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A high pressure die cast magnesium alloy with superior thermal conductivity and high strength

Jian Rong^a, Jia-Ning Zhu^b, Wenlong Xiao^{a,*}, Xinqing Zhao^a, Chaoli Ma^a

^a Key Laboratory of Aerospace Advanced Materials and Performance of Ministry of Education, School of Materials Science and Engineering, Beihang University, Beijing 100191, China

^b Faculty of Mechanical, Maritime, and Materials Engineering, Delft University of Technology, The Netherlands

Abstract

Thermal conductivity is a key parameter for high performance material needed for electronic devices. While most commercially used Mg foundry alloys exhibit low thermal conductivities. In this work, we developed an Mg–3RE–0.5Zn alloy that is suitable for high pressure die cast (HPDC) ultrathin wall cellphone components. The thermal conductivity of this alloy was measured to be 133.9 W/(m·K) at room temperature, approximately 85% that of pure Mg (156 W/(m·K)). Meanwhile, it exhibited acceptable room-temperature mechanical properties with high yield strength of ~153 MPa, ultimate tensile strength of ~195 MPa, and elongation of ~4.3%. The excellent combination of superior thermal conductivity and high strength is attributed to low solute atoms in the α -Mg matrix and the formation of networked (Mg, Zn)₁₂RE eutectic phase. The results from this study will be helpful for developing new HPDC Mg alloys with more excellent performances and promoting the wider application of Mg alloys.

Keywords: Magnesium alloys; Microstructure; Thermal conductivity; Mechanical properties; High pressure die casting.

1. Introduction

Over the last decade, magnesium (Mg) alloys have been attracting increasing interest for applications in electric products, automobile, and aerospace industries because of their low density, high specific strength, and good electromagnetic shielding [1, 2]. Among them, the demand for high-thermal-conductivity Mg alloys with high mechanical performance is increasing rapidly in mobile communication devices, since high thermal conductivity ensures uniform temperature distribution

* Corresponding author: Tel: +86-10-8233 8631; Fax: 86-10-8233-8631; E-mail: wxiao@buaa.edu.cn (Wenlong Xiao).

1 which reduces thermally induced stresses and thus prolongs the service life of structural components
2 [3, 4]. For high efficiency, most Mg alloys in mobile communication devices are high pressure die
3 casting (HPDC) due to its inherent high rate of productivity and consequent relatively low cost for
4 mass production [5, 6]. However, the most commercially used HPDC Mg–Al based alloys, such as
5 [Mg–9Al–1Zn–0.3Mn \(wt.%, AZ91D\)](#), Mg–6Al–0.3Mn (wt.%, AM60) and Mg–2Al–1Si (wt.%,
6 AS21), show poor room-temperature thermal conductivity owing to the solid solution of Al. For
7 example, the thermal conductivity of [AZ91D](#) and AS21 alloy is 51.2 and 68 W/(m·K) at room
8 temperature, respectively [7, 8]. Furthermore, an alloy used as a structural component must possess
9 not only qualified heat dissipation capability but also sufficient strength, while the balance between
10 the thermal conductivity and mechanical strength is a long-term challenge [9–11]. It is necessary to
11 develop high–thermal–conductivity HPDC Mg alloys with good mechanical properties and
12 acceptable castability for the widespread applications.

13 At present, the Mg alloys having thermal conductivities higher than 100 W/(m·K) are mainly
14 based on Mg–Zn alloy system, such as Mg–3RE–3Zn (wt.%, EZ33A), Mg–5Zn–1Zr (wt.%, ZK51A)
15 and Mg–6Zn–3Cu (wt.%, ZC63) foundry alloys [12], and the Mg–2Zn–Zr and Mg–5Zn–1Mn
16 wrought alloys [13, 14]. However, these Mg–Zn based alloys are not suitable for high pressure die
17 casting. The existing strategy of designing heat-dissipating Mg alloys is to reduce solute atoms in
18 the α -Mg matrix and second phases at the expense of strength [9]. Many literatures have confirmed
19 the influence of second phases on the thermal conductivity is much weaker than solute atoms [10,
20 15, 16]. Thus, it is possible to develop high-thermal-conductivity Mg alloys with high strength and
21 acceptable castability by adding proper alloying elements. La and Ce elements seem to be potential
22 elements for the above purpose due to their low solubility in the Mg matrix, showing a weak effect
23 on the thermal conductivity of Mg alloys [17, 18]. Furthermore, they can refine grain structure and
24 mainly exist in the form of $Mg_{12}RE$ intermetallic compounds, which contributes to enhancing the
25 yield strength [10, 19, 20]. The HPDC Mg–La–Zn (HP2) with excellent creep resistance especially
26 at elevated temperatures, has been successfully developed for powertrain application [21]. However,
27 to our knowledge, the thermal properties of these HPDC Mg–RE–Zn alloys have been rarely studied.

28 In this work, a superior thermal conductivity and high strength of Mg–3RE–0.5Zn (EZ30) alloy
29 (weight percentage, also for hereafter not mentioned) was designed for ultrathin walled electric
30 components, where RE is Ce-rich mischmetal (65.24 Ce and 34.45 La) due to its low cost. Previous

1 research confirmed that the addition of Zn could improve the die castability while having a weak
2 influence on the thermal conductivity [9, 22]. The reason for the 0.5wt.% Zn addition is to improve
3 the fluidity of alloy without losing thermal conductivity obviously. The microstructure, thermal
4 properties, and mechanical properties of the HPDC EZ30 alloy were studied. The goal of this study
5 was to develop an HPDC Mg alloy with a good combination of high thermal conductivity and high
6 strength, which may be used as heat dissipation structural components of 5G smartphones.

