

# Heat transfer and solidification behaviour of modified A357 alloy

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Received 3 June 2005; received in revised form 19 July 2006; accepted 24 July 2006

## Abstract

Al–Si alloys are subjected to melt treatment like modification to improve their mechanical properties. Non-destructive technique like thermal analysis is generally used to assess the effectiveness of melt treatment. In the present study, the behaviour of the melt treated Al–7Si–Mg alloy (A357) during solidification with or without chilling was investigated using thermal analysis. Thermal analysis and heat transfer parameters were determined. Thermal analysis parameters were affected significantly by modification and chilling. Modification treatment resulted in the increase of cooling rate, heat evolved, casting/mould interfacial heat flux and eutectic growth velocity. A theoretical model based on undercooling from the equilibrium temperature during eutectic solidification was used to predict growth velocities and eutectic grain size. The eutectic grain sizes estimated using the model and those measured from casting microstructures were found to be in good agreement.

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*Keywords:* Al–Si alloy; Modification; Solidification; Thermal analysis

## 1. Introduction

Aluminium alloy castings have achieved wide usage in automobile, aerospace and other applications because of their high strength to weight ratio [1]. Al–Si–Mg alloys, a group of heat treatable cast Al–Si alloys, exhibit good castability and corrosion resistance in addition to high strength to weight ratio. However, the pursuit of high quality castings with consistent mechanical properties depends upon the proper processing of the alloy, including grain refinement, modification and precipitation heat treatment, etc. [2–5].

It is well known fact that mechanical properties of Al–Si–Mg alloys are strongly related to the size, shape and distribution of eutectic silicon present in the microstructure. The eutectic silicon morphology, dendrite arm spacing (DAS), grain size, etc., play vital role when the alloy is put into specific use. In order to improve mechanical properties, these alloys are generally subjected to modification melt treatment, which transforms the acicular silicon morphology to fibrous one resulting in a noticeable improvement in elongation and strength. Sodium and strontium are the common modifiers in use [5–8].

Modification melt treatment combined with chilling is found to be synergetic to enhance the refinement and modification of

microstructure [9]. This is an important factor as majority of Al–Si–Mg alloys are cast in permanent or metallic moulds rather than sand moulds. Modification increases the heat transfer rate from the solidifying alloy to the mould wall which is the direct consequence of morphology transformation.

In the present work the effect of modification melt treatment and/or chilling on heat transfer and solidification behaviour of A357 alloy is investigated by pouring the alloy into metallic moulds or by allowing it to solidify in the fire clay crucible.

## 2. Experimental

Table 1 gives the composition of the alloy used in the present investigation. Fig. 1 shows the sketch of the experimental set-up. The alloy was taken in a crucible and melted in an electric furnace. Elemental sodium was used as the modifier to ensure complete modification. Predetermined amount (0.015%) of sodium was added to the melt and kept for 10 min to ensure complete modification. After the melt treatment, the alloy was poured into metallic moulds. The metallic moulds used in the present investigations were of copper, mild steel and stainless steel. Experiments were also carried out with unmodified and modified alloy subjected to air cooling in a crucible. Twin bore ceramic beaded thermocouples were inserted into the casting to get temperature data during solidification. The metallic moulds were instrumented with two additional thermocouples inserted in the mould wall to record temperature data of the mould. The temperature data was used as an input for estimating heat flux transients by inverse analysis. The mathematical details of the inverse analysis are given in Ref. [10]. The thermocouples were connected to a high-speed online data acquisition system NI SCXI 1000. Temperature data was acquired at 100 samples per second in these experiments. Metallographic test specimens were prepared using

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Table 1  
Composition by wt.% of A357 alloy

Si	6.86
Mg	0.53
Fe	0.12
Zn	0.03
Ti	0.14
Al	Balance

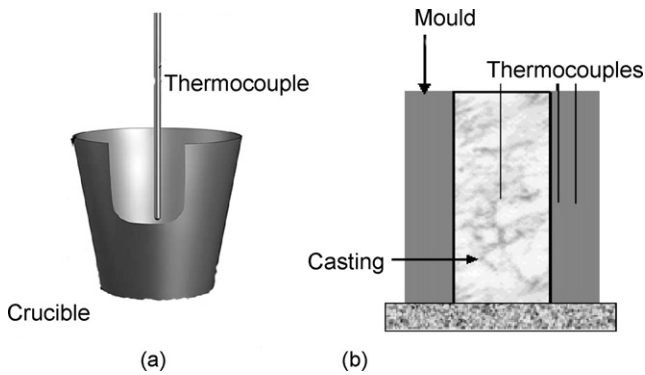


Fig. 1. Schematic representation of experimental set-up for thermal analysis.

a section of the casting. The sections cut from samples were metallographically polished and etched. The samples were then subjected to microexamination.

