The precipitation behavior and mechanical properties of Cu–Ni–Mn–Fe alloy during aging under elevated compression stresses

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Abstract
The Cu–Ni–Mn–Fe alloy is prepared to study its precipitation behavior and mechanical properties during aging under elevated compression stresses. Transmission electron microscopy (TEM), scanning electron microscopy (SEM), x-ray diffraction (XRD), hardness and tensile tests are used for this investigation. The results indicate that the stress-aging treatment leads to the formation of fine, equiaxed grains and twins within the matrix of Cu–Ni–Mn–Fe alloy. During the stress-aging, the density of θ-MnNi precipitates initially decreases and then increases. The stress is further increased from 220 to 385 MPa, the vacancy concentration and dislocation density increase rapidly owing to creep deformation, while the strain-induced effect accelerates the precipitation of θ-MnNi. The results show that the tensile strength and hardness of Cu–Ni–Mn–Fe alloy initially decrease and then increase, and the tensile strength and hardness of the Cu–Ni–Mn–Fe alloy are 890 MPa and 274 HB under the compressive stress of 385 MPa, and the elongation remains at 8.72%. It is mainly attributed to the application of compression stresses during aging, which can enhance both the strength and ductility of the alloy.

1. Introduction
Cu–Ni–Mn–Fe alloys are polybasal precipitation hardening copper alloys, owing to the high level of elasticity [1, 2], high strength, good heat resistance and excellent magnetic isolation properties, Cu–Ni–Mn–Fe alloys are often used as the structural materials in aerospace and automotive applications [3, 4]. For example, Cu–Ni–Mn–Fe alloys are primarily selected for fabricating the high-performance rotor bushing of aerogenerator and some elastic copper alloy devices. Furthermore, due to the urgent needs of aerospace industries, the excellent properties, such as the high strength, good plasticity, are increasingly demanded [1–5]. In order to improve the mechanical properties of Cu–Ni–Mn–Fe alloys, many researchers have focused on their heat treatment. The current method used to strengthen Cu–Ni–Mn–Fe alloys primarily involves a combination of solid-solution and aging strengthening, which results in the formation of θ-MnNi precipitates in the alloy matrix. It has been determined that the strength and hardness of Cu–Ni–Mn–Fe alloys increased as a consequence of the discontinuous θ-MnNi precipitation during aging. In addition, many thermodynamic and kinetic models have been developed to study precipitate evolution in recent years [4–8]. However, although the existing Cu–Ni–Mn–Fe alloy has high strength, brittle fracture often occur during service, which greatly affects the stability of components. Hence, conventional heat treatment unsuitable for the manufacture of Cu–Ni–Mn–Fe alloys with both high strength and ductility.

Stress-aging is a novel technique that combines conventional age-hardening with the application of stress field simultaneously. This technique has been utilized during the manufacture of large, integrally stiffened, lightweight structures intended for aerospace applications [9, 10]. Previous studies have shown that the aged microstructures of Al–Cu alloys could be significantly altered under stress-aging [11–13]. Meanwhile, the externally applied stress during aging significantly affects the precipitation distribution and generates preferential orientation of precipitates in several age-hardenable alloy systems [14]. Chen [15–17] and Guo [18]
investigated the application of various compressive stresses under various loading orientations to determine their effects on the precipitation of Al–Cu and Al–Cu–Mg alloys, and the distribution of precipitates were significantly affected by the loading orientation during stress-aging. Moreover, it has been shown that the application of compressive stress results in the relatively uniform distribution of precipitates within stress-aged samples. Quan et al. [19] studied 2524 Al alloys during creep-aging with the application of an external stress, the value of the peak hardness increased, while the peak hardness was achieved at a faster rate. Meanwhile, the density of the precipitates can be increased by the application of an external stress. Guo [20] and Cao [21] explored the fact that the stress-aging process has an appreciable influence on the intermetallic precipitates of Al–Zn–Mg–Cu alloys, and found that their preferential growth will promote the precipitation of the $\eta'$ phase but inhibit the precipitation of the $\eta$ phase. Furthermore, the behaviour of Al–Cu–Mg and Al–Zn–Mg–Cu alloys was investigated during a two-stage stress-aging process. It was found that owing to the interaction between dislocations and aging precipitates, the diffusion kinetics of solute atoms could be accelerated [22–25]. Song et al. [26] found that such accelerated precipitation is related to the influence of the applied compressive stress during the compressive stress-aging of $\beta$-Ti alloys.

