Effect of Magnesium on Microstructure Refinements and Properties Enhancements in High-Strength CuNiSi Alloys

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Abstract

Microalloying is an effective method to improve the comprehensive properties of copper alloys. The effects of magnesium on the microstructure, mechanical properties and anti-stress relaxation properties of CuNiSi alloys have been investigated. Results demonstrated that magnesium plays significant roles in refining the dendritic microstructure of the as-cast ingot, accelerating the precipitation decomposition, improving the mechanical properties and increasing the anti-stress relaxation properties. The incremental strength increase is due to the Orowan strengthening from the nanoscale Ni₂Si and Ni₃Al precipitates. As compared with the Cu–6.0Ni–1.0Si–0.5Al (wt%) alloy, the ultimate tensile strength of the designed Cu–6.0Ni–1.0Si–0.5Al–0.15Mg (wt%) alloy increases from 983.9 to 1095.7 MPa, and the electrical conductivity decreases from 27.1 to 26.6% IACS, respectively. The stress relaxation rates of the designed Cu–6.0Ni–1.0Si–0.5Al–0.15Mg alloy are 4.05% at 25 °C, 6.62% at 100 °C and 9.74% at 200 °C after having been loaded for 100 h, respectively. Magnesium significantly promotes nucleation during precipitation and maintains small precipitate size.

Keywords Copper alloy · Precipitation · Aging · Mg addition · Strength · Stress relaxation

1 Introduction

Elastic conductive materials increasingly require high strength and low-stress relaxation rate due to the rapid development in the automation [1], communication [2], electrical and electronics industries [3]. CuBe alloys have been used to manufacture elastic conductive products because of their ultrahigh strength and high electrical conductivity [4]. However, the high-temperature properties of CuBe alloys are limited, i.e., the stress relaxation rate of Cu–1.7Be alloy is larger than 25% as served at 150 °C for 100 h [5]. Therefore, researchers are trying to develop new copper alloys with high strength, low electrical resistance and reduced

stress relaxation rate. CuTi [6], CuNiSn [7], CuNiZn [8] and CuNiAl [9] system alloys have been reported recently, which exhibit ultrahigh ultimate tensile strength (> 1000 MPa) and excellent stress relaxation resistance (relaxation rates are less than 10% at room temperature); however, electrical conductivities of the above alloys (< 12% IACS) are much less than that of the CuBe alloy (~ 25% IACS). Fortunately, CuNiSi system alloys with high strength and excellent electrical conductivity have attracted much attention recently [10]. Meanwhile, CuNiSi alloys with high Ni and Si concentrations have high ultimate tensile strength (> 1000 MPa) and yield strength (> 800 MPa), which are potential candidates for replacing the CuBe alloys [11–13].

Effects of Sn [14], Al [15], Cr [16], Zn [17], P [18], V [19], Co [20, 21], Fe [22], Ag [23] and Mg [24] on the structural refinements and property enhancements of CuNiSi alloys have been reported, revealing that the addition of above microalloying elements effectively improved CuNiSi alloys' physical properties. Ni₃Si and Ni₃Al are precipitated from the copper matrix during thermal treatment [11, 15], and the addition of Cr could arouse a significant precipitation strengthening effect by forming ordered f.c.c. (Ni, Cr, Si)-rich phase precipitates [16]. P refines the grain size, increases the nucleation rate of the precipitates and prevents

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the coarsening of the precipitated particles. Thus, the CuNiSiP and CuNiSiZn alloys show an effective improvement in the mechanical properties [17, 18], and Zn also improves the welding property [17]. Fe is an element that could improve both the strength and electrical conductivity [20, 22]. Ag can refine the grain and optimize the hot workability of the CuNiSi alloys [23]. V accelerates the precipitation during aging and contributes positively to the resultant hardness and electrical conductivity [19]. Co provides additional strength by forming Co_2Si precipitates [20]. Mg decreases the stress relaxation of the Cu–Ni–Si alloys [24]. The CuNiSiMg alloys with low Ni and Si concentrations have been optimized to be high electrical conductive medium strength copper alloys. However, the investigation about the effect of Mg on microstructure, properties and stress relaxation performance in the ultrahigh-strength conductive elastic CuNiSi alloys is limited.