7 **2. Alloy design and Experimental Procedures**

8 **2.1 Thermo-Calc calculation**

9 The commercial Thermo-Calc software and TTMG4 database were employed for designing
10 alloying compositions. To investigate the content of Zn on the maximum solid solubility of RE and
11 precipitation temperature of MgZn, isopleths along RE axis through Mg-RE (Ce+La)-Zn pseudo-
12 ternary phase diagram were calculated. The solidification sequence was evaluated by using the
13 Scheil model. The temperature increment was set to be 10 °C in all simulations.

14 **2.2 Fabrication and characterizations**

15 Commercial pure Mg (99.9%), pure Zn (99.9%), and Mg-30 wt.% RE master alloy were used
16 as the raw materials. the alloy was melted in an electric resistance furnace under the protection of a
17 gas mixture of CO₂ and SF₆. Before HPDC, the melt was refined, degassed, and held at ~720 °C.
18 The actual composition of the Mg-3RE-0.5Zn alloy was measured to be Mg-2.45Ce-0.59La-
19 0.48Zn-0.043Mn-0.013Si using inductively coupled plasma-atomic emission spectrometry (ICP-
20 AES). HPDC was carried out on an IMPRESS 200T cold-chamber machine using a steel mold for
21 producing Mg smartphone components with a minimum wall thickness of ~0.4mm. The parameters
22 of the HPDC process used in this study were as follow: injection pressure, 60 MPa; ram velocity,
23 2.5 m·s⁻¹; die holding time, ~5 s; mold temperature, ~260 °C. After HPDC, the real product with no
24 hot cracking was achieved and presented in Fig. 1, indicating good die castability.

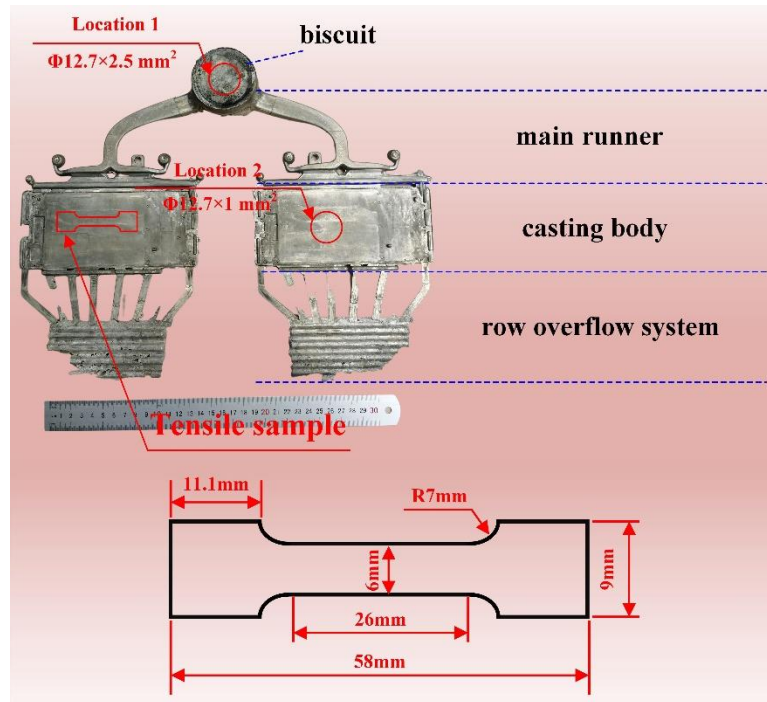


Fig. 1. The product image of the HPDC EZ30 alloy; inset: the size and location of testing samples for thermal conductivity and tension.

The size and location of testing samples for thermal conductivity are shown in Fig. 1. To avoid the influence of casting defects, the sample with rarely less defects was chosen for testing. The thermal diffusivity (α) for the biscuit (location 1) and the thin-wall casting body (location 2) was measured by the laser flash method at room temperature using a NETZSCH LFA 457 instrument. Room-temperature density (ρ) measurement was carried out by the Archimedes method. The heat capacity (C_p , $\Phi 4.5 \times 0.5\text{mm}$) was determined using a NETZSCH DSC 204F1 Phoenix calorimeter. The testing of thermal diffusivity, density, and heat capacity was conducted at least 3 times for accuracy of the results. The thermal conductivity (λ) of the studied alloy was calculated by the following equation:

$$\lambda = \alpha \cdot \rho \cdot C_p$$

The tensile sample was cut from the location marked by the red line in Fig. 1 and the size of tensile samples was also presented in Fig. 1. The room-temperature tensile test was conducted on an Instron 8801 universal testing machine equipped with an extensometer at room temperature under a strain rate of $3.0 \times 10^{-3} \text{ s}^{-1}$. To keep good repeatability, six tensile samples were tested and average mechanical properties were recorded. Phase constituents were examined by X-ray diffraction (XRD, Rigaku RINT-2000, Japan) with Cu K α radiation at the voltage of 40 kV at a scanning speed of 4

