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Samuel, Ehab; Garat, Michel; Major, Fred

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Mechanical Property Behaviour of Rheocast 319 Alloys with and without Iron Additions

E. Samuel¹, M. Garat² and J.-F. Major²

CNRC-Centre des technologies de l'aluminium¹, Rio Tinto Alcan²

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ABSTRACT

Cast aluminum alloys have witnessed a notable increase in use in the automotive and transport industry by virtue of their mechanical properties. The ability of these alloys to be easily cast into complex shapes coupled with a favourable strength-to-weight ratio has given them an edge over cast irons. Cast 319 alloys, for example, are heavily used in the fabrication of engine blocks. The addition of copper in these alloys aids in maintaining high mechanical properties in service, at temperatures of more than 200°C. One particular area of casting which has received further and further attention is the area of semi-solid casting, where an alloy casting is prepared as a slurry with flow properties that resemble both solid and liquid. In the present work, the effects of iron additions and heat treatment to the mechanical properties of a 319 semi-solid alloy were studied. This alloy was prepared using the SEED process, developed by Rio Tinto Alcan (RTA) in collaboration with the Aluminium Technology Centre (ATC) of NRC Canada. The SEED (Swirled Enthalpy Equilibration Device) process is a novel rheocasting method which yields a semi-solid slurry from the mechanical stirring and cooling of the molten aluminum. The rheocasting process is characterized by very good mechanical properties owing to a globular microstructure, which was confirmed via optical and scanning electron microscopy. Typical values of peak stress and elongation recorded in this work were ~350 MPa and ~5%, respectively. The SEED method has already proven successful in the case of 356/357 and 6061 alloys, for example, with fluid die filling, sound castings and good mechanical properties being reported. In addition, the production of large slugs of 6'' diameter and more, capable of engine blocks, has been demonstrated and the possibility to dilute the standard high Fe and 9% Si die casting alloy A380 into a medium Fe, 6 to 7% Si alloy of the 319 type well suited to SEED has been successfully tested. The eventual aims of this study are to use recycled aluminum, containing iron traces, for future casting as well as to develop a fully industrial version of the SEED casting method.

Keywords: aluminum casting, semi-solid, semi-solid casting, rheocasting, SEED process

INTRODUCTION

Cast aluminum alloys, more specifically aluminum-silicon alloys, have established themselves as a suitable alternative to gray cast iron in the automotive industry, by virtue of their low density and good strength [1]. Copper additions to the Al-Si system allow the cast alloy to maintain a favourable level of strength in service, where temperatures can exceed 200°C (392F) [2]. In turn, these aluminum-silicon-copper alloys, or '319 alloys', are of significance in the fabrication of engine blocks.

Although many have investigated cast 319 alloys in terms of microstructure and mechanical properties, there have been fewer studies devoted to semi-solid cast 319 alloys [3-5]. Semi-solids have been of interest to many researchers owing to the uniform flow of the globular microstructure, which facilitates die-filling. Semi-solid alloys can be formed into high integrity castings with a much lower potential for major casting defects and flaws such as porosity and shrinkage. The 319 semi-solid billets used in the present work were rheocast using the SEED process [6, 7]. Essentially, the liquid molten aluminum is cooled to slurry via heat extraction through mechanical agitation in a metallic cylinder (swirling). Therefore, this allows the slurry to be formed on site, thus enabling one to avoid any extra processing steps and costs, such as transportation, storage and reheating of the semi-solid billets in thixocasting. This flexibility saves time and allows for immediate casting of the billet. The SEED method has proven successful in the case of 356/357 and 6061 alloys, among others [8-13].

In keeping with the idea of lowering processing costs, often, recycled aluminum is used instead of pure, where the former is known to contain traces of iron. The presence of iron in 319 aluminum alloys, for example, results from the use of recycled material. However, iron can also severely affect the alloy's mechanical properties due to the formation of the β -Al₅FeSi (' β -Fe') phase, which appear as needles in the microstructure [2, 14-19]. The size of these needles is a function of the iron content in the alloy, with further reductions observed in mechanical properties for increased iron additions. Although the use of primary aluminum stock to cast any alloy ensures a high level of purity, it is the authors' eventual aim to use recycled aluminum, i.e. containing iron, in the casting of future alloys. In this work, the mechanical property behaviour of three 319 SEED-processed semi-solid alloys with varying Fe content was studied.