Previous studies have primarily focused on the improvement of microstructure and properties of typical Al alloys and Ti alloys during stress-aging [27–33], and the mechanical properties of these alloys were significantly improved. Hence, it is necessary to investigate the stress-aging behaviour of Cu–Ni–Mn–Fe alloys intended for engineering applications. In this study, stress-aging experiments were conducted on Cu–Ni–Mn–Fe alloy under a wide range of applied compressive stresses. The effects of the compressive stress on the microstructures were investigated using scanning electron microscopy (SEM), transmission electron microscopy (TEM), and x-ray diffraction (XRD) analysis. Furthermore, the mechanical properties of Cu–Ni–Mn–Fe alloy aged under various compressive stresses were also studied.

2. Materials and methods

Alloy with a nominal composition of Cu–20 wt.% Ni–20 wt.% Mn–5 wt.% Fe was produced by vacuum induction furnace under argon atmosphere. The electrolytic Cu (purity $\geq 99.5\%$), pure Ni (purity $\geq 99.5\%$), pure Mn (purity $\geq 99.5\%$) and pure Fe (purity $\geq 99.5\%$) were melted in a graphite crucible and then cast in a water-cooled copper mold at 1150 °C. The as-cast samples were solution treated at 520 °C for 50 min followed by water quenching. Subsequently, the samples were artificially aged at 430 °C for 24 h with and without the application of compressive stress. The stress-aging was conducted using a SLQ-16B vertical hot-pressing sintering furnace with nitrogen atmosphere. A schematic diagram of the stress-aging experimental setting is shown in figure 1. External stress was applied along the longest axis of each specimen. The dimensions of the stress-aged specimens were 16 $\times$ 16 $\times$ 100 mm$^3$.

Brinell hardness measurements were performed on a HB-3000 hardness tester with 750 Kg load for 30 s dwelling time. For each specimen, the test was repeated 5 times to obtain an arithmetical mean value. Tensile tests were performed by a HT-10 universal testing machine at a displacement rate of 0.2 mm min$^{-1}$ at room
temperature. Phase identification of the specimens was conducted by 7000S x-ray diffractometer (XRD) with Cu Kα radiation. An accelerating voltage of 40 kV, current of 40 mA, and step-scanning rate of 8° min⁻¹ were used. The microstructures were characterized using a JSM-6700F field-emission scanning electron microscope (FE-SEM) and JEM-3010 transmission electron microscope (TEM) with an accelerating voltage of 300 kV. The specimens were mechanically ground, polished and etched with a corrosive agent (5 g FeCl₃ + 20 ml HCl + 100 ml H₂O) for metallographic examination. The TEM foil specimens were obtained by double-jet electro-polishing at −30 °C within a mixture of 67% methanol and 33% hydrogen nitrate.

3. Results and discussion

3.1. Effects of stress-aging on the microstructure morphology

Figure 2 shows the SEM images of Cu–Ni–Mn–Fe alloy specimens aged at 430 °C for 24 h with and without applied compressive stress. The phase composition of Cu–Ni–Mn–Fe alloy is mainly α-Cu dendrites (marked with white ellipses) and θ-MnNi precipitated phase. The white particles in the high-magnification image is θ-MnNi precipitates after stress-aging, as shown in figure 2(f). Based on figure 2(a), it can be determined that

