Effect of Si content on microstructure, mechanical, and thermal/electrical conductivities of Al-xSi-0.3Mn-0.3Mg-0.14Fe alloy prepared by super-slow-speed die-casting

Lu Zhang^{1, 2}, *Heng-cheng Liao¹, and Jiang Li²

1. School of Materials Science and Engineering, Southeast University, Nanjing 211189, China 2. Nanjing Chervon Auto Precision Technology Co., Ltd., Nanjing 211106, China

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Abstract: In this study, Al-xSi-0.3Mn-0.3Mg-0.14Fe alloys (x=6.5, 7.5, 8.5, wt.%) were prepared by super-slow-speed die-casting, and the effects of Si content on the microstructure, mechanical, and thermal/ electrical conductivities in as-cast, T5, and T6 states (DIN EN 1706:2020) were investigated. It is found that the increase of Si content in the alloy enhances the formation of eutectic segregation band in the casting surface microstructure. Within the Si content range of 6.5%–8.5%, as a comprehensive evaluation criterion of mechanical properties, the quality index (QI) of 376.1 MPa can be obtained in the as-cast state of the alloy with about 7.5% Si content, 373.4 MPa in T5 state of the alloy with 6.5% Si content, and 432.2 MPa in T6 state of the alloy containing 8.5% Si. The heat treatment state significantly affects the thermal conductivity and electrical conductivity of the alloys. The eutectic silicon in the alloy is segemented and further spheroidizaed during the solution process, and the solute atoms of Mg and Si are more adequately precipitated during the aging process. Both of these greatly reduce the probability of electron scattering. Thus, T6 treatment significantly improves the electrical and thermal conductivities. With the increase of Si content, both thermal conductivity and electrical conductivity decrease slightly, demonstrating a strong correlation with the Si content in the alloy.

Keywords: Al-Si alloy; microstructure; mechanical property; thermal conductivity; electric conductivity

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1 Introduction

With the rapid advancement of science and technology, there is an increasing demand for lightweight new energy vehicles to save energy and reduce emissions. In response to these demands, aluminum alloys have become more and more important in human production and life due to their excellent characteristics ^[1, 2]. Among them, Al-Si-Mg alloys have good castability, high corrosion resistance, good machinability, and high strength due to the precipitation hardening of Mg₂Si transition particles ^[3, 4]. In addition, due to the

E-mail: hengchengliao@seu.edu.cn Received: 2024-04-19; Revised: 2024-11-04; Accepted: 2024-11-25 rapid development of new energy vehicles (NEVs), the demand for high-performance die-castings of aluminum alloys is increasing. More and more Al-Si-Mg system alloys are being widely used and included in the relevant standards. In the latest European standard DIN EN 1706:2020^[5], the newly included grade EN AC-AlSi7MnMg alloy has a Si content between 6.5% and 8.5%, in which the range of Si content is wider than other AlSi7Mg series alloys (Si content between 6.5% and 7.5%). At the same time, with the rise of electric air conditioning compressors of carbon dioxide refrigerants in NEVs, casting aluminum alloys are required to have higher strength and better thermal conductivity. Therefore, in order to guide production practice, it is necessary to study the influence of Si content on the microstructure, mechanical properties, and thermal/electrical conductivities of AlSi7MnMg alloy by die-casting.

With the rapid development of NEVs, the integral

^{*}Heng-cheng Liao

Ph. D., Professor. His research fields focues on solidification processing, microstructure analysis, and mechanical properties of Al-base alloys and high-entropy alloys.

die-casting technology of front/rear frames, chassis, and body made of aluminum alloy has been widely concerned and gradually obtained practical application. By injecting molten metal into a permanent mold at high speed, die-casting can obtain castings with short cycle time and high surface quality ^[6]. Traditional die-casting technologies require eutectic or near-eutectic alloy melts to enter the mold through a relatively narrow gate at high speed and high pressure. However, the turbulence created at such high fluid velocities can trap air in the melt leading to high porosity in the casting despite exerting pressurization^[7]. For some castings with thicker walls, a slower filling speed combined with an expanded gate design can efficiently reduce air-trapping. Super-slow-speed (abbreviated as 3S) die-casting rises in response. These castings can be subjected to solution heat treatment to further control the microstructure and performances [8].