EXPERIMENTAL PROCEDURE

The 319 alloys were cast using the SEED process at the Aluminium Technology Centre (ATC-NRC) facility in Chicoutimi, Quebec, Canada. The steps involved in the SEED process are given in Figure 1.

Figure 1 - The SEED rheocasting process (a-d) with an HPDC press (e) [12]

Heat extraction of the molten material is achieved through mechanical swirling (step (b)) in a confined cylinder. Primary alpha aluminum phase begins to form. In step (d), the billet is ready for casting (step (e)). A lower pouring temperature will allow for a more globular structure (step (a)) due to an increase in the nucleation sites for the alpha aluminum phase to form. Any recycled material can be reused in the cycle shown in Figure 1.

In the present study, wedge shape plates were cast. These plates were subjected to a T7 heat treatment (solution heat treatment: 495°C(923F)/5 hours, room temperature water quench, artificial aging: 210°C(410F)/4 hours). From these plates, round bar ASTM B557 tensile test samples were machined. A general schematic of the tensile sample locations along the wedge is shown in Figure 2. The chemical compositions of the three cast alloys are given in Table 1.

Figure 2 - General schematic of the tensile sample locations along the wedge mold.

Table 1 - Chemical composition (wt%) of the 319 alloys used in this study.

Si	Mg	Cu	Fe	Ti	Sr	Al
6.46±0.05	0.19±0.02	2.80±0.10	0.09, 0.40, 0.60	≤ 0.10	0.004±0.0005	Balance

Casting was carried out using a 750-kg capacity electrical resistance furnace, and the mold was kept heated at 240-300°C (464-572F). A filling speed of 0.1 m/s and injection pressure and 900 bars were also used. Tensile tests were carried out using a 50-kN capacity MTS servohydraulic tensile tester, with an extensometer attached. Microscope work was performed using an Olympus BX51M optical microscope coupled with a Clemex camera and image analysis system, as well as a Hitachi SU-70 field emission scanning electron microscope coupled with an Inca 300 series EDX detector. Samples for either optical or scanning electron microscopy were taken from a point just below the tensile fracture area.

RESULTS AND DISCUSSION

The SEED process is a cost-effective means of creating semi-solid cast parts. One is able to produce the billet on site and subsequently cast it into shape. As the SEED process has been extensively used in the case of 356/357 and 6061 alloys, for example, it was only natural to extend to other materials such as 319, which already has an established reputation in the automotive industry.

TENSILE PROPERTIES

Several room temperature tensile tests were carried out in order to ascertain the mechanical property behaviour of the SEED processed 319 semi-solid alloys. Results were acquired in the form of force-displacement data and then converted into stress-strain curves. The peak and yield stress for each curve obtained was recorded. Since the peak stress will vary with strain, depending on the degree of strain hardening in the sample prior to fracture, the average peak stress for each alloy and corresponding strains were used. The reason for this is that none of the samples tested showed any signs of necking. Therefore, one cannot use the term 'ultimate tensile stress'; instead, the maximum or 'peak stress' observed (always at the failure strain) is used instead. The results are given in Figure 3.

Figure 3 – Stress vs. failure strain for the 319 alloys used in this study.

A near parallel behaviour between the 319 alloy and the 319 + 0.4%Fe alloy is observed, in terms of average peak stress. Therefore, an addition of 0.4%Fe to the alloy has no deleterious effect on ductility. The increase in peak stress with strain demonstrates that strain hardening was achieved for elongations exceeding 7%. If the iron content is then increased to 0.6%, however, there is a noticeable decrease in peak stress owing to the increased brittleness of the alloy. Moreover, the elongation is reduced from ~7% to ~4%. In fact, it is only at strains below 2% where an increase in peak stress is observed. The 319 + 0.6%Fe alloy, however, does demonstrate a greater value of yield stress. This indicates that the increased iron content hardens the alloy, especially at 0.6%. This is, however, at the expense of ductility. There appears to be a slight linear decrease in yield stress for each alloy, with strain.

Despite any reductions in mechanical properties, it can be seen that a maximum peak stress of 360 MPa, at a strain of 7.5%, was still obtained for the 319 alloy. The level of strength achieved in the 319 alloy tested after being overaged, coupled with a very good value of ductility, has been obtained in using the SEED process; this is remarkable an overaged (T7) alloy is expected to demonstrate a decrease in strength, compared to peak aged (T7) alloys. In other words, the globular microstructure and the flow of the material were such that very good values of mechanical properties were realized in this work.

