Microstructure and mechanical properties of moderate-speed extrudable Mg-Gd-Sm-Zr alloy

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Abstract
Extrusion Mg-Gd based alloy has the excellent mechanical properties, but usually at the sacrifice of productivity. In this paper, we designed a Mg-10Gd-1Sm-0.5Zr alloy, and have successfully extruded it with a die-exit speed of 3.2 m min−1, and the extrudability is improved by indirect extrusion under differential-thermal condition. The microstructure and mechanical properties of as-extruded and peak-aged alloys were investigated in detail. The results indicate that as-extruded alloys exhibit the bimodal structure composed of dynamic recrystallized (DRXed) grains and UnDRXed grains with good ductility, and its elongation (El) is 14.3% at room temperature and 26.2% at 250 °C. Aging treatment remarkably enhance the strength of as-extruded alloy, and the peak-aged alloy exhibits ultimate tensile strength (UTS) of 345 MPa, yield strength (YS) of 287 MPa and EL of 6.3% at room temperature, and has UTS of 316 MPa, YS of 268 MPa, and EL of 7.8% at 250 °C. β′ precipitation strengthening, fine-grain strengthening, solution strengthening, texture strengthening have responsible for improving strength of peak-aged alloy.

1. Introduction
Magnesium and its alloys have attracted more attention in auto, airplane and other fields due to their low density and good thermal dissipation [1–3]. Most Mg alloy structural parts are produced by casting, indicating that many unavoidable casting defects would further limit their application worldwide. Wrought Mg alloys exhibit the superior performance than that of cast counterparts, and researchers have invested considerable passion in wrought Mg alloy in recent years [4, 5].

Among the wrought methods, extrusion process with strong three-dimensional stress is an easier manipulation with just a single pass [6]. Furthermore, extrusion give play to the maximum plasticity of the metal compared to deformed method such as rolling and forging. It is well-known that direct or indirect extrusion are the two main extrusion method, while indirect extrusion is more effective to improve the extrudability due to relatively less friction and less pressure [5, 6]. In addition, the proper extrusion temperature and mould temperature would bring unexpected results for properties and microstructure such as isothermal extrusion [7, 8] and differential-thermal extrusion [9, 10]. Differential-thermal extrusion refers to that the mould temperature is lower than billets temperature, and it is considered as an effective way to reduce heat and improve mechanical properties in Mg-Gd based alloy during extrusion process to further improve the extrudability [9]. Differential-thermal extrusion can prevent surface hot cracking during the extrusion process by reducing the surface temperature of billet. Rong et al [9] reported that the Mg-RE alloy had the superior mechanical properties under differential-thermal condition compared to isothermal condition.

Extrudability refer to the maximum of extrusion speed of extrudates without appeared visible crack [6]. Good extrudability is related to the productivity of the fabricated products. However, Mg alloys cannot be extruded at a faster rate like aluminum alloy due to their unique structure. It is well-known that extrusion
parameters (including extrusion temperature [11], extrusion ratio [12] and extrusion speed [13]) are significant to develop good extrudates with smooth surface quality and excellent mechanical properties. It is worth mentioning that especially extrusion speed is an important parameter to measure the efficiency of industrialized production [6]. In terms of commercial alloys, maximum extrusion speed of AZ31 alloy is less than 20 m min\(^{-1}\) [6, 14], and AZ91 [14, 15] alloy has an extrusion speed limit of 5 m min\(^{-1}\), since the low-melting phase is the main obstacle to extrusion efficiency. In addition, high alloying content would increase susceptibility to hot cracking due to increase of second phase. Breakthrough in extrudability has been a challenge for as-extruded Mg alloys.

