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Full length article

Understanding effects of microstructural inhomogeneity on creep response − New approaches to improve the creep resistance in magnesium alloys

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Abstract

Previous investigations indicate that the creep resistance of magnesium alloys is proportional to the stability of precipitated intermetallic phases at grain boundaries. These stable intermetallic phases were considered to be effective to suppress the deformation by grain boundary sliding, leading to the improvement of creep properties. Based on this point, adding the alloying elements to form the stable intermetallics with high melting point became a popular way to develop the new creep resistant magnesium alloys. The present investigation, however, shows that the creep properties of binary Mg−Sn alloy are still poor even though the addition of Sn possibly results in the precipitation of thermal stable Mg_2Sn at grain boundaries. That means other possible mechanisms function to affect the creep response. It is finally found that the poor creep resistance is attributed to the segregation of Sn at dendritic and grain boundaries. Based on this observation, new approaches to improve the creep resistance are suggested for magnesium alloys because most currently magnesium alloys have the commonality with the Mg−Sn alloys.

Keywords: Magnesium alloy; Creep resistance; Microstructure; High temperature deformation; Mechanical properties

1. Introduction

The high demand in the automotive industry for weight savings has resulted in the great interests for magnesium alloys due to their high specific strength and low density. Some Mg−Al based alloys, such as AZ91D and AM60B, have been used in automotive products since these alloys exhibit superior die castability and a good balance of strength and ductility [1]. However, the applications of magnesium alloys are still limited. One of the problems is the restriction of creep resistance at elevated temperatures. More investigations have revealed the poor elevated temperature creep resistances of Mg−Al based alloys is due to discontinuous precipitation of $Mg_{17}Al_{12}$ (β phase) from the supersaturated α-Mg solid solution and coarsening of β in the interdendritic eutectic region at high temperatures. The β phase has a b.c.c. structure with a melting point of 437 °C and its thermal stability is low. Its hardness decreases by 50−60% when the temperature increases from 25 to 200 °C. How to suppress the formation of β phase plays a key role in improving the creep resistance. Two major approaches were taken to develop new creep-resistant magnesium alloys in the past [2−4]:

- Adding the alloying elements to remove Al by forming Al-containing intermetallics.
- Development of aluminum free magnesium alloys.

The general principle is to form the thermal stable precipitates so that the deformation by grain boundary sliding can be suppressed by these precipitates.

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Recently, heterogeneity of creep deformation was observed by Dargusch et al. [5]. They observed higher creep deformation at grain boundaries than within dendrites. The same phenomenon was also investigated by Han and coworkers [6]. They explained that this inhomogeneous deformation was caused by the enhanced deformation in the primary α(Mg) phase and eutectic α(Mg) phase adjacent to the grain/dendritic boundaries. In these regions, both the stresses and homologous temperature are higher. On the basis of these observations, they put forward a new approach to develop creep-resistant magnesium alloys: adding alloying elements increasing the local homologous temperature or reducing the volume fraction of the eutectic α(Mg) phase. Unfortunately, in their paper no technique routes were given. The present paper will first report the effects of microstructural inhomogeneity on the creep response of Mg–Sn and AZ91 alloys. Based on the analysis of experimental results, new approaches to improve the creep resistance are proposed and discussed, and in the meantime, their practical availability is preliminary examined.

2. Experimental procedures

The selected alloys were Mg–Sn and AZ91 alloys. Mg–Sn is a promising system to develop new cheap creep resistant magnesium alloys by alloying with other elements [7]. The reason to select AZ91 alloy is that this alloy is the most extensively commercially used alloy at present. Table 1 lists the nominal composition of the alloys; all values are given in wt.%. Pure magnesium (99.98%, Hydro Magnesium), pure tin (99.96%, MCP HEK) and pure Al were used for the production of alloys. The alloys were cast using permanent mold casting. During the melting process a protective atmosphere of Ar + SF6 was employed. The melt was poured into a permanent steel mold (preheated up to 350 °C) and then cooled down to room temperature by air cooling.

Tensile creep tests were performed to investigate the creep response of Mg–3Sn and AZ91 alloys in the as-cast and heat-treated states. Tensile specimens with a diameter of 6 mm, gauge length of 30 mm and with M10 threads, were prepared from the as-cast material. Creep tests were performed using an ATS lever arm creep testing machine in air under a constant stress and temperature. The temperature was measured using Ni–CrNi thermocouple calibrated to an accuracy of ±3 °C.

The microstructure was examined using an optical microscope, scanning electron microscope (SEM), transmission electron microscope (TEM). The specimens were ground with the help of silicon carbide emery papers. Then they were polished using OPS containing 0.05 μm colloidal silica. These specimens were chemically etched in a solution of 8 g picric acid, 5 ml acetic acid and 10 ml distilled H_2O in 100 ml ethanol for 10 s. SEM investigations were performed using a JSM 5310, with an accelerating voltage of 15 kV. Back scattered images and energy dispersive X-rays were used for characterizations. Specimens for TEM were ground mechanically to about 70 μm and then thinned using electropolishing in a twin jet system using a solution of 5% $HClO_4$ and 95% ethanol at about −30 °C and a voltage of 40 V. TEM observations were performed on a JEOL 2000 transmission electron microscope with an energy dispersive X-ray analysis (EDX) system operating at 200 kV. X-ray diffraction (XRD) investigations were also carried out using a Siemens diffractometer operating at 40 kV and 40 mA with Cu K_α radiation. Measurements were obtained by step scanning 2θ from 20 to 120° with a step size of 0.02°. A count time of 3 s per step was used.

Table 1
Nominal compositions of investigated alloys (wt.%).

Alloy no.	Alloys	Sn	Ca	Al	Zn	Mg
1	Mg3Sn	3	0			Bal.
2	Mg5Sn	5	0			Bal.
3	Mg3Sn1Ca	3	1			Bal.
4	Mg3Sn2Ca	3	2			Bal.
5	AZ91			9	1	Bal.

