CHAPTER 5

EFFECT OF Mn, Be AND Sr AND THEIR INTERACTION WITH Ca IN Al-7Si-0.3Mg-xFe ALLOY

5.1 INTRODUCTION

Manganese is the most commonly used and the least expensive element for Fe neutralization in Al-Si alloys. Be has been found to be the most effective element in improving the mechanical properties of Al-Si-Mg alloys. However, Be is carcinogenic and the higher amounts of Mn required for iron neutralization lead to sludge formation affecting machinability. Further, it has been reported that Mn does not significantly reduce the deleterious effect of Fe on fracture toughness. It is evident from the previous chapter that Ca greatly increases the impact strength of Fe containing Al-7Si-0.3Mg alloy. Hence, a combination of Ca and Mn can improve both the tensile and impact properties of Fe containing Al-7Si-0.3Mg alloy. It has been observed that higher amount of Ca leads to porosity in the castings. A small amount of Be in Al-Si alloys is known to have several benefits. It preferentially oxidizes forming BeO on the surface of the casting, reducing Mg loss and making Mg available for forming the required volume of Mg2Si for strengthening. Further, there are considerable evidences for the Fe neutralizing effect of Sr in Al-Si-Cu alloys. In view of these beneficial effects of individual elements Mn, Be, Sr and Ca, the study mainly focuses on their combined effects.

5.2 METHODOLOGY

Al-7Si-0.3Mg-xFe alloys were prepared in 20 kg capacity crucible using ALTAB Fe compact (75% Fe and 25% Al). For each experiment, about 3 kg of the alloy was melted in a clay-graphite crucible of 5 kg capacity using an electric resistance furnace. The melt was subjected to fluxing and degassing using commercially available fluxes and degassing tablets respectively. Amount of Be to be
added was chosen as per the optimum level reported for the alloy in the literature.\textsuperscript{26} The effect of Be, Mn, Sr and Ca in Al-7Si-0.3Mg alloy both individually and in combination have been studied with an iron content of 0.8Fe. The main reason for the choice of a high Fe content alloy (about thrice the usual limit) was to promote the formation of large intermetallics in the microstructure. It has been found in the previous chapter that the optimum amount of Ca for Fe neutralization lies in the range of 0.03-0.04%. Since the modification effect of Ca and Sr is found to be nearly equal in the present work, the amount of Ca and Sr for the comparison and interaction study has been taken as 0.04%. The melting, casting, heat treatment and testing procedures are given in chapter 3.

5.3 RESULTS

5.3.1 Microstructure

It is evident from chapter 4 that increasing Fe content in Al-7Si-0.3Mg alloy results in the precipitation of long platelet \( \beta \)-phase, whose amount and size increase with increasing Fe content. Figures 5.1 (a) and (b) show the typical as cast microstructures of permanent mould cast Al-7Si-0.3Mg-0.8Fe alloy. It has been seen that \( \beta \)-Fe platelets are present in the interdendritic regions as well as precipitated along with the eutectic Si. Smaller \( \beta \)-needles are branching out from the parent needle spanning across the matrix (Figure 5.1(a) circled area). This network of \( \beta \)-needles extending over large distances in the matrix may affect the feedability of the alloy during casting. Figure 5.1 (b) is a typical micrograph showing eutectic Si nucleating at several locations along the length of a large \( \beta \)-platelet.

5.3.1.1 Effect of Mn, Be and Sr

\textit{Manganese}

Figure 5.2(a) shows the typical as cast microstructure of permanent mould cast Al-7Si-0.3Mg-0.6Fe alloy with 0.3% Mn addition. It is seen that platelet Fe-intermetallic phases have been replaced by Chinese script phases. At a higher amount of Mn (0.5%), Fe-rich intermetallics [Figure 5.2(b)] appear in Chinese script, star like and other compact shapes. On the other hand, 0.4% Mn addition to Al-7Si-0.3Mg-0.8Fe alloy results in both platelets and Chinese script Fe-intermetallics with the
Figure 5.1: Typical microstructures of permanent mould cast Al-7Si-0.3Mg alloy with (a) 0.8% Fe. Circled area showing the branching of β-platelet and (b) Higher magnification of the branched platelet, arrow showing nucleation of eutectic Si by β-platelets
Figure 5.2: Typical microstructures of permanent mould cast Al-7Si-0.3Mg-0.6Fe alloy treated with (a) 0.3% Mn and (b) 0.5% Mn [Arrows show the Mn-Fe phases]