As described above, Al-Si-Mn-Mg system alloys have been die-cast for use as parts of NEVs. Si content has an important influence on microstructure and hence mechanical properties. Some researchers studied the influence of Si content on the microstructure and mechanical properties of aluminum alloys made by gravity casting. For example, Wang et al. [9] examined the microstructure and mechanical properties of as-cast and as-extruded Al-Si-Mg alloys with different Si content. The ultimate tensile strength of as-cast and as-extruded alloys was increased with the Si content. Liu et al. [10] reported the effect of Si content on solidification microstructure and mechanical properties of Al-Si binary alloy through experiments and phase-field simulation. As Si content increased from 0wt.% to 7wt.%, the microstructure of Al-Si alloy changed from coarse columnar dendrites into uniformly distributed equiaxed dendrites, and the tensile strength increased from 56.9 MPa to 111 MPa, while the ductility decreased from 30% to 2.5%. For Al-Mg-Si system alloys, the adjustment of Si content is used to balance the strength with conductivity ^[11, 12]. Research on the thermal conductivity of cast aluminum alloys primarily focuses on how to improve the morphology of Si phase [13, 14]. However, in 3S die-casting, the effect of Si content on microstructure and mechanical and thermal properties of Al-Si-Mn-Mg alloy isn't well known.

Thus, in this study, three Al-Si-Mn-Mg alloys with different Si content were prepared using self-made die-casting test molds. The effects of Si content on the microstructure, mechanical property, and thermal/electrical conductivities were studied by microstructure observation, tensile testing, and measurement of electrical/thermal conductivities.

2 Experimental procedures

2.1 Sample preparation

Based on the calculated proportion of Al-xSi-0.3Mn-0.3Mg-0.14Fe alloys (x=6.5, 7.5, 8.5, named as A1, A2, and A3, respectively), pure aluminum ingot, Al-10Mn master alloy, and pure Mg ingot were added into the commercial AlSi10MnMg aluminum alloy melt in a furnace with a capacity of about 800 kg to prepare the studied alloys. After degassing and de-slagging, the melt was die-cast in BD-350V6EX machine (Toyo Co., Ltd.) under the super-slow-speed laminar-flow mode. The pouring temperature was set at 725±5 °C, and the specimens were made by using a self-made test die (as shown in Fig. 1). The die was preheated to 250 °C. The sleeve inner diameter was 60 mm, and the punch speeds were set as $0.10\pm0.02 \text{ m}\cdot\text{s}^{-1}$ in the low-speed range and 0.20±0.02 m·s⁻¹ in the high-speed range. The chemical composition of the prepared alloys was determined by a Bruker Q4 TASMAN spectrometer, as listed in Table 1.

2.2 Heat-treatment and tensile test

In a Nabertherm hot air circulation furnace, cylinder samples of A1, A2, and A3 alloys with 8 mm in diameter (as shown in Fig. 1) experienced two types of heat treatment. One is T5 treatment (DIN EN 1706:2020, artificial aging of as-cast samples) at 180 °C for 6 h. Another is T6 treatment (DIN EN 1706:2020): solution treated at 520 °C for 4 h, then water-quenched at about 40 °C, and followed by artificial aging at 180 °C for 6 h.



Fig. 1: Self-made die-casting test die

Table	1:	Chemical co	mposition	(in wt.%) of the ex	perimental allo	vs. measured	by a Bruker	O4 TASMAN S	pectrometer
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Alloy	Si	Mn	Mg	Fe	Sr	ті	Cu	AI
A1	6.50	0.33	0.32	0.135	0.025	0.06	0.01	Bal.
A2	7.44	0.32	0.31	0.136	0.027	0.058	0.01	Bal.
A3	8.40	0.33	0.32	0.136	0.022	0.063	0.01	Bal.