3. Results and discussion

In this section, the relationship between the microstructure and creep response of Mg–Sn alloy is first presented and discussed, with an emphasis on the effect of interfacial microstructure on the creep response. Then the commonality of currently most magnesium alloys with Mg–Sn alloys is described. Based on these observations, the possible approaches to improve the creep resistance are put forward.

3.1. Effect of interfacial microstructure on the creep response of Mg–Sn alloys

The microstructure of Mg–Sn alloys has been described in detailed elsewhere [7,8]. Small amount of the phase Mg_2Sn with a globular shape was observed at the dendritic and grain boundaries in the as-cast Mg–3Sn alloys. The increment in the content of Sn increases the amount of the phase Mg_2Sn (Fig. 1). The subsequent ageing treatment also increases the volume fraction of Mg_2Sn phase [7]. After the addition of Ca, the phase CaMgSn forms instead of the phase Mg_2Sn.

Further microstructural investigations demonstrate that the distribution of Sn is very inhomogeneous in the binary Mg–Sn alloy. SEM observations show the existence of diffusive bright bands at the dendritic and grain boundaries (Fig. 2). These bright bands are enriched with Sn. EDS analysis indicated that the content of Sn is up to 7 wt.% (1.5 at.%) in these regions which is much higher than that in the matrix (approximately 1.2 wt.% (0.25 at.%)). Fig. 3 shows the microstructure after creep rupture for the Mg–3Sn alloy. Voids are observed at the dendritic boundaries. The cracks propagate along the dendritic boundaries, indicating that deformation by boundary sliding plays an important role in the creep of this alloy (Fig. 3(b)).

The creep life is only zero for the binary Mg–3Sn and Mg–5Sn alloys at 135 °C and a constant tensile load of 85 MPa (Table 2). The addition of Ca largely improves the creep properties of Mg–Sn alloy. The creep life increases to 358.4 h after 2 wt.% Ca was added to Mg–3Sn alloy (Table 2).

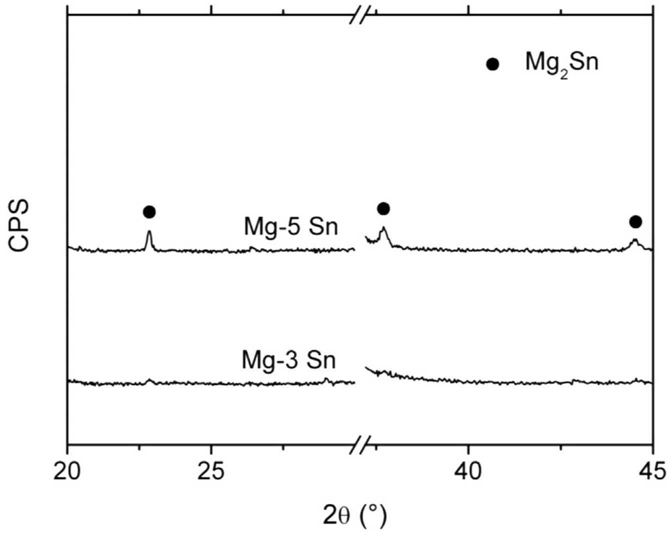

Fig. 1. XRD patterns showing the peaks of the phase Mg_2Sn in the as-cast Mg–3Sn and Mg–5Sn alloys.

Previous investigations unveiled that one of the criteria for developing creep-resistant magnesium alloys is to precipitate thermal stable particles at grain boundaries [3,4]. Aforementioned phase Mg_2Sn has a high melting point of 770.5 °C which is even higher than another thermal stable phase Mg_2Ca (715 °C) [9]. Regularly, a better creep property should be obtained for the Mg–Sn alloys with Mg_2Sn phase precipitated. However, this situation is not observed. Apparently, the thermal stability of precipitates is not the exclusive factor to affect the creep properties of magnesium alloys.

The poor creep properties of the binary Mg–Sn alloys are caused by the segregation of Sn at the dendritic and grain boundaries. Due to the segregation of Sn, the local solidus temperature is lower in these regions than inside the grains (Fig. 4). Correspondingly, in these regions, the homologous temperature is high. In addition, as pointed out by Han and Kad et al. [6,10], these regions exhibit high local stresses due to the grain misorientation and the mismatch in the coefficient of thermal expansions between the precipitates and matrix. The creep deformation preferentially starts at these sites. After

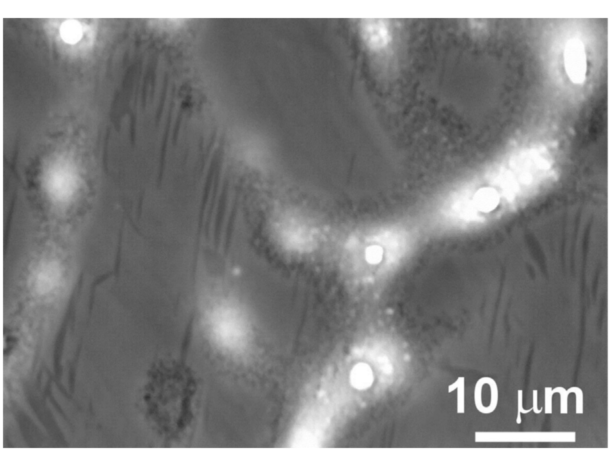

Fig. 2. Microstructure of the as-cast Mg–3Sn alloy without creep.