### 3. Results and discussion

Fig. 2 shows the cooling behaviour of the alloy solidified in the clay crucible for both unmodified and modified conditions. The modified alloy showed the characteristic eutectic depression and reduction in arrest time. Addition of elemental sodium affected not only thermal analysis parameters at eutectic region but also in the liquidous region considerably. Primary nucleation

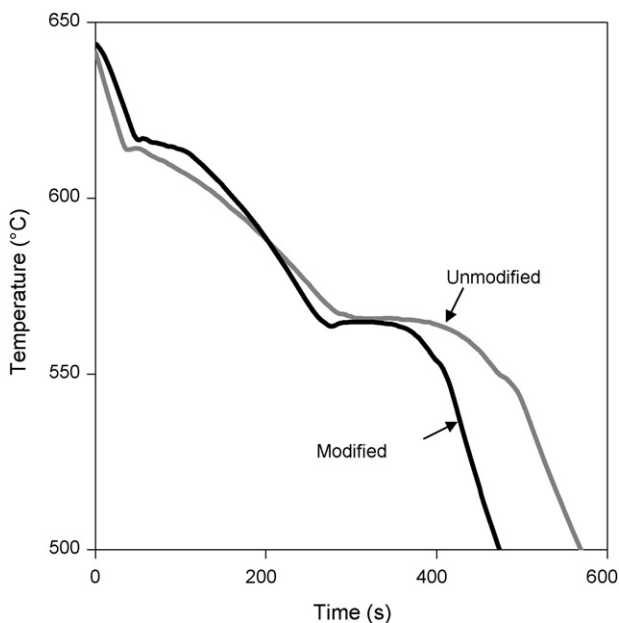


Fig. 2. Cooling curve for the alloy solidified in the clay crucible for both unmodified and modified conditions.

started at a higher temperature with decreased undercooling. An increase of 3 °C was observed in the  $\alpha$ -nucleation temperature. This suggests that the presence of sodium modifier in the melt promoted  $\alpha$ -aluminium to nucleate at a higher temperature. The modification also reduced the undercooling during primary solidification indicating a reduction of barrier for primary nucleation. There was a sharp increase in the cooling rate in the primary region when the alloy was modified with elemental sodium. This factor contributed to the significant reduction in the total solidification time. This cooling curve is used to calculate the heat of solidification ( $Q$ ) for the alloy. The rate of heat released due to solidification,  $Q$  can be obtained by using the following expression [11]:

$$Q = C_P \left[ k(T - T_f) + \frac{\Delta T}{\Delta t} \right] \quad (1)$$

where  $Q$  is the rate of heat released by the solidification process (J/s),  $C_P$  the specific heat of the alloy (J/kg K),  $k$  the slope of  $-\log((T - T_f)/(T_i - T_f))$  versus time curve,  $\Delta T$  the change in temperature (°C),  $T$  the temperature at any given time (°C),  $T_f$  the final temperature (°C),  $t$  the time (s), and  $\Delta t$  is the change in time (s).

The resultant plots of the heat of solidification versus time for both unmodified and modified alloys are shown in Fig. 3. The modification melt treatment resulted in higher amount of heat release during eutectic solidification. The peak value of heat of solidification in the eutectic range for the unmodified alloy was 528 J/kg s where as the corresponding value in the alloy modified with elemental sodium was 548.5 J/kg s.

The mould thermal history obtained during solidification of the alloy in metallic moulds is used for calculating heat flux transients by inverse analysis. The variation of heat flux with respect to time for the alloy casting solidified in copper mould in both unmodified and modified conditions is presented in Fig. 4.

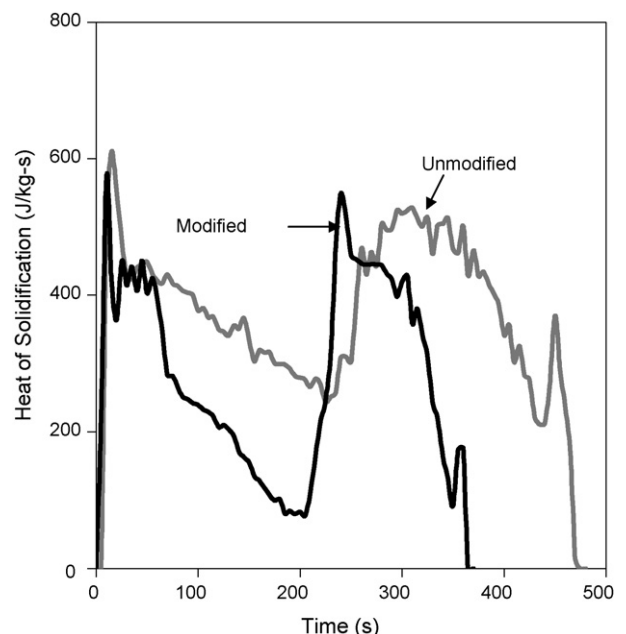


Fig. 3. Heat of solidification vs. time for both unmodified and modified alloys.

