Microstructure and enhanced mechanical properties of Mg-3Sn alloy with Mn addition

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Abstract

The effects of Mn content on the microstructure and mechanical properties of the extruded Mg-3Sn-xMn (x = 0, 0.5, 1.5, 2.5) alloys was systematically investigated in this study. More importantly, the relationship between microstructure and mechanical properties of Mg-Sn-Mn alloy was revealed in detail by calculating the various strength contribution value. The microstructure and mechanical properties of the alloys were characterized by x-ray diffraction (XRD), scanning electron microscope (SEM), electron backscatter diffraction (EBSD) and universal testing machine. The results revealed that the average grain size (AGS) decreased from 21.45 μm to 10.51 μm and then increased to 13.41 μm with increasing Mn content. It was observed that the second phases are dispersed in Mg-Sn-Mn alloys, namely the granular Mg2Sn phase and the rod-shaped α-Mn. Furthermore, the Mg-3Sn-1.5Mn alloy exhibits the optimal comprehensive mechanical properties with ultimate tensile strength (UTS), yield strength (YS), and elongation to fracture (EL) of 249.5 MPa, 203.3 MPa, and 19.3%, respectively. The YS of Mg-3Sn-1.5Mn alloy was significantly enhanced by 42.5 MPa than that of Mg-3Sn alloy, accompanied by a moderately improved elongation from 15.4% to 19.3%. The higher strength of the Mg-3Sn-1.5Mn alloy was attributed to grain refinement (25.2 ~ 28.8 MPa) and second phase strengthening (17.097 ~ 17.147 MPa), while the enhanced plasticity of the alloy is due to the weakening of the basal texture, and the higher SF of the prismatic (a) slip.

1. Introduction

Magnesium (Mg) alloys have great potential to be used in aerospace, automotive, digital products, and other fields due to their lightweight, high specific stiffness, excellent damping capacity, and easy to recycle [1–4]. Unfortunately, Mg alloys usually exhibit poor plastic deformation capacity due to their hexagonal close-packed (HCP) structure with a finite slip system, which hinders the wider application of Mg alloys [2, 5–7]. Studies have found that adding certain alloying elements to pure Mg is an effective strategy to improve their strength and ductility [5, 6, 8–12].

According to the Mg-Sn binary phase diagram [13], when the temperature reaches the eutectic temperature of approximately 561 °C, the solid solubility of Sn in α-Mg is 14.48 wt.%, but at 200 °C, the solid solubility of Sn in α-Mg is only 0.45 wt.%, indicating that Mg-Sn alloys exhibit significant precipitation strengthening effect. Therefore, in recent years, Mg-Sn-based alloys have received special attention [14–17]. Previous studies have shown that the adding of Sn to the Mg matrix can play a role in precipitation strengthening owing to the Mg2Sn precipitated phase being a hard phase (Vickers hardness, 119 HV) [14], it can effectively improve the mechanical properties of Mg-Sn-based alloys. In addition, the melting point of the Mg2Sn intermetallic compound is much higher than that of the Mg17Al12 phase, thus Mg-Sn has a better thermal stability [18, 19]. Therefore,
Mg-Sn-based alloys have higher creep resistance than Mg-Al-based alloys [20]. Liu et al [15] found that the excellent thermal stability of the Mg2Sn improves the mechanical properties of the Mg-Sn alloy at elevated temperatures. Moreover, the compressive creep resistance of the Mg-7%Mn alloy is similar to that of AE42, and the performance of Mg-10%Mn is better than that of the AE42 alloy at 150 °C. Therefore, the addition of Sn can improve the compressive creep resistance of Mg alloys. Although Mg-Sn-based alloys have better creep resistance than other binary Mg alloys, their mechanical properties are poor at room temperature, which impedes their wider application [14, 15, 21]. By adding other alloying elements to the Mg-Sn alloy, its mechanical properties can be further improved in addition to ensuring exceptional creep resistance.