Mg-RE system alloys, especially Mg-Gd based alloys have been developed to be high-performance heat-resistant wrought Mg alloys due to capacity of heat-treatable strengthening [16–18]. The developed Mg-Gd-Y [19], Mg-Gd-Zn [20] and Mg-Gd-Y-Zn [21, 22] extrusion alloys have high strength in recent years. For example, the Mg-Gd-Y-Zr alloy have an ultra-high strength with UTS of 539MPa and YS of 500MPa by extrusion and aging treatment [23]. Although extrusion Mg-Gd based alloys have excellent properties, it is normally achieved at the sacrifice of extrusion efficiency. With respect to Extrudability, Mg-RE alloys have considerable sensitivity to heat deformation than ordinary Mg alloy. Hence, they usually must be extruded at a higher temperature with low speed due to their low extrudability. For example, its ram speed limit is 0.32 m min\(^{-1}\) for Mg-9Gd-3Y-1.5Zn-0.8Zr (wt%) alloy [24]. Moreover, it has still relatively few studies on improving the extrudability of Mg-RE alloy. Hence, it is very essential to expand the extrudability of Mg-RE alloy to further its commercial viability.

Sm belongs to cerium subgroup and has a maximum solubility of 5.8% in solid Mg [25]. It has been reported that Sm and Gd could reduce their solubility in Mg alloys each other indicating a good ageing respond in Mg alloys theoretically [26, 27]. Liu et al [27] reported that cast Mg-Gd-Sm-Zr alloys have the excellent strength, and Mg-10Gd-2Sm-0.5Zr alloy has a UTS of 347MPa and YS of 237 MPa. However, relatively few works on extruded Mg-Gd-Sm-Zr alloy have been made in previous studies, and only compression behavior is conducted recently [3, 28]. It is very imperative to study the current Mg-Gd-Sm-Zr system alloys. Here, we have tried to extrude the Mg-10Gd-1Sm-0.5Zr alloy at a high speed with a high temperature. Microstructural characteristics and tensile properties of the alloy are discussed after hot extrusion.

### 2. Experiment producers

The Mg-10Gd-1Sm-0.5Zr alloy ingots were fabricated by melting high purity Mg (99.95 wt%), Mg-30 wt% X (X = Gd, Sm and Zr) master alloys in a medium frequency induction furnace with passing a protective atmosphere of CO\(_2\) and SF\(_6\) mixed gas. The melt at 750 °C was held for 10 min, and static cool down to 730 °C, pouring then the melt into a pre-heated metal mold. Billets with Φ50 mm × 40 mm cut from ingots were homogenized treatment at 515 °C for 10 h followed by hot water quenching. To avoid the hot shortness, the billets were pre-heated for 2h at 500 °C, and the mold were preheated to 350 °C before extrusion. To reduce the deformation friction, the graphite and engine oil were used as a lubricant adding into die before during extrusion. Extrusion was conducted at die-exit speed of 3.6 m min\(^{-1}\), 3.2 m min\(^{-1}\) respectively employing four-column hydraulic press. H13 hot-work die steel was selected as extrusion mold, and extrusion ratio was 9.67. After extruding, the extrudates was immediately quenched with hot water. Age hardening behaviors were investigated at 225 °C.

The sheet-shaped samples with a gauge length of 15 mm, width of 3 mm and thickness of 2 mm were processed for tensile tests at room temperature (25 °C), 250 °C. The tensile tests were carried out by using SHIMADZU AG-1250 N universal testing machine with a speed of 1 mm min\(^{-1}\). Before testing, the samples were heated to specified temperature and held for 8 min. The values of Brinell hardness were measured along the extrusion direction (ED) under a load of 250 Kgf for 30 s. The actual chemical compositions of Mg-10Gd-1Sm-0.5Zr alloy were measured by inductively coupled plasma (Optima 8000), and the results were listed in table 1. The phase structures were detected by the D8 Advance X-ray diffractometry (XRD) at 40 kV and 40 mA with a scanning speed of 2 degree min\(^{-1}\) using Cu K-\(\alpha\) radiation. The microstructure observed using JSM-7800F scanning electron microscope (SEM) equipped with Oxford C-nano electron backscatter diffraction (EBSD) apparatus at 15 kV, and JEM-2100 high resolution

