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INHOMOGENEITY IN MICROSTRUCTURES AND TENSILE PROPERTIES IN HIGH PRESSURE DIE-CASTINGS OF Al-Mg-Si ALLOYS

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ABSTRACT

High pressure die casting is a promising process to make thin-wall components with high dimensional accuracy and excellent surface quality. However, the non-uniformity is one of the important features in castings. The inhomogeneity in tensile properties and microstructural characteristics were studied for the castings with different thicknesses made by high pressure die-cast Al-Mg-Si alloys. For the local mechanical properties in the die-castings, the casting skin provided the highest, whereas the casting centre provided the lowest yield strength, UTS and elongation. The yield strength, UTS and elongation were significantly decreased towards the casting centre. The microstructural inhomogeneity is characterised by the segregation of coarse fragmented dendrites, the reduction of solute concentrations of Mg and Si, and the increase of porosity level from the surface to the centre of castings. The growth velocities of eutectic α-Al phase estimated using the Jackson-Hunt model confirm that the growth variation is larger in the thin castings than that in thick castings.

KEYWORDS

Aluminium alloys, Castings, Microstructure, Mechanical properties, Inhomogeneity

INTRODUCTION

Inhomogeneity is an inherent characteristic in as-cast microstructures of a variety of alloys because of their non-uniform chemistry, microstructure, defects, topology and the resultant mechanical properties in the different ranges of localization (Campbell, 2003). Generally, three types of inhomogeneities characterise casting microstructures, which include (a) the casting defects of shrinkage and gas porosity, misruns, cold shuts, inclusions, hot tears and hot spots; (b) the non-uniform microstructure from the surface to the centre of castings; and (c) the non-uniform microstructure from the dendrites grown from primary solidification and the structure formed during the subsequent eutectic solidification in inter-dendritic regions. The chemical composition and the morphology between the primary dendrites and the inter-dendritic eutectics also differ in the as-cast microstructure. Obviously, the inhomogeneity is different in the castings obtained from different casting processes.

High pressure die-casting (HPDC) is a casting process extensively used for manufacturing thin-wall components and lightweight structures (Vinarcik, 2003; MacLean, 2003). The die-cast components have high inhomogeneity compared to that produced by many other casting processes (Ji, 2012; Otarawanna 2009), which predominately originates from high speed and high turbulent melt flow during die filling, high cooling rate and two-stage solidification in HPDC (Barone, 2000).This results in the formation of non-uniform microstructure and uncertain distribution of defects in die-castings. In general, the as-cast microstructure is significantly refined under high cooling rates, but is differentiated in several regions. Near the casting surface, fine grains form a defect-free layer, which is described as 'the skin' or 'the surface layer' (Otarawanna, 2009; Yang, 2013). The zone located adjacent to the surface layer toward the centre, which provides an increased volume fraction of eutectic phase is usually defined as 'the band zone' (Otarawanna, 2009). In the central region, a bimodal grain microstructure consisting of a mixture of fine grains solidified inside the die cavity and coarse grains formed due to the external solidification in the shot sleeve (denoted as ESGs) is usually observed over the cross section of castings. The segregation of ESGs in the central region characterises the variation of the microstructure in die-castings. On top of the accumulation of ESGs, the central region always contains increased levels of defects such as porosities and inclusions, and intermediate phases.

For many years, the phenomenological study on microstructure and mechanical properties continues to cover a variety of die-cast alloys. The studies on the subject of inhomogeneity largely involve the effect of defects, gating system and casting parameters (Seo, 2006; Cleary, 2006), solidification and microstructural evolution (Ji, 2013; Chen, 2003),the effect of skin (Taylor, 2012), the identification of various phases (Ji, 2013),concentration profiles in the castings (Ferraro, 2015), quality mapping (Timelli, 2008), the relationship between microstructure and mechanical properties under different casting conditions and under different heat treatments(Lumley, 2011). However, there is still lack of information on what the mechanical properties are at different locations in castings. To the knowledge of the authors, no direct quantitative determinations exist of the non-uniform yield strength, ultimate tensile strength (UTS) and elongation at different points on the cross section of castings, especially regarding the relationship among the microstructure, solute concentrations, defect levels and the mechanical properties at different locations.

In the present paper, the die-cast rectangular tensile samples with different thicknesses were studied for the characteristics in microstructure and mechanical properties. A number of 2 mm thick samples were obtained by machining 3 mm and 5 mm thick samples at different locations, and the mechanical properties were examined and compared with directly cast samples in associated with the local microstructure, solute concentrations and porosities. The discussion was focussed on the relationship among inhomogeneity, solidification and mechanical properties. The growth velocity of eutecticα-Al phase was estimated using the Jackson-Hunt model for different locations in the casting samples.