As a widely used alloying element in Mg alloys, Mn can improve the mechanical properties [22–24], corrosion resistance [25, 26] and creep resistance [20, 27, 28] of the alloys. Specifically, during the Mg alloys smelting and casting, a small amount of Mn is added as a degergency agent, and increasing the amount of Mn in the alloy can refine its grains, further improving the mechanical properties. For instance, She et al [29] reported that the tensile strength, compressive strength and yield asymmetry of the extruded Mg-Ca alloy were significantly enhanced with the addition of Mn. Fang et al [30] also found that a certain amount of Mn was added to the Mg-Gd alloy to form α-Mn precipitates, which can be used as a refiner to inhibit the growth of recrystallized grains during hot extrusion. However, current studies on Mg-Sn-Mn alloys focus on as-cast alloy and corrosion resistance. The strengthening and deformation mechanisms of Mg-Sn-Mn alloy still need to be further studied to promote the application of Mg alloys. In addition, Sn and Mn have significant cost advantages compared to rare earth elements [3, 10–12]. Therefore, in this work, the influence of the addition of Mn on the microstructure of extruded Mg-Sn-Mn alloys and the strengthening mechanism is investigated in detail.

### 2. Materials and methods

As-cast Mg-3Sn-xMn (x = 0, 0.5, 1.5, 2.5) alloys were prepared by smelting and casting, using industrial pure Mg (99.95 wt.%), high-purity tin (Sn) (99.99 wt.%), and Mg-Mn (4.3 wt.%) master alloy as raw materials. The melt was placed in a resistance furnace at 750 °C and kept for 20 min. After cooling to 730 °C, the melt was cast into a steel mold with a size of Φ500 mm × 1000 mm. Ar (99%) and SF6 (1%) were introduced to protect the Mg alloy melt during the melting of Mg alloys. The chemical composition of the ingot was further measured by ICP-OES, and the results are shown in Table 1. The ingot was homogenized at 450 °C for 24 h, then quenched in water (25 °C), finally processed into a cylinder with a size of Φ44 mm × 40 mm and subsequently preheated at 350 °C for 90 min in a 20:1 extrusion die. The ingots used for extrusion are taken from the most central part of the original ingot. Ultimately, it was subjected to hot extrusion molding by a four-column hydraulic press to obtain a bar with a diameter of 10 mm.

The phase identification of the Mg-3Sn-xMn alloys was conducted by taking a 6 × 6 × 5 mm block sample from the extruded bar and using a BRUKER D8 ADVANCE A25 x-ray diffractometer (XRD). The specific parameters were as follows: Cu target Kα line, acceleration voltage 40 kV, scanning angle 20° ~ 90°, and scanning speed 0.02°/s. Furthermore, Zeiss Merlin compact field emission scanning electron microscope (ZEISS-6035) equipped with an EDS and EBSD detector was used to characterize the element distribution, second phase, and grain orientation of the sample. SEM images were imaged by secondary electron with an operating voltage of 20 kV. Finally, an Instron-5982 universal testing machine was used to test the tensile properties of the extruded rod-shaped specimens with a diameter of 10 mm and a gauge of 15 mm. The tensile rate was 0.2 mm min−1 and five specimens of each alloy were prepared for testing to ensure the accuracy of the results.

### 3. Results

#### 3.1. Microstructure

Figure 1 shows the XRD patterns of the as-extruded Mg-3Sn-xMn alloys. It can be observed that the Mg-3Sn alloy is composed of α-Mg and Mg2Sn. After adding the Mn, the diffraction peaks of the α-Mn phase appear in

| Table 1. Chemical compositions of the experimental alloys (wt.%). |
|----------------|--------|--------|--------|
| Alloys         | Sn     | Mn     | Mg     |
| Mg-3Sn         | 3.6    | 0      | Bal.   |
| Mg-3Sn-0.5Mn   | 3.4    | 0.6    | Bal.   |
| Mg-3Sn-1.5Mn   | 3.5    | 1.4    | Bal.   |
| Mg-3Sn-2.5Mn   | 3.6    | 2.1    | Bal.   |
the XRD patterns of Mg-3Sn-1.5Mn and Mg-3Sn-2.5Mn alloys, indicating that the α-Mn exists in the alloys. In the Mg-3Sn-0.5Mn alloy, no diffraction peak corresponding to the α-Mn phase is observed, which illustrates that the Mn element is solubilized into the Mg matrix.

