Materials Research Express

PAPER

Effect of Sn addition on the microstructure and mechanical properties of AZ31 alloys

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Keywords: Mg–Al–Zn–Sn alloy, microstructural evolution, strengthening mechanisms, work hardening

Abstract

An Mg–3Al–1Zn–xSn (x = 0, 3, 6, 9) alloy was prepared by die-casting and analyzed by XRD, SEM, and EBSD. The microstructure, second phase, and grain orientation of the AZT31x alloy were characterized. As the Sn content increased from 3 wt.% to 9 wt.%, the tensile and yield strength of the alloy were effectively improved. With the addition of Sn, the grain size of alloys decreases gradually blocking the dislocation by the grain boundary and the dispersion of the Mg2Sn second phase in the AZT31x alloy contributes to the strength via grain boundary pinning. According to theoretical analysis and calculation, the high strength of AZT319 alloy is partly attributed to the grain fine strengthening (σt90 = 13.76 Mpa) and second phase strengthening (Δσ = 9.27 ~ 12.67 MPA). The total increased strengthening value is lower than experimental value (Δσ = 31.82 MPA). The ratio about τprio / τbasal and τc+a / τbasal in die-cast alloys tensioned to 0.08 deformation gradually decrease, which can reflect that high-Sn content contributes to the strain hardening behavior. The ductility of AZT313 alloy was lightly improved due to the {10-12} tensile twins. When excessive Sn was added, the Mg2Sn second phase coarsened and acted as the nucleus of micro-cracks during the stretching process, thereby reducing the ductility of the alloy.

1. Introduction

As modern transportation becomes lighter and faster, the raw materials used for their manufacture require high-damping properties with vibration and noise reduction characteristics [1, 2]. Magnesium and magnesium alloys with high specific strength and excellent vibration resistance have become a dominant research area in the development of advanced materials. Unfortunately, most magnesium alloys display low strength, which severely limits their application as structural components [3]. As a conventional cast magnesium alloy, Mg–Al–Zn alloys are widely used in automobiles and electronics due to their low cost, high specific strength, and acceptable corrosion resistance. High Zn–content alloys are prone to hot cracks during the casting process, therefore, the content of Zn in the Mg–Al–Zn alloy is usually less than 1 wt.%. This has led to the development of commercial AZ alloys, such as AZ31, AZ61, AZ91, and AM60 magnesium alloys [4]. In recent years, researchers have been interested in Mg–Sn alloys due to their low cost and potential high performance at both room and high temperatures. The solubility of Sn in the α-Mg solid solution drops sharply from 14.85% at the eutectic transition temperature of 561 °C to 0.45% at 200 °C. In addition, Mg2Sn (fcc) is a hard phase with a melting point of 771.5 °C [5]. Li [6] found that the addition of Sn can form the discontinuous precipitation phase at the grain boundary of the state-aged AZ91D alloy. When the amount of Sn reaches 2 wt.%, the discontinuous precipitation phase disappears. Sevik [7] studied the microstructure and mechanical properties of the AM60 alloy with 0.5 ~ 4 wt.% Sn, and the results showed that the addition of 0.5 wt.% Sn significantly improved the tensile strength. The AMT604 alloy has the best tensile strength of 212 MPa reported so far. Kim [8] found that adding 3 ~ 5 wt.% of Sn to alloy AZ51 can introduce twins in the α-Mg grains and inhibit the initiation and propagation of cracks to improve the tensile and fracture properties based on in situ fracture observation.
Recently, Turen [9] found that after adding 0.5 wt.% of Sn to the AZ91 magnesium alloy, the tensile strength and elongation increased, and the improvement of the mechanical properties was attributed to the addition of Sn, which transformed the lamellar phase to fully divorced eutectic $\beta$ phases. Overall, Sn is an effective additive, which can improve the mechanical properties of AZ alloys. However, the influence of high Sn content on the microstructure and mechanical properties of die-cast AZ alloys is still unclear.

In this study, a commercial AZ31 magnesium alloy was selected as the base alloy for die-casting experiments due to its moderate castability and excellent mechanical properties, and its microstructure and phase composition with 3–9 wt.% Sn content was studied to understand the effect of Sn on the solidification structure, precipitation strengthening, and tensile properties.