EXPERIMENTAL

The experimental casting samples of Al-Mg-Si alloys were made using commercial alloys. One mushroom casting sample with $\phi60\times10$ mm testing part was made for composition analysis using an optical mass spectroscopy. The measured compositions of the experimental alloy were 7.5±0.4wt.%Mg and 2.4±0.3wt.%Si with the balance Al and inevitable impurities. A 4500kN cold chamber HPDC machine was used to make ASTM standard tensile samples. The pouring temperature was 650°C. Six samples (3 of $\phi6.35$ mm and 3 rectangular samples with a cross section of 2×6.35, 3×6.35 and 5×6.35 mm, respectively) were made in each shot, but only the rectangular samples were used in this study. The tensile samples were randomly divided into two groups for each thickness. One group was for direct mechanical testing and microstructural analysis. Another group was used to machine new samples at different locations on the cross section. Two 2-mm thick rectangular samples were machined from 3-mm castings and four 2-mm thick rectangular samples were machined from 5-mm thick castings. The distance from the casting centre is defined as the central line distance between the machined samples and the original die-cast samples.

The tensile tests were conducted following the ASTM B557 standard using an Instron 5500 Universal Electromechanical Testing Systems equipped a ±50 kN load cell. The gauge length of the extensometer was 25 mm and the ramp rate for extension was 2 mm/min. The specimens for microstructure and porosity examination were cut from the middle of each tensile test sample. The microstructure and porosity were examined using a Zeiss optical microscope, and a Zeiss Supra 35VP scanning electron microscopy (SEM) equipped with energy-dispersive X-ray (EDX). The porosity was measured on an unpolished sample without etching. The porosity, grain size, and the volume fraction of the solid phase were measured using an AxioVision 4.3 Quantimet digital image analysis system.

RESULTS

Figure 1 shows the mechanical properties of the die-cast rectangular samples with different thicknesses. It is seen that the yield strength slightly decreased but the UTS and elongation increased with increasing sample thickness. When the sample was 2-mm thick, the yield strength, UTS and elongation were 177.1 MPa, 302.1 MPa and 6.7%, respectively. However, when the sample thickness was increased to 5 mm, the yield strength decreased to 168.9 MPa, but the UTS and elongation increased to 350.2 MPa and 9.6%, respectively. The decrease of the yield strength was only 4.5%, but the increment of the UTS and elongation was 16% and 40%, respectively, when the sample thickness was increased from 2 to 5 mm, confirming that the increase of casting thickness was beneficial for the UTS and elongation with a slight sacrifice of the yield strength.

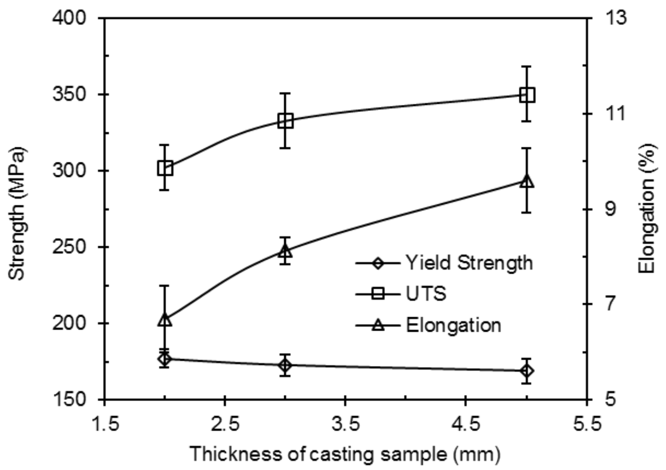

Figure 1. Effect of the thickness of die-cast rectangular samples on the tensile properties of the alloy.

To investigate the inhomogeneity in the mechanical properties of die castings, several 2-mm thick rectangular samples were machined from the 3 mm and 5 mm thick rectangular castings. The tensile properties of 2 mm thick machined samples at different locations are shown in Figure 2. When machining the 3 mm thick castings, only two 2 mm thick rectangular samples could be obtained; one was in the middle with two machined surfaces (denoted as casting centre in Figure 2a) and the others was at the casting surface with one machined surface (denoted as casting skin in Figure 2a). The yield strength, UTS and elongation of the samples at the casting centre were 167.4 MPa, 310.6 MPa and 6.3 %, respectively, but they were 175.3 MPa, 325.7 MPa and 8.9% for the samples at the casting skin. The variation was 5% for the yield strength and UTS but 41% for the elongation. This suggested modification is based on the consideration that the die-cast skin has improved tensile properties but its effect on the overall performance of the cast part needs further elaboration. It would probably have a definite positive impact on fatigue but its overall contribution the tensile properties of the part or alloy is only local.