The EDS images of the Mg-3Sn-2.5Mn alloy (figure 2) and SEM images (figure 3) of the Mg-3Sn-xMn alloys were obtained to further analyze the microstructure, and the granular and slender rod-like of second phase was observed. More specifically, the granular second phase is Mg$_2$Sn phase and the rod-like second phase is α-Mn, as...
shown in figure 2. When Mn is not added, the alloy is mainly composed of the granular Mg2Sn phase and the α-Mg matrix. In addition, the number of rod-shaped α-Mn phases increases with increasing Mn content, as shown in figures 3(c) and (d). It is due to the α-Mn phase with good thermal stability and extremely low solid solubility during the homogenization heat treatment before extrusion which is hardly dissolved into the α-Mg matrix. These phases are broken during the hot extrusion, becoming rod-shaped and concentrated in the extrusion zone. Figure 3 also suggests that when the Mn content is 1.5 wt.%, the second phase in the alloy presents the smallest size and is more uniformly distributed in the matrix, while the second phase of the other two alloys with Mn exhibits a coarse size and irregular distribution.

Additionally, the IPF maps of the Mg-3Sn-xMn alloys (figure 4) reveal that the microstructure of the Mg-3Sn and Mg-3Sn-1.5Mn alloys is more uniform compared with that of the Mg-3Sn-0.5Mn and Mg-3Sn-2.5Mn alloys. The microstructural homogeneity of the hot extruded alloy is mostly determined by the degree of dynamic recrystallization. The recrystallization distribution maps of the alloys (figure 5) show that Mg-3Sn and Mg-3Sn-1.5Mn samples are almost completely recrystallized, so the microstructure of the two alloys is more uniform. Contrarily, the microstructure of the other two alloys contains coarse non-recrystallized grains.

The grain size distribution maps (figure 4) reveal that when the content of Mn increases from 0 wt.% to 1.5 wt.%, the grain size of the alloy gradually decreases from 21.45 μm to 10.51 μm. This is due to the addition of Mn during the hot extrusion not only provides heterogeneous nucleation sites but also increases the content of precipitated phases. These precipitates are squeezed into pieces during the extrusion. The fine particles with zonal distribution along the extrusion direction can well hinder the sliding of the grain boundaries during the extrusion and inhibit the grain growth of the alloy, so the grain size decreases. However, when the content of Mn is increased to 2.5 wt.%, the grain size of the alloy increases slightly and the AGS of the Mg-3Sn-2.5Mn alloy is 13.41 μm. Combined with the analysis of the SEM image of the alloy, it can be seen that this may be due to the increase in the size of the rod-shaped second phase particles in the Mg-3Sn-2.5Mn alloy and the decrease in the degree of spatial dispersion, which leads to a decrease in pinned particles on a section of the grain boundaries. The pinning effect is weakened, so the grain size increases slightly. The third column in figure 4 shows the band contrast images of the extruded Mg-3Sn-xMn alloys, in which the black lines represent the high-angle grain boundaries (HAGBs) and the green lines represent the low-angle grain boundaries (LAGBs). It can be seen that the HAGBs are dominant in the Mg-3Sn and Mg-3Sn-1.5Mn alloys, indicating that sufficient dynamic recrystallization occurs in the alloys. Meanwhile, a large number of LAGBs are found in the irregular coarse grains of the Mg-3Sn-0.5Mn and Mg-3Sn-2.5Mn alloys, which were caused by uneven deformation during the extrusion.
Figure 6 shows the pole figures and inverse pole figures of the extruded Mg-3Sn-xMn alloys. There are concentrated pole density points observed in the $\{0001\}$ pole figures distributed along the transverse direction (TD), which indicates that the four extruded alloys exhibit $\{0001\}$ typical basal fiber texture [31, 32]. Moreover, the texture type of the alloy does not change significantly with the increase of Mn content, but when the Mn content is 1.5 wt.%, the texture intensity of the alloy is the smallest at 9.32. This could be attributed to the random orientation of a large number of fine recrystallized grains in Mg-3Sn-1.5Mn alloy, which leads to the weakening of texture. Specifically, the number of second phase particles in this alloy increases (figure 3). During the hot extrusion, these particles could hinder the movement of dislocations and increase the deformation energy storage, which increases the driving force for recrystallization, leading to an increase in the number of recrystallized grains [6, 33]. The orientation of the newly formed recrystallized grains is random, so the texture intensity of the Mg-3Sn-1.5Mn alloy was weakened.

