Microstructures and Mechanical Properties of Mg–xAl–1Sn–0.3Mn (x = 1, 3, 5) Alloy Sheets

Chao-Yue Zhao¹ · Xian-Hua Chen^{1,2} · Peng Peng¹ · Teng Tu¹ · Andrej Atrens³ · Fu-Sheng Pan¹

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Abstract

The influence of Al alloying on the microstructures and the mechanical properties of Mg–xAl–1Sn–0.3Mn alloy sheets was investigated. The microstructure of Mg–xAl–1Sn–0.3Mn consisted of α -Mg and Mg₁₇Al₁₂ precipitates. Alloying with Al increased the amount of Mg₁₇Al₁₂ and the average grain size. Uniaxial tensile tests were carried out along the extrusion direction (ED), the transverse direction (TD) and 45° toward the ED. Mg–5Al–1Sn–0.3Mn alloy sheet exhibited the best combination of mechanical properties along the ED: a yield strength of 142 MPa, an ultimate tensile strength of 282 MPa and an elongation of 23%. The good performance of Mg–5Al–1Sn–0.3Mn sheet was mainly attributed to the large quantity of Mg₁₇Al₁₂ precipitates and a weak basal texture. Annealing caused static dynamic recrystallization, refined the grain size and enhanced the mechanical properties: yield strength of 186 MPa, ultimate tensile strength of 304 MPa, elongation of 21% along ED. Both strength and ductility were enhanced by Al alloying.

Keywords Mg–xAl–1Sn–0.3Mn alloys · Extrusion · Microstructure · Mechanical properties

1 Introduction

Magnesium alloys are the lightest structural material, which can save energy effectively and reduce environmental pollution [1–3]. Their good properties—low density, high specific strength/stiffness and good damping capacity—make them a potential substitute for heavier materials such as steel and aluminum [4, 5]. However, the application of magnesium alloys has been restricted because of their inadequate mechanical properties [6, 7]. Therefore, it is pressing to develop high-performance magnesium alloys to extend their commercial applications.

A lot of research has indicated that room-temperature mechanical properties of Mg–RE alloys have been increased

by alloying and micro-alloying [8–10]. However, although their mechanical properties were improved, the alloying with rare earth elements greatly increased the cost. Fortunately, recent investigations demonstrated that the Mg–Sn alloy system has the potential to have good room-temperature mechanical properties [11–14], but few works have studied extruded Mg–Sn alloys, even though Sasaki [14] achieved a Mg–Sn alloy with fine grains, Mg₂Sn and a high-strength/low-yield asymmetry at room temperature.

It is commonly recognized that the major obstacle for extruded Mg alloys is high production cost arising from low extrusion temperatures and low extrusion speeds [15, 16]. Some recent work [17] showed that commonly produced extruded Mg–Al–Sn alloys have room-temperature mechanical properties better than conventional alloys. These Mg–Al–Sn alloys have a high fracture elongation, relatively high strength and low cost. However, the research on extruded Mg–Al–Sn alloy sheet is limited.

Alloying with Al refined the grain size and homogenized the distribution of secondary phases in as-cast Mg–Sn alloys [18] and thereby increased the ductility and strength, due to the grain refinement and precipitation strengthening [17–19]. Consequently, Al is a promising element to improve the mechanical performance of extruded Mg–Sn-based alloy sheets.

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✉ Xian-Hua Chen
xhchen@cqu.edu.cn

¹ College of Materials Science and Engineering, Chongqing University, Chongqing 400045, China

² Shenyang National Laboratory for Materials Science, Chongqing University, Chongqing 400045, China

³ School of Mechanical and Mining Engineering, The University of Queensland, St Lucia, QLD 4072, Australia

Hence, this work extruded a set of Mg–xAl–1Sn alloys with various Al contents into sheets and studied the influence of Al on the microstructure and mechanical properties of the extruded alloy sheets, to obtain novel low-cost magnesium alloy sheet for commercial applications.