The 319 alloy has been reported to suffer from 'thermal growth' when a cast part is put into service, with changes in volume of up to 0.1% being recorded [2, 20, 21]. In parts requiring strict dimensional tolerances, changes such as this can severely affect the performance of cast parts in service. It is suggested that these dimensional changes are due to a supersaturation of Cu after solution heat treatment. Due to the various elements and complex precipitation reactions found in the 319 alloy system, the simpler binary Al-Cu system is often referred to:

For peak aging conditions, the thermodynamically unstable θ' phase forms, yet with further aging time (i.e. overaging), it is replaced by the more stable θ (CuAl_2) phase. From the HRTEM observations of Mancha-Molinar *et al.* [2], it was found that aging at 220-240°C (428-464F) led to the precipitation of secondary phases, with corresponding changes in sample dimensions. Researchers have found that the use of an overaging/T7 heat treatment can minimize these dimensional changes. However, there is little research available to elaborate on the actual mechanism(s) responsible for this growth in aluminum alloy castings. In the the present work, these effects were not taken into consideration, given that mechanical tests were carried out on samples machined from cast plates, rather than on cast parts. For our work, T7 heat treatments were used to obtain a compromise between strength (usually achieved though peak aging (T6)) and ductility (which increases during underaging or overaging, and is lowest at peak aging). However, we were careful not to overage at too high a temperature so as not to drastic decrease the alloy's strength.

METALLOGRAPHY

In the 319 alloy (with 0.09%Fe), evidence of the blocky CuAl_2 phase was observed. During solution heat treatment, the alloy becomes supersaturated with Cu. Copper that is dissolved precipitates out during aging in a fine, eutectic form of the CuAl_2 phase, while undissolved copper forms the blocky form of the CuAl_2 phase. This blocky form is observed in alloys modified with strontium; the presence of Sr promotes the segregation of Cu, often resulting in the clustering of the blocky CuAl_2 phase [14]. Therefore, although the addition of Sr to Al-Si cast alloys is used to improve mechanical properties by modifying the morphology of acicular Si into a finer form [2, 21], the presence of a blocky CuAl_2 phase can restrict this potential improvement in strength. Blocky CuAl_2 phase particles (light gray) are shown in Figure 4(a), where it can be seen that these particles are within vicinity of each other. This may be due to the reported segregation of Cu due to the addition of strontium [14]. Figure 4(b) demonstrates the globular microstructure obtained using the SEED process.

The presence of Fe in the alloy results in the formation of the $\beta\text{-Al}_3\text{FeSi}$ phase. This phase appears in the microstructure as needles, increases the brittleness of the alloy and can lead to detrimental effects on the elongation and peak stress [2, 14-19]. The size of $\beta\text{-Fe}$ needles is a function of the alloy iron content. According to Li *et al.* [14], a 319 alloy with up to 0.4%Fe will demonstrate adequate values of stress and ductility, as the $\beta\text{-Fe}$ needles present are uniformly distributed throughout the alloy. This supports the observations made in Figure 1 where the ductility of the 319 and 319 + 0.4%Fe alloys are nearly identical, and there is only a slight dip in strength with the Fe addition compared to the 319 alone. These $\beta\text{-Al}_3\text{FeSi}$ needles are shown in Figures 5 and 6 in 319 + 0.4%Fe and 319 + 0.6%Fe, respectively; however, they were not apparent in the 319 alloy with only 0.09%Fe (Figure 4(a)).

In Figures 5 and 6, the $\beta\text{-Fe}$ needles, distributed throughout the microstructure, are only evident when an adequate amount of Fe is present (i.e. > 0.09%). The micrograph shown in Figure 6 reveals a stronger presence of $\beta\text{-Fe}$ needles than in Figure 5. Moreover, a large portion of the $\beta\text{-Fe}$ needles observed for the 319 + 0.6%Fe alloy are much longer/larger than for the 319 + 0.4%Fe alloy. This is attributed to the increased iron content, which influences the size of the $\beta\text{-Fe}$ needles, and contributes to the overall reduction in mechanical properties.

