Effects of severe plastic deformation on transformation kinetics of precipitates in CuNi3Si1Mg

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Abstract
This study estimates the effects of grain refinement on the transformation kinetics of precipitates as well as the development of mechanical properties and electrical conductivity of CuNi3Si1Mg. During aging at 450 °C, the mechanical behaviors of coarse-grained (CG) CuNi3Si1Mg and ultrafine-grained (UFG) CuNi3Si1Mg were studied and the Avrami-equations of phase transformation kinetics and electrical conductivity were analyzed. For both the conditions, the maximum strength was observed at 6 and 12 h, for UFG and CG CuNi3Si1Mg, respectively. According to the transformation kinetics curves of CG and UFG CuNi3Si1Mg, over aging reduced the growth rate of precipitates and affected their coarsening, resulting in decreased strength.

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1. Introduction

Copper-matrix alloys have high strength and good electrical and thermal conductivities; therefore, they are widely used in electronics as connectors, IC framework, and lead frames in the integrated circuit packages [1–8]. In microelectronics, these alloys are used for small parts, essentially due to their high strength [1,4]. Furthermore, precipitation hardened copper-beryllium (Cu–Be) alloys, having high electrical conductivity of 25% IACS (International Annealed Copper Standard) and high strength exceeding 1000 MPa, have been broadly used as materials for electrical applications [4,5]. However, beryllium exposure to humans during the production process causes lung cancer. Therefore, a variety of beryllium-free medium conductive copper alloys are being used nowadays [9]. As an alternative to Cu-Be alloys, much attention has been focused on Cu-Ni-Si alloys [3,5]. Cu-Ni-Si alloys are primarily strengthened by precipitation hardening. In addition, a small quantity of alloying elements with low solubility is added to improve further the mechanical properties and electrical conductivity of these alloys [3,4]. For example, adding a small amount of magnesium (Mg) increases the strength of Cu-Ni-Si alloys by decreasing the inter-precipitate spacing or by an Mg-atom drag on dislocation motion [5]. Moreover, Mg atoms improve stress relaxation resistance by solid solution hardening due to the significant difference between the atomic radii of Mg and Cu. Furthermore, Mg promotes silicide formation [may also form mixed (Ni, Mg)-silicides] as precipitates [10].

In the Cu-Ni-Si system, δ-Ni5Si is the precipitating phase responsible for the strengthening effect [2,6]. Energetically, it is the most stable structure even at higher temperatures, confirmed by the calculated formation enthalpies of different precipitate structures in the Cu-Ni-Si system [11]. The nanoscale precipitates enhance the alloy strength due to very fine dispersion and nanosized structure and return the electrical conductivity by depleting a matrix from solute atoms [6].

In the precipitation hardening process, a systematic change can be designed to improve the mechanical properties and electrical conductivity. The strength and electrical conductivity are controlled by microstructural variations occurring during thermomechanical treatments. In addition to chemical composition variations, plastic deformation (PD) application prior to precipitation hardening leads to optimized mechanical properties, thermal stability, and fatigue performance as well as improved electrical properties due to the very fine dispersion of the nano-sized precipitates. During precipitation hardening, high response of the solution-treated and plastically deformed alloy is correlated to the faster rate of precipitates nucleation. Because PD introduces a high density of slip bands, dislocations, and vacancies in the matrix, which accelerates the precipitation kinetics by improving the atomic mobility and contributing more potential sites for precipitates nucleation [6,12]. During aging, the accelerated
precipitate nucleation improves the strength and electrical conductivity [13]. The high strength is maintained by finely dispersed precipitates that act as impenetrable obstacles for dislocation movement [6,10]. In addition, the electrical conductivity is retained by reducing the solute atoms in the matrix as an electron scattering factor [6].

Initially during the aging process, faster precipitate nucleation leads to the rapid increase in tensile strength and electrical conductivity. Following which the strength increases with the increasing number of precipitates, but the precipitate size and square lattice spacing decreases. Prolonged aging (over aging) slowly reduces tensile strength and increases conductivity. The decline in strength after peak hardness contributes to precipitate coarsening and removal of coherency between the precipitates and matrix interface, making the precipitates ineffective barriers against dislocation motion [6]. It is possible to study the mechanical properties by analyzing the phase transformation and precipitation kinetics because the sensitivity of strength and electrical conductivity to the precipitation characteristics [14,15].