Figure 5.3: SEM micrograph of Al-7Si-0.3Mg-0.8Fe alloy with 0.4% Mn addition
dominance of the latter (Figure 5.3). The EDS analysis of the Mn-Fe compound in Al-
7Si-0.3Mg-0.6Fe alloy treated with 0.3% Mn indicates the presence of Al, Si, Fe and
Mn and their distributions are shown in Figure 5.4. The Mn-Fe phase has been
identified as Al_{15}(Fe,Mn)_{3}Si_{2}. This is in agreement with that reported in literature.^{8}

**Beryllium**

Figures 5.5 (a) and (b) show the typical as cast microstructures of permanent
mould cast Al-7Si-0.3Mg-0.8Fe alloy with Be (0.2%) addition. It is seen that Be
addition replaces the platelet/needle like Fe-intermetallic phases by small spherical
shape and Chinese script Fe-intermetallic phases. It is to be noted that Chinese script
phases exist both inside the primary α-Al dendrites as well as in the interdendritic
regions. It is also observed that Be added sample cast in permanent mould shows
refined Chinese script phases. On the other hand, in the case of sand cast Al-7Si-
0.3Mg-1.0Fe alloy with Be (0.2%) addition, these Chinese script phases are found
only inside the α-Al dendrites (Figure 5.6). This observation is in line with that of
Murali et al.^{26} From the XRD pattern (Figure 5.7), the secondary phases identified in
the Al-7Si-0.3Mg-0.8Fe and Al-7Si-0.3Mg-0.8Fe-0.2Be alloys are (Al_{3}FeSi_{3} and
Al_{3}FeSi) and Al_{2}FeBe_{3} respectively.

**Strontium**

Figures 5.8 (a) and (b) show the typical as cast microstructures of permanent
mould cast Al-7Si-0.3Mg-0.8Fe alloy with strontium addition. It is seen that Sr
modifies the eutectic Si to fine fibrous form and reduces the size of the platelet Fe
intermetallics. It has also been observed that the platelet phases are precipitated along
with the eutectic Si.

**5.3.1.2 Interaction of Mn, Be, Sr and Ca**

**Calcium and Manganese**

To study the combined effect of Ca and Mn, varying amounts of Ca (0.02,
0.05 and 0.08%) have been added to Al-7Si-0.3Mg-0.6Fe-0.3Mn alloy. It has been
observed that addition of 0.02% Ca refines the eutectic Si to fibrous form. Some of
the β-platelets have been fragmented as clearly seen in Figure 5.9 (a). However,
0.05% Ca addition has refined both eutectic Si and platelet Fe-intermetallics
Figure 5.4: EDS elemental X-ray mapping and elemental distribution of Chinese script Mn-Fe phase in Al-7Si-0.3Mg-0.6Fe alloy treated with 0.3% Mn
Figure 5.5: Typical as cast microstructures of permanent mould cast Al-7Si-0.3Mg-0.8Fe-0.2Be alloy

Figure 5.6: Typical microstructure of sand cast Al-7Si-0.3Mg-1.0Fe-0.2Be alloy
Figure 5.7: X-ray diffraction patterns of (a) Al-7Si-0.3Mg-0.8Fe and (b) Al-7Si-0.3Mg-0.8Fe-0.2Be alloys
Figure 5.8: Typical microstructures of permanent mould cast Al-7Si-0.3Mg-0.8Fe alloy treated with 0.04% Sr (a) Optical and (b) SEM micrographs
Figure 5.9: Typical microstructures of permanent mould cast Al-7Si-0.3Mg-0.6Fe-0.3Mn alloy with (a) 0.02 (b) 0.05 and (c) 0.08% Ca
[Figure 5.9 (b)]. On the other hand, a higher amount of Ca addition (0.08%) to Al-7Si-0.3Mg-0.6Fe-0.3Mn alloy leads to coarsening of the β-platelets as shown in Figure 5.9 (c). Further, precipitation of Al-Ca-Si intermetallics has also been observed.

