Effect of Solidification Times on Crack Opening Displacement of Aluminum Alloy Castings

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The effects of mold temperature, solidification time, and metallurgical structure on both the crack-tip opening displacement (COD) at maximum load and ultimate tensile strength (UTS) of Al-10.6Si, 0.9Mg, 0.9Cu alloy have been investigated. Seven alloy melt samples poured into a copper mold, which was pre-heated at different temperatures, were allowed to solidify at different cooling rates. Even though it was found that the COD values increase and the UTS decreases as the mold temperature and solidification time increase, these parameters could be related to microstructural differences such as silicon crystals size in an α-aluminum matrix.

Keywords crack opening displacement of aluminum castings, pre-heating temperature, silicon particle size in aluminum matrix, solidification time

1. Introduction

Fracture mechanisms of metallic materials have being widely studied by fracture mechanics testing. Currently, these tests are performed to evaluate industrial components and to develop novel advanced prototypes with improved fracture toughness properties. However, for relatively tough materials it is necessary to use test- pieces of dimensions large enough to yield reliable results according to the Linear Elastic Fracture Mechanics (LEFM) criterion; and also that the test-pieces may not be completely representative of the behavior of the components used in service. Thus, performing reliable alternative measurements of materials resistant to fast fracture, by using small test-pieces, so that the toughness parameter measured may be related directly and quantitatively to the fracture toughness of the material, is of strong interest. In this case, the critical value of the crack-tip opening displacement (COD) at the fracture is a parameter that provides information of the fracture toughness of the material. Cottrell[1] and Wells[2] have studied the crack opening prior to crack extension as a parameter that might be treated as characteristic of the crack tip region, for a given material tested under specific set of conditions. The sliding displacement used by Cottrell may equally be replaced by an opening displacement δcrit, for tensile loading. The possibility arises of measuring the value of δcrit for a material on a small test piece, which breaks well after general yield[3] and using this same value to predict the failure stress of a large structure, which breaks before general yield using the equation:

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$$\delta crit = \frac{8\sigma_y}{\pi E} aLn\left[\sec\left(\frac{\pi\sigma_f}{2\sigma_y}\right)\right] \quad \text{(Eq 1)}$$

where σ_f is the failure stress, E is Young's modulus, and σ_y is the yield stress.

For the same large structure, which breaks well before general yield, the term (σ_f/σ_y) is small and Eq 1 may be written as:

$$\sigma_f = \sqrt{\frac{E\sigma_y\delta crit}{\pi a}} \quad \text{(Eq 2)}$$

Clearly, the critical values of COD as defined by Eq 1 apply only to the initiation of further cracking growth. They do not characterize the point of total instability. There may be a substantial difference between the value of δcrit at initiation and that at total instability.

Dawling and Martin[4] have reported that adding Mn (0.5 wt.%) to Al-Mg-Si alloys has the effect of reducing the tendency for intergranular embrittlement in the fully aged condition. Several attempts to measure the critical stress intensity factor (K_{IC}), as defined by the ASTM testing method for plane strain fracture toughness, have been unsuccessful for 5083-0 aluminum alloy. The parameter K_{IC} applies to linear elastic loading behavior; however, Kaufman et al.[5] have shown that this aluminum alloy, for cross sections up to 20 mm, remains sufficiently plastic as to invalidate K_{IC} measurements. Kaufman and Kelsey[6] also noted relatively isotropic fatigue crack growth in their test of 5083-0 Al-alloy. Plasticity and stable crack extension caused non-linearity in the fracture test records at low temperatures and unstable cracking has never been observed[7] for 5080-0 Al-alloy. The effect of aging time and temperature on the mechanical properties and microstructure of a Duraluminum-type alloy has also been studied[8] and it was shown that K_{IC} values have an inverse relationship to proof stress values and minimum K_{IC} values were obtained from peak-aged specimens. Extensive research on the effect of porosity, chemical composition, and cavity effect on tensile properties is available in the literature.[9-11]

Saigal,[12] using two-phase aluminum-silicon alloys, studied the interaction of silicon particles with the aluminum matrix

