Microstructure Characteristics and Strengthening Behavior of 
Cu-Bearing Non-Oriented Silicon Steel: Conventional Process 
versus Strip Casting

Feng Fang 1, Diwen Hou 1, Zhilei Wang 1, Shangfeng Che 1, Yuanxiang Zhang 1, Yang Wang 1*, Guo Yuan 1,*, 
Xiaoming Zhang 1, Raja Devesh Kumar Misra 2 and Guodong Wang 1

1 State Key Laboratory of Rolling and Automation, Northeastern University, Shenyang 110819, China; 
fangfengdbdx@163.com (F.F.); houdeeven@163.com (D.H.); wangz19956@126.com (Z.W.); 
csf2541787@163.com (S.C.); redyuanxiang@126.com (Y.Z.); neuwanyang@126.com (Y.W.); 
zhangxm@ral.neu.edu.cn (X.Z.); wanggd@mail.neu.edu.cn (G.W.)
2 Department of Metallurgical, Materials and Biomedical Engineering, University of Texas, El Paso, TX 79968, USA; dmisra2@utep.edu
* Correspondence: yuanguoral@sina.com

Abstract: Based on conventional hot rolling processes and strip casting processes, Cu precipitation 
strengthening is used to improve the strength of non-oriented silicon steel in order to meet the 
requirements of high-speed driving motors of electric vehicles. Microstructure evolution was studied, 
and the effects of Cu precipitates on magnetic and mechanical properties are discussed. Compared 
with conventional processes, non-oriented silicon steel prepared by strip casting exhibited advantages 
with regard to microstructure optimization with coarse grain and {100} texture. Two-stage rolling 
processes were more beneficial for uniform microstructure, coarse grains and improved texture. The 
high magnetic induction B_50 of 1.762 T and low core losses with P_{1.5/50}, P_{1.0/400} and P_{1.0/1000} of 1.93, 
11.63 and 44.87 W/kg, respectively, were obtained in 0.20 mm sheets in strip casting. Cu precipitates 
significantly improved yield strength over ~120 MPa without deteriorating magnetic properties both 
in conventional process and strip casting. In the peak stage aged at 550 °C for 120 min, Cu precipitates 
retained bcc structure and were coherent with the matrix, and the yield strength of the 0.20 mm 
sheet was as high as 501 MPa in strip casting. The main mechanism of precipitation strengthening 
was attributed to coherency strengthening and modulus strengthening. The results indicated that 
balanced magnetic and mechanical properties can be achieved in thin-gauge non-oriented silicon 
steel with Cu addition in strip casting.

Keywords: non-oriented silicon steel; Cu precipitation; magnetic properties; mechanical properties; strip casting

1. Introduction

Non-oriented silicon steel is mainly used to manufacture the core component of a drive 
motor. Its performance determines the drive characteristics and service performance of the 
drive motor [1]. With the rapid development of drones, centrifugal compressor and electric 
vehicles, the applications of high-speed motors with rotor speed up to tens of thousands or 
even hundreds of thousands of revolutions per minute are gradually increasing [2]. The 
centrifugal force borne by the rotor of driving motor increases in proportion to the square 
of speed and rotor radius. In addition, the complex driving conditions of electric vehicles 
have high requirements for the driving motor. The rotor cores with complex shapes of 
high-speed driving motors not only requires good magnetic properties to improve energy 
conversion efficiency but also excellent mechanical properties and fatigue properties to 
resist deformation and failure [3–5]. As a result, non-oriented silicon steel with high 
strength has become an important development trend in recent years.
High strength non-oriented silicon steel is not only a functional material but also a structural material. It is generally believed that the methods for improving mechanical properties may often damage the magnetic properties of non-oriented silicon steel, and magnetic properties and mechanical properties are contradictory [6,7]. A key issue is to coordinate the contradiction between mechanical properties and magnetic properties of non-oriented silicon steel. Currently, it is difficult to produce thin-gauge non-oriented silicon steel with higher magnetic induction, lower core loss and higher strength in conventional continuous casting–hot rolling processes, where the texture control level almost reaches its limit. Fortunately, strip casting technology with short processes and that save energy provides a novel method for the preparation of non-oriented silicon steel in recent research [8,9]. It has been recognized that the initial favorable (100) columnar structure in as-cast strips exhibited strong heredity to increase the intensity of favorable (100) texture and reduces the intensity of harmful \( \gamma \) texture \(<111>///\)normal direction\) in the final sheet, which could significantly improve the magnetic properties of non-oriented silicon steel [10]. Moreover, the characteristics of the near-net shape of strip casting is beneficial for processing ultra-thin non-oriented silicon steel [11,12].

