

Article

Study of the Solidification Microstructure and Deformation Behaviour of Cu₂₀Fe Alloy

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Abstract: In this paper, the solidification microstructure characteristics of metastable immiscible Cu₂₀Fe alloys under natural cooling conditions and subsequent cold rolling were analysed. The findings demonstrate that the Cu₂₀Fe alloy underwent a liquid–solid transformation under natural cooling conditions. The equiaxed Cu matrix and the Fe dendrites exhibited elongation into ribbon-like structures parallel to the cold rolling direction. Following cold rolling, the mean grain size of the Cu₂₀Fe alloy was considerably refined, and the mechanical properties were improved. After cold rolling, the Cu matrix formed both {112}<111> copper and {110}<112> brass textures. Furthermore, the strengthening mechanisms of the cold-rolled Cu₂₀Fe alloy are primarily dependent on the strengthening of grain boundaries and work hardening. This provides an economically friendly method for the preparation of Cu–Fe alloys with high Fe compositions.

Keywords: Cu–Fe alloys; solidification microstructure; deformation behaviour; mechanical properties



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1. Introduction

Cu–Fe alloys have gathered wide attention from academicians in recent years as a novel kind of structural–functional material. Cu–Fe alloys are known for their high electrical conductivity, high thermal conductivity, high strength, as well as unique electromagnetic characteristics, and are widely used in automotive, 5G communications, electronics, and magnetic shielding [1–5]. In comparison to Cu alloys that include Cu–Ag and Cu–Nb alloys, Cu–Fe alloys benefit from abundant Fe raw materials and lower production costs. Cu–Fe alloy is composed mainly of Cu matrix and reinforced Fe particles, and is a representative of immiscible dual-phase alloy. In the phase diagram of Cu–Fe alloys, an immiscible gap is exhibited below the liquid-phase line, which renders the alloy susceptible to liquid-phase separation during the process of solidification [6]. Moreover, the solid solubility between liquid Cu and Fe is very low, and there is a high-density difference, which is prone to macroscopic segregation under its gravity [7]. As a result, the industrial application of Cu–Fe alloys is constrained, particularly for those alloys containing a high proportion of Fe.

The solidification process of Cu–Fe alloys has been studied extensively to produce high Fe component alloys with small Fe grains, evenly distributed, and good properties. Nakagawa et al. [8] first discovered that liquid–liquid separation occurred in the solidification process of Cu–Fe alloy under the condition of undercooling. The influence of undercooling on the solidification microstructure of Cu–Fe alloy has been studied by glass electromelting, levitation, and dropper technology [9,10]. Researchers have examined the influence of additional elements, including Si, Ag, and Mg, on the solidification microstructure and mechanical properties of the Cu–Fe alloys [11–13]. In addition, Cu–Fe alloys have been explored by rapid solidification, mechanical alloying, and laser melting techniques to obtain grain refinement and uniform microstructure [14–16]. However, the microstructure evolution of Cu–Fe alloys solidified under natural cooling conditions remains less explored. It is essential to develop Cu–Fe alloys with a high Fe proportion, uniform microstructure, and no macro-segregation at low cost by appropriate solidification methods.

In this research, we investigated the solidification microstructure of the Cu20Fe alloy under natural cooling conditions using self-designed molds, as well as the cold rolling deformation behaviour and mechanical characteristics. The solidification mechanism, microstructure evolution, and related enhancement mechanism of the alloy were also analysed.

2. Materials and Methods

The raw material for the experiment consisted of copper and iron with a purity of 99.99%, which were melted together in an electromagnetic induction furnace under the protection of an argon atmosphere. The raw materials were subsequently melted and poured directly into the self-designed molds to produce Cu20Fe alloy ingots. Table 1 lists the casting Cu20Fe alloy chemical composition. Figure 1 presents a schematic depiction of the solidification mold, which employs natural cooling as its approach. The ingots underwent hot rolling ($\varnothing 450 \times 450$ Reversing hot flat mill) at 950 °C to decrease the thickness to 5 mm. Following the hot rolling process, the hot-rolled specimens were solution-treated (high-temperature box type resistance furnace, HL07-22, Huizhong Electric Furnace Co., Ltd., Shanghai, China) at a temperature of 1000 °C for 1 h. The specimens underwent cold rolling (400 four-high cold rolling mill for metal sheet) after appropriate surface treatment and ultimately thinned to 0.2 mm thickness through multiple passes.

Table 1. The Cu20Fe alloy chemical compositions.

Cast	Cu [wt.%]	Fe [wt.%]
Cu20Fe	Balance	19.97

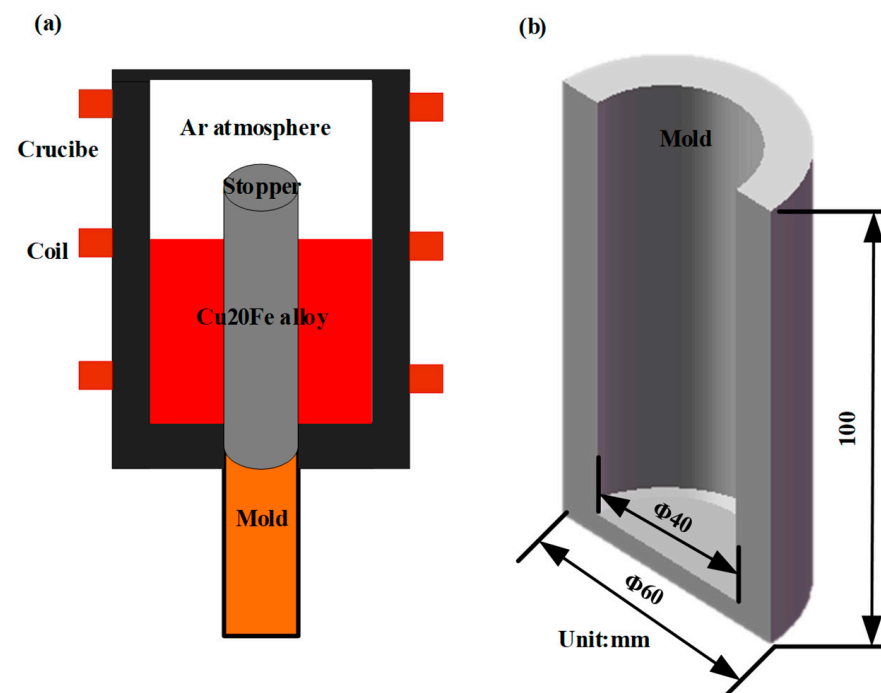


Figure 1. Schematic depiction of the solidification mold: (a) melting devices; (b) solidification mold.

