Effect of heat treatments on the microstructure and mechanical properties of Zn-15 wt% Al alloy

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Abstract
The effect of heat treatment on the micro-structural and mechanical properties of Zn-15wt%Al alloys was investigated systematically in this work. After being annealed below 275 °C, granular \(\alpha\) particles were distributed homogeneously in the \(\eta\) phase matrix, while a lamellar microstructure along with isolated \(\eta\) grains was observed when the alloys were annealed above 275 °C, followed by an air cooling (AC) treatment. A subsequent tensile test showed that the strength of alloys always increased with the annealing temperature. However, when the alloys were annealed above 275 °C, followed by a water-quenched (WQ) treatment, their strength firstly decreased with the aging time and then became unchanged. It was revealed by an \textit{in situ} investigation that the \(\gamma'\) phase retained from quenching decomposed gradually into an equiaxed microstructure during aging, leading to a reduction in the strength. Transmission electron microscopic (TEM) studies showed that dislocations were rarely inspected after a severely plastic deformation in the WQ-treated alloys, indicating that grain boundary sliding and rotation would be a dominant deformation mechanism. However, for the AC-treated alloys, dislocation tangles were easily found in lamellas due to the severe plastic deformations, demonstrating that in addition to grain boundary sliding and rotation, dislocation slip was another important deformation mechanism. A strength model, incorporating phase transformation dynamics with the law of mixtures, was established for the WQ-treated alloys annealed above 275 °C. Meanwhile, the number of \(\alpha\) particles in the \(\gamma'\) phase was assumed to be a key parameter for understanding the evolution of mechanical properties of Zn-15wt%Al alloys during the heat treatments. The more \(\alpha\) particles in a unit volume of the \(\gamma'\) phase, the easier plastic deformation the alloys is prone to, a lower strength and a larger elongation of the alloys will be.

1. Introduction

Zn–Al alloys are widely used in automobile, aerospace, metallurgical, mineral, electrical and electronic industries, due to their good mechanical properties such as exceptional cast-ability, anticorrosion ability, and to their low cost [1–12]. In particular, the Zn–Al alloys have drawn much attention in recent years by their excellent superplastic property at room temperature [6–8]. However, the strength evolution of Zn–Al alloys has not been investigated systematically. Zn–Al alloys have been studied intensively as industries braze cooper and aluminum parts [13–15]. To relieve the residual stress and homogenize the composition of brazed seam, an annealing treatment is usually carried out for the alloys after being brazed. Therefore, the emergence of microstructure and the accompanied mechanical properties during annealing and aging treatments of the alloys should be investigated specifically, which is crucial for a reliable evaluation of brazed structures.

Generally, annealing treatments can lead to a decrease in strength owing to a reduction in dislocation density and an increment of grain size [16–21]. However, work softening and anneal hardening phenomena were
recently reported in rolled (about 90%) Zn(22%)-Al alloys with heat treatments of solutionizing followed by artificial ageing [16]. Moreover, after being heat-treated at 250 °C, the grain size of the alloys increases during the annealing treatment process. Zhang et al proposed [17] that in Zn(22%)-Al alloys, a smaller grain size exhibited an unusually anneal-hardening behavior. When the alloys were subjected to a short-term annealing in the single-phase region followed by a subsequent water-quenching, an elongation was largely improved without decrements in strength of the alloys [17]. Senkov et al [19] found that the dissolution of solute atoms along the grain boundaries enhanced the grain growth in Zn(22)-Al alloys (meaning 22 wt% Al) during the annealing treatments. Yang et al [18] revealed that the work-softening behavior was enhanced by a decrease in grain size which was related to the absorption of dislocation pile-up by grain boundary and when subjected to a grain-coarsening heat treatment, ultra-fine-grained Zn–Al and Zn–Al–Cu alloys exhibit an unusual anneal-hardening behavior. Liu et al [20] applied a high temperature heat treatment to Zn–10Al–2Cu–0.05Ti (ZA10) alloy tubes and revealed that this heat treatment only led to a slightly increase in strength. After an equal channel angular pressing (ECAP) at room temperature, an obvious deformation-induced softening was reported by Hernandez-Rivera et al [21]. Meantime, a slight hardening of Zn–21Al–2Cu was also revealed after the annealing treatment. However, the mechanism for the micro-structural emergence and strengthening after annealing treatments has not been explored in detail yet.