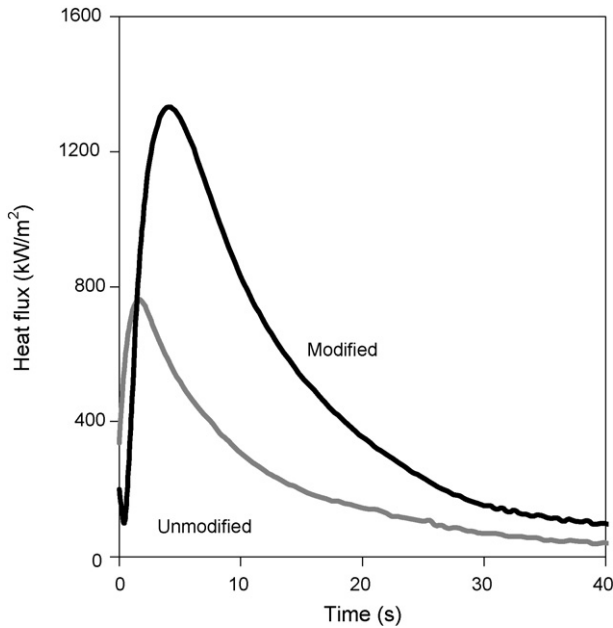


Fig. 4. The variation of heat flux with respect to time for the alloy casting solidified in copper mould in both unmodified and modified conditions.

The heat flux for the modified alloy is considerably higher. For example, the peak flux attained in the unmodified alloy casting solidified in copper mould was about 760 kW/m<sup>2</sup> where as for modified alloy it was about 1331 kW/m<sup>2</sup>. The increase in heat flux was also observed in mild steel and stainless steel moulds. However, the extent of increase was not as pronounced as in case of copper mould.

The heat flux transients indicate the flow of heat from the casting to the mould, which primarily depends on the interfacial contact nature between the casting surface and mould wall. In unmodified alloy, the acicular morphology of the eutectic silicon prevents efficient contact between the mould wall and solidifying casting. The fine fibrous morphology of the silicon of the initial solidified shell in the modified condition in the casting/mould interfacial region facilitates the rate of heat removal from the casting by increasing the electronic heat conduction; hence the cooling rate increases in the modified alloy [9].

Castings showing high heat flux values and those which solidified with small arrest times yielded well modified microstructure. On the other hand, castings with high arrest time and low heat flux produced unmodified acicular structure. A good correlation was found between normalized values of the heat flux and arrest time. This is shown in Fig. 5. A combination of high peak heat flux and low arrest time yielded fully modified fibrous structure. On the other hand, lower heat flux and larger arrest were found to be associated with unmodified eutectic silicon morphology.

Dendrite arm spacing (DAS) exerts a marked influence on the mechanical properties of cast material. In Al–Si cast alloys, it has been shown that the secondary dendrite arm spacing (SDAS) follows power relation with local solidification time [12].

Hence

$$\lambda_2 = A_{Si} t_f^n$$

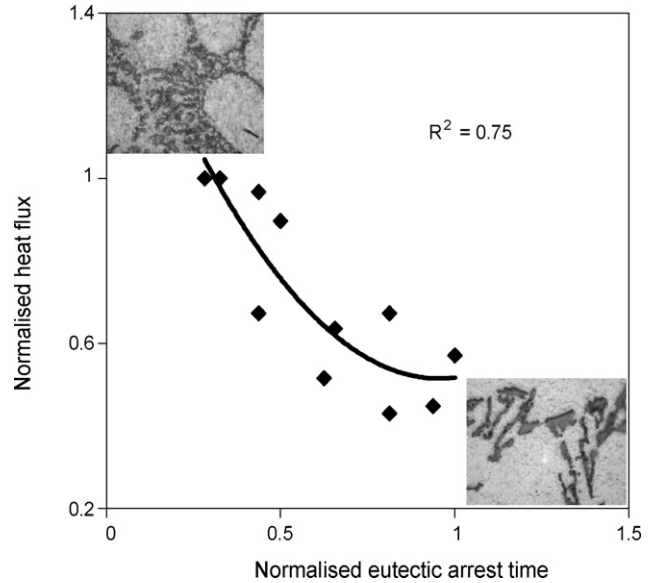


Fig. 5. Variation of normalized values of heat flux with arrest time.

where  $\lambda_2$  is the secondary DAS ( $\mu\text{m}$ ),  $A_{Si}$  the constant dependent on silicon content of the alloy,  $t_f$  the local solidification time (time elapsed from the start of solidification to its completion), and  $n$  is the exponent.