The microstructure analysis was employed by optical microscopy (OM, Olympus BX53M, Leica, Wetzlar, Germany), electron probe microanalyzer (EPMA, JXA-8530F, JEOL, Tokyo, Japan), scanning electron microscopy (SEM, Gemini 300, Carl Zeiss, Oberkochen, Germany), and electron backscatter diffraction (EBSD, Symmetry S3, Oxford instruments, Abingdon, United Kingdom). Before performing microstructure characterization, the surface of each specimen was sanded with 240–2000# sandpaper and then mechanically polished. The Vickers hardness was measured with a hardness tester (MH-500, Everone, Shanghai, China) with a load of 0.2 kg and a dwell time of 10 s. A total of 10 measurements

were taken and the average was taken as the result, the geometry of which is shown in Figure 2a. Room temperature tensile specimens were cut along the parallel rolling direction, and the dimensions of the room temperature tensile specimen are depicted in Figure 2b. A tensile tester (SHIMADZU, Kyoto, Japan) was used to perform tensile tests at room temperature at the tensile rate of 0.5 mm /min, and the test was repeated three times.

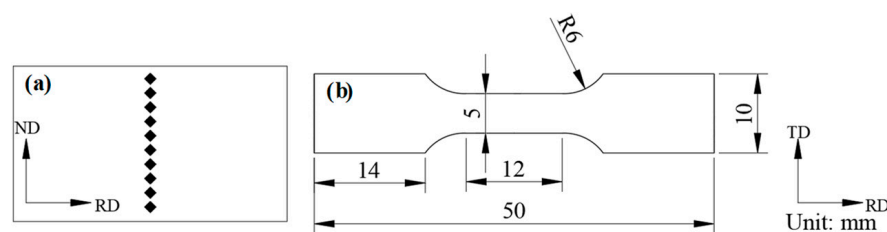


Figure 2. (a) Image of the geometry used in the hardness; (b) dimension diagram of tensile specimen.

3. Results

3.1. Microstructure of As-Cast Cu20Fe Alloy

Figure 3a illustrates the as-cast specimen of the Cu20Fe alloy, while Figure 3b–d depict the microstructure of the corresponding sections in Figure 3a. The darker zones denote the Fe phase, whereas the lighter regions indicate the Cu phase. The microstructure of the as-cast Cu20Fe alloy comprises multiple dendrites composed of directionally aligned Fe particles that are uniformly distributed. No macroscopic segregation of the Fe phase was observed throughout the longitudinal cross-section. Moreover, the Fe dendrites exhibited two distinct characteristics: in the upper part of the ingot, some Fe dendrites formed primary dendrites. At the bottom of the ingot, the Fe phase exhibited broken columnar dendrites. Figure 4 shows a typical two-phase microstructure, which is further confirmed by the distribution maps of Cu and Fe elements.

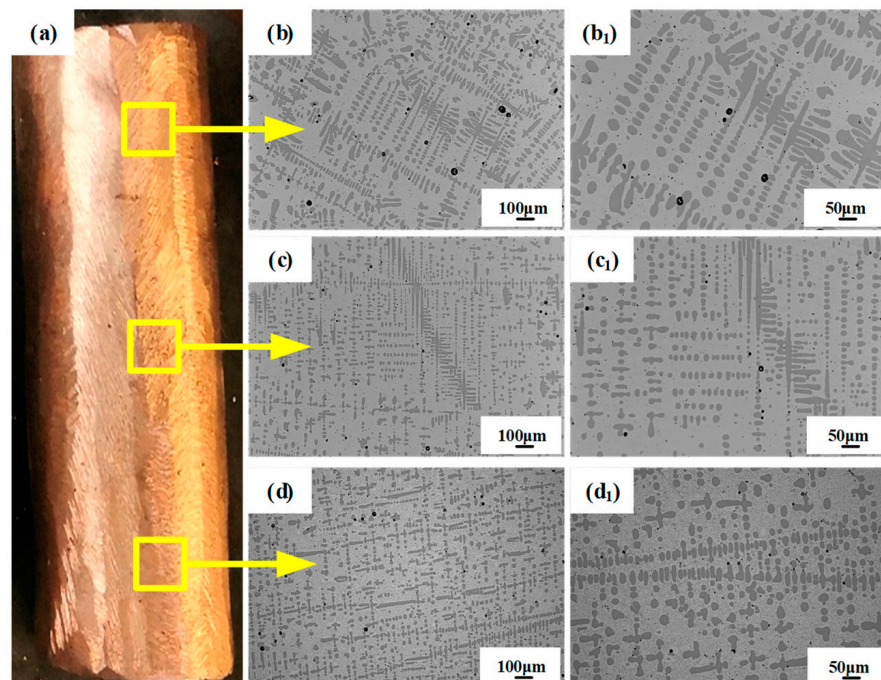


Figure 3. Cu20Fe alloy solidification microstructure: (a) macroscopic specimen of the casting; (b) upper part of the ingot; (c) middle part of the ingot; (d) beneath part of the ingot; (b₁–d₁) are magnified views of (b–d), respectively.

