High Temperature Tensile Deformation Behavior and Failure Mechanisms of an Al-Si-Cu-Mg Cast Alloy – The Microstructural Scale Effect

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Abstract

In this study the high temperature tensile deformation behavior of a commercial Al-Si-Cu-Mg cast alloy was investigated. The alloy was cast with two different cooling rates resulted in average secondary dendrite arm spacing of 10 and 25 μm , which are typical of the microstructure scale obtained from high pressure die casting and gravity die casting. Tensile tests were performed at different strain rates (10^{-4} s^{-1} to 10^{-1} s^{-1}) and over a wide temperature range from ambient temperature to 500 °C. The fine microstructure had superior tensile strength and ductility compared to the coarse microstructure at any given temperature. The coarse microstructure showed brittle fracture up to 300 °C; the fracture mode in the fine microstructure was fully ductile above 200 °C. The fraction of damaged particles was increased by raising the temperature and/or by microstructure coarsening. Cracks arising from damaged particles in the coarse microstructure were linked in a transgranular-dominated fashion even at 500 °C. However, in the fine microstructure alloy the inter-dendritic fracture path was more prevalent. When the temperature was raised to 300 °C, the concentration of alloying elements in the dendrites changed. The dissolution rates of Cu- and Mg-bearing phases were higher in the fine microstructure.

Keywords: EN AC-46000 cast alloy, elevated temperatures, deformation behavior, coarseness of microstructure

1 Introduction

Research on Al-Si based casting alloys has been re-focused as a lightweight structural material for automobile applications, because of their superior specific strength, encouraging mechanical properties, cost-efficient manufacturing and easy recyclability. The Si addition promotes castability and enhances strength, and Cu and Mg are commonly added to improve strength at room and elevated temperatures. Fe additions are made to reduce the die soldering tendency [1]. The microstructure of these alloys contains α -Al dendrites as the main constituents, which are decorated with eutectic Si particles and many intermetallic phases such as Al_2Cu , Mg_2Si and various Fe-bearing phases. Depending on the geometry of the die cast components, the cooling rates vary with the local thickness of the component. This leads to variations in the scale of as-cast microstructural features such as Secondary Dendrite Arm Spacing (SDAS) and the size and morphology of second phase particles. A high cooling rate results in a low SDAS while a lower cooling rate yields a higher SDAS. This affects the plastic deformation behavior and mechanical properties of the castings, which are affected significantly by the scale of the microstructure [2].

The solute level (Mg, Cu and Si), quantity and type of precipitates significantly influence the hardening rate of aluminum alloys [3-5]. Si precipitates from the as-cast super saturated solid solution during aging at temperatures range of 200–250 °C [6]. High temperature exposure of Al-Si-Cu-Mg alloys may lead to changes in the concentrations of alloying elements and of vacancies in solid solution, and to spheroidization of Si particles [7]. The dissolution rate of the various secondary phases is dependent on time and temperature of the exposure, the chemical composition and coarseness of the microstructure.

Unlike many engineering components of Al-Si-Cu-Mg alloys which are designed to be loaded within their elastic range, in several applications, particularly motor components working at elevated temperatures such as cylinder blocks, cylinder heads, pistons and valve lifters, the alloy undergoes plastic deformation. The plastic deformation can occur at different rates of strain depending on service temperature, exposure time and magnitude of the applied load [8]. The influence of strain rate (SR) has been widely studied in wrought materials in order to optimize production rate and processing

temperature during thermo-mechanical processing. Unlike wrought aluminum alloys, experimental data on the effect of strain rate on the deformation behavior of Al-Si based cast alloys is limited.

The effect of SDAS, size and morphology of various microstructural constituents on the tensile properties of Al-Si-Cu-Mg alloys has been thoroughly studied at room temperature [9-12]. Rincon et al. [13, 14] investigated the tensile and deformation behavior of A319 die cast alloy, in both as-cast and heat treated conditions at elevated temperatures up to 400 °C. Several researchers studied the effect of trace element additions on the strength improvement of Al-Si-Cu-Mg [15-18], Al-Si-Mg [19], Al-Si-Cu [20] and Al-Si alloys [21] at elevated temperatures. However an understanding of the role of cooling rate, which determines the scale of various microstructural features, on the high temperature behavior of Al-Si based casting alloys is still lacking. Most published works on Al-Si-Cu-Mg casting alloys have been focused on the room temperature tensile properties. As yet, the role of strengthening mechanisms at elevated temperature has not thoroughly been considered in these alloys. In addition, it is expected that at elevated temperature, the scale of microstructure and strain rate will influence the mechanical behaviour and failure mechanisms. The objective of this work is therefore to study the tensile behaviour, microstructural evolution and failure mechanisms of EN AC-46000 (a typical Al-Si-Cu-Mg casting alloy) by investigating the effect of temperature, microstructural scale and strain rate.

