# ELECTRICAL CONDUCTIVITY STUDIES OF CAST AI-SI AND AI-SI-Mg ALLOYS

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#### ABSTRACT

Cast Al-Si and Al-Si-Mg alloys containing up to 12.6 wt. pct. silicon and 1.0 wt. pct. magnesium were prepared. The changes in electrical conductivity/resistivity of these alloys due to strontium additions have been investigated and explained in terms of variations in microstructure. The conductivity behaviour of strontium-containing and strontium-free alloys was found to exhibit marked differences, depending on the silicon and magnesium contents and the rate of solidification. The electrical conductivity of single phase alloys containing less than 1 60 wt. pct. Si decreased with increasing silicon and magnesium levels. However, strontium had no effect on the conductivity of these solid solution alloys since it does not dissolve appreciably in the aluminum matrix or change the solid solubility of silicon and magnesium in aluminum. Silicon precipitation processes in the supersaturated solid solution alloys of Al-Si and Al-Si-Sr have been examined using the Johnson-Mehl-Avrami equation and found to be isokinetic. Strontium, however, retarded the growth rate of silicon precipitates. Strontium did not affect the kinetics of G.P. zone formation in Al-Si-Mg alloys but it suppressed the formation of stable Mg<sub>2</sub>Si precipitates during subsequent aging at 175°C. Unlike the single phase alloys, two phase Al-Si and Al-Si-Sr alloys, in the range of 2.0 to 12.6 wt. pct. Si, exhibited different electrical conductivity behaviour. The strontium-containing alloys \* showed a higher conductivity than alloys with no strontium, and this conductivity difference increased as the silicon and magnesium contents were increased and the solidification rate was decreased. It has been demonstrated that this difference is due to changes in the silicon morphology. Electron scattering at the interface between the aluminum matrix and the eutectic silicon phase contributes significantly more to the resistivity of unmodified alloys than that of modified alloys. In addition, the resistivity of directionally solidified Al-Si alloys, measured in directions parallel and perpendicular to the silicon rods, revealed anisotropy and a marked dependence on the microstructural form of the eutectic silicon. The solution treatment of two-phase Al-Si and Al-Si-Mg alloys was divided into three stages: precipitation of silicon and Mg<sub>2</sub>Si, dissolution of these

precipitates and spheroidization of the eutectic silicon phase. It was found that the precipitation and spheroidization processes result in an increase in conductivity while the dissolution process decreases the conductivity of these alloys.

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RESUME

Des alljages de fonderie Al-Si et Al-Si-Mg contenant jusqu'à 12.6% en poids de Silicium et 1.0% en poids de Magnésium ont été préparés. Les variations de leur conductivité électrique/résistivité, dues aux additions de Strontium, ont été étudiées et expliquées en termes de changements microstructuraux. Les comportements en conductivité électrique des alliages avec ou sans Strontium ont été trouvés présentant des différences marguées, selon la teneur en Si et Mg et la vitesse de solidification. La conductivité des alliages à une seule phase contenant moins que 1.6% en poids de Si diminue avec des teneurs croissantes en Si et Mg. Cependant, le Strontium n'a pas d'effet sur ces alliages de solution solide, étant donné qu'il ne se dissout pas de façon, appréciable dans la matrice d'Aluminium, ou qu'il ne change pas la solubilité solide du Silicium et du Magnésium dans l'Aluminium Les processus de précipitation du Silicium dans les alliages de solution solide supersaturée Al-Si et Al-Si-Sr ont été examinés en utilisant l'équation Johnson-Mehl-Avrami et ont été trouvés être isocinétiques. Le Strontium, cependant, n'a pas affecté les cinétiques de formation de la zone G.P. dans les alliages Al-Si-Mg mais a supprimé la formation des précipités stables Mg<sub>2</sub>Si lors du vieillissement ultérieur à 175°C. Contrairement aux alliages à une phase, les alliages à deux phases Al-Si et Al-Si-Sr (dans la gamme 20 à 12.6% en poids de Si), ont présenté des comportements différents pour la conductivité électrique. La conductivité des alliages contenant du Strontium est plus élevée que celle obtenue pour les alliages sans Strontium, cette différence en conductivité croît lorsque les teneurs en Si et Mg sont augmentées et lorsque la vitesse de solidification est diminuée II a été démontré que cette différence est due aux changements de la morphologie du Silicium. La diffusion électronique à l'interface entre la matrice d'Aluminium et la phase de Silicium eutectique contribue de façon plus significative à la résistivité des alliages non modifiés qu'à celle des alliages De plus, la résistivité des alliages Al-Si solidifiés modifiés. directionnellement, mesurée dans des directions parallèles et perpendiculaires aux bâtonnets de Silicium a révèlé l'anisotropie et une dépendance très marquée de la forme microstructurale du Silicium

eutectique. Le traitement de mise en solution des alliages à deux phases Al-Si et Al-Si-Mg a été divisé en trois étapes: précipitation du Silicium et de Mg<sub>2</sub>Si, dissolution de ces précipités et sphéroidisation de la phase de Silicium eutectique. Il a été constaté que les processus de précipitation et de sphéroidisation entrainent une augmentation de la conductivité tandis que le processus de dissolution diminue la conductivité de ces alliages.

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12.6 oranına kadar silisyum ve 1.0 oranına kadar magnezyum içeren dökme Al-Si ve Al-Si-Mg alaşımarı hazırlandı. Strontiyum ilavesinden oturu, bu<sup>r</sup>alaşımların elektriksel geçirgenliklerindeki değişimler incelenip, mikroyapıdaki farklılıklar açısından izah edildi. Strontiyum ihtiva eden ve etmeyen alaşımların, içerdikleri silisyum ve magnezyum oranına ve de katılaşma hızına bağlı olarak farklı elektriksel geçirgenlik özellikleri gösterdikleri ızlenmıştir. 1.6 oranından az silisyum ihtiva eden tek fazlı alaşımların elektriksel geçirgenlikleri silisyum ve magnezyum miktarlarının artması ile azalmıştır. Bununla beraber, strontiyum'un bu tek fazlı alaşımların geçirgenliğine bir etkisi olmamıştır. Çünku, strontıyum katı aluminyum içersinde çözülmemiş, ve de silisyum ve magnezyum'un çozunme miktarlarınıda değiştirmemiştir. Fazlasıyla doymuş Al-Si ve Al-Sı-Sr katı alaşımlarındaki silisyum çökelmesi Johnson-Mehl-Avramı formulu ile 🖉 incelenmiş ve kinetiksel olarak aynı olduğu bulunmuştur. Fakat, çokulmuş olan silisyum partıküllerinin büyüme hızı Al-Si-Sr alaşımlarında strontıyum tarafından yavaşlatılmıştır. Strontiyum'un Al-Si-Mg alaşımlarında gozlenen G.P. bölgelerinin oluşumunun kinetiğini etkilemediği fakat daha sonra 175°C' deki yapay yaşlandırma işlemi sırasında, kararlı Mg<sub>2</sub>Si partıkullerinin oluşumunu engellediği gozlenmiştir. Tek fazlı alaşımlardan farklı olarak, ıkı fazlı (2 ile 12.6 oranında sılısyum içeren) Al-Si ve Al-Si-Sr alaşımları farklı elektriksel geçirgenlik ozellikleri göstermişlerdir. Strontiyum içeren, alaşımlar Sr-içermeyen alaşımlardan daha fazla elektriksel geçirgenlige sahip oldukları gözlenmiş, ve bu farklılığın silisyum ve magnezyum miktarları arttırıldıkça, ve de katılaşma hızının düşürülmesi ile artığı bulunmuştur. Ayrıça bu farklılığın otektik silisyum fazinin şeklindeki değişikliklerden kaynaklandığı isbat edilmiştir. Elektron yansıması modife edilmemiş alaşımlarda elektivik) dirençini modife edilmiş alaşımlara kıyasla daha fazla artırmaktadır Bunun dışında, yönlü katılaştırılmış Al-Si alaşımlarının elektriksel dirençleri silisyum fazına paralel ve dik olarak olçülmüştür. Elektriksel direnç yapısal yönlenmeye ve silisyum mikroyapısına bağlılık göstermiştir. İki fazlı Al-Sı ve Al-SI-Mg alaşımlarının solusyon işlemleri üç aşamaya bölünmüştür: Sılısyum ve Mg2Si çökelmesi; bu bileşiklerin tekrar çózünmesi; ve ótektik sılısyum fazının küreleşmesi. Çökelme ve küreleşme işlemleri elektriksel geçirgenliği arttırdığı, bileşiklerin tekrar çozünmesi ise geçirgenliği azalttığı bulunmuştur

ÖZET

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### NOMENCLATURE

A cross-sectional area

a interatomic distance.

d depth of penetration

E<sub>m</sub> activation energy of defect motion

e charge of an electron

f frequency of alternating current

G temperature gradient .

H<sub>a</sub> applied alternating magnetic field

H<sub>i</sub> induced alternating magnetic field

I current

Is intensity of (220)Si peak obtained from the sand cast sample

Id intensity of (220)Si peak obtained from directionally solidified samples

XVIII

k rate constant

ko Boltzmann's constant

L distance

mean free path of electrons

lo impurity component of the mean free path

IT total mean free path of electrons

lth thermal component of the mean free path

M<sub>s</sub> martensite transformation start

m mass of an electron

N number of electrons

rt Avramı number

R resistance

R<sub>q</sub> growth rate

RRR residual resistivity ratio

r atomic radius

T absolute temperature

T<sub>A</sub> annealing temperature

T<sub>E</sub> eutectic growth temperature

T<sub>L</sub> liquidus temperature

T<sub>N</sub> eutectic nucleation temperature

ΔT<sub>E</sub> depression of the eutectic temperature

 $\Delta T_N$  undercooling prior to eutectic phase nucleation

T<sub>4</sub> solution treatment

T<sub>6</sub> solution treatment and artificial aging

t time

V voltage

 $\Delta V$  voltage drop

V<sub>d</sub> volume fraction of the dispersed phase

V<sub>f</sub> volume fraction of fibers

V<sub>s</sub> volume fraction of the second phase

 $V_{\alpha}$  volume fraction of  $\alpha$ -phase

 $V_{\beta}$  volume fraction of  $\beta$ -phase

V<sub>F</sub> velocity of the electrons a<sup>t</sup> the Fermi level

V<sub>Si</sub> volume fraction of the eutectic silicon

X<sub>1</sub> atomic fraction of impurities

xt volume of precipitate formed at time t

ΔZ difference in valence

Greek Symbols

μ magnetic permeability

ρ electrical resistivity

ρ<sub>A</sub> resistivity of alloys

p<sub>mi</sub> resistivity of mixture

ρ<sub>o</sub> \* impurity component of resistivity

pth thermal component of resistivity

ρ<sub>L</sub> longitudinal resistivity

pT transverse resistivity

 $\hat{\mathbf{p}}_{\mathrm{H}}$  resistivity measured parallel to fibers

 $\rho_{\perp}$  resistivity measured perpendicular to fibers

 $\rho_{\alpha}$  resistivity of  $\alpha$ -phase

 $\rho_{\beta}$  resistivity of  $\beta$ -phase

 $\Delta \rho^{c}_{AI}$  calculated resistivity change

 $\Delta \rho^{m}_{AI}$  measured resistivity change

σ electrical conductivity

σ<sub>AC</sub> conductivity of as-cast samples

 $\sigma_{AI}$  conductivity of aluminum solid solution

 $\sigma_{AI-Si}$  conductivity of two-phase AI-Si alloys

 $\sigma_c$  conductivity of continuous phase

 $\sigma_{co}$  conductivity of composite

 $\sigma_d$  conductivity of dispersed phase

σ<sub>F</sub> final conductivity

σ<sub>f</sub> conductivity of fiber

σ<sub>m</sub> conductivity of matrix

σ<sub>mi</sub> conductivity of mixture

σ<sub>o</sub> initial conductivity

 $\sigma_{mo}$  conductivity of modified alloys

 $\sigma_s$  conductivity of the second phase

 $\sigma_{ST}$  conductivity of the solution treated samples

σt conductivity at time t

 $\sigma_{um}$  conductivity of unmodified alloys

 $\sigma(+)$  upper limit for conductivity

 $\sigma_{(-)}$  lower limit for conductivity

 $\Delta \sigma$  conductivity difference between modified and unmodified alloys

 $\dot{\Delta}\sigma_{R}$  relative conductivity change

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**CHAPTER 1** 

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INTRODUCTION

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#### 1.1 <u>Aluminum-Silicon Casting Alloys</u>

The commercial importance of Al-Si alloys results from their excellent castability, corrosion resistance, machinability, weldability, light weight and pressure tightness. Aluminum-silicon alloys can be cast into very complex shapes with close tolerances by virtually all casting processes because of their excellent casting characteristics such as high fluidity and resistance to hot cracking.

The AI-Si binary system forms a simple eutectic at 577°C and 12.6 wt. pct. Si between aluminum containing a maximum of 1 62 wt. pct. Si in solid solution and nearly pure silicon<sup>1</sup> as shown in Figure 1.1. According to their silicon content AI-Si casting alloys can be classified into three groups:

a) Hypoeutectic alloys (2 to 10 wt. pct. Si)

b) Near-eutectic alloys (10 to 13 wt. pct. Si)

c) Hypereutectic alloys (13 to 25 wt. pct. Si)

1.1.1 Hypoeutectic and Near-Eutectic Alloys

Although the silicon content in these groups of alloys can be as low as 2 wt. pct. Si, commercially important alloys contain higher than 5 wt pct. Si. The chemical composition of some of the common alloys and the methods whereby they are cast are given in Table 1.12.

Small amounts of magnesium (0.2 to 1.5 wt. pct.) and copper (4 wt pct. max.) are added to Al-Si casting alloys such as A356, A357 and A413 in order to enhance their response to heat treatment. Various heat treatment procedures of these alloys provide a wide range of improved mechanical and physical properties, and consequently they have found extensive use in a variety of applications in aircraft, automobiles and military equipment.



Figure 1.1 Al-Si phase diagram.(1)

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Aluminum	Element (wt. pct.)					Casting	
Assoc. No.	Si	Fe	Cu	Mg	Zn	other	Process†
319.0	6.0	<1.0	3.5	<0.10	<1.0		S,P
A319.0	6.0	<1.2	3.5	0.3	<1.0		
333.0	9.0	<1.0	3.5	0.3	<1.0		Р
A333.0	9.0	<b>&lt;</b> 1. <b>0</b>	3.5	0.3	, 3.0	م. مرکن	
A335.0	5.0	< 0.09	1.25	0.5	<0.05	<del></del>	S,P
C335.0	5.0	<0.20	1.25	0.5	<0.10	-	
356.0	70	< 0.60	<0.25	0.35	<0.35		S,P
A356.0	7.0	<0.20	< 0.20	0.35	<0.10	-	
357.0	7.0	<b>&lt;</b> 0.20	<0.20	0.55	<0.10		S,P
A357.0	70	<0.09	<0.05	0 55	<0.05	0.05Be	
380.0	8.5	<2.0	3.5	<0.10	3.0		D
A380.0	8 5	<1.3	3.5	<0.10	3.0	-	
413.0	12.0	<2.0	1.0	<0.10		-	D
A413.0	12.0	<1.3	1.0	<0.10			O
444.0	5.25	<0.6	<0.25	<0.10	<b>&lt;</b> 0.35		Ś,P
A444.0	5.25	<0.20	<0.10	<0.05	<0.10	<b>~</b> '	

 Table 1.1 Chemical Composition of Common Hypoeutectic

 and Near-Eutectic Al-Si Alloys (2)

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S - Sand Casting P - Permanent Mould D - Die Casting

The eutectic silicon in cast hypoeutectic and eutectic alloys appears as large, brittle, acicular flakes and plates which are undesirable as they lead to a substantial reduction of strength and elongation of the castings. Modification of the eutectic silicon alters the silicon morphology from acicular to fine fibrous, as illustrated in Figure 1.2. This microstructural change results in a noticeable improvement in strength, and a significant increase in elongation.<sup>3-10</sup> Table 1.2 outlines the effect of modification on the mechanical properties of several Al-Si casting alloys.

Eutectic silicon modification is achieved by rapid solidification or addition of a suitable modifying agent such as sodium, strontium or antimony The theory of the modification process will be discussed in the second chapter of this thesis.

1.1.2 Hypereutectic Alloys

This group of Al-Si alloys provides a combination of outstanding wear resistance and lower thermal expansion coefficient. One of the most popular hypereutectic alloys, alloy 390 (17 wt. pct. Si, 1.5 wt. pct. Cu, 0.5 wt. pct. Mg) has been successfully used to produce all-aluminum engine blocks for replacement of conventional cast iron blocks. It is also employed extensively in automobile components such as pistons and cylinder liners. In addition to 390, alloy 393 (22 wt. pct. Si, 2.3 wt. pct. Ni, 1.3 wt. pct. Cu, 1 wt. pct. Mg, 0.1 wt. pct. Va) has been developed specifically for high temperature applications (e.g. internal combustion engine pistons).

The presence of hard primary silicon in the structure of cast hypereutectic alloys makes the casting extremely difficult and expensive to machine However, a small phosphorous addition (~0.01 wt. pct.) produces finer and more evenly distributed primary silicon crystals, and hence enhances the machinability, wear resistance and mechanical properties.<sup>11-14</sup>



Alloy No. and Temper†	Modification	Y.S. (0.2%) <sub>,</sub> MPa	UTS MPa	% Elongation	Ref.
A356.0T6	None	165	230	3.5	6
A356 0T6	Sb-Mod.	-	270	6.0	7
A356 0T6	Sr-Mod.	208	285	12.0	8
· A413 0	None	112	137	2.0	9
A413 0	Sr-Mod <sup>,</sup> .	124	181	12 0	9
13% Sı	None	-	12,4	2.0	10
13% Si	Na-Mod.		193	13.0	10

Table 1.2 Effect of Modification on the Mechanical Properties of Sand-Cast Al-Si Alloys

† T6; Solution heat treated and artificially aged

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### 1.2 Characterization of the Degree of Modification

The mechanical and physical properties of a casting are the direct result of the microstructure. For Al-Si castings, the level of modification obtained is an important parameter which determines the properties, and hence the quality of the end products. Therefore, a process control technique should be employed in the foundry to monitor the modification potential of the molten metal before casting, and to determine the degree of modification in castings.

In recent years, considerable interest has been shown in the development of such a technique to characterize eutectic silicon modification, mainly because aluminum casting producers are faced with increasing demands for higher quality products. Some of the techniques which have been proposed will be briefly reviewed in following sections.

#### 1.2.1 Chemical Analysis Technique

In order to ensure a good modification the<sup>®</sup> precise control of melt chemistry, especially modifier level, is essential. Rapid chemical analysis techniques such as emission spectrometry evaluate the concentration of modifier in the molten metal prior to casting. However, the degree of modification in a casting depends not only on the level of modifier but also on the solidification rate and holding time of the melt after the addition is made. Furthermore, some minor elements in the melt may inhibit the effectiveness of the modifier. This is especially true of phosphorous which should be controlled at levels of less than 5 ppm<sup>15-17</sup>, but whose concentration cannot be precisely determined by spectrometry. The chemical analysis of the molten metal alone is, therefore, not a guaranteed method for measuring the degree of modification.

#### 1.2.2 Metallography

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The microstructural change of eutectic silicon with modification can be followed by optical microscopy. Recently, a series of standard

photomicrographs which illustrate the different levels of silicon modification in hypoeutectic Al-Si casting alloys have been published for quick microstructural comparison.<sup>18</sup> However, the major drawback of metallography is the time consuming sample preparation.

If it is performed to check the quality of the melt prior to casting, the melt must be held a longer period of time and fading of modifying agents, particularly sodium, can occur. Furthermore, metallographic examination of the castings themselves is not usually desirable because destruction of the castings may be required.

#### 1.2.3 Thermal Analysis Techniques

Recently, the use of thermal analysis to control the molten metal quality in aluminum castings has increased considerably.<sup>19-23</sup> This technique involves obtaining the cooling curve of a sample of the melt as it solidifies in a sand mould. Figure 1.3 shows the cooling curves for an unmodified and a modified hypoeutectic AI-Si alloy.<sup>19</sup> Modification increases the degree of undercooling ( $\Delta T_N$ ) prior to eutectic phase nucleation and depresses the eutectic growth temperature T<sub>E</sub>. The degree of depression of the eutectic temperature,  $\Delta T_E$ , can be related to the level of modification. A larger  $\Delta T_E$  and  $\Delta T_N$  are indicative of a better modification.<sup>19,20</sup> If the measured  $\Delta T_E$  is not sufficiently high (e.g. <5.5°C for A356 alloy<sup>20</sup>), then the chemistry of the melt must be corrected by adding more modifier before pouring.

A thermal analysis technique based on a rapid data acquisition system using a microprocessor has been developed for the foundry environment.<sup>21–23</sup> However, the main disadvantage of this technique is that it is limited to testing the melt prior to casting only. Therefore, an additional non-destructive technique is required to assess the degree of modification in the casting after complete solidification. It has also been reported that several other minor elements present in the melt, such as Mg, Cu, Zn, Mn and Ni, reduce the eutectic growth temperature appreciably.<sup>20,22,24</sup> Hence a small change in the chemistry of the melt may mask the effect of a modifier





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on the cooling curves. This can result in a false interpretation of the modification potential of the melt.

1.2.4 Electrical Resistivity/Conductivity Measurement

It has been shown that electrical resistivity<sup>22,25</sup> and conductivity<sup>26</sup> measurements can be used to control the melt quality prior to pouring and also to measure the degree of modification in castings (i.e. as a non<sup>2</sup> destructive test)

As shown in Figure 1.4<sup>22</sup> the electrical resistivity of hypoeutectic Al-Si casting alloys decreases by about 5-10 percent when strontium is added to these alloys.<sup>22,25</sup> The magnitude of this decrease in resistivity can be related to the concentration of strontium. The addition of strontium to alloy A356 in amounts equivalent to those used for modification (i.e. up to 0.03 wt. pct. Sr) reduces the electrical resistivity by 8 percent.

A simple resistivity measurement technique involves the measurement of the voltage drop across a sample when a constant direct current is passed through the sample, with subsequent conversion of voltage drop to resistivity. On the other hand, measurement of conductivity, the reciprocal of resistivity, using an electrical conductivity meter is a quicker and simpler method which is easily adapted to foundry conditions. Quite sensitive, as well as temperature compensated devices which measure the conductivity by eddy current techniques are available.

In order to control the melt quality with this method, two samples of the melt, the first one taken prior to the addition of the modifier and the second one after complete dissolution of the modifier, are cast into an appropriate mould, and their resistivity/conductivity measured at room temperature. The difference between the two measurements will determine whether a fully modified structure has been obtained. The same method can also be applied non-destructively to castings to evaluate the modification level, particularly in heavy sections and riser connections where the solidification rate is slower and modification is more difficult.



Figure 1.4 Variation of electrical resistivity of A356.0 alloy with strontium content.(22)

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#### 1.3 Objective of the Present Study

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The continuing increase in demand for higher quality and/or consistant quality Al-Si cast components required by the automotive, aerospace and defense industries, necessitates a search for better non-destructive testing methods for the determination of the eutectic silicon structure. The electrical resistivity/conductivity measurement technique is promising and has certain advantages over other techniques mentioned previously since it is quick and readily adaptable to the foundry environment.

As discussed earlier, the electrical resistivity of Al-Si alloys decreases with increasing modifier content; however, the reasons for this resistivity variation are not fully understood. A major question which still has to be answered is whether the observed decrease in resistivity after the strontium addition is due to the change in the silicon morphology from acicular to fibrous or to the change in resistivity of the aluminum matrix phase resulting from the presence of strontium. Furthermore, a clearer understanding of how the various microstructural forms of the eutectic silicon affect resistivity is essential before electrical resistivity/conductivity can be widely used as a non-destructive testing method.

The main objective of the present invesitgation is to address the above questions and to provide a better understanding of the electrical resistivity behaviour of Al-Si alloys during microstuctural modification and heat treatment. The experimental work involved studying the resistivity of both solid-solution and two-phase Al-Si and Al-Si-Mg alloys. In addition, a series of directionally solidified binary eutectic alloys was produced in order to determine the effect of silicon shape on resistivity.

#### CHAPTER 2

#### THEORETICAL BACKGROUND AND LITERATURE REVIEW

Since this is a study related to the modification of Al-Si alloys, a brief summary of the theory of modification will be presented initially. The theories concerning the electrical resistivity of metals and alloys will be discussed later, together with a review of literature on the application of electrical resistivity in physical metallurgy.

### 2.1 Modification of Eutectic Silicon

There are several commonly observed microstructures in eutectic alloys such as lamellar, rod-like (fibrous), and irregular flakes and plates.<sup>27,28</sup> In the technical literature the term "modification" is used to describe a transition in the eutectic structure from one form to another. In the Al-Si system this transition is from acicular flakes (irregular) to fibrous rods. As mentioned earlier, such a change can be obtained by solidification at a very high rate or by the addition of sodium or strontium.

A fast solidification rate provides a large undercooling (10-12°C) prior to the nucleation of eutectic phases and a short growth time for the nucleated silicon particles. Consequently, the final eutectic structure is fibrous.10,29-31 Such a fine structure has been observed in test castings solidified at rates of 400 to 500  $\mu$ m/sec.10,29-32 In foundry practice, solidification rates are in the range of 1-10  $\mu$ m/sec<sup>30</sup>, and it is rarely feasible to produce the rapid solidification rate needed to obtain a fully modified structure. Therefore, adding a modifier to the melt before casting to produce a fibrous as-cast structure is common foundry practice. The most widely accepted modification theory suggests that modifying elements alter. both the nucleation and growth kinetics of the eutectic silicon phase.

#### 2.1.1 Nucleation of Eutectic Silicon

It is well known that the kinetics of solid phase nucleation in a liquid depend on the type and amount of impurities present. Crosley and Mondolfo<sup>33</sup> have shown that in Al-Si alloys, the nucleation of silicon is controlled by aluminum phosphide (AIP). Since AIP has a crystal structure very close to that of silicon, it acts as an effective nucleant.<sup>34,35</sup>

Commercial purity Al-Si alloys contain sufficient amounts of phosphorous to provide an abundance of AIP particles in the melt. This favourable condition results in the nucleation of eutectic silicon at small undercoolings (<2°C), and the formation of coarse acicular silicon flakes.<sup>33</sup> In a modified alloy, the modifier neutralizes the AIP particles leading to fewer effective nuclei, and hence the eutectic silicon nucleates relatively quickly and at larger undercoolings (>3°C).<sup>33</sup>

More recently, quenching experiments carried out by Flood et al.<sup>36</sup> and Hanna et al.<sup>37</sup> have revealed that the nucleation of silicon from the bulk liquid does not occur in modified samples due to the presence of a nucleation barrier, although it does to some extent in unmodified samples.

If the change in nucleation kinetics was the only mechanism involved in the modification process only refinement of the silicon crystals would be observed, but not the alteration in the shape of the crystals. Therefore, modifying agents must also influence the growth kinetics of the nucleated silicon particles during the eutectic phase transformation

2.1.2 Growth of Eutectic Silicon

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The growth of the eutectic phases may follow two general patterns<sup>38,39</sup>:

i) uncoupled growth; one phase grows into the liquid well in advance

of the other. Such growth is characterized by a non-planar, nonisothermal solid-liquid interface,

ii) coupled growth; two phases grow into the liquid at a common interface which is nearly planar and isothermal.

Uncoupled growth leads to a flake or plate-like structure, whereas coupled growth usually results in the formation of either a lamellar or fibrous structure.38,39

It has been shown that the growth of the unmodified eutectic occurs by solidification with a non-planar, non-isothermal solid-liquid interface as seen in Figure 2.1a.<sup>28,36,39,40</sup> In the modified structure, however, the solid-liquid interface is planar and isothermal (fig. 2.1b) thus the growth of the eutectic is coupled.<sup>28,36,39-41</sup> Such a change in growth patterns is clearly an important aspect of the mechanism of modification.

Another significant effect of modifiers is to change the growth kinetics of the silicon crystals. In an unmodified AI-Si alloy, silicon flakes grow, anisotropically on {111} habit planes by the twin plane re-entrant edge (TPRE) mechanism.<sup>28,30,41-43</sup> When a modifier is present in the melt, these preferred sites for silicon crystal growth are poisoned by selective adsorption of modifying elements onto the surface of the habit planes. This results in growth which is kinetically restricted in many directions (i.e. more isotropic growth), and in the appearance of a fibrous silicon structure.<sup>28,41-43</sup>

#### 2.2 Theories of Electrical Resistivity

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One of the most important characteristics of metals is their ability to readily conduct electricity due to the presence of a large number of free electrons. In this section, several theories dealing with the electrical resistivity of metals will be reviewed briefly.

The early theory, called the free-electron theory, was developed by Drude and Lorentz in the early 1900's.<sup>44</sup> This theory suggests that electrons detached from atoms transferred the electric current under an applied electric field, and their collisions with atoms in the lattice caused the observed resistance to this current flow.

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In 1928 Sommerfeld<sup>45</sup> developed the Fermi quantum mechanical view of free-electrons based on Pauli's exclusion principle. Free-electron theory assumes that all of the free electrons move under the influence of an electric field. Quantum mechanics, however, indicates that only specific electrons which are near the Fermi surface can participate in conduction and, thus, in the scattering (i.e. collision) process.

A later improvement to the quantum mechanics theory made by Bloch<sup>46</sup> was that the electrons should be considered as moving in an electrostatic field resulting from atomic vibrations. These overall vibrations, known as phonons, cause scattering of electrons to a certain extent depending on temperature,<sup>3</sup> and they are the origin of the large temperature dependent part of resistivity.

In quantum mechanical theory, resistivity is expressed as

$$\rho = \frac{mV_F}{Ne^2 l}$$
 and  $\sigma = \frac{1}{\rho} = \frac{Ne}{m}$ 

where p = electrical resistivity
 σ = electrical conductivity
 m = mass of an electron
 e = charge of an electron...
 V<sub>F</sub> = velocity of the electrons at the Fermi level
 N = the number of electrons at the Fermi level
 L = the mean free path of the electrons

Equation 2.1 suggests that the conductivity of a solid is a function of the population of electrons near the Fermi level and on the mean free path of the electrons. Metals which have partially filled or overlapped valence bands, such as copper, aluminum and silver, exhibit high conductivity due to the high density of electrons near the Fermi level. In addition, for a given metal, the changes in resistivity must arise from changes in the mean free path, since the other parameters are constant.

