Improved tensile properties of Mg-8Sn-1Zn alloy induced by minor Ti addition

* Wei-li Cheng^{1,2}, Quan-wei Tian², Rui Huo², Liang Tian², and Shou-fan Rong³

1. Key Laboratory of Interface Science and Engineering in Advanced Materials, Ministry of Education, Taiyuan University of Technology, Taiyuan 030024, China;

2. Shanxi key Laboratory of Advanced Magnesium-Based Materials, Taiyuan University of Technology, Taiyuan 030024, China;

3. School of Materials Science and Engineering, Jiamusi University, Jiamusi 154007, China;

Abstract: In this study, the influence of minor titanium (Ti) addition on the microstructure and tensile properties of Mg-8Sn-1Zn based alloys were investigated by means of optical microscopy, X-ray diffraction, scanning electron microscopy, energy dispersive spectrometry, and tensile tests. The results showed that Ti can decrease the secondary dendrite arm spacing (SDAS). The tensile strength of the Mg-8Sn-1Zn-Ti alloys is initially increased by increasing the Ti content up to 0.09wt.%, but subsequently decreased for further increase of Ti content. The improved tensile properties are attributed to the decreased SDAS and refined Mg₂Sn phases, as well as the increased fraction of tin (Sn) segregated regions. The tensile fracture surface of the studied alloys shows mixed characteristics of cleavage and quasi-cleavage fracture. Adding Ti does not significantly change the fracture mode of the studied alloys.

Key words: Mg-8Sn-1Zn alloy; tensile properties; titanium; tin segregation

CLC numbers: TG146.22

Document code: A

Article ID: 1672-6421(2016)03-151-08

Owing to their good mechanical properties at room temperature, conventional Mg-Al and Mg-Zn based alloys have attracted great interest. Unfortunately, they have problems associated with lower creep and corrosion resistance, and cannot meet the requirements of key structural components due to the low thermal stability of the main secondary phase at elevated temperatures ($>150\text{ }^{\circ}\text{C}$)^[1-2]. Therefore, it is imperative to develop novel Mg alloys with excellent performance at high temperatures.

Mg-Sn based alloys have received considerable attention as potential candidates for high performance casting and wrought magnesium alloys with enhanced thermal stability owing to the presence of high melting temperature ($T_m \approx 770\text{ }^{\circ}\text{C}$) Mg₂Sn particles in the microstructure^[3-4]. However, the coarse Mg₂Sn phase and the dendritic structure are often present in as-cast binary Mg-Sn alloys, hindering their industrial application^[5]. Consequently, the modification of the

microstructure and the improvement of the mechanical properties of Mg-Sn based alloys are necessary.

Micro-alloying can ideally help achieve this goal. The element Ti has been widely used owing to its effectiveness in grain refinement of not only Al-based and Mg-Al alloys, but also of Al-free Mg alloys^[6]. Choi et al.^[7] reported that a proper control of trace amounts of Ti below the solubility limit has the potential to improve the corrosion resistance of cast Mg-Al alloys by decreasing the grain size and the volume fraction, as well as the size of lamellar ($\alpha+\beta$) structures and the divorced eutectic β -phase, with the morphology being transformed from a continuous network structure to a semi-continuous one. Similarly, a study by Candan et al.^[8] on AZ91 alloys containing 0.2–0.5wt.% Ti indicated that an improvement in tensile strength can be achieved by introducing Ti to modify the microstructure and refine the grain size. However, until now, only Yang et al.^[9] have investigated the effects of minor Ti additions on the as-cast microstructure and mechanical properties of Mg-Sn based alloys. The results showed that the addition of 0.1–0.3 wt.% Ti to the Mg-3Sn-2Sr alloy can simultaneously improve the tensile and creep properties without clearly affecting the primary grain size.

Previous results revealed that the maximum solid solubility of Ti in solid Mg is approximately 0.2wt.%,

* Wei-li Cheng

Male, born in 1982, Ph. D., Associate Professor. Research interests: high strength magnesium alloys and biomedical magnesium alloys.

E-mail: chengweili7@126.com

Received: 2015-10-25;

Accepted: 2016-02-29

and it can be increased by Zn addition^[9, 10]. Furthermore, our previous work showed that, among the studied ternary alloys, the Mg-8Sn-1Zn alloy exhibits excellent mechanical properties^[11]. Based on the above mentioned results, the morphology, size, and distribution of the eutectic phases and SDAS of Mg-8Sn-1Zn based alloys are expected to be modified by minor Ti alloying in this study, resulting in a further improvement of the tensile properties of the Ti-containing alloys.