2 Experimental procedures

2.1 Materials

The material used was EN AC-46000 aluminum alloy, containing (in wt. %) 10.0 Si, 2.6 Cu, 0.24 Mg, 0.8 Fe, 0.8 Zn and 0.26 Mn. Ingots of the alloy were melted in a 10 kW resistance furnace at 730°C in a silicon carbide bonded graphite crucible. Cylindrical rods (length 20 cm, diameter 1 cm) were cast in a permanent copper mold with a thin graphite coating. The cast rods were then re-melted and heated to 710 °C for 20 minutes under Ar-atmosphere and subsequently solidified using the gradient solidification technique [2]. The gradient solidification set-up enables the production of well-fed and homogenous samples, with low levels of oxides, solidification shrinkage and gas-porosity over the entire length of the sample. The pulling rate of the furnace was used to control the scale of the microstructure and defect-free samples were produced with a microstructure scale similar to high pressure die casting (HPDC) and gravity die casting. The pulling rate of the furnace was set to 3 and 0.3 mms⁻¹, which leads to average SDAS of 10 and 25 µm, respectively.

2.2 Tensile testing

Cylindrical tensile test bars according to ASTM B557M-10 [22] of 6 mm in diameter and 100 mm in length were prepared by machining the as-cast rods. Tensile tests were carried out using a Zwick/Roell Z100 tensile testing machine, at six different temperatures ranging from 25 °C to 500 °C. Four strain rates in the range from 10⁻⁴ to 10⁻¹ were used. The highest strain rate, 10⁻¹, was selected due to the intrinsic limited ductility of the alloy, especially at temperature below 0.5 T_m. The lowest strain rate, 10⁻⁴, was selected to avoid creep-like deformation of the alloy. A wide test temperature range from room temperature up to slightly below the solidus temperature was chosen to allow using the experimental data to define a deformation behavior model of the alloy in a separate work. Stress-strain curves, the maximum stress value (R_m), the stress at 0.2 % offset strain (R_{p0.2}) and the strain at which failure occurs (ε_F) were obtained from the data acquisition system of the machine. In order to have statistical significance in the results, at least four replicates for each case were performed. Prior to tensile testing, the specimens were heated to the pre-set temperature and held for 15 minutes to homogenize.

2.3 Microstructural analysis

The microstructures and fracture surfaces were studied using optical microscopy, (OLYMPUS GX71) and a scanning electron microscope (SEM, JEOL7001F) equipped with an energy dispersive spectrometer (EDS). In order to study concentration profiles and micro segregation of the alloy elements (Mg, Cu and Si) across the dendrite arms, an SEM equipped with wavelength dispersive X-ray spectroscopy (WDS) was employed. The WDS samples were heated to given temperatures, held for 15 min. and quenched immediately in 40 °C water. Three analysis points were measured across a single dendrite arm, and at least nine dendrites were measured for each sample. Image analysis was used for qualitative study of microstructure evolution at elevated temperature. In order to study the evolution of microstructural features at

elevated temperatures a region in the center of a prepared specimen was identified by micro-indentation. The specimen was then heated to temperatures in the range from 300 to 500 °C, held for 15 minutes and studied after each heating without any additional surface preparation. Grain size determination was made through electron backscattering diffraction analysis to (i) ensure that there were no texture effects from the sample manufacturing, and (ii) to measure size of the grains. In order to characterize the failure mechanisms, the broken tensile specimens were mounted with the exposed cross-section parallel to the tensile axis and ground and polished to expose the center of the cylindrical bar. The fraction of fractured particles was examined by optical microscopy and quantitatively assessed using the Stream Motion image analyzer. A rectangular area 5×1 mm on both sides of the fractured surface, (not closer than 0.1 mm to the fracture surface) was measured. No attempt was made in the quantitative analysis to distinguish the type, size and orientation of fractured particles. Differential scanning calorimeter (DSC) analyses were carried out in a purified argon atmosphere using a NETZSCH 404C Pegasus® instrument with the scanning rate of 5°C/min in order to study any non-equilibrium reactions which may occur during heating.