Figure 7 shows the Schmid factor (SF) maps of the basal $\langle a \rangle$ slip ($\{0001\}$ $(11\overline{2}0)$, prismatic $\langle a \rangle$ slip ($10\overline{1}0$) $(11\overline{2}0)$ and pyramidal $\langle c+a \rangle$ slip $(11\overline{2}2)$ $(11\overline{2}3)$ of the Mg-3Sn-xMn alloys, and the corresponding average SF are shown in figures as well. It was determined from figures 7(a)–(c) that the average SF of the basal slip, prismatic slip and pyramidal slip of the Mg-3Sn alloy is 0.35, 0.15 and 0.45, respectively. The pyramidal $\langle c+a \rangle$...
slip exhibits the highest average SF. However, the pyramidal slip has a high critical resolved shear stress (CRSS) at room temperature and is not easy to activate. The basal slip becomes the dominated deformation mode of Mg alloy at ambient temperature owing to its low CRSS [6, 17]. When the content of Mn is 0.5 wt.% (figures 7(d)–(f)), the average SF of basal slip, prismatic slip and pyramidal slip is 0.32, 0.25, and 0.44, respectively. Compared with the Mg-3Sn alloy, the average SF of prismatic (a) slip increases, while the SF of basal slip and pyramidal slip decreases slightly. It shows that the addition of Mn promotes the activities of prismatic (a) slip during the plastic deformation, thereby the plasticity of the alloy was improved. With the further increase of Mn content, the average SF of basal slip and pyramidal slip of the Mg-3Sn-1.5Mn alloy increases slightly compared with the Mg-3Sn-0.5Mn alloy (figures 7(g)–(i)). Overall, the alloys with Mn addition exhibit an increased SF of prismatic (a) slip, which means that the prismatic (a) slip is more easily activated, promoting the deformation [34, 35] and improving the plasticity of the alloys.

Figure 5. Distribution images of different grain types of the four alloys: (a), (e) Mg-3Sn; (b), (f) Mg-3Sn-0.5Mn; (c), (g) Mg-3Sn-1.5Mn; (d), (h) Mg-3Sn-2.5Mn.

Figure 6. Pole figures and inverse pole figures of extruded alloys: (a) Mg-3Sn; (b) Mg-3Sn-0.5Mn; (c) Mg-3Sn-1.5Mn; (d) Mg-3Sn-2.5Mn.
3.2. Mechanical properties