| Alloy       | Mg  | Gd  | Sm  | Zr  |
|-------------|-----|-----|-----|-----|
| Mg:10Gd-1Sm-0.5Zr | Bal. | 10.12 | 0.94 | 0.38 |
transmission electron microscopy (TEM) at 200 kV. The microstructure and texture of as-extruded alloy were characterized by SEM-EBSD method with step size 0.7 μm. The EBSD data are analyzed using Aztec Crystal software and Channel 5 software. The TEM foils were prepared of extruded-T5 alloy by grinding to thickness of approximately 70 μm, and then ion-polished to perforation using a Gatan Precision Ion Polishing System (GATAN691).

3. Results and discussion

3.1. Microstructure of as-cast and as-homogenized alloy

Figure 1 shows the SEM images and EDS results of as-cast alloys. It is obviously seen that the continuous reticular eutectic structures are distributed along grain boundaries and some can be seen within grains as shown in the figures 1(a) and (b). The EDS mappings are used for analyzing elements distribution of eutectic structures qualitatively which is shown in figures 1(c)–(e), and the results are that affluent Gd elements are concentrated in eutectic structures while Sm content is inferior to Gd element. In addition to eutectic structures, some clustered particles are captured as well. Meanwhile, figures 1(f) and (g) show the corresponding EDS results of remarked regions in figures 1(a)–(b), and found that clustered particles are rich in abundant Zr element, indicating that it may be regarded as the heterogeneous nucleation during non-equilibrium solidification. Moreover, the EDS results of eutectic structures B is consisted of Mg, Gd, Sm elements which is consistent of results of mappings.

Figure 2 shows the XRD patterns of as-cast and as-homogenized Mg-10Gd-1Sm-0.5Zr alloy. It indicates that α-Mg and Mg₅(Gd, Sm) compound are the primary phases in as-cast alloy. Combined with the EDS analysis results, the eutectic structures may contain the Mg₅(Gd, Sm) compound. After solution, the Mg₅(Gd, Sm) phase is disappeared, and the GdH₂ phase is detected.

In order to analyze the evolution of microstructure clearly, the SEM microstructure of as-homogenized alloy is depicted in figure 3(a). It is clear shown that the eutectic structures are disappeared, and some cubic particles are observed. The corresponding EDS results of remarked regions are composed of Mg, Gd elements which are
shown in figure 3(c). Interestingly, these cubic phases are frequently observed in Mg-RE alloys [29]. With combination of XRD results, these cubic phases can be identified as GdH$_2$ compounds. It is reported that GdH$_2$ compounds are early formed in cast process, and solution treatment could easily lead the growth of RE hydrides which are deemed to have an unfavorable effect on mechanical properties [30, 31]. After homogenization treatment, the grains shown in figure 3(b) are relatively uniform and its average size is approximately 65 μm, and it exhibits a random texture as shown in figure 3(d).
3.2. Extrudability of the alloy

In general, Mg-Gd based alloys containing numerous rare earth elements are extremely difficult to extrude with a high speed due to strong deformation sensitivity. They are normally extruded in a narrow processing window. Furthermore, extrudability of Mg-RE alloys had been studied limitedly in early stages [28, 32]. Hence, we try to extrude Mg-Gd based alloy with a proper speed under a certain condition using indirect extrusion. To reduce a sharp increase of heat, mould is preheated to 350 °C which is below billets temperature during extrusion process. Figure 4 shows the images of extrudates at die-exit speed of 3.2 m min\(^{-1}\) and 3.6 m min\(^{-1}\). Obviously, extrudates have two types of surface features after extrusion. The hot cracks are observed at the condition of 3.6 m min\(^{-1}\) as shown in figure 4(a) while its surface is smooth and complete at a die-exit speed of 3.2 m min\(^{-1}\) in figure 4(b).