2.3 DSC test and microstructural observation

The heat flow curves were measured by DSC (Mettler DSC3) with nitrogen protection during the cooling process from 700 °C to 400 °C with a cooling rate of 10 °C·min⁻¹. The cross-section of the cylinder samples, used for microstructure observation, in as-cast, T5, and T6 states was ground with P2000 sandpapers, then pre-polished with 3 μ m diamond suspension, finally polished with 50 nm silica gel suspension, and then etched with 0.5% HF water solution. The microstructure was observed by a Leica DMI 6000M optical microscope (OM). The fractography of tensile samples was observed by a Zeiss scanning electron microscope (SEM sigma300). The solidification behavior of Al-*x*Si-0.3Mn-0.3Mg-0.14Fe (*x*=6.5, 7.5, 8.5) alloys and the fractions of each phase during solidification were simulated by CALPHAD (Calculating thermodynamic phase diagram) using PandatTM software and PanAl-2023 database.

2.4 Thermal conductivity and electric conductivity tests

One sample of $10 \times 10 \times 3$ in mm was taken from the 4 mm thick plate shown in Fig. 1. The thermal diffusivity and specific heat of this sample were measured using a laser flash meter (LFA467, NETZSCH). The thermal conductivity, λ , is calculated according to Eq. (1):

$$\lambda = C_{\rm p} \cdot \alpha \cdot \rho \tag{1}$$

where C_p is the specific heat capacity (J·kg⁻¹·K⁻¹), α is the thermal diffusion coefficient (mm²·s⁻¹), and ρ is the density measured by the Archimedean method (g·cm⁻³). The electric conductivity (σ) of the samples in three states was measured

using a Fischer (SMP 350) conductivity indicator equipped with a 13 mm diameter probe at 60 kHz. The values were averaged over three readings.

3 Results

3.1 Solidification behavior

CALPHAD method is a set of calculation models based on thermodynamic principles, which can predict the phase equilibrium and phase diagram of a multicomponent alloy system and output thermodynamic properties such as phase transition temperature and phase constituents. Thermodynamic phase diagram analysis has become a common method for material design at present ^[15, 16]. Lever model (equilibrium solidification) and Scheil model (non-equilibrium solidification) can be used to analyze the solidification process. The Scheil model is used to predict the solidification paths of the studied alloys, as shown in Fig. 2(a). To simplify the simulation process, only Al, Si, Mn, Mg, Fe, and Cu elements were considered in the alloy composition. The differences among the solidification paths of A1, A2, and A3 alloys are on the liquidus temperature corresponding to primary a-Al phase and its phase fraction. Figures 2(b, c, d) show the phase fractions of all phases existing in A1, A2, and A3 alloys, respectively. The solidification microstructure is composed of α -Al, Al₁₅(Fe,Mn)₃Si₂ (π -Fe phase), Si, and Mg₂Si phases. Table 2 lists the liquidus temperature (corresponding to the primary α -Al phase), starting temperature of the eutectic reaction, solidus temperature of A1, A2, and A3 alloys, and the fraction of the primary α-Al phase. The results show that, with the increase of Si content in the alloy, the liquidus temperature and the fraction of the primary α -Al phase decrease obviously, but the eutectic reaction (L $\rightarrow \alpha$ -Al +Si eutectic) starts at almost the same temperature of about 573 °C, and the final eutectic reaction also ends at about 547 °C.