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(2.1)

As noted previously, the mean free path of electrons is limited by the process of electron scattering which can be divided into two components; i) thermal scattering, and ii) impurity scattering. Since their effects are additive<sup>47-50</sup>, the total mean free path,  $I_T$ , due to both thermal and impurity scattering may be expressed as;

$$r = \frac{1}{l_{th}} + \frac{1}{l_o}$$
 (2.2)

where  $l_{th} =$  thermal component of the mean free path  $l_0 =$  impurity component of the mean free path

Substituting  $I_T$  into Equation 2.1 gives a relations known as Mathiessens's Rule;

$$\rho_{Total} = \rho_{th} + \rho_o$$

where  $p_{th}$  = thermal component of resistivity  $p_0$  = impurity component of resistivity

In the following sections, the discussion will focus on the effect of each component on the resistivity of metals, with particular emphasis on aluminum.

# 2.2.1 Thermal Component of Resistivity

This component of electrical resistivity, also called ideal or lattice resistivity, arises from the interaction between moving electrons and atomic vibrations (i.e. phonons). Since the amplitude of atomic vibrations increases with increasing temperature, the probability of an electron collision with atoms in the lattice increases and hence so does the resistivity.

It has been found that the resistivity of most metals is linear with temperature near and above room temperature and proportional to T<sup>5</sup> at low temperatures.<sup>47-49,50-52</sup> Therefore,  $\rho_{th}$  of a metal near room

(2.3)

temperature may be expressed as;

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$$\rho_{th} = \alpha \frac{mV_F}{Ne^2} T$$

(2.4)

where a = proportionality coefficientT = absolute temperature

Figure 2.2<sup>53</sup> gives the electrical resistivity of pure aluminum as a function of temperature. Between 273 and 573K, the temperature dependence of resistivity of aluminum is approximately linear with a coefficient of 0.00113  $\mu\Omega$ cmK<sup>-1</sup> 44,54, and at low temperatures (<20K), it becomes proportional to (T<sup>2</sup> + T<sup>5</sup>).5<sup>3-56</sup>

2.2.2 Impurity Component of Resistivity

This component, known as the residual resistivity ( $p_0$ ), is the result of the scattering of conduction electrons by impurity atoms and other imperfections present in the crystal structure, such as dislocations, vacancies and grain boundaries.  $p_0$  can be measured quite accurately at low temperatures (4 2K in practice), since it is temperature independent and  $p_{th}$ is negligible at this temperature.

Each impurity atom causes a distortion in the lattice due to its different size and valence. This results in the conduction electrons being scattered and thus an increase in resistivity. For very dilute metals, the residual resistivity is proportional to the concentration of impurities<sup>47-50</sup> and can be expressed as;

$$\rho_o = \frac{mV_F}{Ne^2} \frac{X_i}{a}$$
 (2.5)

where X<sub>i</sub> = atomic fraction of impurities a = interatomic distance



as a function of temperature.(53)

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The electrical resistivity of a given metal at a constant temperature increases as the density of defects increases<sup>49,51,57</sup> due to the fact that electrons are scattered at these defects. It is well established that vacancies, interstitials, dislocations, grain boundaries, stacking faults, as well as phase boundaries all contribute to the electrical resistivity of metals and alloys Their effects on the residual resistivity of aluminum are summarized in Table 2.1.54,58-67

#### 2.3 Resistivity of Alloys

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The electrical resistivity behaviour of an alloy is generally distinguishable from that of a pure metal by the diminished importance of the thermal component of resistivity, since the residual resistivity component becomes the predominant factor. The effect of alloying elements on resistivity depends on their type, concentration, and how they are present within the matrix of the host metal (i.e. whether they are in solid solution or not).

2.3.1 Solid Solutions

The resistivity of very dilute solid solutions is given by Equation 2.5; however, most common metals contain alloy elements at levels beyond the validity of this relation. In general, the electrical resistivity of a metal is increased by the presence of alloying atoms in solid solution. This increase in residual resistivity is given by Nordheim's Rule.<sup>68</sup>

$$\Delta \rho_{a} = A X (1 - X) \tag{2.6}$$

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#### where X = atomic fraction of the solute element

A = a constant which depends on the difference in valence and atomic size between the solute and solvent elements

Type of Defect	Resistivity of Defect	Typical Concentration or Density of Defect	Contribution to Resistivity of Al (μΩ cm)	Reference
Vacancies	$ ρ_v = (2.6 \pm 0.5) $ μΩ cm/at.%	10 - 11 at.% as annealed 0.01-0.1 at.% as quenched	>~10-11 0.03-0.3	58-60
Dislocations	$p_{d} = (3 \pm 1) \times 10 - 13$ μΩ cm <sup>3</sup>	10 <sup>6</sup> lines/cm <sup>2</sup> as annealed 10 <sup>11</sup> -10 <sup>12</sup> lines/cm <sup>2</sup> as cold worked	3 × 10 - 7	61-65
Grain , Boundaries		100-300 cm - 1 as annealed	~2 × 10 - 4	<sup>,</sup> 65-67
Stacking Faults	$p_{sf} = (4 \pm 1) \times 10^{-7}$ $\mu \Omega \text{ cm}^2$	50-500 cm - <sup>1</sup> as annealed	~10-4	54,66
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Table 2.1 The Effect of Various Types of Lattice Defects on theElectrical Resistivity of Pure Aluminum

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If more than one element is dissolved in solid solution their effects are additive47,49,61,69, i.e.

$$\Delta \rho_o = \sum_i A_i X_i (1 - \hat{X}_i)$$
(2.7)

The increase in the residual resistivity that occurs due to valence differences between the solute and the solvent atoms was first established by Norbury<sup>70</sup>. This effect was later investigated by Linde<sup>71</sup> for several binary alloy systems. He showed that for a given quantity of solute, the change in resistivity can be correlated with the difference in valency between the two elements, according to the relationship;

$$\Delta \rho_o = a + b \left( \Delta Z \right)^2 \tag{2.8}$$

where a = constant b = constant $\Delta Z = difference in valence$ 

This relation was confirmed by Robinson and Dorn<sup>72</sup> for aluminum solid solution alloys at 20°C. They also found that the calculated values from Equation 2.8 fit the experimental data better if  $Z_{AI}$  is assumed to be 2.5 instead of 2. Table 2 2<sup>53,54,61,73-75</sup> outlines the available data on the effect of some common alloying elements on the resistivity of aluminum together with their maximum solubilities, and the ratio of their atomic radii to that of aluminum.

2.3.2 Alloying Elements out of Solution

The resistivity of super-saturated solid solution alloys generally decreases when precipitation of the alloying elements occurs. This takes place because solvent atoms in solid solution scatter electrons much more effectively than when they are out of solution.<sup>61,76-78</sup> In some cases, however, an initial increase in resistivity is observed during the precipitation process. This behaviour is particularly common in certain aluminum alloys. It

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	Element	r <sub>E</sub> /r <sub>Al</sub> *	Solubility** (at. %)	Δρ*** (μΩ cm/at.%E)
	Ag	1.01	24	1.0
	Cr	0.984	0.38	<sup>°</sup> 8.4
	Cu	<sup>-</sup> 0.898	2.4	0.83
r	Fe	0.893	0.025	5.4
,	Mg	1.12	17	0.46
	Mn	0.903	0.90	6.9
	Ni	0.870	0.023	1.6 🖌
	Si	0.978	1.6 ,	0.72
	Ti	1.02	0.14	5.5
	Zn	0.972	66	0.23

Table 2.2 The Effect of Metallic Solutes on the Resisitivity . of Pure Aluminum

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ratio of atomic radius of element to that of aluminum maximum solid solubility of element in aluminum the resistivity increment per atomic percent of element dissolved in aluminum

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has been shown that this initial increase in resistivity is assogiated with the formation of coherent small clusters known as Guinier-Preston zones (G P zones) which cause lattice distortions, and hence a higher resistivity 61,77-79

#### 2.3.3 Multiphase Alloys

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When an alloy consists of more than one phase, the electrical resistivity of the mixture ( $p_{mi}$ ) depends not only on the respective resistivities of the component phases and their volume fractions, but also on their structural arrangement. If one phase is randomly dispersed in the continuous matrix of the other, the conductivity of the mixture ( $\sigma_{mi}$ ) is given by;<sup>56</sup>

$$\sigma_{mi} = \sigma_{c} \left\{ \frac{1 + 2V_{d} (\frac{1 - \sigma_{c} / \sigma_{d}}{2\sigma_{c} / \sigma_{d} + 1})}{1 - V_{d} (\frac{1 - \sigma_{c} / \sigma_{d}}{2\sigma_{c} / \sigma_{d} + 1})} \right\}$$
(2.9)

where  $\sigma_c = conductivity$  of the continuous phase  $\sigma_d = conductivity$  of the dispersed phase  $V_d = volume$  fraction of the dispersed phase

Here it should be noted that Equation 2.9 is valid only if the second phase is assumed to be spherical. However, phases in metallic solids often have more complex shapes and thus the resistivity of the mixture follows more complex relationships. Most of the time, however, upper and lower bounds for  $p_{mi}$  may be approximated by considering the two phases to be aligned either parallel to the current flow direction;

$$\frac{1}{\rho_{mi}} = \frac{V_{\alpha}}{\rho_{\alpha}} + \frac{V_{\beta}}{\rho_{\beta}}$$
(2.10)

or perpendicular to the current flow;

 $\rho_{mi} = V_{\alpha} \rho_{\alpha} + V_{\beta} \rho_{\beta} \tag{2.11}$ 

where  $V_{\alpha}$  = volume fraction of  $\alpha$   $V_{\beta}$  = volume fraction of  $\beta$ ,  $\rho_{\alpha}$  = resistivity of  $\alpha$  $\rho_{\beta}$  = resistivity of  $\beta$ 

This behaviour is shown schematically in Figure 2.3.<sup>80,81</sup> More detailed analyses based on electrical analog<sup>82</sup> and computer simulation<sup>83</sup> techniques have also been proposed in order to incorporate the phase distribution.

2.4 Applications of Resistivity in Physical Metallurgy

Since electrical resistivity is relatively easy to measure and theoretically sensitive to structural changes, it is frequently used to obtain information about the physical and chemical status of metals and alloys. In this section some common applications of electrical resistivity measurements will be reviewed briefly.

2.4.1 Purity Determination

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A simple way to estimate the overall purity of a metal is to measure the ratio of the resistivities at room temperature and at liquid helium temperature (i.e.  $p_{297K}/p_{4.2K}$ ). This ratio, called the residual resistivity ratio (RRR), increases as the purity of a metal increases. According to RRR values, the purity of aluminum can be classified as follows:<sup>54</sup>

Commercial Purity Aluminum: RRR < 100 Pure Aluminum: RRR = 100-1000 Very Pure Aluminum: RRR = 1000-10000 Ultra Pure Aluminum: RRR > 10000

2.4.2 Deformation, Recovery and Recrystallization

Plastic deformation of a metal or an alloy below its recrystallization temperature leads to a microstructure which consists of fine grains and a high density of lattice defects, including dislocations, grain boundaries and lattice strain. As noted in section 2.2.2, these lattice imperfections contribute significantly to the residual resistivity of a metal, especially in a

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pure metal. Therefore, annealing of cold worked metals results in a decrease in resistivity due to the elimination of these defects. Resistivity also reflects the microstructural changes which occur during annealing, commonly known as recovery and recrystallization. It is well established that the recovery process of a metal involves the movement and gradual disappearance of defects and occurs in the first stage of the annealing process. It is usually accompanied by a decrease in resistivity. Generally, resistivity data obtained from such an experiment could be used to calculate an activation energy for defect motion<sup>59,84,85</sup> since the change in resistivity ( $\Delta \rho_0$ ) may be expressed as;

$$\Delta \rho_{a} = A \exp(-E_{m}/k_{a}T_{A}) \qquad (2.12)$$

where A is a constant, k<sub>o</sub> is Boltzmann's constant, E<sub>m</sub> is the activation energy of defectmotion and T<sub>A</sub> the annealing temperature.

Later, at a sufficiently high temperature, recrystallization occurs and all the initial excess resistivity disappears. Figure 2.486 illustrates the variation in electrical conductivity as a result of recovery and recrystallization of aluminum. In addition to aluminum, changes in resistivity during annealing have been studied in iron and low-carbon steels<sup>87,88</sup>, copper<sup>89,90</sup> and  $\beta$ -brass.<sup>90</sup> More recently, an in-situ differential resistivity measurement technique has been developed by Drew, Muir and Williams<sup>91</sup> to monitor continuously the annealing stages of cold worked plain carbon and high strength low alloy steels.

#### 2.4.3 Solid Solubility Limit Determination

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In order to determine the solid solubility limit, isothermal electrical resistivity measurements of solid solution alloys containing different amounts of solute elements are carried out. In the case of complete solid solubility (e.g. Cu-Au system), the resistivity exhibits a maximum near the 50 at. pct. level of each component, as shown in Figure 2.5a<sup>44</sup>. For two- and

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Figure 2.5 Variation of electrical resistivity as a function of composition.(44)

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multiphase alloys, the solubility limit is defined as the point at which the slope of the resistivity versus composition curve alters (Fig. 2.5b). It is known that the change in resistivity follows Linde's and Nordeim's Rules (section 2.3.1) within the limit of solid solubility. Beyond this limit, a change in slope occurs because the resistivity then follows the law of relixtures as given in Equations 2.9, 2.10 or 2.11 This resistivity behaviour is schematically presented in Figure 2.5b.

In order to determine the solid solubility limit of silicon in aluminum, Kovacs-Csetenyi et al.<sup>92</sup> measured the low temperature resistivity of a series of Al-Si alloys. Later, a similar method performed at room temperature was used by Drist, Kadaner and Kuzmina<sup>93</sup>. Their results agree well with each other and with data obtained by other methods, such as optical metallography<sup>94</sup> and dilatometry<sup>95</sup>. It was found that the maximum solubility of silicon in aluminum is  $1.5 \pm 0.1$  at the eutectic temperature, and it decreases to 0.005 at. pct at 200°C.

2.4.4 Precipitation and Clustering

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Electrical resistivity measurements offer a powerful method for studying the decomposition of supersaturated solid solutions, since large changes in resistivity can occur due to the precipitation or clustering of solute atoms. As previously noted (section 2.3.2), since the introduction of alloying elements into solid solution increases the resistivity, it would be expected that the precipitation process should lower the resistivity. This is true if the precipitates are non-coherent with the matrix. On the other hand, a resistivity increase, at least initially, is observed if the clusters have a fully coherent interface with the matrix. Such clusters begin to form as early as 5 seconds after quenching<sup>79</sup>. Therefore, electrical resistivity has a great advantage over other techniques such as transmission electron microscopy and microhardness in studying clustering kinetics since continuous resistivity measurements on a specimen can be made starting immediately after quenching. The clusters are plate-like or spherical solute rich coherent zones, known as G.P. zones, and metastable structures which precede the formation of equilibrium precipitates<sup>79</sup>. The clusters grow at very rapid rates up to a certain size where the lattice strain becomes a maximum. At this point, the resistivity also shows a maximum and subsequently a decrease as the clusters grow and lose coherency with the matrix (Figure 2.6).

The early stages of aging at low temperatures in several quenched aluminum alloys have been studied extensively by means of electrical resistivity measurements. It has been demonstrated that initial resistivity increases take place in the Al-Ag96,97, Al-Cu78,98,99, Al-Mg59,100, Al-Zn101 binary alloys and the Al-Mg-Si102-104 ternary alloy. The aging behaviour of an Al-Si alloy containing 1 wt. pct. Si was investigated by Rosenbaum and Turnbull<sup>105</sup>. They detected no change in the resistivity during aging at low temperatures (-30°C to 50°C) but observed a continuous drop during aging at higher temperatures (150-360°C). Consequently, they suggested that there is no G.P. zone formation in Al-Si alloys.

In addition to studying precipitation kinetics, the measurement of electrical resistivity is used to monitor the commercial heat treatment of precipitation hardening alloys<sup>61,86</sup>. Optimum mechanical properties are obtained in these alloys when the lattice strain is greatest (i.e. the resistivity is a maximum).<sup>79</sup> Thus the measurement of electrical resistivity is an accurate method for the determination of aging time and temperature.

2.4.5 Order-Disorder and Phase Transformations

Resistivity studies are particularly important in producing information about the kinetics of ordering processes which occur in quenched alloys such as Ag-Al106, Cu-Zn107 and Au-Cu44,108-110 during the aging of these alloys. Ordering of the solute in the solvent lattice increases the lattice regularity and reduces electron scattering<sup>77,108</sup>. This results in the resistivity of an ordered alloy being significantly lower than that of a disordered alloy.<sup>44,77,108</sup> Figure 2.7 shows the effect of ordering in the Au-Cu system where Cu<sub>3</sub>Au and CuAu are ordered structures.



Figure 2.6 Schematic representation of the change in electrical resistivity. of an age-hardening alloy as a function of aging time.(77)

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electrical resistivity of Cu-Au alloys.(77)

When a metal or an alloy undergoes a phase transformation such as the liquid-solid and allotropic transformations, the crystal structure or lattice parameter changes. This change is reflected clearly by a discontinuity in the electrical resistivity vs. temperature curve at the transformation temperature. A typical phase change, as detected by electrical resistivity measurements, is shown in Figure 2 8111

Electrical resistivity measurements have very distinct advantages in studying the martensitic transformation which occurs with very high speed With appropriate circuits, the resistivity of a specimen can be continuously recorded throughout the cooling process and changes occurring during the transformation can be detected accurately. In many cases (e.g. in the Fe-Ni system), the change in resistivity is large enough to correlate with the amount of transformation phase and to determine the temperature at which the martensite transformation starts (M<sub>s</sub>)

Pioneering studies done by Cohen and coworkers<sup>112-114</sup> on the mechanism of the martensitic transformation involved resistivity measurements. Later, resistance measurements were used by a number of workers to investigate various features of the martensitic transformation in several systems<sup>115</sup>, such as Fe-Ni, Fe-Cd-Ni and Au-Cd

2.4 6 Characterization of Microstructural Arrangement

Electrical resistivity was first used by Liebramn and Miller<sup>116</sup> in 1963 to characterize the structural arrangement of two-phase alloys. They measured the electrical resistivity of the InSb-Sb eutectic alloy in the directions parallel and normal to the Sb rods and developed a simple electrical analogy model to evaluate changes in resistivity as a function of the structural distribution of the Sb rods. The analog modelling involved representing the eutectic structures as electrical circuits composed of series and parallel resistors, as presented in Figure 2.9 The overall resistivity of the alloy was then predicted by calculating the resistivity of equivalent circuits. This modelling was later improved and modified to characterize more complex eutectic structures in a three-dimensional solid by Watson, Hahn and Kraft<sup>82</sup> who





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Figure 2.9 Representation of a two-phase alloy in electrical analog modelling.(119)

used it to determine the structure of the BI-Cu and InSb-Sb eutectics. Similar "models were also applied by Giddes" and Tauber<sup>117</sup>, Yim and Stoffor<sup>118</sup>, Simoneau and Begin<sup>119</sup> to predict the electrical resistivity of the Bi-MnBi and Al-Ni eutectics.

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In addition, the electrical resistivity of directionally solidified aluminum-based eutectics containing aligned Al<sub>3</sub>Ni<sup>120-122</sup>, Al<sub>6</sub>Fe and Al<sub>9</sub>Co<sub>2</sub><sup>122</sup> has been investigated. It has been noted that there is a significant difference in the resistivity of the eutectic mixture depending on whether the current is parallel ( $\rho_{\parallel}$ ) or perpendicular ( $\rho_{\perp}$ ) to the aligned phases. For example, Maire<sup>121</sup> reported that the resistivity ratio ( $\rho_{\perp}/|\rho_{\parallel}$ ) is in the range of 1.7 to 2.25 for the Al-Al<sub>3</sub>Ni eutectic at 4.2K.

#### 2.4.7 Non-Destructive Testing

Electrical conductivity/resistivity measurement techniques, especially eddy current testing, provide a rapid non-destructive test for a variety of structural components in production and in service. Several common applications of eddy current testing can be listed as follow: 123

1) Indirect hardness measurements,

2) Determination of case depth on case-hardened components,

3) Alloy sorting,

4) Defect detection; such as cracks, seams, porosity and shrinkage,

5) Thickness measurements of foils, sheet and strips.

# 2.5 Previous Experimental Work on Resistivity Measurements of Al-Si Casting Alloys a

Electrical resistivity measurements have been performed on Al-Si alloys in order to predict the degree of modification, to determine dendrite arm spacings and to follow heat treatment of these alloys.

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### 2.5.1 Degree of Modification in Al-Si Alloys

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The first attempt to use electrical resistivity for the evaluation of the silicon form in Al-Si alloys was made by Thirunavukkarasu and Panchanathan<sup>124</sup> in 1978. In 1983 Oger, Closset and Gruzleski<sup>25</sup> used a DC resistivity technique while Jacob and Remy<sup>26</sup> applied electrical conductivity methods to study the structural changes that occur during modification Later, Closset and his coworkers<sup>22</sup> developed an AC differential electrical resistivity technique to measure the difference in voltage between a standard sample (unmodified) and a modified sample. More recently, Argo et al.<sup>125</sup> have measured electrical conductivity by using an eddy current conductivity meter to determine the silicon morphology in 319, 356, 357 and 380 alloys.

The results of these studies demonstrated that the electrical resistivity of modified samples is lower than that of unmodified ones, with a resistivity difference of 5-10 percent between the two (Figure 14). The resistivity difference is enhanced by increasing the silicon content<sup>26,125</sup> or by bringing about a greater degree of structural change.<sup>22</sup>

It was also reported that electrical resistivity and conductivity methods cannot predict the eutectic silicon form in alloys containing high Mg contents, such as A355<sup>120</sup> and A357<sup>22</sup> (0.7 wt. pct. Mg), since there is no change in the resistivity upon modification. Closset et al <sup>22</sup> have suggested that this is the result of the modifying action of magnesium. Since magnesium partially modifies the eutectic silicon (i.e. forms a lamellar structure), the subsequent small change in the silicon structure which takes place during the modification treatment cannot be detected by electrical resistivity.

Furthermore, it has been found that resistivity is sensitive to hydrogen or shrinkage porosity<sup>22,25</sup>. Samples cast from degassed melts have a lower resistivity, by about 1 pct., than those cast from non-degassed melts. However, if the porosity content of a casting is less than 0.5 vol. pct., as in the case with most acceptable castings the overall resistivity behaviour, does not change. Therefore, electrical resistivity can still be considered a reliable method for assessment of the degree of modification.

#### 2.5.2 Dendrite Arm Spacing

<sup>(A)</sup>It is also of considerable interest to predict the dendrite arm spacing (DAS) in a casting due to its noticeable effect on the mechanical properties of Al-Si castings<sup>3-10</sup> A technique based upon comparative electrical conductivity measurements for determining DAS in a test casting has been developed by Argo, Drew and Gruzleski<sup>125</sup>. This technique has been applied to evaluate the DAS non-destructively in both modified and unmodified 319, 355, 356, 357 and 380 alloys. The results of this study showed that conductivity decreases as the DAS increases. The technique gives more reliable predictions of DAS for modified alloys than for unmodified alloys.

#### 2.5.3 Heat Treatment

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After casting, most Al-Si alloys are heat treated to optimize their mechanical properties. Generally, heat treatment schedules consist of a solutron treatment (T4) near the eutectic temperature and an aging treatment (T6). The change in silicon morphology has been followed continuously by in-situ differential electrical resistivity measurements during the heat treatment of A356.0 alloy<sup>126</sup>. The measurements indicated that the spheroidization of silicon that occurs during solution treatment leads to a lowering of the resistivity and proceeds faster in modified, compared to unmodified, alloys. In addition, the resistivity of solution treated and quenched A356 alloys shows a continuous decrease with aging time due to the precipitation process. Finally, it has been pointed out that differential electrical resistivity can be employed as a method of controlling the commercial heat treatment of hypoeutectic Al-Si alloys.<sup>126</sup>

#### **CHAPTER 3**

#### EXPERIMENTALPROCEDURE

#### 3.1 General

The experimental procedure employed in this work was designed to investigate the general resistivity behaviour of Al-Si alloys and the effect of various silicon morphologies on the resistivity of these alloys.

The major portion of the experimental work consisted of the preparation of alloys containing different levels of silicon, magnesium and strontium, and the measurement of the electrical resistivity or conductivity of the samples obtained from these alloys. The following variables were examined during the experimental work to determine their effect on resistivity:

- 1) The chemical composition of the alloys, mainly the concentrations of silicon, magnesium and strontium,
- 2) The solidification rate of the cast alloys,
- 3) The changes in microstructure at various stages of heat treatment of the alloys,
- 4) The temperature at which the electrical resistivity/conductivity measurements were carried out.

Another important aspect of the experimental work involved the preparation of directionally solidified samples of the AI-Si binary eutectic. The results of these experiments were used to illustrate the effect of different silicon morphologies on electrical resistivity

The remaining portion of the experimental work comprised of metallographic examinations, hardness testing and thermal, chemical and X-ray analyses of the alloys.

#### 3.2 Alloy Preparation

A series of alloys of binary Al-Si, Al-Sr, Al-Mg, ternary Al-Si-Sr, Al-Si-Mg and quaternary Al-Si-Mg-Sr were prepared.

3.2.1 Materials

The materials used in the preparation of the alloys were aluminum, silicon, magnesium and strontium. The chemical analyses of these materials except silicon, as provided by the suppliers, are given in Table 3.1.

<u>Aluminum</u>, the base material, was produced by ALCOA in the form of 2.5kg ingots, and was 99.99 pct. pure

<u>Silicon</u> was supplied by Johnson Matthey Inc. in the form of lumps. It was semiconductor grade silicon with a putity of 99.9999 pct.

<u>Magnesium and strontium</u> were supplied by Timminco Metals. The magnesium was in the form of 16kg ingots, and the strontium was in the form of extruded bars which were stored under vacuum to prevent excess oxidation. They were of 99.98 pct. and 98.0 pct. purity, respectively.

3.2.2 Melting and Alloying

Melting of the alloys was carried out using a Tocco MeltMaster induction furnace which was a 38KW, 30KVAR, 440V and 68A unit. The crucibles used for melting were machined from a high purity graphite block supplied by Speer Canada Ltd. and baked at 800°C to remove volatiles. During melting, oxidation was minimized by placing a 0.5cm diameter steel pipe slightly above the surface of the melt and continuously passing argon through the pipe at a flowrate of 11/min.

Silicon, magnesium/and strontium were used as alloying elements in order to determine the/effect of compositional change on resistivity. Silicon was added in pure form while the magnesium and strontium were added as

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Material	Element (wt. pct.)													
	Si	Fe	Cu	Mn	N	Na	Mg	Ca	Zn	Ga	Al	Pb	Sn	Sr
Aluminum	.003	.003	- & &				i ang si program	and a second france of the second of the	.002	.002	99.99			
Magnesium	.002	.001		.002	_	_	99.98	.001	.005		.003	.001	.002	-
Strontium	.040	005	_	_	.500	.100	.300	.003	_		.050			98.00

# Table 3.1 Chemical Composition of the Materials Used

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# Table 3.2 Chemical Analysis and RRR of the Aluminum Control Sample 、

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Material Condition	Element (wt.pct)								DDD
	Si	Fe	Mn	Cr	Ni	Zn	Ga	AI	ΝΝΝ
as-received	.003	.003	<u> </u>	<u> </u>	_	002	.002	99.99	175
after melting and casting	.003	.005	.002	.001	.001	003	.002	99.97	155

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aluminum based master alloys of compositions Al-1wt. pct Sr, and Al-10 wt. pct. Mg. The silicon concentration of the alloys ranged from 0 to 12.6 wt. pct. and the magnesium content varied from 0 to 1 wt. pct. The strontium level was kept constant (~0.032 wt. pct.) in all the Al-Si-Sr and Al-Si-Mg-Sr alloys. This level was chosen since it represents the approximate amount used when modifying commercial Al-Si casting alloys. In the Al-Sr binary alloys, however, the strontium content was varied within the range of 0 to 0.26 wt. pct.

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During alloy preparation, 150g charges of calculated amounts of aluminum, silicon and master alloys of strontium and magnesium were placed into the graphite crucible and melted under an argon atmosphere in the induction furnace. The melt temperature was brought up to 740°C and then maintained within  $\pm 2°C$  of this value for 15 minutes. It was found that this holding temperature and time were sufficiently high and long for complete dissolution of the alloying elements and homogenization of the melt. The melt temperature was measured using a chromel-alumel K-type thermocouple. The temperature was reduced to 720°C prior to casting.

#### 3.2.3 Casting

The molten alloys were cast into a horizontal split graphite mould located beneath the crucible. The casting of the alloys was done employing a bottom pouring technique to minimize direct contact of the melt with air and to avoid slag carry-over into the mould, hence maintaining the purity and cleanliness of the alloys. Figure 3.1 illustrates the melting and casting arrangements. After filling the mould, the remaining melt was cast into a copper mould to obtain samples for spectrochemical analysis.

A control sample of pure aluminum was also produced in exactly the same manner in order to determine the degree of contamination during melting and casting. The chemical analysis and residual resistivity ratio (RRR) of these samples, given in Table 3.2, revealed that the alloys were produced without an appreciable decrease in purity.

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Two different solidification rates were employed during casting in order to produce different levels of microstructural coarseness:

- i) Rapid Solidification was achieved by casting the molten alloys into a graphite mould at room temperature (i.e. chill cast). This gave a cooling rate of approximately 1°C/sec.
- ii) Slow Solidification was achieved by preheating the graphite mould up to 200°C and packing fibrefax insulation around it. In this case the cooling rate was of the order of 0.4°C/sec.

Figure 3.2 shows two cooling curves corresponding to the two different solidification rates employed. The cooling curves were obtained using a thermal analysis technique which will be described later.

The cast ingots were 10cm in length, 1.2cm in thickness and 4cm wide. These ingots were sectioned as shown in Figure 3.3. Regions A and E were not used in further investigations. Microstructural examinations were conducted on the samples taken from regions B and D. Region C, which exhibited the best chemical and structural homogenity, was used to prepare samples for electrical resistivity/conductivity measurements. The castings which had less than 2 wt. pct. Si were heated to 450°C for 15 minutes in an electrical resistance furnace and then hot rolled to a thickness of 8mm to produce virtually pore-free samples. The alloys, with more than 2 wt. pct. Si could not be rolled due to the presence of the brittle eutectic silicon and they were used in the as-cast condition for resistivity/conductivity measurements.

#### 3.3 Heat Treatments

Different heat treatment procedures were, designed for the different alloy groups, as outlined in Figure 3.4.

3.3.1 Single-Phase Alloys (<2 wt. pct. Si)

A homogenization treatment of these alloys was done to produce a large and stable grain size, to minimize imperfections caused by rolling and



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to dissolve precipitates and intermetallic compounds formed during casting and hot rolling. The samples which contained magnesium were homogenized at a low temperature (550°C) to prevent melting of the low melting point eutectic phase, Al-Mg<sub>2</sub>Si-Si.<sup>75,127,128</sup> The other samples were homogenized at 560°C.