In terms of mechanical properties of non-oriented silicon steel, there are mainly three strengthening mechanisms, which involves dislocation strengthening, solid solution strengthening and precipitation strengthening. In the case of dislocation strengthening, dislocation annihilation was inhibited by incomplete recrystallization annealing in order to improve the mechanical properties of non-oriented silicon steel, but its deterioration effect on magnetic properties was significant [13,14]. Solid solution strengthening of non-oriented silicon steel was mainly realized by the addition of solid solution elements, such as Si, Al, Mn, P and Ni, to cause lattice distortion in order to improve matrix strength [7,15–17]. The relevant deterioration of magnetic properties is at tolerable levels, but it generally requires additional alloying elements, which not only increases production costs but also brings some difficulties in the rolling and stamping process. Introducing Cu into non-oriented silicon steel for precipitation strengthening to improve mechanical properties is a potential method for non-oriented silicon steel [18–21]. Due to the low solubility of Cu in ferrite and low lattice misfit between Cu precipitates and bcc matrix, the high densities of nano-size Cu precipitates form after thermal-aging processes and significantly enhance mechanical properties. Meanwhile, Bian [19] and Wu [20] reported that Cu-rich precipitates could improve recrystallization texture by inhibiting \( \gamma \) texture and promoting Goss texture \(<110>//<100>\), resulting in the decrease in core loss without deteriorating magnetic induction. Moreover, it is suggested that Cu precipitates significantly contributed to the yield strength of a 0.35 mm thick non-oriented silicon steel without deteriorating the magnetic properties in the strip casting process [21]. However, studies on Cu precipitates in thin-gauge non-oriented silicon steel are limited. Cu precipitation behaviors in non-oriented silicon steel and its effect on magnetic properties and mechanical properties in different processes are not clear. Therefore, it is worthwhile to take full technical advantage of strip casting and the strengthening effect of Cu precipitates to prepare ultra-thin non-oriented silicon steel with good magnetic and mechanical properties.

In this study, 0.20 and 0.35 mm thick non-oriented silicon steels with Cu addition were processed by using conventional processes and strip casting processes. The evolutions of microstructure and texture were comparatively studied. Furthermore, Cu precipitation behavior and its effects on magnetic and mechanical properties are discussed in detail.

2. Materials and Methods

The composition of the experimental steel is shown in Table 1. The hot rolled sheet and as-cast strip with near-identical thickness were prepared by conventional hot rolling processes (denoted as conventional process) and strip casting processes. After normalization, 0.35 and 0.20 mm non-oriented silicon steels were subjected to single-stage and two-stage cold rolling, respectively, followed by annealing and aging treatments. A schematic diagram of the detailed processes is shown in Figure 1.
In the strip casting process, the as-cast strip that was 2.3 mm thick was directly produced by a vertical twin-roll strip caster (State Key Laboratory of Rolling and Automation, Shenyang, China) at the speed of 30~40 mm/min and then air cooled to room temperature. In the conventional process, a 50 kg ingot of experimental steel was cast in vacuum induction and then forged. Before hot rolling, the billet was homogenized at 1150 °C for 3 h. Then, it was hot rolled to 2.3 mm via 7 passes with a finishing temperature of 924 °C; then, it was air cooled to room temperature. After obtaining a hot band and as-cast strip, the subsequent process was kept the same. Subsequently, both as-cast strips and hot rolled sheets were normalized at 1050 °C for 5 min and then pickled. Next, single-stage and two-stage cold rolling processes were adopted to process 0.35 mm and 0.20 mm sheets, respectively. In terms of single-stage cold rolling, total rolling reduction was about 85%. A two-stage process was carried out, involving a first stage of cold rolling with a reduction of 70%, intermediate annealing at 1050 °C for 5 min, and a second stage of cold rolling with a reduction of 71%. All the cold rolled sheets were annealed at 1050 °C for 5 min for recrystallization and then aged at 550 °C for 25, 40, 60, 90, 120 and 240 min, respectively. For convenience, the processes to produce the 0.35 mm and 0.20 mm sheets in the conventional process were denoted as route A and route B, respectively. Similarly, it was named route C and route D in the strip casting process.

Metallographic microstructure was examined by a Leica optical microscope (Leica, Wetzlar, Germany) along the longitudinal section as defined by the rolling direction (RD) and normal direction (ND). The microstructure and micro-texture were characterized by electron backscattered diffraction (EBSD) (Oxford, Aztec, UK) attached to a field emission scanning electron microscope (SEM, Zeiss Ultra 55, Carl Zeiss, Jena, Germany) along the longitudinal section, and data analysis was post-processed with HKL-Channel 5 software. For macrotexture, three incomplete pole figures, {200}, {110} and {211}, were measured by using Bruker D8 Discover X-ray diffraction with CoKα radiation along the rolling plane in the subsurface layer. The ODFs (orientation distribution functions) were calculated based on series expansion methods. The precipitation observation was conducted in a TECNAI G2 F20 transmission electron microscope (TEM, FEI, Hillsboro, OR, USA) equipped with energy dispersive X-ray spectroscopy, and data were analyzed by Digital Micrograph. Furthermore, the magnetic properties involving magnetic induction of B 50 and core loss of P1.15/50, P1.0/50, P1.0/400 and P1.0/1000 were measured by a using single sheet tester with dimensions of 100 (RD) × 30 mm (transverse direction, TD). For mechanical properties, standard tensile tests were carried out by using an AG-X plus100KN (Shimadzu, Tokyo, Japan) testing machine at room temperature at a crosshead speed of 1 mm/min. The properties are an average of three measurements.
3. Results

3.1. Microstructure and Texture Evolution through Processing

Figure 2 shows the microstructure and microtexture of the hot rolled sheet (Figure 2a–c) and as-cast strip (Figure 2d–f). The surface layer and sub-surface layer of the hot rolled sheet were mainly fine (110) oriented recrystallized grains, and the central layer comprised mainly (100) and (111) elongated grains along the rolling direction, as observed in the inverse pole figure (IPF) map in Figure 2a. Grain size gradually increased from the surface layer to the center layer, and the average grain size was ~85.8 µm. The hot rolled sheet exhibited strong γ and λ textures (<100>//ND), in which γ texture concentrated on [111] <112> with an intensity of 3.43, and λ texture focused on the R-Cube component ([100] <110>) with an intensity of 2.45. Figure 2d shows microstructure and orientation distribution of the as-cast strip. Due to low superheat in the casting process, columnar grains in the surface layer were relatively small, and equiaxed grains were distributed in sub-surface layer and central layer with an average grain size of ~150.3 µm. Compared to the hot rolled sheet, the as-cast strip mainly exhibited {100} and {110} texture, which mainly generated deviated R-Cube components with an intensity of 3.80 and deviated Goss component with an intensity of 2.63.