In the present work, CuNiSi alloys with and without Mg were synthesized and investigated. Vickers hardness, electrical conductivity and stress relaxation rates were measured to explore the variations of physical properties of the fabricated alloys. Finally, the effects of Mg on the microstructural evolution, stress relaxation mechanism and strengthening mechanism are also discussed.

2 Experimental

Table 1 lists the nominal compositions (wt%) of the studied CuNiSi alloys. Pure Cu, pure Ni, pure Si, Cu–10 wt% Mg master alloy and pure Al were used as raw materials. Cast ingots with dimensions of 120 mm (in length) $\times$ 65 mm (in width) $\times$ 35 mm (in thickness) were fabricated in a steel mold. After removing the surface oxide defects, the ingots were homogenized at 940 °C for 4 h and subsequently hot-rolled at 900 °C by 80% to reduce the thickness from 30 mm to 6 mm. The resultant strips were solid solution treated at 970 °C for 4 h. Then the strips were cold-rolled by 50% (3 mm in thickness) and finally aged at 450 °C for various durations.

Vickers hardness values were measured with a load of 3 kg and a dwelling time of 15 s, and the values were averaged from at least five measurements in each sample. The electrical conductivities were measured at 20 °C, taking the average

value of five measurements. Tensile tests were performed on an MTS 810 tester machine with a constant strain rate of 10^{-3} s^{-1} . Stress relaxation experiments were performed on a self-made device according to the standard of ASTM E328-02(2008). The stress relaxation homemade device and the stress relaxation measurement method were reported in previous work [25]. Metallurgical images were captured by a Leica DM ILM HC optical microscope. Microstructures of thin foil samples were observed in a JEOL JEM 2100F transmission electron microscopy (TEM). Thin foils for TEM observations were prepared by a standard electropolishing method and then were finally processed by argon ion milling.

3 Results

3.1 Microstructural Evolution During Thermomechanical Treatments

Figure 1 shows the microstructure of as-cast ingots of studied alloys. Interdendritic secondary-arm space of the as-cast Cu–8.0Ni–1.8Si alloy is $152 \pm 8.2 \text{ }\mu\text{m}$ (Fig. 1a), while those of Cu–8.0Ni–1.8Si–0.15Mg alloy is $82 \pm 5.7 \text{ }\mu\text{m}$ (Fig. 1b). In the meantime, those of Cu–6.0Ni–1.0Si–0.5Al alloy and Cu–6.0Ni–1.0Si–0.5Al–0.15Mg alloy are $98 \pm 7.2 \text{ }\mu\text{m}$ (Fig. 1c) and $64 \pm 6.1 \text{ }\mu\text{m}$ (Fig. 1d), respectively. As seen in Fig. 1, Mg refined the dendritic microstructure in as-cast ingots. The fine dendrites indicated that the segregation degree in the specimens is less than that of ones with coarse dendrites. Some non-equilibrium compounds appeared in the specimens, which have been determined to be Ni_2Si and Ni_3Si [15]. After homogenization treatment at 940 °C for 4 h, studied specimens were hot-rolled by 80% at 900 °C. Then they were solid solution treated at 970 °C for 4 h, supersaturated solid solution (SSS) strips formed and their microstructures are shown in Fig. 2. The non-equilibrium compounds were re-dissolved into the matrix during solution treatment. Then the grain size varied with the chemical composition of the studied alloys. After the same solid solution treatment, the average grain sizes of the Cu–8.0Ni–1.8Si and Cu–8.0Ni–1.8Si–0.15Mg alloys are $64 \pm 12.2 \text{ }\mu\text{m}$ (Fig. 2a) and $46 \pm 8.1 \text{ }\mu\text{m}$ (Fig. 2b), respectively. Those of the Cu–6.0Ni–1.0Si–0.5Al alloy and Cu–6.0Ni–1.0Si–0.5Al–0.15Mg alloy are $82 \pm 8.5 \text{ }\mu\text{m}$ (Fig. 2c) and $78 \pm 9.4 \text{ }\mu\text{m}$ (Fig. 2d), respectively. Therefore, Mg effectively refines the dendritic microstructure of CuNiSi and CuNiSiAl alloys ingots.