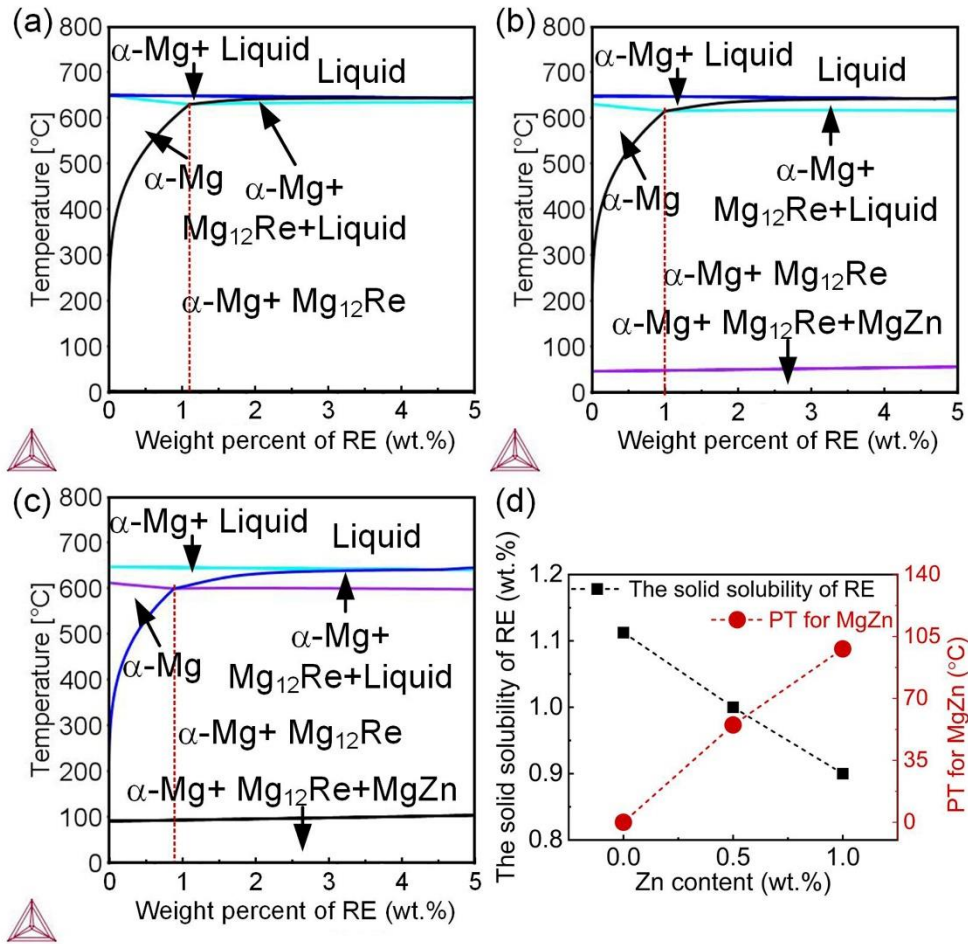
1 degrees/min. The microstructures were characterized by BX51 optical microscope (OM), JSM-
2 7001F field emission scanning electron microscopy (FESEM) and JEM-2100F transmission
3 electron microscopy (TEM) equipped with an INCA-X-Max energy dispersive spectrometer (EDS).
4 The grain size distribution and average grain size were measured by Nano Measure software based
5 on fifteen SEM-BSE images with low magnification. The solidification behavior was analyzed
6 using the differential thermal analysis (DTA, SDT600, America) with cooling and/or heating rate of
7 10 K/min under the protection of Ar gas atmosphere.

8 **3. Result and discussion**

9 Pseudo-binary Mg-RE phase diagram (i.e. 0.0 wt.% Zn addition) was firstly calculated. As
10 shown in Fig. 2 (a), the maximum solid solubility of RE in α -Mg is ~1.12 wt.%, and only α -Mg and
11 $Mg_{12}RE$ will be formed after solidification within the RE content range from 0 to 5 wt.%. For
12 avoiding the embrittlement induced by the large-size RE-containing intermetallic phase, the
13 addition amount of RE was designed as 3 wt.% in the present study. To evaluate the effect of Zn
14 content on the solidification behavior of the Mg-3RE alloy, the Pseudo-ternary Mg-RE-Zn phase
15 diagrams (0.5 and 1.0 Zn addition) were calculated and presented in Fig. 2(c) and (d). One can see
16 that with the addition of 0.5 and 1.0 Zn, solidification of the Mg-3RE alloy starts with the formation
17 of primary α -Mg, followed by the formation of $Mg_{12}RE$ phase, and ends with the formation of α -
18 $Mg+Mg_{12}RE+MgZn$ eutectic. Besides, with increasing Zn content from 0 to 1, the liquidus
19 temperature keeps almost invariable while the solidification ranges of the Mg-3RE alloy increase.
20 It has been reported that the increased solidification ranges because of increasing Zn content could
21 result in the increase of hot-tearing susceptibility during HPDC [23]. Moreover, the effect of Zn on
22 the maximum solid solubility of Mg-RE alloys and precipitation temperatures of Zn-containing
23 phases were investigated.

24 As shown in Fig.2(d), the maximum solid solubility of RE in α -Mg is decreased from 1.2 to
25 0.9 wt.% with increasing Zn from 0.0 to 1.0 wt.%, indicating that Zn can promote the precipitation
26 of the $Mg_{12}RE$ phase, which benefit to improving the thermal conductivity due to the decrease of
27 solute atoms in the α -Mg matrix. The precipitation temperature (PT) of the MgZn phase increases
28 from 55 to 98 °C with increasing Zn content from 0.5 to 1.0 wt.%. For avoiding the hot tearing and
29 formation of the MgZn phase, the 0.5 wt.% Zn was added and the casting mold was preheated to
30 ~260 °C. Based on the above thermo-calc calculation and analysis, the alloying compositions were

1 designed as Mg–3.0RE–0.5Zn.



