The influence of Mn on the tensile properties of SSM-HPDC Al-Cu-Mg-Ag alloy A201

by H. Möller*, E.P. Masuku*, U.A. Curle*, P.C. Pistorius†, R.D. Knutsen‡, and G. Govender*

Synopsis

A201 aluminium alloy is a high strength casting alloy with a nominal composition of Al-4.6Cu-0.3Mg-0.6Ag. It is strengthened by the Ω(Al_2Cu) phase and the θ'(Al_2Cu) phase during heat treatment. Further strengthening of this alloy system can be obtained through the addition of transition elements, but care must be taken as other elements might have adverse effects on the mechanical properties. The objective of this study is to determine the influence of Mn on the tensile properties of rheo-processed Al-Cu-Mg-Ag alloy A201. ThermoCalc software was used to predict the different phases that can be expected in the alloys, and scanning electron microscopy (SEM) with energy dispersive spectroscopy (EDS) was used to investigate the actual phases that formed. The influence of these phases on tensile properties is quantified. SEM and ThermoCalc revealed that there is an increased amount of the $Al_{20}Cu_2Mn_3$ with increasing Mn. The tensile properties showed that high amounts of Mn do have adverse effects on the tensile properties of alloy A201, especially the ductility.

Keywords
Semi-solid metal (SSM) forming, alloy A201, intermetallics, $Al_{20}Cu_2Mn_3$

Introduction

Semi-solid metal (SSM) processing is a manufacturing method capable of producing near-net shape products for a variety of industrial applications. A semi-solid structure (free of dendrites) is produced with the solid alloy particles present in a near spherical form. This semi-solid mixture flows homogeneously, behaving as a thixotropic fluid with viscosity depending on the shear rate and fraction of solid in the liquid, by either thixocasting or rheocasting. With thixocasting, a specially prepared billet of solid material with a globular microstructure is reheated into the semi-solid range, followed by a forming process, such as high pressure die casting (HPDC). With rheocasting, an SSM slurry is prepared directly from the liquid, followed by HPDC. The higher costs associated with thixocasting have resulted in rheocasting becoming the preferred semi-solid process[1]. Aluminium alloy A201 (with nominal chemical composition Al-4.6Cu-0.3Mg-0.6Ag) possesses the highest mechanical strength of all the aluminium casting alloys between room temperature and 200°C[2]. In Al-Cu alloys, the main strengthening precipitate is Al_2Cu (θ'). Small amounts of Ag change the precipitation process in these alloys, causing a form of Al_2Cu (referred to as Ω) to precipitate as thin plates on the {111} matrix planes (primary slip plane) rather than the {100} planes. This results in remarkable high tensile properties being obtained[3]. Elements such as Fe ($AlCu_2Fe$) have been reported in literature[2] to have an adverse effect on the mechanical properties of the alloy, as they tend to form phases on the grain boundaries which give a lower supersaturation of copper during artificial ageing and a decrease in strength in the T6 condition (solution heat treated, quenched and then artificially aged). Of particular interest in this study is Mn. It has been shown that, in conventional A201 castings at quantities of less than 0.5 wt%, it tends to contribute to the formation of the stable dispersion strengthening phase $Al_{20}Cu_2Mn_3$, which is known to aid in grain size control with little removal of Cu in the form of coarse intermetallics[4]. Dispersion strengthening is a means of strengthening alloys wherein small particles of usually less than 0.1 m of a hard, inert phase are uniformly dispersed within a load-bearing matrix phase[5]. Mn has the additional advantage of combining with Fe to form $Al_6(Mn,Fe)$[4]. At levels beyond a maximum of 0.5 wt%, it has been shown to lead to the formation of large fractions of coarse and brittle constituents which act as crack initiators

* *Materials Science and Manufacturing, CSIR, Pretoria, South Africa.*
† *Department of Materials Science and Engineering, Carnegie Mellon University, Pittsburgh, USA.*
‡ *UCT Centre for Materials Engineering, Cape Town, South Africa.*

and reduce the fracture toughness of the material[4]. The purpose of this study is to investigate the effects of Mn in SSM-processed A201.