2. Materials and methods

Mg-3Al-1Zn-xSn ($x = 0, 3, 6, 9$ wt\%) alloys were prepared by a die-casting method, using pure using industrial pure magnesium (99.95 wt.\%), high-purity Al (99.9 wt.\%), high-purity Zn (99.99 wt.\%), high-purity Sn (99.99 wt.\%) and commercial AZ31 alloy as raw materials, under the protection of Ar and SF$_6$ mixed gas. The melt was placed in a resistance furnace at 710 °C and kept for 20 min. The pouring temperature was 680 °C and the mold...
temperature was 180 °C. The injection speed was set at 6 m s⁻¹ during the die-cast process. Eventually, the nominal and measured compositions of die-cast alloys are shown in Table 1.

The microstructure of the casting specimen was observed using an optical microscope (OM, LEICA DMI3000M) and a scanning electron microscope (SEM) (Zeiss ZEISS-6035 field-emission emission) equipped with energy-dispersive x-ray spectroscopy (EDS). The second phase of the casting sample was examined using a Brooke XD8 ADVANCE A25 x-ray diffractometer (XRD) with a copper target Kα line at a scan rate of 0.02 ° s⁻¹ and 2θ range from 20° to 90°. EBSD data were obtained from a Zeiss ZEISS-6035 field-emission scanning electron microscope (SEM) equipped with TSLMSC. All samples were fabricated by the same process, and the sampling positions of the tension samples are taken from the center of the die-cast alloy. In addition, the size of standard tensile samples (a gauge size of 15 mm × 4 mm × 2 mm) were machined by electrical discharge machining (figure 1), and the tensile tests were carried out using a universal testing machine (Instron-5982) at room temperature with a tensile rate of 0.2 mm min⁻¹. Moreover, each step of the tension was repeated more than three times to ensure the accuracy of the experimental data.

3. Results

3.1. Microstructure

Figure 2 shows the XRD pattern of the AZT31x (x = 0, 3, 6, 9) alloy. It can be observed that the phase of the alloy is composed only of α-Mg, but the alloy with Sn presents both α-Mg and Mg₂Sn phases. The Mg₂Sn diffraction peak is not obvious when the Sn content is less than 3 wt.%, because the Sn atoms may be dissolved in the α-Mg matrix mostly during the solidification process. When the content of Sn in the AZT31x alloy is 6 wt.%, the diffraction peak of Mg₂Sn is easily detected, which appears at a 2-Theta angle of 23°. When the content of Sn is 9 wt.%, the intensity of the Mg₂Sn diffraction peak increases, which indicates that the volume fraction of the Mg₂Sn phase increases with increasing Sn content. The electronegativity of Mg, Al, Zn, and Sn are 1.31, 1.61, 1.65, and 1.96, respectively. Since the difference between Mg and Sn is the largest in the binary system, no Zn-Sn compound is formed, but the Mg₂Sn phase is formed first in the Mg-Al-Zn-Sn system. Wang et al. [10] found that Zn only exists in the α-Mg matrix in the form of a solid solution when the ratio of Zn/Al in Mg-Al-Zn alloy is less than 3, and the Mg-Al-Zn ternary phase cannot be formed.

The low-magnification microstructure of the die-cast AZT31x (x = 0, 3, 6, 9) alloy is shown in figure 3. The micro-structure of the alloys is typical of the die-cast magnesium alloy structure with a bimodal grain size distribution. Non-equilibrium solidification conditions caused non-uniform distribution of the alloying elements in the alloy microstructure. The size of the α-Mg large crystals strongly decreases with an increase in Sn weight fraction in the alloys. And the average grain size statistical analysis was performed on AZT319x (x = 0, 3, 6, 9) alloys is about 10.1 μm, 8.4 μm, 6.2 μm and 6.8 μm by Image Pro, respectively. The irregular morphology of the second phase is mainly distributed at the grain boundaries, and part of the spherical second phase is distributed inside the grains. As the amount of Sn added increases, the number of second phases in the alloy increases significantly. Meantime, it can be observed that the dendrite distribution of the AZT31x alloy gradually
becomes uniform with increasing Sn content. When the weight fraction of Sn reaches 9 wt.%, the distribution of the Mg2Sn second phase around the grain boundary is half-continuous, the precipitated second phase is coarsened, and the number density is higher.