In the precipitation hardened state, the variation in the electrical conductivity of alloys is primarily dependent on the matrix purity and size and amount of nano-sized precipitates formed by the decomposition of the supersaturated solid solution. The electrical resistivity of the material depends on the impurities and PD, and increases linearly with the increasing temperature; however, the influence of PD on electrical resistivity is much weaker [16]. According to the Matthiessen-Fuliminger rule [15], electrical resistivity of a solid solution can be expressed by the following equation:

$$\rho = \rho_0 + \alpha \sigma \sigma$$  \hspace{1cm} (1)

Where $\rho_0$ is the electrical resistance of a pure solvent, $\alpha$ is the percentage of the solute atom, and $\rho$ is the change of resistance due to the addition of 1% solute atoms. There is linearity between resistance and the percentage of the solute atoms (Eq. (1)) [15]. Therefore, electrical conductivity possibly increases with prolonging aging time due to the decrease in electron scattering in the crystal structure through precipitate formation. In addition, electrical conductivity and volume fraction of precipitates possibly have a linear relationship,

$$\sigma = \sigma_0 + A \phi$$  \hspace{1cm} (2)

Where $\sigma_0$ is the initial state of electrical conductivity, $\phi$ is the volume fraction of precipitates, and $A = \sigma_{\text{max}} - \sigma_0$ (maximum electrical conductivity). The electrical conductivity versus fraction of precipitates can be expressed as follows:

$$\sigma = \sigma_0 + (\sigma_{\text{max}} - \sigma_0) \phi$$  \hspace{1cm} (3)

$$\phi = \frac{\sigma - \sigma_0}{\sigma_{\text{max}} - \sigma_0}$$  \hspace{1cm} (4)

At the end of the transformation, $\phi = \phi_{\text{max}}$ and $\phi = 1$. Therefore, the volume fraction of precipitates, $\phi$, can be calculated by measuring the corresponding electrical conductivity of the alloy during precipitation hardening at any time [15–19]. The kinetics of isothermal phase transformation (in this study precipitation) can be described by the well-known Avrami empirical formula for phase transformation as follows [15]:

$$\phi = 1 - \exp(-bt^n)$$  \hspace{1cm} (5)

Where $\phi$ is the volume fraction of precipitates, $t$ is the precipitation hardening time, $b$ and $n$ are time-independent constants [16]. In addition, constant $b$ depends on phase transformation temperature, composition of supersaturated solid solution, and grain size; constant $n$ depends on the type of phase transformation and nucleation sites. Constants $b$ and $n$ can be calculated by transposing Eq. (5) as follows:

$$1 - \phi = \exp(-bt^n)$$  \hspace{1cm} (6)

After logarithmic transformation, Eq. (6) can be expressed as:

$$\ln(1 - \phi) = \ln b + n \ln t$$  \hspace{1cm} (7)

The curve fitting diagram of $\ln(1 - \phi)$ against $\ln t$ is approximately a straight line. $n$ is the slope coefficient of the line and $\ln b$ is the intercept. The transformation kinetics equation can be described by substituting the $n$ and $b$ values in Eq. (5). Furthermore, the kinetic behavior of the phase transformation mostly follows an S-shaped profile when $n$ and $b$ values (Eq. (7)) are used in Eq. (5); Fig. 1 depicts the progress of precipitation as a function of time at a given temperature [15,16,18,20].

Taken together, this study aimed at the following: i) to investigate the effect of post-swaging (SW) aging on microstructure, mechanical properties, and electrical conductivity of CuNi3Si1Mg; ii) to study the effect of microstructure on phase transformation kinetics of precipitates; and iii) to find a relationship between the variation of mechanical properties and coarsening of precipitates during precipitation hardening by calculating the Avrami-equation of phase transformation kinetics and the Avrami-equation of electrical conductivity.