Figure 8 exhibits the tensile engineering stress-strain curves of the four extruded alloys and the relationship between the mechanical properties and the alloy composition. Table 2 summarizes the mechanical properties of the alloys with different Mn content at ambient temperature, including the ultimate tensile strength (UTS), yield strength (YS), and elongation (EL). It can be observed that the YS and UTS of the extruded alloy gradually increase with increasing Mn content. When the Mn content increases from 0 to 2.5 wt.%, YS increases from 160.8 MPa to 205.6 MPa, an increase of 27.9%, while UTS increases from 228.7 MPa to 257.7 MPa, which corresponds to an increase of 12.7%. However, when the Mn content is increased to 1.5 wt.%, the EL of the alloy first increases then reaches a peak and finally decreases as the Mn content further increases. Overall, the comprehensive mechanical properties of the Mg-3Sn-1.5Mn alloy are optimal. This is because the Mn elements in the extruded Mg-3Sn-1.5Mn alloy are mainly distributed in the \( \alpha \)-Mg matrix in the form of dispersed rods while being finely dispersed. This dispersed rod-like second phase could hinder the movement of dislocations during the stretching.

4. Discussion

4.1. Effect of Mn addition on the microstructure

Based on the above analysis of the microstructure of the alloys, it was found that the addition of Mn mainly affects the average grain size and the morphology of the second phase. It can be observed from figure 4 that the
addition of Mn can significantly refine the grains, which is consistent with previous reports [21, 36–38]. At present, the widely accepted mechanism for grain refinement due to the addition of alloying elements is the separation of solute atoms and the heterogeneous nucleation of particles [39]. On the one hand, the addition of Mn may cause excessive cooling of the composition in the diffusion layer at the front of the solid-liquid interface, thereby limiting the growth rate of grains, resulting in grain refinement. On the other hand, the number of second phase particles in the alloy gradually increases with the addition of Mn. More specifically, the higher the Mn element content, the greater the number of rod-shaped α-Mn elementary second phases (figure 3).

According to previous reports, the intermetallic compound particles can hinder the migration of the grain boundary during grain growth [32, 40]. In addition, classical theory [41, 42] demonstrated that the second phase in the metal matrix generates additional strain around the matrix during the thermal deformation, so that fine and dense substructures appear, which stimulates the nucleation of DRX and thus grain refinement. Moreover, it was found that when the content of Mn is 1.5 wt.%, the texture intensity of the alloy is greatly weakened (figure 6). This is attributed to the number of the second phase in the Mg-3Sn-1.5Mn alloy being large, the size of second phase particles being the smallest, and the distribution being diffuse. Therefore, extensive recrystallized grains could be stimulated by particle stimulation nucleation (PSN) [26, 29]. During the deformation, these second phase particles hinder the movement of dislocations, increase the deformation energy storage, and increase the driving force for recrystallization as well [44]. As a result, the number of recrystallized grains increases, and the orientation of the newly formed recrystallized grains are random, so the texture intensity of the Mg-3Sn-1.5Mn alloy decreases.

4.2. Effect of Mn addition on the mechanical properties
The YS and UTS of the four alloys gradually increase with increasing Mn content. When the Mn content increases to 1.5 wt.%, the EL of the alloy first increases, reaches a peak and then decreases as the Mn content further increases, as shown in figure 8(b). According to the Hall-Petch relationship, YS and UTS increase as the grain size decreases. Therefore, the strength enhancement of the alloy is attributed to grain refinement. For the alloy with a Mn content of 1.5 wt.%, there is a high proportion of HAGBs (figure 4). HAGBs refer to grain boundaries with misorientation greater than 15°, which can prevent dislocations from transferring from one grain to another. Dislocation pile-up occurs at the grain boundaries, increasing the macroscopic strength [44]. Furthermore, the Mg-3Sn-1.5Mn alloy exhibits the smallest grain size, that is, the smallest inclusions on the grain boundary per unit volume, thereby increasing the elongation. Moreover, the weakening of the basal texture can reduce the hindrance of basal slip [45], and also increase the elongation of the Mg-3Sn-1.5Mn alloy.