1 Experimental procedure

The alloys were prepared by induction melting using pure magnesium (99.9wt.%), pure tin (99.95wt.%), pure zinc (99.97wt.%), and pure Ti powder (99.93wt.%). The melt was held at 780 °C for 20 min and then poured at 720 °C into a cylindrical steel mold (diameter: 20 mm; length: 180 mm) preheated to 200 °C. Homogenization treatment was conducted at 500 °C for 5 h with protection of graphite powders and then being quenched in cold water. The chemical compositions of the tested Mg-8wt.%Sn-1wt.%Zn-xTi alloys, which were analyzed by inductively coupled plasma atomic emission spectrometry (ICP-AES) are given in Table 1.

Table 1: Chemical compositions of studied alloys (mass fraction %)

Alloy	Sn	Zn	Ti	Mg
A0	8.43	1.58	0	Bal.
A1	8.46	1.52	0.09	Bal.
A2	8.32	1.82	0.16	Bal.

The microstructures of the alloys were observed by using optical microscopy (OM) with a Leica 2700 microscope. Scanning electron microscopy (SEM; JSU-6700F) and energy-dispersive spectrometry (EDS; Oxford) were used to study the morphology and micro-chemical properties of the secondary phases. All samples were ground, polished, and etched in acetic picral (3 g of picric acid, 10 ml of acetic acid, 10 ml of H₂O, and 50 ml of alcohol). SDAS, the size of grain and the area fraction of Sn segregated regions (SSRs) were measured from several micrographs using a computer-aided linear intercept measurement and image analysis technique. X-ray diffraction

(XRD; Y-2000) with Cu K α radiation was used to analyze the phase constitution. The tensile tests were conducted with a DNS100 universal electronic testing machine at initial strain rate of 0.00046 s⁻¹ at room and high temperatures (150 °C). For the high-temperature tensile specimens, the holding time was 30 min. Ultimate tensile strength (UTS), 0.2% yield strength (YS), and elongation (El.) to failure were obtained by averaging over three samples.

2 Results and discussion

2.1 Microstructure, phase and composition of the alloys

The optical micrographs (Fig. 1) show that all as-cast alloys have a typical dendritic structure and a semi-continuous or continuous network of precipitates, most of which are located at the dendrite boundaries. Some dark regions distributed along the grain boundaries could be identified as SSRs based on our previous study^[12]. The area fraction of SSRs (Fig. 2) increased from 20.8% to 28.6% when the amount of Ti was increased from 0 to 0.09wt.%, and then decreased to 19.5% when the content of Ti was further increased to 0.16wt.%. Furthermore, the addition of Ti refined the SDAS of the as-cast alloys and the average SDAS value (Fig. 2) decreased from 44.1 μ m (A0 alloy) to 30.9 μ m (A1 alloy). When the amount of additional Ti was increased to 0.16wt.% (A2 alloy), the SDAS value slightly increased (33.1 μ m), however, it still smaller than that of A0 alloy. This is further confirmed by the grain size measurements of the homogenized alloy (Fig. 3). After minor Ti alloying, the average sizes of grains are about 90.05 μ m and 93.28 μ m for the homogenized A1 and A2 alloys, respectively (A0 is 102.97 μ m). Furthermore, most of the Mg₂Sn phases dissolved into the matrix during homogenization in the A0 and A1 alloys. Some remaining coarse Mg₂Sn particles still can be observed in the A2 alloy, indicating the restricted Sn dissolved ability into matrix due to additional Ti content above 0.1wt.%.

The microstructure refinement due to Ti alloying can be well explained considering the growth restriction factor (GRF) of a solute element, as proposed by Maxwell and Hell^[13-15].

$$\text{GRF} = \sum mc_i (k - 1)$$

where m is the slope of the liquid line, k is the partition

Fig.1: Optical micrographs of as-cast alloys: (a, b) A0, (c, d) A1, and (e, f) A2 alloys with temperature gradient of 15 K·mm⁻¹

