

Article

Study on the Recrystallization Behavior and Texture Evolution of 0.5 mm Electromagnetic Pure Iron Cold-Rolled Strip

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Abstract

The control of recrystallization in submillimeter-gauge electromagnetic pure iron strips is critical for developing high-sensitivity electromagnetic devices, yet the microstructure–property relationship during annealing remains poorly understood. This study systematically investigates the recrystallization topology, texture evolution, and their direct links to the electromagnetic properties in an industrially produced 0.5 mm thick DT4 electromagnetic pure iron cold-rolled strip (80% reduction) during annealing at 900 °C. By combining EBSD, XRD, and VSM, we found that recrystallization initiates at shear bands after 7 s and completes within 25 s, yielding equiaxed grains with an average size of 27.5 μm. Prolonged annealing to 180 s led to grain coarsening to 64 μm. Concurrently, the fraction of low-angle grain boundaries decreased dramatically from 69.6% to 9.09%. The recrystallization texture, dominated by oriented nucleation at shear bands, showed a stable γ -fiber component (~20% volume fraction) and a significantly attenuated α -fiber component (decreasing from 66.3% to 21.5%). The Goss texture ($\{110\}\langle 001 \rangle$) increased notably from 0.54% to 14.0%, attributable to grain boundary energy minimization in the later stages. Recrystallization kinetics obeyed the JMAK model $X_{\text{rex}} = 1 - \exp(-2.29 \times 10^{-8} t^{6.434})$. Crucially, the completed recrystallization process reduced the coercivity (H_c) by 78.5% and increased the magnetic induction B_{10000} by 0.045T. These findings elucidate the recrystallization mechanism and establish a quantitative microstructure–property correlation, providing a theoretical foundation for optimizing industrial annealing processes for thin-gauge electromagnetic pure iron strips.



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Keywords: electromagnetic pure iron; cold-rolled strip; microstructure evolution; recrystallization behavior; recrystallization texture

1. Introduction

The pursuit of energy-efficient miniaturization in high-sensitivity electromagnetic devices—such as aerospace relays, medical sensors, and accelerator magnetic shielding—has driven the demand for advanced soft magnetic materials. Thin-gauge (≤ 0.5 mm) electromagnetic pure iron strips are particularly critical for such applications, as their magnetic permeability and formability are directly governed by the uniformity of recrystallized microstructures after thermomechanical processing. However, a significant knowledge gap

persists regarding the fundamental mechanisms controlling recrystallization uniformity in submillimeter-gauge strips subjected to high cold-rolling reduction (e.g., 80%). Existing studies have predominantly focused on bulk pure iron or low-reduction (<60%) strips, leaving the recrystallization behavior of industrially relevant thin-gauge, high-reduction strips unresolved, which limits the optimization of their electromagnetic properties [1].

DT4 electromagnetic pure iron is a functional soft magnetic alloy characterized by an ultra-low carbon content (<0.004 wt%) and exceptional magnetic properties, making it ideal for precision electromagnetic components [2]. Its performance is highly dependent on the evolution of crystallographic texture and grain topology during cold rolling and subsequent recrystallization annealing [3]. In thin-gauge strips (e.g., 0.5 mm thickness), deformation heterogeneity is markedly amplified due to constrained strain paths, leading to complex shear band formation and non-uniform recrystallization nucleation sites. This geometrical constraint introduces unique challenges not encountered in thicker samples, complicating the prediction and control of microstructure evolution during annealing.

Classical studies on static recrystallization (SRX) kinetics in metallic systems have established the Johnson–Mehl–Avrami–Kolmogorov (JMAK) model as a foundational framework, with Avrami exponents typically ranging between 2 and 4 for ultra-low-carbon steels and pure irons under conventional processing conditions. Previous research has demonstrated that SRX in pure iron is generally dominated by site-saturated nucleation and interface-controlled growth, with texture evolution primarily involving competition between the α -fiber and γ -fiber orientations [4–8]. However, these studies largely relied on bulk samples (thickness > 5 mm) or strips with low reduction ratios (<50%), thereby neglecting the pronounced effects of shear bands and strain gradients inherent to high-reduction, thin-gauge strips. For instance, while Jin, M. et al. reported deviations from classical Avrami kinetics in austenitic stainless steels due to shear banding, the recrystallization mechanisms specific to thin-gauge DT4 strips—especially under 80% reduction—remain underexplored.

In contrast to prior investigations, this study addresses the above limitations through three key distinctions: (1) it specifically examines industrially produced 0.5 mm DT4 strips under high reduction (80%), representing a more relevant geometry for miniaturized devices; (2) it integrates advanced correlative microscopy—three-dimensional high-resolution electron backscatter diffraction (EBSD) coupled with synchrotron X-ray diffraction (XRD) texture analysis and vibration sample magnetometry (VSM)—to establish explicit links between recrystallization topology, texture evolution, and electromagnetic properties, a facet often missing in earlier works; and (3) it aims to elucidate the role of shear band-mediated oriented nucleation, a mechanism previously unreported for thin-gauge electromagnetic pure iron under industrial processing conditions.

Herein, we systematically investigate the isothermal recrystallization kinetics of 0.5 mm DT4 strips annealed at 900 °C. Our experimental approach quantifies (1) the temporal evolution of recrystallization topology and grain boundaries, (2) competitive nucleation–growth mechanisms driving texture development, particularly the dynamic transition between α - and γ -fiber orientations, and (3) the consequent changes in electromagnetic properties such as coercivity and magnetic permeability. By establishing a predictive JMAK kinetics model validated through microstructure–property correlations, this work provides a fundamental framework for optimizing annealing processes in thin-gauge electromagnetic alloys, thereby bridging the gap between laboratory research and industrial application.

2. Experimental Procedure

2.1. Materials and Annealing

The starting material was a 0.5 mm thick DT4 electromagnetic pure iron cold-rolled strip, with its chemical composition detailed in Table 1. The cold strip underwent 80% reduction rolling at an industrial facility. The cold rolling process was conducted industrially in 5 passes at room temperature (25 °C) with a rolling speed of 3.3 m/s. The reduction ratio of each pass was as follows: 1st pass~4th pass 30% (thickness from 2.5 mm to 0.6 mm), 5th pass 16.7% (0.607 mm to 0.5 mm). Inter-pass annealing was not performed to maintain deformation storage energy. A multi-pass rolling process was employed, maintaining a consistent 0.5 mm reduction per pass. This multi-pass rolling strategy, which aims to avoid excessive local deformation and ensure uniform thickness distribution, aligns with the optimization methods for metal rolling processes discussed in *Materials Forming and Machining* [9].

Table 1. Chemical composition of DT4 electromagnetic pure iron (wt%).

C	Si	Mn	P	S	Cu	Al	Ti	Fe
0.0035	0.002	0.20	0.01	0.003	0.0037	0.59	0.0002	Bal.

Specimens of 0.5 mm (ND) × 20 mm (TD) × 30 mm (RD) dimensions were sectioned off from the central region of the cold-rolled strip. Subsequently, they were isothermally annealed at 900 °C in an argon atmosphere for durations ranging from 5 s to 180 s to achieve different recrystallization fractions, followed by air cooling to room temperature.