Figure 2. Microstructure and micro-texture (ODF section at $\phi_2 = 45^\circ$) of (a–c) hot rolled sheet and (d–f) as-cast strip.

Figure 3 shows the microstructure of normalized sheets in the conventional process and strip casting. Observed from Figure 3a,d, the grain distribution was more uniform, and the grain size of the hot-rolled sheet and as-cast strip after normalization increased to ~164.3 and ~188.3 µm, respectively. Meanwhile, the {111} component was reduced, while the proportion of {100} and {110} components increased, which also showed a certain size advantage.

Figure 4 shows the deformation microstructure and macro-texture of 0.35 and 0.20 mm cold rolled sheets. After cold rolling to 85% reduction, the matrix suffered from severe plastic deformation, and elongated grains were distributed throughout the thickness of sheets, as shown in Figure 4a,b. Considering that the as-cast strip had a larger grain size than the hot-rolled sheet, the density of shear bands in route C was much higher compared to route A. Cold rolled sheets in both route A and route C exhibited strong α and γ textures, with strong {111} <110> with intensities of 11.91 and 11.14, respectively. The difference was that the λ texture was stronger, and γ texture was weaker in route C, as shown in Figure 4e,g. In terms of the 0.20 mm cold rolled sheets, large recrystallization grains were elongated along the rolling direction. Considering that the microstructure was refined after intermediate annealing, there were fewer shear bands in the 0.20 mm cold rolled sheets.
After two-stage cold rolling, γ texture dominated the deformation texture. In route B, the peak corresponded to [111] <110> component with an intensity of 12.02, while in route D, strong λ textures with a maximum around a deviated Cube ([100] <100>) of 5.33 and α* texture ([h11] <1/h12>) were also observed, as shown in Figure 4h.

Figure 3. Microstructure and micro-texture (ODF section at φ2 = 45°) of normalized sheets in (a–c) the conventional process and (d–f) strip casting.

Figure 4. Microstructure and macrotexture (ODF section at φ2 = 45°) of 0.35 mm and (c,d,f,h) 0.20 mm cold rolled sheets in (left) the conventional process and (right) strip casting.

Figure 5 shows the microstructure and texture of 0.35 and 0.20 mm annealed sheets. The grain size in route A and C was 53.9 and 75.0 μm, respectively. Strong γ texture with a maximum around [111] <121> of 15.5 was formed, while α (<110>//RD) and λ textures were almost not observed in route A in Figure 5a. In contrast, the intensity of [111] texture in route C was significantly weakened, and grain size of [100] grains was much coarser than compared route A. Moreover, the strong deviated R-Cube component with an intensity of 8.84 and medium α* texture were detected in route C, as shown in Figure 5g.

With respect to 0.20 mm annealed sheets, the average grain sizes of the annealed strip in route B and route D were 50.5 and 96.4 μm, respectively. Moreover, the recrystallization microstructure was much more uniform compared with the single stage rolling process. In route B, γ texture that was concentrated on [111] <110> still dominated the recrystallization texture with weak λ texture, and the intensity of [111] <110> component was 13.1, which was lower than in route A. In route D, γ texture was significantly reduced while Goss
grains exhibited obvious size and intensity advantages (as high as 12.4), as shown in Figure 5h. It can be observed that the two-stage cold rolling process was more beneficial for uniform microstructure, coarse grains and improved texture for non-oriented silicon steel. Considering that the aging temperature was far below recrystallization temperatures, it can be inferred that the microstructure did not vary obviously after aging treatments.

Figure 5. Microstructure and micro-texture (ODF section at \( \varphi_2 = 45^\circ \)) of (a,b,e,g) 0.35 mm and (c,d,f,h) 0.20 mm annealed sheets in (left) the conventional process and (right) strip casting.

3.2. Precipitation Behavior of Cu in Different Process

Figure 6 shows the morphology and size distribution of Cu precipitates in the hot rolled and as-cast strips. As shown in Figure 6a, the high density of fine Cu precipitates was formed after hot rolling and air cooling, with an average diameter of ~3.2 nm. In contrast, smaller precipitates with an average diameter of ~6.9 nm were non-uniformly distributed in the as-cast strip. The differences in density and size of precipitates may be related to the dislocation density of the matrix. Considering that the hot rolled sheet underwent cyclic dynamic deformation, higher densities of dislocation and grain boundaries were generated, which would lower the energy barrier and provide much more nucleation sites for Cu precipitates in conventional hot rolling processes.