3.2 Variations of Microhardness and Electrical Conductivity During Aging

Figure 3 shows the hardness and electrical conductivity variations of the SSS samples that have been cold-rolled by 50% then aged at 450 °C for various durations. All the

Table 1 Chemical composition of the studied alloys (wt%)

Alloys	Ni	Si	Mg	Al	Cu
Cu–8.0Ni–1.8Si	8.0	1.8	–	–	Bal.
Cu–8.0Ni–1.8Si–0.15Mg	8.0	1.8	0.15	–	Bal.
Cu–6.0Ni–1.0Si–0.5Al	6.0	1.0	–	0.5	Bal.
Cu–6.0Ni–1.0Si–0.5Al–0.15Mg	6.0	1.0	0.15	0.5	Bal.

Fig. 1 Optical micrographs of the dendritic microstructure in the as-cast studied alloys. **a** Cu-8.0Ni-1.8Si; **b** Cu-8.0Ni-1.8Si-0.15Mg; **c** Cu-6.0Ni-1.0Si-0.5Al; **d** Cu-6.0Ni-1.0Si-0.5Al-0.15Mg

curves show that the hardness of the samples reached their peak values and then decreased. Precipitation strengthening plays a positive role in increasing the hardness; some nanoscale precipitates block the dislocations motion [26]. Peak values of the hardness of all these studied alloys appear at the same aging time of 60 min. For these studied alloys, peak hardness values of Cu-8.0Ni-1.8Si and Cu-6.0Ni-1.0Si-0.5Al alloys are 330.3HV and 324.9 HV, respectively, while those of Cu-8.0Ni-1.8Si-0.15Mg and Cu-6.0Ni-1.0Si-0.5Al-0.15Mg alloys are 340.4 HV and 345.2 HV, respectively (Fig. 3a). Correspondingly, their electrical conductivities are 27.1%IACS (Cu-8.0Ni-1.8Si), 24.2%IACS (Cu-8.0Ni-1.8Si-0.15Mg), 27.1%IACS (Cu-6.0Ni-1.0Si-0.5Al), and 26.6%IACS (Cu-6.0Ni-1.0Si-0.5Al-0.15Mg), respectively (Fig. 3b).

3.3 Mechanical Properties

Seen in Fig. 3, the studied alloys achieve peak hardness after being aged at 450 °C for 60 min. The mechanical properties of the samples were measured by tensile testing. Figure 4 presents the curves tensile tests and the fractures

of the studied alloys. Table 2 illustrates the physical properties of the studied alloys. From the true stress–true strain curves of Cu-8.0Ni-1.8Si and Cu-8.0Ni-1.8–Mg alloys, Cu-8.0Ni-1.8Si and Cu-8.0Ni-1.8–Mg alloys show a tensile strength (ultimate tensile strength) of 1072.5 MPa and 1110.7 MPa (Fig. 4a), respectively. Considering the Cu-6.0Ni-1.0Si-0.5Al alloy and the Cu-6.0Ni-1.0Si-0.5Al-0.15Mg alloy, their tensile strengths were 983.9 MPa and 1095.7 MPa (Fig. 4d). The precipitates in the studied alloys are unyielding and have been broken away from the matrix grains, which led to cracks propagation [27]. Fractures show that quasi-cleavage fracture is the dominated fracture mechanism in the studied alloys (Fig. 4), and some shallow dimples form on the fractures of the tensile specimens (Fig. 4). In Fig. 4, fractures in the Cu-8.0Ni-1.8Si-0.15Mg and the Cu-6.0Ni-1.0Si-0.5Al-0.15Mg tensile specimens showed smaller dimples than those in Cu-8.0Ni-1.8Si alloys and the Cu-6.0Ni-1.0Si-0.5Al alloys. Mg has refined the dimple size, reduced fracture toughness and maintained a quasi-cleavage fracture mechanism.