Figure 5 depicts the EBSD analysis results for the as-cast Cu20Fe alloy. Figure 5a illustrates that the blue area in the phase distribution diagram corresponds to the Cu phase,

while the red zone denotes the Fe phase. The Fe grains are distributed evenly inside the Cu matrix with a dendritic structure. The IPF plot in Figure 5b indicates that the Fe phase is predominantly located within the Cu grains, which is a consequence of peritectic transformation occurring during solidification [17]. Further analysis has shown that the Cu matrix is comprised of coarse equiaxed crystals, with a mean grain size measuring $69 \pm 3 \mu\text{m}$. Figure 5c illustrates the polar diagrams for both the Fe phase and the Cu matrix. The upper diagram refers to the Cu matrix, while the lower one relates to the Fe phase. The findings suggest that both Cu and Fe phases have random textures, exhibiting no discernible preferred orientation, with a maximum texture intensity of 26.

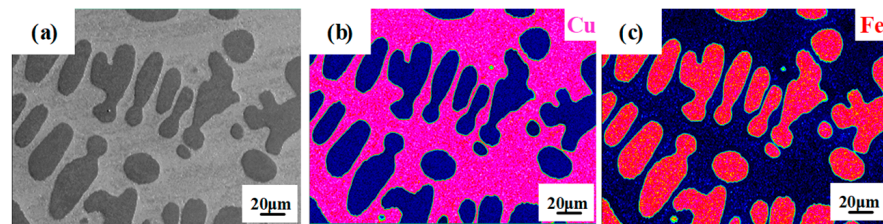


Figure 4. (a) BSE diagram of the as-cast Cu₂₀Fe alloy; (b) Cu elemental mapping; (c) Fe elemental mapping.

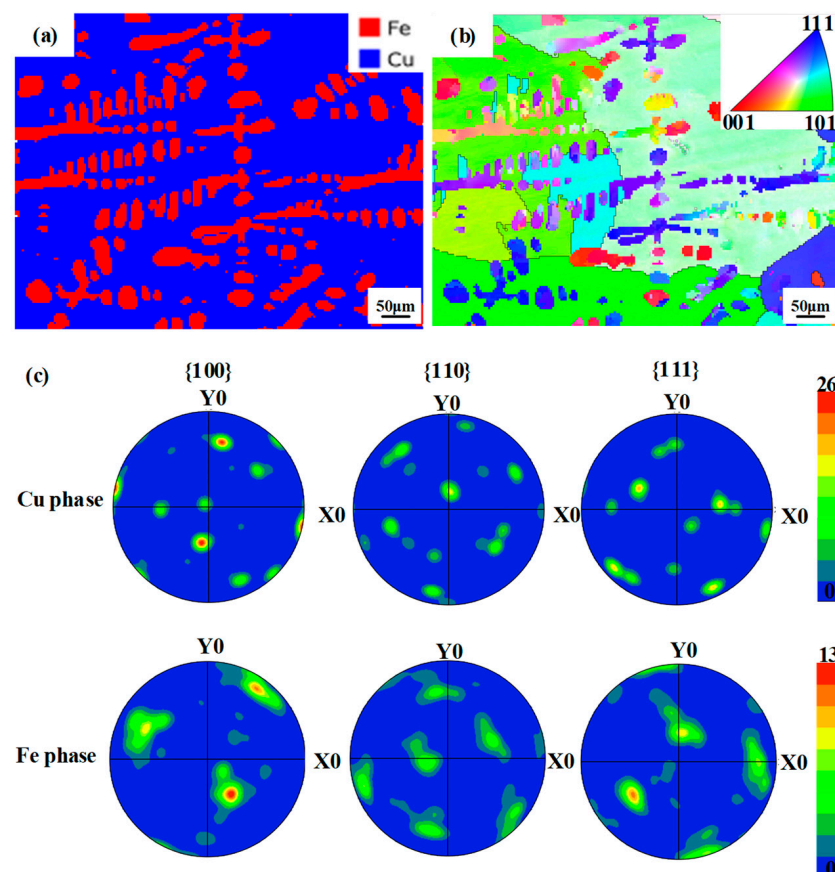


Figure 5. EBSD results for the as-cast Cu₂₀Fe alloy: (a) phase distribution; (b) IPF; (c) {100}, {110} and {111} polar plots for Cu and Fe phases.

3.2. Microstructure of As-Cold Rolled Cu₂₀Fe Alloy

Figure 6 exhibits the BSE image of the as-cold rolled Cu₂₀Fe alloy. As illustrated in the figure, the Fe dendrites are completely fractured after cold rolling and have been stretched into morphologies parallel to the rolling direction, with fibre lengths ranging from 4–90 μm. Figure 6b shows an enlarged view of the red box in Figure 6a. The Fe grains show different

morphologies after rolling, with spindle and tadpole-shaped Fe grains in addition to the elongated Fe grains. In addition, under the influence of coarse Fe grains, the elongated Fe phase around the large grains is bent during plastic deformation. However, the variation in the initial grain sizes of the as-cast Cu20Fe alloy results in differing thicknesses of the Fe fibers after cold rolling, ranging from 0.1 to 5.5 μm .

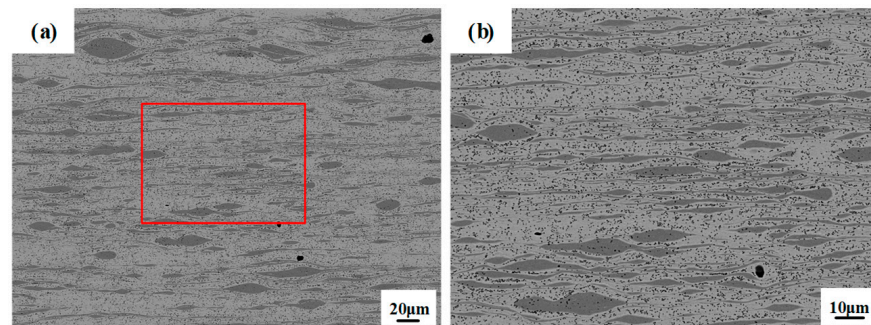


Figure 6. Images of as-cold rolled Cu20Fe alloy: (a) 500 $\times$; (b) 1000 $\times$.