It is well-known that the hot crack is extremely sensitive to extrusion speed due to more heat generation and surface stress with increase of extrusion speed [33]. The shortness may be occurred on the surface of extrudates, when the heat exceeded the incipient melting point [33, 34]. Similarly, the bad surface quality might occur when the maximum stress acting on surface beyond limit of phase [6, 34]. Moreover, the localized solute-enriched regions are likely to have a lower local solidus temperature, indicating that hot cracking may occur easily during extrusion. The figure 4 show that hot cracks exist in the upper part of the extruded bar but not in other parts, indicating that heat increasement may exceed solidus temperature of localized unstable regions at speed of 3.6 m min\(^{-1}\). However, the alloy could be extruded with a smooth surface at speed of 3.2 m min\(^{-1}\), indicating that increased heat not exceed the incipient melting point, and maximum stress is also reduced in this condition. Moreover, the obtained extrudates at such a speed are also attributed to differential-thermal extrusion, since mould with the lower temperature could absorb more heat which means that it might further reduce the possibility of shortness existing during extrusion.

3.3. Microstructure and properties of as-extruded alloy

Figure 5 shows the SEM images and EDS results of as-extruded alloy. A majority of cubic phase is broken up while several are still complete in the matrix. The fair-shaped solute has a redistribution along ED forming intermetallic as shown in figure 5(a). The average size of the intermetallic is in the range of 0.1–2 µm. Moreover, it should be noting that no dynamic precipitations behavior occurs in as-extruded Mg-10Gd-1Sm-0.5Zr alloy. The EDS analysis results of the second phases which are marked region indicate that these intermetallic contains the Gd and Sm elements as shown in the figures 5(c)–(d). Moreover, it should be noting that the relatively high concentration of Gd and Sm atoms are contained in Mg matrix, indicating that it might play an important role for solution strengthening.

To further illustrate the grain structure and its orientation of as-extruded Mg-10Gd-1Sm-0.5Zr alloy, the EBSD IPF maps along with ED are depicted in figures 6(a) and (b). The results reveal that the microstructure is remarkably refined after hot extrusion by means of DRX behavior. The as-extruded Mg-Gd-Sm-Zr alloy
Figure 5. (a) and (b) SEM images of as-extruded Mg-10Gd-1Sm-0.5Zr alloy; corresponding EDS results of marked region in (a) and (b).

Figure 6. EBSD maps of as-extruded Mg-10Gd-1Sm-0.5Zr alloy. (a) and (b) IPF maps; (c) KAM map; (d) (0001) (11−20) Schmid factor maps when the tensile stress is applied along the ED.
exhibits bimodal grain structures which consist of DRXed and UnDRXed grains which is shown in figure 6(b), and DRX ratio is approximately 83%. Moreover, the average grain size is about 8.5 μm. The kernel average misorientation (KAM) maps indicate that less local plastic deformation occurs in UnDRXed grains as shown in figure 6(c) indicating a certain density of dislocations may be accumulated in UnDRXed grains. It should be noting that the formation of bimodal grain structures might be related to differential-thermal extrusion. The lower mold temperature might result in inadequate DRX behavior due to provided insufficient energy. It is reported that the bimodal grain structures possessed higher tensile strength compared with the fine recrystallized grains, this is due to strong and hard un-recrystallized texture strengthening where its basal planes nearly parallel to the loading direction [10, 11, 35]. Figure 6(d) shows the (0001) (1120) Schmid factors (SF) distribution map of the as-extruded alloy when the tensile stress is applied along the ED. It is obviously that the UnDRXed grains exhibit the lowest average SF for basal slip of as-extruded alloy, and it suggests that the basal slip for most unDRXed grains displays a hard orientation and is difficult to be activated indicating that it has a favorable effect on improvement of strength.