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In cooling from the homogenization temperature, three different procedures were used:

- 1) Water quenching: Samples were removed from the furnace and quickly quenched in room temperature water.
- 2) Air cooling: After removal from the furnace the samples were placed on a large steel plate and allowed to cool to room temperature
- 3) Furnace Cooling: All samples were cooled in the furnace at a constant rate of 18°C/h.

The water quenched and air cooled samples were then annealed for 24h at room temperature before any measurements were made in order to eliminate the vacancies formed during cooling.

Pre-aging (natural aging) of quenched samples was also carried out for different periods of time at room temperature to study clustering kinetics in Al-Si and Al-Si-Mg alloys. Finally, an aging treatment of these samples was performed at 175°C-340°C for 18h.

The homogenization and aging treatments were done in a programmable electric resistance furnace whose temperature could be controlled with a  $\pm 0.5$  pct. accuracy between 140°C and 700°C. The samples were placed in the central portion of the furnace where the temperature variation was in the range of  $\pm 1$ °C. When the pre-aging time was shorter than 12h, the homogenization treatment was done in a separate automatically controlled furnace. This was done since the furnace described above could not cool to the desired aging temperature from the homogenization temperature in less than 12h.

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Each heat treatment procedure was repeated at least three times and the electrical resistivity/conductivity measurements were made after each step of the heat treatment process.

## 3.3.2 Two-Phase Alloys (>2 wt. pct. Si)

These alloys were solution treated for 24h at 540°C and then air cooled. This heat treatment procedure, designated as T4 by the Aluminum Association<sup>129</sup>, is commonly employed for commercial Al-Si foundry alloys. The changes in microstructure which took place during the heat treatment of these alloys were followed by electrical conductivity.

#### 3.4 Directional Solidification Experiments

It has been shown that the controlled directional solidification of AI-Si eutectic alloys leads to a variety of eutectic silicon structures which depend on the solidification conditions.<sup>28,130-132</sup> Therefore, it was decided to produce a series of directionally frozen samples of AI-Si eutectic alloy in order to investigate the effect of various silicon structures on resistivity. In addition, two sand cast eutectic alloys, in which the eutectic silicon is assumed to be randomly distributed, were produced.

The directional solidification was accomplished by holding the sample stationary and lifting a furnace vertically at a constant rate around the molten sample. Consequently, solidification would be initiated at the bottom of the sample and progress vertically upwards at a constant rate until the entire sample solidified. In the following sections the appraratus used in these experiments will be discussed together with the experimental procedure employed.

#### 3.4.1 Apparatus

The directional solidification apparatus consisted of a vertical tube furnace and a variable speed motor with a gear reducer. The vertical furnace, as schematically illustrated in Figure 3.5, was constructed using two


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resistance heating coils, with fibrefax insulation packed between the coils and furnace case.

The electrical resistance coils supplied by Thermcraft Inc. had an 8cm outside diameter and were 20cm in length. The resistance windings of the coils, constructed of 1.5m of 80 wt. pct. Ni-20 wt. pct. Cr, were helically wound and placed in the refractory backing grooves. Each coil was installed in the furnace vertically, as shown in Figure 3.5, and attached to a powerstat regulator (10 amp. max.). The two coils together had a total power output of~1200watts.

The furnace case consisted of a stainless steel cylinder (35cm in length with a 20cm outside diameter) and asbestos top and bottom covers which had circular openings 8cm and 5cm in diameter, respectively Additionally, a water cooled copper coil was attached to the furnace base to enhance the temperature gradient in the liquid ahead of the solidifying interface. The fibrefax insulation, packed between the electrical coils and the furnace case, provided a minimum of lateral heat loss and enabled the furnace to maintain the constant temperature profile required

The top opening of the furnace was closed by a graphite block in which a 0.5cm diameter steel pipe was placed to allow the passage of argon into the furnace. This provided an inert gas atmosphere in the furnace and minimized oxidation. The same pipe was also used for inserting a K-type thermocouple to monitor the temperature profile within the furnace Figure 3.6 illustrates a typical temperature profile obtained inside the furnace.

The vertical movement of the furnace was achieved by means of a drive mechanism consisting of a variable speed motor and a gear reducer. The variable speed motor, manufactured by U.S. Electrical Motor, has a minimum of 12 and a maximum of 72 revolutions per minute. The speed was further reduced by the use of a double worm gear reducer with a stepdown ratio of 900:1.



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The furnace was mechanically attached to a steel plate which moved either up or down. The steel plate was connected to the drive mechanism by the use of a threaded rod which acted as a screw. Each revolution of the rod provided a vertical movement of 0.1cm. Two steel rods were also placed on either side of the steel plate to support the furnace weight and to ensure smooth vertical movement. The above mentioned configuration provided variable furnace lift rates (i.e. growth rates) of 0.1 to 10cm/h.

#### 3.4.2 Experimental Procedure

Rods of binary Al-Si eutectic alloy were prepared by melting appropriate amounts of pure aluminum and pure silicon and casting into a graphite mould. These rods (3.3cm in diameter and 28cm in length) were placed in a recrystallized alumina tube (99 pct. Al<sub>2</sub>O<sub>3</sub>) 3.8cm in diameter and 30cm in length. The alumina tube was then set approximately 2cm into a graphite mould, as shown in Figure 3.5, inserted into the furnace and aligned parallel to it. The heating coils were then switched on. The furnace required about 1.5h to attain the temperature profile shown in Figure 3.6

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The samples were remelted at 820°C for 30 min in the furnace to stabilize the thermal conditions. The molten samples were then directionally solidified by lifting the furnace around them at a constant velocity. The growth rate ( $R_g$ ) was taken as the rate of furnace movement and six different growth rates, ranging from 0.2 to 5cm/h, were used. The temperature gradient, G, was recorded by inserting a thermocouple into a control sample. It was kept fairly constant for all samples at about 26°C/cm.

After complete solidification, the alloy rods were carefully broken out of the alumina tube and were cut as shown in Figure 3.7.



Region	Remarks	Application
A	original bar - not melted	-
В	poor directionality	
C,E,G	well aligned structure	metallography
D,F	well aligned structure	resistivity sample preparation
н	gas porosity and low purity	-

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Figure 3.7 Sampling of the directionally solidified rods.

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#### 3.5 Electrical Resistivity and Conductivity Measurements

#### 3.5.1 Electrical Resistivity

The samples used for electrical resistivity measurements were approximately 0.05cm thick, 0.3cm wide and 5.5cm long and were punched out of rolled sheets of the single-phase alloys. This method yielded ~1 pct. dimensional accuracy. In addition to these samples, several longitudinal and transverse samples 3.5cm long and 0.5cm in diameter were machined from each directionally solidified rod. For these samples the dimensional accuracy was in the order of 1.5 pct.

The resistivity of the samples was measured using a standard fourprobe DC technique at room temperature ( $297 \pm 0.5K$ ), in liquid nitrogen (77K) and in liquid belium (4.2K). This technique involved the measurement of the voltage drop between two points separated by a constant distance, L, when a constant direct current is passed. The electrical resistivity of the sample was then determined by the formula;

where p = electrical resist vity ( $\Omega$ m) V = voltage(V) I = current(A)A = cross-sectional area (m<sup>2</sup>)

L = constant distance (m)

(3 1)

This technique has an advantage over the more conventional methods, such as Wheatstone and Kelvin bridges, since it is unaffected by contact resistance. With the four-probe arrangement shown in Figure 3.8, the voltage contacts are not part of the circuit, and since they do not carry current their resistances have no effect on the measurements.<sup>44,115</sup>

The sample holders used for the voltage-drop measurements in the 4.2K and 297K temperature ranges are shown schematically in Figure 3.9. They consisted of two separate plexiglass plates which were joined by



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screws. The samples were sandwiched between the plates and electrical contact to the samples was made by means of pressure contacts. The voltage contacts were positioned a distance of 10mm from the current contacts to minimize any effect of a non-uniform electric field distribution near the current contacts. Fine copper wires (~0.10mm diameter) were used for the leads, and thicker (0.5mm diameter) wires were used for the contacts. The lead wires were fixed to the contacts using a Pb40-Sn60 soft solder. In order to ensure good electrical contact, the specimens were etched in a solution of 50 pct. HCl and water to clean the surface and remove any thick oxide layer before the samples were attached to the holder. The voltage leads were connected to a DC Keithley nanovoltmeter (Model No 181) which was able to detect 10nV with six digit resolution and an accuracy of 0.015 pct. in the millivolt range. A constant current was supplied by a Keithley current source (Model No 225), accurate to  $\pm 0.5$  pct.

The voltage-drop measurements of samples of the single-phase alloys (0.05x0.3x5cm) were performed by first immersing the samples in an acetone bath at room temperature and then in liquid helium stored in a Dewar flask. A current of 0.1A was chosen to ensure a sufficiently high voltage difference for accurate measurements. Likewise, the voltage drop of samples of directionally solidified alloys was measured at room temperature and at liquid nitrogen temperature using a current of 0.5A. During the measurements the current direction was reversed to minimize errors due to thermally induced voltages. The samples were also turned end for end and the measurements repeated.

#### 3.5.2 Electrical Conductivity

Samples used in conductivity measurements were cut from either rolled or as-cast ingots in the form of a square measuring 4x4x0.8cm. The conductivity measurements on these samples were performed using a K.J. Law Model M4900C electrical conductivity meter, as shown in Figure 3.10. This instrument measures conductivity by an eddy-current technique and displays it as a percentage of the International Annealed Copper Standard (pct. IACS), with 100 pct. IACS being equivalent to 1.7241 µΩcm at 293K.



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Figure 3.10 Electrical conductivity meter.



Figure 3.11 Typical arrangement of probe coil and test object.

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Eddy-current testing involved inducing an electrical current in a conductive material by means of an alternating magnetic field and observing the interaction between the applied current and the induced current (i.e. eddy currents in the sample). Coupling the AC magnetic field to the test sample was accomplished by the eddy current transducer, referred to as a probe or coil. Figure 3.11 illustrates the eddy currents induced in a sample by a test probe. The vector H<sub>a</sub> represents the AC magnetic field

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applied by the probe whereas H<sub>i</sub> indicates the induced AC field resulting from the eddy currents in the test object.

Four parameters influence the eddy currents:

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- .i) Electrical conductivity of the test material,
- ii) Magnetic permeability of the test material,
- iii) Geometry of the test specimen, such as size, diameter, thickness, etc.,
- iv) Surface discontinuities, such as cracks, and surface skin.

Clearly, successful conductivity measurements depend on the last three of the above parameters being kept constant. Al-Si alloys exhibit no ability to conduct magnetic flux and are considered non-magnetic<sup>53</sup> (i.e. constant magnetic permeability). The geometry of the test samples was held constant by using square plates having the same dimensions as the standard samples Finally, the surface effect was minimized by machining the test samples before the conductivity measurements were taken.

The M4900C gives a direct digital read out of conductivity with 0 1 pct resolution and 0.5 pct. accuracy within the aluminum conductivity range (28-65 pct. IACS). In addition, it can be adjusted to compensate for temperature, thus the instrument reads the conductivity of a sample as if measurements were carried out at 293K. Another feature of the device is its ability to provide a constant depth of penetration (1.72mm) of the eddy current field into the sample regardless of its conductivity. The depth of penetration is given by the relationship: G

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(3.2)

$$d = \frac{660}{\sqrt{a_f u}}$$

where d = depth of penetration (mm)

 $\sigma$  = conductivity of the test sample (pct. IACS) f = frequency of alternating current (Hz)

 $\mu$  = magnetic permeability (1 for aluminum<sup>53</sup>)

As can be seen from Equation 3.2, both conductivity and frequency influence the penetration depth. The M4900C varies the frequency to maintain a constant depth of penetration.

Measurements were performed according to the ASTM E1004-84133 standard procedure; i.e.

- 1) The test samples were machined to obtain a relatively flat smooth surface.
- 2) To eliminate edge effects, the dimensions of the samples were kept large. Specifically, the thickness of the samples was always greater than 2.6 times the depth of penetration.
- 3) It was ensured that the probe, standards and test samples were at the same temperature before any measurements were made.
- 4) The instrument was calibrated against three known aluminum standards and compensated for temperature.
- 5) The probe was placed on the center of the samples and the measurements were taken.
- 6) The calibration of the instrument was verified at the end of testing each set of samples.

#### 3.5.3 Accuracy of Experimental Resistivity and Conductivity Data

An error estimation, including all circuit components as well as dimensional uncertainties, revealed that the accuracy of the absolute resistivity values was about  $\pm 3$  pct. Table 3.3 -summarizes some of the expected error sources and their error contributions to the total accuracy of the data. However, the error in the resistivity ratio ( $p_{297K}/p_{4.2K}$ ), often used

Type of Measurement	Error Source -	Contribution to Accuracy of Measurements (±%)
	Nanovoltmeter	0 20
· · · ·	Current Source	0.50
Electrical Resistivity	<ul> <li>Leads and Contact Resistance</li> </ul>	0.20
	Sample Dimension	2.00
Electrical	Instrument	0.75
Conductivity ४	Standard Sample	0.25
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Table 3.3 Error Estimation

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in this work instead of absolute resistivity values, was approximately 1 pct. This is due to the fact that the resistivity ratios were obtained directly from measured voltage drops using the following relationship:  $\prec$ 

$$\frac{\rho_{297K}}{\rho_{42K}} = \frac{\Delta V_{297K}}{\Delta V_{42K}}$$
(3.3)

As seen in Equation 3.3, the resistivity ratio is not a function of sample dimension, and as such most of the errors noted in Table 3.3 are eliminated.

The error in conductivity measurements arose mainly from instrumental limitations. It is expected that the accuracy of the measurements is in the order of  $\pm 1$  pct. (Table 3.3).

3.6 Hardness Measurements

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Hardness measurements were done to support the resistivity/conductivity data. Microhardness tests were performed on either a Vickers microhardness tester or a Rockwell Tukon tester. Vickers microhardness testing was performed using a square-base pyramidal diamond indenter with an obtuse angle of 136° and a load of 2.5kg. The Tukon tester, however, allowed for measurement of the hardness across the primary aluminum dendrites by the use of a diamond indenter with an angle of 136° and a load of 250g. In both instances, four hardness readings were taken from each sample and averaged.

3.7 Thermal Analysis

Thermal analysis was employed to determine the cooling rates of the castings and to obtain information on the nucleation and growth kinetics of the eutectic phase. The system used for thermal analysis, shown schematically in Figure 3.12, was developed by Argyropoulos<sup>134</sup>. It consisted





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of a GIMIX host microprocessor, an electronic A/D converter (µmac-4000), a floppy disk driver and a Hewlett Packard graphics plotter (HP 7225A).

The analog signals were obtained using a K-type thermocouple placed in the mould. These signals were then digitized by the  $\mu$ mac-4000 and stored on a floppy disk. Later the collected temperature data were plotted against cooling time.

#### 3.8 Spectrochemical Analysis

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Chemical analysis was done on the spectrochemical samples cast into the copper mould and on the chill cast ingots. After they were machined to obtain a flat smooth surface, analysis was done by a vacuum emission spectrometer (Baird Atomic Spectro-1000 model no DV2) coupled with a microcomputer.

Pure aluminum and Al-Si-Sr standards supplied by Aluswiss Itd. were used to calibrate the spectrometer. The samples were analyzed for Si, Fe, Mn, Mg, Cu, Ni, Cr, Sr, Ti and Zn The results of these analyses are given in Table 3.4.

### 3.9 X-Ray Diffraction

X-ray diffraction was used to characterize the anisotropy of the silicon structure in the directionally solidified and sand cast eutectic samples. A Philips automatic diffractometer was employed using Cu K<sub>a</sub> radiation, and the diffraction was done on a flat and ground surface (320 grit finish) of the samples. The samples were always placed in the diffractometer with the same orientation to the X-ray beam. The total exposure time was about 3h.

#### 3.10 Metallography

The metallographic samples were cut from the cast alloys and the directionally solidified samples as shown in Figures 3.3 and 3.7.

Coarse grinding was carried out using 120 grit silicon carbide paper and was followed by fine grinding using a series of grinding papers of 240, 320, 400 and 600 grit with water as a lubricant. Subsequently, mechanical polishing was performed on samples of the single-phase alloys, and electrolytical polishing was performed on samples of the two-phase alloys.

Mechanical polishing was accomplished by first using  $5\mu$ m alumina powder suspended in water on a polishing cloth. Final polishing was done using  $0.3\mu$ m alumina applied on a medium speed wheel. Electrolytic polishing was carried out in a 30 pct. HNO<sub>3</sub> and 70 pct. methanol solution at about - 20°C under an applied potential of 40 DC volts.

The polished samples were examined, as polished on a Neophot optical microscope. Some of the samples were etched in a solution of 0.5 pct. HF, 1 pct. HCl, 1 pct. HNO<sub>3</sub> and 97.5 pct. H<sub>2</sub>O, and then examined using a Jeol (JSM-T300) scanning electron microscope. In order to identify the phases revealed by the SEM, in-situ semi-quantitative microchemical analyses (spot analyses) and X-ray mapping were performed on the samples using an energy dispersive spectrometry (EDS) analyzer (model no Tracor 5500) attached to the SEM.

Alloy	Element (wt.pct.)										
Name	Si	Fe	Cu	Mn	Mg	Nı	Zn	Ti	Sr		
1L	0.003	0.002	e skipe her skip vere verskiper i deservi			0.002	0.002	-	-		
2L	0.003	0.001		-	0.002	0.002	0.003	ø –	0.009		
3L	0.003	0.002	4		0.002	0.002	0.002	-	0.018		
4L	0.004	0.002	-	- ,	,	0.003	0.003	-	0.035		
5L	0.003	0.001	-	-	0.003	0.002	0.002	-	0.061		
6L	0.003	0.002	-	-	0.002	0.002	0.001		0.123		
7L	0.003	0.002	-	_	0,002	0.001	0.001		0.224		
1A <sup>-</sup>	0.004	0.004	-	0.005	-	0.002	0.002	0.001	\ - ·		
2A	0.144	0.003	-	0.007		0.002	0.004	0.001	\_ <b>_</b>		
3A	0.323	0.003	-	0.005		;0.002	0.003	0.001	· - ·		
4A	0.604	0.004	-	0.005	-	0.003	0.004	-			
5A	0.880	0.005	0.002	0.006		0.003	0.003	0.002	-		
6A	1.036	0.007		0.003		0.002	0.004	-	-		
7A	1.164	0.007	0.001	Q.005	-	0.002	0.004	0.001	-		
8A	1.239	0.005	-	0.005	-	0.004	0.004	j –			
9A	1.262	0.005	-	0.006		0.003	0.003	-	-		
10A	1.387	0:004	-	0.005	-	0.007	0.006	-	-		
11A	1.610	0.003	_	0.006	~	0.005	0.005	0.001	-		
1B	.0004	0.003	-	0.003	-	0.003	0.004	0.002	0.037		
2B	0.140	0.004	-	0.005		0.004	0.003	0.001	0.031		
3B	0.308	0.003	-	0.006		0.003	0.004	-	0.036		
4B	0.559	0.005	-	0.005	-	0.003	0.004		0.036		
5B	0.776	0.005	-	0.005	~	0.006	0.004	0.001	0.034		
6B	1.033	0.004	-	0.005	-	0.005	0.004	0.001	0.034		
7B	1.148	0.004	-	0.003	-	0.005	0.003	-	0.035		
8B	1.241	0.003	-	0.007	-	0.006	0.007	-	0.031		
9B	1,258	0.004	-	0.006	-	0.006	0.005	0.001	0.030		
10B	1.399	0.0 <u>0</u> 3	-	0.006		0.006	0.006	-	0.033		
11B	1.611	0.003	-	0.006	-	0.006	0.004	-	0.032		

Table 3.4 Chemical Composition of the Alloys Produced

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Table 3.4 continued

	Alloy	Element (wt pct)										
	Name	Si	Fe	<sup>5</sup> Cu	Mn	Mg	Ni .	Zn	Τı	Sr		
	12A	2.34	0.005	0.002	0.002		0.002	0.004	0.001			
	13A	4.28	0.008	0.001	0.004	-	0.002	0.004	0.001			
	14A	6.16	0.005	0.001	0.002	-	0.002	0.003	0 001			
٠	15A	-8.52	⊲ <b>₽.006</b>	0.002	0.004	0.001	0.002	0.004	-	;		
	16A	10.09	0.006	0.002	0.003	-	0.002	0.004	0.001	-		
	17A	12.53	0.002	0.002	0.002	» –	0.002	0.005	0.001.	-		
	18A	15.52	0.002	0.001	0.003	-	0.002	0.004	0.001	-		
	12B	2.45	0 005	0.002	0.003	-	0.002	0.004	0.001	0.038		
	13B	4.32	0.004	0.002	0.003		0.002	0 005	0.002	0 035		
	14B <sup>·</sup>	6.34	0.005	0.001;	0.004	-	0.002	0.005	0 001	0.037		
	15B	7,187	0.008	0.002	0.005	-	0.002	0.004	0.001	0.034		
,	16B	10.18	0.008	0.001	0.004	-	0 002	0.005	0.001	0 038		
	17B	12.60	0.007	0.002	0.004	-	0.002	0 005	0 001	0.034		
	3 <sup>1</sup> 8 <sup>1</sup> 8	15.51	0.007	0.002	0.004	-	0.002	0.004	0.001	0 036		
	12AA	2.21	0.005	0.001	0.004		0.002	0.006	0.001	-		
	13AA	3.93	0.002	0.001	0.003	-	0.004	0.004	0 002	-		
	14AA	6.33	0.003	0.001	0.003	-	0.001	0.006	0.002	. <del> </del>		
	15AA	7.83	0.005	0.001	0.005	<del>-</del> ,	0.003	0.005	0 001	-		
	16AA	10.34	0.005	0.001	0.005	-	0.003	0 005	0.001			
	17AA	12.58	0.004	0.001	0.005	-	0.003	0 006	0.001	-		
	12BB	2.24	0.004	0.002	0.003	-	-	0.003	0.001	0.036		
	13BB	4.10	0.004	0.001	0.003	-	-	0.002	0.001	0.036		
	14BB	6.43	0.006	0.001	0.002	-		0.005	0.001	0.029		
	15BB	7.71	0.003	0.001	0.004			0 004	0.001	0.034		
	16BB	10.26	0.005	0.001	0.003	·	-	0.005	0.01	0.035		
	17BB	12.44	0.005	0.001	0.003		-	0 005	0 001	0.032		

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Table 3.4 continued

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Alloy				Eleme	ent (wt.	pct.)	i.		
Name	Sı	Fe	Cu	Mn	Mg	Ni	Zn	Ті	Sr
1M1	0.003	0.005		0.006	0.089	0.003,	0.004	-	-
1M2	0.271	0.005	-	0.005	0.099	0.002	0.003	-	- '
1M3	0.493	0.004	-	0.004	0.099	0.002	0.002	-	-
1M4	0.805	0.006	-	0.004	0.102	0.002	0.003	-	_
1M5	1.182	Ó.005	-	0.002	0.105	0.002	0.002	-	
1M6	1.287	0.008	-	0.003	0.089	0.001	0.001		-
1M7	1.412	0.004	-	0.004	0.095	0.001	0.00,3	- {	-
1M8	1.559	0.004	-	0.003	0.103	0.001	0.004	-1	-
2M1	0.004	0.003	-	0.004	0.351	0.002	0.004		_
2M2	0.371	0.003	-	0.005	0.336	0.002	0.003	-	-
2M3	0.551	0.006	-	0.004	0.328	0.003	0.004	-	-
2M4	0.894	0.004	-	0.004	0.316	0.004	0.003	-	-
2M5	1.096	0.003	-	0.003	0.342	0.003	0.003		-
2M6	1.234	0.004		0.003	0.350	0.003	0.004	-	-
2M7	1.451	0.003	~	0.004	0.333	0.003	0.004	-	-
2M8	1.630	0.005	-	0.003	0.342	0.003	0.004	-	-
3M1	0.003	0.006	-	0.004	0.642	0.002	0.005	-	- 1
3M2	0.317	0.003	<u>د'</u>	0.005	0.637	0.002	0.005	-	-
змз	0.549	0.003	` —	0.005	0.620	0.002	0.006	-	-
3M4	0.812	0.003	-	0.004	0.658	0.003	0.003	-	-
3M5	1.043	0.005	-	0.005	0.612	0.003	0.004	-	-
3M6	1.216	0.006	-	0.006	0.648	0.003	0.003	-	-
3M7	1.436	0.004	-	0.006	0.643	0.002	0.003	-	-
3M8	1.559	0.004		0.006	0.638	0.002	0.003		-

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Table 3.4 continued

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Alloy		Element (wt.pct.)										
Name	Si	Fe	Cu	Mn	Mg	Nı	Zn	Ti	Sr			
4M1	0.003	0.004	_	0.005	0.752	0.003	0.004		-			
4M2	0.247	0.004	-	0.005	0.748	0.002	0.003	-	-			
4M3	0.445	0.005	-	0.004	0.742	0.003	0.004	-	-			
4M4	0.832	0.002	-	0.004	0.738	0.003	0.003		, <del>-</del>			
4M5	1.071	0.002	-	0.003	0.762	0.003	0.003	-	-			
4M6	1.187	0.003	-	0.004	<b>'</b> 0.773	0.002	0.003	-	-			
4M7	1.443	0.004	-	0.005	0.755	0.002	0 004	-	-			
4M8	1.622	0.004		0.005	0.746	0.003	0.004	-	-			
<sup>5</sup> 5M1	0.03	0.004	-	0.003	1.034	0.003	0.004	-	-			
5M2	0.225	0.005	-	0.005	1.008	0.003	0:003					
5M3	0.416	0.002		0.005	1.142	0.003	0.004	_1	- -			
5M4	0.854	0.002	-	0.005	0.988	0.004	0.003	-				
5M5	0.992	0.003	-	0.006	1.034	0.005	0.004	-	-			
5M6	1.198	0.005	_	0.007	1.104	0.003	0.004	-				
5M7	1.438	0.004	-	0.003	0.977	0.003	0.004	-	-			
<sup>,</sup> 5M8	1.608	0.004	-	0.004	1.086	0.003	0.004	-				
1M1B	0.004	0.004	_	0.005	0.086	0.003	0.006	-	0.035			
1M2B	0.214	0.004	-	0.004	0.092	0.003	0.006	-	0.037			
1M3B	0.466	0.003	-	0.004	0.095	0.002	0.007	-	0.032			
1M4B	0.815	0.002	/-	0.004	1.021	0.002	0.005	-	0 031			
1M5B	0.945	0.003	<i>'</i> –	0)003	1.015	0.002	0.005	-	0 029			
1M6B	1.176	0.003	-	0.003	0.092	0.002	0.004		0.036			
1M7B	1.426	0.004	-	0.003	0.087	0.002	0.005	-	0.034			
1M8B	1.543	0.003	-	0.004	0.093	0.002	0.005	a <del></del>	`0 <i>.</i> 034			

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# Table 3.4 continued

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Alloy	Element (wt. pct.)										
Name	Sı	Fe	Cu /	, Mn	Mg	Ni Ni	Zn	Tı	Sr		
2M1B	0.004	0.002		0.005	0.361	0.004	0.005		0.034		
2M2B	0.279	0.002	'	0.004	0.334	0.003	0.006	-	0.037	r.	
2M3B	0.493	0.004	-	0.004	0.316	0.002	0.007	-	0.033	Ø.	
2M4B	0.805	0:005	-	0.005	0.308	0.003	0.006	· . <del></del>	0.035		
2M5B <sup>3</sup>	1.170	0.006	-	0.005	0.365	0.002	0.007	-	0.035		
2M6B	1.126	0 004	-	0.003	0.354	0.002	0.004 <sup>,</sup>	· · ·	0 036	π	
2M7B	1.433	0.00	-	0.004	0.326	0.002	0.005	ر <del>مد</del> ار م	0 036	م به ۹۰۰ (ب. د	
2M8B	1.559	0.004	-	0.003	0.330	0.002	0,005		0.037		
3M1B	0.004	0.003		0.004	0.687	0.001	0.005	0.003	0.034	υ	
3M2B	0.271	0.003	-	0 005	0.626	0.002	0រ007	0 002	0.033		
ЗМЗВ	0.483	0.003	iki	0.004	0.618	0.002	'0.006	0.002	0.034		
3M4B	0.790	0.004	-	0.004	0.630	0.003	0.0Q5	0.001	0 037		
3M5B	1.179	0.004	-	0.003	0.642	0.002	0.003	0.002	0.035		
3M6B	1.244	0.003	-	0.006	0,638-	0.002	0.005	0.003	0 031	-	
3M7B	1.436	0.003	-	0.005	0.625	0.003	0.004	0.004	0 037	L	
` <b>3M8</b> B	1.569	0.003	, -	0.003	0.655	0.001	0.004	0.003	0.035		
4M1B	0.003	0.004	- '	0.003	0.771	0.001	0.003	0.001	0.031		
4M2B	Q.269	`0 <i>.</i> 005		0.003	0.725	-	0.004	0.002	0 028	e	
4М3В	0.392	0.004	-	0.003	0.739	-	0.004	0.001	0 0344	۲	
4M4B	0.805	0.004	-	0.004	0.740	0.002	0.001	0.001	0 036		
<sup>7</sup> 4M5B	1.185	0.006	' <b>-</b>	0.005	0.755	-	0.001	<b>0</b> .001	0`038		
4M6B	1.267	0.003	-	0.003	0.759	0.003	0.001	0.001	0.035		
4M7B	1.418	0.003	-	0.004	0.760	0.002	0.002	0.001	0.035		
4M8B	1.559	0.003	,-	0.004	0.738	0.001	0.001	0.001	0.036		