Figure 2. SEM images of stress-aged specimens that were treated under compressive stresses of: (a) 0 MPa, (b) 110 MPa, (c) 220 MPa, (d) 330 MPa, (e) 385 MPa and (f) high-magnification image.
dendrites exist in the Cu–Ni–Mn–Fe alloy sample that was not subjected to stress. As shown in figures 2(b)–(d), the specimens were subjected to compressive stresses of 110 MPa to 330 MPa, the lengths of the primary dendrite arms decrease and secondary dendrite arms coarsen. It is noteworthy that when the specimen is subjected to a compressive stress of 385 MPa, the grains of the alloy are completely transformed into fine, equiaxed grains, as shown in figure 2(e). A stress-aging temperature of 430 °C was used to treat the Cu–Ni–Mn–Fe alloy, which is greater than the creep and dynamic-recrystallization temperatures of the alloy [22]. Therefore, the heat and pressure are constantly applied for a long period of time during the treatment, and the specimens tend to creep. This causes the dynamic recovery and recrystallization of the alloy microstructures, resulting in refinement of the alloy grains.

Figure 3 shows a bright-field TEM image and selected area electron diffraction (SAED) pattern of the stress-aged specimen subjected to a compressive stress of 385 MPa. As shown in figure 3(b), the diffraction pattern has a symmetrical distribution, this is a typical of an electron-diffraction pattern of twins. In addition, twins can also be observed in figure 3(a). The indexing of the pattern shows that the twinning plane is \( \{111\} \), which is consistent with that of twins typically found in face-centred cubic (FCC) alloy. The features of the twins could be attributed to the creep deformation of the alloy under compressive stress, which results in the generation of deformed twins [34]. The Cu–Ni–Mn–Fe alloy matrix consists of FCC Cu-based solid solution, which has a low stacking fault energy (SFE). This results in the formation of wide stacking faults, which renders it difficult for the dislocations to move via cross slip or climb [35]. Therefore, it is easy for deformation twins to form within the Cu–Ni–Mn–Fe alloy during the stress-aging. In the present study, the presence of twins can result in the rotation of the matrix grains to a certain angle. This promotes the activation of the non-basal slip system, which consequently increases the plasticity of the material [36]. Moreover, the twins can subdivide the grain, resulting in the refinement of the matrix grains. In addition, the twinning surfaces are also favourable nucleation locations for recrystallized grains, and thus promote the dynamic-recrystallization process [34–36]. Hence, the generation of deformed twins indicates that the stress-aging treatment can be used to refine the Cu–Ni–Mn–Fe alloy grains and accelerate the dynamic-recrystallization. This is consistent with the conclusions derived based on the SEM images of the stress-aged samples, shown in figure 2.

3.2. The precipitation behavior of stress-aging

In the case of Cu–Ni–Mn–Fe alloy, precipitation-hardening is the main strengthening mechanism [1]. Hence, the strengthening effect depends largely on the method used, as well as the densities and sizes of the precipitates. Previous studies have shown that \( \theta\)-MnNi precipitates, with a face-centred tetragonal structure (FCT), are formed in Cu–Ni–Mn–Fe alloy during aging [3]. Figure 4 shows the interaction that occurs between the dislocations and precipitates when Cu–Ni–Mn–Fe alloy specimens are subjected to compressive stresses among 0 and 385 MPa. Figure 4(a) shows the TEM image and SAED pattern of the precipitates formed in Cu–Ni–Mn–Fe alloy following the stress-free aging treatment. This reveals that a small amount of \( \theta\)-MnNi precipitates can be found within the microstructure of the alloy, which has a low dislocation density and the sizes of precipitates smaller than 100 nm. As shown in figures 4(b)–(e), the dislocation density slightly increases with the increasing of compressive stress. It can be observed that some flexural dislocations exist within the grain interiors, and attached to the precipitates. Typically, the dislocation segments are pinned by the precipitates. The results clearly show that creep is the dominant deformation mechanism of the stress-aging [22]. Figures 4(b) and (d) show that stress- and strain-induced effects occur during the stress-aging of Cu–Ni–Mn–Fe alloy. In general, when stress...