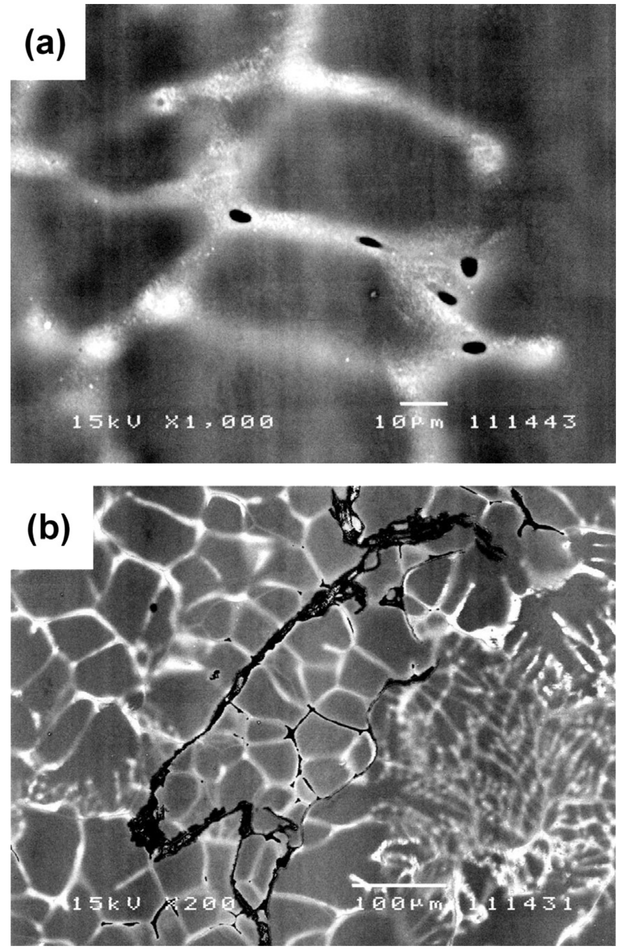

Fig. 3. Microstructures after the rupture of creep at 135 °C and 85 MPa for the as-cast Mg–3Sn binary alloy.

recognizing the effect of Sn segregation on the creep response, it is not difficult to understand why the binary Mg–Sn alloys have a poor creep resistance even though the thermal stable phase Mg_2Sn is precipitated.

3.2. Commonality of currently most magnesium alloys with Mg–Sn alloy

The Mg–Sn phase diagram shows that the eutectic reaction occurs at 561.2 °C for low tin alloys. As observed in the Mg–Sn system, most of currently magnesium alloys such as Mg–Al, Mg–RE, Mg–Gd and Mg–Zn systems also exhibit eutectic reaction (Table 3 and Fig. 5). They have a similar phase diagram as shown in Fig. 4 when the content of primary alloying elements is low. This proposes a similar microstructural situation could be observed in these alloys, like the

Table 2
Creep life of Mg–Sn alloys tensile tested at 85 MPa and 135 °C.

Alloy no.	Alloys	Creep life (hrs)	Elongation (%)	Secondary creep rate (s^{-1})
1	Mg3Sn	0	0	
2	Mg5Sn	0	0	
4	Mg3Sn2Ca	358.4	3.6	4.5×10^{-9}

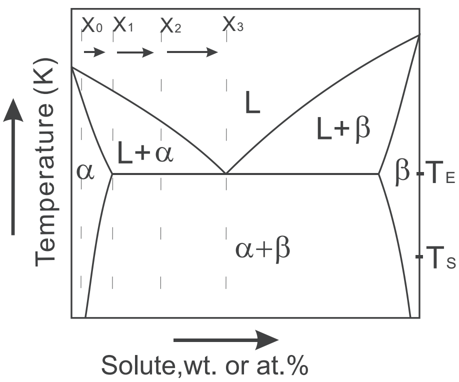

Fig. 4. Schematic of eutectic phase diagram.

segregation of primary alloying elements. Fig. 6 displays the microstructure of AZ91 alloy with and without heat treatments. In the as-cast AZ91 alloy, the regions around the dendritic and grain boundary phase, which appear brighter than the grain center, are observed. EDX results indicated that these regions contain higher amounts of Al and Zn than the center of the grain. After T5 treatments, more precipitates $Mg_{17}Al_{12}$ form around the dendritic boundaries (Fig. 6(b)). These observations confirm the segregation of aluminum near the dendritic boundaries in Mg–Al alloys. Therefore, the effect of microstructural inhomogeneity on their creep response may also be expected in these alloy systems.

It is hereby more interesting to compare the Mg–Sn and Mg–Th systems. First, these two alloy systems have a close eutectic temperature [9]. Second, the intermetallics Mg_2Sn and $Mg_{23}Th_6$ formed by the reaction of Sn and Th with Mg almost have the same melting temperature. If based on the previous point that the creep resistance is proportional to the thermal stability of intermetallics, similar creep properties should be obtained for these two alloy systems. However, their creep properties really have a large difference. The Mg–Th alloys were reported to have an excellent creep resistance even at 340 °C due to the formation of intermetallics $Mg_{23}Th_6$ [11]. This indicates other mechanisms such as interfacial microstructure etc. plays an important role in affecting the creep response. Due to the radioactivity of Th, less data was available about the microstructure and creep response of Mg–Th alloys. It is difficult to carry out a complete comparison with the Mg–Sn alloys. The possible reason may be the less maximum solubility of Th in Mg if compared with Sn (Table 4). The intermetallics $Mg_{23}Th_6$ can easily be formed and then the segregation of Th is difficult in the Mg–Th alloy.

The segregation of primary alloying element, in the meantime, can result in the local oversaturation. The strengthening by oversaturation should also be taken into account. The solid solution strengthening ($\Delta\sigma_s$) is proportional to the content of solute atom (C) and misfit strain representing the size difference between the solute and solvent atoms (ε_s), as follows [12]:

$$\Delta\sigma_s \propto \varepsilon_s^{\frac{3}{2}} C^{\frac{1}{2}} \tag{1}$$

Table 4 lists the properties of primary alloying elements and intermetallics formed by their reactions with magnesium in the currently most magnesium alloys. Al, Zn, Y, Gd and Ag are the most efficient strengtheners due to the large difference of their atomic radius with magnesium [13]. The other elements such as Th and Sn have a less strengthening effect because of their smaller solubility or less difference of their atomic radius with magnesium. The segregation of these elements contributes less solid solution strengthening in the regions close to the boundaries. As a result, the effect of particle strengthening may be particularly important for the Mg–Th and Mg–Sn systems.