Fig. 2: Average SDAS value and fraction of Sn segregated regions of as-cast alloys

coefficient between solid and liquid, and C_i is the concentration of the element. According to the grain growth restriction theory, a large GRF value (4×10^8) for the Ti element implies that Ti as a solute atom agglomerates in the front of the growing crystals, thus inducing a constitutional under-cooling at the solidification front and leading to the primary grain refinement [9, 16]. In addition, the agglomerates of Ti element which has a hexagonal close-packed (HCP) crystal structure may act as heterogeneous nuclei of α -Mg phases, as their lattice parameters are similar to those of HCP Mg [17]. However, as shown in Fig. 2(e and f), the SDAS of the A2 alloy increases and is bigger than that of the A1 alloy, caused by the excessive addition of Ti. As the massive Ti particles in the melt aggregate and grow, the nucleation particles decrease, greatly reducing the crystal nucleus and leading to the growth of the grains [18].

Fig. 3: Optical micrographs of homogenized alloys: (a) A0, (b) A1, and (c) A2 alloys

The XRD patterns (Fig. 4) show that all the alloys are mainly composed of the α -Mg and Mg_2Sn phases. It should be noted that no Ti-containing phase was detected in the A1 and A2 alloys. Figure 5 shows the SEM micrographs of the alloys, and the EDS analysis results in points A to F in Fig. 5 are listed in

Fig. 4: XRD patterns of as-cast alloys

Table 2. All the microstructures of the as-cast alloys comprise α -Mg grains and some intermetallic particles distributed along the grain boundaries and occasionally within the grains. Combining the XRD and EDS results, the semi-continuous network of precipitates of the A0 and A1 alloys are found, which consist of the α -Mg and Mg_2Sn phases, while the massive-like and sphere-like phases comprise divorced Mg_2Sn . For the alloy with Ti content up to 0.09wt.%, the distribution of the eutectic is homogeneous, and the partial eutectic phase presents a rod-shaped morphology with a length of approximately 10–30 μm and an aspect ratio above 15. However, in the case of the A2 alloy, the morphology of the secondary phases changed to a continuous network structure, and coarser lamellar eutectic phases were observed. What is more, based on the EDS analysis results (Table 2 and Fig. 6), Zn concentration regions could be found in the coarse second phases of the A2 alloy, owing to the formation of Zn-containing phases. The reason may be related to the massive Ti particles present in the solid-liquid interface, which hindered the solution of Zn and led to the formation of Zn-rich regions in the A2 alloy.

Fig. 5: SEM micrographs and EDS analysis of as-cast alloys: (a, d) A0, (b, e) A1, and (c, f) A2

Table 2: EDS analysis results of points A to F in Fig. 5

Positions	Contents of elements (at%)				
	Mg	Sn	Zn	Ti	Possible compounds
A	65.5	24.9	0.9	0	α -Mg+ Mg_2Sn
B	63.1	36.6	0.3	0	Mg_2Sn
C	85.9	14.1	1.3	0.01	α -Mg+ Mg_2Sn
D	59.6	25.2	0	0	Mg_2Sn
E	88.6	11.4	1.2	0.03	α -Mg+ Mg_2Sn
F	47.2	24.9	16.8	0.09	Mg_2Sn +Zn rich phase

Fig. 6: EDS mapping results for A2 alloys

2.2 Tensile properties

The tensile properties of the as-cast Mg-Sn based alloys (including YS, UTS, and El.) measured at room temperature and at 150 °C, are shown in Fig. 7(a and b). The UTS and YS increased with the Ti content up to 0.09 wt%. However, the A2

alloy exhibited a deterioration of the tensile properties, while all the Ti-containing alloys revealed relatively low El. values, compared with the base alloy. In addition, an evident reduction in strength and a clear increase in ductility could be observed at 150 °C.

Fig. 7: Tensile properties of studied alloys: (a) room temperature and (b) 150 °C

The improved tensile strength at room temperature and at 150 °C due to 0.09 wt% Ti alloying can be mainly attributed to two aspects: (1) The grain boundary strengthening resulting from the smaller SDAS, according to the Hall-Petch equation, which is supported by a number of findings^[19-21]. (2) Solid solution strengthening caused by Sn and Zn, and precipitate strengthening caused by the refined thermally stable Mg₂Sn phases, which can effectively impede mobile dislocation, pin grain boundaries and hinder crack propagation^[22-24].