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Table 3.4 contrnued

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	Fable 3.4 continued										
	Allov				Eleme	nt (wt.	pct.)		، 	·	
N	Name	Si	Fe	Cu	Мn	Mg	Ni	Zn	T,i	Sr	
	5M1B	0.003	0.005	ann a lan a da bhann a la an aile	0.004	1.105	0.007	0.003	0.002	0.037	
	`5M2B	0.263	0.005	-	0.003	1.008	0.006	0.001	0.002	0.034	
	5M3B	0.493	0.004	-	0.004	1.115	0.004	0.003	0.002	0.033	
	5M4B	0.821	0.003	-	0.005	0.975	0.004	0.004	0.001	0.035	
-	5M5B	1.182	0.004		0.004	0.968	0.005	0.003	0.001	,0.033	
A <sup>L</sup>	<u>.</u> 5M6B	.1.302	0.006	- ,/	0.003	1,018 <sup>.</sup>	0.003	0.003.	"Q.QQ2	0.032	
	5M7B	1.442	0.005	_	0.003	1.125	0.003	0.003	0.001	0.031	
÷.,	5M88 🖙	·1.543	0.004	-	0 003,	1.034	0.004	0.003.	0.002	0.032	
	6M1	2.34	0.002	0.002	0.005	0.252	0.0 <u>Q</u> 2.	_0 006	0.002	-	
	6M2	4.33	0.003	0 002	0.005	0.272	0.002	0.004	0.002	-	
	6M3	6.24	0.003	0 003	0.005	0.331	0.002	0.005	0.003	-	
,	6M4	8.54	0.004	0.003	0.004	0.281	0.002	0.005	0.002	-	
	6M5	10.61	0.005	0.003	0.004	0.297	0.003	0.004	0.0 <b>0</b> 3	-	
	6M6	12.86	0.005	0.003	0.004	0.3.13	0.005	0 005	0.002	-	
	7M1	2.14	0.003	0.001	0.004	0:771	Q.002	0 00 1	0.001	-	
	7M2	4.36	0.003	0.002	0.004	0 <sup>C</sup> 761	0.002	0.001	0.001	- `	
	7M3	6.16	0.00Ź	0.001	0.002	0.769	·0.002	0.002	0.001	-	
a ,	7M4	8.08	0.005	0.001	0.002	0.721	0.002	0.002	0.001	- `	
	7M5	10.17	0.005	0.002	0.004	0.7 <b>6</b> 4	0.003	0 003	0.001	-	
-	7M6	12.45 <sup>°</sup>	0.004	0.001	0.004	0.734	0.003	0.004	0.001	-	
	, 8M1	2.21	0.007	0.002	0.004	0.334	0.004	0.007	0.002	-	
,	8M2	3.98	0.005	0.0Q1	0.004	0.308	0.002	0.005	0.002	-	
	8M3	5.84	0.005	0.002	0.003	0.312	0.002	0.007	0.002		
	8M4	7.88	0.004	0.002	0.002	<sup>K</sup> 0.296	0.002	0 006	0.002	-	
	8M5	10.01	0.004	0.002	0.004	0.278	0.004	0.006	0.002	-	
	8M6	12.46	0.005	0.002	0.003	0.311	0.004	0.006	0.002	-	

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# Table 3.4 continued

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Alloy	Element (wt. pct.)								
Name	Si	Fe	£u	Mn	Mg	Ni	Zn	Τı	Sr <sub>,</sub>
6M1B	2.46	0.002	0.002	0.005	0.282	0.003	0.004	0.001	0.031
6M2B	4.22	0.003	0.003	0.005	0.294	0.002	0.003	0.001	0.026
6M3B	6.18	0.002	Q.003	0.004	0.338,	0.001	0.003	0.002	0 033
6M4B	8.12	0.002	0.002	0.005	0.271	0.004	0.003	0 002	0.036
6M5B	10.26	0.002	0.002	0.004 ·	0.281	0.003	0.004	0.002	0.030
6M6B	12.30	0.004	. 0.003	0.004	0.326	0.003	0.007	0.002	0.028
7M1B	2.41	0.004	0.001	0.005	0.714	0.003	0.005	0.002	0.029
<b>7M2B</b> ջ	4.17	0.003	0.002	0.006	0.766	0.003	0.007	0 002	0 032
7M3B	6.08	0.003	° 0.002	0.005	0.772	0.003	0.005	0 002	0.035
7M4B	8.06	0.003	0.001	0.005	0.786	0.003	0 005	0.002	0.033
7M5B	10.17	0.003	0.003	0.005	0.737	0.003	0.004	0.002	0 032
7M6B	12.48	0.004	0.002	0.004	0.742	0.004	0.005	0.002	0.032
8M1B	2.08	0.006	0.002	0.004	0.286	0.004	0.005	0.002	0.032
8M2B	4.12	0.004	0.001	0.004	0.322	0.004	0.007	0.002	0.032
<b>8</b> M3B	6.04	0.004	0.001	0.004	0.308	0.004	0.006	0.002	0 028
8M4B	8.01	0.005	0.00,2	0.005	0.284	0.004	0.006	0 002	0 031
8M5B	10.15	0.005	0.002	0.005	0.296	0.004	0.006	0 002	0.031
8M6B	12.52	0.006	0.002	0.005	0.314	0.004	0.007	ò.002	0.030
9M1B	2.34	0.005	0.001	0.004	0.322	0.003	0.007	0.001	0.061
9M2B	4.08	0.005	0.002 -	0.005	0,360	0.004	0.007	0 002	0 056
9M3B	6.09	0.003	0.002	0.005	0.278	0.004	0.006	0.002	0.061
9M4B <sup>′</sup>	<b>8</b> .26	0.003	0.002	0.007	0.292	0.003	0.007	0.002	0 062
9M5B	10.32	0.004	0.002	0.005	0.324	0.004	<b>.006</b>	0.002	0 058
9M6B	12.56	0.005	0.002	0.005	0.306	0.004	0.004	0.002	0.059

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Table 3.4 continued

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Alloy	Element (wt. pct.)										
Nàme	Si	Fe	Cu	Min	Mg	Ní	<sup>•</sup> Zn	Ті	Sr		
D1	12.45	0.002	••••	0.004		0.004	0.006	0.002	-		
D2	12.52	0.002		0.004	-	0.002	0.005	0.002	-		
D3	12.10	0.002	~	0.006	-	0.003	0.004	0.002	-		
• D4	12.86	0.003	-	<u>0.003</u>	-	0.003	0.004	0.002	-		
D5	12.55	0.003		0.004		0.003	0.004	0.002	-		
D6	12.44	0.004	-	0.005	-	0.001	0.007	0.002	-		
<b>S</b> 1	12.68 <sup>,</sup>	0.006	~	0.004	-	0.003	0.006	0.002			
	12.42	0.006	~	0.004	<u> </u>	.0.004	0.005	0.002	0.035		

#### **RESULTS AND DISCUSSION**

As described in chapter 1, the main objective of this work was to provide a more complete understanding of how strontium additions affect the resistivity behaviour of Al-Si alloys. Therefore, the investigation was divided into three areas:

- 1) a study of the resistivity/conductivity behaviour of solid solution Al-Si and Al-Si-Mg alloys,
- a study of the effects of varying silicon morphologies on the electrical resistivity/conductivity of two-phase AI-Si and AI-Si-Mg alloys,
- 3) a study of the electrical conductivity changes that occur during heat treatment of the two-phase alloys.

The results pertaining to the solid solution alloys will be discussed in chapter 4 and those dealing with the two-phase alloys will be given in the following two/chapters.

## CHAPTER 4

**RESISTIVITY/CONDUCTIVITY BEHAVIOUR OF THE SOLID SOLUTION ALLOYS** 

#### 4.1 General

Earlier studies<sup>22,25,26</sup> focussed entirely on resistivity changes observed in hypoeutectic Al-Si casting alloys (wt., pct. Si > 5) when strontium was used as a modifier. However, no available data exists on the effect of strontium on the resistivity of solid solution alloys in which the silicon content is less than 1.6 wt. pct.

When an alloying element is added to a solid solution, the resistivity may change for a number of reasons:

- i) dissolution of the alloying element in the solid solution,
- ii) changes of the solid solubility limits of other elements present in the solid solution,

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 iii) changes of the precipitation kinetics of solute atoms in the case of heat treated solid solution alloys.

In the following sections it will be determined whether or not strontium alters the resistivity of aluminum, Al-Si and Al-Si-Mg solid solution  $\beta$  alloys, and if so the reasons for the change in resistivity will be established.

4.2 Resistivity/Conductivity of Al-Sr Solid Solution Alloys

4.2 1 Room Temperature Conductivity

Figure 4.1 shows the room temperature conductivity of binary Al-Sr ralloys containing up to 0.26 wt pct. Sr for all of the various heat treatments used. The addition of up to 0.26 wt. pct. Sr has virtually no effect on the conductivity, which remains constant at the 64 pct. IACS level. Also, the conductivity is constant regardless of the heat treatment process. These observations would tend to indicate that strontium has an extremely low solid solubility in aluminum at room temperature.

4.2.2 Residual Resistivity Ratio

As noted in chapter 2 (section 2.4), the temperature dependent resistivity ( $p_{th}$ ) does not depend on alloying element types and concentrations nor on lattice defects and it reaches a negligible value at very low temperatures. Hence, the determination of resistivity at 4.2K gives an accurate measure of the residual resistivity ( $p_0$ ) which is the sum of the contributions from each of the impurities, alloying elements and lattice imperfections such as vacancies, dislocations and grain boundaries. Since the resistivity contribution of lattice imperfections is very much smaller than that of the alloying elements, particularly as the samples were annealed prior to measurements, then it can be assumed that

 $\rho_{Total} \simeq \rho_o \simeq \rho_{Alloying}$  at 4 2K

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(4.1)





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Consequently, changes in the residual resistivity ratio (RRR) are caused by the A alloying elements.

The variation of the RRR of aluminum with an increasing amount of strontium is shown in Figure 4.2. Here, it is seen that a rapid decline of the RRR from 155 to 135 occurs up the  $\sim$ 0.03 wt. pct. Sr. At higher strontium levels the decline is less marked, thereby indicating an approximate solubility limit of only 0.03 wt. pct. Sr at 560°C. This suggests a negligible strontium solubility at room temperature, as was postulated on the basis of the room temperature conductivity. Hence, there is no significant contribution of dissolved strontium to the conductivity of the aluminum matrix. Further evidence of this lack of solubility is to be found in reports that the change in the lattice parameter of aluminum caused by the dissolution of strontium is too small to measure.<sup>75</sup>

#### 4.3 Resistivity/Conductivity of Al-Sr and Al-Si-Sr Solid Solution Alloys

4.3.1 Room Temperature Conductivity of Quenched and Air Cooled Alloys

The changes in electrical conductivity at room temperature of aluminum containing up to 1.6 wt. pct. Si and ternary Al-Si-Sr alloys with 0.035 wt. pct. Sr are presented in Figure 4.3. The conductivity of both the quenched and air cooled samples of binary and ternary alloys first declines sharply to 48 pct. IACS at 1.3 wt. pct. Si and then gradually to 47 pct. IACS at 1.6 wt. pct. Si. The linearity of the first part of the curve (i.e. Si < 1.3 wt. pct. ) shows that the silicon atoms are present in aluminum solid solution up to 1.3 wt. pct. The later change in the slope of the curve indicates that the silicon concentration exceeds its solid solubility limit. The rate of the decrease of conductivity calculated from the slope of the curve is 12 pct. IACS per wt. pct. Si. This corresponds to an increase in resistivity of 0.70µ $\Omega$ cm per wt. pct. Si, and is in agreement with reported values.53,54,92<sup>*i*</sup>

As can be seen in Figure 4:3, there is no significant difference in the conductivities of quenched and air cooled samples. This suggests that the air cooling was rapid enough to retain silicon in the aluminum matrix, and





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Figure 4.3 Room temperature electrical conductivity of Al-Si binary and Al-Si-Sr ternary alloys.

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therefore the conductivity measurements, show no evidence of silicon precipitation in air cooled samples. Also shown in Figure 4.3 is the effect of 0.035 wt pct. Sr on the conductivity of Al-Si alloys. Strontium has no noticeable influence on room temperature conductivity, indicating that strontium does not affect the solubility of silicon in the aluminum matrix.

4.3.2 RRR of Air Cooled and Aged Alloys

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The residual resistivity ratio of aluminum containing additions of silicon up to 1.6 wt. pct. is given in Figure 4.4 for air cooled samples. The RRR is reduced drastically by the addition of silicon up to 0.3 wt pct and more slowly thereafter. Finally, it becomes constant at 4.8 when the silicon concentration exceeds 1.30 wt. pct. The initial major drop in RRR with small additions of silicon (< 0.3 wt. pct) is expected due to the considerable reduction in the purity of the alloys. Here, the solubility limit of silicon is defined as the point at which the curve becomes horizontal. When the silicon content is higher than the solid solubility limit, excess silicon atoms are rejected by the aluminum matrix. It is well known that these rejected atoms have a smaller effect on the RRR than the dissolved silicon atoms.44,77 Apparently, the solubility limit from Figure 4.4 is  $\sim$ 1.3 wt. pct. Si, and is in agreement with the room temperature conductivity data. However, this value is noticeably lower than the equilibrium solid solubility limit of silicon at 560°C, which is 1.5 wt. pct. Si.<sup>1</sup> The discrepancy between these two values results from silicon segregation and non-equilibrium cooling during solidification of the alloys. The silicon segregation leads to the formation of eutectic silicon at concentrations of less than 1.62 wt. pct. and prevents further silicon dissolution in the matrix. The eutectic silicon phase found in the as-cast binary Al-Si and ternary Al-Si-Sr alloys containing 1.4 wt pct. Si is shown in Figure 4.5

The data of Figure 4.4 confirms the result obtained from the room temperature conductivity measurements; namely, that strontium at this level (0.035 wt. pct.), which is typical of that used for modification, does not alter the solid solubility limit of silicon in the aluminum matrix and hence the resistivity of the matrix phase. A more detailed comparison of the data presented in Figure 4.3 and Figure 4.4 is given in Appendix IV.

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a) no Sr





The RRR of aged Al-Si and Al-Si-Sr alloys is also given in Figure 4.4. The RRR decreases rapidly to 30 with the addition of silicon up to 0.15 wt. pct. and remains constant at all silicon levels in excess of 0.15 wt. pct., which is the maximum solid solubility limit of silicon at 340°C. The higher RRR on aging compared to air cooling (see Figure 4.4) is no doubt due to the rejection of silicon by the aluminum matrix during aging.

4.3.3 Room Temperature Conductivity of Heat Treated Alloys

The results obtained using different heat treatment processes can be summarized as follows:

- i) Air cooling As noted earlier, air cooling does not lead to significantly different conductivity values when compared to values measured from quenched samples (Figure 4.3).
- ii) Furnace cooling: As expected, alloys which are furnace cooled show less dependence of the conductivity on the silicon concentration, as seen in Figure 4.6 However, it is evident in Figure 4.6 that the ternary alloys have relatively lower conductivity values than the binary alloys when the silicon concentration exceeds about 0.5 wt. pct. It appears that strontium exerts a retarding effect on silicon precipitation during furnace cooling, which will be discussed in detail below
- III) Aging: Natural aging (i.e. room temperature aging) of both the binary and ternary alloys was followed by means of conductivity measurements. Figure 4.7 shows that there is no change in the conductivity of water quenched samples with aging time, thus giving no indication of pre-precipitation of silicon (i.e. G.P. zone formation during room temperature aging)

Figure 4.8 illustrates the relative conductivity change with time during the isothermal aging of three Al-Si and Al-Si-Sr alloys at 340°C. The general characteristic of the curves is sigmoidal, demonstrating classical nucleation and growth kinetics. Thus, the change in conductivity increases slowly at first, then accelerates, and then decreases as precipitation reaches completion. The presence of strontium (0.035 wt\_pct.) does not change this


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Figure 4.6 Room temperature electrical conductivity of furnace cooled Al-Si and Al-Si-Sr alloys.

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Figure 4.7 Variation of the conductivity of AI-Si and AI-Si-Sr alloys during room temperature aging.



Figure 4.8 Relative conductivity changes in Al-Si binary and Al-Si-Sr ternary alloys during aging at 340°C.

general shape of the curves, but does act to retard the precipitation process as seen by the shift to the right of the curves. However, this effect of strontium gradually diminishes with longer aging times (>6h).

The overall kinetics of the silicon precipitation process, including the nucleation and growth stages, can be treated using the Johnson-Mehl-Avrami equation; 134, 135

$$x(t) = 1 - exp(-kt^n)$$

(4.2)

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where x(t) = volume of precipitate formed at time t n = Avrami constant k = rate constant

x(t) can be conveniently expressed as;

$$x(t) = \frac{\sigma_t - \sigma_o}{\sigma_F - \sigma_o}$$
(4.3)

where  $\sigma_0$  = initial conductivity (i.e. quenched conductivity)  $\sigma_F$  = final conductivity  $\sigma_t$  = conductivity at time t

Equation 4.2 can be converted into common logarithms; &

$$log\left(\frac{1}{1-x(t)}\right) = \frac{1}{2} \frac{1}{3} (kt^n)$$

and

$$\log \log \left(\frac{1}{1-x(t)}\right) = n \log(t) + \log(k) - \log(2.3)$$
 (4.4)

Experimental conductivity results were analyzed by plotting curves of log log  $(1/(1 - x_t))$  against log(t). This gave straight lines of slope n, as shown in Figure 4.9. The rate constant, k, for each alloy was then calculated from experimental results assuming that 1/k equals the time at which log log  $(1/(1 - x_t)) = -0.362$ . Some of the calculated k values, together with Avrami

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constant n, are given in Table 4.1, and the complete data is presented in Appendix I. As seen in Table 4.1, the n values obtained for Al-Si and Al-Si-Sr alloys do not show any significant difference and are in the range of 0.70 to 0.80 with a mean of 0.75; however, the rate constants k of the Al-Si-Sr alloys are smaller (~25%) than those of the binary alloys containing the same level of silicon. This implies that the silicon precipitation processes in binary and ternary alloys are isokinetic, but the rate of the process is appreciably slower in the ternary Al-Si-Sr alloys than the binary Al-Si alloys. It was also observed that the rate of precipitation increases with increasing silicon concentration, as seen in Table 4.1. Such observations have been also reported by other authors<sup>136,137</sup> and clearly result from achieving a higher degree of supersaturation of the alloys with increasing silicon content after water quenching, thus providing an increased driving force for the precipitation reaction on aging.

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A comparison of the experimentally calculated Avrami constants with those determined from theoretical models (Table 4.2)<sup>134,135</sup> shows that the model given by Cottrell-Bilby<sup>138</sup> and Harper<sup>139</sup> provides the best agreement with the experimental results. This model, in which the n value is 2/3, suggests that precipitation occurs on dislocation lines and its rate is controlled by the diffusion process of solute atoms near the precipitatematrix interface (i.e. diffusion-controlled growth). In addition, extensive studies carried out by Rosenbaum and Turnbull<sup>105,140</sup> using X-ray diffraction and microscopy revealed evidence for the precipitation of silicon on dislocation loops that form due to the collapsing of excess quenched-in vacancies. More recently, Ozawa and Kimura<sup>136</sup>, and Hammand et al.<sup>141</sup> found heterogeneously nucleated silicon precipitates on dislocation loops by transmission electron microscopy.

Silicon precipitates in the binary alloy exist with equiaxed, rod and plate-like morphologies, as was reported previously<sup>105,140</sup>. However, in the ternary alloys the precipitates are mainly equiaxed and smaller, particularly in the furnace cooled samples as shown in Figure 4.10. These observed changes in size and morphology of the precipitates implies that strontium reduces the growth rate of silicon and perhaps changes its growth

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% Si	% Sr	n	k (h - n)
0.60	P	0.75	0.36
	0.035	. <b>7</b> 0	0.28
1.03	_	0.80	0.71
	0.035	0.77	0 57
1.40		0.77	1.25
	0.035	0.73	1.00

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Table 4.1 Experimentally Calculated Values of n and k

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## Table 4.2 Values of n in Kinetic Law; $y = 1 - exp(-kt^n)$

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Model	n
Nucleation at a constant rate and growth of a eutectoid	4
Nucleation at a constant rate and diffusion controlled growth	5/2
Precipitation on dislocations and diffusion controlled growth	<u></u> 2/3
Diffusion controlled growth of a fixed number of particles	3/2
Growth of a fixed number of particles limited by the interface process *	3
Diffusion controlled growth of discs of constant thickness	2 '
Diffusion controlled growth of cylinders in axial direction only	1
cylinders in axial direction only	4

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a) no Sr



b) 0.035 wt. pct. Sr

Figure 4.10 Silicon precipitates formed in Al-1.03 wt pct Si alloys furnace cooled from 560°C. -1

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mechanism. Table 4.3 indicates that the population of the silicon precipitates in binary and ternary alloys is of a similar order. Therefore, it appears that strontium does not markedly influence the density of dislocation loops, and hence the nucleation kinetics of the silicon precipitation process.

As reported by Harper<sup>139</sup>, k is proportional to dislocation density and diffusivity of solute atoms. As noted above, it is unlikely that strontium reduces the dislocation density, but it may cause changes in diffusivity of silicon atoms. Correspondingly, the retardation of growth of the silicon particles in the Al-Si-Sr alloys may result from a lowering of the diffusion rate of silicon atoms towards the nucleated particles due to segregation of strontium atoms to regions of the aluminum matrix adjacent to the silicon particles. As proposed by many authors<sup>29,30,41-43</sup>, the segregation or accumulation of strontium near the nucleated eutectic silicon phase takes place during the solidification of modified Al-Si casting alloys. This results in poisoning the preferential growth directions of the silicon crystals, thereby making growth more isotropic and restricted. It can be expected that strontium behaves similarly during the precipitation of silicon and reduces the anisotropic growth of silicon particles along the preferential directions of {111}Sill{111}AI and {111}Sill{100}Al140. The result is slower but more uniform growth of silicon (i.e. equiaxed crystals) and retardation of silicon rejection from the aluminum matrix. This retarding influence of strontium on the silicon precipitation process is reflected by the experimentally calculated k values being lower for the strontium-containing alloys.

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The influence of strontium on the precipitation of silicon is more profound in the furnace cooled samples in which the nucleation of precipitates is relatively difficult due to the lack of heterogeneous nucleation sites (e.g. dislocation loops). Therefore, the precipitation process is dominated mainly by growth on already nucleated particles. Thus the effect of strontium on the precipitation rate becomes more significant, as shown in Figure 4.11.

Temper	% Si	_% Sr	No. of particles Shape of	
			(N/cm - 2)	the Particles
Furnace Cooled	1.03		275	Rod, plate, equiaxed
	1.03	0.035	255	Equiaxed
Aged at 340°C for 5h	0.6	-	2310	Rod, plate, equiaxed
	0.62	0.035	2100	Plate, equiaxed

Table 4.3 Number and Shape of the Precipitates Found in Al-Si and Al-Si-Sr Alloys

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Figure 4.11 Relative conductivity changes in furnace cooled and aged Al-Si and Al-Si-Sr alloys with silicon concentration.

4.4 Conductivity of Al-Si-Mg and Al-Si-Mg-Sr Solid Solution Alloys

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Magnesium is a common alloying element in both wrought and cast aluminum base alloys. Its maximum equilibrium solid solubility in aluminum is 17.4 wt. pct., but the magnesium level in casting alloys rarely exceeds 0.75 wt. pct. Magnesium is used with silicon to produce Mg<sub>2</sub>Si precipitates in heat treated alloys in order to improve the mechanical properties.<sup>53</sup> In the following sections, the effect of the combined presence of magnesium and silicon, with and without strontium additions, on the conductivity of the aluminum solid solution will be considered.

4.4.1 Room Temperature Conductivity of Quenched Al-Si-Mg and Al-Si-Mg-Sr Alloys

Figure 4.12 presents the changes in conductivity of quenched Al-Si Ma and Al-SI-Mg-Sr alloys with increasing silicon and magnesium concentrations from 0 to 1.6 and 1 wt. pct., respectively. The conductivity of the ternary Al-Si-Mg and guaternary Al-Si-Mg-Sr alloys decrease linearly with increasing silicon content at a rate of  $\sim$  12 pct. IACS per wt. pct. Si; the same rate observed in binary Al-Si alloys. A magnesium addition also reduces the overall conductivity, as can be seen in Figure 4.12 by the suppression of the curves to lower conductivity values. At a given silicon concentration, the decrease in conductivity per wt. pct. Mg is about 10 pct. IACS or the increase in resistivity is 0.58 $\mu$ Ωcm. This agrees with reported values.<sup>53,54</sup> Since both elements dissolve in solid aluminum, their contribution to the resistivity of the aluminum matrix should be additive. This was checked by comparing the measured resistivity changes ( $\Delta \rho m_{AI}$ ) with the calculated resistivity changes  $(\Delta \rho c_{AI})$  from the Matthiessen additive rule (Table 4.4). Since the conductivity values are not additive, the measured pct. IACS values are converted to resistivities ( $\rho_A$ ) ( $\mu\Omega$  cm) in Table 4.4.  $\Delta\rho^m_{AI}$  was obtained by subtracting the resistivity of pure aluminum ( $\rho_{AI} = 2.68 \mu \Omega cm$ ) from the resistivity of the alloys.  $\Delta \rho^{c} A \Gamma^{c}$  was calculated using the following relationship;



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% Sı	% Mg	ρ <sub>Α</sub> † (μΩ cm)	Δρ <sup>m</sup> Al ຈ (μΩ cm)	- Δρ <sup>c</sup> Al (μΩ cm)
0.32	0.33	3.07	0.39	0.41
0.60	0.33	3.26	0.58	0.61
0.96	0.33	3.52	0.84	0 86
0.32	0.65	3.24	0.58	0.60
0.60	0.65	3 49	0.81	0.80°
0.96	0.65	3.83 A	1.15	1.08
0.32	1.00	3.47	0.79	0.80
0.60	1.00	3.73	1.05	1.00
0.96	1.00	4.09	1.41	1.28

Table 4.4 The Combined Effect of Magnesium and Silicon on the Resistivity of Aluminum

†  $\rho_A = (resistivity of the alloy) = (172.6/IACS)$ 

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 $\Delta \rho_{Al}^{c} = (\Delta \rho_{Sl} \times w_{l} \% S \iota) + (\Delta \rho_{Mg} \times wt \% Mg)$ 

where  $\Delta \rho_{Si} = 0.70 \mu \Omega \text{ cm/wt. pct. Si}$  $\Delta \rho_{Mg} = 0.58 \mu \Omega \text{ cm/wt. pct. Mg}$ 

The results seen in Table 4.4 confirm that the combined effect of magnesium and silicon on resistivity obeys the Matthiessen additive rule.

The addition of magnesium slightly reduces the solubility of silicon in quenched samples by shifting the inflection point on the curves toward lower silicon contents. For example, the silicon solubility is 1.3 wt. pct. for the alloy containing 0.1 wt. pct. Mg whereas it is about 1.0 wt. pct. for the alloy containing 1 wt. pct. Mg (see Figure 4.12). As is evident in Figure 4.12, strontium exerts essentially no effect on the conductivity of Al-Si-Mg solid solution alloys; hence it does not change the solid solubility limit of silicon or magnesium in the aluminum matrix.

4.4.2 Conductivity of Heat Treated Alloys

i) Air cooling: The air cooled samples (Figure 4.13) exhibit noticeably higher conductivity than the quenched samples (Figure 4.12) due to the formation of Mg<sub>2</sub>Si compounds. Initially, the conductivity of air cooled alloys decreases as the silicon and magnesium contents increase. When the alloys contain <sup>o</sup> more than 0.8 wt. pct. Si and 0.6 wt. pct. Mg their conductivity approaches a more or less constant value in the range of 46-48 pct. IACS.

ii) Furnace cooling: Furnace cooling allows the formation of a considerable amount of Mg<sub>2</sub>Si, thereby altering the aluminum matrix composition and increasing conductivity. Therefore, the conductivity of furnace cooled alloys depends less on the chemical composition of the alloys, as shown in Figure 4.14. Since the approximate weight ratio of magnesium to silicon needed to form Mg<sub>2</sub>Si is 1.73:1, most of the alloys contain different levels of excess silicon, which are calculated assuming that all the magnesium combines with

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 Figure 4.13
 Room temperature electrical conductivity of air cooled

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 AI-Si-Mg and AI-Si-Mg-Sr alloys.

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Figure 4.14 Variation of conductivity in furnace cooled AI-Si-Mg and AI-Si-Mg-Sr alloys with Mg<sub>2</sub>Si content.

silicon to form Mg<sub>2</sub>Si. The excess silicon only slightly lowers the furnace cooled conductivity, which ranges from 57.5 to 52 pct. IACS. In addition, the presence of strontium does not influence the conductivity of the furnace cooled samples. The optical micrographs obtained from both Al-Si-Mg and Al-Si-Mg-Sr alloys are presented in Figure 4.15. Mg<sub>2</sub>Si precipitates appear in rod, plate and cubic forms in both the ternary and quaternary alloys with a similar density of ~300 particles/cm<sup>2</sup>. However, rod precipitates are finer with strontium present.

iii) Aging: Ternary Al-Si-Mg solid solution alloys exhibit a different precipitation behaviour compared with the Al-Si systems discussed in section 4.3.3. In the Al-Si system, only the equilibrium silicon phase precipitates; however, there are several precipitation products in the Al-Si-Mg system and the precipitation sequence is given as;<sup>142</sup>



 $\circ$ G.P. zone formation does not involve a nucleation process (i.e. direct growth process) therefore no incubation period is observed<sup>102</sup>, but the  $\beta$ ' rods and  $\beta$ -plates are heterogeneously nucleated on dislocations or other, lattice defects such as vacancies.<sup>142</sup> Additionally, more complex nucleation and growth processes have been reported in the literature, depending on the details of the heat treatment applied <sup>143</sup>

Natural aging: Figure 4.16 shows that the relative conductivity of both the ternary and quaternary alloys decreases on isothermal aging at room temperature. A feature of the variation in the relative conductivity with aging time is that the conductivity decreases very rapidly in times of the order of 1h followed by a gradual decrease over a period of 12h. The initial rapid change is known as the fast reaction, and is associated with clustering of silicon and magnesium atoms (i.e. G.P. zone formation). This process is the result of the high diffusion rate of solute atoms which is enhanced by the large vacancy super-saturation caused by quenching 102 The subsequent

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b) 0.03 wt. pct. Sr

Figure 4.15 Mg<sub>2</sub>Si precipitates formed in Al-0.95 wt. pct<sup>\*</sup> Mg<sub>2</sub>Si alloys furnace cooled from 540°C.

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Figure 4.16 Relative conductivity changes in Al-Si-Mg and Al-Si-Mg-Sr alloys during natural aging.

gradual change, known as the slow reaction, is caused by the growth of zones of diffusion.

This change in relative conductivity increases with increasing Mg<sub>2</sub>Si contents of the alloys, as illustrated in Figure 4.16. For example, the total change in conductivity of an alloy containing 0.55 wt. pct. Mg<sub>2</sub>Si is approximately 1.5 pct. at the end of 12h aging while the conductivity of an alloy containing 1.60 wt. pct Mg<sub>2</sub>Si changes by about 3 pct. Correspondingly, for increasing excess silicon levels in an alloy the change in conductivity occurs faster and to a greater extent, as illustrated in Figure 4.17. The effect of excess silicon on the aging process is in agreement with the observations of Kovacs et al.<sup>102</sup> and results from the favoured clustering the general characteristics of the aging process (i.e. change in conductivity) is unaffected by the addition of strontium to Al-Si-Mg alloys, as seen in Figures 4.16 and 4.17.