2 Experimental

The nominal compositions of the Mg–xAl–1Sn–0.3Mn (ATMx10, $x = 1, 3, 5$) alloys are listed in Table 1. Based on a chemical composition of Mg–xAl–1Sn–0.3Mn (wt%), commercially pure Mg (99.95 wt%), Al (99.90 wt%) and Sn (99.90 wt%) were selected to prepare the alloys. In addition, the 0.3 wt% Mn was added to diminish the negative effect of the impurity Fe on the corrosion behavior of the alloys [20]. The raw materials (commercially pure Mg (99.95 wt%), Al (99.90 wt%) and Sn (99.90 wt%)) were melted in a crucible in a furnace protected by a mixed gas of CO₂ and SF₆ and poured into a steel mold pre-heated to 300 °C to obtain ingots for extrusion, 80 mm in diameter and 100 mm in length. These ingots were homogenized at 420 °C for 24 h, cooled in water, pre-heated in a resistance furnace at 350 °C for 1 h and extruded at an initial ingot temperature of 350 °C and a ram speed of 1.3 mm/s using an XJ-500 horizontal extrusion machine to give a sheet profile with a thickness of 3 mm and a width of 60 mm, at a fixed extrusion ratio of 28.

Microstructural examination of the alloy specimens was conducted using optical microscopy (OM), a JEOL JSM-7800F field emission scanning electron microscope (SEM) equipped with a HKL Channel 5 electron backscattered diffraction (EBSD) system and a FEI TECNAI G2 F20 TEM. To reveal the microstructure, the OM and SEM samples were etched in picric solution for 5–10 s at room temperature, the EBSD samples were electropolished in a polishing solution cooled to –30 °C, at 20 V for 40 s, and the TEM samples were prepared through mechanical polishing and ion-beam thinning using a Gatan precision ion polishing system at –70 °C. The average grain size was analyzed from several micrographs of different sections using software named Image-Pro Plus. X-ray diffraction (XRD, Rigaku D/Max 2500, Cu K α radiation, a voltage of 60 kV and a current of 30 mA) was carried out on selected samples to identify the existing phases. Texture analysis was also performed by the XRD. Uniaxial tensile tests at room temperature were carried out on dog-bone specimens of extruded sheets (nominal length of 45 mm, a width of 5.0 mm and a thickness of 3 mm) with a constant cross-head speed

using a CMT5105 material testing machine at a strain rate of 10^{–3} s^{–1} along three different directions: ED, TD and 45° toward ED. EBSD data were analyzed using the HKL Channel 5 system.

3 Results and Discussion

3.1 1 Microstructures of the As-Cast Alloys

Figure 1a–c shows the microstructures of the as-cast ATM110, ATM310 and ATM510 alloys, which indicate that these samples were composed of the α -Mg matrix and interdendritic second-phase particles of Mg₁₇Al₁₂. Increasing Al content decreased dramatically the average grain size from 1.1 mm to 85 μ m and increased the dendrite density along the grain boundaries. The enrichment of Al in front of the solid–liquid interface caused constituent undercooling, which provided nucleation sites in front of the solid–liquid surface that refined the grains [21]. The XRD results as shown in Fig. 1d indicate that the 1 wt% Sn was dissolved in the Mg matrix and the second phase in these alloys was the Mg₁₇Al₁₂ phase.

3.2 Microstructures of As-Extruded Alloy Sheets

Figure 2a–c shows optical micrographs of the as-extruded alloys, which exhibited microstructures finer than the as-cast alloys shown in Fig. 1a–c attributed to the occurrence of dynamic recrystallization during extrusion. The average grain size of the as-extruded Mg–Al–Sn–Mn alloy sheet increased from 4.3 to 14.3 μ m with increasing Al content, and the grain size distribution became heterogeneous possibly because the recrystallization temperature increased with increasing alloying. The coarse grains in ATM510 were tentatively identified as unrecrystallized grains, and further research is planned to test this hypothesis.

Macro-texture maps of the as-extruded sheets are shown in Fig. 2d–f. The intensity of basal texture of the alloy sheet decreased significantly from 9.72 to 7.31 as the Al content increased from 1 to 3 wt%, and then, it was almost unchanged as the Al content increased up to 5 wt%. This suggested that the (0001) plane of the recrystallized grains was mainly parallel to the ED, and thus, ATM110 had the strongest basal texture of the three alloys. Kim et al. [22] and Zhao et al. [23] also reported that increasing Al alloying weakened basal texture during extrusion by activating more non-basal slip during extrusion. Thus, the {0001} plane of more grains prefers to orientation toward the ED and the texture of the {0001} plane is elongated along the ED. Nevertheless, there were still some original coarse grains which maintained their initial orientations during extruded, as shown in Fig. 2b and c. This caused a slight

Table 1 Nominal chemical compositions of studied alloys (wt%)

Alloy	Al	Sn	Mn	Mg
ATM110	1	1	0.3	Bal
ATM310	3	1	0.3	Bal
ATM510	5	1	0.3	Bal

Fig. 1 Optical microstructures and the XRD patterns of as-cast Mg-xAl-1Sn-0.3Mn alloys: **a** ATM110, **b** ATM310, **c** ATM510 **d** XRD patterns

enhancement of the basal texture intensity in ATM510 compared with ATM310 [24].