Calcium and Beryllium

Figure 5.10 shows the as cast microstructure of permanent mould cast Al-7Si-0.3Mg-0.8Fe alloy with Be and Ca additions and Figure 5.11 shows the as cast microstructures of sand mould cast Al-7Si-0.3Mg-1.0Fe alloy with Be and Ca additions. It is seen that addition of a trace amount of Be (0.005%) has no significant effect on the microstructure of the alloys. This observation has been confirmed by the average length of eutectic Si obtained by image analysis (Table 5.1). The variation in the average length of β-phase obtained by image analysis as a function of added elements is shown in Figure 5.12. It is clearly evident that the length of the β-phase is reduced by Ca and Be both individually and in combination.

Manganese, Beryllium and Strontium

As in the case of calcium, addition of a trace amount of Be (0.005%) in Sr (0.04%) added Al-7Si-0.3Mg-0.8Fe alloy has no significant effect on the microstructure. Both Mn-Fe and Be-Fe Chinese script intermetallic compounds have been observed (Figure 5.13) in the combined addition of Be (0.15%) and Mn (0.15%) in Al-7Si-0.3Mg-0.8Fe alloy. On the other hand, combined addition of Mn (0.3%) and Sr (0.04%) in Al-7Si-0.3Mg-0.8Fe alloy (Figure 5.14) leads to modification of the eutectic Si to fibrous form and precipitation of Mn-Fe Chinese script phases. Figure 5.15 shows the EDS elemental X-ray mapping and elemental distribution of Chinese script Mn-Fe phase in Al-7Si-0.3Mg-0.8Fe alloy with Be (0.15%)+Mn (0.15%) addition.

The microstructures of Al-7Si-0.3Mg-0.8Fe alloy treated without and with Mn, Be and Sr in T6 condition are shown in Figures 5.16 and 5.17 respectively and there is no significant change in the β-platelets by heat treatment. Similarly, the persistence of the Chinese script intermetallics even after heat treatment of the alloy with Mn and Be additions is also revealed [Figures 5.16 (b) and 5.17 (a)].
Figure 5.10: Typical permanent mould cast microstructure of Al-7Si-0.3Mg-0.8Fe alloy with Be (0.005%) + Ca (0.04%) addition

Figure 5.11: Typical sand mould cast microstructures of Al-7Si-0.3Mg-1.0Fe alloy treated with (a) 0.015% Ca+0.005% Be and (b) 0.008% Ca + 0.007% Be [“1”- Platelet form of iron intermetallic compound]
Table 5.1: Average length of eutectic Si in Al-7Si-0.3Mg-1.0Fe alloy without and with Ca, Be and Ca+Be additions

<table>
<thead>
<tr>
<th>Alloy Code</th>
<th>Composition</th>
<th>Average length of eutectic Si (µm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>CF3</td>
<td>Al-7Si-0.3Mg-1.0Fe</td>
<td>32.6</td>
</tr>
<tr>
<td>CF3C</td>
<td>Al-7Si-0.3Mg-1.0Fe-0.015% Ca</td>
<td>16.8</td>
</tr>
<tr>
<td>CF3B</td>
<td>Al-7Si-0.3Mg-1.0Fe-0.2% Be</td>
<td>31.5</td>
</tr>
<tr>
<td>CF3CB1</td>
<td>Al-7Si-0.3Mg-1.0Fe-0.015% Ca-0.005% Be</td>
<td>18.6</td>
</tr>
<tr>
<td>CF3CB2</td>
<td>Al-7Si-0.3Mg-1.0Fe-0.008% Ca-0.007% Be</td>
<td>20.8</td>
</tr>
</tbody>
</table>

Figure 5.12: Average length of β phase in Al-7Si-0.3Mg-1.0Fe alloy without and with Ca, Be and Ca+Be additions
Figure 5.13: SEM micrograph of Al-7Si-0.3Mg-0.8Fe alloy with Be (0.15%) + Mn (0.15%) addition

Figure 5.14: Typical microstructure of permanent mould cast Al-7Si-0.3Mg-0.8Fe alloy with Mn (0.3%) + Sr (0.04%) addition
Figure 5.15: EDS elemental X-ray mapping and elemental distribution of Chinese script Mn-Fe phase in Al-7Si-0.3Mg-0.8Fe alloy with Be (0.15%) + Mn (0.15%)
Figure 5.16: Typical microstructures of Al-7Si-0.3Mg-0.8Fe alloy in T6 condition
(a) without and (b) with 0.4% Mn
Figure 5.17: Typical microstructures of Al-7Si-0.3Mg-0.8Fe alloy in T6 condition with (a) 0.2% Be and (b) 0.04% Sr
5.3.2 Solidification Behaviour