Fig. 2: Solidification paths of A1, A2, and A3 alloys (a) and phase fractions of all phases in A1 alloy (b), A2 alloy (c), and A3 alloy (d), calculated by Pandat[™]

Table 2: Parameters calculated by Pandat[™]

Alloys Parameters	A1	A2	A3
Liquidus temperature (°C)	618.0	611.8	605.3
Start temperature of L→α-Al +Si eutectic reaction (°C)	573.5	573.8	574.0
Solidus temperature (°C)	547.6	547.3	547.3
Fraction of primary α-Al phase (mol.%)	54.2	46.4	38.4

The obtained DSC heat flow curves of die-cast A1, A2, and A3 samples during the cooling process from 700 °C to 400 °C are shown in Fig. 3. For A1 alloy, the primary α -Al phase begins to form at about 617 °C, which is in good agreement with the calculation of the phase diagram software PandatTM. For A2 and A3 alloys, the α -Al phase begins to form at about 601 °C and 597 °C, respectively, which is lower than that by calculation. By phase diagram calculation, the solidification of A1, A2, and A3 alloys is predicted to end at 547.3 °C with the final eutectic reaction: L→α-Al+Al₁₅(Fe,Mn)₃Si₂+Si+Mg₂Si (shown in Fig. 2 and Table 2). However, slight differences are observed in the DSC-measured final temperatures of the three samples, which are 543 °C, 546 °C, and 549 °C, respectively. It might be due to the difference between the calculated and actual compositions of the samples, as well as the actual undercooling induced by the cooling with a cooling rate of 10 °C·min⁻¹. The temperature difference (ΔT) between liquidus and solidus of A1 alloy is 74 °C, while for A2 alloy, it is 55 °C, and for A3 alloy, it is 48 °C, indicating that ΔT is greatly decreased with the content of Si. The decrease of the solidification temperature range will affect the solidification process and hence the microstructure, especially the formation of solidification defects.



Fig. 3: DSC heat flow curves of A1, A2, and A3 alloys during the cooling process from 700 °C to 400 °C

3.2 Microstructure observation

Figures 4(a, b, c) show the optical micrographs of the core part of cylinder samples of A1, A2, and A3 alloys in the as-cast state, respectively. The color of primary α -Al dendrites is bright, while the eutectic structure formed during solidification is grey. Figures 4(d, e, f) illustrate the corresponding optical microstructure of the samples after T6 heat treatment. After T6 treatment, as shown in Fig. 5, the morphology of the Si phase changes significantly, from fibrous to particulate, and π -Fe phase dissolves into aluminum matrix partially, while α -Fe phase does not change obviously because the α -Fe phase is thermal stable at the solution temperature, which is consistent with the results in Refs. [17-19]. The eutectic Si phase and other eutectic phases (mainly Fe-rich phases) are often responsible for stress concentration and crack initiation and propagation ^[20, 21]. After the solution treatment at 520 °C for 4 h, the granulation/spheroidization of the eutectic Si phase is expected to improve mechanical properties ^[22, 23].

Figure 6 shows the microstructure at the edge part of cross-section of the cylinders for the three alloys in as-cast state. The zone between the two black dot lines in the images is



Fig. 4: Optical micrographs of A1 alloy (a, d), A2 alloy (b, e), and A3 alloy (c, f) in different states: (a, b, c) as-cast; (d, e, f) T6

Fig. 5: OM images showing the eutectic phases of A2 alloy in as-cast state (a) and T6 state (b)

Fig. 6: Optical images of edge part of cross-section of A1 (a), A2 (b), and A3 (c) cylinders in as-cast state

usually defined as the expansion shear band or defect band $^{[24, 25]}$. In the defect band, eutectic structure is dominant, with the primary α -Al dendrites situated nearby. In high-pressure die casting (HPDC), the variation of eutectic band pattern refers to the irregular distribution of eutectic phases in microstructure of the casting. Eutectic band is the region where eutectic microconstituents (such as eutectic silicon in aluminum alloys) tend to accumulate during the solidification process. Understanding and controlling the variation of eutectic band pattern is crucial in HPDC to ensure a consistent casting quality and desired mechanical properties. Techniques such as optimizing the process parameters (like cooling rates and injection speed), modifying alloy compositions and using grain refiners or nucleating agents can help manage eutectic band characteristics in die castings ^[24, 25].

In A1 cylinder, the bandwidth is narrow, only about 100 μ m, however, in A2 is about 350 μ m, and in A3 is about 450 μ m. Besides the width of defect band, it is also seen that the eutectic fraction in the defect band is also increased. It qualitatively exhibits the influence of Si content on the formation of defect band.