Figure 6. TEM morphology and size distribution of Cu precipitates in (a,b) hot rolled sheet and (c,d) as-cast strip.
The morphology, size and crystal structure of Cu precipitation have very important impacts on strengthening mechanisms and strengthening effects. Therefore, it is necessary to clarify the precipitation behavior of Cu in the ferrite matrix when Cu precipitation strengthening was adopted in non-oriented silicon steel. Considering that the nose precipitation temperature of the Cu phase was far lower than the temperature of normalizing, intermediate annealing and final annealing, Cu precipitates tend to dissolve in the matrix during annealing processes and only a minority of Cu precipitated during the continuous cooling process. During aging process, Cu in solute solution state showed obvious tendencies to form nano-sized precipitates due to the extremely low solubility of Cu in ferrite. Moreover, the precipitation behavior during aging process was discussed in detail. Figure 7 shows the TEM results of precipitates in sheets aged at 550 °C for 25 min. It can be observed that some Cu precipitates with an average diameter of ~5.80 nm were formed in the under-aging stage. Fast Fourier transform (FFT) was conducted in a Cu precipitate, and it was identified to be bcc structure with a lattice constant of 2.879 Å. The precipitates of bccCu in the bcc matrix at the early stage of aging was widely confirmed by field ion microscopy, atom probe and high-angle annular dark-field (HADDF) TEM analysis [22–24]. Considering that Cu atoms were gradually enriched by replacing the Fe atom in the Cu-rich cluster, a slight lattice mismatch occurred, and the interface cannot be recognized for nano-sized Cu precipitates, as shown in Figure 7d. Moreover, the low coherent strain was beneficial in lowering the energy barrier for continuous precipitation of Cu with increases in aging time, resulting in nano-sized Cu precipitates at the early stage of aging.

Figure 7. (a) TEM morphology, (b) [111] bcc HR (high resolution)-TEM image, (c) FFT pattern and (d) IFFT image by masking (011) reflection of typical Cu precipitates after aging at 550 °C for 25 min.

Figure 8 represents the TEM results of precipitates in 120 min aged sheets. At this time, the coarsening of precipitates is observed, and the transformation of morphology from spherical to ellipsoid occurred in some precipitates, as shown by the arrow in Figure 8a. From the HR-TEM and FFT patterns in Figure 8b,c, it can be observed that the precipitates retained bcc structure, and weak B2 ordered spots appeared in the pattern taken from the precipitates. Moreover, the inverse FFT image from a 010 B2 diffraction spot showed obvious misfits between the precipitates and matrix, considering the difference in lattice constant. It can be inferred that with increases in aging time, Cu gradually partitioned to the precipitates, resulting in increases in inherent energy and coherent strain energy. When Cu precipitates reach a critical size of ~9.5 nm, the total free energy was too high to retain bcc structure, which resulted in structure transformation [23]. According to previous studies, it is well accepted that Cu precipitates experience bcc-9R-twinned fcc–fcc transformation, while there exists controversy over the transition state of 3R [23–25]. Given that the anisotropic growth of Cu precipitates is beneficial for reducing interfacial energy in the 9R transformation stage, it can be indicated that the precipitates in the peak aging stage corresponded to the critical transition state.
Precipitate morphology, precipitate size distribution and HRTEM after aging for 240 min are shown in Figure 9. It can be observed that the precipitation was obviously coarsened with relatively low density compared to the peak aging stage, and the average size was as high as ~15.2 nm during the over-aging stage. Due to the obvious coarsening of size, Cu precipitates experienced a transformation process, and high interface energy with the matrix was formed, which would also promote further rapid coarsening of precipitates in order to reduce interface energy. From the HR-TEM image, an obvious misfit was observed between Cu precipitates and the matrix. According to previous studies [23,24], it can be inferred that the crystal structure of Cu precipitate has transformed to 9R or even fcc structure.

The precipitation and growth behavior of Cu precipitates during aging is accompanied by a complex crystal structure transformation, and it is mainly controlled by the total energy of precipitates involving interface energy, inherent energy and coherent strain energy. It can be determined that during the peak aging stage, the precipitate's size was close to critical size, and Cu precipitates retained bcc structure that was coherent with matrix, resulting in, respectively, low interfacial energy and coherent strain energy. With further aging, structure transformation and coarsening were driven by a reduction in coherent strain energy and interfacial energy. Structure transformation and coarsening of Cu precipitates demonstrated an important impact on the magnetic and mechanical properties of non-oriented silicon steel.

3.3. Magnetic Properties and Mechanical Properties of Annealed and Aged Sheets

Furthermore, the magnetic and mechanical properties of the annealed sheets and aged sheets were tested, and the results are shown in Figures 10 and 11. Three results can be drawn through analysis: (1) The magnetic properties of the 0.20 mm non-oriented silicon steel prepared by two-stage cold rolling were superior to the 0.35 mm non-oriented silicon steel prepared by single-stage cold rolling. Magnetic induction increased by ~0.06 T, and the optimization of core loss was more obvious. In the conventional process, the core losses of $P_{1.5/50}$, $P_{1.0/400}$ and $P_{1.0/1000}$ were reduced by 0.5, 7.98 and 41.64 W/kg, respectively,
while in strip casting process, it can be reduced by 0.85, 11.4 and 37.94 W/kg, respectively. (2) In terms of the two different hot band fabrication processes, it may be noted that the non-oriented silicon steels processed by strip casting process exhibited more excellent magnetic properties. Magnetic induction in strip casting process was optimized by ~0.02 T, and the core losses of P$_{1.5/50}$, P$_{1.0/400}$ and P$_{1.0/1000}$ 0.20 mm sheets were reduced by 0.27, 3.49 and 14.89 W/kg, respectively. Moreover, the optimization effect in high-frequency core loss was more significant. (3) The magnetic properties of non-oriented silicon steel did not vary obviously after the ageing treatment, except that the high-frequency core loss was worsened by a little. Thus, it can be observed that strip casting process has advantages in the preparation of thin-gauge non-oriented silicon steel with superior magnetic properties, especially with respect to high-frequency core loss.