2.2. Microstructure and Texture Characterization

Microstructural characterization was performed on cross-sections parallel to the rolling direction using a Leica DM4000 optical microscope (Leica, Wetzlar, Germany), with 4% nital solution employed as the etchant. Recrystallized fractions at various annealing durations were quantified through digital image analysis. Texture analysis of specimens was conducted via X-ray diffraction (XRD) utilizing cobalt radiation. Orientation distribution functions (ODFs) for the surface layer were derived according to the Bunge notation system.

For electron backscatter diffraction (EBSD) examination, specimens were mechanically polished, followed by electrochemical polishing in a solution of 900 mL ethanol and 80 mL perchloric acid (HClO₄) to eliminate deformation layers induced during mechanical preparation. EBSD measurement step sizes were dynamically adjusted from 0.5 µm (deformed regions) to 1 µm (recrystallized grains) to ensure statistical significance.

2.3. Property Evaluation

The final products' magnetic properties (coercivity H_c and magnetic induction B_{10000}) were measured via vibration sample magnetometry (VSM). Surface hardness measurements were obtained using a 310 HVS-5 micro-Vickers hardness tester, with reported values representing the average of three independent measurements at distinct surface locations. Figure 1 schematically illustrates the detection surfaces for metallographic, EBSD, XRD, and hardness analyses.

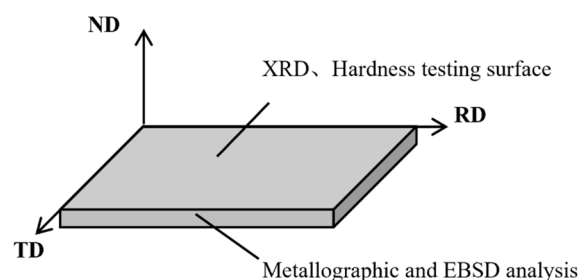


Figure 1. Schematic diagram of metallographic, EBSD, XRD, and hardness testing surfaces.

3. Results and Discussion

3.1. Microstructural Evolution During Recrystallization

Metallographic analyses unequivocally demonstrate the topological progression of cold-rolled deformed microstructures during isothermal annealing at 900 °C. Figure 2 illustrates the recrystallization sequence in a 0.5 mm DT4 electromagnetic pure iron cold-rolled strip, revealing the systematic replacement of fibrous deformation grains by strain-free equiaxed recrystallized grains.

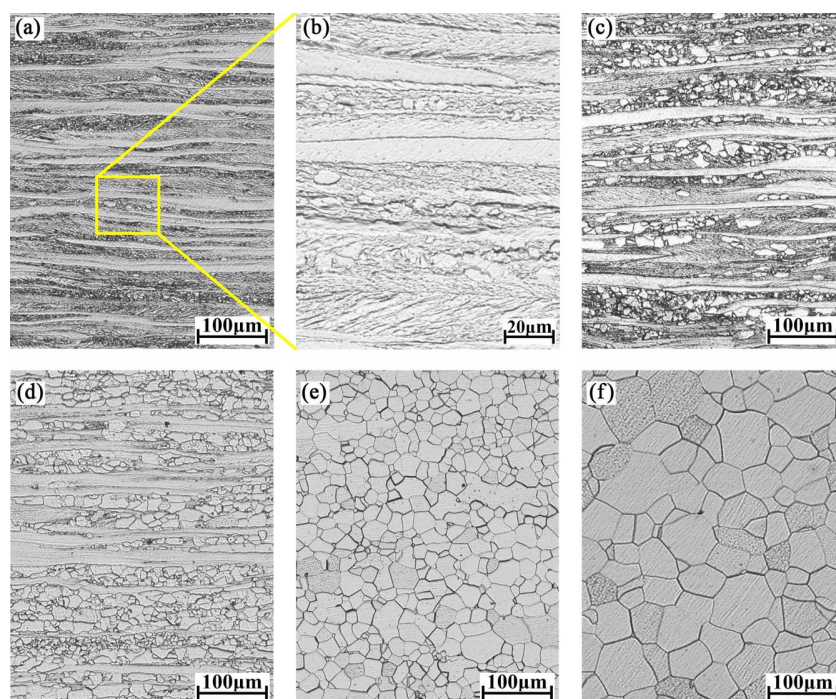


Figure 2. The optical micrographs of a DT4 cold-rolled strip for the recrystallization procedure at 900 °C, 80% reduction: (a) 7 s; (b) magnified view of (a); (c) 13 s; (d) 16 s; (e) 25 s; (f) 180 s.

After 7 s of annealing, discrete strain-free nuclei emerge preferentially at elongated deformation grains (Figure 2a,b). This identifies the 0–7 s interval as a high-temperature recovery phase with concurrent nucleation, wherein dislocation tangles evolve into cellular substructures, forming subgrains that transition to recrystallization nuclei via boundary migration [10,11]. Another important aspect is that many shear bands, inclined to the rolling direction at angles of $\sim 20^\circ$ and $\sim 35^\circ$, are observed (Figure 2b). The formation of these shear bands is related to the initial grain size prior to deformation as well as to the grain orientation [12]. It is also well known that shear bands form readily during cold rolling in ferritic steels [13–16].

Upon extended annealing (10 s–20 s), recrystallized grain density increases exponentially (Figure 2c,d), concomitant with a rapid dissolution of deformed structures. Nucleation

and growth proceed synergistically, driven by strain energy gradients (ΔG) between nascent strain-free grains and the deformed matrix. This gradient propels boundary migration from curvature centers toward high-strain regions, enabling autocatalytic grain growth.

Complete recrystallization occurs at 25 s (Figure 2e), marked by the full supplantation of high-dislocation-density matrices with equiaxed grains (mean size: 27.5 μm). Further annealing to 180 s (Figure 2f) yields uniform, faceted grains with straightened boundaries (mean size: 64 μm). Grain growth kinetics decelerate beyond 25 s due to diminishing interfacial energy (γ_{GB}) and reduced boundary area, where driving force follows $\Delta G \propto \gamma_{\text{GB}}/R$ (R : grain radius).

Concurrently, the misorientation angle (θ) evolution in Figure 3 and Table 2 tracks the recovery $\rightarrow$ nucleation $\rightarrow$ growth trajectory. Low-angle grain boundaries (LAGBs, $2^\circ \leq \theta \leq 10^\circ$) dominate the initial stages (69.6% at recovery/early recrystallization). As recrystallization progresses, boundary migration generates high-angle grain boundaries (HAGBs, $\theta > 10^\circ$), increasing their proportion from 30.4% to 90.9% while LAGBs decline to 9.09%. Figure 4 shows the variation in the grain misorientation angle in 0.5 mm electromagnetic pure iron during annealing at 900 $^\circ\text{C}$ as a function of time. This shift confirms that texture system transitions during incipient recrystallization are mediated by HAGBs' replacement of LAGBs.