The A2 alloy shows relatively poor YS, which is related to the continuous network structure of lamellar eutectic phases, as well as the massive-like Zn-rich regions. As indicated in Fig. 3, the volume fraction of remaining Mg₂Sn particles in the A2 alloy is higher than those of other alloys after homogenization, indicating that the dissolved solid Sn content [Sn_{sol}] in the A2 alloy is relatively low compared to that in the other alloys. It may be inferred that the excessive addition of Ti hindered the solution of Sn into the matrix. As a result, the solid solution strengthening effect caused by [Sn_{sol}] was weakened and

resulted in poor strength.

2.3 Fracture behavior

Figure 8 (a, b and c) shows the SEM tensile fractographs of the studied alloys at room temperature. The fracture surface consists of secondary cracks, cleavage planes, and tearing ridges. Consequently, the tensile fracture surface shows mixed characteristics of cleavage and quasi-cleavage fractures^[18, 25]. Fig. 8(d, e and f) shows the fracture morphology of the as-cast alloys fractured at 150 °C. A large number of narrower cleavage planes, and more tearing ridges and dimples could be seen on the fracture surfaces of the A0 and A1 alloys than on the fracture surfaces of the same alloys tested at room temperature. Notably, numerous cracks initiated from some coarse particles in the fractograph of the A2 alloy, both at room and high temperatures (Fig. 8(c, f)), implying that the coarse particles provided active sites for the fracture initiation. This is attributed to the fact that the A2 alloy has a considerably lower value of El. than the A0 and A1 alloys.

Fig. 8: Tensile fractograph of studied alloys at room temperature (a to c) and 150 °C (d to f): (a, d) A0, (b, e) A1 and (c, f) A2. (1: cleavage plane; 2: tearing ridges; 3: second crack; 4: dimples)

Figure 9 shows the OM images of the longitudinal section of the specimen fractured in tensile tests at room temperature and at 150 °C. The alloys tested at room and high temperatures present an inter-granular fracture mode. Based on the above results, it is inferred that adding Ti to the A0 alloy does not significantly change the fracture mode of the alloy. As

shown in Fig. 9(c, f), some coarse Mg_2Sn phases act as crack initiation sites during the deformation. Therefore, controlling the morphology and nucleation sites of the secondary phase particles is critical to improving the performance of the Mg-8Sn based alloys.

Fig. 9: Optical microstructures of longitudinal sections of studied alloys after failure in tensile tests at room temperature and 150 °C: (a, d) A0, (b, e) A1 and (c, f) A2 alloys

2.4 Tensile properties comparison of Mg-Al, Mg-Zn and Mg-Sn based alloys

Table 3 summarizes the reported tensile properties of as-cast Mg-Al^[26-28], Mg-Zn^[6,29-30] and Mg-Sn^[9,31] based alloys. As indicated, the tensile strengths of Mg-Sn based alloys are lower than those of Mg-Al and Mg-Zn based alloys at RT. With the increase of testing temperature (150 °C), there are obvious declines in strengths for all the alloys as expected. The YS of Mg-Sn based alloys is higher than that of Mg-Al based alloys and is comparable with that of Mg-Zn based alloys. What is

more, it is interesting to find that the YS decreasing ratio (DR) of the Mg-Sn alloy is ranged from 9% to 15%, while that of Mg-Al or Mg-Zn alloy is above 19%, indicating higher thermal stability of Mg-Sn based alloys due to the presence of Mg₂Sn particles at high temperatures. The comprehensive tensile properties of Mg-Sn based alloys can be improved by proper addition of Ti due to refined SDAS and size of Mg₂Sn as well as modified morphology of Mg₂Sn (semi-continuous network structure), indicating that Ti can be acted as grain refiner and heterogeneous nuclei of precipitates in Mg-Sn based alloys.

Table 3: Tensile properties of cast alloys at room temperature and 150 °C

Mg alloys	Compositions (mass%)	Tensile properties							Refs.
		UTS (MPa)		YS (MPa)		EI (%)		DR of YS (%)	
		RT	150 °C	RT	150 °C	RT	150 °C	150 °C	
Mg-Al	Mg-6Al	172	136	91	60	5.52	5.68	34%	[26]
	Mg-3Al-3Cemm	148	105	48	38	3	14	21%	[27]
	Mg-5Al-0.4Mn	168	147	62	50	7	10	19%	[28]
Mg-Zn	Mg-6Zn-2Al	190	150	101	73	8.5	14.5	28%	[29]
	Mg-6Zn-7Al-4RE	168	112	93	70	3.2	5.4	25%	[30]
	Mg-0.4Zn-1.0Zr-1.5 Ca	123	—	—	—	2.1	—	—	[6]
	Mg-0.4Zn-1.0Zr-1.5 Ca-0.8Ti	145	—	—	—	3.2	—	—	[6]
Mg-Sn	Mg-5Sn	129	105	82	72	6.3	12	12%	[31]
	Mg-3Sn-2Sr	124	111	94	82	2.9	12.5	13%	[9]
	Mg-3Sn-2Zn-0.05Ti	139	120	102	93	3.9	16.1	9%	[9]
	Mg-8Sn-1Zn-0.09Ti	129	102	73	62	9	15	15%	This paper