The effects of segregation on the creep response may inspire to consider the eutectic temperature T_E (Fig. 4) as an important parameter when developing new creep resistant magnesium alloys, besides the considerations how to precipitate the thermal stable intermetallics [14]. If the alloy systems have a higher eutectic temperature, a higher creep resistance can be expected, as shown in Fig. 7.

3.3. Approaches to improve creep resistance

To alleviate the segregation undoubtedly improve the creep resistance of magnesium alloys. Several factors affect the

Table 3
Currently most magnesium alloy systems, with an emphasis on the creep resistant magnesium alloy.

Alloy systems		Eutectic or peritectic temperature (°C)	Examples	Alloying elements	Intermetallics
Mg–Al		437	AZ (AZ31, AZ91), AM50, AE42, AS (AS41, AS21), AX, AJ (AJ51, AJ52, AJ62), ACM (522), MRI (153)	RE, Si, Ca, Sr, Ca and RE	Al_4RE, Al_2Si, Al_2Ca, $(Mg,Al)_2Ca$, Al–Ce, Mg–Al–Ca, Mg–Al–Sr
Mg–Zn		341	ZAX (8502, 8506, 8512), ZE, ZC	Al and Ca, RE, Cu	Mg–Al–Zn–Ca
Mg–RE	Mg–Ce	592	MEZ	RE and Zn	$Mg_{12}RE$
	Mg–Y	567	WE (43, 54)	Nd	$Mg_{14}Nd_2Y$, Mg_9Nd
	Mg–Gd	548	Mg–Gd–Y–Mn	Y and Mn	$Mg_5(Gd,Y)$, $Mg_3(Gd,Y)$
Mg–Sc[a]		710	Mg–Sc–Mn, Mg–Sc–Ce–Mn	Mn, Ce and Mn	Mn_2Sc
Mg–Ag		472	QE (22)	RE (Nd)	$MgAl_{12}Nd$
Mg–Th		582	Mg–Th–Zr, Mg–Th–Zn–Zr	Zr, Zn and Zr	$Mg_{23}Th_6$
Mg–Sn		561	Mg–Sn–Ca, Mg–Sn–Al–Si (TAS831)	Ca, Al and Si	CaMgSn

[a] Peritectic reaction at 710 °C.

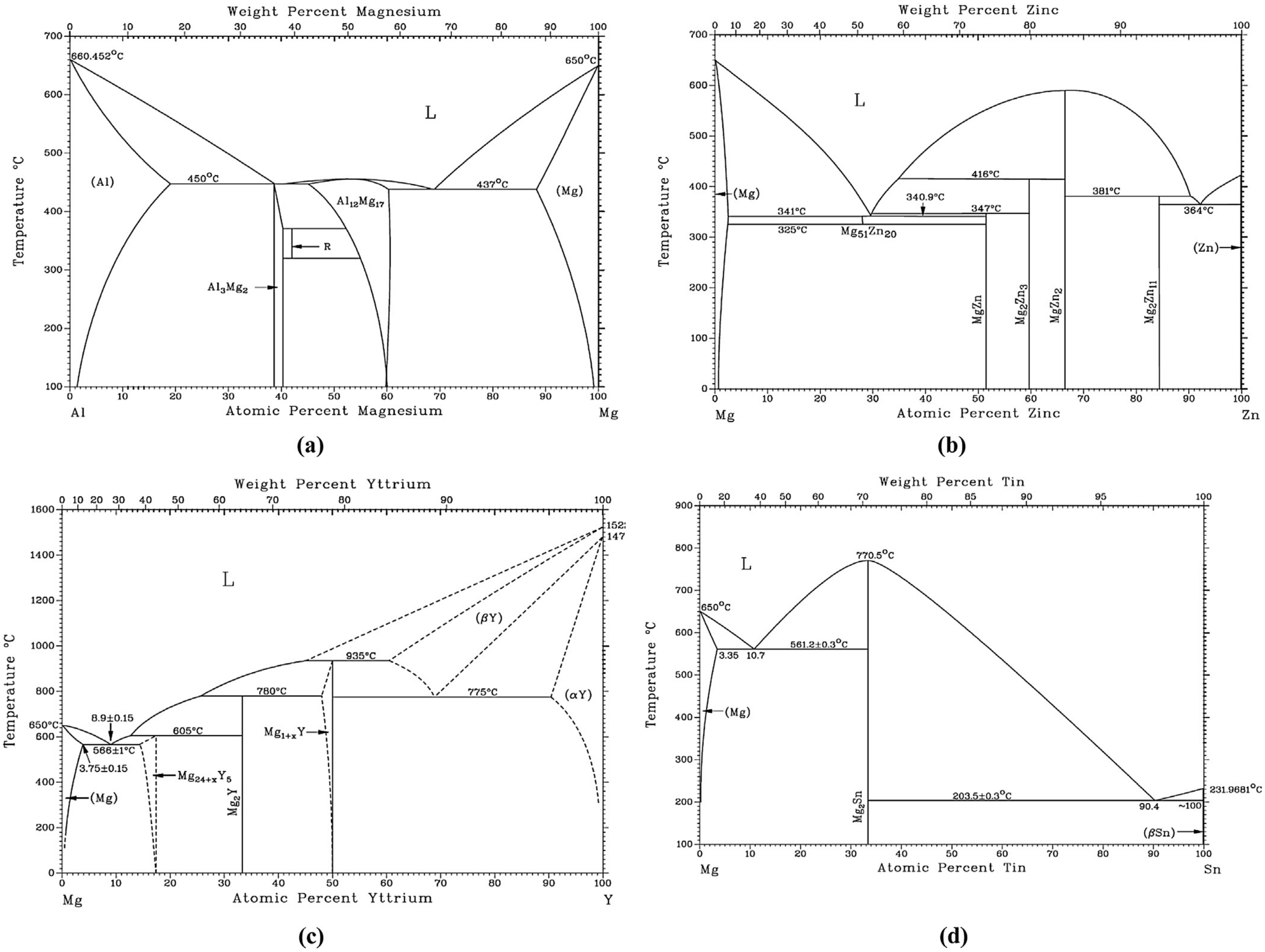

Fig. 5. Reported phase diagrams for (a) Mg–Al, (b) Mg–Zn, (c) Mg–Y and (d) Mg–Sn.