Fig. 2 Optical micrographs of the solid solution-treated studied alloys. **a** Cu-8.0Ni-1.8Si; **b** Cu-8.0Ni-1.8Si-0.15Mg; **c** Cu-6.0Ni-1.0Si-0.5Al; **d** Cu-6.0Ni-1.0Si-0.5Al-0.15Mg

Fig. 3 Properties of the solid solution-treated studied alloys aged at 450 °C for various durations. **a** Vickers hardness; **b** electrical conductivity

3.4 Precipitation Behaviors

Bright-field images of the Cu-8.0Ni-1.8Si alloy sample aged at 450 °C for 15 min, 60 min and 300 min are shown in Fig. 5a–c, indicating that the size of the precipitates is increased with the aging time, with a size of 10 nm to 50 nm.

Bright-field images in Fig. 5d–f indicate that precipitates in the Cu-8.0Ni-1.8Si-0.15Mg alloy are smaller than those in the Cu-8.0Ni-1.8Si alloy after the same aging treatment. Microstructures of the Cu-6.0Ni-1.0Si-0.5Al alloy and the Cu-6.0Ni-1.0Si-0.5Al-0.15Mg alloy aged at 450 °C for various durations are shown in Fig. 6. After aged at 450 °C

Fig. 4 Stress–strain curves and fractures of the studied alloys. **a** Stress–strain curves of Cu–8.0Ni–1.8Si and Cu–8.0Ni–1.8Si–0.15Mg alloys; **b** fracture of Cu–8.0Ni–1.8Si alloy tensile tested sample; **c** fracture of Cu–8.0Ni–1.8Si–0.15Mg alloy tensile tested sample; **d** stress–strain curves of Cu–6.0Ni–1.0Si–0.5Al and Cu–6.0Ni–1.0Si–0.5Al–0.15Mg alloys; **e** fracture of Cu–6.0Ni–1.0Si–0.5Al alloy tensile tested sample; **f** fracture of Cu–6.0Ni–1.0Si–0.5Al–0.15Mg alloy tensile tested sample

Table 2 Physical properties of the studied alloys

Alloys	Hardness, (HV)	Tensile strength, σ_b (MPa)	Yield strength, $\sigma_{0.2}$ (MPa)	Elongation percentage, δ (%)	Electrical conductivity, g (%IACS)	Strain relaxation rate after 100 h, R (%)		
Cu–8.0Ni–1.8Si	330.3	1075.2	905.3	1.82	27.1	10.01 (25 °C)	20.53 (100 °C)	43.19 (200 °C)
Cu–8.0Ni–1.8Si–0.15Mg	340.4	1110.7	940.1	2.41	24.2	5.27	9.49	20.04
Cu–6.0Ni–1.0Si–0.5Al	324.9	983.9	920.5	2.30	27.1	5.08	8.89	11.89
Cu–6.0Ni–1.0Si–0.5Al–0.15Mg	345.2	1095.7	935.2	2.29	26.6	4.05	6.62	9.74

for 5 min, precipitates are detected in the Cu–6.0Ni–1.0Si–0.5Al alloy and Cu–6.0Ni–1.0Si–0.5Al–0.15Mg alloy (Fig. 6a, b). In Fig. 6b, c, e, and f, there are more and finer precipitates in the Cu–6.0Ni–1.0Si–0.5Al–0.15Mg alloy than in the Cu–6.0Ni–1.0Si–0.5Al alloy. The precipitates in the studied alloys have been determined to be δ -Ni₂Si and β -Ni₃Si in previous investigations [29–30]. The δ -Ni₂Si precipitates had six variations in respective perpendicular directions.