To gain further insight into the microstructure of the Cu20Fe alloy during cold rolling, the specimens were examined via EBSD, and the EBSD results are presented in Figure 7. As shown in Figure 7a, the grain boundary distribution diagram illustrates that the black line corresponds to the low-angle grain boundaries (LAGBs, $<15^\circ$), and the red line refers to the high-angle grain boundaries (HAGBs, $>15^\circ$). A large number of HAGBs are observed at the Cu grain boundary and Cu/Fe phase boundary after cold rolling. Figure 7b depicts the IPF diagram, indicating that the Cu matrix undergoes considerable plastic deformation, and the equiaxed Cu grains are elongated and streamlined in the rolling direction, forming a ribbon. Furthermore, the Cu grains were fragmented to yield ultrafine grains with a mean size of 0.6 μm . According to the Fe IPF graph presented in Figure 7d, the Fe phase also deformed in the rolling direction to form a thin strip after large deformation, and the grain size is significantly refined. Moreover, part of the Fe grain colour changes suggests that these grains rotate under the action of stress to maintain the co-deformation with the Cu matrix.

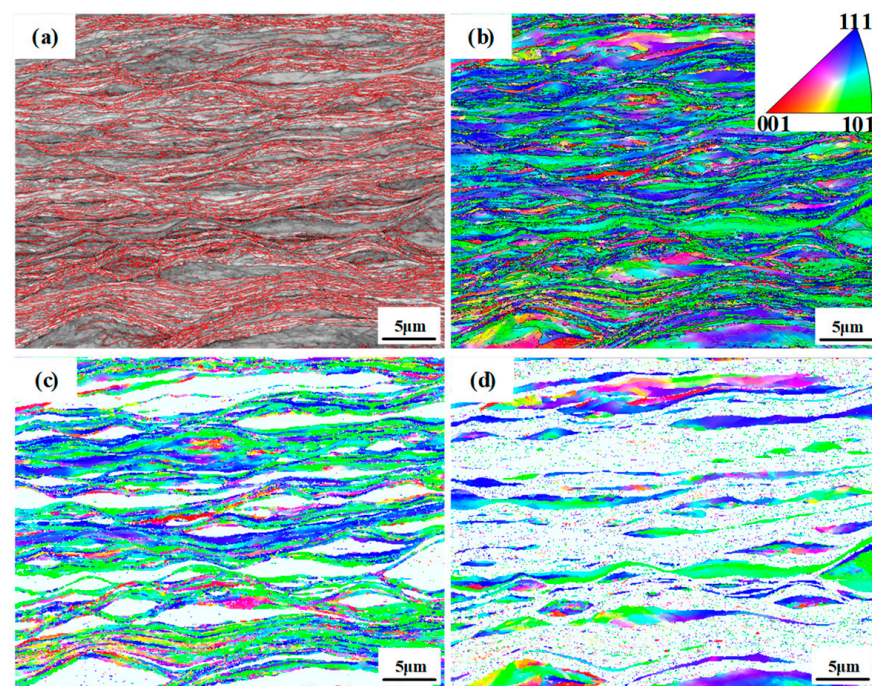


Figure 7. EBSD images of the cold-rolled Cu20Fe alloy: (a) grain boundary distribution; (b) IPF map; (c) Cu phase IPF map; (d) Fe phase IPF map.

The evolution of the texture structure of the Cu20Fe alloy after cold rolling is shown in Figure 8. It can be observed that the Cu grain shows a $\langle 111 \rangle$ orientation, while the Fe grain displays a strong $\langle 110 \rangle$ orientation. Figure 9 shows the texture evolution of the Cu matrix during the rolling process of the Cu20Fe alloy, with section diagrams of orientation distribution function (ODF) for $\varphi_2 = 0^\circ$, 45° , and 90° . The results confirm that the Cu20Fe forms $\{112\}\langle 111 \rangle$ copper texture after cold rolling. During the large deformation cold rolling process, considerable shear stress is generated on the Cu matrix, and the grains turn to the $\{110\}\langle 112 \rangle$ direction, forming brass texture. Additionally, the Cu matrix exhibited a reduction in texture intensity to 9.1 after cold rolling. The copper and brass textures are hard oriented, which increases the strength of the alloy to some extent, but at the same time reduces the plasticity to some extent.

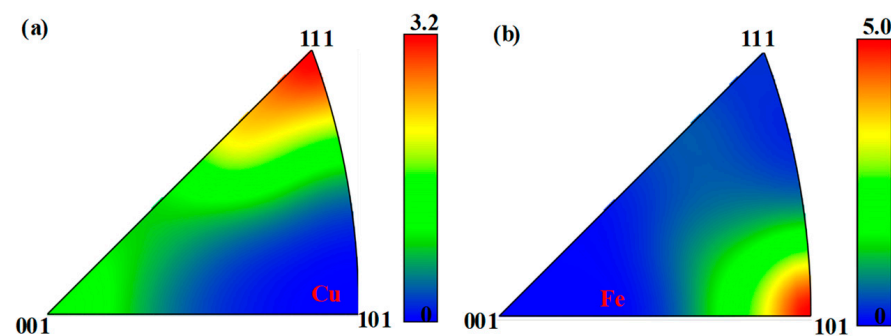


Figure 8. IPF images of cold-rolled Cu20Fe alloy: (a) Cu phase; (b) Fe phase.