segregation of solute atoms at the dendritic and grain boundaries. Such factors include [15]:

- Solidification rate, the higher solidification rate the heavier the segregation due to the unfavorably diffusion of solute atom. For the magnesium alloys, die casting is currently popular fabrication process, which has a relatively high cooling speed. The segregations of solute atoms may be universal in these alloys (as stated in the Section 3.2). Regev et al.'s investigations show that the AZ91D alloy prepared by ingot casting has a better creep resistance than that by die casting [16]. Compared with the die casting, the ingot casting has a lower solidification rate.
- Diffusivity of solute atoms in the solid phases.
- Alloying elements, if the ternary alloying elements decreases the partition coefficient k_0, the degree of segregations increases.
- Horizontal space between the solidus and liquidus lines, the wider the space the heavier the segregation.

Following sections will discuss the practical methods to improve the creep resistance of magnesium alloys by alleviating the segregation or suppressing the sliding of grain boundary after taking these factors into account.

3.3.1. Alloying (additions of non-primary alloying elements)

The contributions of alloying to the improvement of creep properties include two aspects: solid solution strengthening and particle strengthening (obstacles to the movement of dislocations and to the sliding of grain boundaries). Alloying also modifies the microstructural morphologies and then indirectly affects the creep response. In the present section, the emphasis is put on the interactions of alloying elements with the primary alloying elements and how these interactions affect the creep properties.

The principles of selecting alloying elements followed by the previous investigators are based on how to suppress the formation of $Mg_{17}Al_{12}$ (β phase) and how to form the thermally stable phases [4,14]. The alloying elements used in the currently most magnesium alloys are listed in Table 3. As shown in Table 3, almost all these alloying elements can interact with the primary alloying elements to form intermetallics. In the Mg–Al systems, RE, Si, Ca and Sr react

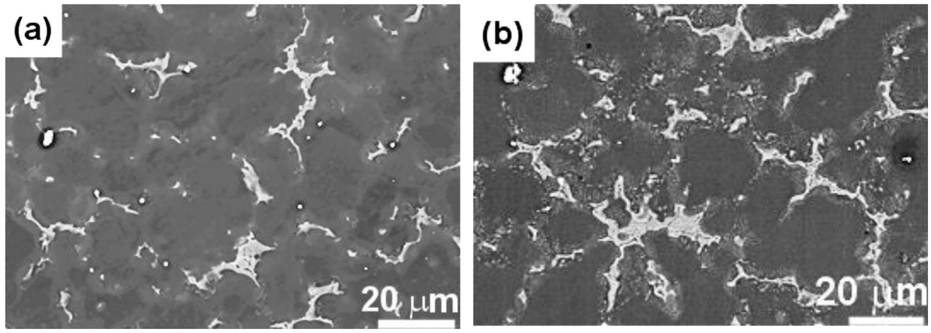

Fig. 6. SEM micrographs of AZ91 showing the segregation of aluminum at the dendritic boundaries, (a) as cast and (b) T5, 210 °C for 200 h.

with aluminum to form the thermally stable Al-containing intermetallics. Similar situations can be observed in the other alloy systems. The alloying elements selected in these systems also interact with the primary alloying element and/or magnesium itself to form the thermally stable intermetallics. For example, the addition of Ca to Mg–Sn systems results in the precipitation of thermal stable phase CaMgSn [7,8]. The detailed information about the intermetallics in the magnesium alloys can be found in the previous paper [14].

Past investigations have shown that all of these stable intermetallics are precipitated at the dendritic and grain boundaries with a size in several micrometers. They cannot be dissolved at high temperatures. The mostly observed morphology is lamellar. All these characteristics demonstrate that these intermetallics are formed during solidification. Then, a question may hereby be asked: could the formation of these intermetallics alleviate the segregation of primary alloying elements. The answer is doubtlessly yes. The evidences are given in Figs. 8 and 9. After AZ91 alloy was modified with the alloying elements Ca and Sr, the segregation of Al is largely alleviated due to the formation of Al_2Ca and Al_4Sr. No more $Mg_{17}Al_{12}$ phase is precipitated during T5 heat treatment (Fig. 8(a) and (b)). For Mg–Sn systems, after the alloying element Ca was added, the band of Sn segregation disappears due to the formation of CaMgSn phase (Fig. 9) [8].

It is well known that after the alloying elements Ca, Sr and RE were added to the AZ91 alloy its creep resistance is improved [2,4,17]. Present results also demonstrate that the creep resistance largely increases after Mg–Sn alloy is modified with the alloying element Ca (Table 2). The improvement of creep resistance by alloying is not only attributed to the precipitation of thermal stable phases, but also to the alleviation (or suppression) of segregation of primary alloying elements in these alloys.

3.3.2. Annealing treatments

The micro-segregation can be alleviated by diffusion annealing treatments in the traditional steels. The same methods are applied in the magnesium alloys. Fig. 10 shows the microstructure of Mg–3Sn alloy after T4 heat treatment. The bands of Sn segregation are not observed which have been observed in the as-cast sample (Fig. 2). During annealing at high temperatures, the solute atom Sn diffuses towards the matrix with the assistance of thermal activation. It is then expected annealing at high temperatures, such as T4 heat treatments, may be another practical way to increase the creep resistance of magnesium alloys.