5.3.2.1 Thermal Analysis

Figures 5.18-5.22 show the cooling curves and their first derivatives and the microstructures (at the centre of the thermal analysis sample, where the thermocouple tip was located) and Table 5.2 gives the alloys, thermal arrests and the phases formed in Al-7Si-0.3Mg-0.8Fe alloy without and with Be (0.2%), Mn (0.4%), Ca (0.04%) and Sr (0.04%) additions respectively. The first derivative is a measure of the instantaneous cooling rate along the cooling curve and is used here to signal the presence of minor slope changes on the curves. For each alloy sample, two cooling curves were taken to check repeatability. Figure 5.18 shows first thermal arrest point at 880.3 K, where the formation and growth of Al nuclei occur. After this arrest, the temperature of the solidifying alloy continues to decrease. Primary Al dendrites grow and the liquid is progressively enriched in Si and Fe. Second thermal arrest occurs at 858 K. This is caused by the latent heat fusion of the β-phase. Next arrest occurs at 845.9 K corresponding to the eutectic Si formation and growth. The final arrest point occurs at 823.8 K when the Mg2Si phase is formed. On the other hand, the peak corresponding to β Fe-intermetallic is completely absent with Be (0.2%) addition. This shows that the β Fe phase has been changed to α-(Be-Fe) phase as shown in the microstructure. However, the peak corresponding to β Fe-intermetallic phase formation is seen in the case of Mn, Ca and Sr additions. This is also evident from the microstructures (Figures 5.19-5.22) showing both platelet and Chinese script phases with Mn addition and refined platelet Fe intermetallics with Ca and Sr additions. Further, Ca and Sr additions decrease the eutectic temperature compared to the untreated alloy and the difference between the eutectic temperature of the untreated and Ca and Sr added sample is 6.6 and 8.7 K respectively. This is due the modification of eutectic Si by Ca and Sr additions to fine fibrous form as shown in the microstructures.

5.3.2.2 Differential Thermal Analysis

Differential Thermal Analysis (DTA) has been carried out in Al-7Si-0.3Mg-0.8Fe, Al-7Si-0.3Mg-0.8Fe-0.2Be and Al-7Si-0.3Mg-0.6Fe-0.5Mn alloys at a heating and cooling rate of 2° K/min and the corresponding DTA curves are given in Figures
Figure 5.18: (a) Cooling curve and its first derivative and (b) microstructures of Al-7Si-0.3Mg-0.8Fe alloy thermal analysis sample.
Figure 5.19: (a) Cooling curve and its first derivative and (b) microstructures of Al-7Si-0.3Mg-0.8Fe-0.2Be alloy thermal analysis sample
Figure 5.20: (a) Cooling curve and its first derivative and (b) microstructures of Al-7Si-0.3Mg-0.8Fe-0.4Mn alloy thermal analysis sample
Figure 5.21: (a) Cooling curve and its first derivative and (b) microstructure of Al-7Si-0.3Mg-0.8Fe-0.04Ca alloy thermal analysis sample (arrows show β-Fe intermetallic phase)
Figure 5.22: (a) Cooling curve and its first derivative and (b) microstructure of Al-7Si-0.3Mg-0.8Fe-0.04Sr alloy thermal analysis sample (arrows show β-Fe intermetallic phase)
Table 5.2: Alloys, thermal arrests and the phases formed in Al-7Si-0.3Mg-0.8Fe alloy without and with Be, Mn, Ca and Sr additions