Fig. 2 Optical microstructures and (0002) pole figures of as-extruded Mg-xAl-1Sn-0.3Mn alloys: **a, d** ATM110, **b, e** ATM310 **c, f** ATM510

3.3 Mechanical Properties and Mechanical Anisotropy

3.3.1 Mechanical Properties

The room-temperature tensile stress–strain curves of the as-extruded alloy sheets are shown in Fig. 3. Increasing Al content improved the mechanical properties of the as-extruded alloy sheets. The tensile strength increased from 230 to 282 MPa and the elongation to fracture increased from 9 to 23%, whereas the yield strength decreased from 144 to 117 MPa with increasing Al content from 1 to 3 wt% and then increased to 142 MPa when the Al content increased to 5%. The increase in strength improvement was attributed to the combined effect of second-phase strengthening and solid solution strengthening [18]. However, the yield strength was not influenced by the above two mechanisms. Figures 2a–c and 3 show that ATM110 with the finer grain size had a yield strength higher than ATM310, which suggests that the Hall–Petch grain size strengthening also had a strong influence on the yield strength. Moreover, the as-extruded sheets had a strong (0002) basal texture, which means that the basal plane of most grains was parallel to ED and in a hard-to-deform orientation (hard orientation). The strong basal texture increased the critical resolved shear stress (CRSS) of basal slip and hindered the movement of dislocations [25]. Thus, the elongation of the as-extruded alloy sheets increased with decreasing intensity of basal texture.

3.3.2 Mechanical Anisotropy

Due to low-symmetry crystal structure and the strong basal texture, the direction of the applied stress influences the

mechanical properties of Mg–xAl–1Sn–0.3Mn alloy sheets. The mechanical anisotropy exerts strong dependence upon further plastic deformation. Figure 4a–c shows the tensile engineering stress–strain curves in three directions (ED, TD and 45° toward ED). The ATM510 alloy sheet had the best strength along the TD and the best elongations along the ED. Compared with ATM110 and ATM310, ATM510 had higher strength, especially the ultimate strength in the three different sheet directions. Furthermore, the increasing Al content decreased the mechanical anisotropy of alloy sheets to some extent.

To quantify the mechanical anisotropy of the as-extruded Mg–xAl–1Sn–0.3Mn alloy sheets, the anisotropy index Δr was evaluated by the equation [26]:

$$\Delta r = r_0 + r_{90} - 2r_{45} \quad (1)$$

$$r = \frac{\varepsilon_b}{\varepsilon_t} = \frac{\ln b/b_0}{\ln t/t_0} \quad (2)$$

where b_0 and b are the width, and t_0 and t are the thickness of the as-extruded sheet before and after the tensile test. The result is shown in Fig. 4d. A larger value of Δr indicates a greater mechanical anisotropy in the plane of the as-extruded alloy sheet. The anisotropy index Δr of the alloy sheet decreased from 0.45 to 0.13 as the Al content increased from 1 to 3 wt% and increased to 0.28 as the Al content further increased to 5 wt%.

The variation of mechanical anisotropy of the alloy sheets was similar to the trend of texture intensity, which means that it is mainly influenced by basal texture. Yu et al. [27] observed that when basal slip was in a hard orientation and difficult to activate due to the strong basal texture, basal slip was easier to activate in the grains with c -axis tilting away from basal poles and these grains can carry the strain. It is clear that the c -axis of grains is more likely to tilt from ND to TD, which means easier activation of basal slip and results in the lowest yield strength in the ED. Moreover, the tendency of c -axis tilting from ND to ED was the smallest compared with that from ND to TD and that from ND to 45°, which led to the highest yield strength in the TD and moderate yield strength in the 45°. In addition, the increase in resistance of basal slip increased the yield strength and tensile strength, but also caused stress concentration, which resulted in crack propagation inside the alloy sheet and decreased elongation. Thus, these alloy sheets exhibit the best strength properties along the TD and the best elongations along the ED. With the increasing Al content, the c -axis of more grains tilted from ND to ED and this phenomenon was the most distinct in ATM310. Moreover, the intensity of the basal texture was also decreased. Therefore, a weakening basal texture and uniform c -axis tilting away from basal poles reduced the diversity in the amount of activated basal slip in different