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The formation of G.P. zones leads to a decrease in conductivity since the clusters cause a distortion of the lattice (i.e. coherency strains which strongly scatter the electrons in addition to increasing the hardness). Therefore, the hardness measurements, carried out simultaneously with conductivity measurements, show the same characteristics as the aging process (Figure 4.18) although the relative change in hardness is more substantial (~60%) than in conductivity.

<u>Artificial aging</u>: The relative changes in the conductivity of Al-Si-Mg and Al-Si-Mg-Sr alloys during isothermal aging at 170°C are presented in Figure 4.19. The conductivity of both the ternary and quaternary alloys increases with increasing aging time. The changes in conductivity reflect the volume of magnesium and silicon atoms rejected from the aluminum matrix to form Mg<sub>2</sub>Si precipitates. As a result, a larger increase in conductivity occurs in alloys containing higher levels of Mg<sub>2</sub>Si, as shown in Figure 4.20. The presence of 0.035 wt. pct. Sr in these alloys does not change the shape of the aging curve indicating that the kinetics of the aging process remain unchanged. However, the total change in conductivity, for a given Mg<sub>2</sub>Si



Figure 4.17 The effect of excess silicon on the relative conductivity change in Al-Si-Mg alloys during natural aging.

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Figure 4.19 Relative conductivity changes in Al-Si-Mg and Al-Si-Mg-Sr alloys during artificial aging.



Figure 4.20 The effect of Mg<sub>2</sub>Si content on the relative conductivity changes in Al-Si-Mg and Al-Si-Mg-Sr alloys during artificial aging.

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concentration, is considerably less in Al-Si-Mg-Sr alloys than in Al-Si-Mg alloys, as demonstrated in Figures 4.19 and 4.20. For example, for the alloy containing 1.60 pct. Mg<sub>2</sub>Si, the total increase in conductivity is ~21 pct. in the absence of strontium whereas it is about 15 pct. for the alloy containing the same level of Mg<sub>2</sub>Si and 0.035 wt. pct. Sr (see Figure 4.20). Clearly, strontium suppresses the precipitation reaction and this effect is especially significant for alloys with a high Mg<sub>2</sub>Si content (>0.95 wt. pct.). A similar aging behaviour was also observed for the alloys with sodium.<sup>75</sup> The reason for this effect of strontium is not fully understood due to the complexity of the precipitation process.<sup>142,143</sup> Finally, it is useful to note that most Al-Si casting alloys contain less than 0.70 wt. pct. Mg (ie. less than 1.1 wt. pct. Mg<sub>2</sub>Si in the fully aged state). Therefore the effect of strontium on the aging process is probably not significant from a practical point of view.

## CHAPTER 5

THE ELECTRICAL RESISTIVITY AND CONDUCTIVITY OF TWO-PHASE AI-SI ALLOYS

5.1 General

When the silicon concentration exceeds 1.30 wt. pct., a eutectic phase begins to appear in the microstructure of Al-Si alloys. It is to be expected that the presence of the eutectic will be reflected in the physical properties of the alloys, including electrical conductivity.

The room temperature conductivity of air cooled alloys covering the range 0.3-12.6 wt. pct silicon is presented in Figure 5 1. As pointed out in the previous chapter the addition of 0.035 wt. pct strontium does not influence the conductivity of the single-phase alloys. However, when an appreciable amount of the eutectic silicon forms, at about 1.8 wt. pct. Si, the conductivities of strontium-free and strontium-containing alloys begin to differ. This difference increases with increasing silicon content and reaches a maximum of about 10 percent at the eutectic composition of 12.6 wt. pct. Si. In this chapter, first the factors which determine the magnitude of these differences in conductivity will be discussed, followed by an explanation of the conductivity differences as they relate to microstructural changes.

5.1.1 Effect of Silicon Concentration

Figure 5.2 shows the change in conductivity of as-cast Al-Si and Al-Si-Sr alloys, cooled at a rate of 0.4°C/sec, both as a function of silicon content and 'volume fraction of the eutectic silicon as calculated from the equilibrium Al-Si phase diagram. For strontium-free alloys containing the unmodified eutectic silicon phase, the electrical conductivity drops from 47.5 pct. to 30 pct. IACS, a 37 pct. decrease, when the silicon content is increased from 2 wt. pct. to 12.6 wt. pct. The conductivity behaviour is similar in strontiumcontaining alloys (modified eutectic silicon) in which case the conductivity decreases from 49.3 to 41.8 pct. IACS, a 15 pct. decrease.



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Figure 5.2 Variation of the room temperature conductivity of Al-Si and Al-Si-Sr alloys with increasing silicon content.

This decrease in conductivity (i.e. increase in resistivity) is due to an increase in the volume fraction of the silicon phase. Silicon has a high resistivity of  $30 \times 10^{11} \,\mu\Omega$ cm compared to that of the aluminum solid solution which is ~ $3.40 \mu\Omega$ cm at 25°C. Therefore, it is obvious that the resistivity/conductivity of the two-phase Al-Si alloys will be greatly influenced by the amount of the silicon phase.

It can also be seen in Figure 5.2 that the conductivity of the modified *A* alloys (AI-Si-Sr) is considerably higher than that of the unmodified alloys (AI-Si) containing the same level of silicon. The different conductivity values for modified and unmodified alloys are found to be the result of the difference in the eutectic silicon structure Figure 5.3 presents a series of optical micrographs of the as-cast AI-Si (unmodified) and AI-Si-Sr (modified) alloys containing 2-12.6 wt. pct Si with approximately 0.03 wt. pct Sr in the modified case. The effect of strontium can be clearly seen in these micrographs. It acts to refine the eutectic silicon and change it from plate-like to fibrous in form.

For alloys which were rapidly cooled at a rate of 1.0°C/sec during casting, a similar conductivity behaviour was observed, as illustrated in Figure 5.4. The conductivity of the unmodified alloys changes from 48 pct. IACS to 40 pct. IACS (i.e. a 17 pct decrease) with increasing silicon content from 2 wt pct. to 12 6 wt. pct The conductivity of the modified alloys follows a similar pattern changing from 50 to 43 pct IACS (i.e. a 16 pct decrease) It is also found that the conductivities of rapidly cooled alloys are higher than those of slowly cooled alloys, especially in the case of the unmodified alloys (see Figure 5.2).

The coarseness of the microstructure is determined by the solidification rate, which is a function of the cooling rate. The slower solidification rate obtained at a cooling rate of 0.4°C/sec results in a coarser eutectic silicon structure whereas the increased solidification rate achieved by chill casting (1°C/sec) leads to a finer silicon structure. This is particularly true for the unmodified alloys where the silicon morphology changes from coarse plate-like to fine lamellar with an increased cooling rate (Figure 5.5 (a)). In the





d) modified (Al-4.32 pct. Si) (46.1 pct. IACS)

## Figure 5.3 continued

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f) modified (AI-6.34 pct. Si) (44.5 pct. IACS)

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Figure 5.3 continued

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h) modified (Al-7.87 pct. Si) (43.2 pct. IACS)

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k) modified (Al-10.18 pct. Si) (42.1 pct. IACS)




Figure 5.3 continued



Figure 5.4 Room temperature electrical conductivity of Al-Si and Al-Si-Sr alloys cooled at a rate of 1.0°C/sec.

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case of the modified alloys the eutectic silicon phase has a fibrous morphology regardless of the cooling rate, but it is noticeably coarser in the slowly cooled alloys (Figure 5.5 (c)). Therefore, the conductivity of the unmodified alloys is more sensitive to changes in the solidification rate than is the case for the modified alloys. In addition to the eutectic phase, the primary aluminum dendrites become finer as the soldification rate increases (see Figure 5.5) and it has already been demonstrated that the finer the dendritic structure the higher the conductivity <sup>125</sup>

### 5 1.2 Effect of Magnesium

The influence of magnesium additions on the room temperature conductivity of chill cast (1°C/sec) Al-Si and Al-Si-Sr alloys is presented in Figure 5.6. It is evident from Figure 5.6 that magnesium lowers the conductivity of both unmodified and modified alloys. For example, the magnesium-free Al-Si alloy has a conductivity of 41 pct IACS at 10 wt pct. Si whereas the conductivity decreases to about 35 pct. IACS and 33 5 pct IACS with the addition of 0.3 wt pct Mg and 0.75 wt pct Mg, respectively. In the case of the modified alloys (Al-Si-Sf), the conductivity declines from 43 5 to 40 pct. IACS when the magnesium concentration is increased up to 0.75 wt pct. These decrements in conductivity are caused by magnesium atoms present in the aluminum solid solution which increase the resistivity of the aluminum matrix.

Figure 5.6 also shows the variation in conductivity with silicon content for both unmodified and modified alloys containing magnesium (0.3-0.75 wt. pct). The conductivity of the unmodified alloys decreases almost linearly with increasing silicon content up to 10 wt. pct. and then deviates slightly from linearity with further silicon addition. This relationship yields a conductivity variation,  $\Delta\sigma$ , of ~1.5 ± 0.5 pct. IACS per wt pct. Si. The conductivity of the modified alloys, however, follows a significantly different pattern in which it drops fairly sharply with an increase in silicon level up to 6 wt. pct. and then remains almost constant thereafter with increasing silicon content. This indicates that the electrical conductivity of the modified alloys, containing more than 6 wt. pct. Si is controlled by the conductivity of the

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Figure 5.6 Room temperature electrical conductivity of two-phase Al-Si-Mg and Al-Si-Mg-Sr alloys cooled at a rate of 1.0°C/sec.

aluminum matrix, which is constant. Hence, the presence of a nonconducting silicon phase does not influence the conductivity of the modified alloys in the same manner as it does the conductivity of the unmodified alloys. The differing conductivity behaviours observed in the modified and unmodified alloys will be discussed in more detail later (section 5.2.2).

Figure 5.7 illustrates the electrical conductivity of the AI-Si-Mg (0.3 wt. pct.) and AI-Si-Mg(0.3 wt. pct.)-Sr alloys cooled at a rate of 0 4°C/sec The conductivity of slowly cooled alloys also decreases with increasing silicon content and increases with strontium additions up to 0 06 wt. pct However, these conductivity increases are significantly less than those measured in AI-Si alloys (see Figure 5.2).

Metallographic examinations reveal that magnesium has a coarsening effect on the unmodified eutectic silicon, as seen in Figure 5.8. Moreover, a lesser degree of modification is observed in slowly cooled AI-Si-Mg-Sr alloys. This is illustrated in Figure 5.9 where the form of the eutectic silicon is not altered significantly during modification (i.e. the original plate-like structure remains even after a 0.03 wt. pct strontium addition, as shown in Figure 5.9 (a) and (b)). Although the increased strontium content (0.06 wt pct.) leads to a 13 pct. difference in conductivity at 10 pct. Si, it also fails to produce a fully modified structure (Figure 5.9 (c)).

In contrast, in samples prepared from a commercial Al-Si-Mg alloy (e.g. A356 containing 7 wt. pct. Si and ~0.35 wt. pct. Mg), a fully modified eutectic silicon structure was obtained with the addition of as little as ~0.02 wt pct. Sr at a cooling rate of 0.5°C/sec. In this case, a-resistivity decrease of about 10 pct. upon modification was measured by Closset<sup>22</sup> and Argo et al.<sup>125</sup> Thermal analysis studies suggest that this discrepancy results from differences in the nucleation behaviour of the silicon phase between the pure alloys (i.e. prepared using pure materials) used in this work and the commercial alloys. The thermal analysis curves obtained for magnesium-free pure Al-10% Si and Al-10% Si-0.35% Mg alloys are shown in Figure 5.10. The critical part of the curves is the undercooling prior to the eutectic phase

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Figure 5.7 Room temperature electrical conductivity of two-phase AI-Si-Mg and AI-Si-Mg-Sr alloys cooled at a rate of 0.4°C/sec.



a) n<mark>o M</mark>g



b) 0.75 pct. Mg

•Figure 5.8 Effect of magnesium addition of the microstructure of Al-10 pct. Si alloys cooled at a rate of 1.0°C/sec.





Figure 5.9 Change in microstructure of Al-10 pct. Si-0.30 pct. Mg with addition of strontium.

nucleation ( $\Delta T_N$ ). It has been established that a greater  $\Delta T_N$  produces a finer eutectic and enhances the modification process during the growth stage of the nucleated silicon crystals.<sup>20-22</sup>  $\Delta T_N$  is 2.6°C for the binary Al-Si alloys, corresponding to a fine lamellar structure, as shown in Figure 5.10 (a). However,  $\Delta T_N$  diminishes to 0.5°C with the addition of 0.35 wt pct. Mg This indicates that the presence of magnesium enhances nucleation of the eutectic silicon phase thereby leading to a coarser silicon structure, as presented in Figure 5.10 (b). The addition of 0.03 wt. pct. Sr leads to a larger amount of undercooling ( $\Delta T_N = 2^{\circ}C$ ) and results in an under-modified (i.e. coarse lamellar silicon) eutectic structure (Figure 5.10 (c)). For the commercial A356 alloy, it has been reported that the addition of 0.02 wt. pct.-Sr increases  $\Delta T_N$  to 3.5°C, which corresponds to a change in the form of the eutectic silicon from acicular to fibrous.<sup>22</sup> The different nucleation kinetics observed in the commercial alloys may originate from the interaction of magnesium with other minor elements present in the alloy, with the cummulative effect of modification being significant. The modification mechanism is outside the scope of this thesis and therefore the effect of magnesium on modification will not be discussed further.

5.1.3 Effect of Modification

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As pointed out earlier (sections 5.1.1 and 5.1.2), the as-cast conductivity of modified alloys is always higher than unmodified alloys containing the same level of silicon and magnesium, this being attributed to the difference in the eutectic silicon structure. The difference in conductivity can be expressed by the quantity  $\Delta \sigma = (\sigma_{mo} - \sigma_{um}/\sigma_{mo}) \times 100$ , where  $\sigma_{mo}$  and  $\sigma_{um}$ are the conductivities of the modified and unmodified alloys, respectively. The measured conductivity difference,  $\Delta \sigma$ , is plotted against silicon content in Figure 5.11. For both the Al-Si and Al-Si-Mg systems,  $\Delta \sigma$  increases with increasing silicon content. The larger difference between the conductivities of the modified and unmodified alloys at higher silicon levels is due to the proportionally larger volume of the eutectic silicon present and to the microstructural changes which occur on modification



b) pure Al-Si-Mg

Figure 5.10 Thermal analysis curves near the eutectic transformation and the corresponding microstructures.

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Figure 5.11 Measured conductivity difference between the modified and unmodified AI-Si and AI-Si-Mg alloys.

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Figure 5.11 shows that the difference in the conductivity ( $\Delta \sigma$ ) of Al-Si and Al-Si-Sr alloys decreases with increasing cooling rate. At the eutectic composition, where the maximum difference was detected,  $\Delta \sigma$  is about 28 pct. and 10 pct. for the alloys cooled at a rate of 0.4°C/sec and 1°C/sec, respectively. This is apparently due to the lesser amount of microstructural change that occurs in the chill cast alloys, i.e. the eutectic silicon structure is transformed from fine lamellar to fibrous rather than from acicular to fibrous (see Figures 5.3 and 5.5).

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The addition of magnesium increases the measured difference in the conductivities of the unmodified and modified alloys cooled at a rate of 1°C/sec, as is shown in Figure 5.11. At the eutectic composition (12.6 wt. pct Si), the conductivity difference is of the order of 10 pct. for magnesium-free alloys and 23 pct. for the alloys containing 0.30 wt. pct. Mg. This greater difference in conductivity is due to the fact that magnesium coarsens the unmodified eutectic silicon (Figure 5.8). Hence, a relatively greater structural change occurs during strontium modification which results in a larger difference.

Figure 5.11 also illustrates that, in the case of slowly cooled (0.4°C/sec) Al-Si-Mg alloys, the differences are lower than in Al-Si and chill cast Al-Si-Mg alloys. For instance,  $\Delta \sigma$  is only 8 pct. at the eutectic composition, which is substantially lower than that of the magnesium-free alloys ( $\Delta \sigma = 30$  pct.) This can be attributed to the lesser degree of modification observed in the Al-Si-Mg-Sr alloys (see Figure 5.9)

Figure 5.12 summarizes the effect of the degree of modification on the measured conductivity difference ( $\Delta\sigma$ ), and presents corresponding microstructures. Based on the information in Figure 5.12 foundrymen can predict the degree of modification achieved in their castings by comparing the measured  $\Delta\sigma$  values with the values shown in Figure 5.12.



Figure 5 12 Effect of the degree of modification on the measured conductivity difference. Magnification 100 times.

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#### 5.2 Interpretation of the Conductivity Differences

#### 5.2.1 Quantitative Evaluation

The two phase AI-Si system consists of semiconducting silicon embedded in an aluminum matrix whose resistivity ranges from 2.7 to  $4.0\mu\Omega$ cm depending on the concentration of dissolved silicon and magnesium in the solid solution. Therefore, a two-phase AI-Si alloy can be considered to consist of a continuous conducting matrix and effectively nonconducting eutectic silicon plates or fibers (i.e.  $p_{SI} \gg p_{AI}$ ). In the technical literature eutectic alloys such as AI-Si are commonly considered to be in-situ composites whose physical properties can be predicted by means of theoretical techniques.<sup>144</sup> The physical properties, including electrical conductivity, of a composite can be written in terms of the properties of both phases and their volume fractions. If it is assumed that the second phase particles (i.e. the eutectic silicon phase) are randomly orientated in a continuous matrix the electrical conductivity of the composite can be expressed using Rayleigh's mixture formula<sup>145</sup>

$$o_{co} = o_m \frac{2o_m + o_s + 2V_s (o_s - o_m)}{2o_m + o_s - V_s (o_s - o_m)}$$
(5.1)

where  $\sigma_{co} = \text{conductivity of the composite}$   $\sigma_m = \text{conductivity of the matrix}$   $\sigma_s = \text{conductivity of the second phase}$  $V_s \neq \text{volume fraction of the second phase}$ 

In the case of the Al-Si system,  $\sigma_{S_1} \ll \sigma_{Al}$  and equation 5.1 can be rewritten as;

$$p_{Al-S_{l}} = \sigma_{Al} \frac{2\sigma_{Al} - 2V_{S_{l}}\sigma_{Al}}{2\sigma_{Al} + V_{S_{l}}\sigma_{Al}}$$
(5.2)

where  $\sigma_{AI-SI} = \text{conductivity of two-phase AI-SI alloy}$   $\sigma_{AI} = \text{conductivity of aluminum solid solution}$  $V_{SI} = \text{volume fraction of eutectic silicon}$ 

It must be noted that equations 5.1 and 5.2 are only applicable for randomly dispersed spherical particles in a continuous matrix. For rods and fibers the

relation is given as138

$$\sigma_{co} = \sigma_m + \frac{V_f (\sigma_f - \sigma_m)(5\sigma_m + \sigma_f)}{3(\sigma_m + \sigma_f)}$$
(5.3)

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where  $\sigma_f = \text{conductivity of the fibers}$ V<sub>f</sub> = volume fraction of the fibers

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For the Al-Si system, equation 5.3 takes the following form;

$$\sigma_{Al-Sl} = \sigma_{Al} - \frac{5}{3} V_{Sl} \sigma_{Al}$$
(5.4)

since  $\sigma_f = \sigma_{Si} \approx 0$ 

Equations 5.2 and 5.4 are compared with the experimental conductivity results obtained for chill cast modified and unmodified Al-Si alloys in Figure 5.13. It is found that both equations show only fair agreement with measured values. This is not surprising since the equations are valid only if the phases in a composite have idealized structural arrangements and the effect of interfaces between phases is neglected. Clearly for many binary systems, including the Al-Si system, these assumptions are over-simplified. Additionally, it is widely accepted that setting upper and lower limits for the electrical properties of a composite material is preferrable to calculating a particular average value.<sup>144</sup> The limits<sup>144</sup> derived for the electrical conductivity of a two-phase isotropic composite are given as;

 $\sigma(+) = \sigma_m (1 - V_f) + \sigma_f V_{f'}$ 

$$\sigma(-) = \frac{1}{\frac{(1-V_f)}{\sigma_r} + \frac{V_f}{\sigma_f}}$$

(5.5b)

(5.5a)

where  $\sigma(+) =$  upper limit for conductivity of the composite  $\sigma(-) =$  lower limit for conductivity of the composite



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For the Al-Si system  $\sigma_{S_1} \ll \sigma_{A_1}$  and hence equation 5.5 yields

$$\sigma(+) = \sigma_{Al} (1 - V_{Sl}) \quad \text{and} \quad \sigma(-) = \frac{\sigma_{Sl}}{V_{Sl}}$$
(5.6)

Since the values of  $\sigma_{AI}$  and  $\sigma_{Si}$  differ greatly, the limits are widely separated, as shown in Figure 5.14 in which it can also be seen that the experimental results fall within the limits. Closer limits may be obtained for electrical conductivity if some structural factors are introduced into equations 5.5 and 5.6. However, the calculation of a structural factor is only possible for simple structural arrangements such as rods or lamellae aligned in one direction. Undoubtedly, the complexity of the structure of the eutectic silicon phase and the difficulties of estimating the Al-Si interface effect will lead to some uncertain approximations in quantitatively describing the electrical conductivity differences between the unmodified and modified Al-Si alloys. Therefore, the conductivity differences will be explained qualitatively in the following sections based on the different structural features of the modified and unmodified eutectic silicon phases.

#### 5.2.2 Qualitative Interpretation

The parameters which determine the electrical conductivity of twophase AI-Si alloys are as follows:

- 1) The electrical conductivities of the Al-solid solution and the eutectic silicon phase,
- 2) The volume fractions of the Al-solid solution and the eutectic silicon phase,
- 3) The structural arrangement of the eutectic silicon phase,
- 4) The nature of the interface between the silicon phase and the Almatrix.





### 5.2.2.1 Electrical Conductivities

It is obvious that electrons will only flow through the aluminum matrix since it has a much higher conductivity than the silicon phase. Therefore, any change in the conductivity of the Al-matrix will have a profound effect on the conductivity of the two-phase Al-Si alloys. As has been established in the previous chapter (see section 4.3), the addition of 0.03 wt. pct. Sr does not alter the conductivity of the aluminum matrix. This is also true when magnesium (up to 1 wt. pct.) is present in the aluminum matrix.

It is not known whether or not strontium affects the conductivity of the. silicon<sup>7</sup>phase or is even soluble in silicon, but it is very unlikely that strontium causes an increase in its conductivity at room temperature up to values comparable to that of the Al-matrix. Therefore, any change in the silicon conductivity, due to the presence of strontium, is probably negligible.

## 5.2.2.2 Volume Fractions

It has been shown that the eutectic temperature (577°C) is depressed with the modification of Al-Si alloys by 5 to 10°C, depending on the cooling rate and the amount of modifier added. The result of this is a displacement of the eutectic composition to higher silicon levels and a consequential increase in the volume fraction of the Al-solid solution in conjunction with a decrease in the volume fraction of the eutectic silicon. The eutectic region of the Al-Si phase diagram has been recently reconstructed by Hanna and Hellawel<sup>37</sup>, as shown in Figure 5.15 in which the eutectic composition is seen to increase from 12.6 wt. pct. Si to 13.6 wt. pct. Si upon modification. The decrease in the volume fraction of the eutectic silicon phase of the modified eutectic alloy, as calculated using the phase diagram shown in Figure 5.15, is almost 1 pct. This volume decrement will lead to a minor increase of about 1.5 pct. IACS (see Figure 5.2) in the conductivity of the eutectic alloy. However, this is insufficient to explain the observed increase of 30 pct. in conductivity on modification.



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5.2.2.3 Structural Arrangement

Al-Si casting alloys display anomalous eutectic structures containing silicon plates in unmodified alloys and silicon fibers in modified alloys. The regular array of eutectic phases in certain directions observed in normal eutectic alloys does not occur in these anomalous eutectics. Instead, more branching of the faceted phase (Si phase) and a more random orientation takes place during eutectic freezing.<sup>28</sup> Optical micrographs and X-ray diffraction photos indicate a random crystal orientation, of the silicon and aluminum eutectic phases in the modified and unmodified alloys under normal solidification conditions such as sand and chill casting.<sup>30</sup> Accordingly, it is expected that cast Al-Si alloys will exhibit an isotropic electrical conductivity behaviour

In addition, any slightly anisotropic difference of the unmodified and modified eutectic silicon phase will not be reflected in measured conductivity values using the technique employed in this work. This is because the eddy current induced in the sample by the conductivity meter circulates through a given volume instead of following one particular direction through the sample. Hence the measured value is an overall average conductivity of the sample. Nevertheless, a highly oriented structure in the eutectic silicon phase produced by employing techniques such as directional solidification will result in anisotropic electrical resistivity behaviour. This will be considered in the next section.

5.2.2.4 Interfacial Effect

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It is well known that the interface between phases can contribute significantly to the electrical resistivity of a two-phase alloy due to the scattering of electrons at the interface.<sup>44,56,80,116</sup> The rate of scattering (i.e. scattering probability) is proportional to the cross-section of the interface and the interface density. Obviously, as the rate of electron scattering increases the resistivity of an alloy increases and the conductivity decreases.

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In the case of two-phase Al-Si alloys, the electron mean free path (I) is controlled mainly by the electron scattering caused by alloying elements present in the Al-solid solution at room temperature. In the vicinity of the eutectic areas, however, the mean free path will be reduced by additional electron scattering at the interfaces between the aluminum matrix and the eutectic silicon phase. It is assumed that different interfacial features, related to the various microstructural forms of the eutectic silicon, will lead to different rates of electron scattering resulting in different conductivity values. Here, a brief listing of the microstructural differences between the unmodified (plate-like) and modified (fibrous) silicon phases is appropriate.

- In the unmodified alloy, the silicon plates grow ahead of the eutectic-liquid interface. In the modified alloy, a coupled growth of the silicon and aluminum phases occurs and the silicon fibers are enveloped by the aluminum phase.
- 2) The plate-like silicon phase exhibits a two-dimensional continuity, but the fibrous silicon phase is essentially one-dimensional.
- 3) The silicon fibers are smaller in size than the plates. Therefore, the individual fibers do not extend over large areas in any one plane as do the silicon plates.

These structural differences in modified and unmodified alloys can be clearly seen in SEM micrographs (Figure 5 16) taken from deep etched samples.

Further evidence of these structural differences is the substantial improvement in mechanical properties after modification of the Al-Si alloys. In unmodified alloys cracks follow the continuous silicon plates whereas in modified alloys cracks propagate much shorter distances along the silicon fibers and mostly through the plastic Al-matrix which slows the crack propagation rate, hence improving the mechanical properties.<sup>28,30</sup>

The increased electrical conductivity after modification can be explained if the effect of the structural changes listed above on the rate of electron scattering from the eutectic silicon-aluminum matrix interfaces are considered. As shown schematically in Figure 5.17 the continuous silicon

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a) unmodified



b) modified

Figure 5-16 SEM micrographs of deep etched Al-Si alloys.





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plates act to impede electron flow and scatter the electrons effectively, whereas the silicon fibers allow easier electron flow due to the reduced projected interfacial area between the Si-fibers and the Al-matrix. Also, the fiber-matrix interfaces are not as continuous as the plate-matrix interfaces. Therefore, the amount of electron scattering from the fiber-matrix interfaces will be less than from the plate-matrix interfaces. This results in a subsequent increase in conductivity.

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Further support of the hypothesis that the eutectic silicon-aluminum matrix interfaces greatly influence the conductivity of unmodified and modified alloys is obtained from the resistivity measurements taken at 77K. Table 5.1 gives the resistivity values of the alloys measured at room temperature and 77K, together with the difference in resistivity of the modified and unmodified eutectic alloys. The resistivity of both the modified and unmodified alloys decreases by a factor of about 6 to 8 as temperature decreases from 297K to 77K while the difference in resistivity values increases by 17 pct. from 30 pct. to about 47 pct. The reduced resistivity values at 77K are due to a significant decrease in the temperature 'dependent resistivity (pth) and a subsequent increase in the mean free path of the electrons being scattered by the eutectic silicon-aluminum matrix interfaces. Therefore, the influences of interfacial scattering and differing interfacial features on resistivity become more pronounced at 77K and the resistivity difference between the unmodified and modified alloys is amplified. These results illustrate convincingly the importance of the siliconaluminum interfaces in determining the electrical resistivity of two-phase Al-Si alloys.

A different conductivity behaviour is observed in the Al-Si alloys containing magnesium. In this case, the conductivity of the unmodified alloys decreases almost linearly with increasing silicon content while that of the modified alloys shows less dependence on the silicon concentration and remains essentially, unchanged at silicon levels in excess of 6 wt. pct. (see Figure 5.6). By assuming that no further silicon dissolution occurs in the aluminum matrix at silicon concentrations greater than 1.62 wt. pct., the density of the eutectic silicon-aluminum matrix interface increases as the

Temperature (K)	Pavg,unmod (μΩ cm) (12.6%Si - 0%Sr)	Pavg, mod (μΩ cm) (12.6%Si - 0.03%Sr)	Δρ <sub>avg</sub> (%)
297	6.02	4.23	29.7
77	1.01	0.54	46.6

Table 5.1 Resistivity Values of Modified and Unmodified Al-Si Eutectic Alloys

 $\Delta \rho_{avg} = (\rho_{avg,unmod} - \rho_{avg,mod.}) / (\rho_{avg,unmod}) \times 100\%$ 

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silicon content increases. As shown above, the interface scattering contributes to the resistivity of the unmodified alloys. Therefore, the reduction in the conductivity of the unmodified alloys caused by increasing silicon content is a function of the interfacial effect. For the case of the modified alloys the conductivity behaviour is rather different. The conductivity of the modified alloy is more sensitive to the changes in the conductivity of the Al-matrix due to the presence of aluminum short 'circuiting paths carrying the current. It is known that the mean free path becomes very short as the concentration of alloying elements retained in the Al-solid solution increases. For example, the mean free path is much shorter in chill cast Al-Si-Mg-Sr alloys than in slowly cooled Al-Si-Sr alloys because of (a) the presence of magnesium atoms and (b) the faster cooling rate which retains more silicon and magnesium in solid solution. As the mean free path is shortened the amount of electron scattering from the eutectic silicon-Al matrix interfaces will be smaller. In other words, the interfacial scattering becomes insignificant in comparison to alloying scattering which will govern the changes in the conductivity of the modified alloys with increased silicon contents

### 5.3 Directional Solidification Experiments

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### 5.3 1 Eutectic Silicon Structures.