2

3 Fig. 2. Isoleths along RE axis through Mg–RE(Ce and La)–Zn pseudo-ternary phase diagram at
 4 various Zn additions with a RE content of 0–5 wt.%.: (a) 0.0 wt.% Zn, (b) 0.5 wt.% Zn, and (c)
 5 wt.% Zn; (d) the maximum solid solubility of RE in α-Mg and precipitation temperature (PT) of
 6 MgZn as function of Zn contents.

7 Table 1 lists the room-temperature thermal properties and density of the HPDC EZ30 alloy.
 8 The thermal conductivity of the biscuit (location 1) and casting body (location 2) is calculated from
 9 the thermal diffusivity, specific heat capacity, and density. It can be seen that there is a small
 10 difference in both heat capacity and density between the biscuit and casting body, while the casting
 11 body shows lower thermal diffusivity than that of the biscuit. The thermal conductivity of the biscuit
 12 and casting body is measured to be 144.7 W(m·K)⁻¹ and 133.9 W(m·K)⁻¹, respectively, indicating a
 13 decrease in the thermal conductivity of the casting body compared to the biscuit. The thermal
 14 conductivity of the casting body determines the level of heat dissipation for cellphone components.

Table 1 Thermal properties and density of the studied alloy at room temperature

Samples	thermal diffusivity $\alpha / \text{mm}^2 \cdot \text{s}^{-1}$	special heat capacity $C_p / \text{J} (\text{g} \cdot \text{K})^{-1}$	Density $\rho / \text{g} \cdot \text{cm}^{-3}$	thermal conductivity $\kappa / \text{W}(\text{m} \cdot \text{K})^{-1}$
Biscuit	80.528	1.008	1.783	144.7
Casting body	74.167	1.010	1.785	133.9

Fig. 3(a) presents the representative engineering stress-strain curves of the casting body, and the average tensile properties are summarized in Table 2. One can see that the alloy exhibits excellent room-temperature mechanical properties with 0.2% yield strength of ~153 MPa, ultimate tensile strength of ~195 MPa, and elongation of ~4.3%. For comparison, the thermal conductivity and yield strength of some as-cast Mg alloys in published literatures are displayed in Fig. 3(b). It can be seen that the HPDC EZ30 alloy exhibits both high thermal conductivity and high YS than the Mg–Al series, Mg–Mn series, Mg–RE series, and most Mg–Zn series, indicating an excellent combination of high thermal conductivity and high strength. The reason is disclosed based on its phase constitution and microstructure characteristics.

Table 2 Tensile properties of the casting body at room temperature

Alloy	Yield Strength $\sigma_{0.2} / \text{MPa}$	Ultimate Tensile Strength σ_b / MPa	Elongation $\varepsilon / \%$
EZ30	153^{+2}_{-3}	195^{+2}_{-4}	$4.3^{+0.2}_{-0.2}$

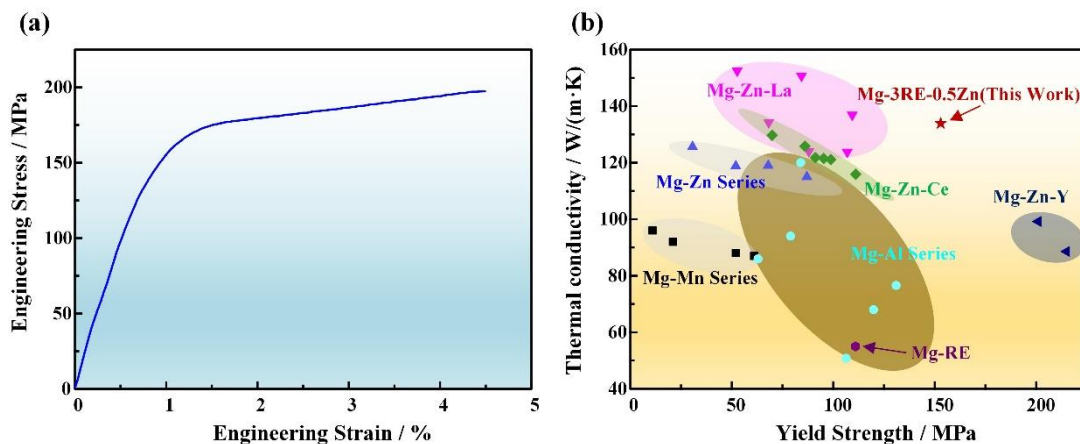
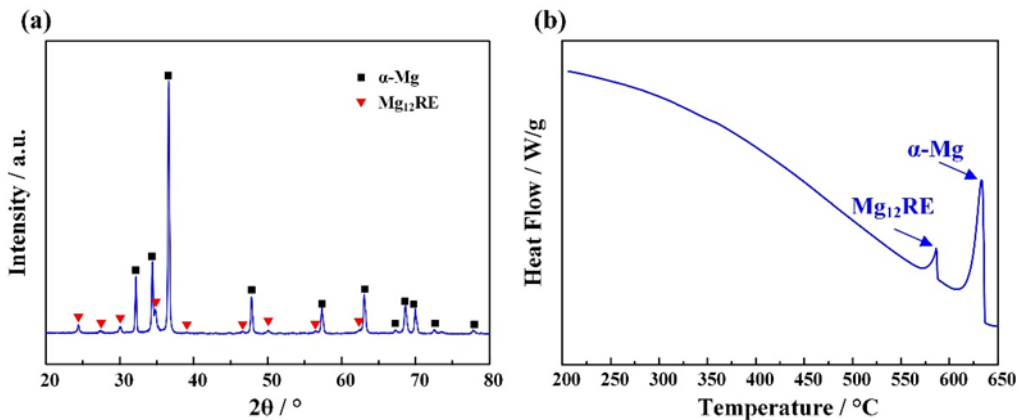