Figure 4 – (a) SEM micrograph (600X) of a typical 319 alloy microstructure where the presence of clustered blocky CuAl_2 phase is evident; (b) optical micrograph (100X) demonstrating the globular microstructure obtained using the SEED process.

Figure 5 – SEM micrograph of a typical 319+0.4%Fe microstructure (400X). The presence of $\beta\text{-Al}_5\text{FeSi}$ needles is evident.

Figure 6 – SEM micrograph of a typical 319+0.6%Fe microstructure (400X). There is a greater presence of $\beta\text{-Al}_5\text{FeSi}$ needles.

It has been observed [22-26] that an increase in the pouring temperature alters the microstructure from globular to dendritic. As this occurs, the viscosity of the semi-solid material is altered and flow is generally more difficult. At lower pouring temperatures, there are more nucleation sites for the formation of the primary phase (α -Al) and at higher pouring temperatures, i.e. larger superheats, growth of the primary phase is directional and results in a dendritic structure. These dendrites interlock, resulting in a non-uniform microstructure. For a more globular microstructure, the flow is facilitated by the presence of equiaxed globules and a reduction in the number of eutectic pools. These pools are present more so for a dendritic structure, given the dendrites tendency to have their branches interlock, thus creating forests of dendrites with large eutectic pools in between them. This liquid segregation is thus encouraged when having a dendritic structure. Liquid segregation is, more often than not, inevitable, owing to the separation of the solid and liquid phases comprising the semi-solid under the application of a force or pressure during flow. For a more globular structure, these globules flow with the eutectic phase and the resultant microstructure maintains its homogeneity.

Given that the flow is influenced by the microstructure (i.e. globular, dendritic, rosette), then the viscosity is also influenced by the microstructure. Globular microstructures demonstrate a lower degree of viscosity and, in turn, much easier material flow. From preliminary observations, it was observed that, for the 319 alloy, an increase in the pouring temperature from 626°C to 633°C brings about a reduction in the yield and peak stress with an increase in pouring temperature. At 636°C, the microstructure observed in the 319 alloy was neither globular nor uniform, contributing to the further reduction in mechanical properties. From the work of Lashkari and Ghomashchi [22], it was reported that regardless of the applied pressure, a higher degree of strain is achieved for an alloy at a lower pouring temperature; this is again attributed to a more uniform, globular microstructure. Lower shear rates also allow the solid phase in the semi-solid material to move more easily. As can be seen from Figure 7, the microstructure of the 319 alloy appears more globular at 626°C than at 630°C or 640°C.

Figure 7 – Optical microscope photos (50X) of the 319 alloy at a pouring temperature of (a) 626°C, (b) 630°C and (c) 640°C.

Table 2 demonstrates the mechanical properties observed for the 319 alloy at two different pouring temperatures. Although the elongation values are comparable, the ductility at a higher pouring temperature is expected to decrease given the flow is more uniform

for a globular microstructure (i.e. at lower pouring temperatures). It is seen that an increase in pouring temperature also brings about a small decrease in the yield and peak stress values.

Table 2. Mechanical properties for the 319 alloy, for two pouring temperatures.

Pouring temperature (°C)	Yield stress (MPa)	Peak stress (MPa)	Elongation (%)
626	285	345	3.9
633	277	337	4.1

The microstructures of the 319 alloy and the 319 + 0.4%Fe alloy, both for a pouring temperature of 626°C, are shown in Figure 8. The microstructure in Figure 8(a) appears more globular, whereas that in Figure 8(b) appears to have a more rosette/dendritic quality to it.

Figure 8 – Optical microscope photo of the (a) 319 alloy and (b) the 319 + 0.4%Fe alloy, both at a pouring temperature of 626°C (50X). The microstructure in (a) appears more globular, whereas that in (b) has a more rosette/dendritic-type quality to it.

Table 3 demonstrates the mechanical properties observed for the 319 alloy as well as for the 319 + 0.4%Fe alloy, both having been poured at a temperature of 626°C. As can be seen, the addition of Fe to the 319 alloy only decreases the mechanical properties slightly.

Table 3. Mechanical properties for two 319 alloys, for a pouring temperature of 626°C.

