sites for the [411] <148> nucleus, which was mainly located at the grain boundaries of {112} <110> and {111} <uvw> and inside the deformed grain {111}uvw, and directional growth and directional nucleation acted together. Hutchinson [16] and Qin [17] showed that high driving force of grain growth, high mobility of 20°~45° grain boundaries, apparent grain size, and quantitative advantages jointly led to selective preferential growth of {114} <481> grains. Zu [18] et al think that the selective growth mechanism caused by orientation pinning was the main reason for the formation of α* recrystallization texture.

In summary, for over 90% reduction, the number of cold-rolled shear bands decreases significantly, but the types of recrystallized texture and mechanisms of nucleation and growth are inconsistent. The reasons why they are different are the large cold rolling reduction rate of ultra-thin non-oriented silicon steel, the thin cold rolled sample, the relatively small plane compressive stress, and the strong shear effect, which leads to the orientation splitting of deformed grains, the rotation of grains around the transverse (TD) and normal (ND) axes simultaneously, and the starting of the new slip system, resulting in the complexity of grain orientation rotation.

In this work, the microstructure and texture in the preparation of 0.23 mm ultra-thin non-oriented silicon steel were systematically studied, and renucleation mechanism and the evolution of microstructure and texture for recrystallization under the 90% reduction rate will be fully clarified.

2. Experimental prototype steel preparation

The preparation process of the experimental prototype steel was: Vacuum furnace melting → vacuum casting → furnace cooling → stripping → soaking → hot rolling → normalizing → acid pickling → cold rolling → annealing. The chemical composition of the experimental steel is shown in table 1. Main technological parameters: billet size 260 mm(length) × 240 mm(width) × 30 mm(height); the billet was prepared to 1150 °C in soaking furnace first and after 5 passes hot rolled to 2.30 mm with the 860 °C final rolling temperature, and then laminar cooled to 620 °C. The hot rolled process was simulated through the above process. The hot rolled plate was normalized at 900 °C for 5 min in N₂ protective atmosphere. The normalizing plate was cold rolled to 0.23 mm after pickling, then annealed in 100% N₂ atmosphere at 980 °C, 1000 °C, and 1030 °C for 3 min, respectively. After that, the annealed sheet was air-cooled to room temperature. The formulation of the annealing process mainly simulates the field continuous annealing furnace, of which the upper limit of the heating temperature is 1040 °C, so the upper limit of the simulation process temperature is...
Table 1. Chemical composition of tested steel (wt.%).

|   | C  | Si | Mn | Al | P   | O, N, S |
|---|----|----|----|----|-----|---------|
|   | 0.0044 | 2.76 | 1.08 | 0.603 | 0.036 | <30ppm |

Table 2. Magnetic properties of annealed sheet annealed at different temperature.

| Annealing time min⁻¹ | 3 |
|----------------------|---|
| Annealing temperature °C⁻¹ | 980/1000/1030 |
| $P_{1.5/50}$ (W kg⁻¹) | 2.68/2.31/2.0 |
| $P_{1.0/400}$ (W kg⁻¹) | 16.33/15.68/14.24 |
| $B_{50}$ (T) | 1.737/1.770/1.781 |

1030°C. The length of the annealing furnace is about 300 meters, and the strip speed is 80 ~ 120 m min⁻¹, so the heating time is set to 3 min. The simulation process is consistent with the field process, so the experimental results are beneficial in guiding the production.

Hot rolling plate, normalizing plate, cold rolling sheet, and annealing sheet were prepared into metallographic specimens measuring 15 mm (RD) × 10 mm (TD), respectively, and they were hot mounting by the XQ-1 metallographic sample mounting press. The samples were ground step by step on 120 # ~ 2000 # water sandpaper, and then they were mechanically polished on FG-1 metallographic sample polishing machine. Subsequently, 4% nitric acid alcohol solution was used to erode their surfaces for about 15 s. After blow drying the samples, metallographic observation and image acquisition were carried out with Olympus GX51F inverted metallographic microscope. The average grain size was calculated by the intercept method, with 10 photographs per sample.

Samples of each process were prepared into 30 mm (RD) × 25 mm (TD) samples by wire cutting. The detection surfaces were ground in turn on 120 # ~ 1200 # sandpaper, and then they were electrolytic polished in 10% perchloric acid alcohol solution to eliminate surface stress. The macro-texture was detected by X’ Pert PRO MPD x-ray diffractometer of PANalytical B.V. The polar maps of {110}, {200} and {211} planes were measured, respectively. Based on the polar map data, the orientation distribution function ODF maps were calculated by TexTools texture software.