Fig. 3 Tensile engineering stress–strain curves of as-extruded Mg–xAl–1Sn–0.3Mn alloy sheets

Fig. 4 Tensile engineering stress–strain curves of as-extruded Mg–xAl–1Sn–0.3Mn alloy sheets in three different sheet directions (0°, 45 and 90°): **a** ATM110, **b** ATM510, **c** ATM510 **d** anisotropy factor

directions and decreased the mechanical anisotropy caused by the low-symmetry crystal structure of these alloys.

3.4 Annealing of ATM510

3.4.1 Microstructure

Figure 2c shows that, different from the microstructure of ATM110 and ATM310, the microstructure of ATM510 was composed of coarse grains surrounded by fine grains. To determine whether the coarse grains were secondary recrystallized grains or incompletely recrystallized grains, as-extruded ATM510 (A0-ATM510) was annealed at 250 °C for 15 min (A15-ATM510), 30 min (A30-ATM510) and 60 min (A60-ATM510). Optical micrographs shown in Fig. 5 reveal the grain structures of the as-extruded and after various annealing times. When the annealing time was 15 min, the average grain size decreased from 14.3 to 12.6 μm and the microstructure started to become more uniform. When the annealing time was increased to 30 min, the grains become uniform and fine, and the average grain size decreased to 7.1 μm . This result shows that coarse grains in ATM510 were unrecrystallized grains and that static

recrystallization occurred in the original coarse grains during the annealing. The fine grains that are recrystallized are continuously engulfed and continually grown to make the grain size uniform. Moreover, with increasing annealing time some twins occurred in the grains [28], as shown in Fig. 5c. However, when the annealing time was 60 min, the grain size increased to 10.3 μm and the shape of the grown secondary recrystallized grain became extremely irregular, which indicated that the grains began to grow when the annealing time was beyond a critical value, as shown in Fig. 5d. Pérez-Prado and Ruano [29] indicated that this is attributed to the grain size heterogeneity rather than to any recrystallization phenomena, like nucleation and growth of new grains, since the initial microstructure was already recrystallized.

Figure 6a and c shows EBSD maps of ATM510 before and after annealing for 30 min. The grain size of A30-ATM510 was remarkably finer than that of A0-ATM510. The kernel average misorientation maps also indicate that the density of coarse grains with low-angle grain boundaries decreased after annealing, which further indicates that these coarse grains were unrecrystallized grains and the static recrystallization occurred during annealing.

Fig. 5 Optical microstructures of annealing-extruded Mg-5Al-1Sn-0.3Mn alloy sheets: **a** A0-ATM510, **b** A15-ATM510, **c** A30-ATM510 **d** A60-ATM510

Figure 6b, d shows the micro-texture of ATM510 before and after annealing for 30 min. The texture type of the alloy sheet did not change after annealing. It was still the (0002) basal texture. In addition, the intensity of basal texture decreased from 12.62 to 11.55 after annealing for 30 min, while the average grain size decreased from 14.3 to 7.1 μm . Moreover, the distribution of the (0002) plane of the unannealed ATM510 sheet was tilted toward the ED slightly. After annealing, the (0002) plane in A30-ATM510 exhibited a greater tendency toward the ED orientation and the distribution of {0002} plane was stretched out longer along the ED, which indicated that the orientation of static recrystallization grains rotated from the ND to the ED during the annealing. That is, the growth direction of the new grains produced by static recrystallization during annealing had a preferred orientation, which was influenced by the residual stress after the hot extrusion. A similar phenomenon was shown by Mackenzie [26]: Orientation selection occurred early in the recrystallization grain growth process. With the plenty of driving force, recrystallization proceeded. The nucleation growth of new grains preferentially grew in a certain direction under the action of residual stress during static recrystallization, which led to the final annealing texture.