Figure 3. (a) TEM image and (b) corresponding diffraction pattern of the stress-aged specimen with an applied compressive stress of 385 MPa.
and thermal fields are applied to the alloy, coupling effects occur between the precipitates and plasticity. The occurrence of creep may result in the introduction of defects, such as dislocations and vacancies. These consequently affect the nucleation and growth of aging precipitates. However, the occurrence of precipitation-hardening inhibits creep via a pinning effect during the stress-aging. Consequently, the strength and hardness of the alloy rapidly increase. Therefore, the mutual coupling stress- and strain-induced effects occur when creep deformation and precipitation simultaneously.

Under compressive stress, the precipitation of θ-MnNi is inhibited and a small amount of θ-MnNi precipitates can be observed within the alloy matrix, as shown in figure 4(c). This stress-induced effect has a dominant effect on the precipitation, the density of θ-MnNi precipitates gradually decreases as the compressive stress increases from 0 to 220 MPa. When the applied compressive stress is greater than 220 MPa, the rate of creep deformation and the vacancy concentration both

**Figure 4.** Dislocation/precipitate interactions that occurred within the stress-aged Cu–Ni–Mn–Fe alloy specimens treated under applied stresses of: (a) 0 MPa, (b) 110 MPa, (c) 220 MPa, (d) 330 MPa, and (e) 385 MPa.
increase rapidly. This directly accelerates the diffusion kinetics of the solute atoms [12–15]. Furthermore, when creep deformation occurs, the dislocation density rapidly increases. The presence of dislocations contributes to the nucleation and growth of the aging precipitates. The dislocations provide a path for the short-circuit diffusion of solute atoms which accelerates the precipitation of $\theta$–MnNi. In this circumstances, the strain-inducing effect caused by creep deformation plays a leading role on the precipitates, and the density of the precipitates gradually increases with the increase of the external compressive stress.

The density of precipitates in Cu–Ni–Mn–Fe alloy is determined by the compressive stress which is applied during stress-aging. The density of the precipitate was evaluated through image processing. Based on the results of thermodynamic analyses performed in previous study [37], the solubility and distribution of alloying elements within the matrix and the precipitation behavior are related to the temperature and applied stress. The effect of stress-aging on the precipitation of $\theta$-MnNi primarily depends on the solubility of the alloying elements, namely Ni and Mn, in Cu–Ni–Mn–Fe alloy matrix under external stress.

When stress is applied, the precipitation of $\theta$-MnNi can be simplified by solely considering the precipitation that occurs within Ni–Mn binary matrix. Here, the variation in the solubility of alloying element ($i$) within the matrix phase ($\alpha$) can be expressed by the following equation [37]:

$$dX_i^\alpha = \frac{dp \cdot V^\theta}{d^\alpha G^\alpha \cdot (X_i^\alpha - X_{i}^{\alpha_0})}$$

where, $X_i^\alpha$ is the dissolved mole fraction of element $i$ within the matrix $\alpha$, $X_{i}^{\alpha_0}$ is the dissolved mole fraction of element $i$ within the $\theta$-MnNi, $V^\theta$ is the volume of the matrix $\alpha$, $P$ is the applied stress and $G$ is the Gibbs energy. $X_i^\alpha > X_{i}^{\alpha_0}$ and $(X_i^\alpha - X_{i}^{\alpha_0})$ can be regarded as constants since the compositions of the $\theta$-MnNi and the matrix are quite different. When the regular solution model is applied to $\theta$-MnNi, and the corresponding approximate processing is performed, the following equation can be obtained [37]:

$$\ln \frac{X_{i}^{\alpha_0}}{X_i^\alpha} = \frac{P \cdot V^\theta}{RTX_i^\alpha} = \frac{A \cdot \sigma \cdot V^\theta}{TRX_i^\alpha}$$

where, $X_{i}^{\alpha_0}$ is the molar fraction of a dissolved element $i$ within the matrix $\alpha$ when no stress is applied, $A$ is a constant, $\sigma$ is the applied stress, $R$ is the gas constant and $T$ is the absolute temperature.