Fig. 3. (a) tensile engineering stress-strain curves of the casting body at room temperature; (b) thermal conductivity and yield strength of some Mg alloys in casting condition [9, 10, 22, 23]

XRD pattern reveals that the casting body is composed of α -Mg and Mg_{12}RE phases (Fig. 4(a)),

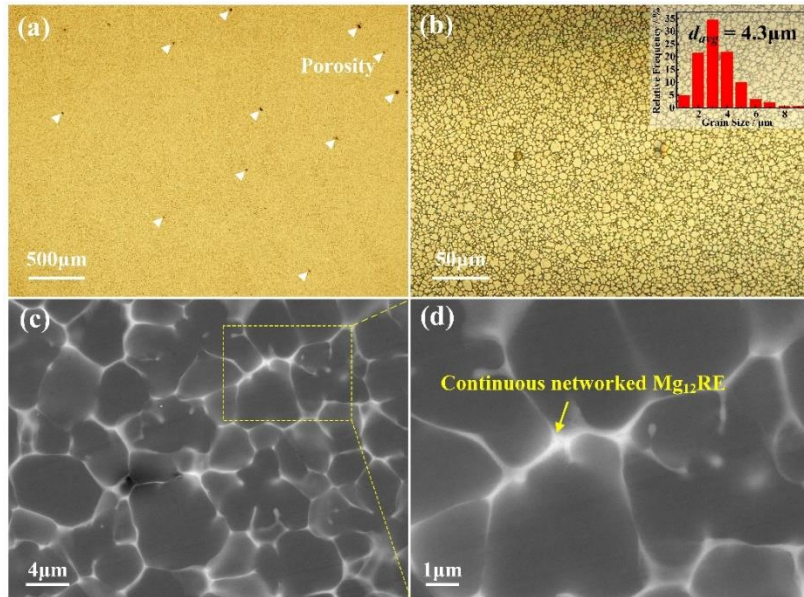
1 which is similar to previous results [24, 25]. MgZn phase cannot be detected in the alloy, which can
 2 be attributed to the combination of low content Zn addition and high mold temperature (~260°C). It
 3 has been reported that the Mg₁₂RE eutectic phase can easily form during the solidification of the
 4 Mg–Ce/La system alloys because of its relatively low barrier to nucleation and relative
 5 thermodynamic stability [26, 27]. Moreover, Zn addition into Mg–RE alloys can increase the
 6 nucleation sites and promote the formation for Mg₁₂RE phase (Fig. 2(d)) [22]. The DTA cooling
 7 curve shown in Fig. 4(b) are used to speculate the phase formation during solidification, where each
 8 exothermic peak denotes the formation of the phase labeled. The exothermic peaks corresponding
 9 to the formation of α-Mg and Mg₁₂RE are approximately 633 °C and 586 °C, respectively.
 10 According to the results of thermos-calc calculation, XRD analysis, and DTA curve, thus, the
 11 solidification behavior of the alloy is basically as following: L→α-Mg + L →α-Mg + Mg₁₂RE +
 12 L→α-Mg + Mg₁₂RE.



13
 14 Fig. 4. (a) X-ray diffraction pattern and (b) DTA cooling curve of the casting body
 15

16 Fig. 5(a) and (b) show optical images of HPDC EZ30 alloy taken from the casting body. One
 17 can see from Fig. 5(a) that a small number of porosities (marked by white arrows) with sizes ranging
 18 from 5 μm to 30 μm can be observed in the alloy, indicating good compactness. As shown in Fig.
 19 5(b), the HPDC EZ30 alloy exhibits a fine homogeneous equiaxed grain structure with an average
 20 grain size of ~4.3 μm, which is comparable with the traditional HPDC Mg–RE based alloys with an
 21 average grain size of 5–20 μm. To further investigated the morphology and distribution of the
 22 Mg₁₂RE phase, typical back-scattered SEM images are presented in Fig. 5(c) and (d). It can be seen
 23 that numerous networked eutectic phases with bright contrast are concentrated at grain boundaries,
 24 as marked by the yellow arrows. According to the calculation and experimental results in the HPDC

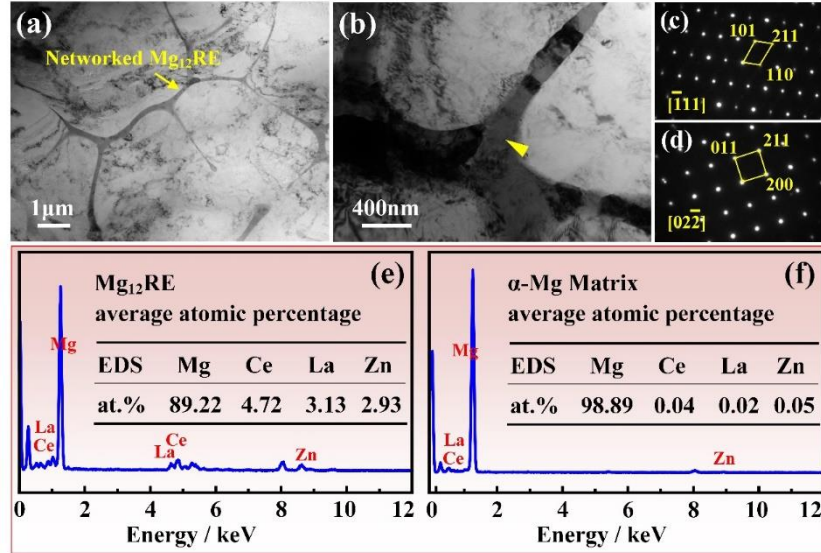
1 Mg-4Ce-0.5Mn alloy, Su et al. [28] confirmed that the primary Mg solution phase was generated
 2 first from the liquid and pushed alloying elements into the interdendritic regions, where the binary
 3 eutectic reactions occur at high temperature. Therefore, primary Mg phases were surrounded by
 4 networked $Mg_{12}Ce$ eutectic phases. For the HPDC EZ30 alloy, the growth of primary α -Mg crystals
 5 is restrained by networked eutectic phases concentrated at grain boundaries and the fast cooling rate
 6 during the solidification, which is responsible for the fine grains ($\sim 4.3 \mu m$).