The TEM bright-field micrographs of A0-ATM510 and A30-ATM510 are shown in Fig. 6e–g. Figure 6e, f

indicates that there were high-density dislocations and a large amount of fine dispersion second phase in the matrix of A0-ATM510. The high-density dislocation disappeared after annealing. The energy released by the annihilation of the high dislocation density promoted static recrystallization during annealing. During annealing, a portion of the recrystallized grains swallow up smaller grains and grow to larger grains and the original unrecrystallized grains undergo static recrystallization due to the high density of dislocations around them. Thus, the grains of ATM510 alloy sheet were refined evenly. Moreover, the droplet-like precipitates taken from $[\bar{1}213]$ in Fig. 6e on the basal plane of the Mg matrix (0001) were identified as $\text{Mg}_{17}\text{Al}_{12}$. The diameter of the droplet-like precipitates was about 100 (± 10) nm, which resulted in a strong pinning effect during annealing. These fine precipitates seriously hindered the kinetics during static recrystallization and slowed the rate of recrystallization. Thus, some annealing twins were produced in the sheet, as shown in Fig. 5c. Besides, there were no typical polygonal Mg_2Sn precipitates, which indicate that Sn atoms were mainly dissolved in the matrix. Figure 6f, g indicates that the second phase did not change after annealing, which indicated that the alloy exhibits good thermal stability which may be attributed to the dissolved Sn [14].

Fig. 6 EBSD maps showing grains and kernel average misorientation in **a** A0-ATM510 and **c** A30-ATM510, (0002) pole figures of alloys, **b** A0-ATM510, **d** A30-ATM510, the TEM images of **e**, **f** A0-ATM510 **g** A30-ATM510 alloy sheets

3.4.2 Mechanical Properties

Figure 7a shows the tensile engineering stress–strain curves after various annealing times. The mechanical properties of ATM510 were improved after annealing, but there was some reduction for an annealing time. Moreover, annealing led to a slight drop of elongation. ATM510 annealed for 30 min showed the best mechanical performance with greater yield and tensile strength and only a negligible decrease in elongation compared with the as-extruded ATM510 alloy. After 30-min annealing, the yield strength increased from 142 to 186 MPa, the tensile strength increased from 282 to 304 MPa and the elongation at fracture decreased from 22.7 to 20.7% at room temperature. The strength improvement

was attributed to the grain refinement (average grain size decreased from 14.3 to 7.1 μm after annealing for 30 min), while the decreased tensile plasticity was possibly attributed to the elongated distribution of (0001) plane along the ED. For the as-annealed ATM510, the finer grain size and some extension twins may have a strong effect on impeding the movement of dislocations. Thus, yield and tensile strength had a significant increase.

Figure 7b–d shows the mechanical properties of Mg–xAl–1Sn–0.3Mn alloy sheets after annealing for 30 min. The tensile properties did not show a clear change before and after annealing treatment for ATM110, which was attributed to the fact that the microstructure was almost unchanged after annealing. In contrast, ATM310 had a large change

Fig. 7 a Tensile engineering stress–strain curves of Mg–5Al–1Sn–0.3Mn alloy sheets with different annealing times, mechanical properties before and after annealing for 30 min: **b** ATM110, **c** ATM310 **d** ATM510

of tensile properties: The yield strength increased, and the elongation at break decreased significantly. There were less original unrecrystallized grains in A30-ATM310 compared with A30-ATM510. Thus, after full static recrystallization occurs, then grain size continues to increase and decreases mechanical performance, which also occurred in A60-ATM510. These results indicate that only when there are enough unrecrystallized grains in the alloy that has been deformed, the annealing at a certain temperature and time can improve the mechanical properties significantly.

Table 2 lists the reported mechanical properties of as-extruded Al-containing or RE-containing Mg alloys. It is clear that ATM510 alloy sheet achieved superior comprehensive mechanical properties compared with as-extruded commercial AZ61 alloy [31]. Moreover, the tensile performance of ATM510 approached or slightly exceeded some as-extruded magnesium alloys which have roughly similar weight percentage of RE alloying [32–34], which indicates that Mg–xAl–1Sn–0.3Mn provides a promising candidate as a low-cost commercial magnesium alloy sheet.