El QUANTA 650 FEG field emission scanning electron microscope equipped with electron backscatter diffraction (EBSD) technology was used to detect and collect the micro-orientation characteristics of annealed samples. The scanned .cpr data was analyzed and processed by HKL Channel 5 software to obtain ODF maps.

3. Result and discussion

3.1. Magnetic property test of experimental steel

MATS-3000M silicon steel measuring instrument was used to measure iron loss $P_{1.0/400}$ and magnetic induction intensity $B_{50}$. The magnetic properties of five samples for each temperature were measured, and the average values are shown in table 2. With the increase of annealing temperature, the average iron loss value gradually decreases, and the magnetic property is best at 1030°C for 3 min, with the average core loss $P_{1.0/400}$ being 14.24 W kg⁻¹ and $B_{50}$ being 1.781 T. The hysteresis line is shown in figure 1.

The microstructure and texture of the experimental materials at 1030°C for 3 min would be further studied.

The samples were conducted along the rolling direction, with a size of 300 mm (RD) × 30 mm (TD). The samples in single sheet measurement will have higher magnetic properties than in square circle detection. Based on the 10% difference of iron loss value in the longitudinal and transverse directions, the total iron loss $P_{1.0/400}$ of the sample at 1030°C for 3 min should be 15 ~ 16 W kg⁻¹.

3.2. Study on microstructure

Figures 2(a) and (b) show the microstructures of hot rolled and normalized plates, respectively. It is found from figure 2(a) that there is obvious microstructural heterogeneity in the thickness direction of the hot rolled plate. The surface layer of the hot rolled plate is equiaxed grain microstructure, the central layer is banded fiber microstructure along the rolling direction, and the transition layer is the mixed structure of recrystallization and banded fiber microstructure. According to statistics, the proportion of fine equiaxed crystals in the surface layer...
Figure 1. Hysteresis loop of annealed sheets at 1030 °C for 3 min.

Figure 2. Microstructure (a) hot rolled plate; (b) normalized plate; (c) cold rolled sheet; (d) annealed sheet.
is 26.38%, and the grain size is 10 μm ~ 20 μm. According to figure 2(b), the normalized plate is completely recrystallized with a grain size of 30 μm ~ 200 μm and an average grain size of 76.82 μm.

The difference in microstructure of hot rolled plate along the thickness is related to the temperature gradient and stress during hot rolling. HongBo Pan [19] et al concluded that the inhomogeneities of microstructure and texture are the microcosmic reflection of stress and strain; in the way of measuring the rolling pressure with and without lubrication during finishing rolling, the results showed that the specimen surface underwent intense shear deformation during non-lubricated rolling, and the thickness of ferrite grains in the center was larger, while that in the surface is very thin, which presents gradient relationship. The hot rolled plate in this experiment was also non-lubricated during hot rolling, and the surface layer underwent great shear force, as can be seen from figure 2(a), the ferrite grain thickness in the surface layer was thin, and the center layer was obviously thicker. Large amounts of friction heat is generated by direct contact between the hot rolled plate and roller during hot rolling, some of which is absorbed by the surface layer, and the shear force and high temperature provide recrystallization conditions for the surface deformation microstructure, which makes the surface layer of the hot-rolled plate form the fine equiaxed crystals. With the deepening of the thickness, the shear force gradually decreases, and the rolling deformation force gradually increases. In addition, the surface layer absorbs friction heat which causes the temperature gradient along the thickness direction, and the high Si content decreases thermal conductivity. These factors cause the transition layer to form the mixed texture of recrystallization and deformation. The shear force of the central layer is basically zero, and the compression deformation force reaches the maximum, with the lowest temperature in the central layer because of the temperature gradient, which prevents the deformation microstructure of the central layer from recrystallizing and only forms the deformed banded microstructure parallel to the rolling direction [20, 21]. These factors above finally lead to the microstructure inhomogeneity in the thickness direction of the hot rolled plate.