Figure 7 shows the (0001) pole figures and IPFs of as-extruded Mg-10Gd-1Sm-0.5Zr alloy. Inverse pole figures refer to the ED. TD indicates the transverse direction. Interestingly, as-extruded alloy exhibits a bimodal texture which consists of (10-10) and weak (0001) texture components, that is, (10-10) axis of the Mg matrix parallel to ED, and (0001) axis of the Mg matrix parallel to ED. It is generally accepted that the typical fiber texture component with (10-10) axis of the Mg matrix parallel to the ED is usually formed during hot extrusion. While a weak abnormal texture with the (0001) axis of the Mg matrix parallel to ED is obtained in DRXed region. The maximum texture intensity is primarily dominated by unDRXed regions, while the unusual texture component is mainly affected by DRXed regions. The formation of weak abnormal texture can be related to DRX nuclei at local shear bands during hot extrusion. The formation of shear bands promoted by REs can leads to the formation of more homogeneously distributed shear bands than in conventional Mg alloys during hot extrusion [36]. In addition, solute drag effect is also a feasible mechanism resulting in texture modification. The solute segregation at grain boundaries or dislocation core could impede the migration of grain boundaries (GBs) and movement of dislocations due to slow diffusion rate of RE [37, 38]. Stanford et al [38] reported that the addition Gd to Mg could effectively reduce the angle between the c-axis and ED. Lyu et al [39] also found this abnormal texture with the (0001) axis of the Mg matrix parallel to ED in as-extruded Mg-Y-Sm-Zr alloy, and they concluded that drag effect, growth of the specific orientation, and transformation of deformation modes might be the main mechanisms. As a result, it is reasonable to suppose that various factors such as shear band formation, transformation of deformation modes, and the solute drag effect might be resulting in the formation of abnormal texture with c-axis parallel to the ED in as-extruded Mg-10Gd-1Sm-0.5Zr alloy.

Figure 8(a) shows the nominal tensile stress-strain curves of as-extruded Mg-10Gd-1Sm-0.5Zr alloy at room temperature (25°C) and 250 °C, and corresponding mechanical parameters of tensile curves are depicted in figure 8(b). The as-extruded alloy exhibits good ductility, and it has UTS of 237 MPa, YS of 165 MPa and EL of
aging precipitation sequence of Mg-Gd based alloys are: S.S.S.S. - maximum hardness reaches 128 HBW at aging 12h, and then decrease with aging time increase. It is well-known initial hardness is 90 HBW for as-extruded alloy, and the hardness increases dislocations at elevated temperature effect on dislocations and produce DSA effect at elevated temperature solute drag force to produce Cottrell atmosphere near the dislocation core, while solute atoms have great drag and EL of as-extruded alloy are 234MPa, 160MPa and 26.2%, respectively at 250 °C. In addition, it should be pointed that the tensile curve at 250 °C exhibits obvious sawtooth shape, and this is caused by dynamic strain aging (DSA) behavior. It is reported that DSA behavior is caused by the interaction between solute atoms and dislocations at elevated temperature [40]. At room temperature, the slow solute diffusion velocity is no enough solute drag force to produce Cottrell atmosphere near the dislocation core, while solute atoms have great drag effect on dislocations and produce DSA effect at elevated temperature [40]. As-extruded Mg-10Gd-1Sm-0.5Zr alloy having high-thermal stability may be associated with drag effect of solute. The faster diffusion of Gd solute atoms at high temperature can form large enough atomic gas clusters near the dislocations, and solute atoms have inhibition effect on the dislocations by dragging effect although YS has a certain decline.