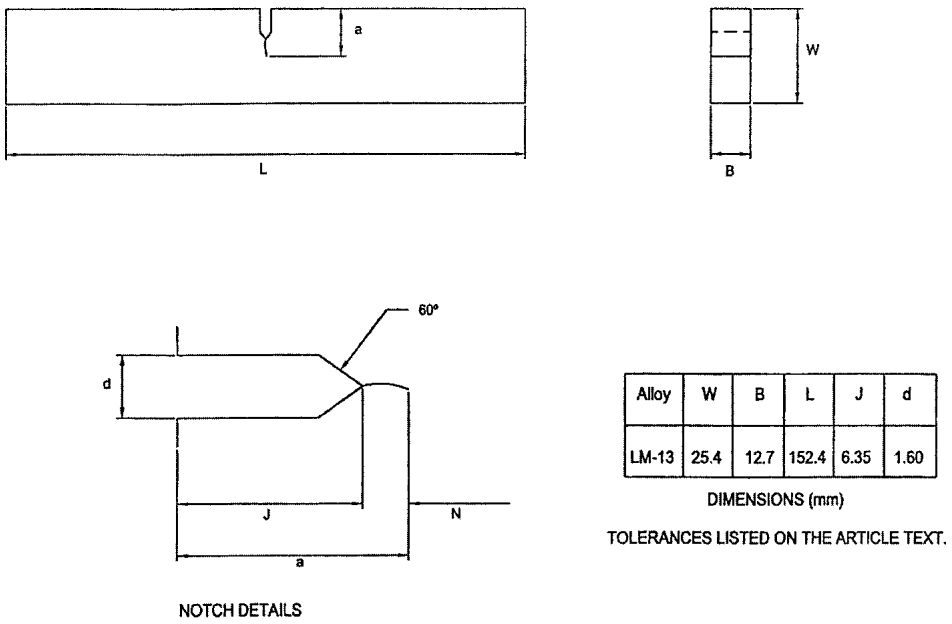

Fig. 1 Dimensions of samples used for three point bending test

using a finite element method. He concluded that silicon particle size is one of the most important microstructural parameters controlling the bulk mechanical properties. On the other hand MacAllister,[13] studying the effects of cooling rates on the mechanical properties of A206.0-T4 and A206.0-T71 aluminum alloy, reported that grain size was a function of cooling rate, and noted that samples showing the smallest grain size produced the highest mechanical properties.

In the same trend as other researchers, we believe that the mechanical properties of cast products are not only affected by cooling rates but also by many other variables such as their chemical composition, heat treatment, principal constitutive phases, dendrite arm spacing, and the size and shape of second phase particles. However, application of fracture mechanics criteria to evaluate the effect of solidification times on fracture toughness and microstructure has not been reported. Accordingly, in this work, several specimens of an Al-Si-Mg-Cu alloy were melted and allowed to solidify under different regimens, aiming to study the effect of microstructure formed during solidification on both the tensile strength and the COD.

2. Experimental

2.1 Alloy Melting

A commercial Al-10.6Si, 0.9Mg, 0.9Cu aluminum alloy was used in this study. The alloy melting for all the samples was carried out in an electrical resistance furnace, with no treatments such as degassing, modification, or nucleation. The pouring temperature was kept constant at 720 °C for all of the experiments.

2.2 Casting

A copper mold was used to produce a set of 17 rectangular bars of 15.9 × 31.8 × 171.6 mm aluminum alloy samples. The