Microstructural heterogeneity in the thickness direction of the hot rolled plate disappears in the normalizing process, and its microstructure has fully recrystallized and achieves homogenization along the thickness direction. When studying the effect of grain size of hot rolled plates after normalization on the microstructure, texture and magnetic properties of finished products, Zeng et al [22] pointed out that the increase in grain size of the normalizing plate is beneficial to increasing the grain size of the resulting sheet and reduce the iron loss. At the same time, the increase in grain size of the normalizing plate is also beneficial to increasing the proportion of Goss texture in the resulting sheet and reducing the proportion of γ-fiber texture to enhance the magnetic induction. However, too large grain size and poor microstructure uniformity would make the cold rolling process difficult, resulting in cracks and even broken strips in the cold rolled sheet. Mao [23] et al also reached similar conclusions in the study about the effect of grain size of hot rolled plates after normalization on the texture and magnetic properties of non-oriented silicon steel. In this study, the average grain size of the normalizing plate is 76.82 μm, and the microstructure uniformity is better, which achieve normalization purposes and are beneficial to improving the magnetic properties of the finished product.

Figure 2(c) shows the microstructure of cold rolled sheet. Normalized plate formed the typical fiber structure after cold rolling. Cold rolled fiber structures are mainly divided into three categories: shear bands (area A) with 30° to rolling direction; the black area (area B) with deep etching is mainly {111} γ-fiber, and the white area (area C) with shallow etching is mainly α-fiber [24]. The plastic deformation of normalizing plate is mainly carried out by slip during cold rolling. Slip occurs first on the easy slip surface inside the grains. At the same time, with the increase of passes for cold rolling and defects inside the grains, the dislocation density also increases. The continuous change of Taylor factor during cold rolling causes an increase in the number of sliding systems in slip state, and the plastic deformation changes from local deformation to uniform deformation of the total thickness. With the increase of cold rolling reduction rate, the dislocation movement also moves from the grain interior to the grain boundaries. The blocked dislocation will form dislocation pileup groups at the grain boundaries and eventually burst out of the grain boundaries, destroying the crystal structure of the grains and forming substructural grains, presenting the whole sheet thick fibrous texture [25, 26].

After annealing at 1030 °C for 3 min, completely recrystallization occurred in the cold rolled sheet. Some grain sizes had reached the thickness of the sheet. The average grain size of the annealed sheet was 147.11 μm, as shown in figure 2(d). In this study, after annealing, the complete recrystallized coarse grains not only eliminated the complex internal stress state which was not conducive to magnetization in the cold rolled sheet but also reduced the magnetization resistance and iron loss of the resulting sheet by reducing the grain boundaries and dislocation tangles.

3.3. Study on texture

Figure 3 shows the macro-textures in the surface, 1/4 and 1/2 layer of the hot rolled plate. It can be seen from figure 3(a) that the surface textures of hot rolled plate are mainly {112} <111> copper texture, {110} <223> partial brass texture, and the weak {110} <001> Goss texture. {112} <111> texture intensity is 2.4,
and the texture content is 7.75%. \{110\}<223> texture intensity is 2.0, and texture content is 7.78%. It can be seen from figure 3(b) that the texture in 1/4 layer of hot rolled plate is mainly \{110\}<001> Goss texture, with the texture intensity of 7.6 and the texture content of 9.04%. It can be seen from figure 3(c) that the texture components in 1/2 layer of the hot rolled plate are mainly Goss texture, \{114\}<110> α texture and \{100\}<001> Cube texture, and the strongest texture component is \{110\}<001> texture, with texture intensity of 5.3 and the texture content of 5.33%.