3.4. Microstructure and properties of peak-aged Mg-10Gd-1Sm-0.5Zr alloy
The optical image, TEM bright field images and its corresponding selected area electron diffraction pattern (SAED) are provided in figure 9 for peak-aged Mg-10Gd-1Sm-Zr alloy. The electron beam is parallel to the [1210]α direction. After aging treatment, bimodal grain structures are largely unchanged, and average grain size is about 9 μm. Observed by TEM, large amounts of precipitations with elliptoid shape are captured in peak-aged alloy as shown in the figure 7(b) and (c). The diffraction spots at 1/4, 2/4 and 3/4 α-Mg are clearly observed in the corresponding SAED patterns as shown in figure 9(d). After the calibration, the precipitation is identified as β′ phase which has a bottom center orthogonal structure (a = 0.64 nm, b = 2.22 nm, c = 0.52 nm) [41]. To capture the structure of β′ phase clearly, the β phase is lens-shaped in the at high resolution as shown in figure 9(e). The orientation relationship between the β′ phase and the α-Mg matrix is [100]β′//[0001]α, (001)β′//(0001)α. The abundant β′ precipitations have the significant role on mechanical properties. Figure 10 shows the aging hardening curve of as-extruded Mg-10Gd-1Sm-Zr alloy treated at 225°C. The initial hardness is 90 HBW for as-extruded alloy, and the hardness increases firstly with the aging time. The maximum hardness reaches 128 HBW at aging 12h, and then decrease with aging time increase. It is well-known that aging precipitation sequence of Mg-Gd based alloys are: S.S.S.S. - β′ - β′ - β1 - β, and β′ phase is the main strengthening contributor at peak-aged state for improving mechanical properties [42]. Hence, the improvement of hardness is due to precipitated β′-type phase at early aging stage, and β′ precipitation is the main strengthening phase to enhance the strength at peak-aged state. With the increase of aging time, β′ precipitation transformed as equilibrium β′ phase, and its strengthening effect is inferior to β′ precipitation.

Figure 11 shows the nominal tensile stress–strain curves of extruded-T5 alloy at room temperature (25°C) and 250°C, and corresponding mechanical parameters are shown in figure 11(b), indicating that aging treatment has a remarkable effect on improvement of strength. The peak-aged Mg-10Gd-1Sm-0.5Zr alloy exhibits UTS of 343 MPa, YS of 284 MPa and EL of 6.3% at room temperature. In addition, peak-aged alloy also has a high strength at elevated temperature, and its value of UTS, YS, and EL are 316 MPa, 268 MPa and 7.8%, respectively at 250 °C. To date, extrusion Mg-Gd based alloys have been reported having high strength such as extruded-T5 Mg-14Gd-0.5Zr alloy has a UTS of 446MPa and YS of 305MPa [43]. They usually are extruded with an extremely low speed at the sacrifice of extrusion efficiency. Here, the Mg-10Gd-1Sm-0.5Zr alloy can be extruded at die-exit speed of 3.2 m min⁻¹, and it exhibits the good mechanical properties after aging treatment, indicating that its great potential for industry feasibility.
The excellent strength of Mg-10Gd-1Sm-0.5Zr alloy is attributed to solution strengthening, fine-grain strengthening and texture strengthening. Hume-Rothery rule requires the difference in the atomic radii for the formation of a replacement solid solution to be less than 15%. The atomic radius difference between RE elements (Gd, Sm) and Mg is within 15% meeting the Hume-Rothery rule requires, indicating that RE elements has solution strengthening effect in Mg. As seen from the EDS results of region B in figure 5, as-extruded Mg alloy contains 12% RE atoms, indicating they could produce solid solution strengthening. The RE atoms dissolved into the \( \alpha \)-Mg matrix causes the asymmetrical aberrations for the lattice of the \( \alpha \)-Mg matrix, further forming an elastic strain field that interacts with the dislocations. The fine DRXed grains provide the large density of grain boundaries which means that they could provide numerous obstacles to the movement of dislocations. In comparison to the fine DRXed grain boundaries, strengthening of coarse unDRXed grains have a negligible effect on strength according to the Hall–patch equation. Although coarse unDRXed grains have less grain boundary strengthening effect, the strong fiber texture component of coarse unDRXed grains may have

**Figure 9.** Optical and TEM images of peak-aged Mg-10Gd-1Sm-0.5Zr alloy; (a) optical image, (b) and (c) bright field images, (d) its corresponding SAED pattern, (e) HRTEM image. Electron beam is parallel to \([2\bar{1}0]_\alpha\).

