Materials and Design 57 (2014) 366–373

Contents lists available at ScienceDirect

Materials and Design

journal homepage: www.elsevier.com/locate/matdes

Effect of Fe content on the fracture behaviour of Al–Si–Cu cast alloys

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ARTICLE INFO

Article history:
Received 10 October 2013
Accepted 18 January 2014
Available online 22 January 2014

Keywords:
Aluminum alloys
Impact testing
Fractography
Intermetallics
Porosity

ABSTRACT

Castings were prepared from 319.2 alloy melts, containing Fe levels of 0.2–1.0 wt%. Sr-modified (~200 ppm) melts were also prepared for each alloy Fe level. The end-chilled refractory mold used provided directional solidification and a range of cooling rates (or dendrite arm spacings, DAS) within the same casting. Impact test samples were machined from specimen blanks sectioned from the castings at various heights above the chill end provided DASs of 23–85μm. All samples were T6-heat-treated before testing keeping with Aluminum Association recommendations. The results show that at low Fe levels and high cooling rates (0.4% Fe, 23 μm DAS), crack initiation and propagation in unmodified 319 alloys occurs through the cleavage of β-Al_5FeSi platelets (rather than by their decohesion from the matrix). The morphology and the size of the platelets (individual or branched) are important in determining the direction of crack propagation. Increasing the DAS to 83μm leads to cleavage fracture. In this case, the fracture path follows a transgranular plane that is usually a well-defined crystallographic plane as judged by the relatively large smooth surfaces of the β-Al_5FeSi phase platelets. Cracks also propagate through the fracture of undissolved $CuAl_2$ or other Cu-intermetallics, as well as through fragmented Si particles. In Sr-modified 319 alloys, cracks are mostly initiated by the fragmentation or cleavage of perforated β-phase platelets, in addition to that of coarse Si particles and undissolved Cu-intermetallics.The amount of undissolved Cu- intermetallics is directly related to the applied cooling rate. Slow cooling rate (DAS ≈83μm) results in the precipitation of Cu- containing phases on the β-platelets, amplifying the likehood for crack propagation through these loacations.

1. Introduction

Fractography is defined as the study and documentation of fracture surfaces [1]. The purpose of fractography is to analyze fracture features and attempt to relate the topography of the fracture surface to the causes and/or basic mechanisms of fracture [2]. The knowledge of fracture behavior is important in upgrading material specifications, improving product design, and analyzing failures for improved reliability [3].

Four major types of fracture (or failure) modes have been discussed in the literature. Briefly, these modes and the sources of their occurrence are as follows:

(1) Dimpled rupture (or microvoid coalescence) caused by ductile fracture or rapid overload fracture;
(2) Cleavage or quasi-cleavage caused by brittle fracture or premature or overload failure by catastrophic rapid fracture;
(3) Intergranular fracture caused by stress corrosion cracking, hydrogen embrittlement, or subcritical growth under sustained load; and
(4) Ductile striations caused by fatigue cracking or subcritical growth under cyclic load.

Ibrahim et al. [4] studied the fracture behavior of Al–Si–Mg (A356) alloys using tensile samples of different section sizes and under different conditions (as-cast/heat-treated, unmodified/Sr-modified), where *in situ* observations of the surface crack initiation and propagation using scanning electron microscopy were presented. Cracks were found to initiate with the fracture of Si particles at relatively low strain values. Crack propagation also proceeded by the same mechanism, preferably through the eutectic region of the microstructure, avoiding the primary Al dendrite cores wherever possible.

Hafiz and Kobayashi [5] conducted a study on the microstructure-fracture behavior relations in Al–Si casting alloys using tensile testing. In general, the voids were found to initiate at silicon particles. The individual voids then grew and coalesced, creating microcracks in the eutectic region. These microcracks linked up to form the main crack, resulting in the final fracture. According to Ragab et al. [6] in commercial A356 alloy, the fracture path of tensile-tested specimens proceeds mainly through the largest silicon particles, which constitute less than 1% of the overall population of silicon

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http://dx.doi.org/10.1016/j.matdes.2014.01.037
0261-3069/

particles in the bulk microstructure. Cáceres and Griffiths [7,8] pointed out that the larger and longer silicon particles are more prone to cracking. In coarser structures, silicon particle cracking occurs at low strains, while in finer structures, the progression of damage is more gradual. They also found that in the former case, broken particles were observed at both cell and grain boundaries, with no evident preference for either, whereas in the case of finer structures, cracking initiated on the grain boundaries. Ammar et al. [9] also reported that the fatigue behavior of Al–Si–Mg casting alloys depends on the size, orientation, and local distribution of the Si particles.

As strontium modification is commonly employed to alter the morphology of the eutectic Si particles in Al–Si alloys, the process is expected to affect the fracture behavior of the alloy. Hafiz et al. [10] studied the role of microstructure in relation to the toughness of hypoeutectic Al–Si casting alloys, using U-notched Charpy impact test samples. Additions of 0.017 or 0.03 wt% Sr were employed to modify the alloys. In the unmodified alloy, the fracture followed a path marked by the eutectic Si, circumventing the primary Al dendrites, where Si particles could be seen adhering to the sides of the crack path.