In die-cast alloy AZT31x \((x = 0, 3, 6, 9)\), the second phase is distributed along the grain boundaries (discontinuous precipitation). In the high-magnification SEM image of alloy AZT31x \((x = 0, 3, 6, 9)\) (figure 4) the characteristics of these Mg2Sn phase precipitates can be observed more clearly, pointed by the yellow arrow in the figure. As the Sn content increases, the number of Mg2Sn phases increases, which is in good agreement with the XRD results. Therefore, it can be assumed that Sn atoms are mainly solid-dissolved in the AZT313 alloy. As the Sn content increases to 6 wt.% and 9 wt.%, a large amount of precipitation forms the second phase.

![Figure 4](image)

**Figure 4.** High-magnification SEM images of die-cast alloys (a) AZ31, (b) AZT313, (c) AZT316 and (d) AZT319.

![Figure 5](image)

**Figure 5.** Pole figures of the \{0001\}, \{10-10\}, and \{11-20\} planes showing the textures of die-cast AZT31x alloys tensioned to 0.08 (a) AZ31, (b) AZT313, (c) AZT316, and (d) AZT319.
composed of Mg$_2$Sn. In addition, it can be seen that the relatively coarse range of Mg$_2$Sn particles is 1 ~ 3 $\mu$m. Therefore, when the Sn content is too high, it may lead to the formation of coarsened Mg$_2$Sn particles [11]. A Mg$_{17}$Al$_{12}$ phase also can be observed in the high-magnification SEM image which differs from the XRD results. The addition of Sn also leads to changes in the morphology and quantity of the Mg$_{17}$Al$_{12}$ phase, as shown in figures 4(a)–(d). For the AZT31$_x$ ($x = 0, 3, 6, 9$) alloys, most discontinuous Mg$_{17}$Al$_{12}$ precipitates are elliptical along the grain boundary and of sizes less than 1 $\mu$m. With increasing Sn content, the number of Mg$_{17}$Al$_{12}$ particles in the ATZ361 and ATZ391 alloys decreases and is gradually sparsely distributed. The suppression of discontinuous Mg$_{17}$Al$_{12}$ precipitates by Sn is mainly due to the richness of the Mg$_2$Sn phase at the grain boundary, which limits the nucleation and growth of discontinuous Mg$_{17}$Al$_{12}$ precipitates. Li et al [6] found that the AZT641 alloy contains the least Mg$_{17}$Al$_{12}$ precipitate phases, which shows that a 1.2% Sn content can inhibit the precipitation of the Mg$_{17}$Al$_{12}$ phase.

To further understand the strengthening mechanism of the AZT31$_x$ ($x = 0, 3, 6, 9$) alloys, the microstructure and texture evolution with different Sn content were analyzed by EBSD. The texture is an important strengthening mechanism in magnesium alloys. Magnesium alloys usually have grains with a preferred crystal orientation due to unidirectional deformation, recrystallization (dynamic/static), and possible grain growth [12, 13]. The texture of alloys with different Sn content at 0.08 deformation is shown in figure 5, which reveals that the texture of all samples is similar. When a Mg alloy is tensile-deformed at room temperature, the base-plane slip becomes the main deformation system, because the [10-12] tensile twins are not conducive to deformation and the critical shear stress (CRSS) of [10-11] compression twins and non-base slip are much higher than the CRSS of base-plane slip [14, 15]. The addition of Sn affects the texture strength of
the alloy lightly. The basal texture of the die-cast AZT31x (x = 0, 3, 6, 9) alloy gradually increases with increasing Sn content, and the maximum strength value of the basal texture increases from 2.03 to 2.26. In addition, the basal surface texture gradually becomes more concentrated. In general, the strength of the texture is affected by the deformation of the alloys, alloying elements, and grain size [16–18]. A stronger basal texture can improve the yield strength of alloy ATZ31x (x = 3, 6, 9) by texture strengthening.