3 Conclusions

(1) The as-cast microstructures of all studied Mg-Sn alloys are mainly composed of the α -Mg and Mg₂Sn phases. Ti refines the SDAS and Mg₂Sn phases of the as-cast alloys. The average SDAS values decrease from 44.1 μ m to 30.9 μ m when the amount of Ti is increased from 0 to 0.09wt.%.

(2) The tensile fracture surface of the studied alloys shows mixed characteristics of cleavage and quasi-cleavage fracture. Adding Ti does not significantly change the fracture mode of the studied alloys.

(3) The tensile test results at room and high temperatures indicated that the Al alloy has the best combination of tensile properties. The improved tensile properties are mainly attributed to the refined SDAS and Mg₂Sn phases as well as the increased fraction of SSRs.

References

- [1] Suresh K, Rao K P, Prasad Y V R K, et al. Microstructure and mechanical properties of as-cast Mg-Sn-Ca alloys and effect of alloying elements. *Trans. Nonferrous Met. Soc. China*, 2013, 23: 3604–3610.
- [2] Cheng W L, Wang M, Que Z P, et al. Microstructural characterizations and mechanical properties of Mg-8Sn-1Al-1Zn-xCu alloys. *J. Wuhan Univ. Tech.*, 2014, 29: 803–807.
- [3] Fu H, Guo J X, Wu W S, et al. High pressure aging synthesis of a hexagonal Mg₂Sn strengthening precipitate in Mg-Sn alloy. *Mater. Lett.*, 2015, 157: 172–175.
- [4] Avraham S, Katsman A, and Bamberger M. Optimization of composition and heat treatment design of Mg-Sn-Zn alloys via the CALPHAD method. *J. Mater. Sci.*, 2011, 46: 6941–6951.
- [5] Liu H M, Chen Y G, Tang Y B, et al. The microstructure, tensile properties, and creep behavior of as-cast Mg-(1–10)%Sn alloys. *J. Alloys Comp.*, 2007, 440: 122–126.
- [6] Chen J A, Sun Y, Zhang J S, et al. Effects of Ti addition on the microstructure and mechanical properties of Mg-Zn-Zr-Ca alloys. *J. Magn. Alloy*, 2015(3): 121–126.
- [7] Choi J Y, and Kim W J. Significant effects of adding trace amounts of Ti on the microstructure and corrosion properties of Mg-6Al-1Zn magnesium alloy. *J. Alloys Comp.*, 2014, 614: 49–55.
- [8] Candan S, Unal M, Koc E, et al. Effects of titanium addition on mechanical and corrosion behaviours of AZ91 magnesium alloy. *J. Alloys Comp.*, 2011, 509: 1958–1963.
- [9] Yang M B, Li H L, Duan C Y, et al. Effects of minor Ti addition on as-cast microstructure and mechanical properties of Mg-3Sn-2Sr (wt.%) magnesium alloy. *J. Alloys Comp.*, 2013, 579: 92–99.