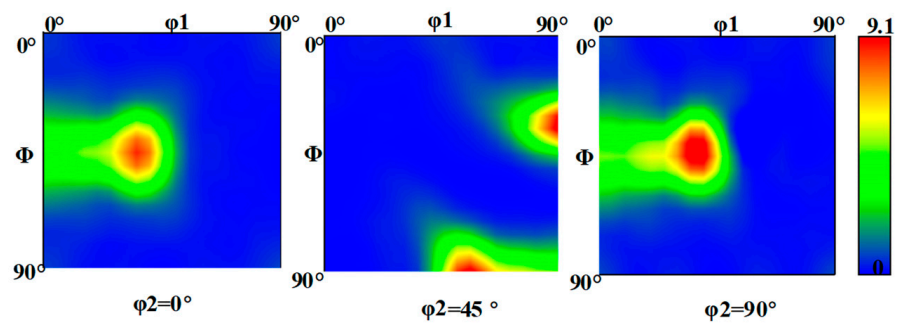


Figure 9. ODF distribution for $\varphi_2 = 0^\circ$, 45° , and 90° sections.

3.3. Mechanical Properties

Figure 10a shows the hardness of the Cu20Fe alloy in cast and cold-rolled conditions, and Figure 10b shows the tensile curve for the cold-rolled Cu20Fe alloy. Table 2 provides a detailed list of the specific mechanical properties of different specimens. The cast Cu20Fe alloy had a hardness of 106.9 HV. Following cold rolling, the Cu20Fe alloy's hardness increased dramatically to 165.6 HV. Furthermore, the tensile strength and elongation of the Cu20Fe alloy were 620 MPa and 4.4% after cold rolling. To elucidate the fracture characteristics of the Cu20Fe alloy, the tensile fracture morphology was observed and analysed using SEM. As shown in Figure 10b, the red box depicts the fracture morphology of the Cu20Fe alloy in the cold rolled condition. The presence of multiple dimples of varied sizes and depths at the fracture site indicates that the alloy exhibits ductile fracture.

Table 2. Summary of mechanical properties of different specimens.

Specimens	Hardness [HV]	Tensile Strength [MPa]	Yield Strength [MPa]	Elongation [%]
Cast	106.9 ± 10.1	-	-	-
Cold rolled	165.6 ± 8.5	620 ± 2.3	562 ± 6	4.4 ± 0.4

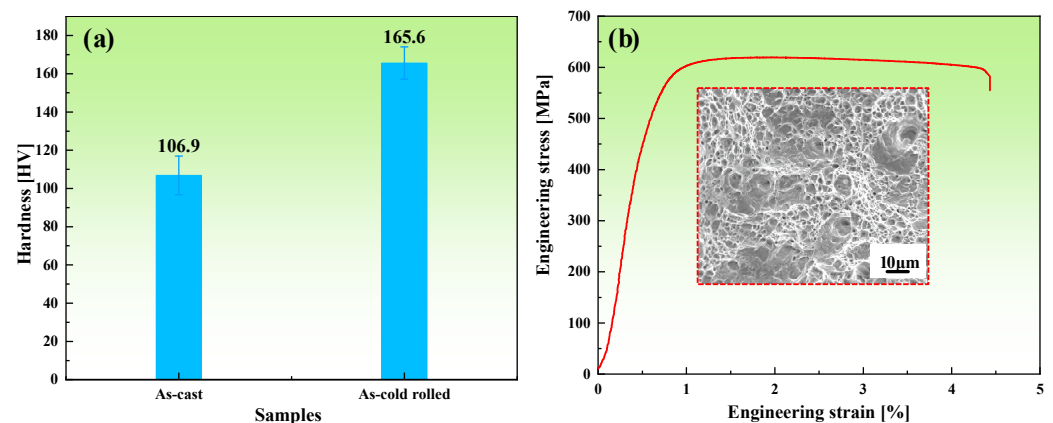


Figure 10. Microhardness and tensile curve of Cu20Fe alloy: (a) microhardness of as-cast and as-cold rolled; (b) tensile curve and fracture morphology of as-cold rolled.

4. Discussion

4.1. Solidification Mechanism of Cu20Fe Alloy

It is well known that the Cu-Fe alloy is recognised as an immiscible alloy, and its phase diagram is depicted in Figure 11. The presence of a miscible gap beneath the liquid phase line renders the alloy susceptible to liquid phase separation during the solidification process. Critical undercooling (ΔT_M) refers to the region between the liquid-phase line and the miscibility gap. When the ΔT_M is greater than actual undercooling, the alloy solidifies along the liquidus line instead of the miscibility gap, resulting in liquid–solid transformation. On the contrary, the liquid phase of the alloy will be transformed into two distinct phases: Cu-rich and Fe-rich.

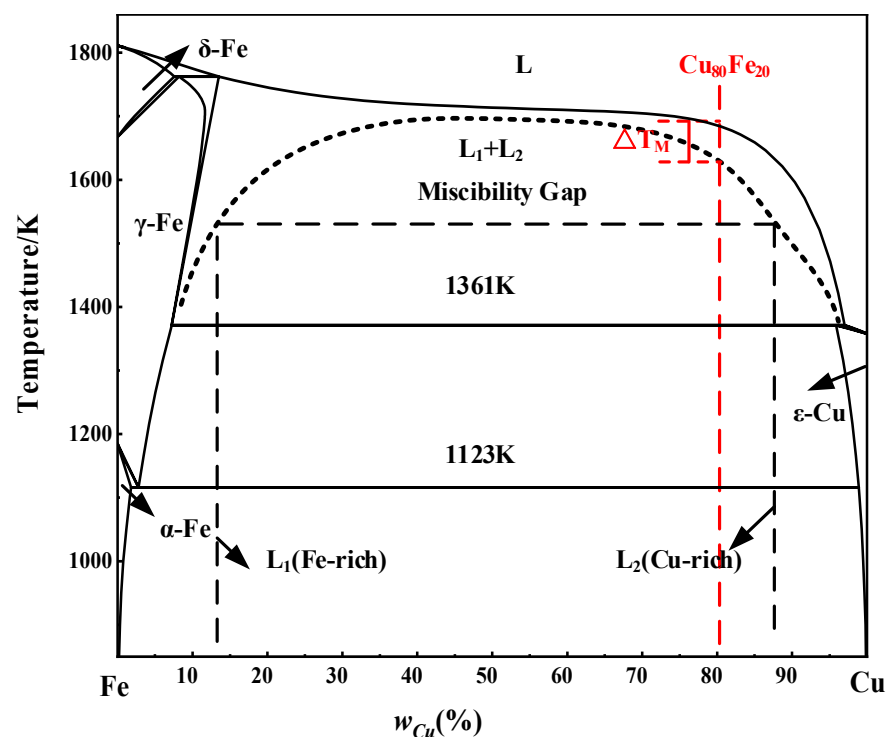


Figure 11. Cu-Fe alloy phase diagram.

