



Effect of grain refiner on the tensile and impact properties of Al–Si–Mg cast alloys



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ABSTRACT

The present study aims to investigate the influence of the addition of Ti and B in the form of five different grain refiners/aluminium master alloys (Al–10%Ti, Al–5%Ti–1%B, Al–2.5%Ti–2.5%B, Al–1.7%Ti–1.4%B and Al–4%B) in conjunction with that of Sr (as modifier) added in the form of Al–10%Sr master alloy to A356.2 alloy. Grain refinement of an A356.2 alloy with Ti and B additions in the ranges of 0.02–0.5% and 0.01–0.5%, respectively, was examined using these different types of grain refiners. Strontium additions of 30 and 200 ppm were made. All alloys were T6-heat treated before mechanical testing. Tensile and impact tests were conducted to evaluate the influence of the interaction between grain refiner and modifier on the mechanical properties. The properties were determined for both the as-cast and heat-treated conditions.

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1. Introduction

The literature reports a number of studies on mechanical properties of the A356 alloy. Some studies have suggested that the ductility should increase as the dendrite cell size decreases [1]. An alternative view is that ductility is controlled by the size of the largest silicon particles [2]. Yet a third viewpoint is that the ductility of unmodified alloys is controlled by the mean size of the silicon particles while that of modified alloys is determined by their distribution [2].

In the A356 alloy, the ductility is determined not only by the dendritic structure (or what is left of it after solution treatment) but also by other parameters, notably silicon particle size and shape [1]. It has been reported that the ductility of this alloy is not a simple monotonic function of either the dendrite cell size or the particle size and shape as is often assumed. More important are the facts that the fracture mode depends on whether or not the cell structure retains some integrity after solution treatment and that for the coarser structures, when the fracture is transgranular, the ductility increases with increasing cell size if the particle size and shape remain constant [3–5]. For Al–Si alloys, the impact strength can provide an interesting estimation of the ductility of an aluminum alloy. In addition, as it is influenced by the fineness of the microstructure and the morphology of the constituents, it can be a good indicator of the Al–Si alloy modification. For an

Al–6.5%Si–0.3%Mg alloy, in T6 condition, the values for U-notched, V-notched and unnotched samples are respectively 7.26 J/cm², 5.10 J/cm² and 22.16 J/cm² [3].

In addition, inappropriate melting and casting conditions can lead to the occurrence of gross defects such as shrinkage and gas porosity or to entrapped dross and oxide films. Hydrogen is the only gas with significant solubility in molten aluminum. This gas plays a major role in the development of unsoundness due to porosity in castings. Hence, degassing is a basic step in the aluminum casting process. Moreover, in comparison with sand casting, permanent mold casting usually produces finer microstructures and fewer and smaller pores due to reduced solidification time, which should lead to an improvement in mechanical properties, e.g., fatigue performance [6].

Grain refinement of the casting may also alter the amount and the morphology of pores in the casting. The presence of the grain refiner leads to a redistribution of porosity. In many cases, there is also an overall reduction in the amount of porosity upon grain refinement [7,8]. Since the radius of the interdendritic liquid pool decreases with the addition of Ti, the average pore size is lowered. Under these conditions, the gas pressure in the liquid is not strong enough to overcome the contribution of interfacial energies and pore formation becomes difficult. The formation of pores in a casting occurs on hetero-geneous nucleation sites, as homogenous nucleation of pores would require very high gas content and shrinkage pressure [9]. It has been proposed that nucleation occurs on oxide films [10] and particles [11]. Titanium diboride particles,

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especially TiB_2 clusters, have also been found to nucleate porosity [3].

A significant improvement in the mechanical properties of aluminum alloys may be obtained by means of the addition of small amounts of Al–Ti–B master alloys which contain microscopic TiB_2 and Al_3Ti nucleating particles [12]. When an Al–Ti–B grain refiner is stirred into the melt, the Al_3Ti particles dissolve rapidly and introduce solute titanium into the melt to assist growth restriction after the nucleation event. The TiB_2 particles remain stable in the aluminum melt and provide sites for heterogeneous nucleation. That is why these particles act as the nucleating substrates. These master alloys are effective grain refiners whereas those containing only TiB_2 or Al_3Ti particles are much less effective. Various theories regarding the grain refining mechanism of Al–Ti–B refiners have been proposed, including the particle theory, the phase diagram theory, the duplex nucleation theory, and the peritectic hulk theory [13].

In the case of the Al–Ti type grain-refiners containing only Al_3Ti particles, these particles must act as the nucleating substrates; their effectiveness, however, is relatively poor when compared with that of Al–Ti–B type grain-refiners [14]. Al–Si alloys may also be efficiently grain refined by Al–4%B master alloys. It has been suggested that the grain refinement mechanism for this master alloy is the heterogeneous nucleation of aluminum crystals on intermetallic AlB_2 particles, combined with a more common growth restriction process resulting from the segregation of silicon during solidification. As regards the Al–B type grain refiner, Sigworth and Guzowski [15] proposed that TiB_2 particles act as nucleants, and the presence of Si enhances their nucleating potential.