3 Results and discussion

3.1 Microstructural characterization

3.1.1 The scale and morphology of microstructural features

Figure 1 shows the microstructural features of the EN AC-46000 alloy as-cast at two different cooling rates. The cooling rate governs size, morphology and distribution of the eutectic Si particles and intermetallic compounds [23, 24]. The microstructure consists of primary α -Al dendrites surrounded by eutectic Si particles and intermetallic phases. The intermetallic phases identified (Al_2Cu , α - $\text{Al}_8\text{Fe}_2\text{Si}$, β - Al_5FeSi) are indicated in Figure 1, which illustrates the effect of cooling rate on the scale of the microstructure. Increasing the solidification rate refines all microstructural features, decreases the SDAS and changes the morphology of eutectic silicon from large and elongated plates to small and rounder particles. This is supported by quantitative measurements of eutectic particle aspect ratio, see Table 1.

Figure 1. a) Intermetallic phases in as-cast microstructure identified through EDS analysis. Optical photomicrograph showing the effect of cooling rate on the SDAS, size, shape and distribution of eutectic Si in the alloy; b) SDAS 10 μm , c) SDAS 25 μm

Table 1. Size and morphology of the microstructural features for high and low cooling rates.

Furnace pulling rate (mm/s)	SDAS (μm)	Grain Size (mm)	Si particles, aspect ratio	Al_2Cu , area fraction (%)	Al_3FeSi , length of largest needle (μm)
0.3	25.9 ± 3.4	2.10 ± 0.36	3.91 ± 0.69	1.9 ± 0.5	76
3	10.1 ± 2.3	1.19 ± 0.28	1.73 ± 0.25	2.1 ± 0.3	18

3.1.2 Cu- and Mg-bearing phases

The change in microstructure after heat treatment (15 mins.) at various temperatures for both cooling rates is shown in Figure 2. It can be seen that increasing the temperature results in dissolution of the Cu-containing particles (mostly Al_2Cu) formed during solidification. A final polishing step using colloidal silica (OP-S) was used to enhance the Cu-bearing phases, turning their color to one that is readily detectable.

Figure 2. Micrographs after heat treatment at various temperatures for fine (SDAS 10 μm) and coarse (SDAS 25 μm) microstructures.

The Al_2Cu particles indicated in the 10 μm SDAS samples are completely dissolved after heating to 400 °C. In the as-cast condition, the Al_2Cu phase is present in the form of both a eutectic phase and a block-like phase. It has been reported that the former dissolves by fragmentation to smaller segments and eventually dissolves by radial diffusion of Cu in the matrix, the latter however dissolves by spheroidization and shrinkage without fragmentation [25]. These two types of

Al_2Cu phase are indicated in the coarse microstructure, Figure 2. In order to understand the dissolution of (Cu- and Mg-bearing) phases at elevated temperature, quantitative measurements of solute concentrations were made. Figure 3 shows the concentration of Cu and Mg in the center of the dendrite arms.

Figure 3. Solute concentration a) Cu and b) Mg at dendrite center measured after exposure to high temperature for SDAS = 10 and 25 μm . The error bars, which are not applied, are just as small as the points. Error bars calculated based on 95% confidence interval. This applies to all of the figures.

The increase in Cu concentration in the dendrite center, illustrated in Figure 3(a) could be a consequence of dissolution of Al_2Cu phase (the most abundant intermetallic) or fragmentation of other Cu-bearing phases such as $\text{Q-Al}_5\text{Mg}_8\text{Si}_6\text{Cu}_2$, which is noted in other studies [7, 26]. A large fraction of the Al_2Cu phases in the 10 μm microstructure dissolved at 500 $^\circ\text{C}$, and the solute concentration reaches 2.22 wt. %. The blocky Al_2Cu particle in the 25 μm microstructure remained stable even after exposure at 500 $^\circ\text{C}$, suggesting a lower dissolution rate, which is consistent with the lower concentration of Cu (Figure 3a) compared to the fine microstructure.