Zn–Al alloys have a low melting point only doubling room temperature (RT), which makes it possible for thermodynamically unstable phase to transform at RT [22–24]. Therefore, a micro-structural stability during aging at RT is a crucial factor to determine the mechanical properties of Zn–Al alloys [25–27]. Demirtas et al reported that a Zn(0.3%)-Al alloy lost its super-plasticity after a long term of natural aging [27], which was attributed mainly to a grain boundary corrosion during a natural aging process due to the presence of Al precipitates at the grain boundaries. Therefore, an investigation of the effect of a natural aging on the microstructure of Zn–Al alloys and their correspondingly mechanical behavior is important for their applications [15, 28–32]. Therefore, this work was focusing on the investigations of the aspects of the Zn–Al alloys, and presents the obtained results together with our mechanistic analyses.

2. Experimental

Ingots of high purity metals Zn and Al (≥99.9%) with a weight ratio of 18:85 were melted in a resistance furnace with a graphite crucible. Zn(15%)-Al billets with a diameter of 80 mm were acquired by casting the melt into a cylindrical steel model. These billets were extruded into wires (diameter of 4.5 mm) through an extruder, where the cooling medium was water of RT. These wires were ultimately reduced to 2.5 mm by a subsequent drawing. Wire specimens (200 mm in length and 2.5 mm in diameter) were annealed both below and above the eutectoid temperature, i.e., 275 °C, for 30 min, respectively. Then, they were cooled down to RT in air (AC) and water (WQ), respectively. Subsequently, tensile tests were carried out on these specimens (200 mm × 2.5 mm) at RT with a strain rate of 8.3 × 10−3 s−1 using the SANs machine. A scanning electron microscope (SEM, JEOL IT-100, Japan) and a transmission electron microscope (TEM, JEOL 2010, Japan) were used to investigate the microstructures of the Zn(15%)-Al alloys.

To in situ investigate the phase transformation of WQ samples, the polished samples were heat-treated at 295 °C for 30 min, and then quenched by water. Within 5 min, the quenched samples were subjected to SEM recording by use of a mode of backscattered electron imaging. To observe the deformed surface morphology of the alloys, plate-shaped samples with the size of 2 mm × 5 mm × 40 mm were fabricated and heat-treated. Before submitted to tensile deformation, the plate-shaped samples were polished and then etched by a alcoholic solution containing 4% nitric acid.

3. Results

3.1. Extruded specimen

The microstructures of the casted and extruded Zn(15%)-Al alloys were examined by SEM images, and are shown in figure 1. Clearly, the microstructures consist of two phases which are the white area and gray area in the figure and correspond to a η phase (Zn-rich solid solution) and an α phase (Al-rich solid solution), respectively. In the alloy samples, a typical dendrite structure was usually observed in the eutectic matrix, as shown in figure 1(a). In the extruded wires, the equiaxed grains with ~0.3 μm in diameter in the α phase orient along the extrusion direction in the η phase shown in figure 1(b).

Figure 2 shows a typical engineering stress-strained curve for the extruded wires. Unlike conventional engineering alloys, no obvious strain hardening effect can be observed from figure 2. In addition, a large elongation, i.e., 125%, is also noticed.
Figure 3 shows the fractural morphologies of the extruded wires after tensile tests. The average size of the fractural surface is measured to be $\sim 0.4$ mm in diameter in the image shown in figure 3(a), indicating a reduction of about $\sim 97\%$ in area. In addition, dimples are a typical characteristic on the fractural surface, as shown in figure 3(b). These results demonstrate that the extruded wires exhibit super-plasticity when deformed at RT.

3.2. Effects of the annealing temperature

Figure 4 displays the microstructures of the extruded alloys from AC-treated samples and annealed at 250 °C and 325 °C. The size of $\alpha$ phases was markedly increased from $\sim 0.3$ $\mu$m to $\sim 0.9$ $\mu$m while the relatively straight grain boundaries were also show in comparison with that in figure 1(b). However, after annealed at 325 °C, lamellar eutectoid structures mixed with isolated $\eta$ phases are observable, cf figure 4(b).

As shown in figure 5(a), tensile strength keeps increasing slowly with the annealing temperature below 250 °C. When the annealing temperature exceeds 250 °C, a sharp increase is taking place. Figure 5(b) shows a decreasing trend of elongation with increasing the annealing temperature; a sharp drop appears when the annealing temperature exceeds 250 °C. In addition, exceptionally large elongation can be observed for the samples annealed below 250 °C.