Table 2 Reported mechanical properties of Al-containing or RE-containing Mg alloys in comparison with the as-extruded alloy in this work

Alloys (wt%)	Process	YS (MPa)	UTS (MPa)	ε (%)	References
ATM510	Extruded	142	282	22.7	This work
ATM510	Extruded + annealed	186	304	20.7	This work
AZ61	Extruded	206	297	13	[27]
AZ61	Extruded + aged	204	294	15	[27]
Mg–5Al–2Nd–0.2Mn	Extruded	158	271	22.2	[28]
Mg–6Gd–2.5Y–0.5Zr	Extruded	167	264	25.1	[29]
Mg–6Gd–2.5Y–0.5Zr	Extruded + aged	177	266	18.1	[29]
Mg–6.5Gd–1.3Nd–0.7Y–0.3Zn	Extruded + aged	201	310	5.8	[30]

4 Conclusions

The microstructure and mechanical properties of Mg–xAl–1Sn–0.3Mn (x = 1, 3, 5) alloy sheets were investigated in the present work. The following conclusions were drawn:

1. As the Al content increases from 1 to 5 wt%, the average grain size of the as-extruded Mg–Al–Sn–Mn alloy sheet increased from 4.3 to 14.3 μm . The amount of the second phase, $\text{Mg}_{17}\text{Al}_{12}$, also increased significantly. The intensity of basal texture of the alloy sheet decreased significantly from 9.72 to 7.31 as the Al content increased from 1 to 3 wt% and then increased to 7.38 when the Al content increased to 5 wt%. The variation of the basal texture was caused by the grain size and the degree of recrystallization.
2. When the Al content increased from 1 to 3 wt%, the yield strength decreased from 144 to 117 MPa and then increased to 142 MPa with Al content increased to 5 wt%. Moreover, with increasing Al content, the tensile strength increased from 230 to 282 MPa and the elongation increased from 9.1 to 22.7%. The anisotropy of mechanical properties of the as-extruded Mg–Al–Sn–Mn alloy sheets decreased due to the variable texture with increasing Al content.
3. The mechanical properties of the as-extruded Mg–5Al–1Sn–0.3Mn alloy sheet were improved due to the grain size refinement and texture variation caused by static recrystallization during annealing. The yield strength increased from 142 to 186 MPa, the tensile strength increased from 282 to 304 MPa, and the elongation decreased from 22.7 to 20.7%. The Mg–5Al–1Sn–0.3Mn alloy is a promising candidate for low-cost commercial magnesium alloy sheet.