According to equation (2), when a compressive stress is applied, the solubility of the solute elements Mn and Ni in Cu–Ni–Mn–Fe alloy matrix is increased since $\sigma$ is a positive value. Therefore, the density of $\theta$-MnNi is decreased, i.e. the inhibition of the precipitation of $\theta$-MnNi. When the compressive stress is lower than 220 MPa, the dislocation and only $\theta$-MnNi subjected to the stress-induced effect. Thus, when the alloy is subjected to compressive stresses among 0 and 220 MPa, the quantity of $\theta$-MnNi within the alloy gradually decreases. This explains the correlation between the mechanical properties of Cu–Ni–Mn–Fe alloy and the change in the microstructure during compressive stress-aging. When the applied compressive stress is further increased, creep deformation of the Cu–Ni–Mn–Fe alloy occurs. It may be considered that the $\theta$-MnNi are subjected to a strain-induced effect and that creep may introduce defects, such as dislocations and vacancies, which consequently affect the nucleation and growth of aging precipitates. Hence, the quantity of $\theta$-MnNi within Cu–Ni–Mn–Fe alloy gradually increases with the increase of compressive stresses.

The evolution of precipitates within Cu–Ni–Mn–Fe alloy during stress-aging can also be demonstrated by the obtained XRD patterns, as shown in figure 5. The Cu–Ni–Mn–Fe alloy is a two-phase alloy, which consists of the matrix ($\alpha$–Cu), and $\theta$–MnNi precipitates. At initial state, three major peaks can be clearly observed, which were calibrated as crystal planes $\alpha$–(111), $\alpha$–(200), and $\alpha$–(220). Comparatively, in the case of $\theta$-MnNi precipitates, two peaks which corresponds to crystal planes $\theta$–(110) and $\theta$–(111) can be observed. Here, the diffraction angle of $\theta$-MnNi precipitates is very close to that of the matrix. When the applied stress is increased during the stress-aging, the intensities of the peaks changed attributed to crystal planes $\alpha$–(111) and $\alpha$–(200). It can be inferred that the $\theta$-MnNi are predominantly precipitated along crystal planes $\alpha$–(111) and $\alpha$–(200) during stress-aging. Meanwhile, it can also be observed that the diffraction peak of crystal plane $\alpha$–(111) initially shifts to the left and subsequently to the right when the applied stress is increased, as shown in figure 5(b).

When the applied stress is lower than 220 MPa, the diffraction peak shifts to left, there is an increase in the crystal-plane spacing of the matrix according to Bragg equation. It is indicated that the compressive stress inhibits the precipitation of $\theta$-MnNi. With the increase of applied stress ($>220$ MPa), the diffraction peak shifts to right, and there is a decrease in the crystal-plane spacing of the matrix. It is indicated that a mass of $\theta$-MnNi is precipitated from the matrix. Moreover, the intensity of each diffraction peak decreased first and then increased with the increasing of applied stress during stress-aging, as shown in figure 5(b). This observation correlates with the number of precipitates within the microstructure of Cu–Ni–Mn–Fe alloy following the stress-aging treatment, as shown in figure 4. It is also indicated that the stress-aging treatment promotes the precipitation of $\theta$-MnNi phase.
3.3. The mechanical properties change during aging under elevated stresses

Figure 6 shows the relationship between the Brinell hardness and the compressive stress applied during stress-aging conducted at 430 °C for 24 h. The Brinell hardness initially decreases and then increases as the applied compressive stress is increased. The maximum hardness increases to 247 HB when a compressive stress with 385 MPa is applied. It is attributed to the precipitation of θ-MnNi, which results in significant hardening and strengthening effects. The result is consistent with the trend associated with the density of θ-MnNi within the matrix of the alloy, as shown in figure 4. Besides, with the increase of applied stress, dynamic recovery and recrystallization are initiated, and the hardness of the alloy increases gradually.

