Influence of heat treatment on the tensile properties and fatigue properties of Mg-8.8Gd-3.5Y-1.5Zn-0.5Zr alloy

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Abstract
The microstructure, tensile properties and fatigue properties of Mg-8.8Gd-3.5Y-1.5Zn-0.5Zr alloy under T4 and T6 heat treatment have been studied. The as-cast alloy was homogenized to eliminate the inhomogeneous structure and the strength of the alloy was increased by aging treatment. The microstructure evolution and mechanical properties of the alloy were investigated by optical microscope, scanning electron microscope, energy dispersive spectrometer and universal mechanical testing machine. The as-cast alloy is composed of dendritic α-Mg matrix, metastable lamellar stacking fault, eutectic Mg24(Gd, Y, Zn)5, massive long-period stacking ordered (LPSO) phase Mg12(Gd, Y)Zn and a few rare earth rich phases. In the process of homogenization, the lamellar phase and Mg24(Gd, Y, Zn)5 eutectic phase dissolved in the α-Mg matrix gradually, and the massive LPSO phase disappeared gradually, while the lamellar LPSO phase steadily grew into grains and formed some precipitate particles near the grain boundary. As the tensile temperature increases, the ductility of solid solution alloy increases, but decreases at 250 °C, which is related to the melting reaction of γ phase at high temperature and the cracking caused by grain boundary sliding. Intergranular fracture occurs at 250 °C. When the tensile temperature rises to 150 °C, the ultimate tensile strength of the alloy increases first before 150 °C and decreases after 150 °C, which is related to two strengthening mechanisms, namely, dissolution of Gd and Y elements into α-Mg matrix results in solid solution strengthening and the co-lattice strengthening between petal β′ and α-Mg matrix. Compared with the fatigue properties of the alloy after solution treatment and aging treatment, the fatigue life of the alloy after aging treatment is longer.

Introduction
In the last few years, Mg alloys have attracted the attention of many scholars and been studied extensively. With the emergence of Mg alloys rich in rare earth elements, especially those containing gadolinium (Gd), more and more scholars are devoted to the study of rare earth magnesium alloys with good ductility and excellent thermal stability. According to related studies, Gd dislocation and grain boundary segregation can improve the high temperature strength of the alloy, and the growth of recrystallization is hindered by solute dragging mechanism [1–6]. In addition, Mg alloys are mainly used in aviation, aerospace, transportation, chemical industry, rocket and other industrial sectors, automotive parts and so on due to their many advantages such as high strength, large elastic modulus and good heat resistance. Mg-Gd alloy is one of the most valuable alloys because of its good aging strengthening effect (up to 250 °C) [7–11].

The addition of zinc in Mg-gadolinium or Mg-yttrium systems can form 18 R, 24 R, 6H, 10H, 14H LPSO structures, which are of great help to improve the comprehensive mechanical properties of magnesium alloys [12–17]. The existence of LPSO structure can improve the mechanical properties of rare earth Mg alloys at room
or high temperatures. After solidification of the alloy will inevitably appear defects, such as alloy composition segregation. Therefore, homogenization is necessary. Homogenization can not only eliminate the dendritic segregation of the as-cast structure, but also adjust the plastic deformation by changing the amount and distribution of the second phase. Both solution treatment and aging treatment can improve the mechanical properties of the alloy [12, 18, 19]. The aging process of Mg-Gd-Y-Zn-Zr series alloys was studied, the order of precipitation has a great relationship of the elements contained in the alloy, especially the Zn content [12, 20, 21]. \( \beta \) phase is usually precipitated in alloys with high Zn content, such as Mg-11Gd-4Y-2Zn-0.5Zr alloy. At the same time, \( \beta \) phase also precipitates in the alloy with less zinc content, such as Mg-9Gd-3Y-0.6Zn-0.5Zr alloy. The order of precipitation of this alloy is \((SSSS) \rightarrow (D0_{19}, Mg_3RE) \rightarrow (bcco,Mg_{15}RE_3) \rightarrow 14H-Mg_3RE) \rightarrow (bco)\) [22]. In addition, Xue et al [23] reported that Mg-9Gd-3Y-1.5Zn-0.5Zr alloy precipitated at 200 °C and 225 °C with neither \( \beta \) nor \( \beta^\prime \) phase and only a small number of 14H-LPSO. In this paper, the aging temperature is 175 °C, the \( \beta \) phase can exist stably below 200 °C, and the precipitation order of the alloy is \((SSSS) \rightarrow (D0_{19}) \rightarrow (bco)\). It is reported that \( \beta \) is an important strengthening phase in Mg-Gd-Y alloy, and its morphology is a convex lens shape, and its thermal stability is very strong [24].