The value of  $A_{Si}$  varies inversely with silicon content of the alloy. A value of 12.8 has been proposed by Bamberger et al. for the aluminium alloy containing 7.5% silicon. In the present investigation, the secondary dendrite arm spacing for the cast alloys were measured from microstructures of both unmodified and modified samples solidified at varying cooling rates by using image analysis. Fig. 6 gives the plot of SDAS versus local solidification time. The best-fit equation obtained for the measured

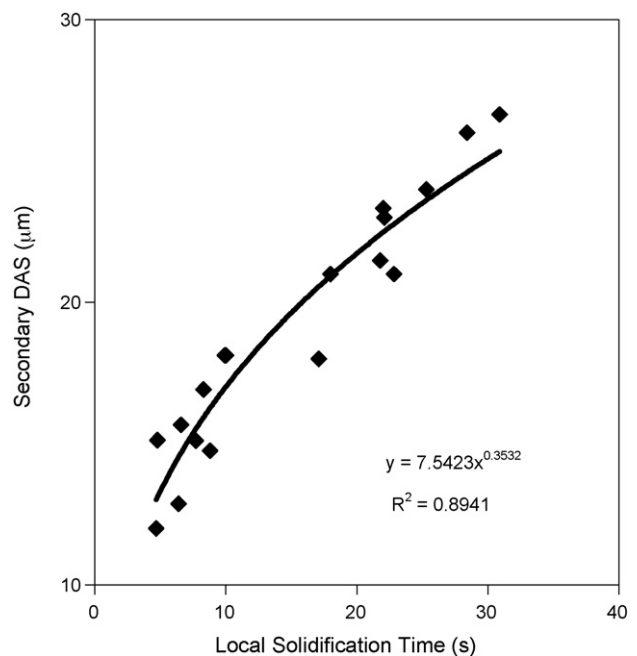


Fig. 6. SDAS vs. local solidification time.

values of SDAS is

$$\lambda_2 = 7.54(t_f)^{0.352} \quad (2)$$

The correlation coefficient for the best-fit equation is found to be 0.89.

A growth model for the eutectic grain, proposed by McDonald et al. was used [13] to estimate grain growth. This model assumes spherical growth during eutectic solidification. The final grain size at the end of solidification depends on growth velocity, which in turn a function of undercooling relative to the equilibrium eutectic temperature. Velocity–undercooling relationships have been established experimentally using unidirectional solidification for both unmodified and modified Al–Si alloys.

Growth velocity is given by

$$V_G = \mu(\Delta T)^n \quad (3)$$

where  $V_G$  is the growth velocity,  $\mu$  the alloy dependent growth constant (0.041 for unmodified alloy and 0.33 for chill/impurity modified alloy),  $n$  the exponent (4 for unmodified alloy and 2 for chill/impurity modified alloy), and  $\Delta T$  is the undercooling from equilibrium temperature.

Typical velocity profile estimated by equation (3) during growth of a eutectic grain is shown in Fig. 7. The growth velocity was not constant throughout solidification but varies significantly. Starting from a low value, in both unmodified and modified alloys, it reached local maximum corresponding to minimum temperature prior to recalescence on the cooling curve. The velocity then dropped to a minimum value corresponding to the maximum solid/liquid interface area. The growth velocity varies inversely with the solid/liquid interface area. With further increase in grain radius, the growth velocity increased continuously till the end of solidification.

Prior to impingement, the radius of a grain that nucleates at time  $t = t_n$  can be calculated by integrating the expression for growth velocity since,  $V_G = dR/dt$ .