Experimental work

In order to study the influence of Mn with the other elements being kept constant, a batch of alloy A201-A (composition given in Table I) was melted in a 20 kg resistance heated tilting furnace and degassed with Ar. Manganese was added according to a mass balance. A sample was poured each time and quenched to analyse the chemical composition by optical emission spectroscopy (Thermo Quantris OES). The CSIR-OES (Council for Scientific and Industrial Research) does not have the Al-Ag analysis database. As a result, a different accredited laboratory was used for the analysis of Ag. Table I shows the chemical composition obtained for each of the alloys. The thermodynamic properties of each alloy were then calculated with an aluminium thermodynamic database (ProCast 2009.1) from the specific OES compositions given in Table I. Stainless steel cups (~ 400 g) were used for processing in the rheocasting system. The mush metal was processed using the CSIR rheocasting process where induction stirring with simultaneous forced air cooling is employed[7]. Plates (6 mm × 55 mm × 100 mm) were cast in steel moulds with a 130 ton clamping force high pressure die casting (HPDC) machine (LK DCC130). ThermoCalc, a commercially available software package used to perform thermodynamic and phase diagram calculations for multi-component systems of practical importance, was used to investigate the possible effects of Mn levels on the equilibrium phases in the alloy, using the Al-DATA ver.2 database. Because of its absence from the available database, the element Ag was not included in the calculations.

The cast plates were heat treated to the traditional T6 condition that consisted of solution treatment at 513°C for 2 h, followed by 527°C for 17 h, quenching in water at ambient temperature and artificial ageing (AA) at 153°C for 20 h[8]. It has been shown by the authors that these solution treatments are not necessarily the optimum solution treatments for rheo-processed material[9]. The samples in the as-cast and heat treated conditions were X-rayed to determine the types of defects present. Subsize rectangular tensile test specimens were machined from the plates to final specimen dimensions with the nominal width of 6mm and gauge length of 25 mm according to ASTM standards and the tensile properties were evaluated (Instron Model 1342)[10]. A total of 9 tensile tests were done for each alloy. Scanning electron microscopy (SEM) with energy dispersive spectroscopy (EDS) at an acceleration voltage of 20 kV was used for microstructural characterization and to quantify the actual intermetallics that had formed.

Results and discussion

Chemical composition

The chemical compositions of the A201 alloys used in this study are shown in Table I, as well as the specification of alloy A201[6]. The difference in the copper content in all alloys is minimal; this simply means the volume fraction of the strengthening precipitates should be the same in all alloys. It has already been stated that Ag additions promote the formation of Ω rather than θ'[2,3]. It is also known that Mg is a critical component for the nucleation of Ω[11]. Magnesium-rich clusters are believed to act as heterogeneous nucleation sites for Ω plates. Silicon, on the other hand, hinders the precipitation and stability of Ω[12]. Trace additions of Si have also been shown to stimulate matrix precipitation of θ', S (Al_2CuMg) and $Al_5Cu_6Mg_2$, which are phases competing for solute. Ω precipitation is suppressed in alloys with Mg to Si ratios of < 2.0, regardless of Mg and Si content[11]. It is seen from Table I that the Mg to Si ratios of all the studied alloys were > 2.0 in this study. Ti plays a major role as grain refiner for the primary α-phase. The Mn content was purposely varied to a maximum of 1.0 (wt per cent) for the scope of the study.