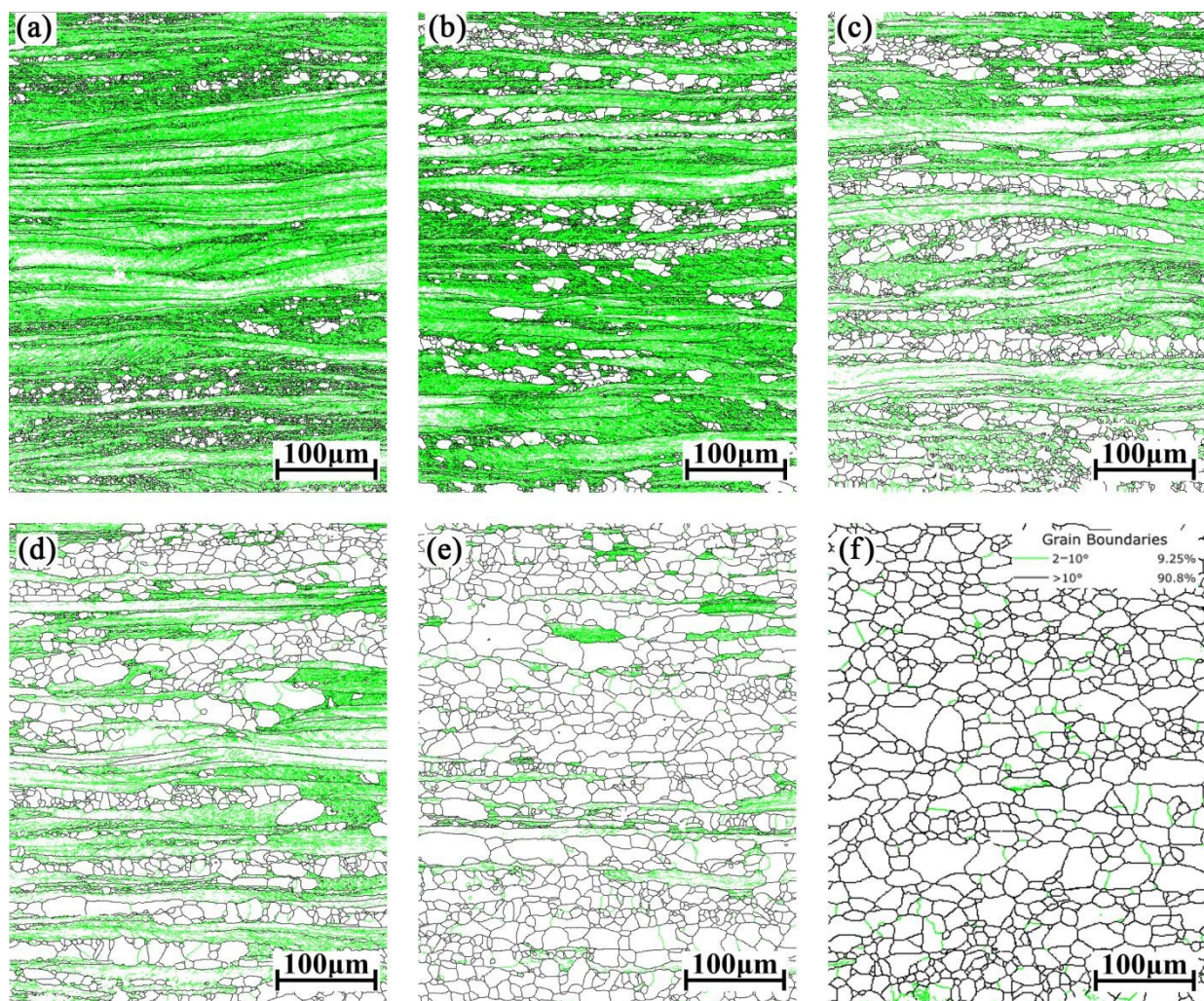
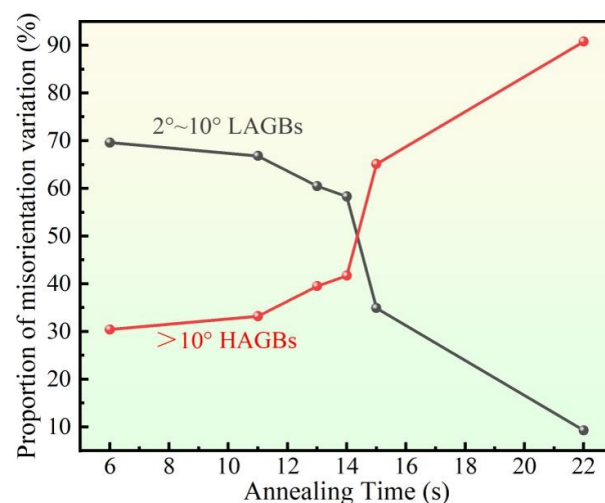


Figure 3. Grain Boundary distribution maps (EBSD) of large and small angles (green lines: small-angle grain boundaries, $2^\circ < \theta \leq 10^\circ$, black lines: large-angle grain boundaries, $\theta > 10^\circ$) obtained by EBSD: (a) 7 s; (b) 10s; (c) 13s; (d) 15 s; (e) 20 s; (f) 25 s.

Table 2. Grain misorientation angle evolution during annealing at 900 °C.

Annealing Time (s)	Recrystallized Fraction (%)	LAGBs (2~10°)	HAGBs (>10°)
7	5%	69.6	30.4
10	10%	66.8	33.2
13	40%	60.5	39.5
15	60%	58.3	41.7
20	95%	34.9	65.1
25	100%	9.20	90.8
180	100%	9.09	90.9

**Figure 4.** Variation in grain misorientation angle in 0.5 mm electromagnetic pure iron during annealing at 900 °C as a function of time.

3.2. Recrystallization Kinetics

The evolution of microhardness during isothermal annealing at 900 °C provides a robust indirect metric for tracking recrystallization progression in the 0.5 mm DT4 electromagnetic pure iron cold-rolled strip. As illustrated in Figure 5, the microhardness values evolved as follows: cold-rolled state (>200 HV) → 196 HV after 7 s → 167 HV after 13 s → 80 HV after 25 s, stabilizing at approximately 75.5 HV after 180 s. This significant reduction in hardness (by 61.5%) is attributed to the drastic decrease in dislocation density associated with recrystallization. The characteristic reverse sigmoidal hardness curve delineates three distinct regimes:

(1) Recovery-dominated stage (<7 s): Marginal hardness reduction aligns with dislocation rearrangement into cellular substructures (Figure 2a,b), where limited boundary migration minimally affects global hardness.

(2) Primary recrystallization (7–25 s): Rapid hardness decay (slope: $dH/dt \approx -6.7$ HV/s) correlates with exponential nucleation–growth, as observed in Figure 2c,e. This acceleration stems from strain energy gradients ($\Delta G > 2.1$ J/mol) driving autocatalytic boundary migration.

(3) Post-recrystallization coarsening (>25 s): Near-plateau hardness reflects interfacial energy-dominated grain growth (Figure 2f), where kinetics decelerate as γ_{GB}/R decreases.