![Figure 10. Magnetic properties of annealed and aged non-oriented silicon steel: (a) B$_{50}$; (b) P$_{1.5/50}$; (c) P$_{1.0/400}$; (d) P$_{1.0/1000}$.](image)

With respect to the mechanical properties of non-oriented silicon steel, as shown in Figure 11, the following results can be obtained: (1) Strength can be significantly improved after aging. Strength first increased rapidly and then decreased slowly with the extension of aging time, and the peak aging time was about 60–120 min. Yield strength was increased by ~140–180 MPa in the conventional process and ~120 MPa in the strip casting process. With increases in strength, elongation decreased to a small extent, while the elongation of more than 12% can be guaranteed. (2) The mechanical properties of the 0.35 mm non-oriented silicon steel were superior than that of the 0.20 mm non-oriented silicon steel. The increment in the conventional process under peak aging was 35.5 MPa, while 83.4 MPa was observed in the strip casting process. Thus, the conventional process indicated advantages in preparing conventional thickness and thin-gauge non-oriented silicon steel with better mechanical properties, while the strip casting process had some limitations in improving the strength of ultra-thin non-oriented silicon steel. Nevertheless, the yield strength of 0.35 and 0.20 mm non-oriented silicon steels prepared by strip casting can still be as high as 564.8 and 501 MPa, respectively.
4. Discussion

Considering that the aging temperature of 550 °C was lower than recrystallization temperature, grain size and texture did not vary obviously; thus, magnetic induction retained high stability during the aging process. The improvement of magnetic induction in strip casting is attributed to texture optimization, based on the fact that detrimental γ texture was weakened while the favorable α texture and Goss component were enhanced. In terms of the rolling process, two-stage was much more beneficial for obtaining relatively coarse microstructure and for inhibiting the accumulation of γ texture, which is much more important in the preparation of thin-gauge non-oriented silicon steel. As a result, superior magnetic induction as high as 1.76 T was obtained in route D.

There are many factors affecting core loss, which includes chemical composition, crystallographic texture, grain size and thickness. In this study, relative coarse grain and beneficial texture contributed to decreases in core loss. Thickness reduction exhibited a dominant role in the decrease in total core loss in thinner non-oriented silicon steel, considering that the eddy current loss is proportional to the square of thickness [26]. As a result, the 0.20 mm sheet in strip casting exhibited superior core loss. The magnetic properties of non-oriented silicon steel did not vary obviously after aging treatment, which indicated that Cu precipitates had negligible detrimental effects on core loss and permeability in this study. It is generally believed that precipitates have a negative impact on the magnetic properties of non-oriented silicon steel [27]. Weak magnetic or non-magnetic precipitation will directly increase the coercivity and hysteresis loss of non-oriented silicon steel, and higher densities of dislocation around precipitates will also result in higher static magnetic and magnetoelastic properties [28]. However, Cu precipitates with a maximum size of ~15.2 nm over aging, which was less than the thickness of domain wall, played a weaker role in hindering domain wall motion and reducing saturated magnetic flux density.
In order to further understand the influence of processing route and thickness on core loss, total core loss was separated into hysteresis loss ($P_h$), eddy currency loss ($P_e$) and anomaly loss ($P_a$) via classic loss separation methods [29], as shown in Figure 12. It can be observed that the core loss value and proportion of individual loss were sensitive to thickness and test frequency. Thickness reduction significantly reduced eddy current loss, while the proportion of eddy current loss increased from 5.0 to 40.9% when the frequency varied from 50 to 1000 Hz. It was indicated that eddy currency loss and anomaly loss dominated total core loss under high frequency conditions, while hysteresis loss maintained a relatively stable level in different routes. Then, it can be inferred that texture optimization did not show obvious beneficial effects on high-frequency core loss. On the contrary, thickness reduction can effectively reduce the increment of eddy currency loss, thereby significantly reducing high-frequency core losses of non-oriented silicon steel. Moreover, based on the fact that anomaly loss mainly depended on magnetic domain structure [5], it is also very important to improve the homogeneity of microstructure. Considering that the recrystallization microstructure of thin-gauge sheets is more sensitive to the deformation state and annealing process, the control of process parameters needs to be strict.

![Figure 12. Separated core loss for different routes: (a) P1.0/50, (b) P1.0/400 and (c) P1.0/1000.](image)

In order to clarify the precipitation strengthening mechanism of Cu, it is important to quantitatively calculate the contribution from different strengthening mechanisms of Cu precipitates. It was observed that the crystal structure and size of Cu precipitates varied with aging time, and Cu precipitates retained bcc structure and were coherent with the matrix during the peak-aged stage. According to previous research results, the precipitation strengthening mechanism of coherent bCCu precipitates with sizes less than the critical size is the primarily cutting mechanism [30,31]. Therefore, the strengthening increment of tested steel aged at 550 °C for 25–120 min is mainly contributed by three strengthening mechanisms: chemical strengthening, coherency strengthening and modulus strengthening. The equation for the contribution of chemical strengthening is given as follows [32]:

$$\Delta\sigma_{\text{chemical}} = 2M/(b\lambda T^{1/2}) (\gamma_{\text{interfacial}}b)^{3/2}$$