Alloy	Yield stress (MPa)	Peak stress (MPa)	Elongation (%)
319 + 0.09%Fe	285	345	3.9
319 + 0.4%Fe	281	333	3.7

As with iron content and pouring temperature, the alloy microstructure was also seen to change with filling speed. However, a thorough investigation into the use of various filling speeds was not carried out to the same extent as the effect of pouring temperatures on globularity and iron content on mechanical properties. Figures 9 and 10 demonstrate some observations made when the filling speed was changed from 0.1 m/s. In Figure 9, for example, the 319 alloy is shown at 0.1 m/s and 1.0 m/s, for a pouring temperature of 626°C. As can be seen, for the lower filling speed (Figure 9(a)), the structure appears more globular, whereas at the higher filling speed (Figure 9(b)), the structure appears to have a slightly more dendritic/rosette form. This behavior is observed in Figure 10 as well for the 319 alloy, at a pouring temperature of 630°C and filling speeds of 0.1 m/s and 0.5 m/s. At the lower filling speed (Figure 10(a)), the structure appears more globular, whereas at the higher filling speed (Figure 10(b)), the structure appears more dendritic.

Figure 9 – Optical microscope photos (100X) of the 319 alloy at a pouring temperature of 626°C and a filling speed of (a) 0.1 m/s and (b) 1.0 m/s.

Figure 10 – Optical microscope photos (50X) of the 319 alloy at a pouring temperature of 630°C and a filling speed of (a) 0.1 m/s and (b) 0.5 m/s.

CONCLUSION

In this study, three Al-Si-Cu alloys were examined in terms of mechanical property behaviour and microstructural response to varying iron contents, pouring temperatures and filling speeds. Although the presence of iron, especially at 0.6%, introduces some difficulty in achieving a suitable combination of properties, it was found that the alloys at 0.09 and 0.4% Fe behaved in a similar manner. This implies that one can substitute pure aluminum with that containing a small amount of iron in terms of alloy fabrication. In summary, it was observed that:

1. The addition of iron to 319 hardens the alloy to a degree, as evidenced by the increase in yield stress with iron content.
2. The addition of 0.4%Fe does not affect the 319 alloy's ductility. A near parallel behaviour in the peak stress vs. failure strain curves for 319 and 319 + 0.4%Fe was observed, with both alloys attaining a maximum failure strain of 7.5%.
3. For 0.6%Fe, the failure strain reduced from ~7% to ~4%. Moreover, an increase in peak stress was only observed at strains below 2%.

4. The presence of β -Al₃FeSi needles was observed primarily in the 319 + 0.4%Fe and 319 + 0.6%Fe alloys. At 0.6%Fe, these β -Fe needles were greater in size and concentration in the alloy, resulting in the increased brittleness observed for the 319 + 0.6%Fe alloy. It is believed that the β -Fe needles which formed in the 319 + 0.4%Fe alloy (less than for the 319 + 0.6%Fe alloy) are uniformly distributed in the microstructure (and smaller in size than at 0.6%Fe). This, in turn, does not lower the ductility of the 319 alloy.
5. A peak stress value of 360 MPa at a failure strain of 7.5% was achieved for the 319 alloy in the overaged (T7) condition; this is comparable to similar values of peak stress obtained by others in the peak-aged (T6) regime. Therefore, the SEED processed alloys' flow was such that the mechanical properties obtained in the overaged state are comparable to the same alloy in the peak-aged state.
6. The CuAl₂ phase was seen to form as a blocky phase and cluster together with other CuAl₂ phase particles. It is believed that the presence of Sr promotes the segregation of Cu in the alloy, although this requires further investigation.
7. An increase in the pouring temperature was observed to alter the microstructure of the 319 alloy from globular to dendritic. At a pouring temperature of 640°C, for example, the 319 microstructure appeared very dendritic. This brought about a slight reduction in the yield and peak stress, for example, when the temperature was increased from 626°C to 633°C. The best mechanical properties obtained for the 319 alloy was at 626°C, which can be attributed to the globular microstructure.
8. The 319 alloy microstructure was also observed to change from globular to a more dendritic/rosette type with increased filling speed, for a given pouring temperature. This is evident, for example, at a pouring temperature of 630°C and filling speeds of 0.1 m/s and 0.5 m/s.

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CONTACT INFORMATION

EHAB SAMUEL,
National Research Council of Canada, Aluminum Technology Centre (ATC-NRC),
501, boulevard de l'Université est.
Chicoutimi (Québec)
G7H 8C3
Ehab.Samuel @imi.cnrc-nrc.gc.ca

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