Figure 6 shows the EBSD microstructure and misorientation angle distribution of the die-cast AZT31x (x = 0, 3, 6, 9) alloy with 0.08 deformation. When the strain is 0.08, the coarse crystal grains are largely deformed, and twins are observed in a part of the coarse crystal grains, but almost no twins are observed in the fine crystal grains. The average misorientation angle is mainly concentrated at low-angle GBs (LAGBs < 10°), while the distribution of high-angle GBs (HAGBs > 10°) is extremely low (figures 6(b), (d), (f), (h)). In addition, the ratio of the average misorientation angle in LAGBs decreases with increasing Sn content. Specifically, the peak value of LAGBs decreases to 0.37, and the HAGBs gradually increases, but the misorientation angle peak at 86° decreases accordingly (figures 6(d), (f), (h)). The peak of the misorientation angle present at 86° confirms the appearance of twins (figures 6(a), (c), (e), (g)). When the stretching is 0.08, the twins in the original coarse grains are more likely to be consumed after the crystal grains are broken to a certain extent with increasing Sn content, resulting in a decrease in the number of twins.

The type of twins and their proportion during the stretching process have an important influence on the tensile properties and texture evolution of the alloys. Therefore, it is necessary to further study the twin of alloy AZT31x (x = 0, 3, 6, 9) during deformation. The twin boundary diagram and the distribution of the integral number of twins of the AZT31x alloy are shown in figure 7. The observed [10-12] tensile twins formed during the deformation process are drawn with red lines and the [10-11] compressed twins are marked with a green line. More [10-12] tensile twins were observed in the AZT31x (x = 0, 3, 6, 9) alloys, indicating that these are the main form of twins during the uniaxial tension of the die-cast AZT31x (x = 0, 3, 6, 9) alloy. This is

![Figure 7. Twin boundary maps of die-cast AZT31x alloys tensioned to 0.08 (a) AZ31, (b) AZT313 (c) AZT316, (d) AZT319, and (e) area fraction distribution of twin.](image)

![Figure 8. Tensile engineering stress-strain curves of the die-cast AZT31x(x = 0, 3, 6, 9) alloys.](image)
because the CRSS of the \{10\-12\} tensile twins is lower than that of \{10\-11\} compression twins, and can coordinate deformation on the \(c\)-axis under normal tensile conditions. Therefore, \{10\-12\} tensile twins play a momentous role in the plastic deformation of Mg alloys [19]. It can also be seen that the twin area fraction in the AZT31\(x (x = 3, 6, 9)\) alloy decreases with increasing Sn content. Kim [8] found that adding \(3 \sim 5\) wt.% of Sn to alloy AZ51 can introduce twins in the \(\alpha\)-Mg grains and inhibit the initiation and propagation of cracks to improve tensile and fracture properties. Zhao et al [20, 21] also reported similar observations, suggesting that this may be caused by grain refinement. Because finer grains in the alloy have a large number of grain boundaries, the growth and expansion of twins is affected.

3.2. Mechanical performance analysis
The typical engineering tensile stress-strain curve of the die-cast AZT31\(x (x = 0, 3, 6, 9)\) alloy is shown in figure 8. The addition of Sn can increase the yield strength (YS) and ultimate tensile strength (UTS) at the same time. On the contrary, the elongation (EL) will decrease to a certain extent. In the AZT313 alloy, YS, UTS, and EL increase, but when the Sn content increases to 6 wt.% and 9 wt.%, UTS increases from 191.9 MPa to 224.8 MPa, YS increases from 107.9 MPa to 135.3 MPa, and EL is reduced from 20% to 16%. The addition of excessive Sn adversely affects the ductility of the alloy, maybe an excessive amount of Sn results in the coarsening of the Mg,Sn and Mg\(_{17}\)Al\(_{12}\) phases. These coarsened second phases can prevent grain boundary slippage during tensile deformation, but stress concentration is more likely to occur at the contact position with the grain boundary, causing microcracks. It has been reported that excessive Sn (exceeding the maximum solubility of 4.75 wt.% at 468 °C) will result in the formation of coarse undissolved Mg,Sn particles [22], which may act as the nucleus of micro-cracks and easily induce crack initiation, causing the alloy to fracture under stress. Cracking reduces the plasticity of the alloy. This can be seen from figure 4, especially for ATZ31\(x (x = 6, 9)\) alloys, when considering the relatively large drop in elongation compared with alloy AZ31. The ATZ319 alloy is considered to have the best mechanical properties (224.8 MPa UTS, 135.3 MPa YS, and 16%EL) among the experimental alloys.