3.3 Mechanical properties

The tensile mechanical properties of the three alloys in different states are listed in Table 3. The calculated QI [Quality Index, QI=UTS+150*log(EL%)] values, which is usually used to evaluate the comprehensive mechanical property of metallic materials ^[26, 27], are also listed in Table 3.

In the as-cast state, with Si content increases from 6.5% to 8.5%, the value of ultimate tensile strength (UTS) is increased from 243 MPa to 258 MPa and yield strength (YS, $\sigma_{0.2}$ here) increases from 116 MPa to 124 MPa. However, the fracture elongation (EL) is decreased from 7.4% to 5.8%. In T5 state, the values of YS of A1, A2, and A3 samples are 190 MPa, 205 MPa, and 209 MPa, respectively, increased by 63.8%,

72.3%, and 68.5% compared to that in as-cast state. This reveals a significant age-hardening effect during T5 treatment. The aging precipitation has a negative impact to the elongation of the three alloys. In studied three alloys, Mg content is the same. The difference of mechanical properties is due to the change of Si content. After T6 heat treatment, the strength values of A1, A2, and A3 alloys are significantly increased compared to those in the T5 state. At the meantime, there is no significant difference in elongation among three alloys in T6 state. Notably, A3 alloy exhibits superior comprehensive mechanical properties, with a QI value up to 432.2 MPa.

SEM was used to observe the fractography of tensile specimens in as-cast, T5, and T6 states. The observation locations approximately correspond to the red box zones drawn in Fig. 6 for each alloy. In Fig. 7(a), Si particles in as-cast A1 alloy is pulled off and interspersed between the α -Al matrix, and the α -Al matrix is deformed along the tension direction. In contrast, the fracture surface of A2 and A3 as-cast alloys is relatively even [Figs. 8(a) and 9(a), respectively], because the observation locations of A2 and A3 alloy are in the expansion shear band where eutectic structure is dominant in the microstructure. For A1 alloy, it is in the zone where the primary α-Al dendrites are dominant. Figures 7(b), 8(b), and 9(b) show the fractography of A1, A2, and A3 alloys in T5 state, respectively. It is seen that the fracture mode is mainly quasi-cleavage with obvious tear ridges and the Si particles that have been pulled off are attached to the surface of the α-Al matrix. From Figs. 7(c), 8(c), and 9(c), it can be seen the fracture surfaces exhibit ductile fracture characteristics. After T6 treatment, the eutectic Si phase is broken up and spheroidized into fine particles. These particles are seen to be covered by α-Al matrix during the tensile deformation process, exhibiting many dimples. The above observation is consistent with the results in Table 3.

Alloy	UTS (MPa)	YS (MPa)	EL (%)	QI (MPa)
A1 as-cast	243±1.6	116±1.4	7.4±0.3	372.4±2.5
A2 as-cast	253±1.7	119±0.9	6.7±0.4	376.1±5.2
A3 as-cast	258±1.4	124±0.6	5.8±0.3	372.0±4.8
A1 T5	276±1.0	190±0.8	4.5±0.2	373.4±3.9
A2 T5	288±2.5	205±0.8	3.2±0.3	363.5±8.5
A3 T5	294±1.4	209±1.1	2.9±0.1	364.4±3.4
A1 T6	304±1.6	253±1.8	6.2±0.3	422.3±4.7
A2 T6	307±0.6	255±1.2	5.9±0.3	422.0±3.3
A3 T6	316±0.9	262±0.6	5.9±0.3	432.2±3.3

Table 3: Mechanical properties of A1, A2, and A3 alloys under different heat treatment states