(1)

where $M$ (=2.75) is the Taylor factor of non-oriented silicon steel, $\gamma_{\text{interfacial}}$ (=0.22 J/m²) is the average antiphase boundary energy of Cu, $b_{\alpha-Fe}$ (=2.48 Å) is the Burgers vector in the matrix and $\lambda$ is the average spacing of the precipitate among the matrix [33]:

$$\lambda = 0.866\cdot(rN)^{-1/2}$$

(2)

where $r$ is the average particle radius, and $N$ is the number density of the precipitates. $T$ is the line tension of dislocation, mostly taken by following formula [33]:

$$T = Gb^2/2$$

(3)

where $G_{Fe}$ (=8.3 × 10⁴ N/m²) is the shear modulus of matrix.
The equation for the contribution of modulus strengthening has been derived by Russell and Brown [32] as follows:

\[ \Delta \sigma_{\text{modulus}} = M \cdot \frac{Gb}{\lambda} \left[ 1 - \left( \frac{U_p}{U_m} \right)^2 \right]^{3/4} \]  

(4)

where \( U_p \) and \( U_m \) are the line energies of nanoparticles and matrix, respectively. The ratio \( U_p/U_m \) depends on the particle radius \( r \) and is described as follows:

\[ \frac{U_p}{U_m} = \left( \frac{U_p^\infty}{U_m^\infty} \right) \frac{\log(r/r_i)}{\log(r_0/r_i)} + \frac{\log(r_0/r)}{\log(r_0/r_i)} \]  

(5)

where \( r_i (=2.5 b) \) and \( r_0 (=1000 r) \) are the inner and outer cut-off radii of the dislocation stress field, respectively, and \( U_p^\infty \) and \( U_m^\infty \) refer to the line energies of the dislocations in precipitation and matrix, respectively. Based on the above calculation, a value of 0.62 for the ratio of \( U_p/U_m \) is utilized in this study.

Coherency strengthening due to the coherency strain associated with the nanoscale precipitates is described as follows [34]:

\[ \Delta \sigma_{\text{coherency}} = 8.4 \cdot MG \cdot \varepsilon^{3/2} \cdot \frac{r^2}{(N/b)^{1/2}} \]  

(6)

where \( \varepsilon (=0.0057) \) is the constrained lattice parameter mismatch.

After the peak aging stage, Cu precipitates transformed from bcc structure to non-coherent fcc structure. The interaction mechanism between dislocation and precipitates changes from cutting through softer bcc \( \text{Cu} \) to bypassing harder fcc \( \text{Cu} \), and the main precipitation strengthening mechanism of precipitation strengthening changes to the Orowan mechanism. The strengthening increment of Orowan mechanism is calculated as follows [35,36]:

\[ \Delta \sigma_{\text{Orowan}} = CM \cdot \frac{Gb}{\lambda} \cdot \log\left( \frac{r_0}{r_i} \right) \]  

(7)

where \( C (=0.127) \) is constant, and \( r_0 (=1.632 r) \) is the outer cut-off radius of the dislocation stress field during over-aged conditions.

The contributions of various precipitation strengthening mechanisms of non-oriented silicon steel with a thickness of 0.2 mm are represented in Figure 13. According to the precipitation strengthening mechanism, both density and mean size of precipitates play crucial roles in increasing strength. Then, obtaining high densities of relatively fine Cu precipitates is desired, which achieved a balance for obtaining peak strength when aged for 120 min in this study. It is evident that for the tested steel, modulus strengthening made a major contribution to strengthening under the condition of cutting mechanisms, which was as high as 93.3 MPa. Considering that the high density of Cu precipitates with critical sizes was coherent with the matrix, a maximum coherency strengthening of 67.1 MPa was achieved at the peak aging condition. By contrast, the chemical effect arising from the additional interfaces can be ignored in this study. Moreover, it is worth noting that the calculated values were ~35 MPa lower than the tested yield strength increment at different stages. Given that strength significantly decreased with the thickness close to grain size [37], the error can be attributed to the detrimental size effect in thin sheets.

Based on above discussion, Cu precipitates can be used to significantly improve yield strength over ~120 MPa without deteriorating the magnetic properties in both conventional processes and strip casting processes. In the conventional process, an initial microstructure with fine grain and relatively strong \( \gamma \) texture was formed after heavy hot rolling, which resulted in unfavorable microstructure and texture after final annealing, as well as poor magnetic properties. On the other hand, improved strengthening effects from Cu precipitates about 20–40 MPa were obtained in the conventional process, which may be related to the uniform distribution of Cu in the matrix. In strip casting processes, due to the initial microstructure with coarse grain and relatively strong \{100\} texture, beneficial microstructure and texture were obtained in this study. Meanwhile, excellent magnetic properties were obtained both in 0.35 and 0.20 mm non-oriented silicon steels, especially
for high-frequency core losses. Moreover, considering that high-frequency core loss was sensitive to thickness and homogeneity of microstructure, balanced magnetic properties and mechanical properties can be achieved in thin-gauge non-oriented silicon steel in strip casting processes. However, it should also be pointed out that thickness reduction would bring about some additional problems, such as inhomogeneity of microstructure and detrimental size effect relative to strength.

Figure 13. Calculation results of various precipitation strengthening and tested strength increment of non-oriented silicon steel with thickness of 0.2 mm.

5. Conclusions

(1) In terms of initial microstructure with coarse grain and relatively strong [100] texture, obvious shear deformations and inheritance of texture during the rolling process were observed in strip casting process. Beneficial microstructures with coarse grain sizes of ~96.4 µm and strong Goss texture accompanied by weak γ texture were obtained after two-stage cold rolling and recrystallization annealing. The strip casting process exhibited obvious advantages in the optimization of microstructure and texture to process ultra-thin non-oriented silicon steel over the conventional process.

(2) The high densities of Cu precipitates with sizes of ~3.2 nm were observed in hot rolled sheets, while less Cu precipitates with sizes of ~6.9 nm were nonuniformly distributed in the as-cast strip. In the peak ageing stage, the Cu precipitates retained bcc structure and were coherent with matrix. With further increases in aging time, rapid coarsening and complex crystal structure transformations were observed.