2. Materials and method

The CuNi3Si1Mg samples were supplied by Wieland-Werke AG as 24-mm diameter rods. The material condition was extruded, homogenized (solution annealed), swaged, and subsequently, precipitation hardened. As the initial state, CuNi3Si1Mg was hot extruded at 850–900 °C. Prior to PD, the specimens were solution heat treated (SHT) at 800 °C for 2 h and then cooled to ambient temperature by water quenching. In general, at this temperature, a complete dissolution of coarse Ni-silicides is not possible because the required solution anneal temperature is approximately 1000 °C, which is industrially unrealistic [21]. Subsequently, the quenched specimens were subjected to PD by SW. During SW of solutionized samples at room temperature, the specimen diameter was reduced from 24 to 7 mm, corresponding to an exact deformation degree of 2.5 [22,23]. To determine the optimum aging
condition, the specimens were subjected to artificial aging at 450 °C for different aging times after SW. Microstructural investigations were conducted using light optical microscopy (Model: Zeiss Axioskop). Electron channeling contrast imaging (ECCI) was performed using an Angle selective backscatter detector (AsB) in a Zeiss ULTRA scanning electron microscope equipped with a thermal field emission cathode. An aperture lens of 60 or 120 μm and acceleration voltages of 15–20 kV at a working distance of 2–6 mm were used. For thin-film TEM examination, the samples were cut into disc and ground to 150-μm thickness before thinning by electropolishing in 20% solution of nitric acid in methanol. The solution was maintained at −35 °C, with a voltage of 10–20 V. Electropolished samples were further cleaned from oxides in Gatan PIPS for 30 min at 2 keV. Samples were examined using JEOL JEM–2000FX, and the energy filtered diffraction patterns were obtained using JEOL JEM–2200FS. Both the microscopes were run at 200 kV. Electrical conductivity measurements were performed using the eddy current method (SIGMATEST 2.069). The conductivity was measured and evaluated according to IACS, where 100% IACS = 58 MS m⁻¹ – 1.7241 μΩ cm. Tensile tests were performed using threaded cylindrical specimens on Instron model 5582 universal testing machine using the constant crosshead speed of 1 mm/min, corresponding to a strain rate of \(6.67 \times 10^{-4} \text{s}^{-1}\). Dia Tester 2RC, with a square base pyramid shaped indenter, normal force of 30 Kgf (HV 30), and loading time of 10 s, was used to determine the Vickers hardness. To construct the hardness profile, an average of 10 measurements was taken for each data point.

3. Results and discussion

3.1. Microstructures

After thermo-mechanical treatment, optical microscopy was used to analyze CuNi3Si1Mg microstructures. Fig. 2a–c compares the initial condition of CuNi3Si1Mg microstructure, with conditions after SHT and SHT followed by SW. The initial structure was made up of rather homogeneous coarse equiaxed grains. The average grain size, estimated by the linear intercept method, was approximately 27 μm. The microstructure was composed of the recrystallized coarse grain structures with twins within the grains and an average grain size of 33 μm (Fig. 2b). After SHT, the microstructure contained coarse Ni-silicides (diameter > 200 nm) that were not dissolved during homogenization [21]. After the deformation degree of 2.5, ultrafine grains were formed due to grain subdivision, evolving dislocation accumulation, tangling, and rearrangement [24]. However, the contour of the refined grains in the developed structure during SW was not resolvable in the optical microscope.

Going forth, “UFG CuNi3Si1Mg” is used for the condition in which precipitation hardening was investigated on the swaged samples and “CG CuNi3Si1Mg” for the situation in which precipitation hardening followed SHT (non-swaged condition).

Fig. 3a and b shows ECCI backscatter micrographs in the center and near-surface regions (distance to surface 50–100 μm) of the sample with a deformation degree of 2.5. The center and near-surface microstructure could be easily distinguished by the grain structures. The grain structure in the center of the specimen had a significant content of subgrains and low-angle boundaries with a
grain size of > 1 μm. On the other hand, the grains or subgrains size of the near-surface region with the distance to the surface of approximately 50–100 μm ranged from 200 to 800 nm.

During aging and over aging of the UFG condition, light microscopy micrographs did not clearly reveal the evolution of substructure and precipitates. Fig. 4a–c indicate the ECCI backscatter micrographs of the under-aged, peak-aged, and over-aged conditions at high resolution (magnification: approximately ×10,000) from the area approximately 100 μm beneath the specimen surface. During aging, characteristic structural variations can occur not only in the grain and precipitates size but also in their arrangements. Increasing the aging time increased the precipitate yield.

3.2. Mechanical properties

Fig. 6a and b show the effect of grain refinement on the
mechanical properties (hardness and strength) of CuNi3Si1Mg upon isothermal aging at 450 °C.

During the early stages of aging, the hardness increased due to Ni-silicide precipitation. For the UFG condition, the maximum hardness (approximately 240 HV 30) was almost attained after 1 h aging. A similar trend was shown by CG CuNi3Si1Mg, but the hardness increment was much smaller (approximately 40 HV30) than that observed in the UFG condition. During intensive PD, the significant increase in hardness can be attributed to the remarkable substructure refinement [26]. During PD, high-density deformation bands, slip lines, low energy boundaries and dislocations formed in the matrix, acting as potential sites for further precipitate nucleation (Fig. 5b). The heterogeneous precipitation along dislocations, grain, and subgrain boundaries retard the arrangement of dislocations, thus in turn increasing the strength rapidly [6,12,27,28]. The ECCI backscatter micrograph in Fig. 7 clearly indicates precipitate nucleation along grain and subgrain boundaries in UFG CuNi3Si1Mg.