ThermoCalc analysis

The calculated phase equilibria (minor phases) for the three alloys used in this study (Table I) are shown in Figure 1 (a–c). In all cases the major phases were liquid, Al-based FCC solid solution (the primary phase upon solidification), and Al_2Cu (formed by precipitation after solidification). However, of interest to this study is the formation of $Al_{20}Cu_2Mn_3$ phase. Figure 1 (a–c) shows that mass percentage of the $Al_{20}Cu_2Mn_3$ phase increases from ~1.2% in the low Mn alloy to around 4.5% in the high Mn alloy. In the two higher-Mn alloys (Figure 1 (b and c)), $Al_{20}Cu_2Mn_3$ is predicted to form during solidification. Some manganese is predicted to be taken up also by the 'alpha' phase, which is an Al-Mn-Fe-Si solid solution based on Al_8Fe_2Si. The other detrimental phases in alloy A201 predicted by ThermoCalc are Cu_2FeAl_7 and Mg_2Si. However, these are predicted to be present at mass percentages of less that 0.1% in all three alloys, thus their impact on the mechanical properties is expected to be minimal.

Microstructure

Scanning electron microscopy (SEM) coupled with energy dispersive X-ray spectroscopy (EDS) was used for microstructural investigation and to determine the compositions of the phases. Secondary electron images of the three alloys in T6 temper condition are shown in Figure

Table I

Chemical composition (wt%) of Al-Cu-Mg-Ag alloys (balance Al)

Alloy	Cu	Fe	Si	Mn	Mg	Ti	Ag
Specification[6]	4.0-5.2	<0.15	<0.1	0.2-0.5	0.15-0.55	0.15-0.35	0.40-1.0
A201-A	4.15	0.04	0.03	0.29	0.23	0.16	0.65
A201-B	4.36	0.05	0.07	0.57	0.26	0.22	0.65
A201-C	4.28	0.07	0.10	1.01	0.26	0.26	0.65

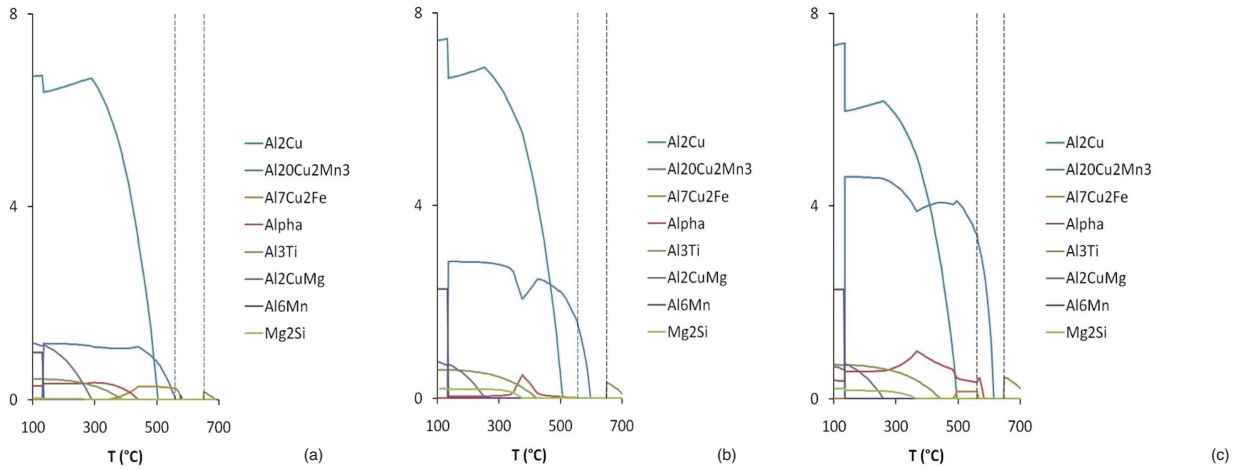
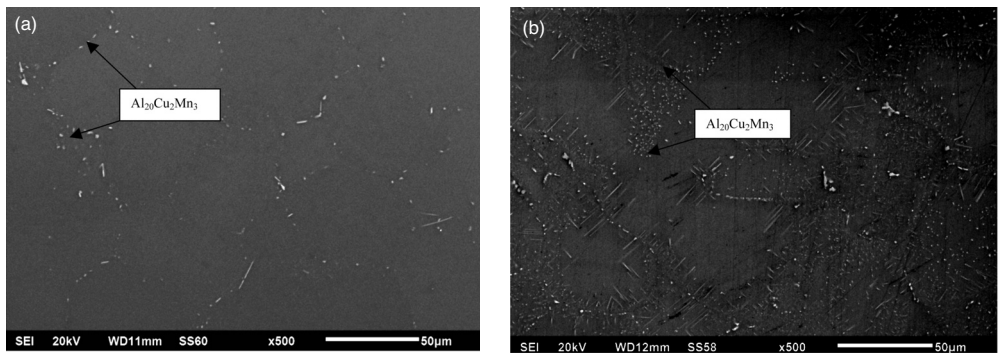