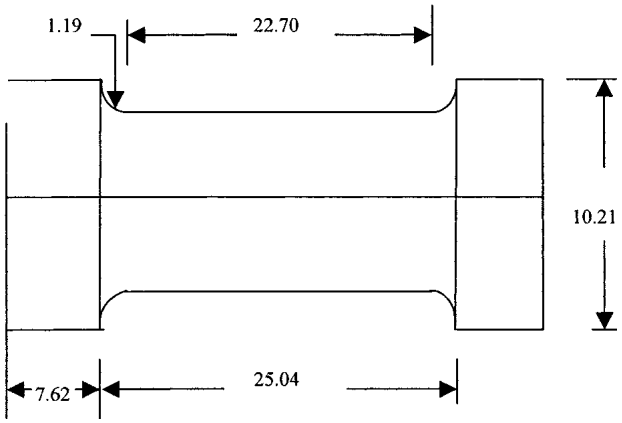

Fig. 2 Dimensions of samples used for tensile testing

mold was pre-heated at different temperatures to achieve different solidification times; the melts were then poured into it, and allowed to solidify. These samples were then heat treated and machined to obtain samples adequate for fracture mechanic studies. Sample cooling curves were attained by producing time-temperature plots. To measure the temperature data, a chromel-alumel thermocouple was placed 2 cm above the bottom of the mold. The temperature was recorded by using a digital display device (±5 °C accuracy). Sample preparation and fracture toughness testing.

Samples were prepared according to the procedures recommended for fracture toughness sample machining and testing as described in the specialized literature. Specifically, ASTM[14,18] and BISRA[15] recommendations were closely followed in this study. The testing technique involves three-point bending and tensile testing on notched fatigue pre-cracked sample. The aim of the testing is to establish the critical displacement value of the crack opening tip while a load is being