According to the larché theory [28], the catalyzing effect of dislocations during heterogeneous precipitate nucleation is possibly associated with the interaction between the dilatation strain field of the edge component of dislocation and the strain field resulting from precipitate formation without destroying the dislocation core. The compressive side of an edge dislocation can assist the coherent nucleation of spherical precipitates with a slightly misfit strain. The accommodation of the misfit strain by the stress field of the dislocation lowers the elastic energy, resulting in a significant reduction in the nucleation barriers and an enhanced nucleation rate along dislocation [12,14,28]. Moreover, the UFG samples exhibited a hardness peak, which continuously decreased during over aging. After 72 h, the hardness of the UFG samples was lower than that of the CG samples. During post-SW aging, both recrystallization and precipitation hardening possibly occurred concomitantly. The accelerated recrystallization, associated with a fast growth of Ni-silicide precipitates, occurs upon annealing if the deformation degree is too high. Therefore, a severely deformed supersaturated solution accelerates the precipitation and coarsening kinetics [12,26,29]. After extended aging time, the microstructure contained both coarse grains and precipitates, resulting in a decline of hardness (Fig. 4). During aging, characteristic structural variations could occur not only in the grain and precipitate size, but also in their arrangements. Furthermore, precipitate formation increased with the aging time. EBSD confirmed that the preferentially formed Ni-silicides consist of Ni2Si, which are nucleated in subgrains and grain boundaries [21]. The average precipitate size was approximately 14 nm after peak aging, whereas the average precipitate size increased to 47 nm after aging for 16 h (over-aged condition) [25]. According to a previous study [25], the mean precipitate size in the CG samples was approximately 5 nm in the peak-aged condition. The faster precipitate growth in the swaged condition is due to the faster diffusion of alloying elements within the core of the dislocation than in the bulk. The existence of ample space in the dilatational zones of dislocations provided a fast diffusion path that was controlled by the pipe diffusion of alloying elements, leading to faster coarsening kinetics. Therefore, precipitate coarsening at the
boundaries may affect the hardness by reducing the total number of precipitates (Ostwald ripening), and consequently, the reduction in the density of pinning points for mobile dislocations. The decrease in hardness with time was dependent on precipitate coarsening, recovery, and recrystallization occurring during precipitation hardening of the severely strained samples. This is entirely consistent with the hardness measurement (Fig. 6a) showing that the thermal stability of the CG CuNi3Si1Mg alloy is better than UFG CuNi3Si1Mg [12,26,29]. In the CG samples, a significant increase in hardness (approximately 71 HV 30) occurred after 30 min, reaching a hardness plateau (approximately 202 HV 30) roughly at 12 h. The constant hardness during this time was due to the reducing effect of the recovery process and recrystallization that may have occurred during annealing [26].

Fig. 6b shows the effect of SW on the ultimate tensile strength (UTS) of CuNi3Si1Mg during the age-hardening process. The UTS of UFG CuNi3Si1Mg was 583 MPa; 76% higher than the UTS of the solution-treated condition (331 MPa). When the SW process was combined with 1–6 h aging treatment at 450 °C, a further increase in UTS was achieved. The UTS of the UFG sample reached 865 MPa, which was 161% and 24% higher than that of the solution-treated and just-aged condition (450 °C for 6–12 h), respectively. The effect of precipitation hardening on strength depends on the initial microstructural parameters such as dislocation density, defects, and grain size. The higher tensile strength of CuNi3Si1Mg with smaller grain size is explained below.