The addition of Mn can also increase the average SF of prismatic (a) slip, promote non-basal dislocation slip, and

| Table 2. The mechanical properties of the as-extruded alloys. |
|-----------------|----------------|-----------------|
| Alloys          | σ_{YS} (MPa)   | σ_{UTS} (MPa)   | EL (%)         |
| Mg-3Sn          | 160.8 ± 3.7    | 228.7 ± 5.3    | 15.4 ± 0.6     |
| Mg-3Sn-0.5Mn    | 185.8 ± 3.4    | 241.5 ± 4.1    | 16.1 ± 0.3     |
| Mg-3Sn-1.5Mn    | 203.3 ± 3.5    | 249.5 ± 4.9    | 19.3 ± 0.7     |
| Mg-3Sn-2.5Mn    | 205.6 ± 2.9    | 257.7 ± 3.6    | 16.8 ± 0.4     |

Figure 8. (a) Engineering stress-strain curves of the as-extruded alloys; (b) Relationship between mechanical properties and alloy composition.
improve the plasticity. On the other hand, the strong fiber texture causes the grains to be in 'hard orientation', which means that basal slip is difficult to activate, that is, the sample has poor ductility [46, 47]. In addition, the coarser second phase particles in Mg-3Sn-2.5Mn alloy tend to cause stress concentration [2], which is not conducive to coordinated deformation. As a result, compared with Mg-35n-1.5Mn alloy, the strength of Mg-3Sn-2.5Mn alloy increases slightly, but the elongation decreases.

4.3. Strengthening mechanisms

Theoretical analysis and calculations were carried out on the Mg-3Sn-1.5Mn and Mg-3Sn alloys to further analyze the strengthening mechanism of the Mn addition. The major strengthening mechanisms consist of fine grain strengthening, second phase strengthening, and solid solution strengthening [2]. However, only replacement solid solutions occur in Mg alloys and their strengthening effect is not significant and can be ignored. Therefore, this article mainly discusses fine grain and second phase strengthening effect. Studies have shown that the Hall-Petch relationship can be used to calculate the contribution of fine grain strengthening [48, 49]:

\[ \sigma_{HP} = \sigma_0 + kd^{-1/2} \]  

where \( \sigma_0 \) and \( k \) is experimental constants and \( d \) is the AGS. Therefore, based on equation (1), the contribution of Mg-3Sn-1.5Mn alloy to the fine-grain strengthening of Mg-3Sn alloy can be calculated as:

\[ \Delta \sigma_{HP} = k(d^{-1/2}_{Mg31} - d^{-1/2}_{Mg2}) \]  

where \( k \) is the H-P coefficient, and \( d \) is the AGS. When \( d > 1 \mu m \), the \( k \) value of Mg is constant \((280 \sim 320 \, \text{MPa} \cdot \sqrt{\mu m}) [48] \), calculated as \( \Delta \sigma_{HP} \approx 25.2 \sim 28.8 \, \text{MPa} \).

In addition, second phase strengthening mainly includes [50]: (1) dislocation bypass mechanism (Orowan strengthening mechanism), \( \Delta \sigma_O \), (2) the load shifts from the matrix to the strengthening generated by the second phase particles, \( \Delta \sigma_T \), and (3) strengthening caused by dislocations due to the different thermal expansion coefficients of the matrix and the second phase particles, \( \Delta \sigma_G \).

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

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

For Mg alloys, Zhang et al [51] proposed an improved equation of the Orowan equation:

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

where \( G \) represents the shear modulus (Mg alloy generally take the value of \( 1.66 \times 10^4 \, \text{MPa} \)), \( b \) is the Perkins vector (\( 3.21 \times 10^{-10} \, \text{m} \)), \( f_p \) represents the volume fraction of the second phase particles, and \( d_p \) represents the average particle size of the second phase particles. The \( f_p \) and \( d_p \) were obtained by analyzing three different SEM images of the same alloy using Image-Pro-Plus software. \( f_p \) and \( d_p \) were 0.26%, 0.3 \( \mu m \) respectively. Therefore, \( \Delta \sigma_O \) can be calculated as 10.097 MPa.