- [10] Buha J. Characterisation of precipitates in an aged Mg-Zn-Ti alloy. *J. Alloys Comp.*, 2009, 472: 171–177.
- [11] Cheng W L, Park S S, Tang W N, et al. Influence of alloying elements on microstructure and micro-hardness of Mg-Sn-Zn-based alloys. *Non. Met. Soc. China*, 2010, 20: 2246–2252.
- [12] Park S S, Kim Y J, Cheng W L, et al. Tensile properties of extruded Mg-8Sn-1Zn alloys subjected to different heat treatments. *Philos. Mag. Lett.*, 2011, 91: 35–42.
- [13] Maxwell I and Hellawell A. Simple model for grain refinement during solidification. *Acta Metall.*, 1975, 23: 229–237.
- [14] Dahle A K, Lee Y C, and Nave M D. Development of the as-cast microstructure in magnesium-aluminum alloys. *J. Light Met.*, 2001, 1: 61–72.
- [15] Wang Y X, Zeng X Q, and Ding W J. Grain Refinement of AZ31 Magnesium Alloy by Titanium and Low-Frequency Electromagnetic Casting. *Metall. Mater. Trans. A*, 2007, 38: 1358–1366.
- [16] Xu D K, Liu L, Xu Y B, et al. The influence of element Y on the mechanical properties of the as-extruded Mg-Zn-Y-Zr alloys. *J. Alloys Comp.*, 2006, 426: 155–161.
- [17] Wang Q D, Lin J B, and Ding M P. Effect of trace Ti on the microstructure and mechanical properties of AM50 alloy. *J. Mater. Res.*, 2008, 99: 761–765.
- [18] Chen T J, Wang R Q, Ma Y, et al. Grain refinement of AZ91D magnesium alloy by Al-Ti-B master alloy and its effect on mechanical properties. *Mater. Des.*, 2012, 34: 637–648.
- [19] Cai S H, Lei T, Li N F, et al. Effects of Zn on Microstructure, mechanical properties and corrosion behavior of Mg-Zn alloys. *Mater Sci Eng A*, 2012, 32: 2570–2577.
- [20] Homayun B and Afshar A. Microstructure, mechanical properties, corrosion behavior and cytotoxicity of Mg-Zn-Al-Ca alloys as biodegradable materials. *J. Alloys Comp.*, 2014, 607: 1–10.
- [21] Du H, Wei Z J, Liu X W, et al. Effects of Zn on the microstructure, mechanical property and bio-corrosion property of Mg-3Ca alloys for biomedical application. *Mater. Chem. Phys.*, 2011, 125: 568–575.
- [22] Kim B H, Park K C, Park Y H, et al. Investigations of the properties of Mg-4Al-2Sn-1Ca-xCe alloys. *Mater. Sci. Eng. A*. 2010. 527: 6372–6377.
- [23] Kim B H, Jeon J J, Park K C, et al. Microstructural characterization and mechanical properties of Mg-xSn-5Al-1Zn alloys. *Int. J. Cast. Met. Res.*, 2008, 21: 186–192.
- [24] Balasubramani N, Pillai U T S, and Pai B C. Optimization of heat treatment parameters in ZA84 magnesium alloy. *J. Alloys Comp.*, 2008, 457: 118–123.
- [25] Ortega Y, Leguey T, and Pareja R. Tensile fracture behavior of aging hardened Mg-1Ca and Mg-1Ca-1Zn alloys. *Mater. Lett.*, 2008, 62: 3893–3895.
- [26] Lv Y Z, Wang Q D, Zeng X Q, et al. Effects of rare earths on the microstructure, properties and fracture behavior of Mg-Al alloys. *Mater. Sci. Eng. A*, 2000, 278: 66–76.
- [27] Tian X, Wang L M, Wang J L, et al. The microstructure and mechanical properties of Mg-3Al-3RE alloys. *J. Alloys Comp.*, 2008, 465: 412–416.
- [28] Wang J L, Yang J, Wu Y M, et al. Microstructures and mechanical properties of as-cast Mg-5Al-0.4Mn-xNd (x = 0, 1, 2 and 4) alloys. *Mater. Sci. Eng. A*, 2008, 472: 332–337.
- [29] Chen J H, Chen Z H, Yan H G, et al. Effects of Sn addition on microstructure and mechanical properties of Mg-Zn-Al alloys. *J. Alloys Comp.*, 2008, 461: 209–215.
- [30] Xiao W L, Jia S S, Wang L D, et al. The microstructures and mechanical properties of cast Mg-Zn-Al-RE alloys. *J. Alloys Comp.*, 2008, 480: 133–136.
- [31] Poddar P and Sahoo K P. Microstructure and mechanical properties of conventional cast and rheocast Mg-Sn based alloys. *Mater. Sci. Eng. A*, 2012, 556: 891–905.

This work was supported by the National Natural Science Foundation of China (Grant no. 51404166), Shanxi Province Science Foundation for Youths (2013021013-4), Research Project Supported by Shanxi Scholarship Council of China (2014-023), Technological Innovation Programs of Higher Education Institutions Shanxi (Grant no. 2014120), and the Advanced Programs of Department of Human Resources and Social Security of Shanxi Province for Returned Scholars (2013101).