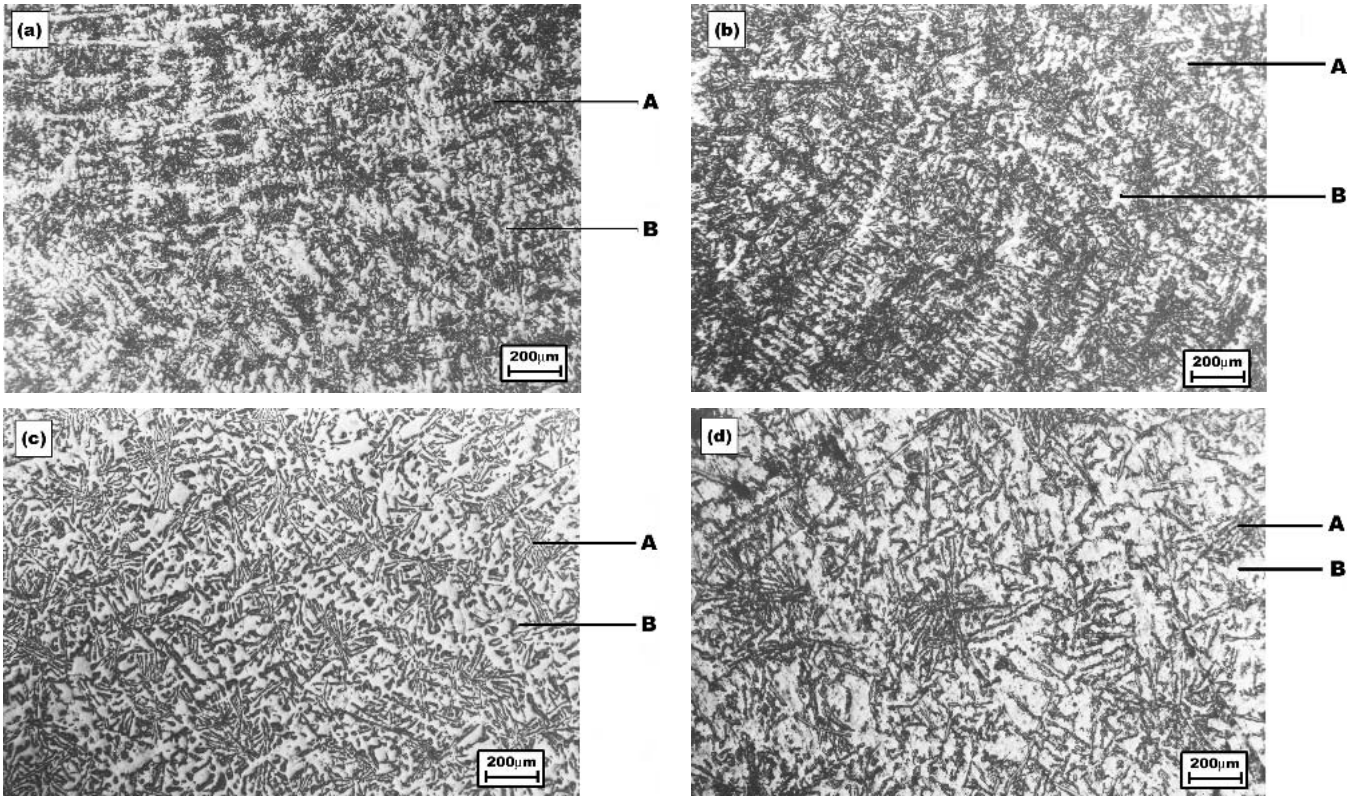

Fig. 3 Optical micrographs of polished sections of samples, at different solidification times: **(a)** 1.4 min; **(b)** 2 min; **(c)** 3.1 min; **(d)** 8.7 min. A: silicon phase (black), B: α-aluminum phase (white)

Table 1 Summary of Sample and Properties Results

Sample	Silicon Particle Size, μm	Crack Opening Displacement δ, mm	Crack Opening Displacement Average δ_m, mm	Ultimate Tensile Strength, N/mm²	T mold, °C	Solidification Time, min
1	10-20	0.0254				
2		0.0218	0.0233	213.9	50	1.2
3		0.0228				
5	10-20	0.0254				
6		0.0228	0.0241	184	100	1.4
0-1	20-40	0.033		N/A		
0-2		0.0254	0.0283		150	1.3
0-3		0.0266				
7	20-40	0.033				
8		0.0238	0.0282	172.3	200	2.0
9		0.0279				
0-4	40-60	0.0317	0.0317	N/A	250	2.5
10	60-100	0.0254				
12		0.0228	0.0241	179.1	300	3.1
13	120-180	0.0279				
14		0.0304	0.0296	138.9	400	8.7
15		0.0304				

applied. Automatic recording of load versus displacement is achieved through a strain gauge located at the edge of the crack notch.

The 17 bars were heat treated at 520 °C for 8 h followed by quenching in hot water, and precipitation treatment at 180 °C for 6 h and air cooled. The bars were then machined for fracture mechanics testing samples. Details of sample dimensions are schematically shown in Fig. 1. Sample fatigue pre-cracks were produced by using an Amsler vibrophone machine according to COD[16] recommendations. Fracture toughness testing was done using a 98.1 KN (10 t) Electromechanical Instron tensile testing machine according to standard recommendations.

2.3 Tensile Testing

From the heat treated COD samples (after the COD testing was over), cylindrical tensile test specimens 6.43 mm. in di-

ameter (Fig. 2) were machined and tested at room temperature in the electromechanical machine described previously, at constant crosshead displacement speed of 1 mm/min (0.0166 mm/s).

2.4 Metallographic Samples

All tested samples were metallography examined after being sectioned from the heat-treated bars, polished, and 0.1% HF etched, according to the standard microscopy optical procedures.

3. Results and Discussion

Size differences in silicon rich phases within an α-aluminum matrix related to the cooling times can be observed in Fig. 3, which only shows representative micrographs.

Differences in silicon size particles through the different solidification times were detected (Table 1). Also, significant differences in the precipitated silicon morphology were observed as a function of the solidification time. Small particles of silicon, interdendritically agglomerated, were typical of short solidification times (Fig. 3a and b); in contrast, at slow solidification times, larger silicon particles agglomerated into more "rosette-like" type of precipitates (Fig. 3c and d). Clearly, the longer the solidification time, the coarser the silicon precipitates.

The crack opening displacement, at first attainment of maximum load, the ultimate tensile strength (UTS), the mold temperature (°C), and the solidification time, along with the corresponding silicon particle sizing are summarized in Table 1.

To calculate δm (crack opening displacement) from data obtained from three point samples the following formula was used.

$$\delta_m = \frac{V_n(W-a)}{W+2a+3z} \quad \text{(Eq 3)}$$

Where *Vn* is the clip gauge displacement at first attainment of maximum load, *a* is the crack length (notch plus fatigue crack), *W* is the sample width, and *z* is the distance between gauge clips and sample surface.