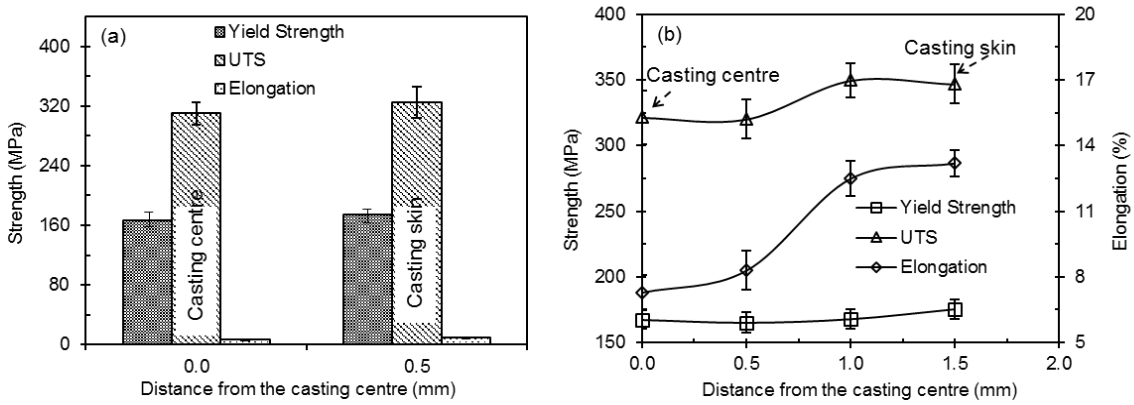

Figure 2. The tensile properties of the alloy with different surface conditions and locations of the 2 mm thick rectangular samples machined from (a) 3 mm thick rectangular casting samples and (b) 5 mm thick rectangular casting samples.

The similar observations were also found in the 2 mm samples machined from 5 mm thick castings. The mechanical properties are shown in Figure 2b for the samples at different locations. As seen, four 2 mm thick samples could be obtained from 5 mm thick castings with 0.5 mm intervals of locations. The variation of mechanical properties was significant from the surface to the centre of castings. The yield strength, UTS and elongation at the casting centrewere167.3 MPa, 321.4 MPa, and 7.3%, but those at the casting skin were increased to 175.5 MPa, 346.8 MPa and 13.2%, respectively. The increment was 8% for the UTS, 5% for the yield strength, and 80% for the elongation. It is important to note that the machined samples from the 5 mm thick casting surface provided very similar yield strength, but significantly improved UTS and elongation in comparison with those from the 3 mm thick casting surface. The variation was 7% for the UTS and 48% for the elongation between the skin samples obtained in the 3 mm thick castings and the 5 mm thick castings.

Three observations were obvious from the experimental results. (1)The thickness of samples could significantly differentiate the tensile properties of the die-castings. (2) The machined samples from the casting skin provided higher yield strength, UTS and elongation than those from the casting centre. (3) Tensile properties were different at the same location in different thick castings. Therefore, the inhomogeneity in the mechanical properties was significant in the experimental alloy and castings.

Figures 3–5 show the as-cast microstructure on the cross section of rectangular die-cast samples. The segregation of primary α-Al phase was apparent and more ESGs were found in the central region on the cross section. At the local area where the tensile samples were taken, two types of primary α-Al phases were seen in the centre region in all the samples, but the ESGs were rarely observed in the skin region. The quantitative analysis showed that the volume fraction of the primary phase was at a level of 68% in the centre and 40% in the skin of casting samples. The coarse primary phase (α_1) was almost absent in the casting surface and gradually increased from the surface to the central region. The variation was up to 30% in the experimental samples, which represented the significant inhomogeneity in casting microstructure. This phenomenon was also observed in the samples with different thicknesses. The analytical results of grain sizes are shown in Figure 6a, in which the coarse grains (α_1) became smaller near the casting skin than that in the casting centre, but the fine primary grains (α_2) showed no visible variations in size. The fine α-Al phase exhibited similar sizes from the surface to the centre of castings with different thicknesses. By combining the grain size and the volume fraction, the calculated overall grain sizes are shown in Figure6b for the experimental alloys. The grains sizes were obviously larger in the casting centre and smaller in the casting skin. However, the grain sizes of the primary α-Al phase were at similar levels at the skin and the centre of different samples.