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Thermal arrests</th>
<th>Phases formed</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Peak</td>
<td>Temperature (°K)</td>
</tr>
<tr>
<td>Al-7Si-0.3Mg-0.8Fe</td>
<td>A</td>
<td>880.3</td>
</tr>
<tr>
<td></td>
<td>B</td>
<td>858.0</td>
</tr>
<tr>
<td></td>
<td>C</td>
<td>845.9</td>
</tr>
<tr>
<td></td>
<td>D</td>
<td>823.8</td>
</tr>
<tr>
<td>Al-7Si-0.3Mg-0.8Fe-0.2Be</td>
<td>A</td>
<td>880.7</td>
</tr>
<tr>
<td></td>
<td>B</td>
<td>838.1</td>
</tr>
<tr>
<td></td>
<td>C</td>
<td>823.7</td>
</tr>
<tr>
<td>Al-7Si-0.3Mg-0.8Fe-0.4Mn</td>
<td>A</td>
<td>879.6</td>
</tr>
<tr>
<td></td>
<td>B</td>
<td>856.5</td>
</tr>
<tr>
<td></td>
<td>C</td>
<td>845.8</td>
</tr>
<tr>
<td></td>
<td>D</td>
<td>823.8</td>
</tr>
<tr>
<td>Al-7Si-0.3Mg-0.8Fe-0.04Ca</td>
<td>A</td>
<td>882.1</td>
</tr>
<tr>
<td></td>
<td>B</td>
<td>855.9</td>
</tr>
<tr>
<td></td>
<td>C</td>
<td>839.3</td>
</tr>
<tr>
<td></td>
<td>D</td>
<td>826.4</td>
</tr>
<tr>
<td>Al-7Si-0.3Mg-0.8Fe-0.04Sr</td>
<td>A</td>
<td>879.3</td>
</tr>
<tr>
<td></td>
<td>B</td>
<td>857.2</td>
</tr>
<tr>
<td></td>
<td>C</td>
<td>837.2</td>
</tr>
<tr>
<td></td>
<td>D</td>
<td>826.7</td>
</tr>
</tbody>
</table>
Figure 5.23: DTA curves of Al-7Si-0.3Mg-0.8Fe alloy (a) Heating and (b) Cooling at 2°K/min.
Figure 5.24: DTA curves of Al-7Si-0.3Mg-0.8Fe alloy with 0.2% Be addition
(a) Heating and (b) Cooling at 2°K/min.
Figure 5.25: DTA curve of Al-7Si-0.3Mg-0.6Fe alloy with 0.5% Mn addition
(a) Heating and (b) Cooling at 2°C/min.

(a)

(b)
5.23-5.25. The peak corresponding to platelet Fe intermetallic could not be detected since it merges with the eutectic Si reaction as shown in the cooling curve (Figure 5.23). It is seen from the Figure 5.24 that the temperature of Be-Fe phase dissolution (heating curve) and the precipitation (cooling curve) are higher than the peak corresponding to β-Fe intermetallic phase formation (858 K from thermal analysis). As a result, the amount of free iron available to precipitate in the form of β-Al₅FeSi is diminished. This is supported from the microstructural observation of the replacement of platelet Fe intermetallics by Chinese script phases. Similarly, the peak at 863 K (heating curve) in Figure 5.25 corresponds to Chinese script Mn-Fe-intermetallic phase reaction. The observation of sludge formation at higher Mn (0.5%) addition in the microstructure is supported from the appearance of a peak at 900 K (cooling curve).

5.3.3 Physical characteristics

5.3.3.1 Porosity

Table 5.3 shows the effect of Ca, Mn and Ca + Mn on the porosity and electrical conductivity of Al-7Si-0.3Mg-0.6Fe alloy. It is seen that % porosity is increased with Ca (0.05%) addition compared to the untreated and other additions studied. The % porosity is lower with Mn followed by Ca + Mn additions. This may be due to the morphological change of large platelet Fe-intermetallic compounds to star like, Chinese script and other compact shapes. This result is in accordance with the work of Samuel et al\textsuperscript{195} on the effect of Be, Mn, Cr and Sr on 319 (Al-6.2Si-3.7Cu) alloy, which has shown that the contours of the script phase facilitate the filling up of the liquid metal in between the dendrite arms of the phase (compared to the blocking action of the β-needles/platelets) and thereby reducing the amount of porosity.

5.3.3.2 Electrical conductivity

The electrical conductivity of Ca added samples are higher than those of the untreated, Mn and Ca+Mn added samples (Table 5.3). This is due to the modification of eutectic Si by Ca addition to fibrous form, which allows easier flow of electrons compared to the platelet morphology.
Table 5.3: % porosity and electrical conductivity of permanent mould cast Al-7Si-0.3Mg-0.6Fe alloy without and with Ca, Mn and Ca + Mn additions

<table>
<thead>
<tr>
<th>Alloy Code</th>
<th>Addition</th>
<th>% Porosity</th>
<th>Electrical conductivity</th>
</tr>
</thead>
<tbody>
<tr>
<td>CF1</td>
<td>No addition</td>
<td>0.82</td>
<td>34</td>
</tr>
<tr>
<td>CF1C</td>
<td>0.005% Ca</td>
<td>1.27</td>
<td>36</td>
</tr>
<tr>
<td>CF1M</td>
<td>0.3% Mn</td>
<td>0.78</td>
<td>33</td>
</tr>
<tr>
<td>CF1MC</td>
<td>0.05% Ca +0.3% Mn</td>
<td>0.86</td>
<td>34</td>
</tr>
</tbody>
</table>