4. Discussion
The change of dislocation density during plastic deformation has a great influence on the tensile behavior of the alloy, and the dislocation entanglement caused by the increased dislocation density during alloy deformation becomes an obstacle to the movement of dislocations [23]. To discuss the influence of grain size and dislocation density in the microstructure on the strength of the alloy, the \(\sigma\) can be calculated as follows [24, 25]:

\[
\sigma = \sigma_0 + \sigma_{H_P} + \sigma_d
\]  

(1)

Where \(\sigma_0\) is the friction coefficient, \(\sigma_{H_P} = kd^{-1/2}\) is the Hall-Petch equation, and \(\sigma_d = \frac{M \sqrt{G \psi b}}{d}\) is the Taylor dislocation; where \(\psi\) is the dislocation density, \(\eta\) is a constant, \(M\) is the Taylor factor, \(G\) is the shear modulus, and \(b\) is the Burgers vector. Since \(\sigma_{0.2} = \sigma_0 + kd^{-1/2}\), equation (1) can be written as:

\[
\sigma \approx \sigma_{0.2} + \sigma_d
\]  

(2)

Then the stress contribution related to the dislocation density can be obtained as:

\[
\rho^{1/2} \propto \sigma_d \approx \sigma - \sigma_{0.2}
\]  

(3)

This means that the applied stress required during the plastic deformation of the alloy is proportional to the dislocation density inside. Therefore, the second phase, grain size and grain orientation factors that affect the internal dislocations of the grain influence the tensile behavior of the alloy.

4.1. Effect of the second phase on mechanical properties
It can be seen from figure 3 that a large amount of Mg,Sn hard phase is dispersed and distributed in the AZT31\(x (x = 3, 6, 9)\) alloy. These precipitates effectively hinder the movement of dislocations and the sliding of grain boundaries during the stretching process, so the first two-phase strengthening plays an important role in the improvement of YS. With the addition of Sn, the precipitation of Mg,Sn increases, as can be seen from figure 3, especially for the ATZ319 alloy. Moreover, the increase in the number of well-dispersed second phases in the grains can prevent the movement of dislocations and hinder their recovery, which is beneficial for the improvement of the YS of ATZ31\(x (x = 3, 6, 9)\) through the strengthening of the second phase. It can be observed the ATZ319 alloy has higher YS than AZ31. The strengthening of the second phase mainly includes [26]: (1) the dislocation bypass mechanism (Orowan strengthening mechanism), \(\Delta \sigma_{Ow}\); and (2) the strengthening of the load from the matrix to the second phase particles, \(\Delta \sigma_{PS}\).
Table 2. Parameters for SF and Slip system of die-cast AZT31x(x = 0, 3, 6, 9) alloys.

| Alloy    | \( m_{\text{b}} \) | \( \tau_{\text{bmax}} \) | \( m_{\text{s}} \) | \( \tau_{\text{prom}} \) | \( m_{(\text{s}+\text{t})} \) | \( \tau_{(\text{s}+\text{t})} \) | \( \tau_{(\text{s}+\text{t})}/\tau_{\text{bmax}} \) | \( \tau_{(\text{s}+\text{t})}/\tau_{\text{bmax}} \) |
|----------|-----------------|-----------------|-----------------|-----------------|-----------------|-----------------|-----------------|-----------------|
| AZ31     | 0.3062          | 31.69           | 0.3458          | 66.37           | 0.4488          | 86.14           | 2.09            | 2.72            |
| AZT313   | 0.3066          | 33.08           | 0.3456          | 71.47           | 0.4484          | 92.73           | 2.16            | 2.8             |
| AZT316   | 0.3057          | 39.25           | 0.3466          | 75.80           | 0.4488          | 98.16           | 1.93            | 2.5             |
| AZT319   | 0.3070          | 41.54           | 0.3401          | 76.44           | 0.4479          | 100.67          | 1.84            | 2.42            |