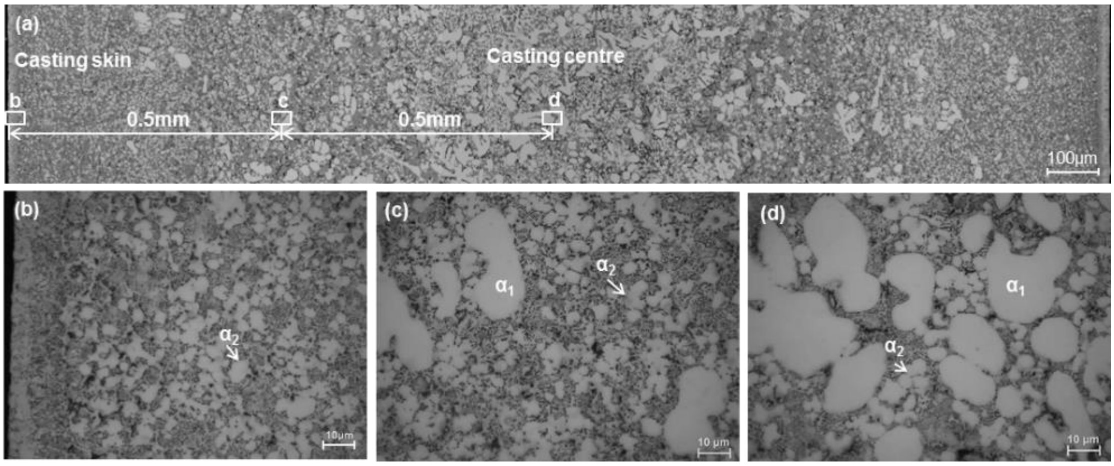

Figure 3. Optical micrographs showing the microstructure of the die-cast alloy, (a) on a cross section of 2 mm rectangular tensile sample, (b) on the skin, (c) 0.5 mm from the casting centre, (d) in the centre.

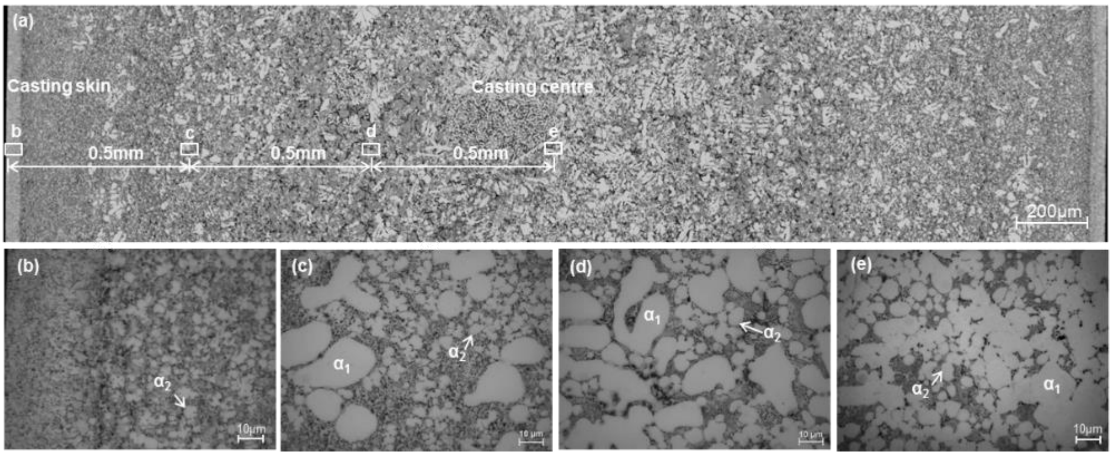

Figure 4. Optical micrographs showing the microstructure of the die-cast alloy, (a) on a cross section of 3 mm square sample, (b) on the skin, (c) and (d) 0.5 mm intervals from the casting centre, (e) in the centre.