Fig. 7: Fractography of A1 alloy in as-cast (a), T5 (b), and T6 (c) states

Fig. 8: Fractography of A2 alloy in as-cast (a), T5 (b), and T6 (c) states

Fig. 9: Fractography of A3 alloy in as-cast (a), T5 (b), and T6 (c) states

3.4 Thermal conductivity and electrical conductivities

3.4.1 Thermal conductivity

The thermal conductivity values of A1, A2, and A3 alloys in different states, measured at room temperature (25 °C), are listed in Table 4, and the relationship of the thermal conductivity (λ) with Si content in the alloy is drawn in Fig. 10. In the same state, the thermal conductivity of alloy decreases with the increase of Si content. For the same alloy, the thermal conductivity in the T6 state is the highest, then in T5 state, and finally in as-cast state. Linear regression fitting between the thermal conductivity and Si content in the alloy was performed. R-Sq=95.9% (R-Sq: Square of R) in as-cast state, R-Sq=96.7% in T5 state, and R-Sq=99.9% in T6 state, indicating a high correlation.

3.4.2 Electrical conductivity

Eddy current conductivity test is a non-destructive method to measure the electronic conductivity (σ) of non-ferrous metals. Table 5 records the conductivity values of A1, A2, and A3 alloys in different states, measured at room temperature (25 °C), while Fig. 11 illustrates the electronic conductivity against Si content in the alloy. σ value decreases with the

Table 4: Thermal conductivity values of A1, A2, and A3 alloys in different states, measured at room temperature (25 °C), W·m⁻¹·K⁻¹

Sample	A1	A2	A3
As-cast	152.90	151.93	151.47
T5	161.77	160.39	159.68
Т6	186.86	185.74	184.73

Fig. 10: Relationship of thermal conductivity with Si content in the alloy

Table 5: Electronic conductivity values of A1, A2, and A3 alloys in different states, measured at room temperature (25 °C), ms·m⁻¹

Sample	A1	A2	A3
As-cast	20.003	19.543	19.334
T5	21.744	21.468	21.073
Т6	24.628	23.898	23.225

Fig. 11: Relationship between electronic conductivity of A1, A2, A3 alloys and Si content

content of Si in the alloy in case of the same state. For the same alloy, the conductivity in T6 state is greater than that in T5 state, and the latter greater than that in as-cast state. In the same way, a linear regression was performed to analyze the relationship between the electronic conductivity and Si content in the alloy. Highly linear relation is obtained, with R-Sq values of 95.5% (as-cast), 99.0% (T5), and 99.9% (T6).

4 Discussion

According to the thermodynamic phase diagram, the primary phase in A1, A2, and A3 alloys is α -Al. With the increase of Si content, the liquidus temperature decreases and the liquidussolidus interval becomes narrow. However, there are some differences between the DSC test results and the calculation results by thermodynamic phase diagram software of Pandat[™]. According to the metallographic images of the as-cast rod section (Fig. 6), the defect bands in A2 and A3 alloys are wider. The DSC test results are affected by the composition of the sampling area (as marked in Fig. 6 by frame) where high Si content is expected. This results in a lower liquidus temperature in DSC measurement than that by phase diagram calculation. At the same time, a cooling rate of 10 °C ·min⁻¹ during the cooling process of DSC samples leads to a relatively larger undercooling of the final eutectic reaction. These factors cause the aforementioned differences.

Regarding the expansion shear band/defect band, many researchers have reported this phenomenon. Gourlay and Dahle^[24] observed the formation of expansion shear bands in Al-7Si-0.3Mg alloy made by die-casting. During the solidification process of the alloy, the solid phase and the liquid phase are deformed at the same time. In the case of low solid fraction, the solid grains in the liquid are pushed away from each other under the applied pressure, forming an expansion shear band. The macro segregation phenomenon in the expansion shear band makes the composition uneven. Laukli et al.^[28] found that Si content in the alloy had an important influence on the formation of defect bands, from clear and narrow bands at low Si content to vague and broad bands at high Si content. The defect zone contains more eutectic phases than the surroundings. In this study, similar results are observed.