5.3.4 Mechanical Properties

5.3.4.1 Tensile properties

The ultimate tensile strength and % elongation of Ca, Mn and Ca+Mn added Al-7Si-0.3Mg-0.6Fe alloy cast in permanent mould are compared in Figures 5.26 (a) and (b) respectively. Mn (0.3%) addition has increased the ultimate tensile strength (UTS) nominally [4.2% improvement] compared to the untreated alloy. However, the addition of Ca and Ca+Mn has decreased the UTS by 11 and 8.8% respectively compared to untreated alloy. It has also been observed that the addition of Ca and Mn both individually and in combination increases the elongation of the alloy [Figures 5.26 (b)] compared to the untreated alloy. However, the maximum improvement in elongation (140%) is obtained with Mn (0.3%) followed by Ca+Mn and Ca addition, the latter showing only a nominal increase.

Figures 5.27 (a) and (b) show the variation of ultimate tensile strength and % elongation of Al-7Si-0.3Mg-1.0Fe alloy with Ca, Be and Ca + Be additions. Be (0.2%) addition has improved the UTS (26.3%) and elongation (71.4%) due to the morphological change of β needles to Chinese script form. The results obtained in the present study are in good agreement with the published data26 concerning the effectiveness of Be as an iron neutralizer in Al-7Si-0.3Mg alloy. It has been observed
Figure 5.26: Effect of Ca, Mn and Ca+Mn additions on the (a) UTS and (b) % elongation of Al-7Si-0.3Mg-0.6Fe alloy cast in permanent mould
Figure 5.27: Effect of Ca, Be and Ca+Be additions on the (a) UTS and (b) % elongation of Al-7Si-0.3Mg-1.0Fe alloy cast in permanent mould
that a trace amount of Be addition to Al-7Si-0.3Mg-1.0Fe-0.015% Ca alloy improves both the UTS and elongation compared to Al-7Si-0.3Mg-1.0Fe and Al-7Si-0.3Mg-1.0Fe-0.015Ca alloys. The maximum improvement in tensile strength (34.5%) is obtained with Ca (0.008%)+Be (0.007%) addition. However, % elongation is low compared to the Be added alloy. This is due to insufficient amount of Ca to reduce the length of β needles completely. Hence, a higher amount of Ca with trace Be may give the best combination of tensile properties.

The ultimate tensile strength and % elongation of Al-7Si-0.3Mg-0.8Fe alloy treated with Be, Ca, Mn and Sr individually or in combination and cast in permanent mould are shown in Figures 5.28 (a) and (b) respectively. It has been observed that all additions except Ca, Sr and Mn increase the UTS compared to the untreated alloy. The highest improvement in UTS has been observed in the combined addition of Be (0.15%) and Mn (0.15%) to Al-7Si-0.3Mg-0.8Fe alloy. On the other hand, Ca (0.04%), Sr (0.04%) and Mn (0.4%) additions have reduced the UTS by 9.2, 5.3 and 8.6% respectively compared to that of untreated Al-7Si-0.3Mg-0.8Fe alloy. It has also been observed that the addition of Be, Ca, Sr and Mn both individually and in combination increases the % elongation of the alloy [Figures 5.28 (b)] compared to the untreated alloy. This is due to the morphological change of platelet Fe intermetallics to Chinese script form with Be and Mn additions and the refinement of platelet Fe phases with Ca and Sr additions respectively. Addition of Mn (0.3%) to Al-7Si-0.3Mg-0.8Fe-0.04% Ca and Al-7Si-0.3Mg-0.8Fe-0.04% Sr alloys leads to 75 and 167% improvement in elongation respectively compared to individual addition of Ca and Sr.

Be (0.2%) addition has shown significant improvement in elongation (125%). It is interesting to note that trace amount of Be (0.005%) with Ca and Sr (0.04%) has improved the UTS (2 and 5.5% respectively) and elongation (100 and 167% respectively) compared to the untreated and Ca and Sr alone added alloys.