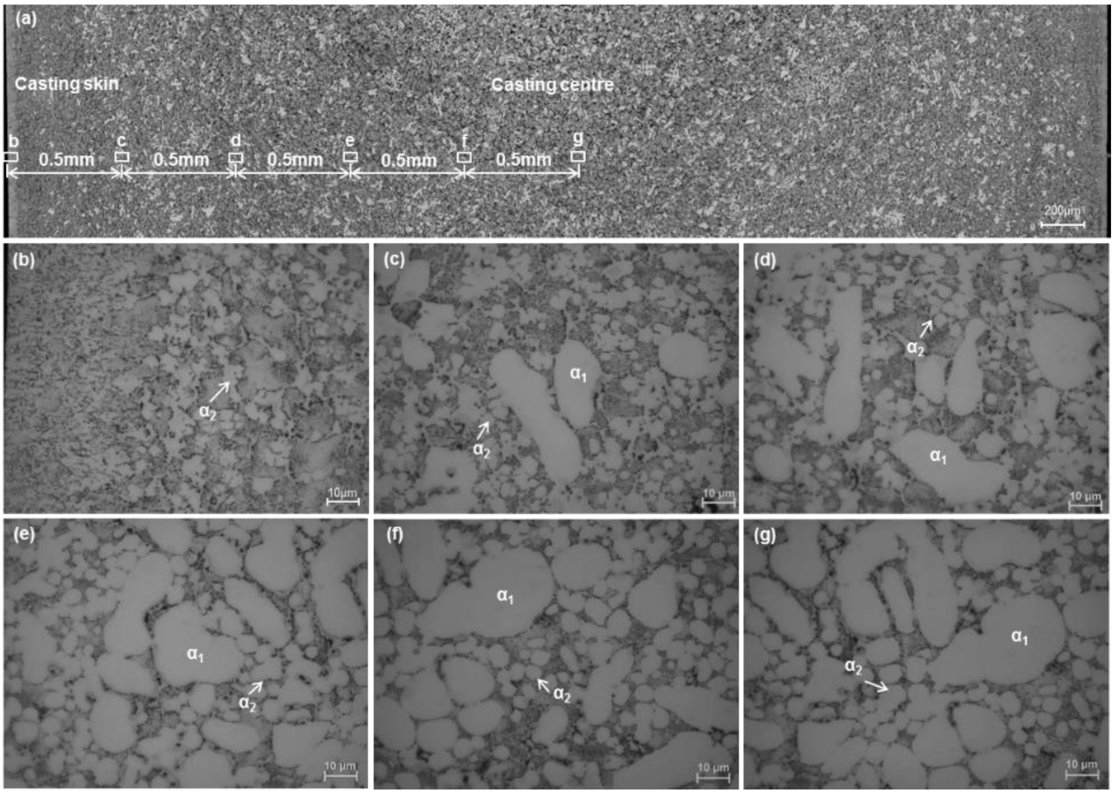

Figure 5. Optical micrographs showing the microstructure of the die-cast alloy, (a) on a cross section of 5 mm rectangular sample, (b) on the skin, (c) to (f) 0.5 mm intervals from the centre, (g) in the casting centre.

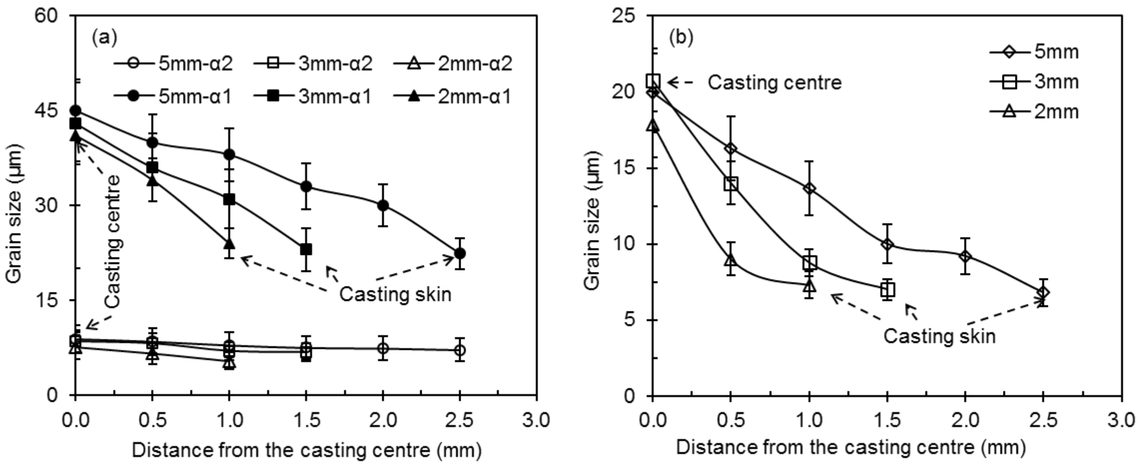

Figure 6. Grain sizes of the primary α-Al phase on the cross section of the die-cast alloy with different thicknesses of rectangular tensile samples, (a) the average size of the coarse primary α-Al phase (α_1) and the fine coarse primary α-Al phase (α_2), (b) overall average size of the primary α-Al phase.