3.3. Effects of cooling treatments

During heat treatments, different cooling velocity usually leads to different microstructures and mechanical properties of alloys [30]. In this study, Zn(15%)-Al alloys annealed at 290 °C were cooled in air and water, respectively. Subsequently, these samples were aged at RT and tested by a tensile testing machine. Figure 6 shows the variations of tensile strength with aging time of the samples annealed at 290 °C and followed by AC and WQ, respectively. The tensile strength for the AC samples is little changed with the aging time, but sharply decreases
sharply firstly and tends to be stable after 11 h for the WQ samples. The total reduction in tensile strength for the WQ samples is about 115 MPa during the aging process at RT.

To reveal the underlying mechanisms of strength reduction in WQ samples, in situ microstructure observation by SEM was performed. In order to get better results of the in situ microstructure evolution, pre-polished samples were annealed at 290 °C for 30 min and quenched in water were utilized. The time from quenching to the SEM observation was kept within 5 min. Images captured at 6, 30, 60 and 960 min are shown respectively in figure 7. Examining the images reveals that the microstructure of the WQ samples is composed of two phases: a η phase (bright isolated Zn-rich) and a gray γ phase. Unlike the AC samples (shown in figure 4(b)), no lamellar structure was observed, indicating that the WQ treatment suppressed the eutectoid transformation, i.e. \( \gamma \rightarrow \eta + \alpha' \) at 275 °C. Subsequently, the thermodynamically unstable γ phase then decomposed into η and \( \alpha' \) phase at RT, resulting in fine white and black grains with a size of 100 ~ 200 nm with a prolonged time, cf figures 7(b)–(d). Meanwhile, the bright colored η phase spreading along the extrusion direction kept unchanged during the whole in situ observation period. Therefore, the tensile strength decrement with prolonged aging time observed in the WQ treated samples can be attributed to the decomposition of the thermodynamically unstable γ phase.

3.4. Surface observation of the deformed Zn(15%)-Al alloys
To investigate the surface morphologies of Zn(15%)-Al alloys after deformation, two plate-shaped samples were firstly annealed at 290 °C for 30 min, followed by the WQ and AC treatments, respectively. After 24 h, the samples were polished and deformed to 0.4 and 0.7 strain, respectively. The polished surface morphologies of the two samples were investigated by SEM with a secondary electron imaging mode; the obtained images are shown in figure 8. For the WQ treated sample, cracks are initiated between η grains and at the interfaces between η grains and the granular structures, as indicated by the arrows in figure 8(a). For the AC treated sample, cracks

Figure 3. Fractural morphologies of the extruded Zn(15%)-Al alloy wires after tensile testing: (a) low magnification, (b) high magnification.
are initiated between the lamellas in the lamellar eutectoid structures, which are indicated by the arrows in figure 8(b).

3.5. TEM examination of the deformed Zn(15%) - Al Alloys

As shown in figure 9, TEM observation was carried out to further confirm the relationship between microstructure and the corresponding deformation behaviors of the Zn(15%) - Al samples annealed at 290 °C and followed by the WQ or AC treatments. The samples had been deformed to a 40% plastic strain before observation. No obvious dislocations are found in the WQ treated sample (figure 9(a)). In addition, unusual contrast can be observed in figure 9(a), as indicated by the arrows. However, in the AC treated sample, dislocation tangles are easily observed in the lamellas, as indicated by the arrow in figure 9(b).

3.6. Strength modeling of the WQ Zn-15Al samples

To further understand the strength changes of WQ samples annealed above 275 °C, a model based on the mixed law can be established. The strength changes were proposed to obey the following equation (1):

$$\sigma = \sigma_{E_{Zn}} f_{E_{Zn}} + \sigma_{\gamma} f_{\gamma} + \sigma_{\gamma} (1 - f_{E_{Zn}} - f_{\gamma})$$  \hspace{1cm} (1)

where $\sigma_{E_{Zn}}$ and $f_{E_{Zn}}$ are the strength and volume fraction of the primary eutectoid Zn-rich phase, respectively; $\sigma_{\gamma}$ and $f_{\gamma}$ are the strength and volume fraction of the decomposed $\gamma$ phase, respectively; $\sigma_{\gamma}$ is the strength of $\gamma$ phase. According to the Zn(15%) - Al phase diagram, the volume fraction of primary eutectoid Zn-rich phase is calculated to be 0.275. The values of $\sigma_{E_{Zn}}$, $\sigma_{\gamma}$ and $\sigma_{\gamma}$ were measured to be 158 MPa, 190 MPa and 348 MPa, respectively, by use of Zn(1%) - Al and Zn(22.3%) - Al alloys. The last variable $f_{\gamma}$ is proposed to comply with Avrami equation [33]. Therefore, based on the values estimated from the SEM images, $f_{\gamma}$ was fitted to be the following equation:

![Figure 4. Microstructures of Zn(15%) - Al alloy: AC treated samples annealed at (a) 250 °C and (b) 325 °C.](image-url)
Figure 5. Variations of the ultimate tensile strength (a) and of the elongation (b) with the annealing temperature.

Figure 6. Variations of tensile strength with aging time at room temperature of AC and WQ treated samples annealed at 290 °C.
The measured $f_{v,t}$ and corresponding fitted curve is shown in figure 10(a). Substituting equation (2) into equation (1), tensile strength evolution of the WQ treated samples with prolonged aging time can be expressed by equation (3):

$$\sigma(t) = \sigma_{EZn} f_{EZn} + \sigma_v 0.725[1 - \exp(-(0.725t)^{1.35})] + \sigma_0 (1 - f_{EZn} - 0.725[1 - \exp(-(0.725t)^{1.35})])$$

(3)

The predicted tensile strength according to equation (3) and measured values are shown in figure 10(b). The good fit indicates that equation (3) can predict accurately the tensile strength evolution with aging time for the Zn(15%)-Al alloys annealed at 290 °C and quenched in water.

4. Discussion

An exceptionally large elongation and reduction of area in Zn(15%)-Al alloys were observed. Meanwhile, unlike conventional engineering alloys, almost no strain hardening effect is found in the stress-strain curves for the alloys prepared in this work. In addition, no obviously dislocation tangle structures were observed in the Zn(15%)-Al alloy with a 0.4 plastic strain (figure 9(a)), indicating that dislocation slip could not be the dominant step in the deformation mechanism. It is well known that plastic deformation is usually accomplished by sliding and rotating of equiaxed grains along their boundaries in the superplastic alloys \[34-36\]. Therefore it can be inferred that the Zn(15%)-Al alloys prepared are typical super-plastic alloys.

According to the Zn–Al phase diagram \[37\] shown in figure 11, the microstructure of Zn(15%)-Al alloys are composed of $\eta$ grain (rich in zinc) and $\gamma'$ structure (the decomposed $\gamma$ phase). The $\gamma'$ structure is composed of $\alpha$ particles (rich in aluminum) and $\eta$ matrix shown in figure 4. Based on the lever rule, the weight fraction of $\gamma'$ can be calculated to be 56.7 wt%. According to the phase diagram and the microstructures observed by SEM, the size distribution and volume fraction of $\eta$ grains are basically in the same level under all the heat treatments. Therefore, the $\gamma'$ phase may play an important role in the mechanical performance of the alloys. Subsequently,
the different morphologies of $\gamma'$ phase, where granular and lamellar structures were observed in the WQ and AC treatments, should be responsible for the differences in the mechanical properties of the heat-treated alloys.

In the $\gamma'$ structure (figure 8), the interface sliding between $\alpha$ particles and $\eta$ matrix should be the dominant way in deformation mechanism. In other words, every $\alpha$ particle is a deformation element since the deformation always occurs around its boundaries. Therefore, the number density of $\alpha$ particles in $\gamma'$ phase (i.e. the number of $\alpha$ particles in a unit volume of $\gamma'$ phase) is assumed to be an important parameter to influence the mechanical properties of Zn(15%)-Al alloys. From the above assumption, it can be inferred that in the samples with a larger number density of the $\alpha$ particles, deformation should be more favorable, leading to a lower tensile strength and a higher elongation. On the contrary, an increased tensile strength and a decreased elongation could be observed when the number density of $\alpha$ particles in $\gamma'$ is decreased. According to the Zn–Al phase diagram, the solid solubility of Zn in the $\alpha$ phase increases with temperature below 275 °C, resulting in a continuous size increment of $\alpha$ particles. The diameter of spherical $\alpha$ particles increased from 0.2 μm to 0.5 μm when annealing temperature increased to 250 °C (figures 1(b) and 4(a)). Based on the particle size estimated from the SEM images and the phase diagrams, the number density of $\alpha$ particles in $\gamma'$ phase decreases from 100 to 8.2 $\mu$m$^{-3}$ when annealing temperature increased to 250 °C (figures 1(b) and 4(a)). Based on the particle size estimated from the SEM images and the phase diagrams, the number density of $\alpha$ particles in $\gamma'$ phase decreases from 100 to 8.2 $\mu$m$^{-3}$ when annealing temperature increased to 250 °C, leading to a continuous increment in tensile strength and a decrement in elongation shown in figure 5. Nevertheless, the strength increment disappeared slowly in the next several weeks, due to shrinkage of Al-rich grains caused by the exsolution of Zn atoms from Al-rich grains at RT.