For alloys cooled naturally in the mold, the cooling rate is relatively slow, about 50–100 K/s, resulting in a small actual subcooling, which makes it challenging to achieve the required ΔT_M (about 50 K for Cu20Fe alloy [18]). This indicates that the Cu20Fe al-

loy exhibits a liquid–solid transformation during the solidification process, resulting in the generation of γ -Fe dendrites as the predominant microstructure (in Figure 3). It can be determined that the Cu20Fe alloy experiences very little undercooling during natural cooling, and the solidification did not enter the miscibility gap. Consequently, there is no liquid phase separation; instead, the γ -Fe dendrites nucleate and grow immediately from the undercooled melt. According to the LKT/BCT dendrite growth theory [19,20], the relationship among the dendritic growth rate V , dendrite tip radius R , and bulk undercooling ΔT is defined by the following equations:

$$\Delta T = \Delta T_t + \Delta T_c + \Delta T_r + \Delta T_k \quad (1)$$

$$R = \frac{\Gamma / \sigma^*}{(\Delta H / C_{PL}) P_t \xi_t - \{2 m'_1 C_0 (1 - k_v) P_c / [1 - (1 - k_v) I_v(P_c)]\} \xi_c} \quad (2)$$

$$\Delta T_c = m_1 C_0 \left[1 - \frac{m'_1 / m_1}{1 - (1 - k_v) I_v(P_c)} \right] \quad (3)$$

$$I_v(P_c) = P_c \exp(P_c) E_1(P_c) \quad (4)$$

where ΔT_t is the thermal undercooling, ΔT_c is the solutal undercooling, ΔT_r is the curvature undercooling, and ΔT_k is the kinetic undercooling, Γ is the Gibbs-Thomson coefficient, ΔH is the melting enthalpy, σ^* the stability constant, C_{PL} is the liquid phase specific heat, P_t and P_c are the thermal and solute Peclet number, m'_1 the actual liquidus slope, m_1 the liquid phase slope, C_0 the initial alloy composition, k_v the actual solute partition coefficient, ξ_t and ξ_c the stability parameters, E_1 the exponential integral function, and I_v the Ivantsov function. Given that the solidification process approaches equilibrium, the component undercooling ΔT_c is around 42.6 K [6], which has a significant impact on the formation of dendrites. This shows that solute diffusion has a substantial impact on the growth of Fe dendrites during solidification, and is less affected by the undercooling of the other three parts. As can be seen in Figures 3–5, the dendritic microstructure is mostly segregated, showing that the same dendrite backbone consists of multiple Fe particles orientated. In certain regions, the Fe dendrite backbone exhibits pronounced branching, ascribed to the concentration gradient disparity induced by solute accumulation near the dendrite's tip during its growing. During solidification, substantial components are undercooled in localised regions, resulting in the formation of primary dendrites.

4.2. Cold Rolling Deformation Behaviour of Cu20Fe Alloy

After cold rolling, both the Cu and Fe grains of the Cu20Fe alloy are stretched into ribbon-like structures along the rolling direction. Due to a distinction in modulus of elasticity and yield strength between the two phases, the Cu phase experiences a greater extent of plastic deformation than the Fe phase under identical deformation conditions, leading to the fragmentation and fracture of Cu grains, generating a multitude of ultrafine grains. Furthermore, the Fe phase exhibits a high yield strength, and its diffusely distributed Fe particles within the Cu matrix impede the movement of dislocations. The dislocations first start to accumulate gradually within the Cu matrix. As strain needs to be continuous at the Cu/Fe non-uniform phase interface, geometrically necessary dislocations (GNDs) will be produced to fit the strain gradient [21], as illustrated in the KAM image in Figure 12b. During rolling deformation, the grain direction of the Fe phase shifts to the $\langle 101 \rangle$ orientation. Grain boundaries impede dislocation migration to some degree, resulting in dislocation accumulation at the grain boundaries and the generation of localised high-stress concentrations [22]. The Fe phase will plastically deform when the local stress concentration exceeds the yield strength of the Fe. As illustrated in Figure 12d, the interior of Fe grains exhibits predominantly bright green, reflecting the presence of numerous dislocations within the Fe grains. The dendritic Fe grains are entirely fractured, resulting in banded and spindle-shaped morphologies. Some of the elongated Fe bands undergo 'necking' fracture

to coordinate the strain compatibility of the two phases, which can effectively inhibit the residual stress and crack formation. Simultaneously, the Cu matrix fills in the fracture region of Fe grains. The microstructure of cold-rolled Fe grains is closely related to the initial Fe grain size and morphology. The microstructure of the Fe phase after cold rolling is closely related to the initial microstructure of the Fe grains. With regard to the isolated Fe dendrites, they are predominantly in the shape of short rods, which are subjected to the highest stress during the rolling process and are most prone to plastic deformation in a band-like distribution. It is noteworthy that the smaller the initial Fe grain, the more slender the Fe strips after cold rolling. In contrast, for the complete dendrite, the dendrite trunk is thicker and coarser in grain size, and the shear stress required for plastic deformation is greater relative to that of short rod-shaped dendrites, which perform as spindle-shaped after rolling.