The Mg-richest compound, Mg_2Si was rarely found in the micrographs due to the low Mg content in the alloy (0.24 wt. %). Nevertheless Mg could be present in other intermetallic compounds such as $\pi\text{-(Al}_8\text{Mg}_3\text{FeSi}_6)$. Figure 3(b) shows the Mg concentration in the center of the dendrite arms for both SDAS after high temperature exposure. Increased Mg concentration in the dendrite center for both scales of microstructure is thought to be due to dissolution of Mg_2Si and also transformation of $\pi\text{-(Al}_8\text{Mg}_3\text{FeSi}_6)$ to $\beta\text{-(Al}_5\text{FeSi)}$ and Mg in solid solution [26]. The Mg concentration in the fine microstructure dendrites increases from 0.015 wt. % in the as-cast condition to around 0.9 wt. % at 400 $^\circ\text{C}$. Figure 3 shows that dissolution of Mg-bearing phases is more pronounced at temperatures above 400 $^\circ\text{C}$. In the fine microstructure, a large fraction of the available Mg content was transferred to the matrix in the form of a solid solution within 15 min at 500 $^\circ\text{C}$. The rate of Mg concentration increase was slower for the coarse compared to the fine microstructure as a result of slower diffusion kinetics caused by the longer diffusion distance [26]. It is therefore shown that the rate of dissolution of Cu- and Mg-bearing phases depends on exposure temperature and the size of those phases.

3.1.3 Eutectic Si

The morphology of eutectic Si particles at elevated temperature is shown Figure 2. During heating, Si particles spheroidize and then coarsen. The change in Si concentration (see Figure 4) suggests a rapid spheroidization rate for fine Si particles (SDAS 10 μm) above 400 $^\circ\text{C}$. The coarse plate-like Si particles in the 25 μm SDAS microstructure however are less responsive to spheroidization, and demand longer exposure times at a given temperature to become spheroidized. Si particles showed almost no change in their internal architecture below 300 $^\circ\text{C}$. Coarse Si particles resist coarsening up to 400 $^\circ\text{C}$. Figure 4 shows that the Si concentration at the center of a dendrite arm at room temperature is lower for the fine

microstructure rather than for the coarse one. Increasing the solidification rate reduces the time during which solid state diffusion state may take place which results in a lower concentration of Si in solid solution [27]. The marginally higher concentration of Si in the dendrite of SDAS 25 μm (see Figure 4) may be due to that reason.

Figure 4. Si concentration in the center of dendrite arms measured after different exposure temperatures for SDAS 10 and 25 μm .

According to the Al-Si equilibrium phase diagram [28], Si has a maximum solid solubility in Al of 1.6 wt. % at 576 °C, which drops to 0.8 and 0.2 wt. % at 500 °C and 400 °C respectively. Coarsening of eutectic Si as a result of increased temperature is due to diffusion-controlled reduction of Si in solid solution in the vicinity of the eutectic cells and migration to pure Si particles. The concentration of Si in the center of small dendrite arms (SDAS 10 μm) drops dramatically above 400 °C, as Si is gradually depleted from solid solution and contributes to coarsening of nearby eutectic Si particles, see Figure 4. Coarse dendrite arms (SDAS 25 μm), however experienced no meaningful change in Si concentration. The larger diffusion distance in 25 μm SDAS is presumed to hinder solute diffusion from the matrix to the particles; and from small to large Si particles. The spheroidization of Si by solution heat treatment is one of the steps of post-solidification processing Al-Si based cast alloys, and hence has been thoroughly discussed in that respect. Nevertheless, it is useful to note that the findings in this work are in agreement with previous studies [26, 29].

3.2 Mechanical Properties

It is well-known that casting parameters (melt cleaning, oxide films, pores etc.) have marked effects on the tensile properties of Al-Si casting alloys at room and at elevated temperatures [14]. The presence of defects is associated with significantly degraded tensile ductility which consequently shortens the plastic strain region. Such defects prevent the direct determination of stress-strain relations in the plastic region and also result in significant scattering in measured tensile properties [30]. In order to avoid this problem the levels of these defects have been minimized by using the directional solidification technique. Any effects of casting defects have therefore been neglected. Figure 5 shows the R_m , $R_{p0.2}$ and ϵ_F as a function of temperature for samples with different SDAS. There is a general tendency toward decreasing R_m and $R_{p0.2}$ with increasing temperature due to facilitated dislocation movement [13-15, 31, 32]. The movement of dislocations through a lattice is aided by thermal vibrations. Consequently, resistance to dislocation slip increases as the temperature falls. In Al alloys, room temperature falls in a range between $0.2T_m$ and $0.5T_m$, where the deformation mechanism is usually slip [33]. Figure 5 shows that at temperatures below 300 °C ($\sim 0.5T_m$), R_m slightly decreases with increasing temperature. However at 300 °C and above the strength of the alloy degrades steeply. As the temperature increases, the potential for dislocation climb, and dislocation slip increases [33]. The fine microstructure (SDAS 10 μm) displayed superior ultimate tensile strength compared to the coarse microstructure (25 μm SDAS) at RT. This supremacy