Both δ -Ni₂Si and variations had an orthorhombic structure with the same structure parameters: $a = 0.706$ nm, $b = 0.499$ nm and $c = 0.372$ nm [31]. The β -Ni₃Si phase had a primitive cubic structure, $a = 0.3351$ nm [30]. Crystal orientation relationships between the Cu and precipitates (β -Ni₃Si, δ -Ni₂Si and δ' -Ni₂Si) have been determined as that [15, 25]: $(02\bar{2})\text{Cu} \parallel (01\bar{1})\beta \parallel (010)\delta$, $[100]\text{Cu} \parallel [100]\beta \parallel [001]\delta$; $(02\bar{2})\text{Cu} \parallel (01\bar{1})\beta \parallel (100)\delta'$, $[100]\text{Cu} \parallel [100]\beta \parallel [001]\delta'$.

Fig. 5 Bright-field micrographs of microstructural evolutions in Cu-8.0Ni-1.8Si-(0.15Mg) alloys having been aged at 450 °C for various durations. Cu-8.0Ni-1.8Si alloy: **a** 15 min; **b** 60 min; **c** 300 min. Cu-8.0Ni-1.8Si-0.15Mg alloy: **d** 15 min; **e** 60 min; **f** 300 min

3.5 Stress Relaxation Behaviors

Figure 7 shows the stress relaxation rate as a function of the loading time, and Table 2 illustrates the strain relaxation rate of the samples after being loaded for 100 h. The stress relaxation rates increased with the loading time in the first 2 h, then increased slightly or kept at a constant value. The stress relaxation rates increase with environmental temperature. For the Cu-8.0Ni-1.8Si alloy, the stress relaxation rates achieve 10.01%, 20.53% and 43.19% after being loaded at 25 °C, 100 °C and 200 °C for 100 h, respectively. In the Cu-8.0Ni-1.8Si-0.15Mg alloy, the stress relaxation rates achieve only 5.27%, 9.49% and 20.04% after being loaded at 25 °C, 100 °C and 200 °C for 100 h, respectively. In the meantime, for the samples of Cu-6.0Ni-1.0Si-0.5Al alloy, stress relaxation rates of specimens achieve 5.08%, 8.89% and 11.89% after being loaded at 25 °C, 100 °C and 200 °C for 100 h, respectively. Moreover, in the Cu-6.0Ni-1.0Si-0.5Al-0.15Mg alloy, stress relaxation rates achieve only 4.05%, 6.62% and 9.74% after being loaded at 25 °C, 100 °C and 200 °C for 100 h, respectively. In summary, the addition of Mg

significantly decreases the stress relaxation rate of CuNiSi alloys.

4 Discussion

4.1 Effect of Mg on the Microstructural Refinement

The addition of Mg in CuNiSi alloys contributed significantly to microstructural evolution. The interaction between solute atoms and dislocations in the Cu-8.0Ni-1.8Si-0.15Mg alloy brings about solid solution strengthening. Evaluation of the data in Fig. 1b and d led to the conclusion that: Mg effectively refines the dendritic microstructures. Figure 2b, d indicates that Mg can promote intermetallics to re-dissolve into the matrix. As the Cu-8.0Ni-1.8Si alloy supersaturate solid solution was aged at 450 °C (Fig. 5), precipitates formed during decomposition. The precipitates size of Cu-8.0Ni-1.8Si-0.15Mg alloy is smaller than that of Cu-8.0Ni-1.8Si alloy, indicating that the precipitation processing was depressed because of the addition of Mg element. Comparatively,

Fig. 6 Bright-field micrographs of microstructural evolutions in Cu–6.0Ni–1.0Si–0.5Al–(0.15Mg) alloys having been aged at 450 °C for various durations. Cu–6.0Ni–1.0Si–0.5Al: **a** 5 min; **b** 60 min; **c** 120 min. Cu–6.0Ni–1.0Si–0.5Al–0.15Mg: **d** 5 min; **e** 60 min; **f** 120 min

as seen in Fig. 6, the Cu–6.0Ni–1.0Si–0.5Al–0.15Mg and Cu–6.0Ni–1.0Si–0.5Al–0.15Mg alloys showed close precipitate size when aged under the same treatment conditions.