**Figure 10.** Brinell hardness evolution as a function of aging time during isothermal aging at 225 °C.
positive effect on strength. The specific crystallographic orientation leads to a distinct low SF for basal slip. It is well-known that the value of SF is normally used for evaluating capacity of slipping systems, and low value of SF represents high CRSS for slipping which means it is easy to be activated [44]. Hence, it is relatively difficult to activate their basal slip further resulting in improvement of strength. As shown in figure 6(d), almost unDRXed grains have the low SF, implying that basal slip of unDRXed grains is difficult to be activated. Hence, unDRXed grains also play a significant role for strength of Mg-10Gd-1Sm-0.5Zr alloy.

\( \beta' \) precipitations also play a significant role for enhancing the strength. It is widely acknowledged that Mg-RE alloys are the typical heat-treatable alloys due to high solid solubility of rare earths in Mg alloy. The precipitation has good interface relationship with matrix indicating that it has favorable inhibition effect of dislocation movement. With respect to Mg-Gd based alloy, it is previously reported that the \( \beta' \) phases formed on the prismatic plane can effectively hinder basal slip in the dominant slip mechanism of Mg alloy [7, 45]. Moreover, it has a high thermal stability and have a great contribution to strengthen mechanical properties in Mg alloys. The abundant \( \beta' \) strengthening phase observed in Matrix distribute in matrix uniformly as shown in the figure 9. The orientation relationship between \( \beta' \) phase and \( \alpha\)-Mg matrix is \([100]_{\beta'}//[0001]_{\alpha}, (001)_{\beta'}//(0001)_{\alpha}\). It indicates that benign interface relationship exists between \( \beta' \) phase and \( \alpha\)-Mg matrix. The good coherent strain field has the interaction with dislocation to further result in high yield strength. In addition, the excellent strength at elevated temperature is also contributed to existence of \( \beta' \) phase for Mg-10Gd-1Sm-0.5Zr alloy. It is reported that \( \beta' \) phase could exist stably below 250 °C which means that \( \beta' \) phase also stabilize the strength at elevated temperature [46]. Hence, the alloys could exhibit the good strength at elevated temperature through \( \beta' \) phase inhibiting on dislocation slip.

4. Conclusion

In this work, we have successfully extruded the Mg-10Gd-1Sm-0.5Zr alloy with a moderate-speed extrudable speed (3.2 m min\(^{-1}\)) under differential-thermal condition. The current results are expected to broaden our knowledgebase to develop extruded Mg alloys with good extrudability to expand their viable commercial value. The Mg-10Gd-1Sm-0.5Zr alloy exhibits excellent mechanical properties, and following conclusions can be drawn:

1. The Mg-10Gd-1Sm-0.5Zr alloy can be extruded successfully without any defects at die-exit speed of 3.2 m min\(^{-1}\). The decrease of heat generated is achieved through differential-thermal extrusion method during indirect extrusion.

2. As-extruded Mg-10Gd-1Sm-0.5Zr alloy exhibits the bimodal grain structures and bimodal texture, that is, \(\{10-10\}\|\{0001\}\) ED texture components and \(\{001\}\|\{0001\}\) ED texture components. The abundant \( \beta' \) precipitations are introduced after aging treatment. The orientation relationship between the \( \beta' \) phase and the \( \alpha\)-Mg matrix is \([100]_{\beta'}//[0001]_{\alpha}, (001)_{\beta'}//(0001)_{\alpha}\).

3. Solution strengthening, fine grain strengthening and texture strengthening are considered to be the main strengthening mechanisms of as-extruded alloy, while \( \beta' \) precipitation strengthening have responsible for the excellent mechanical properties for peak-aged alloy besides the above strengthening mechanisms.
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Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

Declaration of competing interest

The authors declare that they have no competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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