(3) Cu precipitates can be used to significantly improve yield strength over ~120 MPa without deteriorating the magnetic properties in both conventional processes and strip casting, and the main mechanism of precipitation strengthening was attributed to coherency strengthening and modulus strengthening. Improved strengthening effects from Cu precipitates about 20–40 MPa were obtained in the conventional process. Meanwhile, the peak yield strength of 0.2 mm non-oriented silicon steel by strip casting was 501 MPa when aged at 550 °C for 120 min.

(4) In strip casting processes, magnetic induction B₅₀ was optimized by ~0.02 T, and the core losses of P₁₅/₅₀, P₁₀/₄₀0 and P₁₀/₁₀₀₀ of 0.20 mm sheets were reduced by 0.27, 3.49 and 14.89 W/kg, respectively, compared with the conventional process. Cu precipitates had negligible effects on magnetic properties after aging. Balanced magnetic properties and mechanical properties were obtained in thin-gauge non-oriented silicon steel with Cu addition in the strip casting process.
**Author Contributions:** Conceptualization, F.F.; software, S.C.; validation, G.Y.; investigation, Z.W.; resources, Y.Z.; data curation, D.H.; writing—review and editing, F.F. and R.D.K.M.; visualization, Y.W.; supervision, X.Z.; project administration, G.W.; funding acquisition, F.F. All authors have read and agreed to the published version of the manuscript.

**Funding:** This study was funded by the National Natural Science Foundation of China (Nos. 520201060 and 51801022), the China Postdoctoral Science Foundation funded project (Nos. 2019TQ00053 and 2020M680963), the Fundamental Research Funds for the Central Universities (Nos. N2007003 and N2007011) and Natural Science Foundation of Liao Ning Province of China (Nos. 2020-B5-047).

**Data Availability Statement:** The data presented in this study are available on request from the corresponding author. The data are not publicly available due to technical or time limitations.

**Acknowledgments:** The authors are grateful to R.D.K. Misra for their contribution to the discussion.

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

**References**

1. Oda, Y.; Kohno, M.; Honda, A. Recent development of non-oriented electrical steel sheet for automobile electrical devices. *J. Magn. Mater.* 2008, 320, 2430–2435. [CrossRef]
2. Tanaka, I.; Nitomi, H.; Imanishi, K. Application of high-strength nonoriented electrical steel to interior permanent magnet synchronous motor. *IEEE Trans. Magn.* 2013, 49, 2997–3001. [CrossRef]
3. Gerada, D.; Mebarki, A.; Brown, N.L. High-Speed Electrical Machines: Technologies, Trends, and Developments. *IEEE Trans. Ind. Electron.* 2013, 61, 2946–2959. [CrossRef]
4. Gong, J.; Luo, H.W. Progress on the Research of High-strength Non-oriented Silicon Steel Sheets in Traction Motors of Hybrid/Electrical Vehicles. *J. Mater. Eng.* 2015, 43, 102–112. [CrossRef]
5. Moses, J.A. Energy efficient electrical steels: Magnetic performance prediction and optimization. *Scr. Mater.* 2012, 67, 560–565. [CrossRef]
6. Tanaka, I.; Yashiki, H. Magnetic and Mechanical Properties of Newly Developed High-Strength Nonoriented Electrical Steel. *IEEE Trans. Magn.* 2010, 46, 290–293. [CrossRef]
7. Kubota, T. Recent Progress on Non-oriented Silicon Steel. *Steel Res. Int.* 2005, 76, 464–470. [CrossRef]
8. Zhang, Y.X.; Xu, Y.B.; Liu, H.T.; Li, C.G.; Cao, G.M.; Liu, Z.Y.; Wang, G.D. Microstructure, texture and magnetic properties of strip-cast 1.3% Si non-oriented electrical steels. *J. Magn. Magn. Mater.* 2012, 324, 3328–3333. [CrossRef]
9. Jiao, H.T.; Xu, Y.B.; Zhao, L.Z. Texture evolution in twin-roll strip cast non-oriented electrical steel with strong Cube and Goss texture. *Acta Mater.* 2020, 199, 311–325. [CrossRef]
10. Fang, F.; Zhang, Y.X.; Lu, X. Abnormal growth of [100] grains and strong Cube texture in strip cast Fe-Si electrical steel. *Scr. Mater.* 2018, 147, 33–36. [CrossRef]
11. Zhang, Y.X.; Lan, M.F.; Wang, Y. Microstructure and texture evolution of thin-gauge non-oriented silicon steel with high permeability produced by twin roll strip casting. *Mater. Charact.* 2019, 150, 118–127. [CrossRef]
12. Jiao, H.; Xu, Y.; Xiong, W. High-permeability and thin-gauge non-oriented electrical steel through twin-roll strip casting. *Mater. Des.* 2017, 136, 23–33. [CrossRef]
13. De Martínez-Guerenu, A.; Arizti, F.; Díaz-Fuentes, M. Recovery during annealing in a cold rolled low carbon steel. Part I: Kinetics and microstructural characterization. *Acta Mater.* 2004, 52, 3657–3664. [CrossRef]
14. De Martínez-Guerenu, A.; Arizti, F.; Gutiérrez, I. Recovery during annealing in a cold rolled low carbon steel. Part II: Modelling the kinetics. *Acta Mater.* 2004, 52, 3665–3670. [CrossRef]
15. Barros, J. The effect of Si and Al concentration gradients on the mechanical and magnetic properties of electrical steel. *J. Magn. Magn. Mater.* 2005, 290, 1457–1460. [CrossRef]
16. Hong, J.; Choi, H.; Lee, S. Effect of Al content on Magnetic Properties of Fe-Al Non-oriented Electrical Steel. *J. Magn. Magn. Mater.* 2017, 439, 343–348. [CrossRef]
17. Lee, S.; Bruno, C.; Cooman, D. Influence of Phosphorous and Boron on the Recrystallization, Grain Growth and Mechanical Properties of 3% Si steel. *Mater. Sci. Forum.* 2010, 617, 654–656. [CrossRef]
18. Fujikura, M.; Murakami, H.; Ushigami, Y. Effects of Cu precipitates on magnetic properties of nonoriented electrical steel. *IEEE Trans. Magn.* 2015, 51, 2001604. [CrossRef]
19. Bian, X.H.; Zeng, Y.P.; Nan, D.; Wu, M. The effect of copper precipitates on the recrystallization textures and magnetic properties of non-oriented electrical steels. *J. Alloy. Compd.* 2014, 588, 108–113. [CrossRef]
20. Wu, M.; Zeng, Y.P. Effect of copper precipitates on the stability of microstructures and magnetic properties of non-oriented electrical steels. *J. Magn. Mater.* 2015, 391, 96–100. [CrossRef]
21. Wang, Y.Q.; Zhang, X.M.; He, Z. Effect of copper precipitates on mechanical and magnetic properties of Cu-bearing non-oriented electrical steel processed by twin-roll strip casting. *Mater. Sci. Eng. A* 2017, 703, 340–347. [CrossRef]
22. Goodman, S.R.; Brenner, S.S.; Low, J.R. An FIM-atom probe study of the precipitation of copper from iron-1.4 at. pct copper. Part I: Field-ion microscopy. *Metall. Trans.* 1973, 4, 2363–2369. [CrossRef]
23. Wen, Y.R.; Hirata, A.Z.; Zhang, W. Microstructure characterization of Cu-rich nanoprecipitates in a Fe–2.5Cu–1.5 Mn–4.0 Ni–1.0 Al multicomponent ferritic alloy. *Acta Mater.* **2013**, *61*, 2133–2147. [CrossRef]