4.2 Effect of Mg on the Stress Relaxation Response

Stress relaxation occurs in the elastic materials in elevated temperature environments. Stress relaxation increases rapidly in the first stage and then approaches stable value in the second stage (Fig. 7). Self-diffusion is temperature dependent, and therefore, the stress relaxation rate is temperature dependent. The self-diffusion rates vary exponentially with the temperature. The activation energy for self-diffusion is material dependent. In the studied CuNiSi alloy system, a multitude of precipitates formed in the matrix, introducing a second-phase strengthening. These precipitates act to block the dislocation motion. The geometric effect is much less sensitive to thermal exposures, so the studied alloys have a much higher stress relaxation rate at high temperature than at lower temperature. By adding Mg in the CuNiSi alloys, the precipitates densities are improved, precipitates size is limited and the dislocation motion becomes more difficult. Therefore, the stress reduction is decreased and the stress relaxation rate is lowered. Another aspect of the precipitates' behavior is its effect on the yield strength. The inhibiting

dislocation motion increases the yield strength of CuNiSi alloys.

Figure 8 shows the microstructural evolution of Cu–6.0Ni–1.0Si–0.5Al–0.15Mg alloy specimens loaded at different temperatures for a different time. As stress relaxation testing specimens are loaded at 25 °C, the size and density of precipitates varied very little or negligible. While the sample is loaded at 200 °C for 2–100 h, the size of the precipitates coarsened slightly. When the samples are loaded at 25 °C, which is relatively low, the size of the precipitates remains unchanged. At 200 °C, the samples experience both straining and thermal activity states, and the strain induces the growth of precipitates while at that temperature.

4.3 Strengthening Mechanisms

Firstly, focusing on the solid solution strengthening mechanism, the Mg solute atoms increase the strength of CuNiSi alloys in two ways: (1) they directly pin the dislocations and impede dislocation motion (dislocation–solute interaction) and (2) they indirectly hinder the annihilation of dislocations during plastic deformation and increase the dislocation density (dislocation–dislocation interaction). Secondly, focusing on the precipitation strengthening mechanism, the high yield stress of

Fig. 7 Stress relaxation rate of the studied alloys stress loaded at different temperatures for various durations. **a** Cu-8.0Ni-1.8Si; **b** Cu-8.0Ni-1.8Si-0.15Mg; **c** Cu-6.0Ni-1.0Si-0.5Al; **d** Cu-6.0Ni-1.0Si-0.5Al-0.15Mg

the studied alloys is due to the nanoscale δ -Ni₂Si particles (and β -Ni₃Si), which is dominated by the Orowan mechanism [25]. The increase in yield strength by the precipitates can be evaluated by the Orowan strengthening mechanism [32, 33].

The effect of precipitates on the strengthening is dependent on the precipitate size. The strengthening effect decreases with the size of precipitates, i.e., nanoscale precipitates significantly strengthen the matrix while large scale ones slightly increase the strength. As seen in Figs. 6 and 7, the precipitate size (δ -Ni₂Si and β -Ni₃Si) in Cu-8.0Ni-1.8Si-0.15Mg alloy and the Cu-6.0Ni-1.0Si-0.5Al-0.15Mg alloy is smaller than that of Cu-8.0Ni-1.8Si alloy and Cu-6.0Ni-1.0Si-0.5Al alloy. Additionally, the volume fraction of the precipitate increases dramatically and scatters evenly. As a result, the yield strength increase in Cu-8.0Ni-1.8Si-0.15Mg alloy and Cu-6.0Ni-1.0Si-0.5Al-0.15Mg alloy is larger than those of Cu-8.0Ni-1.8Si alloy and Cu-6.0Ni-1.0Si-0.5Al alloy.