was retained at temperatures up to 500 °C. It was also observed that elongation to failure (ϵ_F) significantly increases at temperatures above 200 °C for both coarseness of microstructure. However, the fine microstructure always showed higher ϵ_F values at any given temperature, Figure 5, and the difference increased with increasing temperature.

The contribution of solute level on $R_{0.2}$ of the alloy at room temperature has been well documented. Higher amount of solute

Figure 5. R_m , $R_{p0.2}$ and ϵ_F values of EN AC-46000 alloy with different SDAS at different temperatures. Strain rate is constant and equal to 10^{-3} s^{-1} .

results in increased yield strength in Al-Si-Cu-Mg alloys [7, 34]. It was found that $R_{0.2}$ is marginally greater for the case of SDAS 25 μm rather than those in SDAS 10 μm , see Figure 5. Olofsson et al. [2] also reported similar observations comparing SDAS 10 and 25 μm . Figure 3 and Figure 4 show the concentration of solutes in the coarse microstructure is slightly greater in the fine microstructure, which could explain the increase in $R_{0.2}$. In other works, the aspect ratio of Si-particles [23] and size of the eutectic cell [35] were suggested to make a significant contribution to the strength of Al-Si alloys in the early strain regime (<0.003). The contribution of those features however is not pronounced according to the findings of this study. Figure 5 shows that the marked difference in $R_{0.2}$ fades at 300 °C and above. Neither morphology of Si-particles nor the eutectic cell size showed any indicative change up to 400 °C. Nevertheless, 300 °C is the temperature where the significant

change in solute concentration profile initiated, see Figure 3. It is therefore suggested that the yield strength is solely matrix dependent irrespective of the size of microstructural constituents.

Flow stress curves of the alloy at different temperatures for SDAS 10 μm are shown in Figure 6. This shows an increase in hardening rate at 200 $^{\circ}\text{C}$ in the low strain regime which is apparently due to precipitation hardening. A typical DSC trace for as-cast EN AC-46000 samples, scanned at a heating rate of 5 $^{\circ}\text{C}/\text{min}$ showed an exothermic peak with onset temperature of 206 and 215 $^{\circ}\text{C}$ for SDAS 10 and 25 μm respectively. The peak evidently corresponds to precipitation of Si from supersaturated solid solution [6]. According to the equilibrium phase diagram, Si has a maximum solubility of 1.6 wt. % at 576 $^{\circ}\text{C}$ in Al, which is lowered to 0.26 wt. % at 400 $^{\circ}\text{C}$. Results from WDS analysis indicated average concentrations of 1.45 and 1.65 wt. % Si in the α -Al matrix for fine and coarse microstructures respectively, Figure 4. An initial average composition of 1.55 wt. % supersaturated Si was considered as reasonable estimate in the as-cast condition. Increasing the temperature to 200 $^{\circ}\text{C}$ results in precipitation of Si from supersaturated solid solution and a concomitant increase in strength through precipitation hardening. In a work by Rincon et al. [13], a precipitation hardening behavior was observable in the flow stress curve at 240 $^{\circ}\text{C}$ in A319 cast alloy.

Figure 6. Flow curves corresponding to 10 μm SDAS at different temperatures. SR is equal to 10^{-3} s^{-1} .

Figure 7 shows the strain rate dependency of stress-strain behavior of the alloy at 400 $^{\circ}\text{C}$. The flow stress curves showed that the alloy became sensitive to strain rate at 300 $^{\circ}\text{C}$ and above irrespective of coarseness of microstructure. The strain rate sensitivity ($\text{SRS} = \delta \log(\sigma) / \delta \log(\text{SR})$) at constant strain (0.01) was calculated for both coarseness of microstructure. The value of 0.015 ± 0.004 was obtained for both SDAS 10 and 25 μm at 200 $^{\circ}\text{C}$ and below. The corresponding value rises to 0.12 and 0.11 at 400 $^{\circ}\text{C}$ for the fine and coarse microstructures respectively ($R^2 > 0.95$). Both coarseness of microstructure retain rather low strain rate sensitivity even at 400 $^{\circ}\text{C}$, which is plausible due to the relatively large grain size ($\sim 1\text{-}2 \text{ mm}$).