For Al-Si-Mg alloys with Si content in the range of 6.5% to 8.5%, most of the previous studies have focused on the mechanical properties of AlSi7Mg (A356). Yamamoto et al. ^[29] prepared Al-7Si-0.5Mg alloy by thixotropic casting and reported that the addition of 0.5% Cu caused UTS to increase to 296 MPa and YS to 209 MPa with T5 heat treatment. In our study, A3 alloy (Al-8.5Si-0.3Mn-0.3Mg-0.14Fe) prepared by super-slowspeed die casting (in T5 state) exhibits approximately equal strength to that reported by Yamamoto. The addition of Cu element can increase the strength; however, it can also lower the corrosion resistance ^[30]. Therefore, in high-corrosion environments, it is advisable to limit the addition of Cu. The addition of rare earth elements, such as Ce, was reported to improve UTS (290.45 MPa), YS (238.27 MPa),

and elongation (2.8%) of A356 alloy after T6 treatment [31]. In our study, UTS and YS are increased with the rise of Si content in the alloy. After T6 treatment, A3 alloy has a good combination of strength and elongation (UTS, 316 MPa; YS, 262 MPa; elongation, 5.9%). In European standard DIN EN 1706:2020^[5], the EN AC-AlSi7MnMg alloy has a large and slack range of Si content, between 6.5% and 8.5%. However, from this study, it is seen that the change of Si content in this range has an obvious influence on strength and elongation. In as-cast state, as Si content in the alloy increasing from 6.5% to 8.5%, small increments in UTS (about 15 MPa) and YS (8 MPa) are obtained at a large expense in reduction in elongation (from 7.4% to 5.8%). Similar trend appears in T5 state. T5 treatment does not alter the eutectic Si structure. Thus, as expected, with an increase in Si content, the fraction of eutectic Si phase is increased that results in an increase in strength and a reduction in elongation. In T6 state, Si content increasing from 6.5% to 8.5% leads to a rise in UTS and YS (both about 10 MPa), but does not cause a great reduction in elongation as that in as-cast and T5 states. The elongation approximately keeps the same level (6.2%-5.9%), which is related to the spheroidization of eutectic Si phase during the solution treatment. Compared with the eutectic Si fibers, the stress concentation effect around Si phase is greatly weakened when eutectic Si phase is spheroidized into particles. Thus, the influence of the eutectic Si fraction on elongation is not considerable. For the same Si content, for an example A2 alloy, it is also seen that T5 and T6 treatments lead to a great increment in YS, 86 MPa and 136 MPa, respectively. During die-casting, non-equilibrium solidification leads to more Mg and Si solutes existing in the Al solution than conventional gravity casting. Larger supersaturation of them makes more metastable particles of Mg₂Si phase precipitated during T5 treatment that greatly strengthen Al matrix, thus, YS is increased and elongation is decreased. During T6 treatment, on the one hand, the solution process makes the eutectic Si phase segmented and spheroidized. On the other hand, most eutectic Mg₂Si phases are re-dissolved into the matrix that produces higher supersaturation of Mg and Si solutes, thus, more metastable particles of Mg₂Si phase are precipitated during the further aging process than the single T5 treatment. Therefore, YS of T6 treatment is more greatly increased than T5 treatment.

Although the content of Si in the DIN EN 1706:2020 standard falls within a range of 2%, die-casting manufacturers often order materials with a narrower composition range. This study provides some references to select alloy composition and heat treatment regulations for manufacturers. For an as-cast product, selecting an alloy with a Si content of approximately 7.5% can obtain 376.1 MPa of QI value. If the product is used in T5 state, a Si content of about 6.5% can yield 373.4 MPa of QI. If the product requires a higher QI value, incorporating a Si content of about 8.5% combined with T6 treatment can yield 432.2 MPa.