Therefore, the total intensity contribution of the second phase can be computed as:

\[
\Delta \sigma = \Delta \sigma_O + \Delta \sigma_T
\]  

(4)

For magnesium alloys, Zhang et al [27] proposed an improved Oran equation:

\[
\Delta \sigma_O = \frac{0.13Gb}{d_p} \ln \left( \frac{d_p}{2b} \right)
\]

(5)

where \( G \) refers to the shear modulus (magnesium alloys generally take the value \( 1.66 \times 10^{-4} \) MPa); \( b \) is the Berth vector (3.21 \( \times 10^{-10} \) m), \( d_p \) refers to the volume of the second phase particles, and \( d_p \) refers to the average particle size of the second phase particles (figure 3). The value is calculated as \( \Delta \sigma_O = 2.47 \) MPa.

The strengthening of the load from the matrix to the second phase particles can be calculated by the following equation:

\[
\Delta \sigma_T = \frac{1}{2} \sigma_f \cdot f_v
\]

(6)

\( \sigma_f \) refers to the yield strength of the matrix, and generally takes 100 \~ 150 MPa for magnesium alloys [26], and \( f_v \) refers to the volume fraction of the second phase particles. The value is calculated as \( \Delta \sigma_T = 6.8 \sim 10.2 \) MPa. The calculated second phase strengthening is \( \Delta \sigma = 9.27 \sim 12.67 \) MPa.

4.2. Effect of grain size on mechanical properties

It can be observed that with an increase in Sn weight fraction in the alloys, the grain size become smaller gradually in figure 3. Refinement strengthening can be calculated by Hall-Petch relationship:

\[
\sigma_{\text{HP}} = \sigma_0 + k d^{-1/2}
\]

(7)

Therefore, the contribution of AZT319 alloy to the fine-grain strengthening of AZ31 alloy can be calculated as:

\[
\Delta \sigma_{\text{HP}} = k_{mg} (d_{AZT319}^{1/2} - d_{AZ31}^{1/2})
\]

(8)

Where \( d \) \( \mu \)m is the average grain size; \( k_{mg} \) (=220 Mpa \( \cdot \) \( \mu \)m\(^{-1/2}\)) [28] is constants related to the grain boundary structure; The value is calculated as \( \sigma_{\text{HP}} = 13.76 \) MPa. The total value \( \Delta \sigma \) = 23.03 \~ 26.43 MPa of the yield stress predicted by the strengthening mechanism and grain fine strengthening is lower than the experimental value \( \Delta \sigma = 31.82 \) MPa. The deviation can be attributed to the influence of twin deformation or grain orientation.

4.3. Effect of grain orientation on mechanical properties

The activation of slip and \{10-12\} tension twins depend largely on the grain orientation. Generally, the activation of sliding systems and twins follows Schmid’s law [29]. For the uniaxially stretched AZT31x (x = 0, 3, 6, 9) alloy, the measured \( \varepsilon \) and \( \sigma \) are related to the critical shear stress \( \tau \) and shear strain \( \varepsilon \) on the base plane:

\[
\varepsilon = m_{s,b}
\]

(9)

\[
\sigma = \tau / m_{s,b}
\]

(10)

Where \( m_{s,b} \) is the Schmidt factor of base-plane slip, \( m_{s,b} = \cos \lambda - \cos \phi \) (\( \lambda \) is the stress axis and the slip direction angle, \( \phi \) is the stress axis and the normal line of the slip surface’s angle

The CRSS of the base-plane is:

\[
\tau_{\text{basal}} = m_{s,b} \times \sigma_{0.2}
\]