Hrong et al. [11] studied the fracture behavior of A356 alloys with different iron contents under resonant vibration. According to them, the cracks were found to initiate and grow along the eutectic Si and Fe-rich intermetallic phase particles. According to Ibrahim et al. [12] directly after the onset of plastic deformation, Al–7Si–0.3Mg sand-cast alloys display microcracks that are associated with brittle grain boundary cracking and cleavage of plate-shaped iron intermetallics, although the iron content is as low as 0.1 wt%. The work of Kato [13] on Al–7.0%Si–1.0%Fe alloys showed that the needle-like β-iron intermetallics are cracked easily in the earlier stages of deformation, whereas the α-iron Chinese script intermetallics can withstand a higher stress. Villeneuve and Samuel [14] reported that in Al–13%Si–Fe alloy castings, cracks appeared within the β-Al_5FeSi platelets rather than at the β/Al interface. This is due to the brittle nature of the β-phase, whereby the platelets are easily split into two halves.

In their investigation of the high-cycle fatigue testing of AlSi11 cast aluminum alloy, Stanzl-Tschegg et al. [15] showed that casting voids reduce the fatigue strength and lead to early fracturing. Crack initiation occurs at these voids, particularly if the porosity acts as a stress concentrator. The crack initiation times are strongly influenced by the position of the porosity, *viz.*, whether it is close to the surface or to the center of the specimen. Savelli et al. [16] also reported that, in cast AlSi7Mg0.3 aluminum alloys, internal pores are responsible for the crack initiation of fatigue cracks. Based on this, a mathematical model was proposed by them to predict the fatigue life using pore dimensions as the main parameter. The present authors studied the relationship between tensile and impact properties in Al-Si-Cu cast alloys, taking into consideration the cooling rate and Fe concentration [17,18]. The results revealed that the progress of the crack through the matrix of low Fe content (≈0.19%) and at the highest cooling rate (≈23μm) revealed that fracture of Si particles occurs mainly at the grain boundaries and the fracture path is predominantly transgranular. In samples with large DAS (≈83μm), particle cracking was concentrated at the grain boundaries and the fracture mode is intergranular (along the grain boundaries).In high Fe-content alloys, crack initiation occurs through the fragmentation of β-Al_5FeSi intermetallics, Si particles, and Al_2Cu particles.The crack propagation takes place through the cleavage of the β-Fe platelets. Cracks also propagate through the fracture of undissolved Al_2Cu or other intermetallics such as $Al_{7-}Cu_2Fe$ as well as through the brittle Si particles.The fibrous nature of the Si particles in Sr- modified alloys as well as the fragmentation of the β-Al5FeSi phase due to modification decreases the chances for crack propagation to occur through these particles.

In the present work, the 319 alloy with different Fe content were selected to study the effect of iron based intermetallics and porosity on fracture behavior of impact tested samples, and hence determine the fracture mechanism of these alloys. It should be mentioned here that for fractograpy part of the work, only alloys with DAS of 23 and 83μm were considered.

2. Experimental procedure

The chemical composition of the as-received alloy is listed in Table 1.The ingots were melted in a silicon carbide crucible of 7 kg capacity, using an electric resistance furnace (temperature deviation of ±5 °C). The melting temperature was held at 735 ± 5 °C. The Fe additions were made using Al–25%Fe master alloy in the required amounts Castings were done in a rectangular end-chill mold, as shown in Fig. 1(a). The four walls of the mold are made of refractory material, while the bottom is a water-chilled copper base, to promote directional solidification and ensure proper thermal insulation.

The molten metal was degassed and then poured through 20 ppi (8 pores/cm) ceramic foam filter disks fitted into the riser above the sprue, to avoid incident inclusions. Such a mold has to be assembled each time for a casting. Using this mold arrangement, casting blocks were produced that exhibited a range of dendrite arm spacings (DASs) along the length of the casting, corresponding to solidification rates that decreased (and, hence, DASs that coarsened) with increasing distance from the chill end. Charpy unnotched simple beam impact samples were used in accordance with ASTM: E23 standards. The standard proportions are shown in Fig. 2. All samples were T6-heat-treated before testing keeping with Aluminum Association recommendations. The fracture analysis work was carried out using a scanning electron microscope (SEM) operating at 15 kV.

3. Results and discussion

Table 2 indicates the average DAS that were obtained for the present 319 alloy, corresponding to different distances from the copper chill end. A decrease in impact energy with respect to an increase in iron content and decrease in cooling rate was observed in alloy 319.2, both in the non-modified and in the Sr-modified conditions, as shown in Table 3 and Fig. 3 which is in accordance with the results of other researchers [17,18]. Clearly, the highest cooling rate (DAS 23 μm) gives much better values in the unmodified and Sr-modified conditions than any other cooling rate. At the lowest iron content, the impact energies obtained at the highest cooling rate (DAS 23 μm) are about 2.5 and 2.7 times those obtained at the next cooling (DAS 47 μm) in the unmodified and Sr-modified alloys, respectively. Even at the highest iron content, the same holds true. Apparently, the impact energy is very sensitive to the cooling rate above a certain threshold, with the corresponding DAS representing the fineness of the microstructure and constituents at this cooling rate. Other workers, *e.g.*, Richard [19] and Tsukuda et al. [20] have also reported on the sensitivity of impact strength to the as-cast microstructure and to very small variations therein. In the present study, this threshold value appears to correspond to a DAS value of 23 μm, or thereabouts.

The SEM fractographs of the alloy sample corresponding to the 0.4% iron level and highest cooling rate are presented in Figs. 4 and 5, taken respectively from the sample edge where the sample was hit during impact testing, and the center of the sample surface. The features exhibited in Fig. 4 are typical of cleavage fracture, as is expected to be the case at high deformation rates during impact loading. The fracture plane changes orientation from grain to grain. A certain amount of intergranular cleavage is also observed in the form of secondary cracks. The fracture surface reveals the presence