Figs. 11 and 12 display the creep curves of Mg–3Sn and AZ91 alloys with different heat treatments. The as-cast Mg–3Sn alloy exhibits relatively lower creep resistance than the one after T4 and T6 heat treatments. The T4 heat treated sample had a life of 7.2 h (more than 10 times compared to that of as-cast) with a steady state creep rate of $7.8 \times 10^{-6}\ s^{-1}$. After T6 heat treatment, the sample had a life of 11.4 h with a steady state creep rate of $3.5 \times 10^{-6}\ s^{-1}$. A similar creep behavior as exhibited by Mg–3Sn alloy could be observed in

Table 4
Properties of primary alloying elements and intermetallics formed by its reaction with magnesium.

Alloying system	Primary alloying element	Atomic radius (pm)	Disregistry with Mg	Maximum solubility (at.%)	Possible intermetallics	
					Name	Melting point (°C)
Mg[a]–Al	Al	143.1	−0.11	11.6	$Mg_{17}Al_{12}$	402
Mg–Zn	Zn	134	−0.16	3.3	MgZn	347
Mg–Ce	Ce	181.8	0.14	0.13	$Mg_{12}Ce$	611
Mg–Y	Y	180	0.13	3.6	$Mg_{24}Y_5$	620
Mg–Gd	Gd	180.4	0.13	4.53	Mg_6Gd	640
Mg–Sc[b]	Sc	162	0.01	15.9	MgSc	
Mg–Ag	Ag	144	−0.10	4.0	Mg_3Ag	492
Mg–Th	Th	179	−0.12	0.49	$Mg_{23}Th_6$	772
Mg–Sn	Sn	151	−0.06	3.45	Mg_2Sn	770

[a] Atomic radius of magnesium is 160 pm.
[b] Mg–Sc, peritectic reaction at 710 °C.

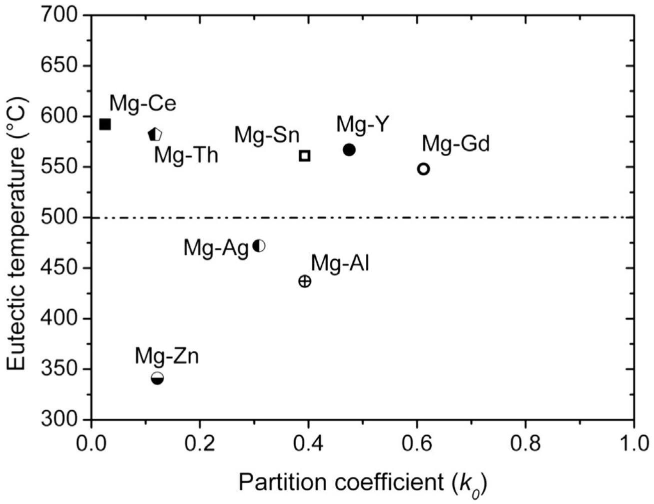

Fig. 7. Eutectic reaction temperature as a function of partition coefficient k_0 for the solutes that exhibit eutectic reactions with magnesium.

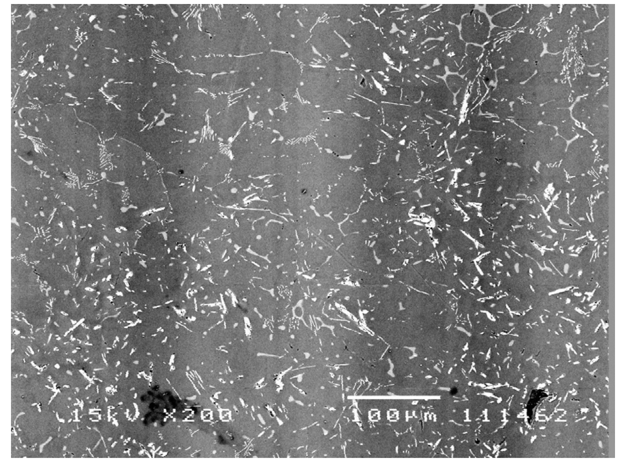

Fig. 9. Microstructure of Mg−3Sn−2Ca after creep rupture showing small amount of porosities. The diffusive bright bands disappear after 2% Ca was added to Mg−3Sn alloy.

AZ91 alloy also. The as-cast sample had a higher steady state creep rate compared to that of T4 and T6 heat treated samples. When comparing the results of T4 and T6 heat treatments, T6 heat treated samples had a slight edge over its T4 counterpart. After ageing at low temperatures, a part of the primary alloying elements, which are supersaturated in the matrix and at the boundaries, can precipitate out in the form of second phases. The segregations are further alleviated. Additionally, these hard second phases act as obstacles to the movement of dislocations and hinder the sliding of grain boundaries. Hence, the creep resistance can be seen to increase further after T6 heat treatment.

The beneficial effects of annealing treatments on the creep resistance of magnesium alloys was also reported in the previous papers [18,19]. Regev et al. reported that T4 heat treatment improves the creep resistance of AZ91 alloy [19]. Mordike's investigations indicated that T6 heat treatment improves the creep resistance of WE43 and Mg−Gd alloys (Fig. 13, [18]). Therefore, annealing at high temperatures is another effective way to alleviate the segregation of primary alloying elements and to improve the creep resistance.

3.3.3. Interfacial engineering

The segregation of primary alloying elements enhances the sliding of grain boundary during creep. Hence, suppression of deformation by boundary sliding improves the creep resistance. To suppress the sliding of grain boundary, two practical ways can be taken [20,21]. The first one is to precipitate the thermal stable particles at the grain boundaries. This method has extensively been used in improving the creep resistance of magnesium alloys [4,14]}. Present results show that the creep resistance is largely improved for Mg−Sn alloys when CaMgSn particles are formed at the dendritic and grain boundaries (Table 2 and Fig. 14). Due to the formation of this phase, the amount of voids and cracks largely decreases at the boundaries in the sample of Mg−3Sn−2Ca alloy after its creep rupture [22]. This indicates that the formation of this phase effectively suppresses the deformation by the sliding of grain boundaries.