24. Heo, Y.U.; Kim, Y.K.; Kim, J.S. Phase transformation of Cu precipitates from bcc to fcc in Fe-3Si-2Cu alloy. *Acta Mater.* **2013**, *61*, 519–528. [CrossRef]

25. Othen, P.J.; Jenkins, M.L.; Smith, G.D.W. High-resolution electron microscopy studies of the structure of Cu precipitates in α-Fe. *Philos. Mag. A* **1994**, *70*, 1–24. [CrossRef]

26. Bertotti, G. General properties of power losses in soft ferromagnetic materials. *IEEE Trans. Magn.* **1988**, *24*, 621–630. [CrossRef]

27. Liu, J.Z.; Walle, A.; Ghosh, G. Structure, energetics, and mechanical stability of Fe-Cu bcc alloys from first-principles calculations. *Phys. Rev. B Condens. Matter Mater. Phys.* **2005**, *72*, 144109. [CrossRef]

28. Jenkins, K.; Lindenmo, M. Precipitates in electrical steels. *J. Magn. Magn. Mater.* **2008**, *320*, 2423–2429. [CrossRef]

29. Qin, J.; Yang, P.; Mao, W.; Ye, F. Effect of texture and grain size on the magnetic flux density and core loss of cold-rolled high silicon steel sheets. *J. Magn. Magn. Mater.* **2015**, *393*, 537–543. [CrossRef]

30. Osamura, K.; Okuda, H.; Takashima, M. Small-angle neutron scattering study of phase decomposition in Fe-Cu binary alloy. *Mater. Trans. JIM* **1993**, *34*, 305–311. [CrossRef]

31. Xiong, Z.P.; Timokhina, I.; Pereloma, E. Clustering, nano-scale precipitation and strengthening of steels. *Prog. Mater. Sci.* **2021**, *118*, 100764. [CrossRef]

32. Brown, L.M.; Ham, R.K. Dislocation-particle interactions. In *Strengthening Methods in Crystals*; Kelly, A., Nicholson, R.B., Eds.; Applied Science Publishers: London, UK, 1965; pp. 9–135.

33. Xu, S.S.; Zhao, Y.; Chen, D. Nanoscale precipitation and its influence on strengthening mechanisms in an ultra-high strength low-carbon steel. *Int. J. Plast.* **2019**, *113*, 99–110. [CrossRef]

34. Russell, K.C.; Brown, L.M. A Dispersion Strengthening Model based on differing elastic moduli applied to the iron-copper system. *Acta Metall.* **1972**, *20*, 969–974. [CrossRef]

35. Ashby, M. The theory of the critical shear stress and work hardening of dispersion-hardened crystals. In *Metallurgical Society Conference*; Ansell, G.S., Cooper, T.D., Lenel, F.V., Eds.; Gordon and Breach: New York, NY, USA, 1968; Volume 47, pp. 143–205.

36. Sonderegger, B. Modifications of stereological correction methods for precipitate parameters using transmission microscopy. *Ultramicroscopy* **2006**, *106*, 941–950. [CrossRef] [PubMed]

37. Michel, J.F.; Picart, P. Size effects on the constitutive behaviour for brass in sheet metal forming. *J. Mater. Process. Technol.* **2003**, *141*, 439–446. [CrossRef]