The microstructural inhomogeneity was also observed from the eutectic phase. The eutectic regions of die-cast samples with different thicknesses are shown in Figure 7. Clearly, the finer eutectic microstructure was observed in the casting skin in comparison with that in the casting centre. In the meantime, it is also seen that the eutectic micro constituent in the casting centre was finer for the as-cast sample with 2 mm thickness than that of the 5 mm samples.

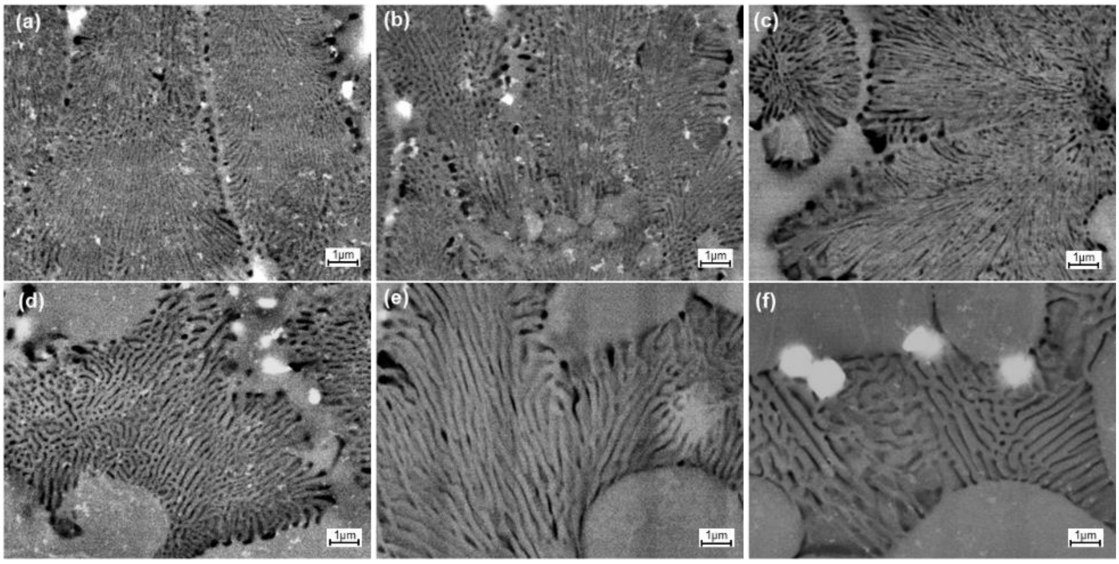

Figure 7. Backscattered SEM images showing the eutectic microstructure of (a) and (d) 2 mm, (b) and (e) 3 mm, and (c) and (f) 5 mm thick rectangular tensile samples at the skin (a,b,c) and in the centre (d,e,f).

The detailed quantitative analysis is shown in Figure 8. The eutectic phase exhibited a higher volume fraction near the casting skin and a lower volume fraction in the casting centre. However, the eutectic spacing was significantly larger in the casting centre than that at the casting skin. It is noted that the eutectic spacing was at a similar level of 0.23 to 0.25 μm at the casting skin for the samples with three thicknesses, but the eutectic spacing at the casting centre was varied from 0.38 μm for the 2 mm thick sample to 0.48 μm for the 5 mm thick sample.

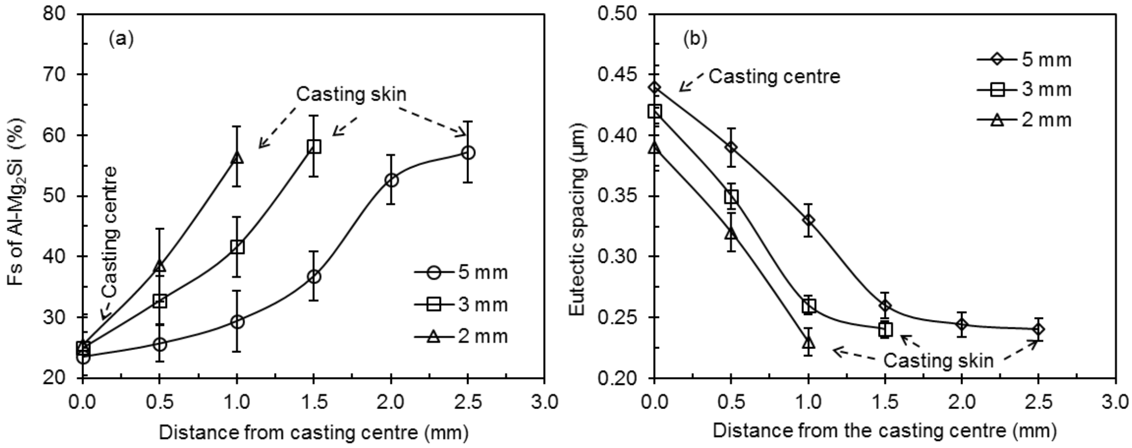

Figure 8. Measured (a) volume fractions and (b) the eutectic spacing of the eutectic Al-Mg_2Si phase on the cross section of the alloy with different thicknesses of the rectangular tensile samples.