When the annealing temperature of the alloys is higher than 275 °C (the eutectoid temperature), the subsequent cooling step affects dramatically the decomposition behavior of the $\gamma$ phase [38]. When the alloys are cooled quickly by water, the decomposition of $\gamma$ phase was suppressed, leading to thermodynamically unstable $\gamma'$ phase at RT. Subsequently, the unstable $\gamma'$ phase decomposed slowly at RT, resulting in formation of fine granular microstructures, as shown in figure 7. Concurrently the strength decreased slowly with aging time, which was observed to comply with Avrami relationship, as shown in figure 10(b). However, when the alloys are
cooled slowly, the $\gamma$ phase decomposed immediately at relatively high temperature, generating coarse lamellar structures, as shown in figure 4(b). It can be attributed to a higher atom diffusion velocity at higher temperature. For the AC treated samples, the number density of plate-shaped $\alpha$ particles in $\gamma'$ is calculated to be $0.4 \, \mu m^{-3}$ based on the lever rule and particle size estimated from figure 4(b), which are high in tensile strength and low in elongation. The number density of $\alpha$ particles in $\gamma'$ is summarized in figure 12, which always decreases with increasing the annealing temperature.

The TEM image for the alloy with $\sim 0.4$ plastic strain is very complex. In the WQ treated alloys, no dislocated structure could be observed. However, unusual contrast is found near grain boundaries and triple grain junctions indicated by the arrows in figure 8(a), which may be caused by a disordered atom arrangement after a drastic grain boundary sliding and a grain rotation during large plastic deformation. However, in the AC treated samples, dislocation tangle structures could be easily found indicating that, besides grain boundary sliding and rotation, dislocation slip is also an important deformation mechanism. Compared to the high number density of $\alpha$ particles in $\gamma'$ phase of the original sample ($100 \, \mu m^{-3}$), the number decreased more than 200 times in the AC treated samples ($\sim 0.4 \, \mu m^{-3}$); in this case, the interface sliding between the plate-shaped $\alpha$ particles and $\eta$ matrix in $\gamma'$ cannot accommodate the plastic deformation completely. Therefore, dislocation slip has to act as a supplementary deformation mechanism, leading to dislocation tangle structures. For all the Zn(15\%)-Al alloys, a higher tensile strength leads to a lower elongation, which is consistent with conventional engineering materials.

5. Conclusions

(1) When the Zn(15\%)-Al alloys with equiaxed microstructure have an exceptionally large elongation; no strain hardening and dislocation slipping are observed, thus charactering superplastic properties of the alloys.
Figure 10. Modeling results of the $\gamma$ phase decomposition and the corresponding tensile strength change with aging time for the Zn(15\%)–Al alloys annealed at 290 °C and quenched in water: (a) measured volume fraction of decomposed $\gamma$ phase and the corresponding fitted curve; (b) predicted tensile strength and the measured values.

Figure 11. Zn–Al phase diagram.
When the extruded alloy wires were submitted to annealing, a tensile strength increment and an elongation decrement were observed with increasing the annealing temperature. The main reason leading to this evolution in mechanical properties is the decrease of the number density of $\alpha$ particles in $\gamma'$ microstructure.

For the Zn$_{15\%}$-Al alloys annealed above 275 °C, the AC treatment resulted in a lamellar microstructure. However, the QW treatment suppressed the decomposition of $\gamma$ phase at relatively high temperature, which decomposed subsequently at RT, resulting in formation of fine and equiaxed grains. Dramatic decrement in the strength occurred during the decomposition of $\gamma$ phase.