At the same time, the fatigue life and stability of magnesium alloy are also worth studying. High cycle fatigue is studied in this paper, the fatigue performance is closely related to the initiation zone and propagation zone of fatigue crack, especially the crack propagation zone. There have been many studies on the fatigue properties of magnesium alloys, but few studies on the high cycle fatigue properties of magnesium alloys by heat treatment. Dong et al. studied the fatigue properties of Mg-10Gd-3Y after heat treatment and concluded that aging treatment could significantly improve the fatigue properties of GW103 alloy [25]. For a deeper understanding of the fatigue category, it is necessary to study the fatigue properties of other kinds of alloys, especially the effect of heat treatment on them. Up to now, there are few studies on the comprehensive mechanical properties of Mg-Gd-Y-Zn-Zr alloy after heat treatment. Based on the above reasons, the microstructure, tensile properties (RT and high temperature) and fatigue properties of Mg-8.8Gd-3.5Y-1.5Zn-0.5Zr alloy after heat treatment (solution and aging treatment) were studied in this paper.

Experimental process

Firstly, alloys with nominal composition Mg-8.8Gd-3.5Y-1.5Zn-0.5Zr (wt %) are prepared in a resistance furnace using commercial pure magnesium (> 99.9%) and Zn, Mg-30Gd (at %), Mg-30Y (at %) and Mg-25Zr (wt %) master alloys in a CO\(_2\) and SF\(_6\) atmosphere, made by semi-continuous casting in a ratio of 100 : 1. The as-cast alloy is then subjected to two heat treatments, T4 and T6, respectively. During the T4 heat treatment of the as-cast alloy, the solution temperature was 530 °C and the time was 12 h. In order to retain the microstructure of the alloy after the solution, water quenching was carried out at the temperature of 70 °C. The aging treatment was subsequently conducted at 175 °C to peak hardness. The hardness was determined by UHL VMHT Vickers hardness tester. The loading load is 200 g, and the holding time is 15 s.

The microstructure of as-cast and heat-treated alloys was observed using an Axio Observer A1m Carl Zeiss optical microscope. Samples examined under a light microscope need to undergo chemical corrosion in the solution of 1.1 g picric acid, 16 ml alcohol, 1 ml acetic acid and 2 ml distilled water. For a better view, the morphology of the alloy phase was observed using Hitachi SU5000 scanning electron microscope (SEM) with a voltage of 20 kV. At the same time, the fracture morphology of the samples was observed by scanning electron microscopy. A 3382 Instron universal material experiment machine was used to conduct tensile tests on the samples along the ED direction at room temperature, 100 °C–300 °C, and the strain rate was 1 \( \times \) \( 10^{-3} \) s\(^{-1}\). The tests were repeated three times at each temperature and averaged to obtain the tensile mechanical properties.

Solution and aging treatment (peak aging) were performed on the as-cast samples respectively, the samples were

![Figure 1: Tensile and fatigue specimens. (a) RT tensile specimen; (b) High temperature tensile specimen; (c) fatigue specimens.](image-url)
processed into smooth bars and then stretched at room and high temperatures. Room temperature and high temperature stretch bars are shown in figures 1(a) and (b).

As shown in figure 1(c), the fatigue sample is an hourglass shape, and its surface is polished to make it smooth. The minimum diameter of the sample is 5 mm. The fatigue test was performed on Instron 8801 servo-hydraulic test frame with an axial load of 100kN. The resonance frequency is selected between 0.25 Hz and 1 Hz according to different stress amplitude conditions and the stress ratio of $R = -1$. The cycle times of specimen fracture were defined as fatigue life and the experiment was stopped. The fracture of fatigue specimen was observed by scanning electron microscope (SEM).
Results and discussion

Microstructure of as-cast alloy
Temperature of solid solution treatment of the Mg-8.8Gd-3.5Y-1.5Zn-0.5Zr alloy was determined according to literature data and differential scanning calorimetry (DSC) curves [26, 27]. The DSC curve of Mg-8.8Gd-3.5Y-1.5Zn-0.5Zr alloy is shown in figure 2. According to the DSC curve, 530 °C is determined as the temperature for the solution treatment in our study. In addition, the choice of solution time is also very important. Therefore, we choose the most suitable time is 12 h [28].

