Effect Of Heat Treatments In The Silicon Eutectic Crystal Evolution In Al-Si Alloys

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Abstract. This paper describes the heat treatment effect on the eutectic silicon evolution in the A357 alloy, obtained by semisolid forming process (SSM). The coarsening rate of the silicon was determined by Image Analysis Technique in specimens from rheocasting ingots and thixocasting components. The study was realized in the temperature range from 450 to 550ºC by applying heating times between 1 and 24 hours. The results show that during the heat treatment the coarsening and sphereodization of the silicon particles is produced and the fragmentation stages, which are observed in conventional alloys, do not appear. Kinetic silicon growth has been adjusted to the Oswald’s ripening equation.

Introduction

Semisolid forming is an alternative to the traditional casting processes of aluminium alloys [1]. These processes consist of the material pseudoplasticity control when it keeps in semisolid state. Under these conditions viscosity decreases with shear rate to which the material is agitated, and the thixotropic behaviour relies on the shear stress and time. The resultant structure consists of globular alpha particles dispersed in the eutectic liquid. The ingot obtained by this process and showing these characteristics, is called rheocasting ingot. For conformation by means of thixocasting, the ingots are cut at adequate size and are heated to the semisolid range, about 7 minutes at 585ºC, through an induction furnace. Immediately after, the ingot is automatically handled like a solid to the injection machine. The resulting microstructure contains a globular alpha phase surrounded by a very fine eutectic microconstituent. Through heat treatments the eutectic silicon can grow and become globular as it happens in conventional casting [2]. The formation of silicon spherulites and growth of larger particles at the expense of smaller ones, and the subsequent redistribution of the crystals in the Al-matrix, brings about an improvement in plastic behaviour and workability.

Grain growth occurs in polycrystalline materials to decrease the system energy by decreasing the total grain boundary energy. The total surface energy of the silicon particles is reduced by coarsening. Under ideals conditions competitive grain growth kinetics should obey a power law proposed by Hu and Rath [3-4] that describe s the growth of the average particle size as a function of time, according to the Ostwald ripening model [5]:

\begin{equation}
D^m - D_O^m = kt
\end{equation}

where D is the length scale, e.g., mean grain diameter, after annealing for time t, k is a temperature-dependent rate parameter and D\textsubscript{O} is the length scale or the grain size at t = 0

\begin{equation}
D^{1/n} - D_{O}^{1/n} = kt
\end{equation}

where n is an empirical constant typically ≤ 0.5 . The grain growth exponent is also often defined as m = 1/n and is then typically ≥ 2. The value of m depends on the grain growth mechanism. The classical value of grain growth exponent for pure metals or ceramics is 2. The coarsening exponent
for classical theoretical papers on Oswald ripening [6], of grains in a stable polycrystalline matrix, is $m = 3$ in case of volume diffusion controlled growth, $m = 4$ if the grain growth is controlled by grain boundary diffusion, and $m = 5$ for diffusion on dislocations; but in practice much higher values for $m$ (5-11) are often found [7]. These higher values of $m$ are usually ascribed to particle growth due to coalescence. In spite of the many studies devoted to this subject there is, however, still no satisfactory understanding of this ageing behaviour.

Assuming that $D_0 \ll D$

$$D = k t^n$$  \hspace{1cm} (3)

The growth rate $k$ varies with temperature according to an Arrhenius-type equation of the form [8]

$$k = k_o \exp \left(-\frac{Q}{RT}\right)$$  \hspace{1cm} (4)

from the rate parameter $k$ one gets access to the activation energy $Q$, which can be used to deduce the grain growth mechanism.

**Experimental procedure**

The material used in the present study was Al alloy A357 manufactured by rheocasting and thixocasting processes. This alloy contains 7.75 wt-%Si, 0.65 wt-%Mg and 0.024 wt-%Sr. Two types of materials were tested. A rheocast A357 aluminium alloy ingot was the starting material. Two areas were studied in this material: **periphery** and **core** of ingot. After, the A357 ingot was formed in semisolid state using a thixocasting method. The semisolid state was obtained in an induction industrial furnace heating the ingot for 7 minutes before water quenching. The specimens obtained using this procedure are named **component**. This component has a homogeneous microstructure and that is why only a zone is studied.

The heat treatments were carried out on samples with a volume of $2 \text{ cm}^3$ in a tubular electric $\varnothing 50\text{mm}$ furnace, with a heat treating region about $200\text{mm}$. The samples were being heated to 450, 500 and $550^\circ\text{C}$ in the range from 15 minutes to 24 h and immediately after they were cooled in the air.

By means of image analysis the equivalent diameter and the aspect ratio of eutectic silicon were determined. A statistical analysis was made starting from a study of 20 regions belonging to every sample.