For the Zn$_{15\%}$-Al alloy wires annealed above 275 °C and quenched in water, decomposition dynamics of the unstable $\gamma$ phase at RT was proposed to comply with Avrami equation. As such, a strength model based on the mixed law was established, which predicted the evolution of tensile strength with aging time very well at RT.

Dislocation tangles were easily observed in the Zn$_{15\%}$-Al alloys with a lamellar microstructure, demonstrating that the dislocation slip was an important deformation mechanism. However, in the Zn$_{15\%}$-Al alloys with the equiaxed microstructure, no dislocation tangle structure could be found. This can be attributed to the very small value of the number density of $\alpha$ particles in $\gamma'$ microstructure, i.e., 0.3 $\mu$m$^{-3}$, which cannot accommodate the plastic deformation. Dislocation slip has to act as a supplementary deformation mechanism.

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References

[1] Zhu Y H, Man H C, Dorantes-Rosales H J and Lee W B 2003 Ageing characteristics of furnace cooled eutectoid Zn–Al based alloy J. Mater. Sci. 38 2925–34
[2] Lo S H J, Dionne S, Sahoo M and Hawthorne H M 1992 Mechanical and tribological properties of zinc-aluminium metal-matrix composites J. Mater. Sci. 27 681-91
[3] Yao H Z and Islas J J 1997 Microstructures and dimensional stability of extruded eutectoid ZnAl alloy J. Mater. Process. Technol. 66 244–8
[4] Zhu Y H, Hernandez R M and Baños I L 1999 Phase decomposition in extruded Zn–Al based alloy J. Mater. Sci. 34 3653–8
[5] Elkhair A and Daoud T M 2004 Effect of different Al contents on the microstructure, tensile and wear properties of Zn-based alloy Mater. Lett. 58 1754–60
[6] Demirats M, Purcek G, Yanar H, Zhang Z J and Zhang Z F 2015 Improvement of high strain rate and room temperature superplasticity in Zn–22Al alloy by two-step equal-channel angular pressing Materials Science & Engineering A 620 233–40
[7] Demirats M, Purcek G, Yanar H, Zhang Z J and Zhang Z F 2015 Achieving room temperature superplasticity in Zn–5Al alloy at high strain rates by equal-channel angular extrusion J. Alloys Compd. 623 213–8
[8] Ha T K, Son J R, Lee W B, Park C G and Chang Y W 2001 Superplastic deformation of a fine-grained Zn–0.3 wt%Al alloy at room temperature Materials Science and Engineering: A 307 98–106
[9] Bobzin K, Ote M and Knobloch M A 2017 Surface pre-treatment for thermally sprayed Zn–15Al coatings J. Therm. Spray Technol. 26 1–9
[10] Yan X, Liu S, Long W, Huang J, Zhang L and Chen Y 2013 Stress corrosion crack of Zn–15Al alloys in hot and humid environment Mater. Lett. 93 183–6
[11] Bobzin K, Ote M and Knobloch M A 2017 Designing the corrosion products of Zn–15Al: a new approach to smart corrosion protection coating Corros. Sci. 155 217–29
[12] de Rincón O, Rincón A, Sánchez M, Romero N, Salas O, Delgado R, López B, Uruchurtu J, Marroco M and Ponsiano Z 2009 Evaluating Zn, Al and Al–Zn coatings on carbon steel in a special atmosphere Constr. Build. Mater. 23 1465–71
[13] Liu L, Chen Z, Zhou Z, Chen G and Liu C 2017 Diffusion barrier property of electrosheet Ni–W–P coating in high temperature Zn–5Al/Cu solder interconnects J. Alloys Compd. 722 746–52
[14] Nagooka T, Morisada Y, Fukusumi M and Takemoto T 2011 Selection of soldering temperature for ultrasonic-assisted soldering of 9506 aluminum alloy using Zn–Al system solders J. Mater. Process. Technol. 211 1334–9
[15] Guo W, Luan T, He J and Yan J 2017 Ultrasonic-assisted soldering of fine-grained 7034 aluminum alloys using ZnAl filler metals Mater. Des. 125 85–93
[16] Prasad K J W B 2004 Influence of Heat Treatment Parameters on the Lubricated Sliding Wear Behaviour of a Zinc-based Alloy Mater. Sci. Forum 257 1137–44