Figure 7. Flow curves corresponding to 25 μm SDAS at 400 $^{\circ}\text{C}$ obtained under different SR.

An increased strain rate leads to a higher strain hardening rate at 400 $^{\circ}\text{C}$ and consequently a postponed necking occurrence. Above 0.5 T_m , the dislocation density is determined by the balance between strain hardening and dynamic recovery. Once the rate of straining increases, a higher steady state dislocation density is obtained which leads to a larger saturation stress level [33]. The effect of strain rate disappears upon lowering the temperature due to the trifling rate of dynamic recovery compared to the rate of hardening.

3.3 Deformation and fracture behavior

As long as the level of defects (e.g. gas porosity and oxide films) in cast aluminum alloys is minimized the failure sequence consists of three consecutive steps; initiation of small damage in particles, damage growth and microcrack formation in the particles, and finally linkage of the microcracks or crack propagation [11, 30]. The failure sequence is illustrated in Figure 8.

Figure 8. Micrographs showing the failure sequence. a) Damage initiation in particles, b) damage growth and c) crack linkage. SDAS 25 μm , test temperature 100 $^{\circ}\text{C}$, SR 10^{-4}s^{-1} . The tensile axis is horizontal in all images.

In order to assess the role of microstructure coarseness, temperature and strain rate on the failure mechanism, the damage frequency (ratio of damaged/cracked particles over all particles) and area fraction of damaged/cracked particles in failed specimens was quantitatively measured using image analysis. Measurements were made at room temperature in the central

plane over the entire gauge length. Micrographs showed that the majority of cracked particles at any given condition were found in the region close to the fracture surface. Hence, the measurement in that region only was taken into account. Figure 9 shows the fraction of cracked particles as a function of temperature for the fine and coarse microstructures.

Figure 9. Fraction of damaged particles as of function of tensile test temperature.

It is evident that in the coarse microstructure increasing temperature leads to a monotonic increase in fraction of damaged particles, and the increase becomes significant above 300 °C. In the fine microstructure however, the fraction of damaged particles below 300 °C was quite limited (fraction lower than 0.02) and at 400 °C and beyond, the fraction increased significantly. Plastic flow of the alloy is governed by stress incompatibilities between plastically deformed matrix and elastically strained particles. Once plastic strain in the alloy starts, the brittle second phase particles begin to form microcracks. This continues until a critical strain in which a critical volume fraction of cracked particles is reached, which leads to micro crack coalescence and failure as the matrix between the fractured particles becomes unstable. The process is primarily controlled by microstructure and matrix-properties [11, 31]. As long as the mode of fracture is brittle (below 300 °C) particles dominate the microcracks linking process. As the matrix softens at higher temperatures, the matrix properties begin to play a role in the failure process. Raising the temperature may reduce the matrix instability between the fractured particles and microcracks. On the other hand, the stress induced in particles during matrix deformation will be intensified which leads to further particle damage and void formation. The end result is an increased fraction of fractured particles and increased strain. This trend was clearly seen in the coarse microstructure of this study (SDAS 25 μm in Figure 9). Rincon et al. [13] reported that increasing the temperature from 250 °C to 400 °C leads to an increase in fraction of cracked particles from 12 % to 50 % in an Al8Si3CuMg cast alloy having SDAS in the range 24-36 μm, which is in agreement with the findings of this work. Mahidhara & Mukherjee [31] also reported that as eutectic Si particles in a Al-Si eutectic alloy become coarser, the level of cavitation at the matrix-Si particle interface was significantly increased at 558 °C.

Inter-particle cracking is a function of size, aspect ratio and type of particles as well their orientation with respect to the direction of applied load. Since larger defects and voids are allowed to exist in larger particles, they have a lower fracture stress and in turn they fracture promptly, as suggested by Wang et al and Hunt [12, 36]. In the microstructure with SDAS 25 μm, damage initiates more readily in elongated particles rather than in particles having a smaller aspect ratio. Particles in the fine microstructure (SDAS 10 μm) were however less prone to damage, as they have a smaller and less scattered aspect ratio, see Table 1. This is supported by the fact that higher aspect ratio particles accumulate stress at a faster rate and accordingly tend to crack earlier, as suggested by Wang [23]. It was observed that the fraction of damaged Al₂Cu and β-Al₅FeSi particles was higher than the fraction of damaged Si particles. This was also noticed in another study [13]. However, in the present study no quantitative measurement was done to distinguish secondary particles according to type.