Figure 3 depicts the microstructure of as-cast Mg-8.8Gd-3.5Y-1.5Zn-0.5Zr alloy. As shown in figure 3, the α-Mg, network-shaped eutectic compounds and lamellar phases are the main constituent phases of the as-cast alloy. According to figure 3, network-shaped eutectic compounds are mainly distributed and grew along grain boundaries, two different contrasts, i.e., bright white and dark gray are observed (figure 3(d) shows the image), marked as point A and point B, respectively. According to some previous studies [29], the bright white phase is Mg$_{24}$(Gd, Y, Zn)$_5$, the dark gray phase is Mg$_{12}$(Gd, Y, Zn). In addition, the bright white phase in cuboid-shaped particles marked C as shown in figure 3(d) are RE-riched phase [30].

Microstructure evolution of homogenized alloy
Figures 4(a) and (b) show the OM structure of Mg-8.8Gd-3.5Y-1.5Zn-0.5Zr alloy after solution treatment. The solution temperature and time are 530 °C and 12 h, respectively. Contrast with the as-cast alloy in figure 3, the microstructure of the as-cast alloy is more homogeneous, and the network-shaped phase in the as-cast alloy gradually dissolves into the α-Mg matrix and forms discontinuous eutectic compounds. Observe figures 3 and 4, it can be seen that the grains change from dendritic to equiaxed, and the grain boundaries become clearer. The coarse and irregular network-shaped eutectic compounds transform into smaller banded crystals. As a result, the grain size is less restricted. Figure 4(c) shows the microstructure image of the solid solution alloy observed by SEM back scattered electron (BSE) mode. After solution treatment for 12 h, the bright white phase Mg$_{24}$(Gd, Y, Zn)$_5$ in eutectic phases are almost invisible and are dissolved in the α-Mg matrix, while the gray phases (block-shaped LPSO phases) were only partially dissolved. In the process of solid solution treatment, in addition to the precipitation of supersaturated α-Mg matrix and the decomposition of eutectic phase, some small particles...
precipitate at grain boundaries. The metastable lamellar phase is dissolved in the $\alpha$-Mg matrix during the solution treatment and formed obvious straight lamellar phase at grain boundary. As shown in figures 4(a) and (b), there are many precipitates in the $\alpha$-Mg matrix and these precipitates are irregularly distributed, as marked by yellow ellipses, especially near grain boundaries. After 12 h of solution treatment, the block-shaped LPSO phase dissolves into the $\alpha$-Mg matrix, resulting in lattice distortion and solution strengthening. In addition, the dislocation slip, that is, the tensile strength of the matrix is improved. During solution treatment, the second phase dissolves into the $\alpha$-Mg matrix, resulting in lattice distortion and solution strengthening. In addition, the grain size does not change significantly. There are many second phase precipitates at the grain boundary, which precipitate at grain boundaries. The metastable lamellar phase is dissolved in the $\alpha$-Mg matrix during the solution treatment and formed obvious straight lamellar phase at grain boundary. As shown in figures 4(a) and (b), there are many precipitates in the $\alpha$-Mg matrix and these precipitates are irregularly distributed, as marked by yellow ellipses, especially near grain boundaries. After 12 h of solution treatment, the block-shaped LPSO phase at grain boundary gradually disappeared and then changed to the lamellar LPSO phase.

According to figure 4(c), there are many rectangular particles distributed in the matrix, which are considered to be rich in REs. Compared with the as-cast alloy, there are more rectangular particles rich in REs after solution treatment than in as-cast alloy. The square-shaped RE-rich phases are relatively stable and play a role of pinning grain boundary, which makes the Mg- Gd- Y- Zr alloy insensitive to high temperature. Thus, the mechanical properties of the alloy are improved. Figures 4(d)–(g) show the distribution of Gd, Y, Zn, Zr after 12 h of homogenization at 530 °C. The four elements are distributed uniformly in the matrix during solution treatment. The segregation of dissolved eutectic compounds can be eliminated by the enrichment of Gd and Zn in the interdendritic eutectic phase. Suzuki et al [31] show that the addition of Y and Zn could reduce the stacking fault energy. In conclusion, the strength of Mg alloy is improved by solution treatment.