**Results and Discussion**

In Fig. 1 the ingot microstructure is shown. The first micrography shows a globular $\alpha$-phase and a very thin eutectic; this structure corresponds to the ingot centre. The second one reveals the ingot periphery with a greater size dendritic structure and the eutectic ratio becomes smaller. The component microstructure is observed in Fig. 2. The silicon size is larger and besides, the ingot heating provokes the appearance of fine eutectic spheroids inside the $\alpha$ grains. That is because when the temperature increases over the eutectic isotherm, the silicon solubility in the $\alpha$ phase decreases, and liquid drops are produced. These occluded drops in the $\alpha$ grains suffer an undercooling process and solidify fast [9]. Obviously, this phenomenon isn’t carried out in the liquid surrounding the $\alpha$-phase.
The micrographs obtained after several heat treatments, at 450°C, are shown in Fig. 3. These treatments produce a change in matrix composition, the silicon particles size increase and become globular. It is possible to see (Fig. 3) how silicon becomes spherical and grows quickly. This fact does not only take place at the core but at the periphery, with different speed in everyone. Besides, a shape and size variation of the α-phase spheres is also to be observed. The α-phase grows increasing temperature and time of annealing. The original grain boundaries are defined by silicon eutectic crystals and they are blurred as the silicon crystals grow and coalesce the α-phase grains. As
the treatment time goes on, the big silicon crystals become an homogeneous distribution inside the continuous alpha matrix.

The silicon equivalent diameter evolution at different temperatures, in the component, has been described in figure 4 and the corresponding power law is detailed in table 1. The silicon grain size grows quickly at the first heating phase and reaches values between 1,5 μm in the core and 1,7 μm in the periphery.

By applying the equation 3 the m value from table 1 is obtained. For a mean value of n=0.20 and applying other time the equation 3, the values of k_T are obtained (Fig. 5). These values are related in Table 1.

![Graph 1](image1.png)

**Fig. 4** The silicon equivalent diameter evolution at different T and times: a) Ingot core, b) Ingot periphery and c) component

<table>
<thead>
<tr>
<th>Temperature /°C</th>
<th>Function (INGOT CORE)</th>
<th>m</th>
<th>k_T</th>
</tr>
</thead>
<tbody>
<tr>
<td>450</td>
<td>$D = 0.409t^{0.23}$</td>
<td>4.3</td>
<td>0.409</td>
</tr>
<tr>
<td>500</td>
<td>$D = 0.605t^{0.15}$</td>
<td>6.6</td>
<td>0.605</td>
</tr>
<tr>
<td>550</td>
<td>$D = 0.774t^{0.20}$</td>
<td>5.0</td>
<td>0.774</td>
</tr>
<tr>
<td><strong>Function (INGOT PERIPHERY)</strong></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>450</td>
<td>$D = 0.565t^{0.20}$</td>
<td>5.0</td>
<td>0.565</td>
</tr>
<tr>
<td>500</td>
<td>$D = 0.603t^{0.23}$</td>
<td>4.3</td>
<td>0.603</td>
</tr>
<tr>
<td>550</td>
<td>$D = 0.973t^{0.19}$</td>
<td>5.2</td>
<td>0.973</td>
</tr>
<tr>
<td><strong>Function (COMPONENT)</strong></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>450</td>
<td>$D = 0.720t^{0.21}$</td>
<td>4.7</td>
<td>0.721</td>
</tr>
<tr>
<td>500</td>
<td>$D = 0.924t^{0.22}$</td>
<td>4.5</td>
<td>0.935</td>
</tr>
<tr>
<td>550</td>
<td>$D = 1.394t^{0.20}$</td>
<td>5.0</td>
<td>1.347</td>
</tr>
</tbody>
</table>
k and Q/R values in the equation 4 are determined by the graphic function of k in front of 1/T (Fig. 6).

The obtained activation energy values appear in table 2.

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**Fig. 5** Graphic function of diameter in front of 1/T: a) ingot core, b) ingot periphery and c) component

**Fig. 6** Graphic function of k in front of 1/T: a) ingot core, b) ingot periphery and c) component
Table 2

<table>
<thead>
<tr>
<th>Component</th>
<th>Q/R</th>
<th>Q kJ/mol</th>
</tr>
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<tbody>
<tr>
<td>ingot core</td>
<td>3808</td>
<td>32</td>
</tr>
<tr>
<td>ingot periphery</td>
<td>3177</td>
<td>26</td>
</tr>
<tr>
<td>component</td>
<td>3700</td>
<td>31</td>
</tr>
</tbody>
</table>

Conclusion

The resulting average activation energy to the silicon coarsening is 30 kJ·mol⁻¹. This value is in concordance with the one described in the bibliography [10], which is $Q = 32.75$ kJ/mol in the range of temperatures 480 ÷ 620 ºC. Therefore the silicon growth exponent is 5 and, consequently, the mechanism can be considered as volume diffusion in matrix.

References