For more precipitation during aging, the dislocations act as diffusion paths for solute atoms and provide nucleation sites. Furthermore, after 1–6 h of aging at 450 °C, the strength of UFG CuNi3Si1Mg is attributed to dislocation strengthening [30] and strengthening due to non-shearable particles [29,30]. For plastically deformed metals, the strength is characterized by the dislocation density. At a certain strain, the microstructure simultaneously consists of low-angle grain boundaries (incidental dislocation boundaries, IDBs) and high-angle grain boundaries (geometrically necessary boundaries, GNBs). Under such conditions, the contribution of dislocation strengthening from two sources should be considered. First is the dislocation strengthening due to the presence of IDBs and GNBs, because the boundaries that act as obstacles of dislocation motion in plastically deformed materials consist of dislocations. Therefore, the strengthening is caused by the interaction between these dislocations [29–34]. The second contribution in UFG CuNi3Si1Mg after precipitation hardening depends on non-shearable particles. A decrease in the inter-particle spacing and an increase in the precipitate size increase the strength by pinning the dislocations and impeding their movement. Practically, inter-particle spacing plays a more significant role than the precipitate size. For example, in the ECAP-processed Al 6082, the yield strength measured by mechanical tests was significantly higher than the strength values measured using the Taylor formula, which only takes into account the interaction between dislocations in strengthening. This deviation can be explained by the interaction between dislocations and Mg₅Si and Mn₁₂Si₆Al₅ precipitates [29,30,33].

### 3.3. Phase transformation kinetics of CuNi3Si1Mg during precipitation hardening at 450 °C

Table 1 shows the electrical conductivity measurements and volume fractions of Ni-silicide precipitates under the CG and UFG conditions during precipitation hardening at 450 °C as a function of aging time. Eq. (4) was used to calculate the volume fraction of precipitates. SW had an accelerating effect on the aging process to enhance electrical conductivity. In general, alloying elements serve as scattering centers for conduction electrons in metals; therefore, increasing their concentration lowers the conductivity. Ni-silicides formation from the copper matrix increased the electrical conductivity due to the decrease of scattering of conduction electrons in the crystal structure. Compared to the electrical conductivity in the CG condition, that in the UFG condition was enhanced by PD (here, SW). This is because the precipitates preferentially appear at grain boundaries and in deformed materials quick occurrence of precipitation occurs due to grain boundaries with higher density. If fast precipitate nucleation on grain boundaries in UFG CuNi3Si1Mg affects the kinetics of the precipitation hardening process, further evaluation can be conducted by analyzing the phase transformed kinetics curves [16,35]. Eqs. (4), (5), and (7) can delineate the kinetics of isothermal precipitation. Fig. 8a and b show the typical Avrami double-logarithmic plots for CuNi3Si1Mg in CG and UFG conditions, respectively. The approximate values of n and b in Eq. (5) could be deducted from the straight lines. A = (σ_max − σ_0), n, and b were listed in Table 2. Table 3 depicts the kinetic Avrami equation of electrical conductivity (Eq. 3) and the Avrami–equation of phase transformation (Eq. (5)) of CG and UFG CuNi3Si1Mg.

Fig. 9 shows the measured electrical conductivity and the calculated ones belonged to the UFG and CG conditions for different precipitation hardening times. Most of the errors between the curves and symbols were <4% and <3% for UFG and CG CuNi3Si1Mg, respectively. The calculated results were in agreement with the experimental ones. In both the conditions, the electrical conductivity gradually increased and reached a saturated level for prolonged precipitation hardening times. The conductivity of the UFG sample was 21.16% IACS, and rapidly increased to

| Table 1 | Volume fraction of Ni-silicides (φ) and electrical conductivity (σ) of CuNi3Si1Mg alloy precipitation hardened at 450 °C for different times in CG and UFG conditions. |
|---|---|---|
| CG | UFG |
| t (min) | σ (vs IACS %) | φ (%) | t (min) | σ (vs IACS %) | φ (%) |
| 0 | 22.7 | 0 | 0 | 21.16 | 0 |
| 10 | 26.6 | 13.36 | 10 | 30.3 | 24.03 |
| 60 | 34.8 | 41.44 | 60 | 36.11 | 39.35 |
| 180 | 38.6 | 54.45 | 180 | 40.21 | 50.08 |
| 360 | 40.3 | 60.27 | 360 | 42.9 | 57.15 |
| 540 | 43 | 69.52 | 540 | 43.6 | 58.99 |
| 720 | 43.5 | 71.23 | 720 | 44.6 | 61.62 |
| 1440 | 46.3 | 80.82 | 1440 | 49.4 | 74.23 |
| 2880 | 48.8 | 89.38 | 2880 | 51.87 | 80.73 |
| 4320 | 50.3 | 94.52 | 4320 | 55.1 | 89.22 |
| 5760 | 50.8 | 96.23 | 5760 | 57 | 94.22 |
| 7200 | 51.4 | 98.29 | 7200 | 58.4 | 97.90 |
| 8640 | 51.9 | 100 | 8640 | 59.2 | 100 |
approximately 36.13% IACS after 1 h (60 min) aging at 450 °C. Subsequently, the electrical conductivity slowly rose up to 96 h (5760 min), following which it became nearly constant with further increase in aging time. In contrast, the aging response of the CG samples was less rapid initially and the conductivity value increased to 34.8% IACS after 1 h (60 min), reaching a value of 50.8% IACS after 96 h and then remaining constant.