Easton and St-John [4] proposed that the optimum chemical grain refiner needs to include both solute titanium (with its very high growth restriction factor, GRF) and nucleant particles (e.g. TiB_2 or $TiAl_3$) in order to produce effective grain refinement. For this reason, the Al–5Ti–1B type of refiner is extremely effective. The GRF of any solute element is given by the factor, $mC_0(k-1)$, where m is the gradient of the liquidus line in the binary alloy phase diagram, C_0 is the concentration of the solute in the alloy, and k is the partition coefficient. It is generally assumed that individual GRFs are additive in multi-component alloy systems, however this method can grossly overestimate the GRF value.

The beneficial effect of boron in conjunction with titanium to grain refine aluminium alloys has also been explored by others [16,17] and alternative mechanisms have been proposed. These have involved various permutations of mixed (Al, Ti) B_2 borides and $TiAl_3$ that become stabilized at sub-peritectic compositions. Easton and St-John [4] also suggested that because aluminium casting alloys, such as A356, already contain high solute levels and a high GRF (the GRF of 7%Si is equivalent to that of 0.17%Ti), the optimum grain refiner really only needs to contain nucleant particles. This means that master alloys such as Al–2.2Ti–1B (i.e. TiB_2 particles with no extra solute Ti) or Al–4B (i.e. containing AlB_2 and/or AlB_{12} particles) are quite suitable as grain refiners for foundry alloys [9]. It has been reported that optimum percentages of Ti and B for the A356 alloy are approximately between 0.06%Ti, 0.01%B and 0.08%Ti, 0.02%B. Higher additions of Ti and B cause the formation of Ti-based intermetallic compounds within the eutectic region [16].

Lu and Dahle [18] have reported that an A356 melt treated with Al–1.5Ti–1.5B loses its Sr much more quickly, particularly in the initial stage after addition, compared to a melt treated with Al–5Ti–1B. This explains the quick loss of eutectic modification in the Al–1.5Ti–1.5B treated melt, i.e. there is insufficient free Sr in the melt to modify all the eutectic Si. It is well known that molten Al–Si alloys can lose their Sr through surface oxidation and/or vaporization [5]. However, whichever mechanism is operating, the Sr concentration in the melt is expected to decrease gradually over

Table 1
Composition of the base A356 alloy used in this study (wt.%).

Al	Si	Mg	Cu	Ti	Mn	Zn	Fe
Bal.	7.21	0.41	<0.01	0.11	<0.01	<0.01	0.08

time. Li et al. [19] reported that when Al–10%Sr was added into an Al–Si–Cu melt together with Al–3%Ti–4%B (at the holding temperature of 725 °C), the precipitation of B from $(Al,Ti)B_2$ in Al–3%Ti–4wt.%B alloy and Sr from Al_4Sr in Al–10%Sr alloy resulted in the formation of SrB_6 , which has a very high melting temperature of about 2500 °C [20]. The weight ratio of Sr to B in SrB_6 is 1.35:1, therefore, the consumption of Sr is larger than that of B in the precipitation process, producing less SrB_6 and more dissolved B, which consequently resulted in a smaller grain size.

To clarify the effect of combined addition of silicon modifier and grain refiners, i.e., Sr–B interaction on the tensile properties and impact toughness of the A356.2 casting alloy, the variation in these properties is discussed and analysed with respect to the addition level and type of master alloys used for this study.

2. Experimental procedure

The A356.2 alloy used in this study was received in the form of 12.5 kg ingots. Its chemical composition is given in Table 1. Melting was carried out in an electrical resistance furnace equipped with a rotary degassing system using pure argon. Addition of Sr was made using an Al–10%Sr master alloy to obtain strontium levels of 30 ppm and 200 ppm. Grain refiner additions were then made to the Sr-modified A356.2 alloy melt using different grain refiner types, and in various concentrations. The master alloy amounts were added in terms of the weight percent of Ti or B required for achieving the required concentrations in the melt. Castings were made to prepare tensile test bars using an ASTM B-108 type metallic mold, and impact samples using a metallic mold, each casting providing ten impact test bars.

Metallography samples were sectioned from the as-cast and T6-heat treated tensile tested specimens. These sections were mounted in bakelite, then polished to a fine finish for examining the corresponding microstructures. For measurements of grain size, the samples were etched to reveal the grains in the macrostructure.

The metallography samples were examined using optical microscopy and scanning electron microscopy (SEM), coupled with energy dispersive X-ray (EDX) and wavelength dispersion spectroscopic (WDS) analysis facilities. Measurements of the eutectic silicon particle characteristics and grain size were carried out using a Clemex image analysis system in conjunction with the optical microscope. Identification of new intermetallic phases formed with the addition of the different grain refiners was done using electron probe microanalysis (EPMA). The WDS analysis provided the means to determine the chemical compositions and formula of the observed intermetallics. The impact tests were performed on unnotched samples. A computer-aided instrumented SATEC SI-1 Universal Impact Testing Machine (model SI-ID3 from Satec system Inc.) was used to carry out the impact tests. The average values of the energies obtained from the six samples tested for each condition were taken as the representative values for that particular condition. Total number of tensile and impact tests carried out in the present study was about 1200.

The tensile and impact samples were heat treated following the T6 temper, which comprised solution heat treatment at 540 °C for 8 h, followed by quenching in 60 °C water. The samples were then aged at room temperature for 24 h, followed by artificial aging at

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