**Mechanical properties**

**Tensile properties of as-cast and homogenized alloys**

The stress–strain curves of as-cast alloy at 20 °C–250 °C are shown in the figure 5(a). By observing the tensile curve of the alloy, the tensile strength of the as-cast alloy reaches its maximum value at 200 °C, which is 248 MPa, however, it is reduced to 212 MPa at 250 °C. And the elongation gradually increases from room temperature to 250 °C, reaching a maximum value of 13.8% at 250 °C. The stress–strain curves of the alloy at 20 °C-250 °C after solution treatment at 530 °C for 12 h are shown in figure 5(b). According to figure 5(b) and table 1, as the temperature goes up, the ultimate tensile strength reaches the maximum value at 150 °C and then decreases gradually. The strength of the solid solution alloy increases significantly to 268 MPa at 200 °C. When the temperature rises to 250 °C, the ultimate tensile strength decreases to 243 MPa. As the temperature increased from room temperature to 200 °C, the elongation gradually increased from 4.9% to 14.6%. But at 250 °C, the elongation drops to 12%. The melting reaction of $\gamma$ phase at high temperature and the crack caused by grain boundary slip can decrease the elongation of GW alloy. By observing the elongation data of the solid solution alloy in table 2, it can be seen that the elongation decreases 12% at 250 °C. Compared with figures 6(a), (b) and (c), it can be seen that the fracture morphology is intergranular fracture, which is related to grain boundary embrittlement, as reported by Li et al [32]. Zheng et al [33] reported that there are many factors leading to intergranular brittle fracture, such as precipitates at grain boundaries, stress concentration at grain boundaries or inclusions and impurity segregation in tensile experiments. In the next section, the tensile fracture characteristics of the solution treated alloy will be analyzed in detail. Table 1 lists the mechanical properties of as-cast and solid solution samples at room and high temperatures. According to figure 5 and table 1, the mechanical properties of the alloy treated by solution treatment at room and high temperatures are better than those of the as-cast alloy. Obviously, the strength of the matrix is enhanced by the inhibition of precipitation relative dislocation slip, that is, the tensile strength of the matrix is improved. During solution treatment, the second phase dissolves into the $\alpha$-Mg matrix, resulting in lattice distortion and solution strengthening. In addition, the grain size does not change significantly. There are many second phase precipitates at the grain boundary, which...
Table 1. Mechanical properties of Mg-8.8Gd-3.5Y-1.5Zn-0.5Zr alloy at different temperature; as-cast and T4 temper.

| Alloy          | Temper | UTS/MPa | TYS/MPa | Elongation/% |
|----------------|--------|---------|---------|--------------|
| Mg-8.8Gd-3.5Y-1.5Zn-0.5Zr |        |         |         |              |
| As-cast        | RT     | 201     | 116     | 2.8%         |
|                | 100 °C | 223     | 157     | 6.0%         |
|                | 150 °C | 234     | 143     | 8.2%         |
|                | 200 °C | 248     | 152     | 12.5%        |
|                | 250 °C | 212     | 129     | 13.8%        |
|                | T4     | 247     | 123     | 4.9%         |
|                | 150 °C | 252     | 143     | 7.3%         |
|                | 200 °C | 265     | 160     | 9.8%         |
|                | 250 °C | 268     | 171     | 14.5%        |
|                |        | 243     | 170     | 12%          |
Figure 6. Tensile fracture diagram of solid solution alloy at different temperatures. (a) 150 °C; (b) 200 °C; (c) 250 °C; (d) 300 °C.

Table 2. Elongation of a solid solution alloy at different temperatures.

| T/°C | 150 | 200 | 250 | 300 |
|------|-----|-----|-----|-----|
| Elongation/% | 9.8 | 14.5 | 12 | 22.5 |

Figure 7. Age hardening curve of Mg-8.8Gd-3.5Y-1.5Zn-0.5Zralloy at 175 °C.
hinders the dislocation movement. In conclusion, the TYS value of solid solution alloy is higher than that of as-cast alloy.

Fracture morphology of homogenized alloys
Table 2 shows the elongation of the alloy after solid solution at 150 °C-300 °C, and the elongation decreases at 250 °C. Figure 6 shows the fracture diagram of the solid solution alloy at the corresponding temperature in table 2. As shown in figures 6(a) and (b), the sample showed transgranular fracture at 150 °C and 200 °C, and the cleavage planes and tearing edges were marked by yellow arrows. These two fracture characteristics are often
observed in low temperature deformed magnesium alloys. According to figure 6(c), the fracture feature of the specimen at 250 °C is intergranular fracture, which corresponds to the low ductility. As the temperature rises to 300 °C, a large number of dimples can be seen at the fracture, which corresponds to increased ductility.