According to the literature [27], when the hot rolled reduction rate was low, large Goss oriented deformed grains were firstly formed by the shear force, and then Brass oriented deformed grains were gradually formed within Goss, while Copper-type texture was formed under larger shear force. The relationship between Copper orientation and Goss orientation was rotating around TD, which was the result of large shear force, while the relationship between Brass orientation and Goss orientation was rotating around ND, indicating that if the rotation around TD was blocked, it would likely to rotate around ND. The relationship between the three shear textures was that the Goss texture was formed first, and the Copper texture was formed last. The Copper orientation has little effect on the formation of Goss texture as it is easily transferred to the rotating cube \{100\}<011> orientation during cold rolling, and the grain size is not coarse. \{110\}<227> ∼ \{110\}<223> grains that are close to Brass texture are not conducive to the abnormal growth of Goss grains as their mobility is higher than that of Goss grain boundaries at low temperature. In this experiment, the Goss oriented grains formed by dynamic recrystallization in the surface layer of the hot rolled plate rotated around the normal (ND) and transverse (TD) directions and transformed into Copper and partial Brass textures, respectively, as shown in figure 3(a), it is proved that the surface shear stress is the largest. Due to the surface layer absorbing the friction heat, the temperature gradient was formed between the surface and the 1/4 layer, the shear force decreased with the depth of the plate thickness, and more Goss texture was retained, as shown in figure 3(b) that Goss texture was the strongest at 1/4 thickness; at the center of thickness, there was the maximum compression deformation force due to the little shear stress, the original α-fiber texture in the casting billet was transferred to α-fiber texture along the \{001\}<100> ∼ \{001\}<110> ∼ \{112\}<110> ∼ \{223\}<110> path under the compression condition, and part of Goss texture was transferred to α-fiber texture along the \{110\}<001> ∼ \{554\}<225> ∼ \{111\}<112> ∼ \{111\}<110> ∼ \{223\}<110> path [28, 29], so α-fiber texture appeared in the center, as shown in figure 3(c).

Therefore, the texture gradient of the hot rolled plate from the surface layer to the center layer was: strong (Copper texture + Brass texture) + weak Goss texture → strong Goss texture → strong Goss texture + weak α-fiber texture.

The normalized plate undergoes complete recrystallization and has better microstructure uniformity, and texture analysis of 1/4 layer and 1/2 layer is shown in figure 4. It can be seen from figure 4(a) that the texture components in 1/4 layer of the normalized plate are mainly \{114\}<221> texture and \{114\}<371> texture, which are concentrated near the α'-fiber. The intensity of \{114\}<221> texture is 5.5, and the texture content is 15.98%. \{114\}<371> texture intensity was 5.5, and texture content was 18.79%; the texture component of the 1/2 layer of the normalized plate is similar to that of the 1/4 layer, as shown in figure 4(b) that the texture in 1/2 layer is mainly on and near the α'-fiber line, and the strongest texture component is \{114\}<481> texture with texture intensity of 7 and the texture content of 18.32%. These indicate that the texture type of 1/4 layer is basically the same as that of 1/2 layer, and the microstructure and texture of the normalizing plate are more uniform along the thickness direction than that of the hot rolled plate.

According to reference [30], the typical \{114\}<481> texture in the α'-fiber texture could nucleate at the grain boundaries of Goss grains, so the α'-fiber texture with \{114\}<221> and \{114\}<371> (similar \{114\}<481>) as the core texture is finally formed. Qin [17] et al. studied strong \{100\}<012> ∼ \{114\}
Figure 4. $\varphi_2 = 45^\circ$ ODF sections of the normalized sheet (a) 1/4 layer; (b) 1/2 layer.

Figure 5. $\varphi_2 = 45^\circ$ ODF sections (a) the cold rolled sheet; (b) the annealed sheet (Graphical Abstract).

$<481>$ recrystallization texture in hot rolled plate under high reduction rate and found that the intensity of $\{100\}<012> \sim \{114\}<481>$ $\alpha$-fiber texture in normalizing plate increased, and the content of texture component increased from 20.4% to 29.8% when normalizing at 1000 °C, with the increase of normalization time from 1 min to 5 min Goss texture in each layer of the hot rolled plate provided nucleation sites for $\alpha$-fiber texture, and $\alpha$ texture in the center of the hot rolled plate also promoted the formation of $\alpha$-fiber texture.

Figures 5(a) and (b) show the macroscopic textures of cold rolled and annealed sheets of non-oriented silicon steel. The cold rolled texture is mainly $\alpha$-fiber texture with the strongest $\{114\}<110>$ texture, and the texture content is 19.77%. The texture of annealed plate is mainly $\alpha$-fiber texture with $\{114\}<481>$ and $\{113\}<361>$ as the core textures. The intensity of $\{114\}<481>$ texture is 4.3, and the texture content is 20.43%.