The concentrations of solute Mg and Si were generally consistent in the skin region but with a gradual drop toward to the centre inside the band. The Mg concentration was close to the nominal composition of 7.5wt.% in the skin region and gradually decreased to 6.4wt.% at the casting centre.

Similarly, the Si concentration was also close to its nominal composition at 2.4wt.% in the skin region but slightly lower in the central region. However, the solute elements showed an apparent increase in the band zone. Mg and Si were enriched to 11.8wt.% and 3.3wt.%, respectively. Obviously, the peak of the solute enrichment in the band zone was much higher than the nominal composition in the alloy. The results confirmed the existence of significant segregation and non-uniform distribution of the solute elements on the cross section of die-castings. Furthermore, a significant difference of the porosity levels was found between the skin region and the central region. The levels of porosity were increased from the skin to the centre of casting samples. It was less than 0.3% of porosity in the skin region, but over 1% of porosity in the central region. Therefore, inhomogeneity in porosity levels was significant in the die-cast samples.

DISCUSSION

In all the casting processes, melt is cooled down to form as-cast microstructure. Therefore, the inhomogeneity in castings is essentially associated with the solidification process. The solidification in high pressure die-casting occurs in two stages in the shot sleeve and in the die cavity. The solidification in the shot sleeve is under a conventional way to form a solid layer at the bottom of the shot sleeve, which is most likely broken and smashed and mixed with liquid during the forward movement of piston. The mixtures of solid and liquid phases are further being sheared when passing through the ingate and filling the die cavity. During die filling, the ESGs are experienced a pressure driven flow with high speed and high turbulence. Because the shear rate near the casting skin is close to zero and that at the central region are significantly higher according to the fundamental theory (Nott, 1994), the ESGs therefore suffer shear-induced migrations from the skin towards the casting centre. These results in suspension of the segregated ESGs in the melt at the end of die-cavity fill, which is different from the melt near the ingate which has a relatively homogeneous dispersion. It needs to be emphasised that the melt flows throughout the whole die cavity and stop in overflows. Therefore, the segregation of ESGs exists in the whole casting. The liquid temperature after shearing during die filling is supposed to be uniform and without a significant superheat. The melt starts solidification immediately after die filling. Because the ESGs have a temperature lower than the melting point of the alloy, the liquid surrounding the ESGs is essentially undercooled at the moment of filling the die cavity. As a result, the solidification of the remnant liquid happens simultaneously throughout the die cavity, forming a number of fine grains in the form of globular particles or fine dendrites. The primary α-Al phase formed in the die cavity shows less significant inhomogeneity because of the fine grain sizes. Meanwhile, as the shear rate at the casting skin (mould or die surface) is zero in principle, the melt is thus stuck on the die surface at the first moment of die filling. The melt away from the mould surface continues the filling process, forming a dynamic balance between the static skin and the turbulent flow in the inner channel. Therefore, the microstructure in the skin region has different morphology to that in the central region. Moreover, the rapid solidification for the segregated mixture of the ESGs and melt inside the die cavity will restrain the release of the gas introduced by turbulence in the melt. Consequently, the porosity in the casting central region is higher than that in the skin region.

The eutectic solidification happens after full cavity fill. Because the solidification inside the die cavity also includes the formation of primary α-Al grains, the eutectic reaction is supposed to occur under a static condition and confined to the small intergranular areas. From the experimental results, the eutectic spacing from the casting skin to the casting centre is slightly different, indicating the cooling rate varies during solidification. The high cooling rate is able to lead the formation of fine eutectic spacing and the size of eutectic cells. Currently, the direct measurement of the cooling rate in the die cavity is still very difficult, in particular for the variation of cooling rate on the cross section of thin wall die-castings. Therefore, it is worth of estimating the difference of cooling rates using the easily measured eutectic spacing.