Hence

$$R(t) = \int_{t_n}^t \mu(\Delta T)^n dt \quad (4)$$

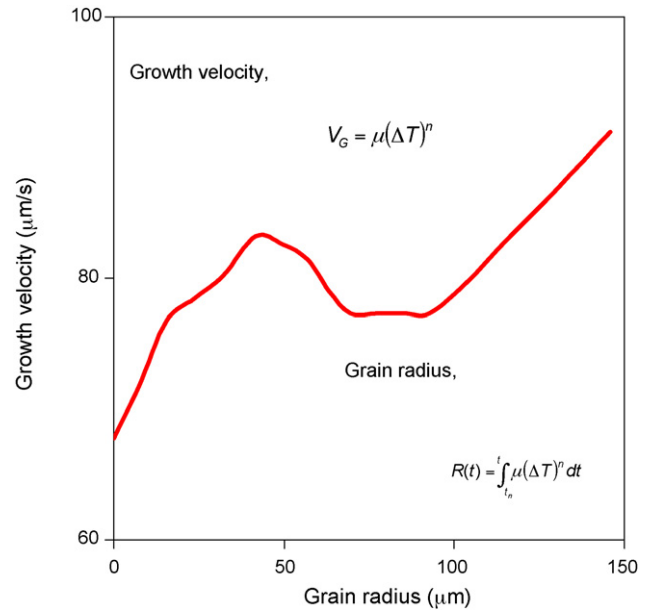


Fig. 7. Growth velocity profile as a function of grain radius.

The average grain size values predicted from the equation were comparable with those obtained from the actual microstructures. For example, for unmodified sample solidified in 50 mm mild steel mould, the predicted average eutectic grain size was about 154  $\mu\text{m}$  where as the measured values from the microstructures were in the range 160–168  $\mu\text{m}$ . Similarly for unmodified alloy casting poured into 50 mm stainless steel mould, the predicted and measured eutectic grain sizes were 62 and 55–60  $\mu\text{m}$ , respectively. Typical microstructures of samples solidified in stainless steel mould are shown in Fig. 8.

Microstructure modification can also be assessed by examining the minimum growth velocities occurred during eutectic growth. In the study made by McDonald et al., fully modified structure was found at growth velocities greater than 7  $\mu\text{m/s}$  where as at velocities lower than this partially modified structures were observed. But, in the present investigation growth velocities lower than 15–20  $\mu\text{m/s}$  resulted primarily in unmodified structure; velocities in the range 20–30  $\mu\text{m/s}$  gave lamellar or near fibrous structure and minimum growth velocities greater

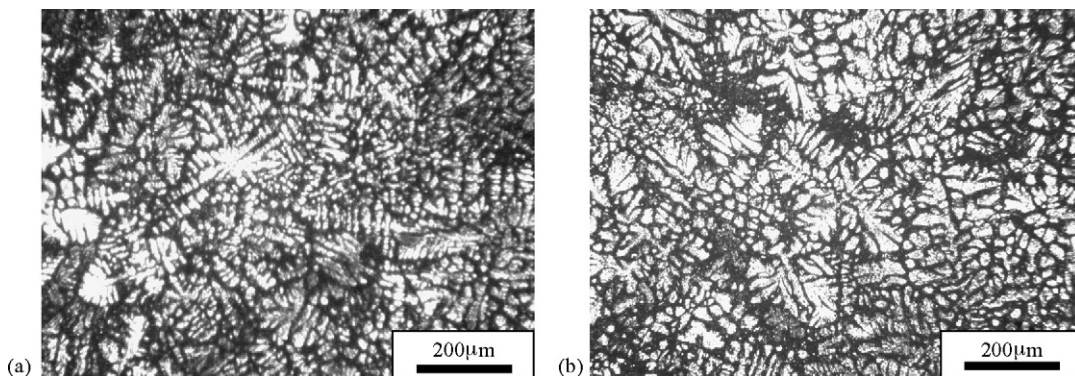


Fig. 8. Photomicrographs showing eutectic cells in casting samples solidified in stainless steel mould: (a) unmodified and (b) modified.

than 30  $\mu\text{m/s}$  generally resulted in fully modified fine fibrous structure.

#### 4. Conclusion

Based on the results and discussion the following conclusions were drawn:

1. Modification melt treatment resulted in higher heat evolution during eutectic solidification.
2. Modification melt treatment also resulted in a significant increase of heat flux transients at the casting/mould wall interface.
3. A good correlation was observed between normalized peak heat flux and normalized eutectic arrest time. A combination of high heat flux and low arrest time yielded fully modified structures.
4. Secondary DAS was found to follow a power relationship with local solidification time. The best-fit equation obtained is  $\lambda_2 = 7.54(t_f)^{0.35}$ .
5. Growth velocities lower than 15–20  $\mu\text{m/s}$  resulted in unmodified structure and velocities of 20–30  $\mu\text{m/s}$  yielded lamellar or near fibrous structure. Minimum growth velocities greater than 30  $\mu\text{m/s}$  resulted in fully modified fine fibrous structure.

#### Acknowledgment

The financial support extended by the Ministry of Human Resources Development (MHRD) Govt. of India, under an R&D project is gratefully acknowledged.

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