7
 8 Fig. 5. (a) and (b) optical images of the casting body; (c) and (d) typical back-scattered SEM
 9 images of the casting body; inset: d_{avg} stands for the average grain size

11 Fig. 6 (a)–(d) shows the typical bright-field TEM images and corresponding selected area
 12 electron diffraction (SAED) patterns taken from the casting body of HPDC EZ30 alloy. As seen
 13 from Fig. 6(a) and (b), the networked divorced eutectic phases are consistent with SEM observations
 14 and similar with the networked $Mg_{12}RE$ eutectic phase reported in the Mg–Ce/La alloys [17, 25].
 15 The SAED patterns (shown in Fig. 6(c) and (d)) taken from the eutectic phase confirm that the
 16 networked phase is $Mg_{12}Ce$ isomorphous phase with body-centered tetragonal structure and space
 17 group of $I4/mmm$. The measured lattice parameters of this phase are $a = 1.0322 \text{ nm}$ and $c = 0.594$
 18 nm . EDS analysis (Fig. 6(e)) illustrates its chemical composition is 89.22Mg, 2.93Zn, 4.72Ce, and
 19 3.13La (at.%), where the atomic ratio of (Mg+Zn)/(Ce+La) is 11.7, close to 12, suggesting that the
 20 networked intermetallic is $(Mg, Zn)_{12}RE$ with Zn atoms partly occupying the Mg sites. A similar
 21 result has been reported in the HPDC Mg–4Zn–2La–3Y alloy, where the Zn segregates in the

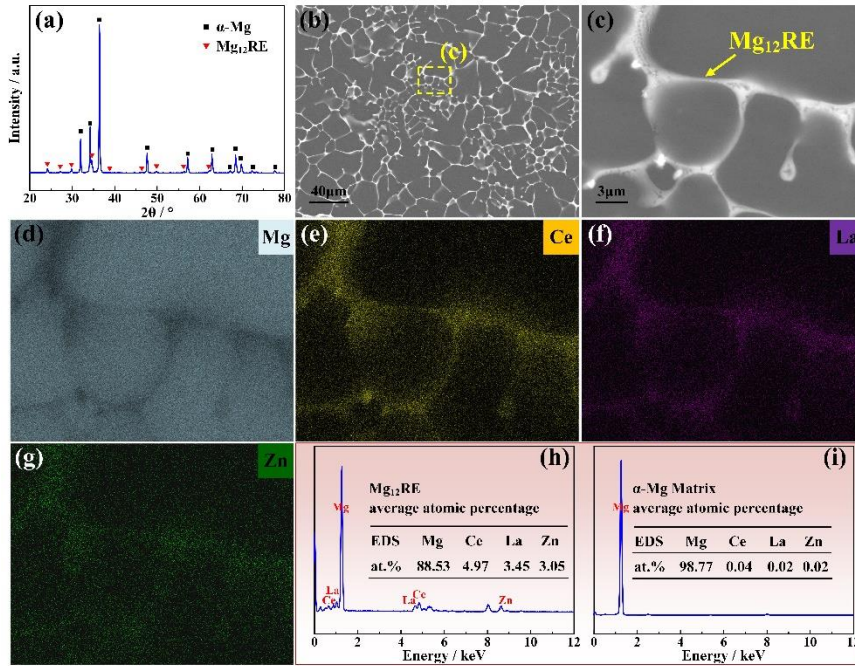
1 $Mg_{12}RE$ intermetallic phase [22]. Moreover, the EDS spectrum in Fig. 6(f) indicates that the α -Mg
 2 matrix contains rare solute atoms. Therefore, it can be concluded that the REs and Zn elements in
 3 the HPDC EZ30 alloy mainly exist in the form of the networked $(Mg, Zn)_{12}RE$ eutectic phase.