The texture characteristics of steel materials during the cold rolling show that the $\{112\}<111>$ slip system plays an important role when the cold rolling reduction rate exceeds 60% and after rolling the stable orientations were $\{223\}<110>$, $\{111\}<110>$, $\{112\}<110>$ and so on [31–33], which lead to the increase of $\{112\}<110>$ and $\{001\}<110>$ components after cold rolling. During the cold rolling process, the rolled texture basically rotated along the following two ways: $\{001\}<100>$ $\rightarrow \{001\}<110>$ $\rightarrow \{112\}<110>$ $\rightarrow \{223\}<110>$ $\rightarrow \{114\}<481>$ $\rightarrow \{114\}<371>$ $\rightarrow \{31\}<112>$ $\rightarrow \{111\}<112>$ $\rightarrow \{111\}<110>$ $\rightarrow \{223\}<110>$, so the textures of cold rolled sheet presented strong $\alpha$-fiber texture and weak $\gamma$-fiber texture [34]. This experiment differed from the above rotation ways, and the texture of the cold rolled sheet was concentrated in $\{114\}<110>$. The rotation process of cube texture during cold rolling was described in document [35], which was divided into three stages: first to $\{001\}<230>$, then to $\{115\}<120>$, and finally stabilized at $\{114\}<110>$. In this experiment, the 1/4 and 1/2 layers of the normalized plate are both dominated by $\alpha$-fiber texture with $\{114\}<481>$ texture as the core. $\alpha$-fiber texture directly rotates to the $\alpha$-fiber texture during the cold rolling, resulting in the strongest $\{114\}<110>$ texture, rather than the typical cold rolling texture $\{223\}<110>$ of traditional non-oriented silicon steel.
3.4. Texture evolution of \{114\} <481>

Different from the annealing texture of ordinary non-oriented silicon steel, the typical texture \{114\} <481> appears in ultra-thin non-oriented silicon steel, but there is no unified conclusion on the nucleation and growth of \{114\} <481> texture at present. The reasons why they are different are the large cold rolling reduction rate of ultra-thin non-oriented silicon steel and a small proportion of the plane compressive stress, and it is mainly affected by the strong shear stress, which leads to the grains rotating around the TD and ND axes and the start of the new sliding system at the same time, resulting in the complication of grain orientation rotation. In this study, the nucleation and growth process of \{114\} <481> texture were studied by the "sampling method" at 1 s, 2 s, 10 s, and 180 s during annealing process.

Figure 6 shows the microstructure, ODF diagram, and main grain orientation of annealed sheet at 1 s. It can be seen from figure 6(a) that there are still non-recrystallization banded structures in the annealed microstructure. Figure 6(b) shows that the texture type of annealed sheet is similar to that of cold rolled sheet. The texture type of annealed sheet are mainly \(\alpha\)-fiber texture, including \{001\} <110> ~ \{111\} <110>, concentrated in \{114\} <110>, and some \(\gamma\)-fiber texture \{111\} <112>. It can be seen from figure 6(c) that the \{114\} <481> texture has formed at the early stage of annealing. Part of \{114\} <481> are banding distributed along the grain boundaries of \{114\} <110> and \{111\} <110>, and some are polygon shaped at the \{114\} <110>, \{111\} <110> and \{111\} <112> grain boundaries. In addition, it is obviously found from figure 6(c) (white rectangular area) that the \{114\} <481> oriented grains nucleate in the \{114\} <110> deformation band. So that the \{114\} <481> oriented nucleus nucleate at the grain boundaries of \(\alpha\)-fiber texture and \{111\} <112> texture, and in the deformation bands of \{114\} <110>. However, \{114\} <110> is the main texture in the cold rolled sheet of ultra-thin non-oriented silicon steel, and \{111\} <112> texture is fewer. Therefore, \{114\} <481> texture mainly nucleates at the grain boundaries and in the deformation bands of \(\alpha\)-fiber texture. At the beginning of annealing, \{114\} <481> oriented grains do not have size advantage, with the diffuse texture and weak strength.

Figure 7 shows the main orientation of annealed sheet at 2 s. The microstructure has completely recrystallized at the 2 s, with the texture dominated by strong \{111\} <112> and weak \{114\} <481>. The average grain size of \{111\} <112> orientation is 25.18 \(\mu\)m, the texture strength is 13.2, and the content is 35.4%. The intensity of \{114\} <481> texture is weak, and the number of grains is small, with an average grain size of 15.23 \(\mu\)m. In addition, it is evident from figure 7(a) in the white rectangular area that \{111\} <112> oriented grains nucleate and grow within the \{111\} <110> deformed bands.