Figure 1—Calculated phase equilibria (minor phases) in mass percentage for Al alloys with compositions given in Table I corresponding to alloy a) A201-A, b) A201-B, c) A201-C. The vertical broken lines give the solidus temperature, and the temperature at which the primary FCC phase first appears upon cooling



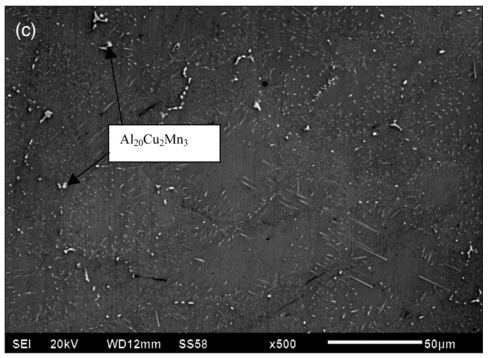


Figure 2—Secondary electron images of samples of alloy (a) A201-A , (b) A201-B and (c) A201-C in T6 temper condition

2(a–c). In all alloys the phases with the spherical and needle-like morphology were tentatively identified by EDS (see typical EDS spectra in Figure 3 representing all three alloys) to be $Al_{20}Cu_2Mn_3$. Figure 2 (a-c) shows that the volume fraction of the precipitates increases with increasing Mn content, as predicted from the equilibrium calculations. This is expected to have an adverse effect on ductility; the effect on strength is not expected to be large since the equilibrium amount of Al_2Cu is predicted to be little affected by the presence of the Mn-containing precipitates (see Figure 1).

Beffort *et al.*[12] also showed that increasing levels of Mn, Cr and V in the matrix will result in the formation of intermetallic compounds that reduce the fracture toughness of the material due to the fact that they act as crack initiators. They postulated that the embrittlement was caused by particles with composition of $(AlCu)_6(Cu,Mn,Cr,V)$. Based on the EDS analysis (see Figure 3) and ThermoCalc (see Figure 1) done in this study, the embrittling particles are believed to rather be $Al_{20}Cu_2Mn_3$.

The SEM fractographs of the tensile samples of the three alloys in T6 temper condition are shown in Figure 4 (a-c). All alloys failed in a ductile manner, as characterized by the dimple nature of the fracture surfaces. Slight variation in dimple size was observed for all three alloys. The presence of $Al_{20}Cu_2Mn_3$ particles is more evident in A201-B and A201-C (see Figure 5 for higher magnification). This is expected to