Figure 10 shows SEM micrographs of the fracture surface of the two microstructures which were strained to failure at room temperature, 300 and 400 °C. It is obvious that the coarse microstructure experienced brittle fracture at 300 °C and below due to the presence of large plate-like brittle phases. Rapid fracture of coarse particles and intergranular propagation of microcracks shield the matrix from plastic deformation (Figure 10 d and e). Lack of plasticity (<5 %) suggests that the cast alloy with SDAS 25 μm is intrinsically brittle below 300 °C.

Figure 10. SEM micrographs of fractured surfaces of the alloy with SDAS 10 and 25 μm tested at different temperatures. SR is equal to 10^{-3} s^{-1} .

Figure 11 shows the tensile instability behavior of the alloy with fine and coarse microstructure at temperatures of 200 °C and 300 °C. Below 200 °C, the derivative of the curves does not meet the flow curves of the fine microstructure suggesting that brittle failure occurred at this temperature. The accumulation of damage in particles leading to failure takes place prior to necking. At 200 °C and above however, the instability criterion is met, indicating that 200 °C marks the brittle to ductile failure transition temperature. In the coarse microstructure, the instability criterion is satisfied at 300 °C, see Figure 11. At 400 °C the coarse microstructure shows the typical dimples of ductile fracture and multiple cracked particles in the fracture surface. In contrast, the fracture surface of the fine microstructure tested at room temperature contains ductile dimples, see Figure 10a. The number of dimples was significantly increased at 300 °C, proving the fully ductile fracture above 200 °C, which is a consistent finding based on instability criteria. In a similar study to this [14], it was found that A319 alloy softens in the interval of 270 to 320 °C, where SDAS ranged from 20-50 μm which agrees well the results for the case of SDAS 25 μm .

Figure 11. Tensile instability plots of the alloy at 200 and 300 °C; for SDAS 10 and 25 μm . SR is equal to 10^{-3} s^{-1} .

In a study by Wang [23] it is suggested that in alloys with large SDAS, the failure occurs prior to necking at RT as a result of the high damage accumulation rate caused by particle cracking. In small SDAS samples however, necking takes place below the fracture strain. The comparison was made between 17 and 54 μm SDAS in A356/357 alloy. It is therefore feasible that in the coarse microstructure damaged particles (Si, Al_2Cu and $\beta\text{-Al}_5\text{FeSi}$ etc.) cause premature failure but that this is deferred to a higher temperature (300 °C) when necking occurs due to increase matrix plasticity. In the fine microstructure however, post-necking damage through particle cracking apparently takes place at 200 °C. Results of tensile instability in the fine microstructure do not agree with the suggestion by Wang [23]. This contradiction may be due to the higher level of Si (10 wt. %) and the presence of Cu (2.6 wt. %) in the EN AC-46000 alloy, which limits the intrinsic ductility of the matrix. The tensile instability point of the fine and coarse microstructure approaches the same value at higher temperature (above 400 °C), indicating that necking occurrence is determined by stress relaxation during plastic deformation instead of damage accumulation in particles [14]. Two typical modes of damage accumulation found in the present study are single micro cracks and microcrack clusters, see Figure 12 (c) and (d).

Figure 12. Optical micrographs showing fracture path of failed specimen a) SDAS 10 μm , RT, 10^{-2} s^{-1} , b) SDAS 10 μm , RT, 10^{-2} s^{-1} ; c) single microcrack and d) microcracks clusters at RT, 10^{-2} s^{-1} .

It has been proposed that the rate of damage formation in single particles at different strain is approximately constant. The accumulation of damage in particles however is rare at low strain levels and becomes significant as the fracture strain is approached [36]. Figure 13 shows the area fraction of fractured particles and cavities as a function of strain rate at 500 °C for different microstructure coarseness.

Figure 13. 2D Area fraction of fractured particles and cavities at particle/matrix interface as function of strain rate at 500 °C. The corresponding elongation to failure and R_m are indicated for each case.