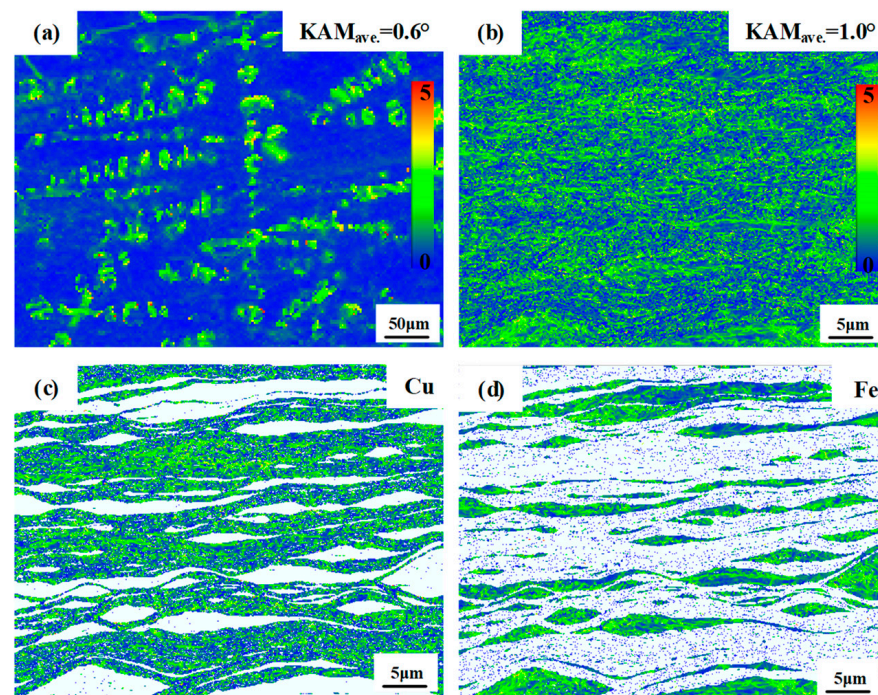


Figure 12. KAM plots of Cu20Fe alloy; (a) as-cast; (b) cold rolled; (c) the KAM plot of the Cu phase in (b); (d) the KAM plot of the Fe phase in (b).

4.3. Mechanism for Strengthening

The above analysis of the microstructure and mechanical properties of the Cu20Fe alloy demonstrated that grain refinement and dislocation accumulation occur during cold rolling. Consequently, it can be inferred that the primary mechanisms responsible for the enhancement of alloy strength are grain boundary strengthening and dislocation strengthening. As can be seen in Figure 13a,b, the grain size of the Cu20Fe alloy was significantly refined after cold rolling, decreasing from the initial mean grain size of 29 μm as-cast to 0.65 μm . In addition, during the cold rolling process, grain rotation, shear fragmentation, and dislocation accumulation at grain boundaries gradually evolve into fine crystals with high-energy grain boundaries, with the proportion of HAGBs increasing from 32% to 58%. Therefore, grain boundary strengthening plays a critical role in enhancing the mechanical properties of cold-rolled Cu20Fe alloys. In Figure 12a,b, the KAM value of the alloy rises from the initial 0.6° to 1.0° after cold rolling. Figure 13c,d indicate that the GNDs density of the Cu matrix increased from $0.6 \times 10^{14} \text{ m}^{-2}$ in the as-cast to $20.35 \times 10^{14} \text{ m}^{-2}$ after cold rolling, indicating that substantial accumulation of dislocation density within the matrix after cold rolling leads to a pronounced work-hardening effect that improved the material's strength while decreasing its plasticity.

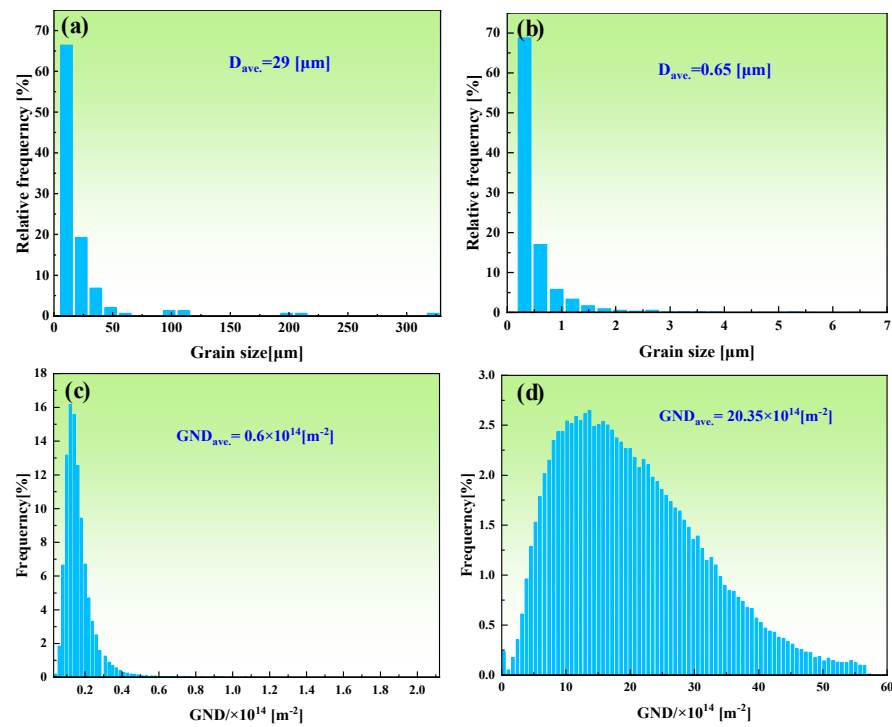


Figure 13. (a,c) are the grain sizes and GND distribution of the as-cast; (b,d) are the grain sizes and GND distribution of the as-cold rolled.