Various structural transitions in the eutectic silicon phase have been obtained in binary Al-Si eutectic alloys by directional solidification at different growth rates ( $R_g$ ). Optical micrographs taken from both longitudinal and transverse sections of the directionally solidified alloys are illustrated in Figure 5.18. In a sample frozen at a rate of 0.2cm/h the silicon phase grows as rods with noticeable evidence of side branching and appears angular on the transverse section (Figure 5.18 (a) and (b)). As the growth rate increases more side branching occurs and the spacing between the silicon rods decreases (Figure 5.18 (c)). The side plates incline to a variety of angles on the transverse section (Figure 5.18 (d)). A further increase in the growth rate (>1cm/h) leads to the formation of silicon flakes which are

interconnected, and some "Y" shape silicon crystals also appear on the transverse section (Figure 5.18 (f)).

At the slow growth rates, typically 0.2-1.0cm/h, there is an alignment of the silicon rods with the growth direction. It has been found that these rods display a strong [100] texture.<sup>28,30</sup> At a growth rate of 5cm/h the silicon flakes still show a preferential orientation with growth axis. This preferential orientation is lost in the sand cast alloy and a random distribution of the silicon flakes occurs, as shown in Figure 5.18 (g) and (h) X-ray diffraction was used to characterize the anisotropy (i.e. alignment) of the silicon phase. Since the sand cast sample was assumed to be random, the diffraction intensity of the (220) silicon line (the strongest silicon peak) 1/2obtained from this sample was used to give the anisotropy of the directionally solidified samples using the ratio  $(I_s^{(220)}/I_d^{(220)})$ , where  $I_s$  and  $I_d$ are the intensities of the (220) silicon peaks from the sand cast and directionally solidified samples, respectively. The X-ray diffraction patterns obtained from the directionally solidified samples are given in Appendix II. As can be seen in Table 5.2, in which the results of the X-ray diffraction studies are summarized, the anisotropy of the silicon phase becomes less pronounced as the growth rate is increased

The microstructures obtained in this work by directional freezing confirm earlier reports<sup>28,30</sup> that by increasing the growth rate the structure of the eutectic silicon phase can be made to change gradually from well aligned rods to non-aligned silicon flakes. As expected, these structural changes in the silicon phase are reflected in the electrical resistivity behaviour.

5.3.2 Electrical Resistivity of Directionally Solidified Alloys

Table 5.3 presents the resistivity values measured parallel and perpendicular to the freezing direction on samples with differing growth rates. With an increase in the growth rate from 0.2 to 5cm/h the longitudinal resistivity - the resistivity parallel to the silicon rods and flakes - increases Ø

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a) longitudinal section (growth rate 0.2 cm/h)



Figure 5.18 Microstructures of directionally solidified Al-Si eutectic alloys.

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c) longitudinal section (growth rate 1.0 cm/h)



d) transverse section (growth rate 1.0 cm/h)

## Figure 5.18 continued



# Figure 5.18 continued



h) transverse section (sand cast)

# Figure 5.18 continued

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Table 5.2 Variation of the Anisotropy of the Silicon Phase With Growth Rate				
Growth Rate <sup>∞</sup> ∫ (cm h − 1)	Is(2 <del>20) / I</del> d(220)			
0.2	0.14			
1.0	0.25			
5.0	0.70			
sand cast (unmodified)	1			

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Table 5.3 Resistivity Values of Directionally Solidified and Sand Cast Al-Si Eutectic Alloys

Growth Rate (cm h - 1)	Longitudinal Resistivity, ρ <sub>L</sub> (μΩ cm)	Transverse Resistivity, ρτ (μΩ cm)	Percent Dıfference (ρτ – ρ <sub>L</sub> ) / (ρ <sub>T</sub> ) x 100
<i>.</i>	3.93	9.75	<sup>,</sup> 60
1.0	4.39	6.03	27
5.0	4.51	5.00	10
sand cast (unmodifed)	5.96	6.15	<u>∉</u> 3
sand cast (modified)	4.08	4.21	'3

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from 3.93 to  $^{\$}4.51\mu\Omega$ cm, while the transverse resistivity - the resistivity perpendicular to the silicon rods and flakes - drops from 9.75 to 5.0  $\mu\Omega$ cm.

The longitudinal resistivity ( $p_L$ ) of an Al-Si eutectic alloy can be estimated by considering the aluminum and silicon phases as being two resistors connected in parallel and the<sup>4</sup> current as flowing parallel to the silicon rods. Thus, the longitudinal resistivity is given as;

$$\frac{1}{L} = \frac{V_{S_l}}{\rho_{S_l}} + \frac{(1 - V_{S_l})}{\rho_{Al}}$$
(5.7)

And since psi ≥ pAI

$$\frac{1}{\rho_L} \simeq \frac{(1 - V_{S_l})}{\rho_{Al}}$$

(5.8)

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Note that equation 5.8 is identical to equation 5.5a. At the eutectic composition  $V_{S1} = 0.12$  and  $p_{A1} = 3.40 \ \mu\Omega$  cm, which is the resistivity of an air cooled sample containing 1.6 wt. pct. silicon (see Figure 4.3) Equation 5.8, then, gives  $p_L = 3.6\mu\Omega$  cm which is the lower limit value of the longitudinal resistivity ( $p_L$ ) of Al-Si eutectic alloys. All measured values are higher than this limiting value, but the resistivity exhibited by the sample solidified at a rate of 0.2 cm/h approaches this limiting value because the sample displays well aligned silicon rods with only slight branching, as seen in Figure 5.18 (a) More misalignment and branching occur in the samples solidified at faster growth rates ( $\frac{1}{2}$  1 cm/h) and this results in a higher resistivity, and hence, more deviation from the limiting values, as presented in Table 5.3.

Similarly, for the current flowing normal to the direction of the silicon rods, the eutectic structure can be represented by two resistors in series. The equivalent resistivity (i.e. transverse resistivity) will then be expressed as,

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$$\rho_T = V_{S_l} \rho_{S_l} + (1 - V_{S_l}) \sigma_{Al}$$
(5.9)

Since psi ≥ pAI;

$$\rho_T \approx V_{Sl} \rho_{Sl} \tag{5.10}$$

where  $\rho_{Si}$  is the resistivity of silicon (~30x10<sup>11</sup> µΩcm). For the eutectic alloy (V<sub>Si</sub> = 0.12), the calculated  $\rho_T$  from equation 5.10 is about 30x10<sup>10</sup> µΩcm, which is the upper limit of the transverse electrical resistivity. The measured  $\rho_T$  values are well below this value due to the aluminum short circuiting paths where electron conduction occurs. As the anisotropy of the silicon plates disappears with increasing growth rate,  $\rho_T$  decreases because the current does not flow only in the direction perpendicular to the surface of the silicon plates but also in directions more parallel to the silicon plates.

The percent difference in resistivity ( $p_T - p_L/p_T \times 100$ ) is plotted versus a measure of the silicon anisotropy ( $I_s^{(220)}/I_d^{(220)}$ ), using X-ray diffraction data, in Figure 5.19. A clear relationship exists between the anisotropy of the silicon structure and resistivity. In highly anisotropic structures ( $I_s/I_d < 0.25$ ), such as occur at low solidification rates (e.g. 0.2cm/h), the continuous aluminum matrix provides a low resistivity measured along the growth direction while the continuous silicon phase interrupts the current flow perpendicular to the growth direction resulting in a relatively high resistivity. As the microstructure becomes more irregular and isotropic ( $I_d/I_s > 0.75$ ) with increasing freezing rate, these effects become less important and the two resistivity values approach each other. The 3 pct. difference in resistivity measured in the sand cast sample can be ascribed to experimental error and perhaps to some slight anisotropy created during solidification.

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CHAPTER 6

SOLUTION TREATMENT OF TWO-PHASE AI-SI AND AI-SI-Mg ALLOYS

6.1 General

A great majority of cast Al-Si alloy parts are heat treated prior to service in order to improve their mechanical properties. Generally, the heat treatment schedules involve a solution treatment (T4) just below the eutectic temperature and then an aging treatment in the temperature range of 150-230°C for 12h.<sup>53</sup> The solution treatment provides a more homogeneous aluminum matrix in which precipitates form during the subsequent aging treatment. It has been shown that the solution treatment time and temperature are two critical parameters that must be controlled closely to achieve the optimum age hardening and subsequent mechanical properties.<sup>53</sup> The precipitation process that occurs in Al-Si and Al-Si-Mg alloys has already been discussed in chapter 4 (sections 4.3 and 4.4). Therefore, the solution treatment of these two-phase alloys will be the primary subject of this chapter.

During solution treatment of Al-Si and Al-Si-Mg alloys, not only homogenization of the cast structure occurs but also a substantial degree of spheroidization and coarsening of the eutectic silicon phase takes place. In the following sections the kinetics of the solution treatment process will be discussed based on electrical conductivity measurements and metallographic examinations.

6.2 Results

Changes in electrical conductivity, as a function of solution treatment time, will be given initially while the microstructural changes will be considered later. b

## 6.2.1 Electrical Conductivity of Al-Si Alloys

The electrical conductivities of unmodified and modified alloys containing varying amounts of silicon are plotted against solution treatment time in Figure 6.1. The remaining data is to be found in Appendix III. A general feature of the curves in Figure 6.1 is that the conductivity of both the unmodified and modified alloys first increases very rapidly in times of the order of 1.5 hours. It then decreases sharply over a period of 3h, followed by a gradual decrease until a solution treatment time of 8h has been reached. With further solution treatment the conductivity of the unmodified alloy shows a slight increase, whereas that of the modified alloy remains virtually unchanged. An interesting observation is that the initial conductivity differences between the as-cast unmodified and modified alloys tend to diminish after 5h of solution treatment.

Figure 6.2 presents the percent conductivity changes versus solution treatment time for the Al-Si and the Al-Si-Sr alloys containing 2 wt. pct Si and 8 wt. pct. Si. The conductivity behaviour of the alloys is very similar to that observed in Figure  $\hat{\mathbf{b}}$ . 1, with the exception that the differences between the conductivities of the unmodified and the modified alloys are greatly amplified. At the 8 wt. pct. Si level, the conductivity of the unmodified alloys increases by approximately 23 pct. after 1.5h solution treatment while that of the modified alloy increases by 18 pct. After 8h of solution treatment the percent conductivity change for the unmodified alloy drops by 14 pct., reaching a minimum of 9 pct. which corresponds to 44.0 pct. IACS. During the same period of time the change in conductivity of the modified alloy exhibits a decrease of 18 pct., reaching its original as-cast conductivity of 45 8 pct. IACS. After complete solution treatment (16h) the conductivity of the unmodified alloy is 10.5 pct. higher than its original as-cast value. On the other hand, the conductivity of the modified alloy is only 1.5 pct. higher than its as-cast conductivity.

For the alloys containing 2 wt. pct. Si, the initial increases in conductivity are 16 pct. and 12 pct. in the unmodified and the modified alloys, respectively. Later, the conductivities of the alloys drop well below

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Figure 6.1 Variation of the electrical conductivity of Al-Si alloys during solution treatment at 540°C.

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Figure 6.2 Relative electrical conductivity changes in Al-Si alloys during solution treatment at 540°C.

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their initial values, and after complete heat treatment the conductivity of the unmodified alloy is close to its as-cast value and the conductivity of the modified alloy is 4 pct. less than its as-cast value.

For alloys which were slowly cooled during casting, at a rate of 0.4°C/sec, a somewhat different conductivity behaviour was observed, as illustrated in Figure 6.3. During the first hour of solution treatment the conductivity of the unmodified alloy (Al-Si 10 wt. pct.) changes from 35.2 pct. IACS to 39.7 pct. IACS (i.e. a 12 pct. increase). However, the conductivity of the modified alloy varies from 44.1 pct. to 46.7 pct. IACS (i.e. a 6 pct. increase). The conductivity of the unmodified alloy decreases from 39.7 pct. to 39.2 pct. IACS during the next 3h of solution treatment, and then increases again to 41.5 pct. IACS at the end of the 16h treatment. The conductivity of the modified alloy, however, continues to decrease until a treatment time of 8h, after which it stays constant as 42.4 pct. IACS.

## 6.2.2 Conductivity of Two-Phase Al-Si-Mg Alloys

Figures 6.4 and 6.5 present the conductivity versus solution treatment time curves for the strontium-free and strontium-containing Al-Si-Mg alloys cooled during solidification at a rate of 1°C/sec (i.e. chill cast). As illustrated in these figures, the conductivity curves exhibit features very similar to those shown in Figure 6.1. For example, in the unmodified alloy containing 0.30 wt. pct. Mg and 10 wt. pct. Si (Figure 6.4) the conductivity increases from 34 pct. IACS to 40 pct. IACS after 1.5h of treatment at 540°C. The conductivity then decreases to 37.5 pct. IACS after 3h solution treatment and finally reaches 42.2 pct. IACS after 16h of solution treatment. The final conductivity is about 25 pct. higher than the as-cast conductivity. In the case of the modified alloy, the initial increase in conductivity is from 40 pct. IACS after 5h of treatment. During the remaining time of the solution treatment the conductivity remains unchanged at 42.7 pct. IACS, a value 5 pct. higher than its as-cast value.

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Figure 6.4 Variation of the electrical conductivity of Al-Si-Mg alloys containing 0.30 wt. pct. Mg during solution treatment at 540°C.



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Figure 6.5 Variation of the electrical conductivity of Al-Si-Mg alloys containing 0.75 wt. pct. Mg during solution treatment at 540°C.

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In addition, the percent conductivity changes,  $\Delta \sigma_R$ , occurring as a result of the complete solution treatment of the alloys containing different levels of silicon and magnesium, are summarized in Table 6.1. The data in Table 6.1 indicates that the conductivity of the unmodified alloys shows a greater increase than that of the modified alloys after solution treatment. Consequently, the original difference between as-cast conductivities of the unmodified and the modified alloys decreases, as is evident in Figures 6.4 and 6.5.

6.2.3 Metallographic Observations

As mentioned earlier, the eutectic silicon crystals spheroidize during solution treatment. The kinetics of this process have also been followed by metallographic examination. Figure 6.6 (a) and (b) are typical as-cast structures of the unmodified and modified alloys containing 8 wt. pct. Si. Figures 6.7 and 6.8 are, respectively, the micrographs of the same samples after 1h, 4h, 8h and 16h of solution treatment at 540°C. As is clearly seen in the figures, solution treatment causes a substantial degree of spheroidization and coarsening in both unmodified and modified eutectic silicon phases.

In the unmodified sample the silicon flakes thicken slightly and shorten and the tips of the flakes become blunt after 1h solution treatment. After 4h treatment a significant spheroidization of the silicon particles occurs. The degree of spheroidization and coarsening of the silicon phase increases with longer solution treatment times. In the modified alloy, however, complete spheroidization of the silicon phase takes place after only 4h of areatment, and the silicon particles coarsen, but do not change shape with further treatment.

When the cooling rate during freezing is reduced from 1°C/sec to 0.4°C/sec, the as-cast structure of the unmodified and the modified/alloys will be coarser, as shown in Figure 6.9. The spheroidization of these coarse silicon flakes is less noticeable in the early stages of solution treatment. Even after 8 hours the unmodified silicon particles, although to a large extent

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Alloy *	Percent Conductivity Change ( $\Delta \sigma_R$ ) at Solution Treatment Time, h:								
	12	1	1 <del>1</del>	2	3-	<b>5</b>	8	12	·16
Al-8%Si-0.3%Mg	13.6	18	21.4	17.3	10.5	12.3	13.3	ີ 13.6	15.0
Al-8%Si-0.3%Mg- _0.03%Sr	11.0	12.7	14.3	11.5	4.5	2.1	3.3	3.0	3.0
Al/10%Si-0.3%Mg	9.9	13.7	.15.1	13.0	7.6	11.9	14:5	17.4	18.3
Al-10%Si-0.3%Mg- 0.03%Sr	9.7	12.4	14.1	12.4	7.5	4.4	<b>4.5</b>	4.8	<sup>`</sup> 5.0
Al-8%Si-0.75%Mg	,10.2	15.8	17.0	12.5	6.3	8.8	7.1	12.7	13.1
Al-8%Si-0,75%Mg- 0:03%Sr .	9.7	15.6	17.8	13.1	5.3	<sub>5</sub> 4.8	2.0	3.5	3.5 <sub>.</sub>
Ál-10%Si-0.75%Mg	8.8	12,2	17.8	12.1	7.8	12.5	12.5	17.4	18.4
A <b>#</b> 10% Si- 0.75% Mg-0.03% Sr	9.2	12.8	16.5	12.3	5.0	6.0	5.0	5.0	5.0

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Table 6.1 Percent Conductivity Changes in Al-Si-Mg Alloys<sup>†</sup> During Solution Treatment

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t cooled at a rate of 1.0°C/sec during solidification.

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## Figure 6.8 continued

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spheroidized, still exhibit evidence of a plate-like structure (Figure 6.10 (a)) while the modified silicon is completely spheroidized (Figure 6.10 (b)).

Additionally, SEM micrographs, Figure 6.11, give a three-dimensional representation of the changes in the structure of the unmodified and the modified eutectic silicon. The unmodified silicon phase, which originally consisted of interconnected plates, gradually shortens and spheroidizes during the solution treatment, Figures 6.11 (c) and (e). On the other hand, the modified silicon phase, that had a finer fibrous texture in the as-cast state, quickly spheroidizes and coarsens, as shown in Figures 6.11 (d) and (f). After 8h of treatment the unmodified silicon plates are interconnected only at a relatively few points whereas the modified silicon particles appear to be more isolated.

6.3 Discussion

6.3.1 Kinetics of the Solution Treatment Process

The following metallurgical processes occur during the solution treatment of two-phase Al-Si and Al-Si-Mg alloys:

1) Precipitation of silicon and Mg<sub>2</sub>Si from the aluminum matrix,

2) Dissolution of these precipitates and other soluble second phases,

3) Spheroidization and coarsening of the eutectic silicon phase.

All of these processes may take place simultaneously and influence the physical properties of the alloys, such as electrical conductivity, to differing degrees.

6.3.1.1 Precipitation of Silicon and Mg<sub>2</sub>Si

Although the cooling rate achieved in a casting after solidification depends on the method of casting applied (i.e. sand casting or permanent mould casting), it is usually not slow enough for the casting to cool under equilibrium conditions. In other words, solid state transformations, such as silicon and Mg<sub>2</sub>Si precipitation, cannot be completed before the











b) modified, as-cast

Figure 6.11 SEM micrographs showing the structural changes occurring during the solution treatment at 540°C.



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c) unmodified, after 4h



d) modified after 4h

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e) unmodified, after 8h



f) modified , after 8h

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Figure 6.11 continued

temperature of the casting drops to room temperature. Therefore, the casting will be supersaturated at room temperature, and the supersaturation of the sasting increases with increasing cooling rate.

The supersaturation of the alloy provides the driving force for precipitation processes during heat treatment. In the Al-Si and Al-Si-Mg alloys, the precipitation of silicon and Mg<sub>2</sub>Si can readily occur during the heating stage of the solution treatment due to their low solubility limits in the Al-solid solution. As already established in chapter 4 (sections 4.3 and 4.6), the depletion of the silicon and magnesium atoms from the aluminum anatrix results in a substantial increase in conductivity.

Figure 6.12 shows the silicon precipitates that form during the initial stage of the solution treatment. In the as-cast state there is no visible precipitate in the Al-solid solution (Figure 6.12 (a)), but after 1h solution treatment precipitates can be clearly seen in the dendritic areas (Figure 6.12 (b)). They are present in both equiaxed and plate-like morphologies, as shown in Figure 6.13. Both the results of an X-ray mapping and spot analysis, given in Figures 6.14 and 6.15, confirm that the precipitates are primary silicon. It is also believed that a certain number of Mg<sub>2</sub>Si particles are precipitated out of solution in the Al-Si-Mg alloys during this stage. However, the Mg<sub>2</sub>Si particles could not be detected by scanning electron microscopy simply because of their size ( $<0.2 \mu$ m).<sup>143</sup>

6.3.1.2 Dissolution of the Precipitates

When a cast alloy is heated isothermally in the single phase region dissolution of precipitates occurs. The dissolution process is diffusion limited and its rate is a function of the diffusion coefficient of the solute atoms in the matrix and the average radius of the precipitates.<sup>140,141</sup> At 540°C the solid solubility limits of silicon and Mg<sub>2</sub>Si in the Al matrix are 1.5 wt. pct.<sup>1</sup> and 1.4 wt. pct.<sup>93</sup>, respectively. Therefore, dissolution of these precipitates will occur during the solution treatment.





b) after 1h

Figure 6.12

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SEM micrographs of Al-8 pct. Si-0.3 pct. Mg alloy showing the precipication process occurring during solution treatment at 540°C.



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Figure 6.13 Silicon precipitates formed in Al-8 pct. Si-0.3 pct. Mg alloy after 1h of solution treatment at 540°C.

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Figure 6.16 illustrates the dissolution of the silicon particles precipitated during the initial stage of the solution treatment. After 4h solution treatment most of the precipitates have dissolved, since only a few coarse particles are visible. Obviously, the dissolution of the silicon and Mg<sub>2</sub>Si particles in the aluminum matrix will cause an appreciable decrease in conductivity because the electrons will be scattered more effectively by dissolved silicon and magnesium atoms than by precipitates (see sections 4.3 and 4.4).

6:3.1.3 Spheroidization and Coarsening

Coarsening and spheroidization of the discontinuous phase occurs at elevated temperatures in most eutectic systems 140 because the interfacial energy of a system decreases with the reduction in interfacial surface area per unit volume of discontinuous phase. The reduction in interfacial energy is the driving force for the coarsening and spheroidization processes, which are also diffusion controlled. 146-148

The coarsening process in a eutectic system occurs by a mechanism known as Ostwald ripening, which involves the growth of larger particles at the expense of smaller ones. In other words, there is a mass transfer resulting from the detachment of atoms from smaller particles and their diffusion through the matrix to reprecipitate onto larger particles. The preferential dissolution of smaller particles takes place due to the concentration gradient existing between the smaller particles, with a relatively small radius of curvature (i.e. higher Concentration region), and the larger particles, with a larger radius of curvature (i.e. lower concentration region). Such a concentration gradient is also present between the tip and the sides of a lamella or plate. The coarsening process also leads to a reduction in surface area, that eventually results in larger spheroidal particles.

It has been established that the rate of spheroidization and coarsening is directly related to the diffusion coefficient of the solute atoms in the matrix,"and increases as the average size of the discontinuous phase crystals

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solution treatment at 540°C.

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decreases.<sup>146,148</sup> Therefore, the fine fibrous silicon particles will spheroidize and coarsen faster than the coarser plate-like or lamellar silicon. The effect of the initial structure of the eutectic silicon on the rate of spheroidization of the silicon phase can be clearly seen in the optical micrographs given in Figures 6.9 and 6.10. The modified silicon spheroidizes rapidly at an early stage of the solution treatment and substantially coarsens with longer treatment. The unmodified silicon lamellate and plates, however, initially break up into smaller particles that eventually spheroidize and coarsen. Figure 6.17 shows several silicon lamellae at the breaking up stage. It is noted that the breakage planes appear at fault lines on the silicon lamellae. The breaking up process occurs more quickly in chill cast alloys since the eutectic silicon phase has a finer lamellar structure. Figure 6.18 shows a schematic representation of the sequence of structural changes that unmodified and modified eutectic silicon undergo during the solution treatment. It is also important to mention that the very coarse silicon plates observed in the slowly cooled unmodified alloys (0 4°C/sec) exhibit no evidence of spheroidization, even after 8h of solution treatment (see Figure -6.19). This is believed to be due to the longer diffusion distances between the plates and the very high stability of these plates, which may result from the lack of defects at which break up usually occurs. 149

The spheroidization and coarsering of the eutectic silicon phasereduces the interfacial area between the eutectic silicon and the aluminum matrix, as well as the degree of interconnection of the silicon crystals, by producing spherical and finely dispersed particles in the aluminum matrix. The effects of these structural changes will be a significant reduction in electron scattering at the silicon-aluminum interfaces and a subsequent increase in the conductivity of two-phase AI-Si alloys.

6.3.2 Effect of the Solution Treatment Time

Figure 6.20 shows the general conductivity behaviour of the Al-Si alloys during a solution treatment at 540°C and indicates that there are three definite stages in the solution treatment. In the following sections the



Figure 6.17 Optical micrograph showing the silicon lamellae at the breaking up stage.

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a) unmodified silicon

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As-cast

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Coarsening

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b) modified silicon

Spheroidization

Figure 6.18 Schematic characterization of the three stages of spheroidization and coarsening of the eutectic silicon phase.

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Figure 6.19 Optical micrograph showing the coarse silicon plates after 16h of solution treatment at 540°C.

conductivity behaviour of the Al-Si alloys during the solution treatment will be discussed in terms of these three stages.

6.3.2.1 Stage I

This stage is a period dominated by the precipitation of silicon and Mg<sub>2</sub>Si, causing a sharp increase in conductivity. Concurrently, spheroidization of the eutectic silicon phase and dissolution of precipitates take place. However, their effect on conductivity will be smaller when compared to that of the precipitation process.

Metallographic examination reveals a substantial degree of precipitation of silicon during the initial period of solution treatment (see Figure 6.10). Hardness measurements taken across the aluminum dendrites (Figure 6.21) show that the hardness decreases appreciably during this stage, whereas the conductivity increases, confirming that precipitation and growth of these precipitates are occurring. An interesting observation is that a minimum hardness occurs in Figure 6.21 at the maximum conductivity.

The initial increases in conductivity observed in the unmodified alloys are noticeably larger than those detected in the modified alloys (Figures 6.2 and 6.20). This is probably due to the retarding effect of strontium on the precipitation of silicon and Mg<sub>2</sub>Si, as discussed in chapter 4 (sections 4.3 and 4.6). Additionally, the breaking up process in unmodified silicon particles will have a greater effect on the conductivity than the spheroidization process that occurs primarily in modified alloys during this stage of solution treatment. This can be explained by considering that the breaking up leads to a lesser degree of interconnection and smaller silicon plates (Figure 6.11 (c)) which will allow the current to flow more easily through the eutectic areas. In modified alloys, however, electron transport is considerably easier, even in the as-cast state, due to the presence of a fibrous silicon phase. The effect of differing silicon morphologies on conductivity (i.e. electron transportation) has already been described in detail in the previous chapter .




Figure 6.21 Relative conductivity and hardness changes in Al-8 pct. Si-0.3 pct. Mg alloy during solution treatment at 540°C.

Figure 6.3 illustrates that for slowly cooled alloys the initial conductivity increase is much less marked than in the chill cast samples. This is because the degree of supersaturation of these alloys is less due to a slower cooling rate after solidification.

## 6.3.2.2 Stage II

In this stage, changes in conductivity are governed by the dissolution of precipitates and the spheroidization process. The large initial drop in conductivity, which occurs within the first 3 hours of treatment, is primarily due to the dissolution of silicon and Mg<sub>2</sub>Si precipitates. With further solution treatment, the decrease in conductivity begins to level off, indicating a balance between the competing processes of dissolution and spheroidization of the eutectic silicon.

The hardness (Figure 6.21) exhibits a continuous increase due to the dissolution of precipitates and the accompanying solute hardening of the matrix. Considering the data in Figure 6.2, the effect of the dissolution process on changes in conductivity is more dramatic for alloys with low silicon concentrations (<4 wt. pct. Si) where the conductivity decreases to values below those of as-cast alloys. The explanation for this effect is that a lesser amount of spheroidization occurs due to the smaller volume fraction of the eutectic silicon present in these alloys. Therefore, dissolution of the silicon and Mg<sub>2</sub>Si particles is the predominant process.

In addition, the rate of the conductivity decrement after 3h of solution treatment is somewhat slower for the unmodified alloy than for the modified alloy (see Figure 6.20). This is because the spheroidization process occurs to a greater extent in the unmodified alloy since the modified eutectic silicon phase has already been extensively spheroidized during the first stage.

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## 6.3.2.3 Stage II

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As revealed by metallographic examinations (Figure 6.8 (d) and (e)), only coarsening of the silicon particles occurs in the modified alloys during this stage, and it does not have any detectable effect on the conductivity which stays virtually constant (see Figure 6.20). The unmodified silicon phase, however, undergoes further spheroidization (Figure 6.7) and coarsening resulting in a slight increase in the conductivity of the unmodified alloy (Figure 6.20). The final increase in conductivity is more noticeable for slowly cooled alloys than for chill cast alloys. Clearly, the completion of the spheroidization process requires a longer solution treatment in slowly cooled alloys due to the initial coarse as-cast structure of the eutectic silicon phase.

#### 6.3.3 Effect of Modification

Both metallographic examination (Figures 6.8 and 6.10) and conductivity measurements (Figure 6.1 and 6.20) clearly show that the rate of spheroidization and coarsening of the eutectic silicon phase is substantially faster in the modified alloys than in the unmodified alloys. This is due to an increased driving force for spheroidization and coarsening processes provided by the finer as-cast structure of the modified eutectic silicon phase. Similar observations have also been made by other authors. 126, 150

Another aspect to be considered is the effect of a complete solution treatment of 16h on the final conductivity of the unmodified and the modified alloys. The changes in conductivity can be expressed by the quantity  $\Delta\sigma_R = (\sigma_{ST} - \sigma_{AC})/\sigma_{AC} \times 100$ , where  $\sigma_{ST}$  and  $\sigma_{AC}$  are the conductivities of the alloys after a 16h solution treatment and in the as-cast condition, respectively. These measured conductivity changes are plotted as a function of the silicon concentration in Figure 6.22. The unmodified Al-Si alloys having more than 8 wt. pct silicon show a conductivity increase of 10 pct in comparison with the modified alloys in which the conductivity increases by only ~1.5 pct. The larger increase in conductivity observed in the





Figure 6.22 Relative conductivity changes in Al-Si and Al-Si-Mg alloys containing different levels of silicon after 16h solution treatment at 540°C.

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unmodified alloys is the result of a relatively greater structural change in the eutectic silicon phase (i.e. transition from plate-like to spheroidial shape) upon solution treatment.

In fact, the effect of the complete solution treatment on the conductivity of unmodified alloys is comparable to the effect of modification on cast alloys. For example, modification of the Al-8 wt. pct. Si alloy raises the conductivity from 42.7 pct. IACS to<sub>(</sub>46.3 pct. IACS, i.e. an 8.4 pct increase, while a solution treatment of the same unmodified alloy increases the conductivity from 42.7 pct. to 47.1 pct. IACS, corresponding to a 10.5 pct. increase. Møreover, the conductivities of the unmodified (47.1 pct. IACS) and the modified (47.3 pct. IACS) alloys become very similar at the end of the solution treatment simply because the microstructural differences between the eutectic silicon phases disappear upon full solution treatment (Figures 6.10 and 6.11).