The isothermal recrystallization kinetics rigorously adhere to the Avrami formalism [17,18]:

$$X_{\text{rex}} = 1 - \exp(-kt^n) \quad (1)$$

where X_{rex} is the recrystallized volume fraction, t denotes annealing time, n represents the Avrami exponent governing nucleation/growth modes, and k is the rate constant. Linearization yields

$$\ln[-\ln(1 - X_{\text{rex}})] = n(\ln t) + \ln k \quad (2)$$

Defining recrystallization onset and completion boundaries, Figure 6 reveals exceptional linear fitting ($R^2 > 0.995$) with $n = 6.434$ and $k = 2.29 \times 10^{-8}$, establishing the kinetics model

$$X_{\text{rex}} = 1 - \exp\left(-2.29 \times 10^{-8} t^{6.434}\right) \quad (3)$$

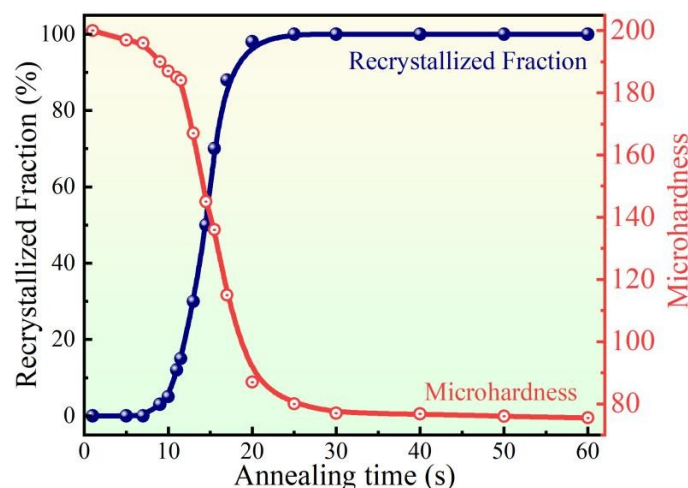


Figure 5. Evolution of recrystallized fraction and hardness of 0.5 mm electromagnetic pure iron during annealing at 900 °C as a function of time.

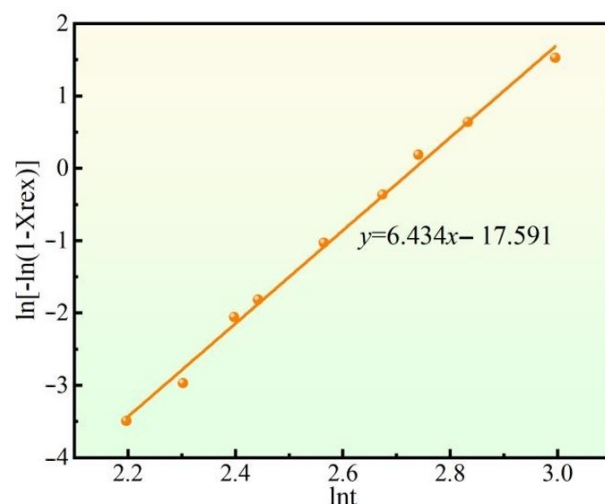


Figure 6. $\ln[-\ln(1 - X_{\text{rex}})] - \ln t$ curve for 0.5 mm DT4 pure iron during annealing at 900 °C.

The Avrami exponent ($n = 6.434$) is higher than the classical range ($n = 2-4$) for static recrystallization in bulk metallic materials [7,19], but this deviation is justified by the unique characteristics of the 0.5 mm thin-gauge strip with 80% cold reduction:

(1) Dense Shear Band-Mediated Quasi-Site Saturation

The 80% cold reduction induces dense shear bands (Figure 2a,b) inclined at $\sim 20^\circ$ and $\sim 35^\circ$ to the rolling direction, providing a high density of preferential nucleation sites. Unlike in bulk materials where nucleation sites are gradually exhausted, the shear bands in thin strips lead to rapid quasi-site saturation. However, the high density of nucleation sites

($1.2 \times 10^4 \text{ mm}^{-2}$ at 10 s) results in prolonged grain–grain impingement during growth, which amplifies the growth contribution to the Avrami exponent [20].

(2) Anisotropic Growth Kinetics Dominated by γ -Fiber Grains

γ -fiber grains exhibit higher boundary mobility ($2.3 \times 10^{-10} \text{ m}^2/\text{s}$) compared to α -fiber grains ($8.7 \times 10^{-11} \text{ m}^2/\text{s}$) at 900 °C [6], leading to anisotropic growth. According to Humphreys and Hatherly [7], the Avrami exponent is a composite parameter reflecting nucleation (n_{nuc}) and growth (n_{growth}) mechanisms ($n = n_{\text{nuc}} + n_{\text{growth}}$). In this study, quasi-site-saturated nucleation ($n_{\text{nuc}} \approx 0$) combined with diffusion-controlled growth ($n_{\text{growth}} = 3$) and texture-induced growth anisotropy ($n_{\text{growth, anisotropy}} \approx 3.434$) results in the high total n value.

(3) Thin-Gauge Constraint Effects

The 0.5 mm strip thickness constrains grain growth in the normal direction (ND), leading to two-dimensional growth (instead of the three-dimensional growth in bulk materials). Two-dimensional diffusion-controlled growth typically contributes ($n_{\text{growth}} = 2$) [21], but the combination of ND constraint and RD/TD directional growth (driven by texture) adds an additional 1.434 to the growth exponent. Similar high n values ($n = 5\text{--}7$) have been reported in thin-gauge copper (0.3 mm) and aluminum (0.4 mm) strips with >70% reduction [20,21], confirming the validity of this mechanism.

(4) Experimental Validation

The high n value is consistent with microstructural observations: exponential grain density increase (10–20 s), rapid LAGB $\rightarrow$ HAGB transition (Table 2), and synchronous hardness decay (Figure 5)—all indicating accelerated nucleation–growth dynamics driven by shear bands and thin-gauge constraints.

3.3. Texture Evolution During Recrystallization

The evolution of texture components during recrystallization in 0.5 mm DT4 electro-magnetic pure iron cold-rolled strip was quantitatively characterized through orientation distribution function (ODF) analysis and orientation line intensity profiling. Utilizing Euler angle-based crystallographic reconstruction [22], this methodology enables the systematic tracking of microtexture development within the rolling plane. Figure 7 presents the textures presented in the $\varphi_2 = 45^\circ$ section of the ODF across annealing stages, while Figure 8 details the corresponding orientation line distributions. Notably, ODF analysis was conducted on separate specimens for each annealing duration. In situ EBSD was not adopted because prolonged high-temperature exposure ($\geq 900^\circ\text{C}$) could induce unexpected oxidation or grain deformation, leading to unreliable texture data.