The obtained trend of δm as a function of pre-heating mold temperatures is graphically shown in Fig. 4. Figure 5 also

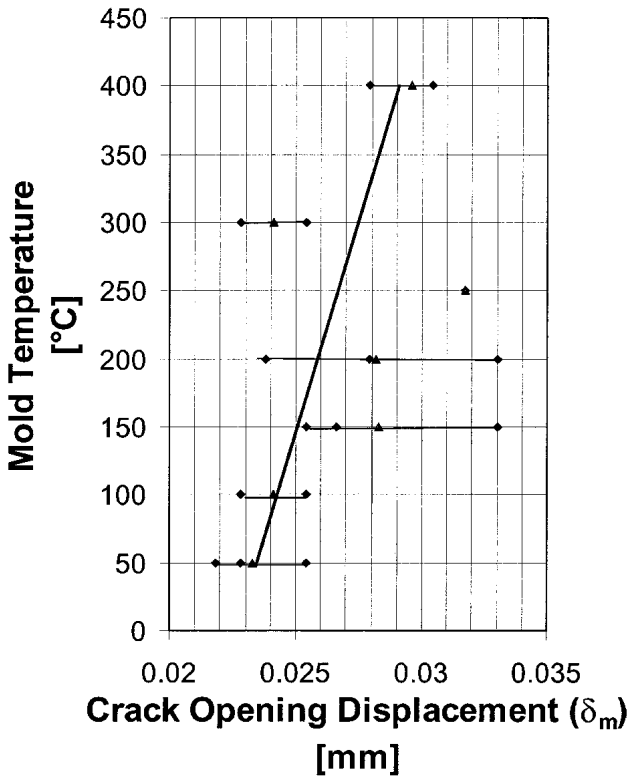

Fig. 4 Trend of the effect of mold temperature on the crack opening displacement

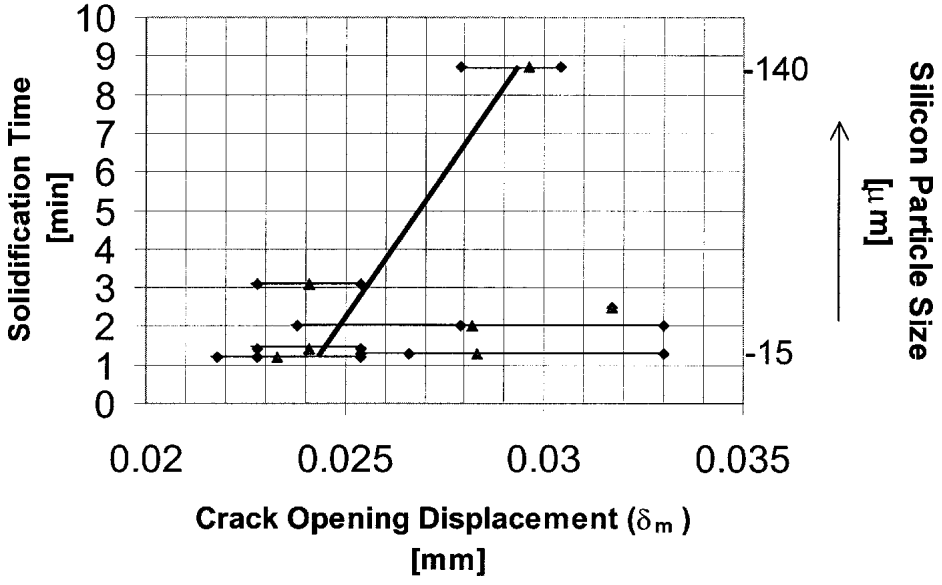

Fig. 5 Trend of the solidification time with the rendered silicon particle size as a function of crack opening displacement

shows the trend of the solidification time with different silicon particles sizes as a function of δm. In these cases, the slower the solidification time (or the higher the mold temperature), the higher the crack opening displacement. Note that larger silicon particles rendered larger crack opening displacements.

Figure 6 shows that the crack opening displacement tends to increase as UTS decreases, while Fig. 7 indicates that the UTS increases as the solidification time decreases. Figure 4, 5, and 6 are the result of plotting the average values with all scatter points. Note a wider scatter in COD values at intermediate mold temperatures and solidification times compared with both, the two fastest solidifying and the slowest to solidify data points. Results showing a low scatter showed a better fit between δm and solidification time or mold temperature (Fig. 8).

The UTS values of samples from mold temperatures of 150° and 250 °C were not possible to be calculated because these samples broke outside the standard ASTM requirements.