4
 5 Fig. 6. (a) and (b) bright-field TEM images of the casting body; (c) and (d) corresponding SAED
 6 patterns taken from the networked intermetallic indicated by the yellow arrow in (b); (e) and (f)
 7 EDS spectra of the networked intermetallic and α -Mg matrix, respectively.
 8

9 For comparison, the phase constituent and microstructure characterization of the biscuit are
 10 also investigated and presented in Fig. 7. As shown in Fig. 7(a), the XRD analysis result of the
 11 biscuit is consistent with that of the casting body (Fig. 4(a)), where only the $Mg_{12}RE$ phase can be
 12 detected in addition to the α -Mg matrix phase. Furthermore, there is an obvious difference in
 13 microstructure between the biscuit and casting body. Compared to the fine homogeneous grain
 14 structure of the casting body, coarser and inhomogeneous grain structure can be observed in the
 15 microstructure of the biscuit (Fig. 7(b)). And the microstructure of the biscuit also consists of
 16 continuous networked eutectic phases distributed along grain boundaries. Corresponding EDS
 17 element mappings (Fig. 7(d)-(g)) indicate that the networked eutectic phases are enriched with Ce,
 18 La, and Zn elements. EDS analysis (Fig. 7(h)) shows the average chemical composition of the
 19 networked eutectic phase and reveals that the atomic ratio of $(Mg, Zn)/(La, Ce)$ is 11.5, which is
 20 close to 12 for $Mg_{12}RE$ phases. According to the XRD analysis and EDS results, it can be confirmed
 21 that the networked eutectic phase of the biscuit is $(Mg, Zn)_{12}RE$, which is the same as that of the
 22 casting body. Moreover, the total concentrations of solute atoms in the α -Mg matrix of the biscuit
 23 are significantly low (0.08 at.%), which is similar to that of the casting body. The reason is that the

1 formation of $(\text{Mg}, \text{Zn})_{12}\text{RE}$ eutectic phase almost completely consumed the REs and Zn elements.



2

3 Fig. 7. (a) X-ray diffraction pattern and (b), (c) back-scattered SEM micrographs of the biscuit;
 4 (d)-(g) corresponding element mapping of Mg, Ce, La, and Zn, respectively; (h) and (i) EDS
 5 spectrums of the networked eutectic phase and α -Mg matrix, respectively.
 6

7 The thermal conductivity of Mg alloys is generally influenced by solution atoms and second
 8 phases, which will induce the lattice distortion and thus scatter free electron and phonon movement,
 9 resulting in the reduction in thermal conductivity [9, 15, 29]. The adverse effect on thermal
 10 conductivity caused by the solute atoms is known to be approximately several orders larger than
 11 that by the second phase [15, 16]. Li et al. investigated the effect of RE elements present in solid
 12 solution and intermetallic on the thermal conductivities of Mg-RE alloys, and revealed that the
 13 reduction in thermal conductivity was approximately 123.0 W/(m·K) with per 1 at.% RE addition
 14 in the form of solute atoms or 6.5–16.4 W/(m·K) in the form of Mg-RE intermetallic [15]. Moreover,
 15 Xie et al. studied the influence of RE-containing intermetallic on the thermal conductivity of Mg-
 16 La/Ce-Zn alloys, and the results indicated that the Mg_{12}RE intermetallic showed a slight impact on
 17 decreasing the thermal conductivities [25].

18 The microstructural observations have further confirmed that the alloying elements (REs and
 19 Zn) in both biscuit and casting body are largely consumed by the formation of the networked (Mg,
 20 Zn)₁₂RE eutectic phases (Figs. 6 and 7), resulting in significantly low concentrations of alloying

1 elements dissolved in the α -Mg matrix. Thus, the lattice distortion produced by solute atoms and
2 networked (Mg, Zn)₁₂RE eutectic phases seems to be weak, which accounts for the high thermal
3 conductivity (144.7 W/(m·K) of the biscuit and 133.9 W/(m·K) of the casting body. Compared to
4 the biscuit, the decrease in thermal conductivity of the casting body is mainly ascribed to the much
5 finer grain size, which means higher volume fractions of grain boundaries. The increased volume
6 fractions of grain boundaries in the casting body can act as the scattering sources blocking the free
7 movement of electrons and thus decreasing thermal conductivity [7]. It has been reported that the
8 contents of Al solute atoms in the thin-walled AZ91D specimen (~6.23 at.%) were much higher than
9 that in the traditional HPDC AZ91D specimens (~3.67 at.%) due to the rapid cooling rate [7, 30].
10 The actual thermal conductivity of the thin-walled AZ91D specimen was lower than that of the
11 traditional HPDC AZ91D specimens due to more serious lattice distortion induced by more solute
12 atoms. Therefore, it can be concluded that Mg-Al based alloys containing high Al contents are not
13 suitable as the thin-walled products with high thermal conductivity. For comparison, the EZ30 alloy
14 shows great potential as thin-walled components with high thermal conductivity, owing to the
15 addition of alloying elements with low solubility and the alloying elements largely consumed by the
16 formation of (Mg, Zn)₁₂RE eutectic phases.

17 It is well known that fine-grain strengthening, second phase strengthening, and solid solution
18 strengthening are the main strengthening mechanisms of Mg alloys [31, 32]. Moreover, Gavras et
19 al. reported that the YS of the HPDC Mg-RE alloys was attributed to the contribution from key
20 strengthening factors, including fine-grain strengthening that is related to the Hall-Petch
21 relationship (σ_{gb}), solid solution strengthening (σ_{ss}), and grain boundary reinforcement present at
22 grain and cell boundaries from the RE-containing intermetallic (σ_{Mg-RE}) [33, 34]. Thus, in this study,
23 the YS of the HPDC EZ30 alloy can be estimated by the following equation:

$$24 \quad \sigma_{Mg-3RE-0.5Zn} = \sigma_{gb} + \sigma_{ss} + \sigma_{Mg-RE}$$

25 According to the Hall-Petch equation [35, 36], the YS contributed from grain boundary
26 strengthening (σ_{gb}) is estimated to be ~125 MPa in this alloy with an average grain size of ~4.3 μ m.
27 Nevertheless, for Mg-Ce/La system alloy, solid solution strengthening from RE atoms is not the key
28 strengthening component due to Ce and La with low solubility in the Mg matrix. For instance, Zhu
29 et al. reported a die-cast Mg-2.5RE-0.6Zn alloy (wt.%, RE: rich in La) with a high YS of 136.3
30 (\pm 5.4) MPa, owing to the fine grain structure with an average grain size of approximately 7 μ m and