According to the Jackson-Hunt theory of eutectic growth (Jackson, 1966), the relationship between eutectic spacing λ and growth velocity V follows

$$\lambda^2 V = constant \tag{1}$$

Although the Jackson-Hunt theory is generally suitable for most eutectic reactions, the determination of the constant is still a challenge and the results vary in the published literature. Because no accurate data were found for AlMgSi alloys, it is expected that the constant at a range of 130–160 μm^3/s is approximately acceptable (Ji, 2017). When $\lambda^2 V$ = 130 to 160 μm^3/s, we can obtain different growth velocities with the variation of λ from the skin to the centre of die-castings. Therefore, the possible growth velocity for the different thick samples can be calculated using the measured eutectic spacing. The results are shown in Figure 9, in which a band provides the growth range for each casting thickness. The typical growth velocity at the skin and the centre is 262 5 μm/s and 750 μm/s for the 5 mm thick casting, 2740 μm/s and 820 μm/s for the 3 mm thick casting, and 2865 μm/s and 1004 μm/s for the 2 mm thick casting, respectively. It is understandable that the variation of the growth velocity is larger for the thin casting than that of the thick castings. However, the growth velocities near the casting surface are very close for the different castings. And, the variation of growth velocity from the skin to the centre is much less for the 5 mm thick castings than that for the 2 mm thick castings.

Generally, the growth velocity is proportional with the undercooling in the melt, which is significantly affected by the cooling rate of solidification (Schwarz, 1997). A high cooling rate always results in a large undercooling. At a large undercooling, the number of nuclei that can grow up in the melt during nucleation is increased, leading to the formation of an increased number of eutectic cells and to decreased cell sizes during solidification. Meanwhile, the growth is controlled by solid-liquid interface movement through the redistribution of solute in front of the interface. The high cooling rate can form an increased thermal gradient ahead of the solid-liquid interface, by which the diffusion is promoted and therefore the growth velocity steeply rises during solidification.

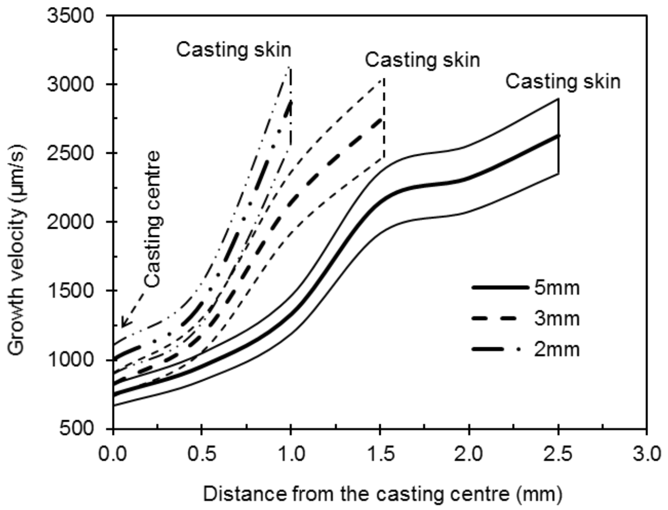

Figure 9. Calculated growth velocity of eutectic α-Al phase according to the eutectic spacing of α-Al phase on the cross section of the alloy with different thicknesses of the rectangular tensile samples.

CONCLUSIONS

Macro-inhomogeneity in tensile properties is closely associated with the microstructure, solute concentrations and defect levels in the as-cast Al-Mg-Si alloys. Under the same ingate velocity and overflow, the 2 mm die-cast samples provide slightly higher yield strength and lower UTS and elongation than the 3 mm and 5 mm die-cast samples. For the local mechanical properties on the thick die-castings represented by the machined 2 mm thick samples, the casting skin provides the highest, whereas the casting centre provides the lowest yield strength, UTS and elongation. The yield strength, UTS and elongation are significantly decreased towards the casting centre.

The inhomogeneity in microstructure is mainly characterised by the significant segregation of ESGs from the surface to the centre of castings, although the primary α-Al phase and the eutectic Al-Mg_2Si phase solidified in the die cavity also show a slight difference in terms of grain size and eutectic spacing. The growth velocity of eutecticα-Al phase can be calculated using the Jackson-Hunt model. The variation of growth velocity is larger in the thin castings than that in thick castings. The typical growth velocity at the skin and the centre is 2625 μm/s and 750 μm/s for the 5 mm thick casting, 2740 μm/s and 820 μm/s for the 3 mm thick casting, and 2865 μm/s and 1004 μm/s for the 2 mm thick casting, respectively. However, the growth velocities near the surface are close in the different thick castings.

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