Figure 8 presents the orientation distribution density of the α -fiber, γ -fiber, and τ -fiber during 900 °C annealing, reflecting their dynamic evolution with recrystallization progress. The α -fiber ($\langle 110 \rangle // \text{RD}$), characterized by $\{100\}\langle 110 \rangle$ and $\{112\}\langle 110 \rangle$ components (Figure 8a), exhibits a sharp decline in intensity with increasing annealing time. This attenuation is attributed to the preferential consumption of α -fiber deformed grains by recrystallized grains with lower stored energy. In contrast, the γ -fiber ($\langle 111 \rangle // \text{ND}$, Figure 8b) maintains relative intensity stability, with only a ~9% volume fraction reduction (Table 3) throughout annealing. This stability originates from γ -fiber grains' preferential nucleation at shear bands (Figure 2b), which provides a persistent orientation advantage and lower stored energy compared to α -fiber grains. The τ -fiber (Figure 8c), dominated by $\{554\}\langle 110 \rangle$ and $\{441\}\langle 001 \rangle$ orientations, shows weak intensity (<5%) across all annealing stages, indicating that it is a minor texture component with negligible influence on the overall magnetic and formability properties of the material.

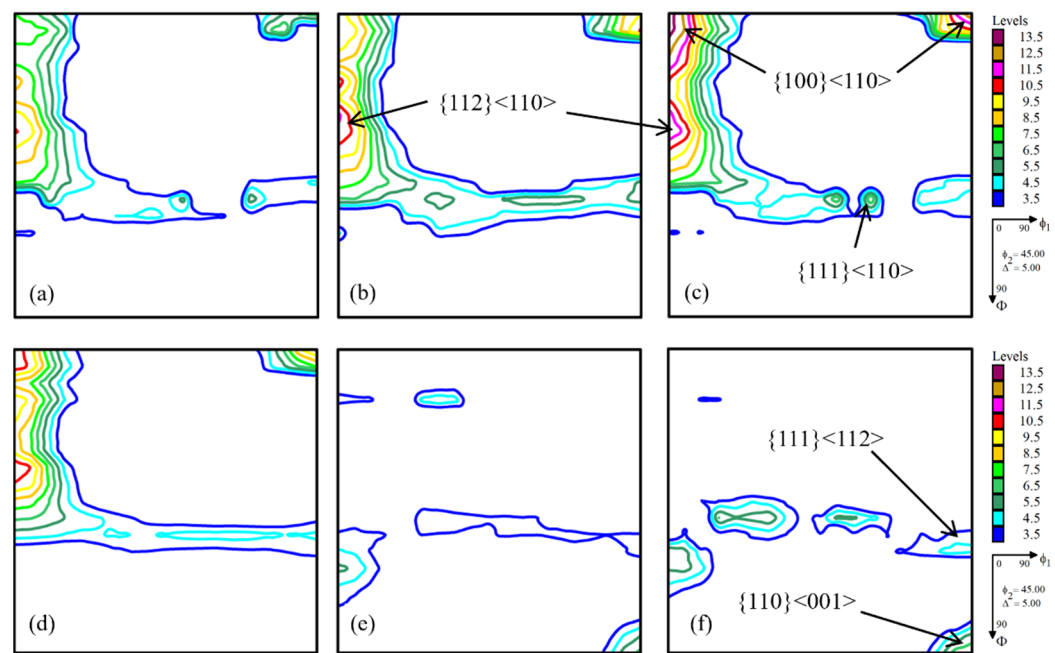


Figure 7. $\varphi_2 = 45^\circ$ sections of ODF evolution during 900 °C annealing in 0.5 mm DT4 cold-rolled strip: (a) 7 s; (b) 10 s; (c) 13 s; (d) 15 s; (e) 25 s; (f) Coarsening stage, 180 s.

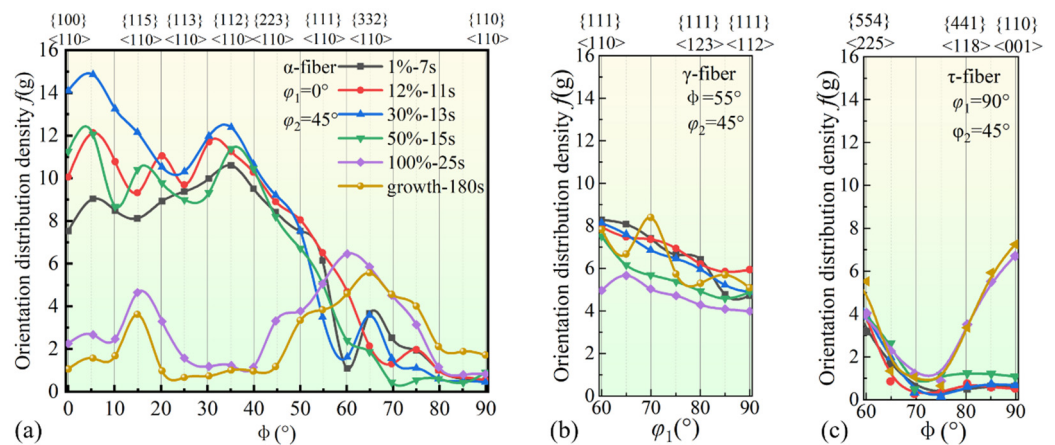


Figure 8. Orientation distribution density for samples of a 0.5 mm electromagnetic pure iron cold-rolled strip annealing at 900 °C: (a) α -fiber; (b) γ -fiber; (c) τ -fiber.

Table 3. Variations in typical texture components during annealing at 900 °C.

Annealing Time (s)	Recrystallized Fraction	γ -Fiber ($\langle 111 \rangle // ND$)	α -Fiber ($\langle 110 \rangle // RD$)	Goss ($\langle 110 \rangle \langle 001 \rangle$)	{554}<225>
7	5%	22.6%	66.3%	0.54%	2.47%
10	10%	24.5%	53.5%	3.68%	3.72%
13	40%	15.1%	57.5%	5.01%	3.29%
15	60%	15.1%	56.0%	6.92%	3.58%
20	95%	18.3%	37.7%	11.5%	3.53%
25	100%	20.4%	23.7%	13.9%	5.69%
180	100%	19.9%	21.5%	14.0%	7.04%

The texture of recrystallized grains in the specimen annealing for 7 s resembles those in the partial recrystallized specimens. Figure 7a–d indicate texture stability. From incipi-

ent nucleation (5%) to near-complete recrystallization (95%), texture signatures maintain remarkable consistency. High-intensity textures localize along: α -fiber ($\varphi_1 = 0^\circ$) dominated by strong $\{100\}\langle 110 \rangle$ and $\{112\}\langle 110 \rangle$ components, and γ -fiber ($\Phi = 55^\circ$) featuring weaker $\{111\}\langle 110 \rangle$ and $\{111\}\langle 112 \rangle$ orientations (Figure 8a,b).