Figure 8 is the main orientation and ODF diagram of completely recrystallized annealing sheet at 10 s. The texture type is \{114\} <481> and \{111\} <112> coexist, but the intensity of \{114\} <481> texture is 11, which is significantly higher than that of \{111\} <112> texture. The \{111\} <112> and \{114\} <481> oriented grains both experienced a continuous growth process when the annealing time continued to increase to 10 s. The average grain size of \{111\} <112> oriented grains was 37.84 \(\mu\)m, and that of \{114\} <481> oriented grains was 54.56 \(\mu\)m. At this stage, \{114\} <481> oriented grains had size and texture strength advantages.

Figure 9 is the main orientation and ODF diagram of the fully recrystallized annealing plate at 180 s. \{114\} <481> is the main texture, and \{111\} <112> texture intensity is further weakened. \{114\} <481> texture content reached 49.4%, and the average grain size is 234 \(\mu\)m. The average grain size of \{111\} <112> oriented grains is 136.8 \(\mu\)m. At this time, \{114\} <481> oriented grains have absolute advantages of size, strength, and quantity compared to \{111\} <112>.

Figure 10 shows the trend chart of grain size during annealing. With annealing time increasing from 2 s to 180 s, the average grain size of all grains increases from 11.53 \(\mu\)m to 118.68 \(\mu\)m, the average grain size of \{111\}
<112> orientation increases from 25.18 μm to 136.8 μm, and the average grain size of {114} <481> orientation increases from 15.23 μm to 234 μm. During the 1 s ~ 3 s annealing stage, {111} <112> grains preferentially nucleate and grow, resulting in {111} <112> with larger grain size in the early annealing stage and being the main texture. During the 3 s ~ 10 s annealing stage, the growth rate and the average size of {114} <481> oriented grains both exceeded that of {111} <112> oriented grains, but the
number of $\{114\} <481>$ oriented grains is small and the content is only 20.5% at this time, which is much lower than that of $\{111\} <112>$ with 35.4%. During the $10 \text{s} \sim 180 \text{s}$, the growth rate of $\{114\} <481>$ oriented grains is significantly higher than that of other orientation grains, so $\{114\} <481>$ oriented grains have obvious size advantages and the content reaches 49.4%, while the growth of $\{111\} <112>$ oriented grains was inhibited, and the content is reduced to 10.4%.

It indicates that $\{111\} <112>$ oriented grains grew mainly and dominated at the initial recrystallization stage. With the completion of recrystallization and the extension of holding time, the texture changed from strong $\{111\} <112>$ to strong $\{114\} <481> \sim \{113\} <361>$ texture, and the $\{114\} <481>$ oriented grains grew rapidly and became the main texture.

It has been found at domestic and foreign that the $\alpha^*$-fiber texture centered around $\{114\} <481>$ texture would be formed in the annealed sheets of ultra-low carbon steel, IF steel, and other steels which were directly annealed after cold rolling with high reduction rate (>85%). Among them, a small amount of $\{114\} <481>$ texture components will be generated at the grain boundaries of cold rolled deformed $\alpha$-fiber orientation grains due to sufficient irregular deformation. $\alpha^*$-fiber orientation grains would use $\{114\} <481>$ as the nucleation sites for recrystallization and growth during subsequent annealing [5, 13, 36]. Wang [15] et al studied the annealing $\{114\} <481>$ texture of 6.5% Si non-oriented silicon steel and found that the $\{114\} <481>$ texture was not only nucleated in the deformed $\{001\} <110>$ grains but also the grain boundaries of the deformed $\alpha$-fiber grains could be used as the nucleation sites for the $\{114\} <481>$ texture. In addition, many scholars have observed that the $\{114\} <481>$ sub-structure that formed within $\{112\} <110>$, $\{111\} <112>$ and other oriented grains and formed by the ‘orientation splitting’ phenomenon during cold rolling can also be used as the nucleation sites for $\alpha^*$-oriented grains [17, 37, 38]. Toge [39] found that $\{100\} <011>$ grains would form a deformation band at 35° to the rolling direction during the warm rolling, where $\{114\} <481>$ oriented grains would nucleate during annealing, and $\{114\} <481>$ oriented grains could grow rapidly because of the favorable orientation relationship between $\{114\} <481>$ oriented grains and nucleation matrix. Studies have shown that $\{111\} <112> \sim <110>$-oriented grains provide abundant nucleation sites for $\{114\} <481>$ oriented grains, and that $\{114\} <481>$ oriented nucleus have the large misorientation angle with deformed matrix and grow up by rapid migration of large angle grain boundaries, eventually forming $\{114\} <481>$ recrystallization texture [17, 38].