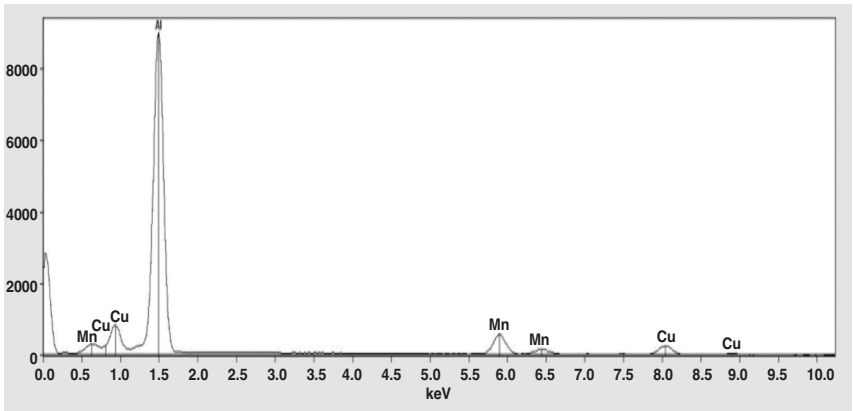
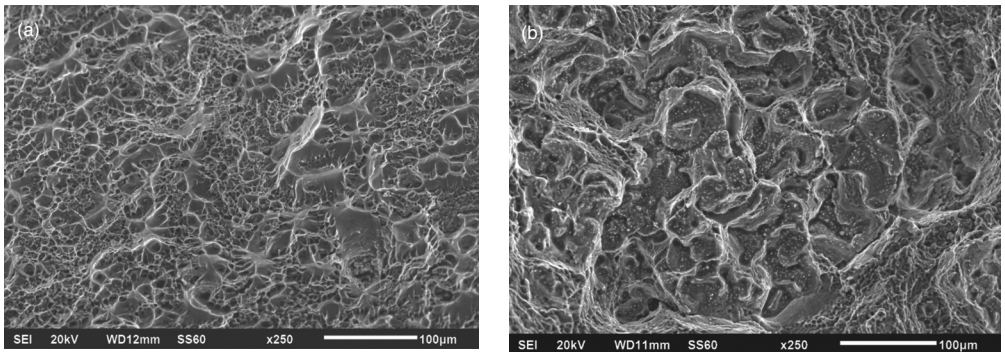


Figure 3—EDS spectra of tentatively identified $Al_{20}Cu_2Mn_3$ particle



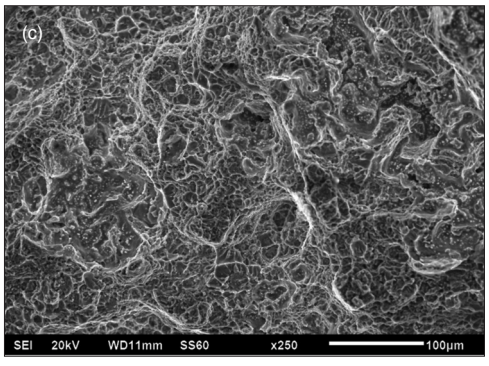


Figure 4—Secondary electron images of fracture surfaces of tensile specimens for alloys (a) A201-A, (b) A201-B and (c) A201-C in T6 temper condition

cause lower ductility as these hard, brittle particles will serve as stress concentrators, making it easy for crack initiation to occur. Close inspection at high magnification reveals the presence of the $Al_{20}Cu_2Mn_3$ (see Figure 5(a-b)).

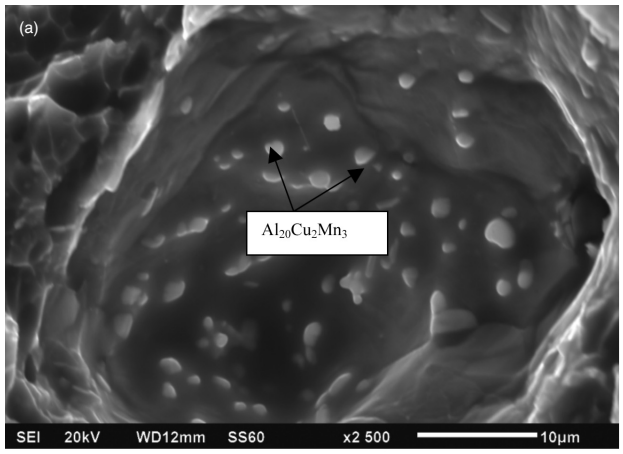
Mechanical behaviour

The average tensile values of the alloys in the T6 temper condition are shown in Table II, with the standard deviation from nine values shown in brackets. Table II shows that the tensile properties (especially ductility) of the alloys decrease with increasing Mn content as expected from SEM analysis. The difference in the elongation of A201-B and A201-C is not significant. Figure 6 clearly shows how the brittle particles cracked during tensile testing. The same phenomenon was experienced by Möller et al using SSM-HPDC F357 alloy[13].