The second one is to modify the morphology of grain boundary by alloying and/or by heat treatments so that a zigzag grain boundary is obtained [23,24]. The grain boundary precipitates, which are coherent with one of the grains on either side of the grain boundary, can normally produce the

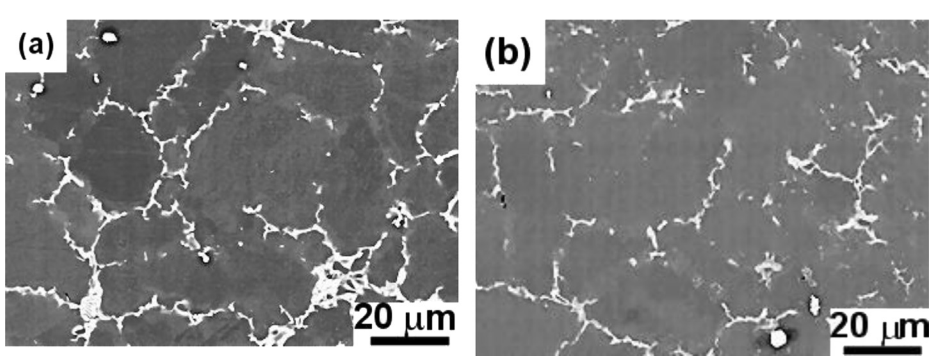

Fig. 8. SEM micrographs of modified AZ91 alloy with Ca and Sr (MRI153 alloy) showing the alleviation of segregation of aluminum at the dendritic boundaries, (a) as cast and (b) T5, 210 °C for 200 h.

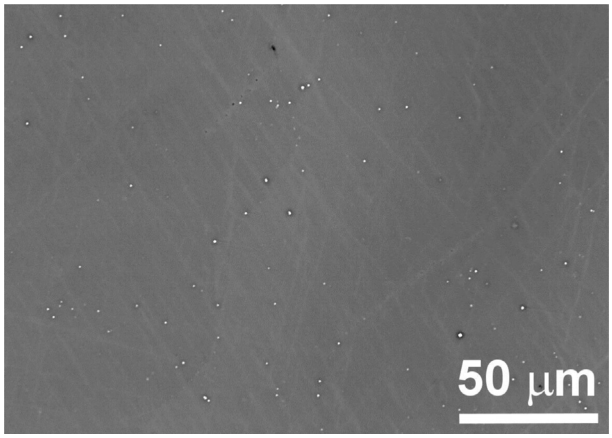

Fig. 10. SEM micrograph showing the alleviation of Sn segregation after Mg–3Sn was solution treated at 500 °C for 6 h.

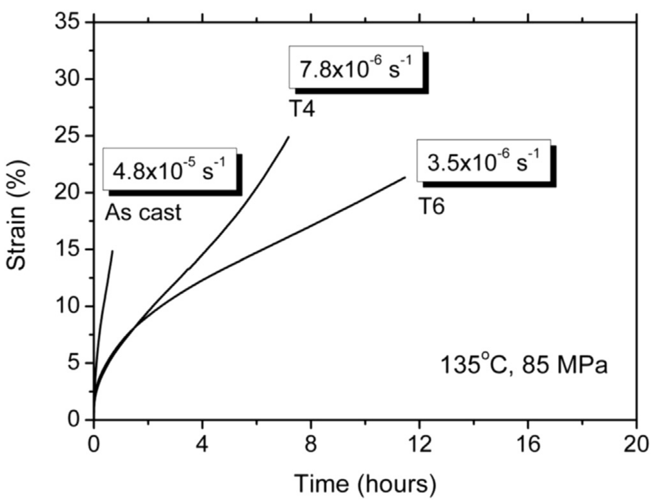

Fig. 11. Creep curves for the Mg–3Sn alloy compressive tested at 135 °C and 85 MPa.

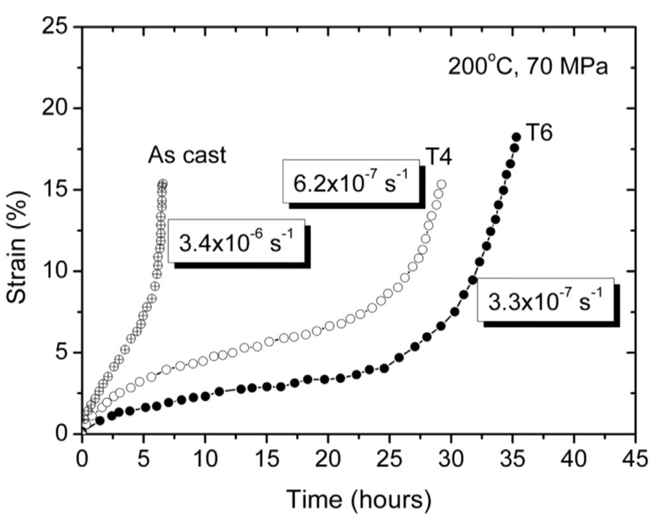

Fig. 12. Creep curves for AZ91 alloy tensile tested at 200 °C and 70 MPa.