An interesting result is the possibility of measuring the value of δcrit for a material by using a small test-piece, which breaks at general yield. δcrit is the critical crack opening displacement and can be calculated under one of the following conditions: (1) crack opening displacement at fracture; (2)

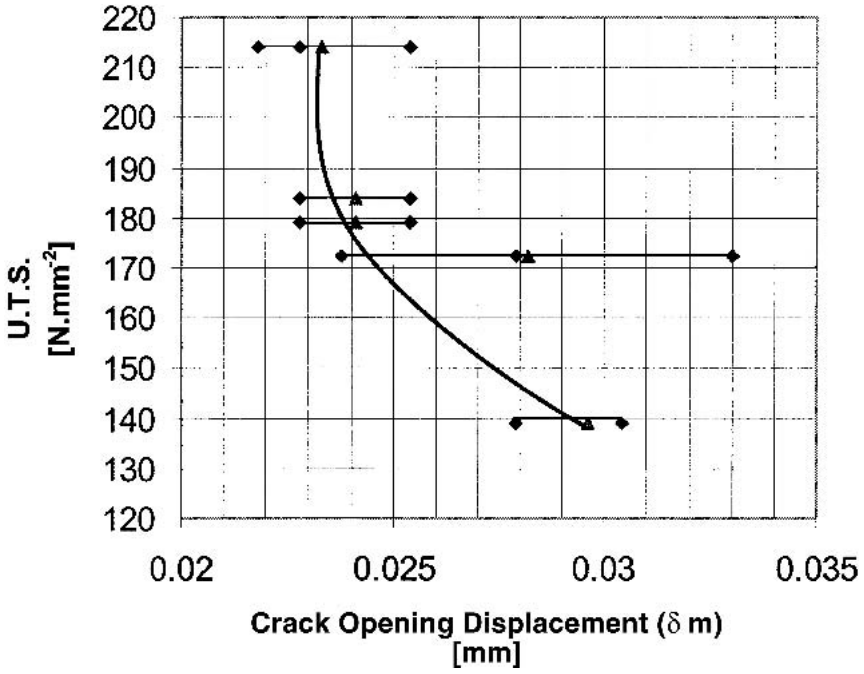

Fig. 6 Trend of the UTS as a function of crack opening displacement

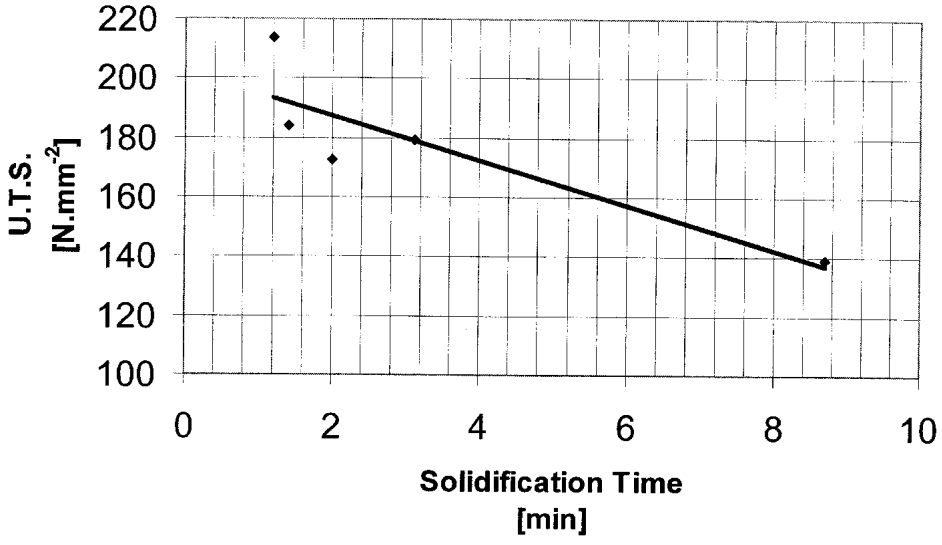

Fig. 7 Effect of UTS as a function of solidification time

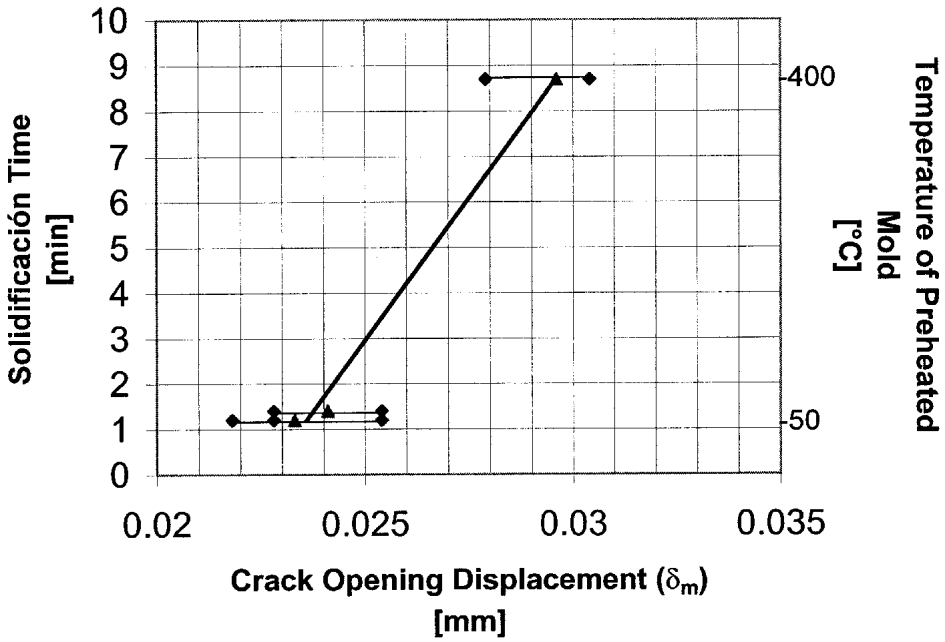

Fig. 8 Crack opening displacement as a function of both fastest and slowest solidification times, and different pre-heating mold temperatures

crack opening displacement at first instability or discontinuity; and (3) crack opening displacement at which an amount of stable crack growth commences. When the applied load/displacement curve reaches a maximum point, followed by further displacement with little or no falling applied load, stable crack growth is occurring. The critical displacement required is that maximum value at the point at which the stable crack growth started.[16]

References 16 and 18 provide a suitable method for measuring the onset of crack growth. In this work, COD for crack initiation could not be calculated, but for comparison, a crack opening displacement (δm) calculated from the clip gauge displacement (Vn) at the first attainment of a maximum load can then be used. Even though these values could provide a rough approximation of the failure stress of the samples to be compared, using Eq 2, for a component that could break prior to general yield, the δcrit must be calculated.