Furthermore, the changes of the microstructure that take place during the aging treatment significantly influence the variation in strength and electrical conductivity of CuNi3Si1Mg. The strength is governed by the size and distribution of precipitates. However, the conductivity is controlled by the amount of solute atoms in the matrix and by free dislocations, subgrains, and grain boundaries to a weaker extent. The electrical conductivity value after solution treatment (aging time 0 min) was approximately 22.7% IACS. The distortion of lattice, caused by solid solution and point defects in matrix, acts as scatterers of conducting electrons; this is considered as the main reason for the decreased electrical conductivity. In addition, the pre-deformed annealed samples have an electrical conductivity of approximately 21.16% IACS, which is lower than that in the just-SHT condition (22.7% IACS), due to the higher density of defects in matrix. The electrical conductivity in metals obeys Matthiessen law [15] that states that electrical resistivity is proportional to the residual resistivity caused by solid solution, precipitation, vacancy, dislocation, and grain boundary. During precipitation hardening, the dispersed distribution of nanoscale Ni2Si precipitates reduces the electron scattering. In addition, the influence of PD-induced crystalline defects is negligible compared to the effect of solute atoms in the matrix [12]. The electrical conductivity measurements were consistent with a previous study that demonstrated that the decrease of alloying elements’ concentration in solid solution during aging is associated with an increase in electrical conductivity [12]. At the early stage of aging, the electrical conductivity of CG and UFG CuNi3-Si1Mg rapidly increased, restoring 74% and 67% of their electrical...
The degree of supersaturation in the matrix (Cu) acts as a driving force for the growth process. Swaging renders a high density of defects in the matrix, which can act as potential sites for further precipitate nucleation, leading to faster depletion of the matrix and a rapid increase in conductivity due to more precipitation of the excess solute atoms than the CG condition [6,7,12,27,36]. Fig. 10 shows the kinetic curves of precipitation hardening of UFG and CG CuNi3Si1Mg by equations in Table 3.

Figs. 9 and 10 indicate that after a prolonged time, the density of solute atoms decreased, but the precipitation rate and electrical conductivity were constant. The time needed for the volume fraction of Ni-silicides to approach 50% for the UFG and CG conditions were different. For the UFG condition, the time required to reach 50% of the volume fraction of precipitates was approximately 24 h (1440 min) and that for the CG condition was approximately 48 h (2880 min). Initially, the precipitation hardening rate of UFG CuNi3Si1Mg is higher; however, the precipitation rate gradually becomes slower than CG CuNi3Si1Mg (Fig. 10). The degree of supersaturation in the matrix (Cu) acts as a driving force for increasing growth of precipitates [20]. The initial higher precipitation rate in UFG CuNi3Si1Mg can be attributed to the higher density of grain boundaries, dislocations, deformation bands, and slip lines formed during swaging. They act as diffusion paths for solute atoms (Ni and Si) and provide nucleation sites for more precipitation. In contrast, as the precipitates start to grow, there will be no effective competition between the particles for solute atoms. Therefore, the content between precipitates for solute particles increases as they grow, decreasing the concentration level of solute atoms in the matrix. In UFG CuNi3Si1Mg, the precipitates are closer to each other; therefore, the depletion of the matrix from solute atoms, i.e., Ni and Si, as potential sites for further precipitate nucleation and faster depletion of the matrix from the solute atoms, i.e., Ni and Si, act as factors of scattering of electrons.

After precipitation hardening at aging time > 6 h, the concentration of solute atoms, as a driving force for continuing the growth process, decreases, resulting in higher reduction of the precipitation growth rate in the UFG condition than the CG condition. On the other hand, the ECCI micrographs of UFG CuNi3Si1Mg indicate that during aging at 450 °C for 16 h, Ni-silicide coarsening occurs and the precipitate size increases from 14 nm to 45 nm. Thus, growth rate reduction and precipitate coarsening possibly occur concomitantly during aging, which is associated with the softening of CuNi3Si1Mg. In other words, precipitate coarsening is considered as the main reason for the reduction of strength of UFG and CG CuNi3Si1Mg during over aging.

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