The tensile properties of Cu–Ni–Mn–Fe alloy specimens aged at 430 °C for 24 h under various compressive stresses were determined at room temperature. Figure 7 shows the variation of ultimate tensile strength, yield strength and elongation values of designed Cu–Ni–Mn–Fe alloy versus the applied stress, respectively. The results indicate that the Cu–Ni–Mn–Fe alloy exhibits a noticeable age hardening effect during stress-free aging at 430 °C for 24 h. It is mainly attributed to the precipitation of θ-MnNi, which provides significant hardening and strengthening effect [1–5]. With the increase of applied compressive stress, the strength first decreases and then increases. It is consistent with the trend associated with the hardness, as shown in figure 6. When the applied stress is 220 MPa, the ultimate tensile strength and yield strength values of the alloy reach the minimum value, namely 625 MPa and 391 MPa, respectively. When the applied stress is increased to 385 MPa, the ultimate tensile strength of the alloy increased to 890 MPa. Comparatively, the elongation values of the alloys exhibit an
opposite trend. When the applied stress is increased, the elongation initially increases from 8% to 11.12%, and then decreases to 8.72%. Owing to the combined effects of large applied stress and high-temperature aging, the strain-induced effect, attributed to creep deformation, has a dominant role regarding precipitation, the strengthening are mainly attributed to the density of precipitates increases under 385 MPa compressive stress and the grains of Cu–Ni–Mn–Fe alloy are completely transformed into fine, equiaxed grains. In addition, twins are formed within the Cu–Ni–Mn–Fe alloy matrix during stress-aging. Twin boundaries can effectively prevent the movement of dislocations, and the existence of the twin planes results in the refinement of the matrix grains. Therefore, the strength of Cu–Ni–Mn–Fe alloy increases and greater elongation is achieved by stress-aging treatment, demonstrating that the plasticity of the alloy increases.

Figure 8 shows the fracture morphology of Cu–Ni–Mn–Fe alloy with stress-aging. Based on the elongation values and the observation of fracture morphology, the primary fracture mode of the alloy is ductile fracture, and the fracture mechanism is primarily attributed to ductile fracture. As shown in figures 8(a)–(c), many curved tear-ridge lines and dimples can be observed in local areas. When the applied compressive stress increases continuously during stress-aging, the depths of dimples increase gradually. It is indicates that the plastic ductility of the alloy gradually increases, which is consistent with the tensile-test results, as shown in figure 7. Besides, a step surface can be clearly observed with small, equiaxed dimples distributed on it. The fracture morphology differs to that was subjected to an applied stress of 330 MPa. When a large external stress is applied, creep deformation occurs. Consequently, dynamic recovery and recrystallization occur under the high-stress and high-temperature conditions. The dendritic grains of Cu–Ni–Mn–Fe alloy matrix become equiaxed, and the strength of the alloy slightly decreases. However, the plasticity is greater than that of the sample subjected to stress-free aging.

4. Conclusions

In summary, we have systematically studied the precipitation behavior and mechanical properties of Cu–Ni–Mn–Fe alloy during aging under elevated compression stresses. The main conclusions are:

1. Compressive stress-aging treatments can be used to refine the grains of Cu–Ni–Mn–Fe alloys and the deformed twins are produced within the alloy matrix.

2. The stress-induced effect inhibits the precipitation and the strain-induced effect accelerates the precipitation of θ-MnNi. As the external compressive stress gradually increased from 0–385 MPa, the density of θ-MnNi first decreased and then increased.

3. Compressive stress-aging can enhance both the strength and ductility of Cu–Ni–Mn–Fe alloy. The ultimate tensile strength of Cu–Ni–Mn–Fe alloy are 890 MPa under the aging compressive stress of 385 MPa, and the elongation remains at 8.72%.
Figure 8. Fracture morphologies of stress-aged Cu–Ni–Mn–Fe alloy specimens treated under compressive stresses of: (a) 0 MPa, (b) 110 MPa, (c) 220 MPa, (d) 330 MPa, and (e) 385 MPa.