Tensile properties of aged samples at room and high temperatures

Figure 7 depicts the stress-strain curve of Mg-8.8Gd-3.5Y-1.5Zn-0.5Zr alloy at 175 °C of peak aging. According to the curves in the figure, the hardness of the alloy reached its peak value at 162 h, with a hardness value of 122HV.

The temperature of the tensile test were 20 °C, 100 °C, 150 °C, 200 °C, 250 °C and 300 °C, respectively. As can be seen from figure 9(a), the stress-strain curve of the alloy at room temperature tensile is steeper than that at high temperature. In this case, the plasticity of the alloy is poor, and when the stress is at its maximum, the alloy breaks. It is found that the plasticity and stress increased with the increase of tensile temperature. When the tensile temperature reaches 150 °C, the stress is the highest. When the tensile temperature is higher than 150 °C, the plasticity of the alloy increases obviously and the stress decreases. The UTS, TYS and elongation of peak-aged specimen at 20 °C, 100 °C, 150 °C, 200 °C, 250 °C and 300 °C, respectively. As can be seen from figures 8(a) is shown in figure 8(b), many precipitates are distributed in the α-Mg matrix (as shown in point C in figure 8(b)). The energy spectrum analysis results of three points A, B and C in figure 8(b) are shown in figures 8(c)–(e). As can be seen from figures 8(c) and (d), the main components of precipitate labeled A are Mg and Gd, with a little Y element, and point B is square phase, with the main elements Mg, Gd and Y. By comparing figure 8(c) and (d), it can be seen that the Gd and Y element at point B are much higher than that at point A. Point C is α-Mg matrix, EDS analysis results show that there is a lot of Gd in the α-Mg matrix, but a little of Y element.

As shown in figure 9, the UTS of the alloy increases first and then decreases, which is different from the previous rule. There are two main reasons. First of all, through the analysis of EDS at point C in figure 8(b) (see figure 8(e)), it can be seen that there are many Gd elements distributed in the magnesium matrix, on the contrary, rare earth element Y is relatively small. It is well known that elements Gd and Y are highly soluble in α-Mg matrix, according to the relevant research, as Wu et al [37]. When Gd and Y diffused into magnesium, α-Mg lattice distortion would be caused to form a stress field and hinder dislocation motion. According to Bailey-Hirsch relation [38, 39], when the dislocation accumulates to a certain extent, it will form dislocation reinforcement. Secondly, the petal-like β phase precipitated after aging is the main strengthening phase at high temperature. It exists stably in the environment below 250 °C. There is a relationship between β phase and α-Mg matrix, that is, the coherent relationship, and the phase is easily cut by dislocation. The surface energy of the material increases as the dislocation is cut. After a certain degree, the strength of the material will increase [7]. Xu et al’s study on the heat-treated Mg-Gd-Y-Zn-Zr alloy shows that the precipitation order of the alloy aged at
about 200 °C are supersaturated solid solution (SSSS) → β′ (D019) → β′ (bco) [39, 40]. It is well known that the crystal structure of Mg is HCP, which is not easy to deform at RT and the sliding system is difficult to be activated. With the increase of temperature, the slip system of magnesium alloy increases gradually, but the α-Mg matrix is easy to soften. Generally, Mg alloys have much lower strength at high temperatures than at room temperature. The mechanical properties of the alloys used in this paper are better at high temperature than at RT.

**Fatigue properties of alloys**

**S-N curves of samples treated by solution and aging**

The S-N curves of Mg-8.8Gd-3.5Y-1.5Zn-0.5Zr alloys in two states are shown in figure 10(a)(b) with a spanned cycle range of 6000–22000. As shown in figure 10(a), with the increase of fatigue strength from 120 MPa to 140 MPa, the fatigue cycle times of the solution treated alloy decreases from 21118 cycles to 7722 cycles. The fatigue cycle times of peak-aged alloy also decreases gradually, from 4221000 cycles to 48422 cycles. The

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**Figure 10.** S-N curves of Mg-8.8Gd-3.5Y-1.5Zn-0.5Zr alloys in two conditions. (a)T4; (b)T6.