Thus, δm could then provide a rating of samples relative toughness at a given temperature. Previous results[12,13] indicate that fracture in ductile materials can initiate long before the maximum load in COD testing is reached.

In this work, molding conditions affected fracture toughness of the alloy under study. Increasing the pre-heating temperatures and increasing the solidification times (longer cooling times) will render larger dendrite size microstructure.[17] This growth in dendrite size is obviously due to diffusional aspects of the solidification itself. Furthermore, for long solidification times, the silicon crystals would tend to agglomerate, impairing the mechanical properties of the casting. The overall result is that dendrites with large silicon crystals are detrimental to tensile properties, as noted in other works,[13] and consequently would have a definitive effect on the fracture toughness. Clearly, large silicon particles, agglomerated as rosettes, act as stress concentration sites, which affect the fracture toughness performance of the alloy.

This plasticity effect was so notorious in our experiments that the stress intensity factor K_{IC} was not possible to be determined. This agrees with previous reports.[5]

4. Conclusions

The mechanical properties of an AlSiMgCu cast alloy are affected by the pre-heating of the casting molds. The higher the pre-heating temperature, or the longer the solidification times, the poorer the mechanical behavior of the alloy. One of the factors to explain this behavior is undoubtedly related to the shape and size of the silicon crystals and the α-aluminum matrix. The size of the silicon crystals increases markedly as the mold temperature and the solidification time increase, affecting the fracture toughness properties of the casted material.

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