Document [5, 40] pointed out that when the reduction rate of cold rolling is more than 85%, the $\alpha$-deformed grains will undergo orientation splitting, forming such as $\{001\} <210>$, $\{114\} <481> \sim \{113\} <361>$ and other deformation matrix which are beneficial to the nucleation of $\lambda$ grains. With orientation splitting, there was a large deformation energy storage, which also created conditions for the nucleation of deformation matrices with low energy storage such as $\{001\} <210>$ and $\{114\} <481>$. Beck [41] and He [30] pointed out that there were various oriented nucleus in the deformed matrix, but only those nucleus with large angle grain boundaries between the deformed matrix could grow rapidly. According to the directional growth theory, Verbeken [42] et al studied the recrystallization behavior of $\alpha$-fiber microstructure in Body-centered cubic metal rolled by large deformation rolling, so the result shows that the relationship between $\{113\} <471>$ and $\{112\} <110>$ is 19.7° <110>, which was beneficial to rapid growth.
It can be seen from figure 6 that the sites of \(\{114\}<481>\) orientation nucleus in this experiment can be divided into two types. The first type of nucleation is located at the grain boundaries and the internal deformed bands of \(\{114\} \sim \{112\}<110>\) oriented grains, and the second type is at the grain boundaries of \(\{111\}<112>\) orientation grains. The experimental results are basically consistent with the previous discussion; for these two types of nucleation sites, the \(\alpha\) deformed matrix, \(\{111\}<112>\) and other oriented grains form the \(\{114\}<481>\) substructure because of the 'orientation splitting' phenomenon during the too large cold rolling reduction, and the \(\{114\}<481>\) substructure was used as the nucleation sites for \(\alpha\) oriented grains. In this study, with the fast heating rate, the temperature quickly over the grain nucleation inoculation temperature and transited to the grain nucleation growth temperature [43], and the grain energy storage was weakened, and the texture of the cold rolled sheet played a genetic role. Because the cold rolled sheet was dominated by \(\alpha\) texture, \(\{114\}<481>\) mainly nucleated at grain boundaries and internal bands of \(\{114\} \sim \{112\}<110>\) oriented grains. When nucleation starts, the grain size grows further with the increase of annealing time, but until the recrystallization was completed, the \(\{114\}<481>\) oriented crystals did not have size, number, and strength advantages, and the \(\{111\}<uvw>\) oriented crystals always dominated. After recrystallization, the \(\{114\}<481>\) oriented crystals begin to have the advantage of growth rate, and their grain size and texture strength gradually exceeded that of \(\{111\}<uvw>\) oriented crystals. Therefore, \(\{111\}<uvw>\) oriented grains dominate in the early stage of recrystallization, and \(\{114\}<481>\) oriented grains dominate in the later stage of grain growth.

Gobenardo [38] also observed similar phenomena in the study of the origin of \(h,1,1<21/1,h,1,2\) fiber texture in single-phase ferrite steel. The analysis showed that the order of energy storage in cold-rolled deformed grains from large to small was \(\{110\} \sim \{111\} \sim \{112\} \sim \{100\}\). In the first stage, due to the transformation of \(\{110\}\) oriented grains in the sliding system during cold rolling, the content of \(\{110\}\) oriented deformed microstructure in the cold rolled sheet was very small, so \(\{111\}<112>\) oriented grains would nucleate and grow at the grain boundaries and deformation bands of \(\{111\}<110>\) deformed grains at the initial stage of recrystallization [44], as shown in figure 7(a).

Generally, the order of energy storage of different orientations deformed grains from large to small is \([44]\):\(E_{\{110\}<110>},E_{\{111\}<uvw>},E_{\{112\}<110>},E_{\{100\}<110>},E_{\{110\}<001>},E_{\{116\}},E_{\{117\}}\) are both low energy storage oriented grains [45], and \(\{114\}\) was close to \(\{116\}\), and the cold rolled deformed microstructure of \(\{111\}\) plane which have more energy storage than that of \(\{114\}\) plane preferentially formed \(\{111\}<112>\) oriented grains. In addition, large angle grain boundaries that are easy to migrate form between them because of the large misorientation angle between the nucleation regions of \(\{111\}<112>\) and \(\{111\}<110>\). The content of \(\{111\}<110>\) texture in the cold rolled sheet was high, so the nucleation and growth of \(\{111\}<112>\) oriented grains were mainly manifested in the early recrystallization. At this time, the number and size of \(\{111\}<112>\) oriented grains are predominant compared to those of \(\{114\}<481>\) oriented grains, as shown in figures 6 and 7.