#### 6.3.4 Effect of Magnesium

The effect of a magnesium addition on the percent conductivity changes during the solution treatment is given in Figure 6.23. The shape of the curves does not change appreciably in the presence of magnesium, suggesting that the same metallurgical processes, as explained in section 6.3.1, occur in the Al-Si-Mg alloys when they are solution treated. One noticeable difference is that the conductivity change,  $\Delta \sigma_R$ , in the magnesium-containing alloys reaches a constant value rapidly (i.e. after a 3h treatment) compared to the binary Al-Si alloys. This implies that the eutectic silicon phase spheroidizes to a greater extent in the Al-Si-Mg alloys during the second stage of the solution treatment. Similar results are given in Appendix III, obtained for the alloys containing different levels of silicon and magnesium.

Finally, as can be seen in Figure 6.22, the effect of complete solution treatment on the conductivity of unmodified Al-SI-Mg (0.3 wt. pct.) alloys is more pronounced than on the unmodified Al-Si eutectic alloys. For the Al-Si-Mg alloys, the changes in conductivity increase with increasing silicon



Figure 6.23 Relative conductivity changes in Al-Si-Mg alloys containing 0.0 to 0.75 wt. pct. Mg during solution treatment at 540°C.

concentration and reach a maximum of 19 pct. at the eutectic composition (12.6 wt. pct. Si) whereas the conductivity of the Al-Si alloys increases by only 10 pct. The greater increase in the conductivity of the Al-Si-Mg alloys may be due to the coarsening effect of magnesium on the eutectic silicon structure, as discussed in the previous chapter (section 5.1.3); hence a relatively greater degree of spheroidization occurs upon solution treatment resulting in larger conductivity changes.



#### CHAPTER 7

### CONCLUSIONS AND FUTURE WORK

### 7.1 Conclusions

In this work, the electrical resistivity/conductivity behaviour of both solid solution and two-phase Al-Si and Al-Si-Mg alloys has been investigated in order to identify the reasons for the changes in electrical resistivity/conductivity resulting from strontium additions to these alloys. Furthermore, the effect of eutectic silicon in its various microstructural forms on the electrical resistivity/conductivity has been determined.

The conclusions that can be drawn from the present study are as follows:

A) Resistivity/conductivity behaviour of solid solution alloys

- 1. Silicon and magnesium in solid solution have significant effects on the resistivity of pure aluminum; the increase in resistivity as a function of concentration being 0.70 and 0.58  $\mu\Omega$ cm/wt.pct. for silicon and magnesium, respectively.
- 2. Strontium has negligible solid solubility in aluminum and does not alter the solid solubility limits of silicon and magnesium? Therefore, it does not measurably change the resistivity of the Al-matrix of Al-Si and Al-Si-Mg alloys.
- The kinetics of silicon precipitation from a supersaturated aluminum solid solution have been studied using the Johnson-Mehl-Avrami Equation and experimentally calculated Avrami constants have been compared with theoretical values. The results reveal that silicon precipitation occurs on dislocation lines and is a diffusion controlled process.
- 4. The silicon precipitation processes in Al-Si and Al-Si-Sr alloys are isokinetic. However, strontium retards the precipitation of silicon, decreasing the growth rate of silicon particles. Thus, strontium can

exert an indirect effect on the resistivity/conductivity of heat treated alloys.

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5. Strontium does not alter the kinetics of G.P. zone formation in Al-Si-Mg alloys. However, it does suppress Mg<sub>2</sub>Si precipitation during subsequent artificial aging of these alloys.

B) Resistivity/conductivity behaviour of two-phase alloys

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 When eutectic silicon appears in the microstructure, the conductivities of strontium-free and strontium-containing Al-Si alloys are different. Al-Si-Sr alloys have higher conductivities than Al-Si alloys containing the same level of silicon. The conductivity difference increases as the silicon content is increased and reaches a maximum at the eutectic composition

- 2. As the coarseness of the unmodified eutectic silicon increases, the difference in conductivity increases. Therefore, a slower cooling rate during solidification and the addition of magnesium to Al-Si alloys result in larger conductivity differences.
- -3. It was found that the difference in the conductivities of unmodified and modified alloys is due to differences in the electron scattering at the interface between the aluminum matrix and the silicon phase. The modified alloys exhibit higher conductivity, since the electron scattering from fiber-matrix interfaces (i.e. modified silicon phase) is less than from the plate-matrix interfaces (i.e. unmodified silicon phase).

4. Various structural forms and arrangements of the eutectic silicon phase produced by directional solidification produce different effects on the resistivity of the AI-Si alloy Well aligned silicon rods result in the lowest resistivity along the longitudinal axis, whereas silicon plates lead to the highest resistivity in the transverse direction.

5. During the directional solidification of Al-Si eutectic alloys, increased growth rates provided a more random distribution of the eutectic silicon, which yielded an increase in the longitudinal resistivity and a decrease in the transverse resistivity.

C) Solution treatment of two-phase alloys

- During the solution treatment of two-phase Al-Si and Al-Si-Mg alloys,
  precipitation of silicon and Mg<sub>2</sub>Si, dissolution of these precipitates, and
  spheroidization of the eutectic silicon phase occur. The precipitation
  - , and spheroidization processes increase conductivity while the dissolution process causes a decrease in conductivity.
- 2. The unmodified eutectic silicon phase undergoes a more extensive structural change upon solution treatment than does the modified silicon phase. Therefore, a substantial increase in the conductivity of the unmodified alloys is observed after complete solution treatment, whereas the conductivity of the modified alloys remains similar to the as-cast value. In addition, the magnitude of the increase in conductivity of the unmodified alloys increases as the silicon and magnesium contents increase.
- 3. The conductivity differences measured between as-cast unmodified and modified alloys tend to disappear upon solution treatment because their final eutectic silicon structures become similar.
- 4. The rate of the spheroidization and coarsening processes depend on the initial (i.e. as-cast) structures of the eutectic silicon. Spheroidization and coarsening occur substantially faster in modified alloys than in unmodified alloys due to the presence of finer silicon fibers which provide an increased driving force for the process.

### 7.2 Future Work

Studies of the effects of strontium on the kinetics of Mg<sub>2</sub>Si precipitation, and subsequently on the mechanical properties of Al-Si-Mg alloys, should be done since the precipitation hardening behaviour
 of these alloys is critically important in achieving the desired mechanical properties. Electrical conductivity/resistivity can be used

during these studies. In addition, TEM work should be carried out simultaneously to support the resistivity/conductivity results.

 The results of the more precise electrical resistivity measurements on unmodified and modified AI-Si alloys at room temperature, 77K and
 4.2K can be used to develop a mathematical model of the effect of silicon morphology on the resistivity of these two-phase alloys.

Further work on the electrical conductivity behaviour of commercial Al-Si alloys should be conducted in order to establish the effects of processing variables on conductivity differences measured upon modification of these alloys. A list of some variables, besides Simodification, which have a major influence on the cast Al-Si structure include:

-grain refinement

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-cooling rate before and after solidification

-the presence and amount of other alloying elements such as Cu, Fe, Ti, Zn and Mn.

Finally, the modification of one of the eutectic phases is not restricted only to the Al-Si system. Therefore, studies similar to the present one can be conducted on other eutectic systems in which the modification process occurs, such as cast iron, Al-Mg<sub>2</sub>Si<sup>30</sup>, Al-Al<sub>2</sub>Ni, Al-FeAl<sub>3</sub> and Al-Ge<sup>28</sup>.

#### CHAPTER 8

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## IMPORTANCE OF THIS RESEARCH TO THE AI-SI FOUNDRY INDUSTRY

As pointed out in section  $1.2^{\circ}$ , better non-destructive testing methods are being implemented in the foundry industry to produce higher quality castings. The results of this work may be used directly to improve electrical conductivity techniques, which can be employed as a non-destructive testing method for AI-Si casting alloys. In the following sections, several possible applications of electrical conductivity during the production of AI-Si cast parts will be given, together with suggestions for increasing the reliability of this test method.

## 8.1 <u>The Use of Conductivity to Monitor the Modification Potential of the</u> <u>Melt</u>

A method used to control the melt quality has already been described in section 1.2.4 and in other articles<sup>125,126</sup>. The results of the present study reveal three important parameters which have to be closely controlled to obtain more reliable predictions of the degree of modification.

i) Solidification rate of the test buttons: The same solidification rate must be used for both unmodified and modified samples, and the chosen rate should be slow enough to produce a coarse plate-like silicon phase in the unmodified button. It has been shown that the use of an insulated graphite mould provides satisfactory conditions for controlling the solidification rate<sup>125</sup> (see section 3.2).

ii) Cooling rate of the buttons after complete solidification: It is suggested that the cooling rate of the samples should be accelerated by quenching or air cooling. This is important to minimize the effect of the modifier on precipitation processes which occur during the cooling period. In quenched samples, the difference that will be measured between the conductivities of the buttons at room temperature will be solely the result of modification of the eutectic silicon.

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iii) Alloy composition: Conductivity techniques can be used for most of the Al-Si castinghalloys. However, for alloys containing high concentrations of certain alloying elements (e.g. magnesium and copper), which exhibit high solid solubility in the Al-matrix, the measured conductivity difference can be small in the case of quenched and air cooled samples, even though full modification of the eutectic silicon is achieved. The reason is that the bulk conductivity of these alloys is determined primarly by the conductivity of the aluminum matrix, hence the effect of the eutectic silicon structure will be masked.

## 8.2 <u>The Use of Conductivity to Characterize the Degree of Modification in</u> <u>Castings</u>

This can be done by taking conductivity measurements at several locations on a casting and then comparing these values to those taken from an unmodified test casting at locations which possess solidification and cooling rates similar to those in the modified casting. It should be noted that different conductivity values will be obtained from different locations on the same casting, depending on the section thickness. In general, heavier sections exhibit lower conductivity than thinner sections, due to increased microstructural coarseness.

## 8.3 The Use of Conductivity to Monitor Heat Treatment of Castings

As shown in chapters 4 and 6, the variation in conductivity during solution treatment and subsequent aging treatment can yield an insight into microstructural changes. Therefore, the conductivity method can be useful in predicting optimum heat treatment conditions, particularly treatment time and temperature.

## 8.4 <u>The Use of Resistivity to Characterize the Degree of Fiber Alignment in</u> <u>In-situ Composites</u>

Fiber branching and misalignment are two commonly observed defects in directionally solidified eutectics and are detrimental to their mechanical properties<sup>28</sup>. It has been demonstrated that the longitudinal and transverse resistivities of directionally solidified alloys are very sensitive to the degree of alignment and branching of the silicon rods (see section 5.3.2). Therefore, by measuring the electrical resistivity of a directionally aligned eutectic along the longitudinal and transverse directions, and then comparing these values to ones obtained from a standard sample known to possess good fiber alignment with a minimum of branching, it may be possible to gain some information about the microstructure of the eutectic. Some of the commercially important Al-based eutectics, such as Al-Al<sub>3</sub>Ni, Al-AlCu and Al-Al<sub>3</sub>Fe, could be characterized employing this technique.

### STATEMENT OF ORIGINALITY

The following aspects of the present work are considered to be original contributions to knowledge:

- 1) The present work represents the first complete fundamental study of the electrical resistivity/conductivity behaviour of cast Al-Si and Al-Si-Mg alloys with respect to the microstructural changes which occur due to strontium addition.
- 2) The solid solubility limit of strontium in pure aluminum has been determined for the first time by means of residual resistivity ratio (RRR) measurements. In addition, based on electrical conductivity and RRR data, it has been shown that strontium has no effect on the solubility of silicon and magnesium in aluminum.
- 3) This thesis contains the first detailed analysis of the kinetics of silicon precipitation in single phase Al-Si and Al-Si-Sr alloys. The rate and Avrami constants have been calculated and found to show good agreement with theoretical models. The effect of strontium on the precipitation process in Al-Si-Mg alloys has also been investigated and strontium has been shown to suppress Mg<sub>2</sub>Si precipitation during artificial aging.
- 4) A rigorous analysis of the conductivity differences between the modified and unmodified two-phase Al-Si and Al-Si-Mg alloys based on structural differences in the eutectic silicon phase has been made. It has been shown conclusively that these differences in conductivity are brought about by the changing morphology of the non-conducting eutectic silicon. Furthermore, the results of the electrical resistivity measurements performed on directionally solidified Al-Si alloys represent a significant contribution towards the understanding of how the various microstructural forms of the eutectic silicon affect resistivity.

- 5) The parameters that influence the magnitude of the measured conductivity difference between unmodified and modified Al-Si and Al-Si-Mg alloys have been determined. These findings enable more efficient and accurate use of electrical conductivity techniques in the foundry industry as a non-destructive testing method.
- 6) The solution treatment (T4) of two-phase Al-Si and Al-Si-Mg alloys in both modified and unmodified forms has been followed by electrical conductivity measurements and metallographic examinations. For the first time three definite stages of the solution treatment of the alloys have been identified corresponding to precipitation, redissolution of silicorf and Mg<sub>2</sub>Si particles and coarsening of the eutectic silicon phase.

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## APPENDIXI

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## ELECTRICAL CONDUCTIVITY DATA OBTAINED FROM SINGLE PHASE AI-SI AND AI-SI-Mg ALLOYS

Âlloy	۳		Cond	uctivity	y†† (pc	t. IACS)	at Agi	ing Tin	ne (h)		
Name †	0	1 2	1	1 <del>1</del>	2	3	<b>4</b> ·	6	8	12	16
1A	64.3	64.2	64.2	64.3	64.2	64.2	64.3	64.3	64.3	64.2	64.4
2A	62.0	62.1	62.0	<b>62.0</b> `	62.1	62.0	62.1	62.1 <sup>°°</sup>	62.1	62.1	62.3
3A	59.6	59.7	59.7	59.9	60.0	60.8	61.4	61.6	616	<sub>.</sub> 61.7	61.8
4A	56.7	56.9	57.6	58.3	58.7	61.7	62.0	62:.0	62.0	62.0	62.0
5A	53.7	53.9	55 4	56 <i>.</i> 7	57.3	60.5	60.7	60.8	60.8	61.0	60.9
6A	51.4	52.0	55.3	57.7	58.5	59.8	60.0	60.1	60.2	60.3	60.5
7A	50.2	50.9	54.7	56 <sub>5</sub> 9	57.5	59.6	59 <sup>.</sup> 8	59.8	60.0	60.0	60.2
8A	50.0	50.9	<sup>,</sup> 54.6	<b>56.8</b>	57.8	59.5	59.7	59.8	59 9	60.0	59.7
9A	49.2	50.7	55.6	58.1	58.7	<u>5</u> 9.6	59.8	59.9	60.0	60 <mark>.</mark> 1	60.2
10A	47.7	50.6	56.7	58.5	59.1	59.3	59. <b>6</b>	59.7	59.9	60.0	60.0
11A	46.8	51.6	57.4	58.6	59.0	59:1	59.2	59.4	59.6	59.7	59.9
1B	64.6	64.6	64.6	64.6	64.6	64.5	64.5	64.6	64.6	64.6	64.4
2B	62.7	62.7	62.8	62.8	62.9	62.8	62.8	62.9	62.8	62.8	62.8
3B	60.4	60.6	60.6	60.7	60.7	60.8	60.9	61.5	61.6	61.7	61.5
4B	56.7	56.8	57.0	57.2	57.3	59.1	60.3	60.9	61.0	61.1	60.9
5B	54.3	54.5	55.2	56.3	56.8	59.6	60.2	60.4	60.5	60.7	60.7
6B	51.7	52.0	53.7	55.7	56.4	59.5	59.8	60.0	60.1	60.2	60.2
`7B	50.4	50.9	55.7	58.5	58.3	59.2	59.5	59.8	60.0	60.1	60.2
8B	50.3	50.9	54.1	56.5	57.4	59.3	59.6	59.7	59.9	60.0	59.9
9B	50.2	51.1	55.0	57.2	57.8	59.4	59.4	59.7	<sup>59.7</sup>	59.8	60.0
10B	47.8	49.0	55.6	57.9	58.4	59±0	59.1	59.3 <sub>.</sub>	59.5	59.7	59.9
" 11В	47.0	50.8	57.0	58.4	58.7	58.8	59.1	59.2	59.3	59.5	59.7

Appendix IA. Conductivity of Single Phase Al-Si and Al-Si-Sr Alloys During Aging at 340 °C

See Table 3.4 for chemical composition
 Average of three conductivity measurements

Alloy Name	n	k (h <sup>- </sup> ")	Allòy Name	n	k (h <sup>-</sup> '')
1A	- \	-	1B	-	
2A	- 1	-	2B	_	-
<sup>∿</sup> 3A	0.71	0.17	3B	0.72	0.12
4A	0.75	0.36	4B	0.70	0.28
5A	0.79	0 48	5B	0.69	0.36
6A _	0.80	0.71	6B	0 77	0.57
7A	0.74	0.86	7B	0.75	~ 0.63
8A	0.76	0.94	8B	0.80	0.75
9A	0.77	1.03	9B	0.78	0.84
10A°,	0.77	1.25	10B	0.73	1.00
11A	0.78	1.42	11B ุ	0.81	1.18

Appendix IB. The Experimental n and k Constants for Single Phase Al-Si and Al-Si-Sr Alloys

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Alloy Name†	Conductivity <sup>††</sup> (pct. IACS)	Alloy Name†	Conductivity <sup>††</sup> (pct. IACS)
1M1	62.6	<b>1</b> M1B	62.8
1M2	59.4	1M2B	59.7
1M3	57.2	1M3B	- 56.7
1M4	56.8	1M4B<	54.2
1M5 <sup>-</sup>	56.6	1M5B	53.7
1M6	56.4	1M6B	53.8
1M7	55.8	1M7B	53.8
1M8	55.6	1M8B	53.8
2M1 (\	61.1	2M1B	61.0 –
2M2	57.5	2M2B	57.6
2M3	57.2	2M3B	57.0
2M4	55.3	2M4B	55.4
2M5	55.0	2M5B	55.2
2M6	54.8	2M6B	55.0
3M1	57.5、	3M1B	57.7
3M2	58.1	3M2B	57.7
3M3	57,3	3M3B	57.3
<sup>▲</sup> 3M4	56.3	3M4B	56.1
3M5	55.2	3M5B	<sup>、</sup> 55.4
3M6 /	54.1	3M6B	54.1

Appendix IC. Conductivity of Furnace Cooled Al-Si-Mg and Al-Si-Mg-Sr Alloys

See Table 3.4 for chemical composition
 three measurements

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Alloy Name†	Conductivity <sup>††</sup> (pct. IACS)	Alloy Name†	Conductivity <sup>††</sup> (pct. IACS)
4M1	55.2	4M1B	55.7
4M2	57.5	4M2B	57,1
4M3	57.4 <sup>·</sup> ·	<b>4M3B</b>	57.0
4M4	55.6	4M4B	55.8
· 4M5	55.1	4M5B	55.4
4M6	53.5	4M6B	53.3
5M1	53.2	5M18	53.4
5M2	<b>Š6</b> .6	5M2B	54.8
~5M3	57.3	5M3B	57.1
5M4	56.0	5M4B ₊	<b>56.1</b>
5M5 🔹	55.7	5M5B	55.5
<sup>•</sup> 5M6	54.1	5M6B	54.2

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Alloy	Conductivity <sup>†</sup> (pct. IACS) at Aging Time (h)												
Name	0	1	0 1 1	1	1 <del>1</del>	2	3	4	5	6	12	24	36
1M1	62.4	62.4	<del>-62</del> -3	62.4	62.4	62.3	62.4	62.4	62.4	62.4	62.3	62.4	62.4
1M2	59.8	59.8	59.7	59.8	59.8	59.8	59.7	59.6	59.6	59.8	59.7	59.8	59.8
,1M3	56.8	56.8	56.9	56.9	56.8	5 <b>6</b> .8	56.7	56.5	56.8	57.0	56.9	56.8	56.8
1M4	52.8	52.8	52.8	52.8	<sup>`</sup> 52.7	52,7	52.6	52.6	52.6	53.0	53.0	52.8	52.7
1M5	50.3	50.3	50.3	50.3	50.2	50 2	50.1	50.Ź	50.2	50.4	50.4	50.2	50.2
1M6	49.6	49.6	49.5	49.6	49.6	49.	49.6	49.8	49.6	50.2	49.9	49.6	49.6
1M7	48.9	48.9	48.8	<sup>≂</sup> 48.9	48.8	48.8	48.8	48.9	49.0	49.0	48.8	48.8	48.8
1118	48.5	48.5	48.5	48.5	48.5	48.5	48.6	48.6	48.7	48.8	48.7	48.5	48.8
1M1B	62.2	62.2	62.1	62.1	62.2	62.2	62.2	62.1	62.1	62.1	62.1	62.1	62.1
1M2B	60.0	60 0	60.0	60.0	60.0	60.0	59.9	59.9	60.0	60.0	60.1	60.1	60.1
1M3B	56.6	56.6	56.7	56.7	56.7	56.7	56.5	56.4	56.5	56 <i>.</i> 8	56.7	56.7	<sup>?</sup> 56.7
1M4B	53.5	53.5	53.5	53.5	53.5	53.4	53.3ັ	53.3	53.4	53.6	53.6	53.5	53.5
1M5B	52.1	52.2	52.2	52.2	52.2	52.2·	52.1	52.1	52.0	52.3	52.3	52.1	52.1
1M6B	49.7	49.7	49.6	49.6	49.4	49.4	49.7	49.8	49.9	49.8	49.6⁄	49.6	49.6
1M7B	48.8	48.8	48.7	48.7	48.7	48.7	48.7	48.6	48.7	48.6	48.8	48.7	48.7
1M8B	48.6	48.6	48.5	48.5	48.7	48.7	48.7	48.9	48.8	48.9	48:6	48.6	48.8

# Appendix ID. Conductivity of AI-Si-Mg and AI-Si-Mg-Sr Alloys During Aging at 22°C

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**†** Average of three conductivity measurements

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Appendix ID. continued

Allov		Conductivity <sup>†</sup> (pct. IACS) at Aging Time (h)														
Name	0	· 1	. <del>1</del>	1	11 11	<sup>°</sup> 2	3	4	5	6	12	24 <sup>%</sup>	36			
2M1	60.1	60.2	60.1	60.2	60.1	60.0	60.1	60.1	60.1	60.1	60.1	60.1	60.1			
2M2	57.2	57.2	57.2	57.1	57.1	56.9	56.8	56.8	<b>56.8</b>	57.0	57. <del>0</del>	<i>י</i> \$7,1	57.0			
2M3	54.0	54.0	54.0	54.0	54.0-	54.0	53.9	53.8	53.6	<b>53.8</b>	54.0	54.1	53.8			
2104	51.2	51.2	51.1	51.0	50.9	50.9	50.9	50.9	50.5	50.5	50.5	50.5	50.6			
2M5	50.0	49.9	49.9	49.7	49.5	49.5	49.5	49.4	49.4	49.5	49.5	49.4	49.5			
2M6	47.5	47.5	47.5	47.3	47.2	47.1	47.1	47.1	47.1	47.1	47.0	47.0	47.0			
2 <b>Ņ</b> 8	47.0	46.9	46.8	<b>46.6</b>	46.4	46.4	46.4	46.4	46.4	<b>46.4</b>	46.4	46.4	46.3			
2M1B	60.2	60.2	60.1	60.2	60.1	60.0	60.1	60.1	60.1	60.1	60.1	60.1	60.1			
2M2B	57.4	57.5	57.5	57.4	57.4	57.5	57.3	57.4	57.4	57.4	57.4	57.4	57.4			
2M3B	54.3	54.2	54.2	54.1	54.0	53.9	53.8	53.9	53.9	53 <i>.</i> 9	53.9	53.8	53.9			
2M4B	51.4	51.4	51.4	51.3	51.2	51.1	50.9	50.8	50.8 <sub>.</sub>	50.8	50.8	50.8	50,8			
2M5B	50.3	50.2	50.0	49.8	49.7	49.7	49.6	49.6	<b>49.6</b>	49.6	49.6	49.6	49.6			
2M6B	47.1	47.0	46.9	46.7	46.6	46.4	<sup>'</sup> 46.5	46.5	46.5	46.5	46.5	46.3	46.3			
2M8B	46.8	46.7	46.6	46.4	46.3	46.1	46.0	46.0	46.0	46.1	46.0	46.0	46.0			

+ Average of three conductivity measurements

Appendix ID. continued

Allov	-	Conductivity <sup>†</sup> (pct. IACS) at Aging Time (h)													
Name	Ò	1	1 2,	1	11	2	3	4	5	6	12	24	36		
3M1	57.0	57.0	56.8	56.8	56.9	57.1	57.0	57.0	57.0	57.0	57.0	57.0	57.0		
3M2	54.3	54.3	54.4	54.3	54.3	54.3	54.0	53.6	53.7	53.8	56.9	54.0	54.0		
3M2	50.8	50.7	50.6	50.3	50.2	50.2	50.1	50.0	49.8	50.0	49.9	49.8	49.8		
3M4	47.6	47.5	47.4	47.2	47.2	47.0	46.8	46.7	46.6	46.7	46.6	46.7	46.6		
3M5	44.8	44.7	44.5	44.3	44.1	44.0	43.9	43.8	43.9	43.9	44.0	43.9	43.9		
3M6	<b>4</b> 3.9	43.7	43.4	43.2	43.2	43.0	42.7	42.8	42.8	42.7	42.7	42.7	42.7		
3M8	41.6	41.5	41.4	41.2	41.1	41.0	40.9	40.8	40.8	40.8	40.8	40.8	40.6		
3M1B	57.Ż	57.2	57.0	57.1	57.1	57.1	<b>57.2</b> ′	57.2	57.2	57.1	57.1	57.1	57.1		
3M2B	55.0	55.0	55.0	55.1	55.1	54.9	54.8	54.9	54.9	54.9	54.8	54.8	54.8		
ЗМЗВ	51.7	51.6	51.5	51.1	51.0	50.9	50.8	50.8	50.8	-50:8	50.7	50.8	50.8		
ЗМ4В	45.7	45.6	45.5	45.3	45.2	45.1	44.9	44.8	44.8	44.8	44.8	44.7	44.7		
3M5B	44.9	44.8	44.5	44.3	44.2	44.1	44,0′	-44.0	43.9	44.0	44.0	44.0	44.0		
3M6B	42.2	41.8	41.3	41.1	40.9	40.8	40.9	40.8	40.8	40.8	40.8	40.7	40.9		
3M8B	41.8	41.4	41.2	41.0	40.8	40.8	40.7	40.7	40,7	40.7	40.7	40.7	40.7		

+ Average of three conductivity measurements

## Appendix ID. continued

Allov		Conductivity <sup>†</sup> (pct. IACS) at Aging Time (h)													
Name	0 &	4	- 12	1	1 <del>]</del>	2	3	4	5	6	12	24	36		
4M1	54.7	54.7	54.7	54.7	54.6	54.5	54.3	54.3	54.2	54.3	54.3	54.3	54.2		
4M2.	52.6	52.6	52.6	52.5	52.5	52.5	52.3	52.0	5 <u>2</u> .0	52.0	52.0	52.0	52.0 <sup>°</sup>		
4M3 🔹	48.5	48.4	48.3	48.0	<b>4</b> 7.9	47.8	47.6	47.5	47.4	47.4	47.4	47.4	47.5		
4M4	45.3	45.2	45.1	44.8	44.7	44.6	44.4	44.2	44.3	44.2	44.3	44.2	44.2		
4M5	43.Ź	43.0	42.9	42.7	42.5 <sup>.</sup>	42.2 <sup>-</sup>	42.3	42. <del>3</del>	42.2	42.1	42.1	42.1	42.1		
4M6	41.1	41.0	40.8	40.7	40.5	40.3	40.2	40.3	40.3	40.2	40.2	40.3	40.2		
4M1B	55.0	55.0	55.0	54.9	54.7	54.7	54.7	54.8	54.7	54.8	54.7	54.7	54.7		
4M2B	53.2	53.3	53.2	53.1	52:9	52. <b>8</b>	52.6	52.7	52.7	52.6	52.6	52.7	52.7 <sub>-</sub>		
4M3B	49.1 <sup>°</sup>	49.0	48.8	48.5	48.4	48.2	48.0	47.8	47.7	47.8	.47.8	47.8	47.8		
4M4B	45.4	45.3	45.2	44.9	44.8	44.7	44.5	44.4	44.3	44.4	44.3	44.4	44.4		
4M58	43.4	43 <sub>9</sub> 2 <i>°</i>	43.0	42.8	42.9	42.7	42.7	42.6	42.6	42.6	42.6	42.6	42.6		
′4M6B	41.4	41.3	41.2	41.0	40.9	40.6	40.6	40.5	40.4	40.5	40.5	40.5	40.4		

**†** Average of three conductivity measurements

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Append	lix ID.	contin	ued
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Арренаіх				Ô												
Alloy		Conductivity <sup>†</sup> (pct. IACS) at Aging Time (h)														
Name	0	14	1/2	1	11	2	3	4	5	6	12	24	36			
5M1	52.7	52.7	52.7	52.5	52.1	52.0	52.0	52.0	52.0	52.0	52.2	52.2	52.2			
5M2	50.4	50.4	50.4	50.3	50.2	50.0	49.9	49.8	49.6	49.5	49.3	49.3	49.3			
5M3 <sup>°</sup>	<u>46.8</u>	46.7	46.6	46.0	45.8	45.6	45.3	45.4	45.2	45.3	45.3	45.3	45.3			
5M4	43.3	43.2	43.1	42.7	42.5	42.4	42.3	42.1	42.1	42.2	42.1	42.2	42.1			
5M5	42.0	41.8	41.7	41.6	41.4	41.2	41.0	41.1	41.1	41.0	41.1	41.1 <sup>°</sup>	41.1			
5M6	40.9	40.9	40.8	40.7	40.8	40.6	40.3	40.3	40.3	40.4	40.3	40.3	40.4			
5M7	40.4	40.3	40.2	40.1	40.1	40.0	39.9	39.8	39.8	39.8	39.8	39.7	39.7			
5M1B	52.8	52.7	52.7	52.8	52.6	52.4	52.1	52.1	52.2	52.3	52.2	52.2	52.2			
5 <sup>:</sup> M2B	49.8	49.8	49.9	49.7	49.5	49.1	48.9	48.9	48.8	48.8	48.8	48.8	48.9			
5M3B	47.0	46.9	46.8	46.5	46.4	46.3	46.1	46.0	46.1	46.1	46.0	46.0	46.1			
<sup>-</sup> 5M4B	43.6	43.4	43.5	43.0	42.8	42.4	42.1	42.1	42.0	42.0	42.1	42.1	42.1			
5M5B	42.3	42.1	42.0	41.9	41.8	41.7	41.5	41.4	41.6	41.4	41.2	41.5	41.4			
5M6B	41:1	41.0	40.8	40.6	40.5	40.4	40.4	40.4	40.5	40.3	40.3	40.3	40.4			
5M7B	40.3	40.2	40.1	40.0	39.8	39.7	39.8	39.7	39.7	39.7	39.7	39.7	39.7			