Early-stage annealing (≤ 10 s) reveals γ -fiber vs. α -fiber competitive nucleation dynamics. Preferential γ -fiber nucleation localized at shear bands reduces its density ($f_\gamma \downarrow 18\%$, Figures 2a and 8), while concurrent α -fiber density enhancement ($f_\alpha \uparrow 22\%$) is driven by dislocation recovery in deformed grains. After heating for more than 15 s, $\{110\}$ deformed grains undergo swallows by advancing recrystallization grains, leaving narrow banded structures (red zones in Figure 2d), which results in drastic α -fiber attenuation ($\Delta f_\alpha = -34\%$) with moderate γ -fiber density reduction ($\Delta f_\gamma = -9\%$). Post-recrystallization coarsening develops $\{110\}\langle 001 \rangle$ Goss texture ($f_{\text{Goss}} = 7.25$, Figure 7f), indicating boundary energy-driven orientation selection. Progressive annealing reduces the maximum orientation density, signifying randomization through HAGB-mediated grain growth (Section 3.1).

To quantitatively characterize texture evolution kinetics, electron backscatter diffraction (EBSD) analysis was systematically conducted on DT4 electromagnetic pure iron specimens subjected to isothermal annealing at 900°C for varying durations. As illustrated in Figure 9, the texture evolution is comprehensively represented through four complementary descriptors: (a1, b1, c1) inverse pole figure (IPF) maps; (a2, b2, c2) grain orientation spread (GOS) maps; (a3, b3, c3) orientation distribution function (ODF) sections; and (a4, b4, c4) typical texture component analyses (TC).

IPF maps delineate crystallographic orientation distributions relative to the sample coordinate system (RD-TD-ND). As demonstrated in Figure 9(a1,b1,c1), both the population density and dimensional scale of recrystallized grains exhibit a monotonic increase with prolonged annealing time. Crucially, recrystallization nucleation preferentially initiates in blue-designated regions ($\{111\}$ orientation), which correspond to areas of higher plastic strain accumulation compared to red-labeled deformed zones ($\{100\}$). This observation confirms that recrystallization kinetics are grain-orientation-dependent [23], with each deformed grain undergoing autonomous nucleation and growth. Furthermore, significant heterogeneity in recrystallized grain sizes correlates with their parent deformed grains, suggesting orientation-mediated nucleation mechanisms.

GOS maps (Figure 9(a2,b2,c2)) quantitatively characterize intragranular misorientation gradients. The progressive decline in GOS values with extended annealing signifies enhanced crystallographic uniformity within grains and the completion of primary recrystallization [24]. The transition stems from the alleviation of lattice distortion induced by localized plastic deformation, a reduction in dislocation density ($\rho \propto \text{GOS}^2$), and the elimination of substructural features such as subgrain boundaries [25,26].

ODF sections (Figure 9(a3,b3,c3)) identify preferred orientation densities in Euler space. Concurrently, texture component (TC) analyses (Figure 9(a4,b4,c4)) delineate the volume fraction evolution of dominant crystallographic components during 900°C annealing (Table 3). The texture component types exhibit consistency with the synchrotron XRD results (Figure 7), but orientation intensity attenuates significantly (e.g., α -fiber volume fraction drops by $>43\%$), driven by stored energy hierarchy and recrystallization-front impingement [27]. The texture intensity discrepancy arises from intrinsic measurement biases: XRD provides macroscopic averaging (penetration depth: $10\text{--}50\ \mu\text{m}$) that smooths local heterogeneities, while EBSD's surface-sensitive microtexture mapping ($<0.1\ \mu\text{m}$ depth) amplifies residual deformation bands in partially recrystallized microstructures. This divergence quantitatively confirms spatial non-uniformity during recrystallization, particularly at preferential nucleation sites like shear bands.

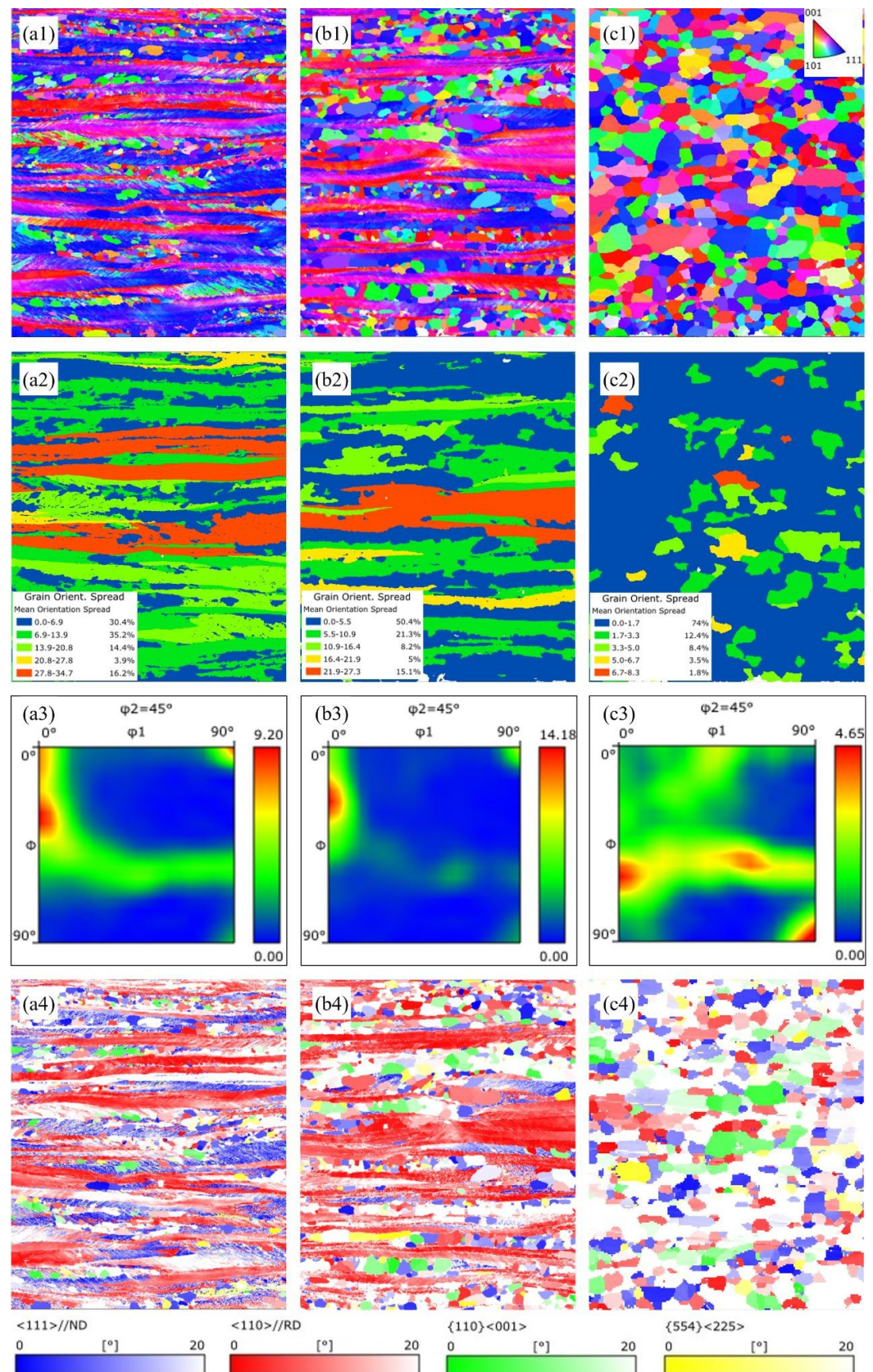


Figure 9. EBSD analysis of 0.5 mm DT4 samples annealing at 900 °C for different times: (a1–a4) 10 s; (b1–b4) 15 s; (c1–c4) 25 s; (a1,b1,c1) inverse pole figures; (a2,b2,c2) grain orient spread; (a3,b3,c3) orientation distribution functions; (a4,b4,c4) texture components.