**Figure 11.** Fracture surface of Mg-8.8Gd-3.5Y-1.5Zn-0.5Zr alloy: (a)–(c) overall fracture surfaces of T4 alloy (120 MPa, 21118 cycles; 130 MPa, 12794 cycles; 140 MPa, 7722 cycles); (d)–(f) overall fracture surfaces of T6 alloy (120 MPa, 4221000 cycles; 130 MPa, 508730 cycles; 140 MPa, 48422 cycles).
comparison between the two figures shows that the fatigue life of peak-aged alloy is longer than that of the alloy after solution treatment at the stress range of 120MPa-140MPa.

Fractography
The fracture morphology of fatigue samples in two states under SEM is shown in figure 11. The cyclic stresses of the solid solution alloy are 120 MPa, 130 MPa and 140 MPa, and the cyclic numbers are 21118, 12794 and 7722 cycles, respectively. The cycle stresses of peak-aged alloy are also 120 MPa, 130 MPa and 140 MPa, and the cyclic numbers are 4221000, 508730 and 48422 cycles. As shown in figure 11, the overall fracture surface of solution and peak-aging samples can be divided into three zones, namely, crack initiation region (Region 1), steady crack propagation region (Region 2) and steady fatigue crack propagation region (Region 3), see the dashed lines in figure 11. By observing figure 11, it can be found that the fracture cracks mainly start from the surface of the sample. Obviously, the first area looks rough and is made up of many planes. Region 2 is coarser than Region 1 and has multiple facets and radial ridges along the direction of crack propagation. In the direction of crack propagation, the area 3 beyond the second semicircle white dotted line is a relatively flat surface.

Figure 12 is a partial enlargement of regions 1, 2 and 3 in figure 11. As mentioned in the previous study [41], during the fatigue test, impurities will appear at the fracture, such as oxide film, will lead to fatigue cracks. Looking at region 1 of the figure, it can be seen that fatigue cracks of solid solution and peak-aged alloys may induce subsurface at the oxide. Because of the large plasticity around impurities such as oxides, cracks are more likely to form here. Therefore, pores or oxide inclusions are the preferred sites for the formation of fatigue cracks. As shown in figure 12(b), the alloy fracture after solid solution treatment presents quasi-cleavage fracture characteristics in region 2. Many cleavage surfaces are seen in the fracture diagram of the aged alloy, showing typical brittle fracture characteristics. The influence of region 1 and 2 on the fatigue life is much greater than that of region 3. The solid solution and peak-aged alloys have almost no dimples in region 3, however, many secondary cracks can be seen at grain boundaries, which lead to intergranular fracture. According to the fracture SEM diagram of the fatigue sample and the fatigue properties in table 3, the fatigue life of the fatigue sample after aging is longer than that of the fatigue sample after solution. The area 2 of the fracture specimen with long fatigue life is larger. In conclusion, the fatigue life is closely related to region 2.

Table 3. Fatigue properties of alloys in different states.

| State | Strength Amplitude (MPa) | Fatigue life, Nf/Cycle |
|-------|-------------------------|----------------------|
| T4    | 120                     | 21118                |
|       | 130                     | 12794                |
|       | 140                     | 7722                 |
| T6    | 120                     | 4221000              |
|       | 130                     | 508730               |
|       | 140                     | 48422                |
Conclusion

(1) α-Mg, network-shaped eutectic compounds (Mg2Zn(Gd, Y)), lamellar phases and RE-rich phases with square shape are the main components of the as-cast alloy. After solution treatment at 530 °C for 12 h, the network-shaped compounds in the alloy are dissolved. In addition, the lamellar LPSO phase grows from grain boundary to grain interior, and some precipitate phases are irregularly distributed near grain boundary. The strength of magnesium alloy is improved during solution treatment because a large number of stacking faults can be precipitated out during solution treatment.

(2) Under high temperature, γ phase is prone to melting reaction, and grain boundary slip will cause cracks, which are the reasons for the decrease of the elongation of solid solution alloy at 250 °C.

(3) The temperature range studied in this paper is from room temperature to 300 °C, the tensile strength of the alloy increases below 150 °C, reaches the maximum value at 150 °C, and decreases above 150 °C. This is mainly related to three reinforcement mechanisms. Firstly, the dissolution of Gd and Y elements in α-Mg matrix causes solid solution strengthening. Secondly, the thermal stability of β phase can also improve the strength of the alloy. Thirdly, the relationship between β and α-Mg matrix is coherent, which leads to coherent strengthening.

(4) The fatigue performance is closely related to the crack growth zone and the fatigue property of aging alloy is better than that of solid solution alloy.

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Data availability statement

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

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