It was observed that at 500 °C increasing the strain rate from 10^{-3} to 10^{-1} s^{-1} profoundly increased the area fraction of fractured particles for the coarse microstructure. The trend was not as clear at 400 °C and no trend was found at lower temperature, nor was this trend evident for the fine microstructure. This suggests that the cleaving rate of the particles which are already fractured at lower strain levels is related to a thermally activated stress relaxation mechanism such as dislocation climb. Increasing the strain rate leads to a higher dislocation density [33] and accordingly an increased number of dislocations enhances the probability of interactions with particles. Micrographs suggest that an increase in the area fraction of damaged particles includes more frequent microcrack clusters and single cleaved particles. However, no meaningful trend was observed between the strain rate and elongation to failure at 500 °C (see Figure 13) or any other test temperatures. This is consistent with somewhat lower ductility of the alloy at higher strain rate (due to the increased probability of microcrack aggregation). The higher strain rate leads to instability in plastic deformation at higher strength and strain at elevated temperature (above 400 °C) see Figure 7. Extended local uniform plastic strain at high strain rates resulted in opening of microcracks and shattering of cracked particles without any pronounced initiation of damage in the particles. Localized shear bands formed in the matrix at the edge of microcracks (mostly inside the grains) provide an opportunity for fractured particles to grow [11]. Mahidhara & Mukherjee [31] report quite the opposite observation for an Al-13Si wrought alloy deformed at 557 °C under different strain rates. It was qualitatively observed that by reducing the strain rate from 10^{-2} to 10^{-5} s^{-1} , the numbers of cavities - particularly in grain boundaries - surprisingly increased and elongation to failure monotonically decreased. Their observation is however not consistent with the findings of the current study. Unlike the Al-13Si wrought alloy [31] in which the grain size ranged from 15-35 μm , grain boundary sliding is not a plausible deformation mechanism in the EN AC-46000 alloy which has a grain size of around 1-2 mm. Therefore, the higher rate of straining does not necessarily facilitate grain boundary sliding and consequently improved plasticity. Another reason which may be relevant is the inevitable uncertainties related to the mode of processing (e.g. casting defects). Since the samples are in the as-cast condition, their plasticity severely suffers from such defects.

4 Conclusions

The tensile properties and deformation behavior of EN AC-46000 alloy in the temperature range of room to 500 °C were studied. The microstructure comprised two distinctive coarseness, a) SDAS 10 µm and b) SDAS 25 µm. The strain rate during tensile testing was varied from 10^{-4} s^{-1} to 10^{-1} s^{-1} .

1- R_m , $R_{p0.2}$, and ϵ_F were largely dependent on temperature. Ductility increased surprisingly above 300 °C, even for the case of coarse microstructure. $R_{p0.2}$ showed no significant decrease from ambient temperature to 200 °C, however above that it decreased monotonically. R_m consistently decreased with increasing temperature for the case of SDAS 10 and 25 µm.

2- The strain rate effects on tensile behavior were not significant below 200 °C for both SDAS 10 and 25 µm. The effect of strain rate became more important at 300 °C where the flow stress of the alloy decreased dramatically. At this temperature and above, strain rate has a marked effect on R_m . The effect on the flow stress curve became even more highlighted at 400 °C and above. Four order of magnitude increase in strain rate at 500 °C (from 10^{-4} to 10^{-1} s^{-1}) resulted in a higher flow stress, even when the temperature was 100 °C lower (10^{-4} s^{-1} at 400 °C).

3- The concentration of Cu and Mg in the dendrites increases as the temperature exceeds 300 °C. The increase becomes significant above 400 °C. The rate of dissolution of Mg- and Cu-bearing phases is higher in the alloy with fine microstructure. Spheroidization of eutectic silicon particles occurred above 400 °C and the rate was greater in the fine microstructure alloy.

4- The brittleness of the alloy below 300 °C was related to accumulation of a high volume fraction damaged particles such as Cu- Fe-bearing phases and Si-particles. The initiation rate of damage in the coarse particles was significantly higher, which enhances the probability of failure and decreasing both R_m and ϵ_F compared to the fine microstructure.

5- The fraction of cracked particles was found to be a function of temperature. The coarse microstructure is more sensitive to crack formation at any given temperature. The area fraction of damaged particles was significantly increased at high strain rate (10^{-1} s^{-1}) at 500 °C. No meaningful relation between the fraction of damaged particles and strain rate was found at any temperature.

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