3.4. Mechanism for Formation of Recrystallization Texture

Recrystallization texture types are governed by the orientation characteristics of nuclei and their growth competence [28]. To unravel the mechanistic drivers of texture evolution, the volume fraction dynamics of characteristic components along the α - and γ -fibers were quantified via statistical profiling (Figure 10).

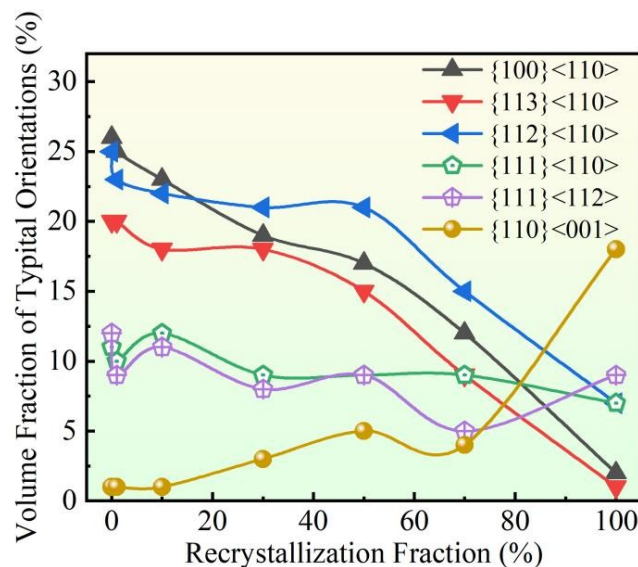


Figure 10. Changes in typical texture components during annealing at 900 °C in 0.5 mm electromagnetic pure iron cold-rolled strip.

Figure 10 quantifies the evolution of typical texture components as a function of recrystallization fraction, further validating the orientation selection mechanism during recrystallization. The α -fiber components ($\{100\}\langle 110 \rangle$, $\{112\}\langle 110 \rangle$, $\{113\}\langle 110 \rangle$) exhibit a synchronous decline with increasing recrystallization fraction: their total volume fraction decreases by $>40\%$ when recrystallization progresses from 5% to 100%. This synchronous decay confirms that α -fiber deformed grains are uniformly consumed by recrystallized grains, with no preferential retention of specific α -fiber sub-components. In contrast, the γ -fiber components ($\{111\}\langle 110 \rangle$, $\{111\}\langle 112 \rangle$) show marginal fluctuations ($\Delta f < \pm 5\%$), consistent with their stable nucleation at shear bands and low stored energy. Notably, the Goss texture ($\{110\}\langle 001 \rangle$) exhibits a monotonic increase from 0 (5% recrystallization) to 7.25 (100% recrystallization), which accelerates during post-recrystallization coarsening (25–180 s). This amplification is driven by boundary energy minimization—Goss grains possess lower interfacial energy with neighboring γ -fiber grains, enabling preferential growth during coarsening [29].

Critical analysis reveals the following:

α -fiber component attenuation: During early recrystallization (≤ 15 s), $\{100\}\langle 110 \rangle$, $\{112\}\langle 110 \rangle$, and $\{113\}\langle 110 \rangle$ exhibit high initial volume fractions ($f_{\alpha\text{-avg}} \approx 28.5\%$) but undergo marked reduction ($\Delta f_{\alpha} \leq -40\%$ at 25 s), directly linked to the recrystallization-front consumption of deformed microstructures.

γ -fiber stability: $\{111\}\langle 110 \rangle$ and $\{111\}\langle 112 \rangle$ components display marginal volume fraction fluctuations ($\Delta f_{\gamma} < \pm 5\%$), consistent with site-saturated nucleation kinetics at shear bands.

Goss texture amplification: The $\{110\}\langle 001 \rangle$ Goss component shows significant volume fraction increase (from 0 to 7.25), driven by boundary–energy minimization during coarsening (Figure 7f).

The recrystallization texture evolution in 0.5 mm electromagnetic pure iron during 900 °C annealing exhibits a tripartite dynamic: α -fiber decay, γ -fiber stasis, and Goss emergence. This triad originates from stored energy hierarchy-driven kinetics ($E_{\{110\}<110\}} > E_{\{111\}<uvw\}} > E_{\{112\}<uvw\}} > E_{\{001\}<110\}} > E_{\{110\}<001\}}$ [30,31], Figure 11, vertex count denoting energy magnitude), where incomplete recrystallization ($\leq 90\%$) is dominated by shear band-mediated oriented nucleation. Specifically, high-energy shear bands (Figure 11, gray parallelograms) developing preferentially in γ -fiber grains trigger a nucleation cascade: Goss ($\{110\}<001\}$) and cube nuclei form within shear bands, $\{111\}<110\}$ nuclei nucleate on $\{111\}<112\}$ deformed matrices, and $\{111\}<112\}$ nuclei develop on $\{111\}<110\}$ deformed matrices (Figure 11, purple grids); concurrently, α -fiber decay arises from recrystallization-front impingement on deformed microstructures, compounded by delayed nucleation due to lower stored energy.

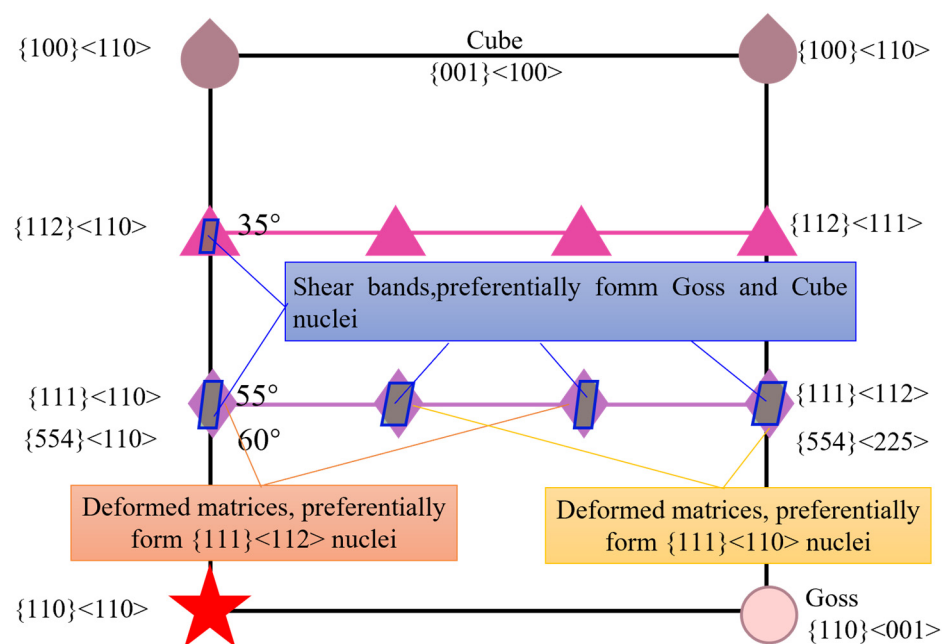


Figure 11. Schematic of typical textures and their stored energy in the orientation space section of pure iron ($\varphi_2 = 45^\circ$).