According to the directional growth theory [18], there are various recrystallized orientation nucleus in the deformed matrix, but only those nucleus with large angle grain boundaries between themselves and the deformed matrix can grow rapidly. In the middle stage of recrystallization, there is a 27°<110> special orientation relationship between the nucleated \(\{114\}<481>\) grains and the \(\{001\}<110>\) oriented substrates, and \(\{114\}<481>\) grains have the rapid growth advantage, which lead to the aggregation of other orientation textures to \(\{114\}<481>\). Since the energy storage of \(\{111\}<112>\) and \(\{111\}<110>\) deformed grains in the cold-rolled sample is higher than that of other deformed grains, and the new \(\{111\}<112>\) grains will nucleate in the \(\{111\}<110>\) deformed matrix during recrystallization annealing: The aggregation of \(\{111\}<112>\) oriented nucleus that nucleated in the \(\{111\}<110>\) deformed matrix will aggregate due to the orientation pinning effect, which seriously hinders the growth of \(\{111\}<112>\) nucleus but meets the selective growth of \(\{114\}<481>\) oriented grains; The relationship between \(\{114\}<481>\) oriented nucleus and \(\{111\}<112>\) is 37.6°<−5,2,1>, and the \(\{114\}<481>\) oriented nucleus grow up by the rapid migration of large angle grain boundaries and swallowing \(\{111\}<uvw>\) oriented grains [15, 42], so \(\{114\}<481>\) oriented grains can grow further. At this time, the grain size of \(\{114\}<481>\) oriented grains is obviously larger than that of \(\{111\}<112>\) grains, but there are still a certain number of \(\{111\}<112>\) oriented grains, as shown in figure 8.

In the later recrystallization stage, the grains continue to grow with the increase of annealing temperature and time, and the grain boundary mobility of the contact \(\{111\}<112>\) oriented recrystallized grains slows down due to the \(\sum 3\) grain boundary, which inhibits the further growth of each other [46], which is also known as the 'orientation pinning' phenomenon. Therefore, at this stage, \(\{114\}<481>\) oriented grains grew selectively and had absolute size, texture strength and content advantages, and finally, \(\{114\}<481>\) became the main annealing texture [14, 47–50], as shown in figure 9.

The texture evolution path of ultra-thin non-oriented silicon steel from cold rolling to the initial stage of recrystallization, the completion stage of recrystallization, the grain growth stage and the end of annealing is:
The highlights of this study lie in:

4. Conclusion

The highlights of this study lie in: (1) The texture evolution characteristics of non-oriented silicon steel for new energy vehicles were clarified; (2) The nucleation sites of α'-oriented grains were confirmed; (3) The competition growth mechanism of α' and {111} <112> texture was revealed. The specific conclusions are as follows:

(1) The hot rolled plate shows typical shear texture characteristics, with the Copper and partial Brass texture in the surface layer, Goss texture in 1/4 layer, and Goss texture and α-fiber texture in center layer. Complete recrystallization occurs in the normalizing plate, with α'-fiber textures around strong {114} <814> in 1/4 layer and the 1/2 layer. The cold rolled plate is mainly α-fiber texture with the strength concentrated in {114} <110>. The annealing plate formed α'-fiber texture with {114} <481> as the core.

(2) The {114} <481> oriented grains mainly nucleate inside the deformed {114} <110> grains and at the grain boundaries of the deformed α-fiber grains, without the advantages of size, quantity and strength.

(3) At the initial stage of recrystallization, {111} <112> oriented grains rapidly grow into the main texture by swallowing α-oriented grains with the advantages of energy storage and misorientation angle with {111} <110>. At the later stage of recrystallization, {114} <481> oriented grains grow up and become the core texture by the special relationship of 37.6°<3,2,1> with {111} <112> and rapidly swallowing {111} <112> oriented grains.

Figure 11. Formation mechanism of {114} <481>.
Data availability statement

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

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