1 high volume fraction of Mg₁₂RE intermetallic phase distributed at grain boundaries [37]. In the
2 present study, the solid solution strengthening (σ_{ss}) can be determined based on the following
3 equation [22, 38]:

$$\Delta\sigma_{ss} = \alpha MG\varepsilon_{ss}^{2/3} C^{1/2}$$

4 where α represents a constant (1/550 in Mg), M represents the Taylor factor (3.06), G is the shear
5 modulus of the Mg matrix (about 16.6 GPa), C is the concentration of the solute atoms, and ε_{ss} is
6 the misfit strain. According to the ESD result (Fig. 6(f)), the strength increment (σ_{ss}) from RE and
7 Zn solutes is finally calculated to be ~6 MPa for the HPDC EZ30 alloy, revealing the weak
8 strengthening effect due to the low content of solute atoms in the α -Mg matrix. Hence, the grain
9 boundary reinforcement (σ_{Mg-RE}), contributed from networked (Mg, Zn)₁₂RE intermetallic phase
10 concentrated at grain boundaries, benefit to the YS is ~22 MPa. Chia et al. demonstrated that the YS
11 of the binary HPDC Mg–La/Ce alloys was depended on the volume fraction of Mg₁₂RE intermetallic
12 and indicated that the HPDC Mg–2.87Ce (wt.%) and Mg–3.44La (wt.%) alloys showed a high YS
13 of ~135 (± 2) MPa and ~141 (± 5) MPa, respectively [39]. The composition of the HPDC EZ30 alloy
14 newly developed in this study is relatively similar to the Mg–2.87Ce and Mg–3.44La alloys, as is
15 the volume fraction of the intermetallic phase. Compared with these two alloys, the studied alloy
16 has finer and more homogeneous grain structures and thus exhibits higher strength. Additionally,
17 Zn addition in the studied alloy can reinforce the strengthening effect by segregating in the Mg₁₂RE
18 intermetallic phase. Therefore, the high YS of the HPDC EZ30 alloy is mainly attributed to the
19 combined strengthening effects of fine grain size (~4.3 μ m) and (Mg, Zn)₁₂RE intermetallic phase
20 (14.7%).
21

22 The tensile fracture morphology of the HPDC EZ30 alloy is presented in Fig. 8. As shown in
23 Fig. 8(a), some porosities marked by the yellow arrows are observed in the fracture surface, which
24 harms the ductility. It should be noted that the formation of porosities is inevitable in the HPDC
25 alloys and can be controlled by optimizing the die-casting technique. Furthermore, it can be seen
26 from Fig. 8(b) that the fractograph of EZ30 alloy exhibits typical intergranular fracture
27 characteristics containing numerous clear ridges and smooth dissociation facets, which is consistent
28 with the low elongation (4.3%) of the alloy. The reason is that the inevitable porosities and high
29 volume fraction of coarse continuous networked Mg₁₂RE intermetallic phases can easily cause stress
30 concentration and acts as the source of micro-cracks, which results in low ductility.

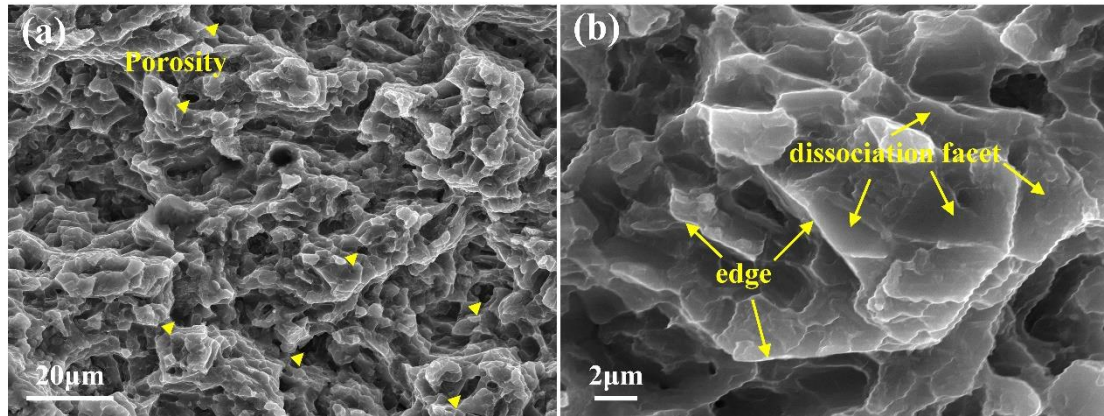


Fig. 8 SEM images of the tensile fracture for the casting body.

4. Conclusions

In summary, the HPDC Mg-3RE-0.5Zn alloy with superior thermal conductivity and high strength was developed. This alloy is suitable for the production of ultrathin wall cellphone components. The HPDC components exhibited high thermal conductivity of 133.9 W/(m·K) and good mechanical properties with YS of ~153 MPa, UTS of ~195 MPa, and elongation of ~4.3%. The high thermal conductivity was mainly attributed to the low content of solute atoms in the α -Mg matrix, and the high strength was responsible for fine grain structure and networked (Mg, Zn)₁₂RE eutectic phase.

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