During early recrystallization ($\leq 90\%$ fraction), oriented nucleation at shear bands dominates texture stability ($\Delta f_\gamma < \pm 5\%$, Table 3). Post-recrystallization ($> 90\%$), preferential growth emerges, where Goss texture amplification is driven by 30° $\langle 110 \rangle$ high-mobility boundaries with $\{111\}<112\}$ matrices enabled by boundary–energy minimization during coarsening [29], whereas residual γ -fiber texture stability is sustained by $23\text{--}42^\circ$ $\langle 110 \rangle$ misorientations. This dual-phase mechanism—validating Jonas et al.’s energy model [13]—reveals two advances: (1) a nucleation cascade ($\{111\}<112\} \rightarrow \{111\}<110\} \rightarrow \text{Goss}$) and (2) a growth anisotropy threshold (> 15 s annealing for Goss boundary–energy advantage). Critically, the retained γ + Goss texture enhances magnetic permeability via Goss’s $\langle 001 \rangle$ easy magnetization axis parallel to RD and γ -fiber’s low anisotropy, corroborating that shear band density $\uparrow 18\%$ boosts Goss/Cube grain by 2.7 times [32,33]. With increasing annealing time, the overall orientation intensity gradually decreases.

3.5. Magnetic Properties

Figure 12 illustrates the evolution of coercivity (H_c) and magnetic induction at 10,000 A/m (B_{10000}) as a function of annealing time during isothermal treatment at 900 °C. With prolonged annealing, H_c exhibits a continuous decreasing trend, dropping from

200 A/m in the as-cold-rolled state to 64.3 A/m after 25 s (complete recrystallization) and further declining to 43 A/m after 180 s (grain coarsening stage). In contrast, B_{10000} shows a moderate yet distinct increase, rising from 1.819 T (cold-rolled) to 1.864 T at 25 s and stabilizing at 1.865 T at 180 s.

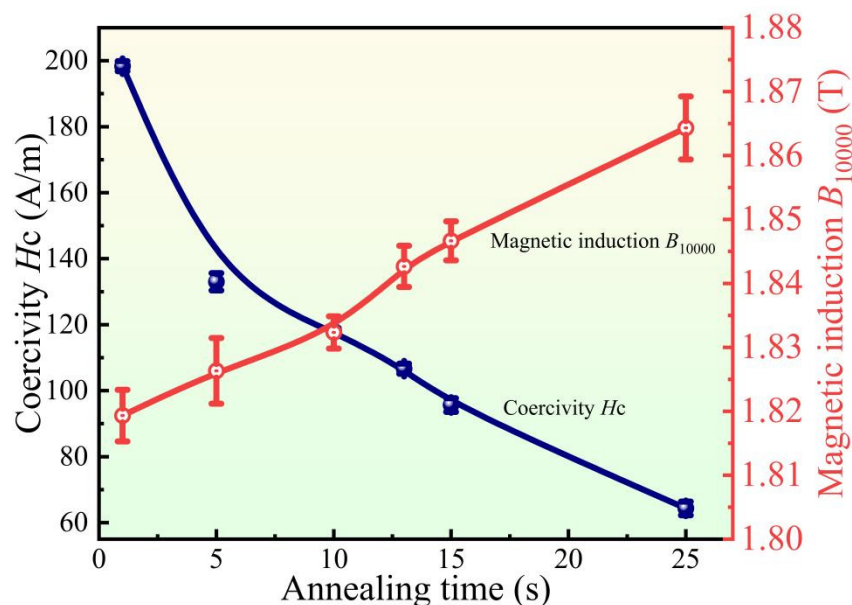


Figure 12. Coercivity (H_c) and magnetic induction (B_{10000}) of 0.5 mm DT4 cold-rolled strip annealing at 900 °C as a function of time.

This correlated variation in magnetic properties is attributed to two synergistic microstructural evolutions inherent to recrystallization and post-recrystallization coarsening: (i) the progressive elimination of dislocations mitigates magnetic domain wall pinning—dislocations act as obstacles to domain wall motion, and their reduction lowers the energy required for magnetization reversal, thereby decreasing H_c . (ii) The formation of a stable γ -fiber ($\langle 111 \rangle // ND$) combined with amplified Goss texture aligns the $\langle 001 \rangle$ easy magnetization axis of α -Fe parallel to the rolling direction (RD). This texture optimization minimizes magnetic anisotropy energy, facilitating domain rotation during magnetization and enhancing magnetic flux density (reflected by increased B_{10000}). Notably, the stability of B_{10000} after 25 s indicates that complete recrystallization is sufficient to achieve near-optimal texture alignment, while the continued decrease in H_c during coarsening is associated with further reductions in the residual lattice distortion and grain boundary.

4. Conclusions

This study systematically elucidates the recrystallization mechanisms of a 0.5 mm thick DT4 electromagnetic pure iron cold-rolled strip (80% reduction) during isothermal annealing at 900 °C.

(1) Microstructure Evolution: The recrystallization initiates at shear bands after 7 s and completes within 25 s, yielding strain-free equiaxed grains (27.5 μm) that coarsen to 64 μm after 180 s, accompanied by a dominant transition from low-angle to high-angle grain boundaries (reaching 90.9% HAGBs).

(2) Texture Evolution Mechanism: The texture evolution is governed by oriented nucleation at shear bands, resulting in a stable γ -fiber component (~20%), a significantly attenuated α -fiber (decreasing from 66.3% to 21.5%), and a notably enhanced Goss texture (increasing from 0.54% to 14.0%) during coarsening, driven by boundary energy minimization.

(3) Recrystallization Kinetics Model: The kinetics strictly follow the JMAK model ($X_{\text{rex}} = 1 - \exp(-2.29 \times 10^{-8} t^{6.434})$) with an unusually high Avrami exponent ($n = 6.434$), attributable to the synergistic effects of shear band-mediated quasi-site saturation, anisotropic growth, and thin-gauge constraints.

(4) Microstructure–Property Correlation: These microstructural transformations directly enhance the soft magnetic properties, evidenced by a 78.5% reduction in coercivity and a 2.5% increase in magnetic induction, while the hardness drops by 61.5%.

(5) Industrial Implications: The established quantitative relationships and the predictive model provide a fundamental basis for optimizing industrial annealing processes to achieve superior magnetic performance in high-sensitivity electromagnetic devices.

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