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References

- [1] M.D. Sameer, A.K. Birru, J. Magnes, *Alloys* **7**, 264 (2019)
- [2] C.Y. Zhao, X.H. Chen, F.S. Pan, J.F. Wang, S.Y. Gao, T. Tu, C.Q. Liu, J.H. Yao, A. Atrens, J. Mater. Sci. Technol. **35**, 142 (2019)
- [3] Y.B. Hu, C. Zhang, T.X. Zheng, F.S. Pan, A.T. Tang, *Materials* **11**, 1942 (2018)
- [4] F. Zhong, H.J. Wu, Y.L. Jiao, R.Z. Wu, J.H. Zhang, L.G. Hou, M.L. Zhang, J. Mater. Sci. Technol. **39**, 124 (2020)
- [5] G.J. Zhang, Y. Wang, Z. Liu, S.M. Liu, J. Magnes, *Alloys* **7**, 272 (2019)
- [6] B.J. Wang, J.Y. Luan, D.K. Xu, J. Sun, C.Q. Li, E.H. Han, *Acta Metall. Sin. (Engl. Lett.)* **32**, 1 (2019)
- [7] L.Z. Liu, X.H. Chen, J.F. Wang, L.Y. Qiao, S.Y. Gao, K. Song, C.Y. Zhao, X.F. Liu, D. Zhao, F.S. Pan, J. Mater. Sci. Technol. **35**, 1074 (2019)
- [8] T. Tu, X.H. Chen, J. Chen, C.Y. Zhao, F.S. Pan, *Acta Metall. Sin. (Engl. Lett.)* **32**, 23 (2019)
- [9] K. Luo, L. Zhang, G.H. Wu, W.C. Liu, W.J. Ding, J. Magnes, *Alloys* **7**, 345 (2019)
- [10] B.J. Wang, D.K. Xu, S.D. Wang, L.Y. Sheng, R.C. Zeng, E.H. Han, *Int. J. Fatigue* **120**, 46 (2019)
- [11] X.H. Chen, F.S. Pan, J.J. Mao, J.F. Wang, D.F. Zhang, A.T. Tang, J. Peng, *Mater. Des.* **32**, 1526 (2011)
- [12] C.Y. Zhao, X.H. Chen, F.S. Pan, S.Y. Gao, D. Zhao, X.F. Liu, *Mater. Sci. Eng. A* **713**, 244 (2018)
- [13] H. Liu, H. Huang, J.P. Sun, C. Wang, J. Bai, A.B. Ma, X.Y. Chen, *Acta Metall. Sin. (Engl. Lett.)* **32**, 269 (2019)
- [14] T.T. Sasaki, K. Yamamoto, T. Honma, S. Kamado, K. Hono, *Scr. Mater.* **59**, 1111 (2008)
- [15] S.J. Meng, H. Yu, S.D. Fan, Q.Z. Li, S.H. Park, J.S. Suh, Y.M. Kim, X.L. Nan, M.Z. Bian, F.X. Yin, W.M. Zhao, B.S. You, K.S. Shin, *Acta Metall. Sin. (Engl. Lett.)* **32**, 145 (2019)
- [16] M.A. Gibson, X. Fang, C.J. Bettles, C.R. Hutchinson, *Scr. Mater.* **63**, 899 (2010)
- [17] J. She, F. Pan, J. Zhang, A. Tang, S. Luo, Z. Yu, K. Song, M. Rashad, *J. Alloys Compd.* **657**, 893 (2016)
- [18] P. Peng, A.T. Tang, X.H. Chen, J. She, S.B. Zhou, J.F. Song, F.S. Pan, *Crystals* **8**, 356 (2018)
- [19] B.J. Wang, D.K. Xu, S.D. Wang, E.H. Han, *Front. Mech. Eng.* **14**, 113 (2019)
- [20] H.X. Liao, J. Kim, T.T. Liu, A.T. Tang, J. She, P. Peng, F.S. Pan, *Mater. Sci. Eng. A* **754**, 778 (2019)
- [21] P. Cao, M. Qian, D.H. StJohn, *Scr. Mater.* **54**, 1853 (2006)
- [22] S.H. Kim, J.U. Lee, Y.J. Kim, J.G. Jung, S.H. Park, *J. Alloys Compd.* **751**, 1 (2018)
- [23] C.Y. Zhao, X.H. Chen, J.F. Wang, T. Tu, Y. Dai, K.S. Shin, F.S. Pan, *Adv. Eng. Mater.* **21**, 1801062 (2019)
- [24] X. Li, P. Yang, L.N. Wang, L. Meng, F. Cui, *Mater. Sci. Eng. A* **517**, 160 (2009)
- [25] C.F. Fang, G.X. Liu, Z.H. Wen, X.G. Zhang, *J. Alloys Compd.* **686**, 347 (2016)
- [26] C. Wang, H.Y. Wang, H.Y. Zhang, X.L. Nan, E.S. Xue, Q.C. Jiang, *J. Alloys Compd.* **575**, 423 (2013)
- [27] Z.P. Yu, Y.H. Yan, J. Yao, C. Wang, M. Zha, X.Y. Xu, Y. Liu, H.Y. Wang, Q.C. Jiang, *J. Alloys Compd.* **744**, 211 (2018)
- [28] M.A. Azeem, A. Tewari, S. Mishra, S. Gollapudi, U. Ramamurty, *Acta Mater.* **58**, 1495 (2010)
- [29] M.T. Pérez-Prado, O.A. Ruano, *Scr. Mater.* **48**, 59 (2003)
- [30] G. Vespa, L.W.F. Mackenzie, R. Verma, F. Zarandiqi, S. Yue, *Mater. Sci. Eng. A* **487**, 243 (2008)
- [31] Y. Uematsu, K. Tokaji, M. Matsumoto, *Mater. Sci. Eng. A* **517**, 138 (2009)
- [32] J.L. Wang, Y.C. Guo, J.P. Li, Z. Yang, S. Kamado, L.M. Wang, *J. Alloys Compd.* **653**, 100 (2015)
- [33] L.Q. Guan, Y.L. Deng, H.J. Shi, L. Yang, M. Chen, *J. Mater. Eng. Perform.* **27**, 243 (2018)
- [34] X.L. Hou, Z.Y. Cao, L.L. Zhang, D.Q. Fang, S.W. Xu, S. Kamado, L.M. Wang, *Trans. Nonferrous Met. Soc. China* **20**, 508 (2010)