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Author contributions

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References

[1] Zou J T, Shi H, Li T, Wang X H and Liang S H 2016 Study on aging strengthened Cu–Ni–Mn–Fe alloy Mater. Res. Express 3 085040
[2] Shen Z Y, Wang X P, Yang Y G and Yang Y Z 2009 Influence of heat treatment process on properties of Cu–Ni–Mn–Fe alloy Rare Met. 33 119–23
[3] Zou J T, Zhao J P, Wang X H and Liang S H 2013 Effect of trace boron addition on microstructure and properties of Cu–Ni–Mn–Fe alloy T. Nonferr. Met. Soc. China 23 1055–11
[4] Zou J T, Shi L, Shi H, Feng Q L and Liang S H 2020 Study on aging strengthening and nano precipitates of Cu–Ni–Mn–Fe alloy Mater. Res. Express 7 055040–1
[5] Xie W B, Wang Q S, Mi X J, Xie G L, Liu D M, Gao X C and Li Y 2015 Microstructure evolution and properties of Cu–20Ni–20Mn alloy during aging process Trans. Nonferr. Met. Soc. China 25 3247–51
[6] Zhang W B, Du Y, Zhang L J, Xu H H, Liu S H and Chen L 2011 Atomic mobility, diffusivity and diffusion growth simulation for fcc Cu–Mn–Ni alloys Calphad 35 367–75
[7] Xie W B, Wang Q S, Xie G L, Mi X J, Liu D M and Gao X C 2016 Kinetics of discontinuous precipitation in Cu–20Ni–20Mn alloy Int. J. Miner. Metall. Mater. 23 323–9
[8] Han D, Wang Z Y, Yan Y, Shi F and Li X W 2017 A good strength–ductility match in Cu–Mn alloys with high stacking fault energies: determinant effect of short range ordering Scripta Mater. 133 59–64
[9] Marlaud T, Deschamps A, Bley F, Lefebvre W and Baroux B 2010 Evolution of precipitate microstructures during the retrogression and re-aging heat treatment of an Al–Zn–Mg–Cu alloy Acta Mater. 58 4814–26
[10] Liu Y Y, Pan Q L, Zhang X L, Liang S X, Zheng L Y, Gao F and Xie H L 2014 Effects of stress-aging on the microstructure and properties of an aging forming Al–Cu–Mg–Ag alloy Mater. Des. 58 247–51
[11] Zha A W and Starke J E 2001 Stress aging of Al–xCu alloys: experiments Acta Mater. 49 2285–95
[12] Zha A W, Chen J and Starke J E 2000 Precipitation strengthening of stress-aged Al–xCu alloys Acta Mater. 48 2329–46
[13] Luo K, Jiang Y, Yi D Q, Fu S and Zang B 2013 Strained coherent interface energy of the Guinier–Preston II phase in Al–Cu during stress aging J. Mater. Sci. 48 7927–34
[14] Zha A W and Starke J E 2001 Stress aging of Al–Cu alloys: computer modelling Acta Mater. 49 3063–9
[15] Chen Q, Chen Z G, Guo X B and Deng Y L 2016 Changing distribution and geometry of $\gamma'$ in Al–Cu–Mg single crystals during stress aging by controlling the loading orientation Mat. Sci. Eng. A 650 154–60
[16] Chen Q, Chen Z G, Deng Y L, Guo X B and Ren J K 2016 Effect of loading orientations on the microstructure and property of Al–Cu single crystal during stress aging Mater. Charact. 117 35–40
[17] Chen Z G, Chen Q, Guo X B, Ren J K and Deng Y L 2016 The precipitation behavior of Al–Cu–Mg–Ag single crystal during aging under elevated compression stresses Mat. Sci. Eng. A 669 33–40
[18] Guo W, Yang M, Zheng Y, Zhang X S, Li H, Wen Y X and Zhang J W 2013 Influence of elastic tensile stress on aging process in an Al–Zn–Mg–Cu alloy Mater. Des. 106 14–7
[19] Quan J W, Zhao G, Tian N and Huang M L 2013 Effect of stress on microstructures of creep-aged 2524 alloy Trans. Nonferr. Met. Soc. China 23 2209–14