4.4 Stress Relaxation Mechanism

Stress relaxation occurs in the elastic materials in elevated temperature environments. Stress relaxed rapidly in the first stage and then approached stable value in the second stage (Fig. 7). A MAXWELL model was employed to explain the stress relaxation in viscoelastic materials with elasticity and viscosity [34]. The top spring referred to a linearly elastic material, which obeys Hooke law,

$$\sigma_e = E \cdot \epsilon_e, \quad (3)$$

E is the elastic modulus of the materials. The Newton liquid states for linearly viscous material, and

$$\sigma_v = \eta \cdot \dot{\epsilon}_\eta, \quad (4)$$

η is the viscosity. Combining Eqs. (3) and (4), a stress relaxation mode was gained.

$$\epsilon = \epsilon_e + \epsilon_\eta = 0, \quad (5)$$

That is,

Fig. 8 Bright-field micrographs of microstructural evolutions in Cu-6.0Ni-1.0Si-0.5Al-0.15Mg alloy samples loaded at different temperatures for different time. **a** 25 °C/2 h; **b** 25 °C/20 h; **c** 25 °C/100 h; **d** 200 °C/2 h; **e** 200 °C/20 h; **f** 200 °C/100 h

$$E\epsilon_e + E\epsilon_\eta = 0, \quad (6)$$

$$\dot{\sigma} + E\frac{\sigma}{\eta} = 0, \quad (7)$$

where $\dot{\sigma}$ is strain rate, let $t_r = E/\eta$, t_r is a constant. Thus, Eq. (7) can be expressed as,

$$d\sigma/\sigma = -dt/t_r, \quad (8)$$

Take integral of Eq. (8) that is

$$\ln \sigma = -t/t_r + C, \quad (9)$$

where t is the stress relaxation time, C is an integral constant. From Eq. (9), stress relaxation was a time-dependent loss in stress at constant strain. The mechanism of stress relaxation occurs at the atomic level. When stress was applied to the studied sample, generally bending stress introduced the individual atoms and planes of atoms away from the initial position. The number and distribution of dislocations were determined by the stress pattern in the samples. Dislocations are non-equilibrium, high-energy regions and the individual atoms in the dislocation will tend to return to lower energy equilibrium position by self-diffusion within

the metallic crystalline matrix with time and temperature. Stress relaxation lowers the energy in the system, and the atom self-diffusion is the core mechanism of stress relaxation. Self-diffusion is temperature dependent, and therefore, the stress relaxation rate is temperature dependent. The self-diffusion rates vary exponentially with the temperature. The activation energy for self-diffusion is material dependent. In the studied Cu-Ni-Si alloy system, plenty of precipitates were formed in the matrix, introducing a second-phase strengthening. These precipitates act to inhibit the motion of dislocations, and therefore, the stress was reduced in the copper matrix. The geometric effect is much less sensitive to thermal exposures, so the studied alloys have a much higher stress relaxation rate at a high temperature than at a low temperature. With the addition of Mg in the Cu-Ni-Si alloys, the precipitates densities were improved, precipitates size was limited, and the dislocation motion became more difficult, as a result, the stress reduction was decreased and the stress relaxation rate was lowered. Another aspect of the precipitates is the effect on the yield strength. The yield strength of Cu-Ni-Si alloys was increased through inhibiting dislocation motion.

5 Conclusions

Magnesium plays significant roles in refining the dendritic microstructure of as-cast CuNiSi alloys. During solid solution treatment, magnesium promotes the dissolution of non-equilibrium precipitation phase. Compared with the Cu–8.0Ni–1.8Si alloy (and Cu–6.0Ni–1.0Si–0.5Al), Cu–8.0Ni–1.8Si–0.15Mg (and Cu–6.0Ni–1.0Si–0.5Al–0.15Mg) alloy shows enhanced hardness as well as strength. The Cu–6.0Ni–1.0Si–0.5Al–0.15Mg alloy achieves a tensile strength of 1095.7 MPa, yield strength of 935.2 MPa, elongation of 2.29%, hardness of 345.2 HV and an electrical conductivity of 26.6%IACS. The stress relaxation rates of the Cu–6.0Ni–1.0Si–0.5Al–0.15Mg alloy are 4.05% at 25 °C, 6.62% at 100 °C and 9.74% at 200 °C after having been loaded for 100 h, respectively. The stress relaxation rates of the CuNiSiMg alloys are much lower than those of the CuNiSi alloys. Magnesium significantly promotes precipitation nucleation as the samples have been undergoing aging treatment.

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