(11)
The deformation of the AZT31x (x = 0, 3, 6, 9) alloy after yielding is generally caused by basal surface slip at the beginning, but in magnesium alloys, prismatic (a) slip and pyramidal (c + a) slip will also be activated as a supplement to basal slip to promote deformation. The CRSS on the prismatic (a) and the pyramidal (c + a) can be obtained by the following equations:

\[ \tau_{\text{prism}} = m_{\text{pr}} \times \sigma \]  
\[ \tau_{(c+a)} = m_{(c+a)} \times \sigma \]

\( m_{\text{pr}} \) is the Schmid factor of prismatic (a) and \( m_{(c+a)} \) is the Schmid factor of pyramidal (c + a) slip. Table 2 shows the average Schmid factor, \( \gamma_{\text{basal}} \), \( \tau_{\text{prism}} \) and \( \tau_{(c+a)} \) of the basal and non-basal slip of the AZT31x (x = 0, 3, 6, 9) alloy. During the stretching process, the die-cast AZT31x (x = 0, 3, 6, 9) alloy exhibits a weaker base (a) slip and a stronger (c + a) pyramidal slip. This demonstrates that the start of the (c + a) pyramidal slip during the alloy stretching process is favored. As Sn content increases, the \( \tau_{\text{prism}}/\gamma_{\text{basal}} \) and \( \tau_{(c+a)}/\gamma_{\text{basal}} \) in AZT31x(x = 0, 3, 6, 9) decreases, affecting the dislocation activity. The value of \( \tau_{\text{non-basal}}/\gamma_{\text{basal}} \) can reflect the difficulty of activation of non-basal slip [30]: the lower the value of \( \tau_{\text{prism}}/\gamma_{\text{basal}} \) and \( \tau_{(c+a)}/\gamma_{\text{basal}} \), the easier it is to activate [31]. Wang et al. [32] found that the SF of the corresponding pyramidal (c + a) increased by 19.8% after 8% strain in AZT643, and (c + a) slip was greatly activated, which can explain that the addition of Sn can enhance non-basal. Moreover, Sn has been shown to significantly reduce the stacking fault energy (SFE) of Mg. The activation of non-basal slip is in favor of dislocation interactions which can accumulate a large number of dislocations in the alloy during uniaxial stretching, which contributes strain hardening ability [33]. The yield strength was increased partly due to the working hardening behavior during tensile deformation.

5. Conclusions

This work studied the effect of Sn content on the microstructure and mechanical properties of die-casting AZT31x (x = 0, 3, 6, 9) alloys. The following conclusions can be drawn from the experimental results:

1. It was found that the average grain size of the die-cast alloys decreases lightly and the tensile strength and yield strength of the AZT31x alloy were effectively improved when the Sn content increases from 0 wt.% to 9 wt.%. With the addition of Sn, the grain size of alloys decreases gradually blocking the dislocation by the grain boundary and the dispersion of the Mg2Sn second phase in the AZT31x (x = 3, 6, 9) alloys contributes to the strength via grain boundary pinning. According to theoretical analysis and calculation, the high strength of AZT319 alloy is partly attributed to the grain fine strengthening (\( \sigma_{fl} = 13.76 \) MPa) and second phase strengthening (\( \Delta \sigma = 9.27 \sim 12.67 \) MPa). The total increased strengthening value is lower than experimental value (\( \Delta \sigma = 31.82 \) MPa).

2. According to theoretical analysis and calculation, the ratio about \( \tau_{\text{prism}}/\gamma_{\text{basal}} \) and \( \tau_{(c+a)}/\gamma_{\text{basal}} \) in die-cast alloys tensioned to 0.08 deformation gradually decrease, which can reflect that high-Sn content contributes to the strain hardening ability relating yield strength. The AZT319 alloy has excellent mechanical properties at room temperature, with a YS of 135.32 MPa and an UTS of 224.77 Mpa.

3. The ductility of the AZT313 alloy was improved due to the large increase in {10-12} tensile twin during tensile deformation. When excessive Sn is added, the Mg2Sn second phase is coarsened, which may act as a nucleus for microcracks during the stretching process and reduce the ductility of the alloys.

Acknowledgments

This research was sponsored by Qinghai Provincial Basic Research Program(2020-ZJ-707).

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