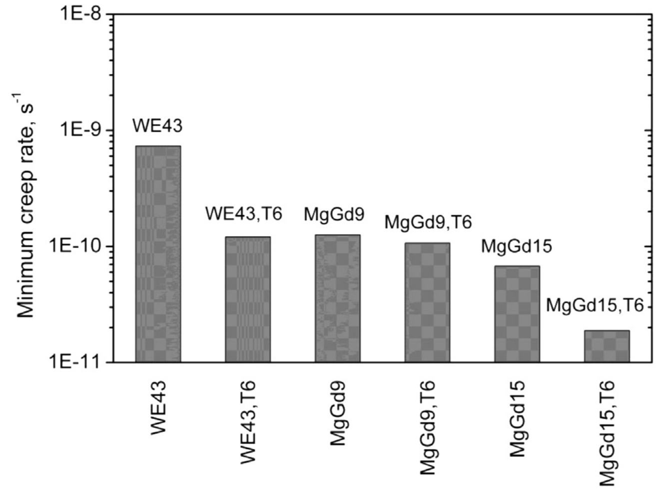

Fig. 13. Comparison of secondary creep rates of binary Mg–Gd alloys and WE43 alloy in as-cast and T6 states. The creep test was performed at 200 °C and 60 MPa [18].

grain boundary serrations during solidification [23]. In addition, the morphology of grain boundaries can be modified by the interaction of grain growth (movement of grain boundary) with the pinning effect of precipitates during heat treatments [21]. Fig. 14 shows the serration morphology of grain boundary formed during the solidifications in the as-cast Mg–3Sn–2Ca alloy (see the arrows in white color in Fig. 14). This morphology of grain boundary has been reported to be beneficial to restrain the grain boundary sliding in superalloys and steels [20]. As shown in Fig. 15, the binary Mg–3Sn alloy has a straight grain boundary. After the addition of 2% Ca to this alloy, the morphology of grain boundary is changed from regularity to irregularity. The change in the morphology of grain boundary may also have a positive effect on the creep resistance in the Mg–3Sn–2Ca alloy.

4. Summary and conclusions

Most of commonly used magnesium alloys exhibit eutectic reactions during solidification, such alloys including Mg–Al,

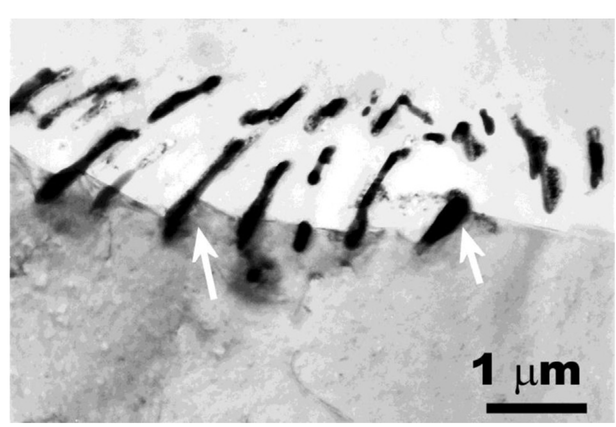

Fig. 14. TEM picture showing the morphology of grain boundary for the as-cast Mg–3Sn–2Ca alloy.

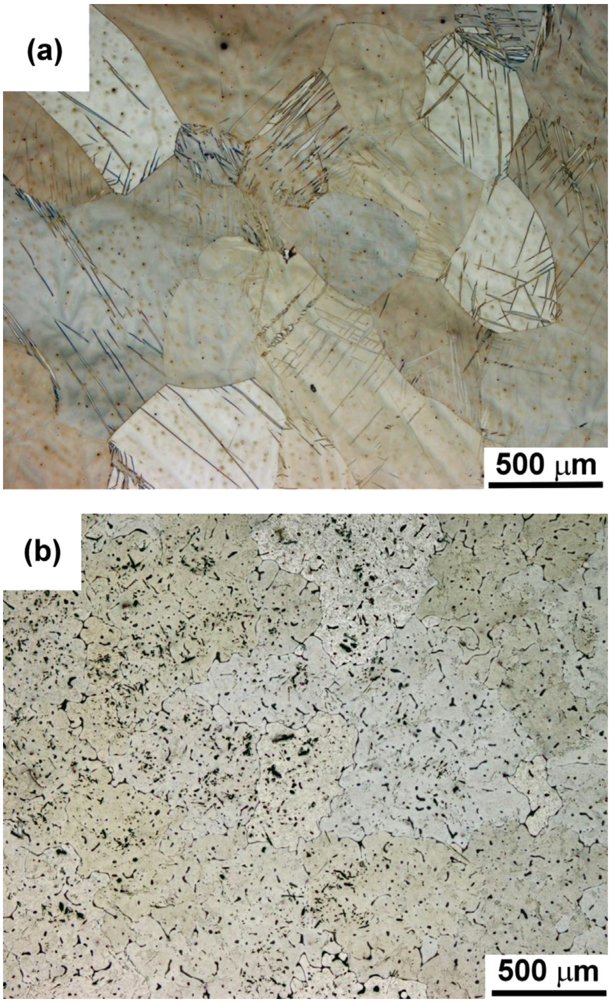

Fig. 15. Optical microstructure for the as-cast alloys, (a) Mg−3Sn and (b) Mg−3Sn−2Ca.

Mg−Zn, Mg−RE and Mg−Sn etc. Die casting is a popular process to prepare magnesium alloys at present which normally results in the severe segregation of primary alloying elements at the dendritic and grain boundaries. Investigations on the effect of primary alloying element segregation on the creep response in the Mg−Sn and Mg−Al alloys illustrate that the segregations of these primary alloying elements deteriorate the creep resistance. Due to their segregations, the local homologous temperature increases at the dendritic and grain boundaries and creep deformation preferentially starts at these sites. How to alleviate the segregation of primary alloying elements and to increase the resistance to the sliding of grain boundaries therefore plays an important role in improving the creep resistance of magnesium alloys. Three practical ways are suggested to improve the creep resistance, including alloying, annealing treatment and interfacial engineering. The first method not only alleviate the segregation by the formation of primary alloying element containing intermetallics, but also in the meantime these intermetallics act as effective obstacles to hinder the movement of dislocations and sliding of grain boundaries. The second way alleviate the segregations by diffusion of primary alloying elements, or by precipitation from the oversaturated solid solution near the dendritic and grain boundaries. The third way is to obtain the zigzag boundaries to increase the resistance to the sliding of boundaries by alloying and/or annealing treatments.

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