The total thermal conductivity (λ) of a metal, in general, includes two parts: (1) the electron thermal conductivity, λ_e ,

generated by the diffusion of free electrons; (2) the phonon thermal conductivity, λ_{ph} , generated by lattice vibration. The calculation formula is shown in Eq. (2) ^[32]:

$$\lambda = \lambda_{\rm e} + \lambda_{\rm ph} \tag{2}$$

The thermal conductivity from phonons is generally very small and can often be ignored. The thermal conductivity from electrons is closely related to the electronic conductivity of a metal, and the relationship between them satisfies the Wiedemann-Franz law^[33]:

$$\lambda_{\rm e} = L\sigma T$$
 (3)

where *T* is the temperature, *L* is the Lorentz number, and σ is the electronic conductivity. It is seen that the thermal conductivity of a metal is usually proportional to the electronic conductivity of the electrons. Thus, the thermal conductivity and electronic conductivity are always discussed together.

When the electron mobility reaches its maximum in a perfect atomic lattice, any strain, defect, or body defect in a metal can act as an electron scattering center that reduces its mean free range, thus the electronic/thermal conductivities is decreased ^[34]. Due to high pressure die-casting process, the solidification rate is high. A large number of alloying elements are obliged to retain in the Al matrix, leading to a larger lattice distortion subsequently a lower thermal/electronic conductivities in as-cast state. Heat treatment is one of the most effective control factors to increase the thermal conductivity of a casting. In this study, the thermal conductivity in T5 state is increased by about 10 W \cdot m⁻¹ \cdot K⁻¹ compared with the as-cast state. By T5 treatment, solute atoms are continuously precipitated from the solid solution (Al matrix), in the form of phase transformation of β -Mg₂Si in the Al-Si-Mg system, which greatly reduces the electron scattering in Al matrix and thus improves the electronic/thermal conductivities ^[35, 36]. It is reported that the interface between the matrix and secondary phase can scatter free electrons and shorten the mean free range of electrons ^[37]. Compared with T5 state, the thermal conductivity of sample in T6 state is improved by about 25 W·m⁻¹·K⁻¹. One of the reasons is that the eutectic silicon phase has experienced segmenting and spheroidizing during the solution process. On the one hand, it largely reduces the fraction of phase interface. On the other hand, Al matrix in the eutectic zones become more smoothly connected than that in as-cast state. These two factors greatly decrease the thermal transmission barrier from eutectic silicon. Another reason, perhaps more important, is from the more adequately precipitation of Mg and Si solution atoms during the aging process [38], because the solution treatment results in a higher supersaturation of solutes. These two changes in microstructure significantly reduce the chance of electron scattering, thus, T6 treatment greatly improves the electronic/thermal conductivity of the studied alloys. In Ref. [39], the thermal conductivity of as-cast Al-7Si binary alloy is 152 W·m⁻¹·K⁻¹. Our results in Table 4 are consistent with it. After solution heat treatment for 48 h, the obtained thermal conductivity of Al-7Si binary alloy was 171 W·m⁻¹·K⁻¹ in Ref. [39]. While in this study, after T6 treatment, all the thermal conductivity values of the prepared three alloys by super-low-rate die-casting are more than 180 W \cdot m⁻¹ \cdot K⁻¹.

5 Conclusions

(1) With the increase of Si content in the alloy from 6.5% to 8.5%, the formed expansion shear band becomes much wider in casting product.

(2) Based on the quality index QI, in as-cast state, a QI value of 376.1 MPa can be obtained by choosing about 7.5% of Si content; in T5 state, 373.4 MPa can be obtained by choosing about 6.5% of Si; in T6 state, 432.2 MPa can be obtained by choosing about 8.5% of Si.

(3) Heat treatment can significantly affect the thermal/ electronic conductivities of the alloy. After T6 treatment, the phase interface fraction between eutectic silicon and Al matrix is greatly reduced because the eutectic Si fibers experience segmenting and spheroidizing during the solution process, and solute atoms are more adequately precipitated during the aging process. These two factors significantly reduce electron scattering. Thus, T6 treatment greatly improves the electronic/thermal conductivities.

(4) Both the thermal and electronic conductivities are highly correlated with Si content. With the increase of Si content in the alloy, both thermal conductivity and electrical conductivity decrease slightly.

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Conflict of interest

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

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