The reinforcement mechanism bypassing non-deformed grains is expressed by the Orowan–Ashby formula [23]:

$$\Delta\sigma_{\text{Orowan}} = 0.81 \frac{MGb}{2\pi(1-\nu)^{\frac{1}{2}}} \cdot \frac{\ln\left(\sqrt{\frac{2}{3}} \cdot \frac{d_p}{b}\right)}{\lambda_p} \quad (5)$$

where M is the Taylor factor, G is the shear modulus, b is the Burgers vector of the Cu, ν is the Poisson's ratio, and d_p and λ_p refer to the Fe grain mean size and spacing. Table 2 summarises the relevant parameters and results from the yield strength calculations. The calculation indicates that the yield strength enhanced by Orowan is 38 MPa.

The Hall–Petch formula can evaluate the relationship between grain boundaries and alloy yield strength [24].

$$\Delta\sigma_{\text{GB}} = k \cdot d^{-\frac{1}{2}} \quad (6)$$

where k is a constant and d is the mean grain diameter of the Cu matrix (estimated from EBSD results). The increased value of the grain boundary effect on strengthening is 181 MPa.

The Taylor formula assesses the influence of work hardening owing to dislocations on yield strength [25]:

$$\Delta\sigma_{\text{Work}} = M\alpha Gb\rho^{\frac{1}{2}} \quad (7)$$

where α represents the geometrical constant, while ρ denotes the dislocation density (calculated by the Williamson–Hall method). The calculation results show that work hardening increases the yield strength by 291 MPa. Given that the solid solubility between Fe and Cu is small [26], the effect of solid solution strengthening is not considered in this analysis. Therefore, the overall yield strength of the cold-rolled Cu20Fe alloy can be determined through the following formula:

$$\sigma_{\text{Total}} = \sigma_0 + \Delta\sigma_{\text{Orowan}} + \Delta\sigma_{\text{GB}} + \Delta\sigma_{\text{Work}} \quad (8)$$

where σ_0 represents the intrinsic lattice strength, approximated at 60 MPa [27]. Table 3 provides a comprehensive overview of the calculation parameters for yield strength and the corresponding results. To provide a clearer illustration of the role of each strengthening mechanism, both experimental and calculated values of yield strength are presented in Figure 14. After cold rolling, the Cu20Fe alloy exhibits an experimental yield strength of 562 MPa, while the estimated yield strength is 570 MPa, indicating a strong correlation between the experimental and calculated values. It is evident that work hardening serves as the predominant enhancing mechanism for the Cu20Fe alloy, contributing 51.1% to the overall yield strength.

Table 3. Parameters and corresponding results for calculating yield strength.

Parameter	Value	Units	Ref.
M	3.08	-	[23]
G	46	GPa	[23]
b	0.256	nm	[23]
α	0.26	-	[23]
k	0.14	MPa·m	[23]
ν	0.34	-	[23]
λ_p	1.2	μm	This study
d_p	0.9	μm	This study
d	0.6	μm	This study
$\Delta\sigma_{\text{Orowan}}$	38	MPa	Calculated
σ_0	60	MPa	[27]
$\Delta\sigma_{\text{GB}}$	181	MPa	Calculated
ρ	9.8×10^{14}	m^{-2}	This study
$\Delta\sigma_{\text{Work}}$	291	MPa	Calculated
σ_{Total}	570	MPa	Calculated
$\Delta\sigma_{\text{Exp.}}$	562	MPa	This study

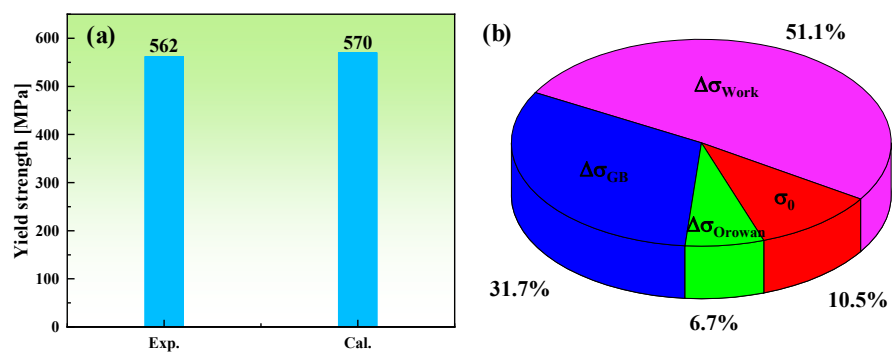


Figure 14. (a) A comparison between calculation and experimental results; (b) the percentage of different reinforcement mechanisms.

5. Conclusions

In this research, we successfully fabricated Cu20Fe alloy ingots using a self-designed mold. The solidification mechanism and microstructural properties of the alloy under natural cooling conditions were investigated. Subsequently, the Cu20Fe alloy ingot was subjected to cold rolling, after which its deformation behaviour and mechanical characteristics were investigated, yielding the following main conclusions:

1. The Cu20Fe alloy undergoes a liquid–solid transition under natural cooling conditions, forming a dendritic microstructure consisting of multiple Fe particles arranged in an oriented manner without liquid phase separation. It provides a new low-cost method for the industrial preparation of Cu–Fe alloys with high Fe compositions.
2. After the cold rolling process, the equiaxed Cu grains and Fe dendrites were extended into ribbon-like structures in the rolling direction, with significant grain refinement,

and the orientation of the Cu and Fe grains shifted from the initial disorder to the $\langle 111 \rangle$ and $\langle 110 \rangle$ directions, respectively.

3. The Cu20Fe alloys formed both $\{112\}\langle 111 \rangle$ copper and $\{110\}\langle 112 \rangle$ brass textures after cold rolling.
4. The mechanical properties of Cu20Fe alloys were markedly enhanced after cold rolling. The enhancement of mechanical properties is primarily ascribed to grain boundary strengthening and work-hardening effects, with work-hardening playing a significant role in the strengthening mechanism.

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