5.3.4.2 Impact Strength

Figure 5.29 shows the impact strength of Al-7Si-0.3Mg-0.6Fe alloy with and without Ca, Mn and Ca+Mn additions. As reported in the previous chapter, impact
A: Al-7Si-0.3Mg-0.8Fe
B: Al-7Si-0.3Mg-0.8Fe +0.2% Be
C: Al-7Si-0.3Mg-0.8Fe +0.04% Ca
D: Al-7Si-0.3Mg-0.8Fe + 0.04% Sr
E: Al-7Si-0.3Mg-0.8Fe +0.4% Mn
F: Al-7Si-0.3Mg-0.8Fe +0.3% Mn+ 0.04% Ca
G: Al-7Si-0.3Mg-0.8Fe + 0.3% Mn+0.04% Sr
H: Al-7Si-0.3Mg-0.8Fe +0.005% Be+0.04% Ca
I: Al-7Si-0.3Mg-0.8Fe + 0.005% Be+ 0.04% Sr
J: Al-7Si-0.3Mg-0.8Fe +0.15% Be+0.15% Mn

Figure 5.28: Effect of Be, Mn, Sr and Ca individually and in combination on the (a) UTS and (b) % elongation of Al-7Si-0.3Mg-0.8Fe alloy
Figure 5.29: Effect of Ca, Mn and Ca+Mn additions on the impact strength of Al-7Si-0.3Mg-0.6Fe alloy
strength has been greatly improved (80%) by Ca addition. However, only 27% improvement in impact strength has been observed in Mn added alloy. Ca+Mn addition shows 85% improvement. Further, the microstructural features observed in the fractographs of impact tested alloys [Figures 5.30 (a-d)] support these results. Ca added alloy shows more dimples, a characteristic of ductile fracture [Figure 5.31 (b)]. On the other hand, fractograph of Mn added alloy [Figure 5.31 (c)] shows the presence of a large number of cleavage fractured surface typical of a brittle failure.

The effect of Be, Mn, Sr and Ca addition individually and in combination on the impact strength of T6 heat-treated Al-7Si-0.3Mg-0.8Fe alloy is shown in Figure 5.31. It has been found that all additions except Mn improve the impact strength compared to the untreated alloy. However, combined addition of Mn with Ca, Sr and Be improves the impact strength.

5.4 DISCUSSION

The large size of β-Al₃FeSi platelet observed in the untreated alloys indicates that it must have formed at a higher temperature, i.e., as a pre-eutectic reaction (providing a longer time for its growth). On the other hand, the much smaller size of the β-platelets observed indicates that they are very likely the products of a co-eutectic reaction. It has been reported[144, 200, 201] that unlike the Si eutectic temperature, which is only marginally affected by variations in cooling rate, the β-phase start temperature decreases with decreasing iron content, increasing cooling rate and increasing melt superheat until it eventually merges with the Si eutectic temperature. Though the β-phase continues to crystallize until the end of the silicon eutectic reaction, the length of the primary β-phase needles greatly depends on the time interval between the β-phase start temperature and the silicon eutectic temperature. With decreasing β-phase start temperature, the β-phase growth time and therefore the length and volume fraction of this phase decrease until the β-phase start temperature merges with the silicon eutectic temperature.

It has been observed that the eutectic Si nucleates on the surface of the β-platelets [Figure 5.1 (b)]. This is supported from the work of Taylor et al.[272] It has
Figure 5.30: Fractographs of impact tested Al-7Si- 0.3Mg- 0.6Fe alloy (a) without and with (b) 0.05% Ca (c) 0.3%Mn and (d) 0.3% Mn+0.05% Ca
Figure 5.31: Effect of Be, Mn, Sr and Ca individually and in combination on the impact strength of Al-7Si-0.3Mg-0.8Fe alloy
been reported that the Si particles are often observed to grow from multiple locations along a single β-platelet. For eutectic Si to nucleate and grow on the β-platelets, the latter must form and present prior to the former. This is the case for the platelets that form as either a primary β-phase or as binary β-phase (a component of the Al-β-Al₅FeSi binary eutectic). Such platelets are well developed by the time the ternary eutectic Si begins to grow. This has been confirmed by the appearance of a peak at 858 K in Figure 5.18 corresponding to β-platelets formation.