Tseng *et al.*[14] showed that Mn-bearing phases such as $Al_{20}Cu_2Mn_3$ caused the solid solution level of copper in the matrix of A206 (the Ag-free version of A201) to decrease. More importantly, they suggested that increasing the Mn-solution level retarded the precipitation of the strengthening phases in the alloy. This resulted in a decrease in strength of the A206 alloy with increasing Mn levels—in agreement with the observed results for A201 in this study (Table II).

Conclusions

High Mn contents in rheo-processed Al-Cu-Mg-Ag alloy A201 resulted in the formation of high volume fractions of $Al_{20}Cu_2Mn_3$ particles. Tensile properties of the alloys decreased with increasing Mn content. Micro-cracking of the $Al_{20}Cu_2Mn_3$ particles occurred during tensile testing, which



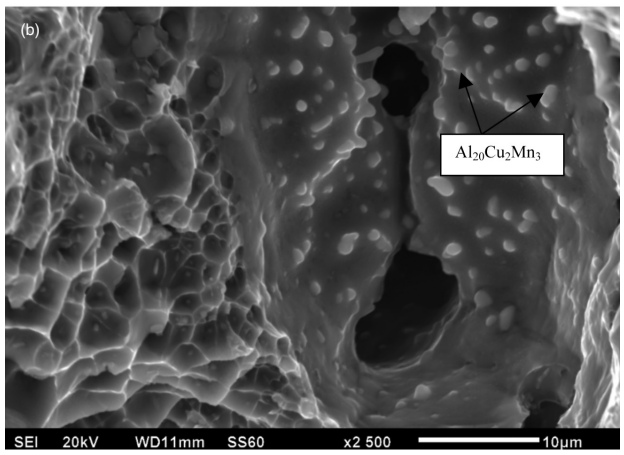


Figure 5—Secondary electron images of fracture surfaces of T6 tensile specimens at magnification of x2500 for alloys (a) A201-B, and (b) A201-C

Table II

Tensile properties of the three A201 alloys in T6 temper condition

Alloy	YS (MPa)	UTS (MPa)	%A
A201-A	297 (20.1)	407 (10.6)	13.4 (4.8)
A201-B	273 (9.7)	371 (12.1)	6.0 (1.6)
A201-C	267 (8.0)	353 (18.9)	4.2 (0.8)

caused a marked reduction in ductility. Increasing the Mn-solution level presumably retards the precipitation of the strengthening phases in the alloy, which results in a decrease in strength with increasing Mn levels.

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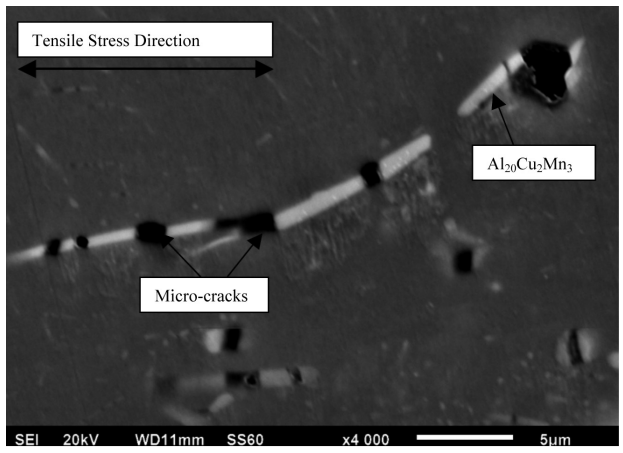

Figure 6—Secondary electron image of 0.6%Mn containing A201 in T6 temper condition after tensile testing to fracture

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