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

\[ \Delta \sigma_T = \frac{1}{2} \sigma \sigma T \]  

where \( \sigma_T \) represents the yield strength of the matrix, generally taken for Mg alloys as (100 \sim 150 \, \text{MPa}) [50]. \( \Delta \sigma_T \) is then calculated as 0.1 \sim 0.15 \, \text{MPa}.

The strengthening caused by the dislocation caused by the different thermal expansion coefficients of the matrix and the second phase particles is given by references [50, 52, 53]:

\[ \Delta \sigma_G = \alpha Gb \left( \frac{12 \Delta T \Delta C G f_p}{bd_p} \right)^{1/2} \]  

where \( \alpha \) is a constant, with a value of 1.25, \( \Delta T \) represents the difference between the extrusion temperature and the actual temperature, and \( \Delta C \) represents the difference in the thermal expansion coefficient between the matrix and the second phase (\( C_{Mg} = 2.61 \times 10^{-6} \, \text{k} \), \( C_{Mg2Sn} = 0.45 \times 10^{-6} \, \text{k} \)). \( \Delta \sigma_G \) is calculated as 6.9 MPa. Therefore, the total strength contribution of the second phase reinforcement is \( \Delta \sigma = \Delta \sigma_O + \Delta \sigma_T + \Delta \sigma_G = 17.097 \sim 17.17 \, \text{MPa} \).

According to the calculation of the above strengthening mechanism, the results of the calculated values including the fine grain strengthening and the second phase strengthening are respectively 25.2 \sim 28.8 \, \text{MPa} and 17.097 \sim 17.147 \, \text{MPa}. The calculated total value 42.297 \sim 45.947 \, \text{MPa} is close to the increase of the experimental
strength $\Delta \sigma_{YS} = 42.5$ MPa. Therefore, the high strength of the Mg-3Sn-1.5Mn alloy is mainly attributed to grain refinement and second phase strengthening.

5. Conclusions

The effects of Mn content on the microstructure and mechanical properties of the extruded Mg-3Sn-xMn ($x = 0, 0.5, 1.5, 2.5$) alloys was systematically investigated in this study. The following conclusions can be drawn:

1. The addition of Mn can refine the grains and the average grain size of the alloys decreases from 21.45 $\mu$m (Mg-3Sn alloy) to 10.51 $\mu$m (Mg-3Sn-1.5Mn alloy). This is attributed to the fact that $\alpha$-Mn in the alloy can act as the nucleation sites of recrystallization while inhibiting the growth of recrystallized grains.

2. The Mg-3Sn-1.5Mn alloy exhibits the optimal comprehensive mechanical properties at room temperature with UTS, YS, and EL of 249.5 MPa, 203.3 MPa, and 19.3%, respectively. The YS of Mg-3Sn-1.5Mn alloy was significantly enhanced by 42.5 MPa (26.4%) than that of Mg-3Sn alloy.

3. According to theoretical analysis and calculation, the enhanced 42.5 MPa strength of the Mg-3Sn-1.5Mn alloy is mainly attributed to grain refinement and second phase strengthening. The increase in EL is due to the weakening of the basal texture and the higher SF of the prismatic $<a>$ slip.

Data availability statement

The data that support the findings of this study are available upon reasonable request from the authors.

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Conflicts of interest

On behalf of all authors, the corresponding author states that there is no conflict of interest.

Availability of data and material

The data in this work is transparency.

Authors’ contributions

Conceptualization, Yongqiang Fang and Zeli Yu; methodology, Yixin Zhang, Bing Zhang and Ke Wang; investigation, Wenjing Lu and Xiaochen Ma; writing—original draft preparation, Qi Wei; writing—review and editing, Shuai Yuan; Resources, Shuxiang Zhang.

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