Addition of Mn changes the morphology of platelet Fe intermetallic to Chinese script. This is because of the replacement of Mn by Fe, which can substitute each other, being transitional elements. However in the case of 0.5% Mn addition, agglomeration of the script and star like particles takes place [Figure 5.2 (b)]. In order to obtain the crystallization of the iron compound in the Chinese script form and avoid the needle like and polyhedral crystal morphology, a certain critical ratio of iron: manganese is required and this ratio depends on the cooling rate. Bakerud et al. have clearly explained the solidification sequence (Figure 2.13) with Mn addition in Al-7Si-0.3Mg alloy. According to Mondolfo, (Fe,Mn)Al₆ is the first phase to form in the Al-Fe-Mn-Si system, which encompasses many commercial alloys. In many alloys, (Fe,Mn)Al₆ reacts peritectically with the liquid to form (FeMn)₃Si₂Al₁₅. The rate of peritectic transformation depends on the equilibrium diagram and kinetic factors.

Be (0.2%) has changed the morphology of platelet Fe compound to Chinese script and fine globules. Both Mn and Be additions lead to the precipitation of either coarse or fine Chinese script form of intermetallic phase with Fe at slow (thermal analysis sample, cooling rate 1.5⁰ K/min) and fast cooling rates (permanent mould) respectively. It has also been observed in the present work that the Chinese script phases with Be addition forms only inside the α-Al at slow cooling rate and both inside the α-Al and in the interdendritic areas at high cooling rates. Murali et al. have reported that in the case Be (0.27%) added sand cast Al-7Si-0.3Mg alloy, the Be-Fe phase is formed inside the α-Al, which is probably the result of a peritectic reaction. It has been observed from the DTA curves (Figures 5.24 and 5.25) that the Be-Fe and Mn-Fe phases precipitate at temperatures much higher than the formation
temperature of the $\beta$-phase. As a result, the amount of free iron available to precipitate in the form of $\beta$-$\text{Al}_5\text{FeSi}$ is diminished leading to the formation of Chinese script phase.

Be (0.2%) addition has significantly improved the tensile and impact properties. This is due to the morphological change of $\beta$ needles to Chinese script form. This has been further confirmed from the cooling curve of Al-7Si-0.3Mg-0.8Fe-0.2Be alloy showing the absence of a peak corresponding to $\beta$-Fe intermetallic compared to the cooling curve without Be addition (Figures 5.18 and 5.19). The microstructures of Mn and Be added alloys in the T6 treated condition show the persistence of Chinese script phases. Further, it has been reported\textsuperscript{277} that $\alpha$-$\text{AlFeMnSi}$ particles have quite perfect crystals and hence the dissolution process is hindered by the lack of defects.

The improvement in elongation and impact strength with Sr addition is due to the fragmentation of the platelet Fe intermetallics and modification of the eutectic Si. This observation is in accordance with the work of Samuel et al\textsuperscript{195} on Sr addition to Al-Si-Cu (319) alloy. The fragmentation of platelet Fe-intermetallic phase is clearly explained in the previous chapter.

It has been found that Mn addition improves the tensile properties. On the other hand, impact strength is not significantly improved. However, Mn in combination with Ca, Sr and Be improves the impact strength significantly. This is due to the modification of the eutectic Si from acicular to fibrous form. The most important observation in the present work is the attainment of the best combination of tensile and impact properties by Ca+Mn, Sr+Mn and Be+Mn additions to Al-7Si-0.3Mg-xFe alloys. Similarly, combined addition of a trace amount of Be (0.005%) to Ca and Sr improves the tensile properties significantly. This is due the fact that Be in trace amount in Al-Si alloys preferentially oxidizes forming BeO on the surface of the casting, cleans the melt surface and reduces gas pick up (and hence porosity) and Mg loss making Mg available for forming the required volume of $\text{Mg}_2\text{Si}$ for strengthening.\textsuperscript{250}
5.5 SUMMARY

1. Mn (0.3%) addition to Al-7Si-0.3Mg-xFe alloy changes the morphology of platelet iron phase to script form leading to significant improvement in tensile properties. However, there is no significant improvement in impact strength.

2. Be (0.2%) changes the platelet morphology of iron phase to Chinese script form, which are seen only inside the α-Al in slow cooling (sand cast), both in the interdendritic as well as inside the α-Al in fast cooling (permanent mould casting). This morphological change is responsible for the significant improvement in both the tensile and impact properties.

3. Combined additions of Ca + Mn, Be + Mn and Sr + Mn lead to improvement in both tensile and impact properties compared to individual additions and a synergistic effect of both the elements is achieved.

4. A trace amount of Be (0.005%) addition to Ca and Sr leads to superior tensile properties compared to Ca and Sr additions alone.