[20] Guo W, Guo J Y, Wang J D, Yang M, Li H, Wen Y X and Zhang J W 2015 Evolution of precipitate microstructure during stress aging of an Al–Zn–Mg–Cu alloy Mat. Sci. Eng. A 654 167–75
[21] Cao S F, Pan Q L, Liu X Y, Lu Z L, He Y B and Li W B 2010 Effects of external stress on aging precipitation behavior of Al–Cu–Mg–Ag alloy Trans. Nonferr. Met. Soc. China 20 1513–9
[22] Lin Y C, Zhang J L and Chen M S 2016 Evolution of precipitates during two-stage stress-aging of an Al–Zn–Mg–Cu alloy J. Alloy Compd. 684 177–87
[23] Lin Y C, Jiang Y Q, Zhang X C, Deng J and Chen X M 2014 Effect of creep–aging processing on corrosion resistance of an Al–Zn–Mg–Cu alloy Mater. Des. 61 228–38
[24] Lin Y C, Zhang J L, Liu G and Liang Y J 2015 Effects of pre-treatments on aging precipitates and corrosion resistance of a creep-aged Al–Zn–Mg–Cu alloy Mater. Des. 83 806–75
[25] Lin Y C, Liu G, Chen M S, Li J, Zhou M and Zhou H M 2015 Effects of two-stage creep–aging processing on mechanical properties of an Al–Cu–Mg alloy Mater. Des. 79 127–35
[26] Song Z Y, Sun Q Y, Xiao L, Liu L, Wang H, Chen W, Sun J and Ge P 2008 The influence of compressive stress on the precipitation process of $\eta$ phase in solution treated TB3 alloys during aging process Rare Metal. Met. Eng. 37 700–3
[27] Guo F, Zhang D F, Fan X W, Li J X, Jiang L Y and Fan F S 2016 Microstructure, texture and mechanical properties evolution of pre-twining Mg alloys sheets during large strain hot rolling Mat. Sci. Eng. A 655 92–9
[28] Yang W C, Ji S X, Zhang Q and Wang M P 2015 Investigation of mechanical and corrosion properties of an Al–Zn–Mg–Cu alloy under various aging conditions and interface analysis of $\eta'$ precipitate Mater. Des. 85 752–61
[29] Fang X, Du Y, Song M, Li K and Jiang C 2012 Effects of Cu content on the precipitation process of Al–Zn–Mg alloys J. Mater. Sci. 47 8174–87
[30] Tao J S, Zhang L, Wu G H, Chen A T, Zhang X L and Shi C C 2018 Effect of heat treatment on the microstructure and mechanical properties of extruded Al–4Cu–1Li–0.4Mg–0.4Ag–0.18Zr alloy Mat. Sci. Eng. A 717 11–9
[31] Cavalerio P, Cabibbo M, Panella F and Squillace A 2009 A 2198 Al–Li plates joined by friction stir welding: mechanical and microstructural behaviour Mater. Des. 30 3622–31
[32] Chen A, Peng Y, Zhang L, Wu G and Li Y 2016 Microstructural evolution and mechanical properties of cast Al–3Si–1.5Cu–0.2Zr alloy during heat treatment Mater. Charact. 114 234–42
[33] Razi A, Khalil-Allali J, Ehmsenfar M R, Pourbabak S, Schyders D and Amin-Ahmadi B 2018 Influence of stress aging process on variants of nano-Ni4Ti3 precipitates and martensitic transformation temperatures in NiTi shape memory alloy Mater. Des. 142 93–100
[34] Zhao S C, Guo F J, Wang L P, Wu T and Feng Y C 2015 Effect of pre-compressive strain on microstructure and mechanical properties of Mg–2.7Nd–0.4Zn–0.5Zr alloy Mat. Sci. Eng. A 647 28–33
[35] Lu L, Sui M L and Lu K 2000 Superplastic extensibility of nanocrystalline copper at room temperature Science 287 1463–6
[36] Lou C, Zhang X Y and Ren Y 2014 Improved strength and ductility of magnesium alloy below micro-twin lamellar structure Mat. Sci. Eng. A 614 1–5
[37] Tian S G, Liu B Z, Yin L D, Yang H C, Xu Y B and Hu Z Q 2013 Stress-induced precipitation of fine $\gamma'$ phase and thermodynamics analysis J. Iron Steel Res. 15 626–9