[17] Zhang Y, Yang L, Zeng X, Zheng B and Song Z 2013 The mechanism of anneal-hardening phenomenon in extruded Zn–Al alloys Mater. Des. 50 223–9
[18] Senkov O N and Myshlyaev M M 1986 Grain growth in a superplastic Zn–22%Al alloy Acta Metall. 34 97–10
[19] Yang C F, Pan J H and Lee Y H 2009 Work-softening and anneal-hardening behaviors in fine-grained Zn–Al alloys Journal of Alloys Compd 468 230–6
[20] Lin G et al 2020 Microstructural and mechanical properties of ZnAl alloy tubes and their weld seams prepared by Conform continuous extrusion Rare Metal 39 707–15
[21] Hernández-Rivera J L, Flores E E M, Contreras E R, Rocha J G, Cruz-Rivera J J and Torres-Villaseño G 2017 Evaluation of hardening and softening behaviors in Zn–21Al–2Cu alloy processed by equal channel angular pressing Journal of Materials Research and Technology 6 329–33
[22] Zhang L, Ding X, Wei Y, Man Z and Song Z 2018 Effect of prestrain on precipitation behaviors of Ti–2.5Cu alloy High Temperature Materials Processes 37 467–93
[23] Yan X, Liu S, Long W, Huang J, Zhang L and Chen Y 2013 The effect of homogenization treatment on microstructure and properties of Zn-15Al solder Materials Design 45 (none) 440–5
[24] Zhu Y H 2001 Phase transformations of eutectoid Zn–Al alloys J. Mater. Sci. 36 3973–80
[25] Zhu Y H, Lee W B and To S 2003 Ageing characteristics of cast Zn–Al based alloy (ZnAlCu,) J. Mater. Sci. 38 1945–52
[26] Lozenko V V and Shepelevich V G 2009 Structure and mechanical properties of rapidly solidified foils of Zn–Al alloys Physics of Metals, Metallography 107 370–4
[27] Demirats M, Kawasaki M, Yanar H and Purcek G 2018 High temperature superplasticity and deformation behavior of naturally aged Zn–Al alloys with different phase compositions Materials Science & Engineering A 730 73–83
[28] Gu M, Chen Z, Wang Z, Jin Y, Huang J and Zhang G 1994 Damping characteristics of Zn–Al matrix composites Scr. Mater. 30 1321–6
[29] Kushibe A, Takigawa Y, Higashi K, Aoki K, Makii K and Takagi T 2007 Application of high-strain-rate superplastic Zn–Al alloy to seismic dampers and its optimised shape design Materials Science Forum 552 552–583–90
[30] Sandovo- Jiménez A, Negrete J and Torres-Villaseño G 2010 Phase transformations in the Zn–Al eutectoid alloy after quenching from the high temperature triclinic beta phase Mater. Charact. 61 1286–9
[31] Ha T K, Koo H W and Chang Y W 2003 Temperature and grain size dependence of superplasticity in a Zn–0.3 wt%Al alloy Met. Mater. Int. 9 29–35
[32] Demirats M, Purcek G, Yanar H, Zhang Z J and Zhang Z F 2016 Effect of natural aging on RT and HSR superplasticity of ultrafine grained Zn–22Al alloy Materials Science Forum 838–839 320–5
[33] Song Z Y, Sun Q Y, Xiao L, Liu L, Wang H, Chen W and Sun J 2009 The influence of prior cold deformation on precipitation of alpha phase and variation of hardness in Ti-10Mo-8V-1Fe-3.5Al during aging treatment J. Mater. Res. 24 452–8
[34] Tokunaga T and Ohno M 2019 Microstructure evolution during superplastic deformation of an Al-coated Mg alloy sheet J. Alloys Compd. 805 436–43
[35] Sayf F, Mahmudi R and Roumina R 2020 Inducing superplasticity in extruded pure Mg by Zr addition Materials Science and Engineering: A 769 138502
[36] Pozdniakov A V, Yu, Barklov B, Amer S M, Levchenko V S and Mikhailovskaya A V 2019 Microstructure, mechanical properties and superplasticity of the Al–Cu–Y–Zr alloy Materials Science and Engineering: A 758 26–35
[37] Okamoto H 1995 Al–Zn (aluminum–zinc) J. Phase Equilib. 16 281–2
[38] Oliveira J P, Santos T G and Miranda R M 2020 Revisiting fundamental welding concepts to improve additive manufacturing: from theory to practice Prog. Mater. Sci. 107 100590