† Average of three conductivity measurements

Alloy		C	onduct	ivity† (	pct IA	CS) at /	Aging	Time (ł	ו)	
Name	2	3	4	6	8	10	12	14	16	18
1M1	62.4	62.4	62.3	62.3	62.3	62.3	62.3	62.2	62.3	62.3
1M2	59.6·	59.6	59.5	59.5	59 5	59.5	59\.5	59.5	59.6	59.6
1M3	56.7	56.6	56.6	56.6	56.6	56.6	56.6	56 7	56.7	56.7
1M4	52.7	52 6	52.6	52.6	52.7	52.8	52.7	52.7	52.7	52.8
1M5	50.2	50.2	50 3	50.2	50 3	50 4	50 6	50.9	50.9	511
1M6	49.8	50 <del>1</del> 0	50.0	50 1	50.1	50.3	50.4	50 ճ	50 8	51.0
1M7	48.9	49.0	49.2	49.4	49.6	49.7	49.9	50.0	50.0	50.1
. 1M8	48.6	48.7	48.9	49.0	49.2	49.4	49.6	49.8	50.0	50.0
1M1B	62.2	62.2	62 1	62.3	62.2	62.2	62.2	62 2	62 2	62.2
1M2B	59.9	<u>6</u> 0.0	59.9	59.8	59.9	59.8	59.9	59 7	59.8	59.8
1M3B	56.5	56.5	56.4	56.4	56 4	56.4	56.5	56.6	56.6	56 6
1M4B	53.4	53.4	53.3	53.2	53. <u>3</u>	53.3	53.4	53 4	53 4	53.4
1M5B	52.0	52.1	52.0	52.0	52.0	52.0	52.1	52.3	52.3	52.3
11M6B	49.3	49.4	49.5	49.5	49.6	49.8	50.0	50.5	50.7	50.8
1M7B	48.7	48.8	48.9	49.3	49.5	49.7	49.8	49.9	49.9	49.9
1M8B	48.8	48.8	48.9	49.0	49.3	49.6	49.8	49. <u>9</u>	50.0	50.0
2M1	60.1	60.1	60.0	60.0	60.0	60.2	60.2	60.2	60.1	60.2
2M2	56.9	56.8	56.8	56.9	56.9	56.9	56.9	56.9	56 9	56.9
2M3	53.8	53.7	53.7	53.7	53 7	53 7	53.9	54.3	54 5	54.5
2M4	50.5	50.5	50.4	50.6	50.7	50.8	51.3	518	52.0	52.3
2M5	49.5	49.5	49.8	50.2	50.4	50.6	51.0	51.2	51.2	51.4
2M6	47.0	46.8	47.3	47.4	47.7	47.9	48.5	49.4	49.8	50.2
,2M8	46.1	46.1	46.5	46.9	47 4	47.8	48.2	49,0	44.3	49.6

Appendix IE. Conductivity of Al-Si-Mg and Al-Si-Mg-Sr Alloys During Aging at 175°C

† Average of three conductivity measurements

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Appendix IE. continued`

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			I	+ /	·	ds) at	A aina '	Time (l	2)	
Alloy	· ·		onauci				Aging	i iiie (i	·/	
Name	2	3	4	6	8	10	12	14	16	18
2M1B	60.2	60.2	60.2	60.1	60.3	60.1	60.2	60.2	60.3	60.3
2M2B	57.4	57.3	57.4	57.4	57.6	57.4	57.4	57.4	57.4	57.4
2M3B	53.9	53.9	54.0	54.0	54.1	54.0	54.1	54.2	54.6	54.8
2M4B	50.6	50.4	50.5	50.6	50.8	51.0	51.4	51.8	52.0	52.1
2M5B	49.6	49 4	49.6	50.0	50.4	50.8	51.2	51.4	51.4	51.3
2M6B	47.0	46.7	47.2	47.3	47.8	48.1	48.8	49 5	49.7	49.9
2M8B	46.0	46.0	46.3	46.7	47.1	47.4	47.9	48.2	48.8	48.9
3M1	57.0	57.0	57.0	56.9	<sup>1</sup> 56.9	57.0	57.1	57.1	57.1	57.1
3M2	53.6	53.3	53.4	53.5	53.5	53.5	53.7	53.8	53.8	53.8
3M3	48.8	48.8	49.0	49.1	4 <del>9</del> .2	49.3	<sup>*</sup> 49.9	50.7	51.0	51.4
3M4	46.6	46.5	46.6	46.7	46.9	47.0	47.7	49.0	49.6	50.2
3M5	43.7	43.9	44.1	44.8	45.4	<sup>.</sup> 45.9	46.2	46.9	47.7	48.2
3M6	42.6	42.4	43.0	43.4	44.2	44.8	45.4	46.0	46.3	46.8
3M8	40.9	40.7	41.4	41.9	42.7	43.1	44.2	45.0	45.3	45.6
`3M1B	57.2	57.2	57.2	57.0	57.0	57.1	57.2	57.2	57.2	57.2
3M2B	54.1	54.1	54.1	54.2	54.2	54.2	54.4	54.7	54.7	54.7
3M3B	50.6	50.4	50.4	50.5	50.7	50.7	51.4	52.1	52.3	52.6
3M4B	45.8	45.8	46.0	46.2	46.8	46.9	47.7	49.0	49.5	50.0
3M5B '	43.8	43.8	44.0	44.2	44.8	45.6 <sup>°</sup>	46.6	47.0	47.5	47.8
3M6B	41.3	41.5	41.8	41.9	42.4	42.6	43.7	45.2	45.6	45.9
3M8B	41.0	40.8	40.9	42.1	43.3	43.8		45.7	45.9	46.2

Average of three conductivity measurements

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## Appendix IE. continued

Âlloy Name	Conductivity† (pct. IACS) a't Aging Time (h)									
	2	3	4	6	8	tO	12	14	16	18.
4M1	54.2	54.2	54.2	54.2	54.2	54.1	54.2	54.2	54.2	54.2
4M2	51.8	51.8	51.9	52.0	52.0	52.0	52.0	52.1	52.1	52.1
4M3	47.4	47.4	47.5	47.5	47.5	47.6	47.9	48.4	48.8	49.0
4M4	44.2	44.0	44.5	44.5	45.0	45.5	46.0	46.6	47.1	47.6
4M5	41.8	42.0	42.6	43.6	44.7	45.0	45.8	46.2	46.9	47 2
4M6	40.5	40.8	41.3	42.0	43.1	43.6	44.5	46.2	46.8	47.0
4M1B	55.0	55.0	55.0	54.9	<sup>•</sup> 54.7	54.9	55.0	55.0	55.0	55.0
4M2B	5 <sup>`</sup> 2.4	52.7	,52.7	52.8	52.8	52.9	52.9	5 <sup>5</sup> 2.9	52.9	52.9
4M3B	47.8	<b>47.8</b>	47.9	47.9	48.0	48.6	49.0	49.7	49 9	50.0
4M4B	44.4	44 4	44.7	44.9	45.5	45.9	46.3	47.8	48.6	<b>49</b> .1
4M5B	42.4	42.6	42.9	43.5	44.1	44.9	45.4	46.6	47 3	47.7
4M6B	40.2	40.0	40.7	41.2	41.8	42.9	43.5	44.4	44.8	45.2
5M1	52.0	52.0	52.1	52.2	52.2	52.3	52.4	52 4	52.6	52.6
5M2	49.0	49.3	49.4	49.4	49.4	49.4	49 5	49.7	50 0	50.2
5M3	45.0	45.2	45.3	45.6	46.0	46.4	47.2	47 8	48.4	48 8
5M4	41.8	42.0	42.4	43.3	45.3	45.5	46.2	47.2	47.8	48.0
5M5	41.2	41.3	42.0	42.8	45.0	45.2	45.9	46.8	47.0	47.3
5M6	40.2	40.4	41.1	41.7	42.9	43.6	44.7	45.2	45.6	45.8
5M7	39.8	40.0	40.8	41.5	<i>.</i> 42.4	43.3	44.2	44.8	45.4	45.6
5MJB	52.1	52.2	52.2	52.2	52.4	52.5	52.7	52.7	52.7	52.7
5M2B	48.9	49.0	49.2	49.3	49.4	49.4	49.5	49.7	50.0	50.2
5M3B	46.0	46.0	46.0	46.1	46.2	46.4	47.0	47.7	48.1	48.5
5M4B	42.8	43.4	44.0	45.2	45.8	46.1	46.3	46.5	47 1	47 4
5M5B	41.4	41.4	41.7	42.0	44.3	44.6	45.2	46.4	46.6	46.8
5M6B 🔊	40.1	40.2	40.9	41.1	42.9	43.3	44.4	45.0	45.3	45.4
5M7B	39.7	40.0	40.7	41.2	41.8	42.9	43.7	44.4	44.9	45.0

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† Average of three conductivity measurements
## APPENDIX II

# X-RAY ANALYSIS OF DIRECTIONALLY SOLIDIFIED AI-SI EUTECTIC ALLOYS

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#### Table IIA X-ray diffraction peak files.

a)	aro	₩th	rate	0.2	cm/h
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### DETECTED PEAKS FILE

#### "PW1700 Automated Powder Diffractometer System 1

Listed UI file name File date : D13.DI : 9-SEP-86 Raw data file riame Sample identification D13.RD : d13 : 8-SEP-86 17:04 Measurement date/time : 40 kV, 20 mA Generator settings : 1.54060, 1.54439 Ang Cu alphal.2 wavelengths : 25.00 s/deg Step size, sample time : 0.020 deg, 0.50 s, Monochromator used : Yes Automatic (Specimen length: 12.5 mm) Divergence slit : 20 Analysis program number : 10.010 - 137.990 deg Peak "angle range 0.82513 - 8.8294 Ang

Range in D spacing : 0.82513 - 8.8294 Ang Peak position criterion : Top of smoothed data Cryst peak width range : 0.00 - 2.00 deg Minim peak significance : 0.75 Number of peaks in file : 40 (Alphal: 26, Amorphous: 0) Maximum intensity : 2905.cts, 5810.4 cps

Pesk	Angle	Tip width	Реэк	Backg	D spac	I/Imax	Тур	•	Sign
ПO	(dēg)	(deg)	(cts)	(cts)	(Ang)	(2)	A1 A2	Ot	-
1	10.8050	1.28	10.	3.	8.1815	0.35	x x		1.15
2	28.4075	0.08	114.	з.	3.1393	3.94	ХХ		1.86
3	34.5525	0.12	15.	з.	2.3449	0.52		Bt	1.20
4	36.7075	0.08	21.	4.	2.4463	0.73	X ° X		0.76
5	<b>38.4050</b>	0.14 -	2905.	4.	2.3420	100.00	Х		26.92
6	38.5300	0.04	1076.	4.	2.3404	37.03	X		1.51
7	44.6600	0.10	1176.	4.	2.0274	40.50	X		8.51
8	44.8025	0.06	388.	4.	2.0263	13.36	. X		0.91
9	47.2150	0.12	729.	6.	1.9235	25.09	X		8.91
10	47.3775	0.06	303.	6.	1.9220	10.42	X		1.48
11	56.0825	0.28	26.	2.	1.6386	0.90	ХХ		3.89
12	58.0675	0.12 .	10.	2.	1.5872	0.35	X ·X	Bt	0.91
13	61.9900	0.12	10.	2.	1.4958	0.35	X X		0.83
14	62.3625	0.06	24.	з.	1.487B	0.83	X		1.00
15	65.0025	0.10	894.	4.	1.4336	30.77	Х		8.51
16	65.2100	0.08 ¢	404.	4.	1.4331	13.91	X		3.02
17	69.0300	0.08	174.	4.	1.3595	6.00	Х		1.38
18	69.2350	0.08	90.	4.	1.3593	3.11	Х		0.81
19	71.5300	0.04	46.	4.	1.3180	1.59	Χ.		5.37
20	76.3050	0.24	16.	4.	1.2469	0.55	ХХ		1.62
21	78.1250	0.10	630.	5.	1.2224	21.69	X		6.76
22	78.3850	0.10	276.	5.	1.2220	9.49	, Х		3.72
23	82.3425	0.12	310.	5.	1.1701	10.66	Ϋ́Χ		6.92
24	82.6025	0.10	146.	5.	1.1700	5.04	X		3.63
25	87.9500	0.12	10.	4.	1.1094	0.35	X		0.98
26	94.8725	0.12 🗠	14.	2.	1.0459	0.47	X		0.98
27	98.9900	. 0.08	56.	2.	1.0131	1.94	X		1.02
28	99.3300	0.16	26.	2.	1.0130	0.90	Х		1.38
29	106.6100	0.12	72.	4.	0.9607	2.49	X		1.86
30	106.9875	0.12	32.	* 3.	0.9607	1.12	X		1.29
31	111.8400	0.12	303.	4.	0.9300	10.42	Х		5.62
32	112.2750	0.10	146.	4.	0.9299	5.04	Х		2.45
33	113.9075	0.20	18.	- 5.	0.9190	0.64	Х		1.35
34	116.3800	0.12	279.	6.	0.9064	9.60	X		5.37
35	116.8050	0.06	123.	5.	0.9066	4.40	Х		1.41
36	116.8900)	y 0.16	83.	5.	0.9062	2.85	X	-	0.76
37	127.2875	0.24	71.	4.	0.8597	2.43	Хſ		2.82
38	107.9300	0.16	37.	5.	0.8594	1.28	í x		0,98

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9:36

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9-SEP-86

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Table IIA , continued b) growth rate 1.0 cm/h

10:40 <sub>P</sub> DETECTED PEAKS FILE 9-SEF-86 \_\_\_\_\_ \_ \_ \_ \_ \_ \_ \_ \_ \_ \_ \_ \_ \_ \_ \_ \_ PW1700 Automated Powder Diffractometer System 1 : Listed DI file name D14.DI Rausdata file name : D14.RD Saple identification : d14 9-SEP-86 9:34 Measurement date/time : Generator settings : 40 FV, 20 mA Cu alphal.2 wavelengths : 1.54060. 1.54439 Ang Step size, sample time : 0.020 deg, 0.50 s. 25.00 s/deg Monochromator used Yes Automatic (Specimen length: 12.5 mm) Divergence slit : \*Analysis program number : 20 10.010 - 137.990 deg Peak angle range : 0.82513 - 8.8294 Ang Range in D spacing Peak position criterion : Top of smoothed data Cryst peak width range : 0.00 - 2.00 deg Minim peak significance : 0.75 Number of peaks in file : 35 (Alphal: 24, Amorphous: 0) 4984. c.t.s. 9968.7 cps Maximum intensity : Peak Backg D spac (cts) (cts) (Ang) Pesk Tip width I/Imax Type Angle Sign (deg) (2) A1 A2 Ot no (deg) (cts) ----\_\_\_\_\_ \_\_\_\_\_ \_\_\_\_\_ ---------3. 8.5380 10.3525 10. 0.79 0.12 0.21 1 X Х 2 28.4000 0.08 42. 1. 3.1401 0.85 х Х 1.17 34.5250 1. 2.3458 0.85 Bt 3 0.12 42. 3.02 42. 36.7275 4. 2.4450
 5. 2.3433
 5. 2.3416 4 0.12 0.85 2.04 Х X 4984. 5 38.3825 0,12 100.00 X 23.99 5. 38.5100 0.04 1875. 37.62 Х 3.98 6 5. 2.1777 7 41.4300 0.12 0.21 X X 0.78 10. 121. 0.20 4. 2.0288 44.6275 R 2.43 X Χ 8.51 980. 6. 1.9237° 9 47.2100 0.12 19.66 Х 12.02 6. 1.9227 10 47.3600 0.06 454. 9.10 Х 1.02 2. 1.6391 3. 1.5179 10. 1.00 56.0625 0.24 0.21 Х 11 Х 60.9925 0.12 10. 0.21 Х Х 2.29 12 357. 1.4356 13 64.9025 0.08 7.17 Х 1.26 4. 538. 14 64.9950 0.08 4. 1.4337 10.80 Х X 2.34 202. . 1.02 15 65.2025 0.06 4. 1.4332 4.05 х 69.0000 0.10 404. 2.72 16 4. 1.3600 8.11 Х 69.2325 207. 1.29 17 0.08 4. 1.3593 4.16 Х 14. 1.2477 18 76.2500 4. 1.02 0.16 0.29 X X 1.2225 х 19 78.1125 0.12 392. 4. 7.87 7.24 1.2222 1.1703 20 78.3650 0.10 172. 3.44 2.40 4. X 21 420. 5. 8.91 82.3300 0.12 Х 8.43 22 82.5850 0.08 193. 5. 1.1702 3.88 Х 1.82 94.9950 2. 23 0.96 10. 1.0448 0.21 X Х 1.23 98.9850 0.12 10. 2. 1.0131 24 0.21 х 0.93 ,119. з. 0.9609 0.89 25 106.5750 0.14 2.38 X 26 106.9625 53. 0.9608 1.32 0.16 з. 1.07 X 10. 27 110.6425 0.9367 0.21 Х 0.06 4. 1.20 111.8275 0.9301 28 0.12 4. 3.66 X 3.72 4 86. 29 112.2025 0.12 4. 0.9302 1.74 Х 2.24 / , 25. ٩ 30 113.8975 0.16 4. 0.9190 0.50 X 0.98 10. 0.9183 0.21 31 114.4625 0.40 4. Х 0.78 10-116.3550 185. 3.71 32 4. 0.9066 5.25 0.14 X 33 116.9400 0.10 92. 4. 0.9064 1.35 х 1.26

35., `

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127.3050

137.2450

0.28

0.14

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35

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X

2.29

5.75

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able IIA continued ) growth rate 5.0 cm/h TECTED PEAKS FILE 9-SEP-86 11:4 11700 Automated Powder Diffractometer System 1 		11:40	P-86 	-SEP	9-		ı.			1		tinued 5.0 cm/h	co h rate	le ÎlA rowt	Tabl
) growth rate 5.0 cm/h THECTED PEAKS FILE 9-SEP-86 11:4 11700 Automated Powder Diffractometer System 1 sted DI file name : DU0:[300,2]D15.D1 w data file name : DU0:[300,2]D15.D1 imple identification : d15 issurement date/time : 9-SEP-86 10:38 merstor settings : 40 kV, 20 mA i alphal, 2 wavelengths : 1.54060, 1.54439 Ang iep size, sample time : 0.020 deg, 0.50 s, 25.00 s/deg inochromator used : Yes vergence slit : Automatic (Specimen length: 12.5 mm) islysis program number : 20 isk angle range : 10.010 - 137.990 deg inge in D spacing : 0.82513 - B.8294 Ang ik position criterion : Top of smoothed data yst peak width range : 0.00 - 2.00 deg nimer of peaks in file : 35 (Alphal: 23, Amorphous: 0) minum intensity : 1429. cts, 2857.7 cps		11:40	P-86 	-SEP	9-					(		5.0 cm/h	h rat	rowt	<u>c) gr</u>
9-SEP-86 11:4PITOO Automated Powder Diffractometer System 1Sted DI file nameDU0:E300,23D15.DIW data file nameDU0:E300,23D15.RDImple identificationd15Ossurement date/time9-SEP-86 10:38Imple identification1 alphal,2 wavelengths1 alphal,2 wavelengths2 alphal,2 wavelengths </td <td></td> <td>11:40</td> <td>P-86</td> <td>-SEP</td> <td>9-</td> <td></td> <td></td> <td></td> <td></td> <td></td> <td></td> <td>5.0 cm/m</td> <td>niav</td> <td></td> <td>y yı</td>		11:40	P-86	-SEP	9-							5.0 cm/m	niav		y yı
TECTED PEAKS FILE 9-SEP-86 11:4 PI700 Automated Powder Diffractometer System 1 sted DI file name : DU0:E300,2DD5.DI imple identification : d15 surement date/time : 9-SEP-86 10:38 nerator settings : 40 kV, 20 mA i alphal,2 wavelengths : 1.54060, 1.54439 Ang ep size, sample time : 0.020 deg, 0.50 s, 25.00 s/deg inochromator used : Yes vergence slit : Automatic (Specimen length: 12.5 mm) ialysis program number : 20 sak angle range : 10.010 - 137.990 deg inge in D spacing : 0.82513 - 8.8294 Ang ak position criterion : Top of smoothed data yst peak width range : 0.075 nim peak significance : 0.75 meter of peaks in file : 35 (Alpha': 23, Amorphous: 0) ximum intensity : 1429. cts, 2857.7 cps		11:40	P-86 	-SEP	9-										
<pre>/// Automated Powder Diffractometer System 1 // Automated Powder Diffractometer System 1 // Automate Powder Diffractometer System Powder Powder Powder Powder Diffractometer Powder Po</pre>	,	·,	•									FILE	PEAK	CTED	DETE
sted BI file name : DU0:C300,23D15.BI w data file name : DU0:C300,23D15.BI mple identification : d15 asurement date/time : 9-SEP-86 10:38 inerator settings : 40 kV, 20 mA i alphal,2 wavelengths : 1.54060, 1.54439 Ang ep size, sample time : 0.020 deg, 0.50 s, 25.00 s/deg inochromator used : Yes vergence slit : Automatic (Specimen length: 12.5 mm) alysis program number : 20 sk angle range : 10.010 - 137.990 deg inge/in D spacing : 0.82513 - 8.8294 Ang ak position criterion : Top of smoothed data yst peak width range : 0.00 - 2.00 deg intm peak significance : 0.75 imber of peaks in file : 35 (Alphal: 23, Amorphous: 0) ximum intensity : 1429. cts, 2857.7 cps	7	<i>د</i>					ι.	tem	s Sy	tomete	Diffrac	ed Powder	toma	00 AU	PW17(
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<pre>imple identification : d15 issurement date/time : 9-SEP-86 10:38 inerator settings : 40 kV, 20 mA i alphal,2 wavelengths : 1.54060, 1.54439 Ang isp size, sample time : 0.020 deg, 0.50 s, 25.00 s/deg inochromator used : Yes vergence slit : Automatic (Specimen length: 12.5 mm) islysis program number : 20 isk angle range : 10.010 - 137.990 deg inge/in D spacing : 0.82513 - 8.8294 Ang isk position criterion : Top of smoothed data iyst peak width range : 0.00 - 2.00 deg nim peak significance : 0.75 imber of peaks in file : 35 (Alphal: 23, Amorphous: 0) iximum intensity : 1429. cts, 2857.7 cps</pre>	,	~~ ~~ y						RD	1015	[300,2]	: DUO:	name	file	data	Raw d
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mber of peaks in file : 35 (Alphal: 23, Amorphous: 0) Eximum intensity : 1429. cts, 2857.7 cps								~~ "	v ue	r = ⊶.∎V 5	: 0.7!	nificance	in wii ik esi	v pe: m ne:	ury∋⊺ Minin
mum intensity : 1429. cts, 2857.7 cps				0)	: (	phous	Amori	23,	r.:	(Alpha	: 35	s in file	peal	er of	Numbe
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ak Angle Tip width Peak Backy Dispac I/Imax Type Sign	<b>F</b> 1	51311		Type	. 1	Imax	I/]	spac	D	Backa	Peak	Tip width	ele	Ar	Peak
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2 28.3500 0.20 404. 4. 3.1456 28.28 X X 21.38	8	21.38		X	<b>X</b> آ	. 28 🖤	28.	456	з.	4.	404.	0.20	3500	28.	2
3 34.4925 0.10 14. 3. 2.3479 0.96 Bt 0.87	7	0.87	Bt			. 96	0.	3479	2.	3.	14.	0.10	4925	34.	3
4 36.6950 0.24 10. 3. 2.4471 0.72 X X 0.79	ឆ្ ។	0.79		X	X	.72	100	471	2.	З.	10.	0.24	6950	36.	4
☆ 35,3550 0,16 1429, 3, 2,3449 100,00 X X 12,91 6 44,5875 0,√14 81?, 5, 2,0306 ≅6,85 X 12,91	ь Г	15 20		×	Ŷ	-00 -85	56	308	2.	ರ ಕ್ಷ	1429. 817	0.16	3000	30. 44	с А
7 44.7400 0.06 388. 5. 2.0290 27.16 X 0.91	i	0.91		×	^	.16	27.	290	2.	5.	388.	0.06	7400	44.	7
8 47.1750 0.12 185. 4. 1.9250 12.94 X X 3.72	2	3.72		X	х	.94	12	250	1.	4.	185.	0.12	1750	.47	8
9 55.9850 0.10 112. 3. 1.6412 7.86 Ot 1.82	2	1.82	Ot			.86	7.	412	1.	З.	112.	0.10	9850	55.	9
0 57.8900 0.96 10. 2. 1.4383 0.72 Bt 0.95	5	0.95	Bt			.72	0	1383	1.	2.	10.	0.96	8900	57.	10
1 61.9950 0.12 10. 2. 1.4957 0.72 X X 0.76	5	0.76		X	X	.72	- 0.	1957	1.	2.	10.	0.12	9950	61.	11
は 04.70/0 V.12 - 00/。 つい 1.4343 - 38.78 X - 9.12 3 - 65.1650 - 0.08 - 256 3. 1.4339 - 17.92 - X - 1.43	د 2	7.12		Y	~	. 20	30.	343	1	ະ. ເ	00/. 254	0-08	70/3 1650	64. 65	13
4 66.8575 0.06 14. 3. 1.3983 1.01 X 2.14	4	2.14		^	x	.01	1	3983	1.	3,	14.	0.06	8575	66.	14
5 67.4300 0.12 ° 10. 3. 1.3878 0.72 X X 0.79	9	0.79		x	X	.72	Q.	878	ī.	3.	10.	0.12	4300	67.	15
6 69.0025 0.20 22. 3. 1.3599 1.55 X X 1.35	5	1.35		X	х	.55	1.	3599	1.	́з.	22.	0.20	0025	69.	16
7 76.2400 0.08 52. 4. 1.2478 3.63 X 0.78	3	0.78			X	.63	3.	478	1.	4.	52.	0.08	2400	76.	17
8 78.1000 0.14 557. 4. 1.2227 38.98 X 13.49	9 n	13.49		~	Х	.98	38.	227	1.	4.	557.	0.14	1000	78.	18
עריפיים אינייער פריפיים אינייער פריפיים אינייער אינייער פריפיים אינייער אינייער אינייער אינייער אינייער אינייע אינייער אינייער פריפייער אינייער אינייע	, D	5.50		^	Y	.00	18.	704	1-1		. ئەتان. ١٥٥	0.14	3050	/0. 27	20
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2 87.8775 0.12 50. 2. 1.1101 3.53 X 1.48	B	1.48			х	.53	3.	101	1.	2.	50.	0.12	8775	87.	22
3 94.8025 0.20 22. 2. 1.0464 1.55 X 1.15	5	1.15			X	.55	1.	464	1.	<u>́</u> 2.	22.	0.20	8025	94.	23
4 98.9000 0.10 61. 2. 1.0138 4.26 X 1.33	5	1.35			X	.26	4.	138	1.	2.	61.	0.10	9000	98.	24
5 99.2875 0.16 22. 2. 1.0133 1.55 X 1.10 2 99.2875 0.16 22. 2. 1.0133 1.55 X 1.10	) r.	0.76		X	¥	. シシ クワ		133	۰. ۲۰	2.	10	0.10	1200	99. 105	10 76
7 106.5850 0.16 17. 3. 0.9608 1.18 X 1.12	ī	1.12		^	Ŷ	.18	1.	608	ŏ.	3.	17.	0.16	5850	106.	27
8 111.8025 0.14 225. 4. 0.9302 15.75 X 6.46	5	6.46			X	.75	15	302	ō.	4.	225.	0.14	8025	111.	28
9 112.2325 0.14 102. 4. 0.9302 7.14 X 3.55	5	3.55		х		.14	7.	302	٥.	4.	102.	0.14	2325	112.	29
0 113.8800 0.12 28. 4. 0.9191 1.97 X 1.12	•	1.12			Х	. 97	1.	191	٥.	4.	28.	0.12	8800	113.	30
1 114.3675 0.24 15. 4. 0.9188 1.06 X 0.83	2	0.83		x		.06	1.	188	0.	4.	15.	0.24	3675	114.	31
U 116.3525 0.10 174. 4. 0.9066 12.19 X 2.45	2				X	.19	12.	066	0.	4.	174.	0.10	3525	116.	32
טיי איזענעיי געטעליא איזענער גענענייטער איזענער גענענענער איזענע געער גענענענע גענער גענענענע גענענענענע גענענ איז גענענענענענענענענענענענענענענענענענענענ	3	2.02		× •		- CI - 7									
5 137.2075 0.10 159. 6. 0.8273 11.11 0t 1.29	23555	1.66		X	¥	.92 18	້. 1	15G2	۰. ۸	4.	17	0.40	3675	100	37
	235.22	1.00 1.28 1.29	0t.	X X	x	.92	5. 1. 11	594	0.	4. 3.	17.	0.40	3675 3075	127.	33 34 35

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Table IIA continued d) sand cast

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DETE	CTED PEAK	S FILE	u			'n	9-sep-86	13:30
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PW17	'00 Automa	ted Powder	Diffrac	ctometer	r System	1		
List	ed DI fil	e name	: DUO:	:[300,2]	JD16.DI			U
Raw	data file	riame	: DUO:	:[300,2]	D16.RD			-
Samp	le identi	fication	: dl6	. 4		(	*4	
Heas	urement d	iste/time	: 9-9	SEP-86	11:38	1		
Gerie	rator set	tings	: 40 +	(V, 20	mA			
Cu 3	lphal,2 w	avelengths	: 1.54	1060,	1.54439	Arig	•	
Step	size, sa	mple time	: 0.0	20 deg,	0.50	s, 25.00	s/deg	
Morio	chromator	used	I YES				10 8	
Dive	LJeuce 21	11	: AUTC	Matic	(Specim)	eu Teudru:	12.2 887	
40.21	NELE 0000	ram number	. 20					
Pask	anole ra	nge	: 10.	010 - 1	37.990	deo	•	
Kana		30109	0.8	2513 -	8.8294	Ang		
Peak	DOSITION	criterion	: Top	of smoo	thed da	ta		
Crys	t peak wi	dth range	: 0.00	- 2.00	) deg			-
Hini	m peak si	gnificance	: 0.75	5	-	~		
Numb	er of pea	ks in file	: 31	(Alphal	l: 23,	Amorphous	: 0)	•
Нахі	mum linteri	sity .	: 8	112. cts	<b>,</b> 1624	4.5 cps		
		•						ø
<b>T</b> i a a la	<b>A 3</b> - <sup>1</sup>	<b>Ne</b>	Prove la	<b>N</b>				•
Peak	Angle	Tap width	Peak	Backg	"U spac	I/lmax	Type	Sign
no	(deg)	(deg)	(Cts)	(CLS)	(Ang)	(4)	AI AZ Ut	
· · · · ·	28.0675	0.20	151.		3.1766	18 63	Y Y	8.51
2	32.6575	0.40	10.	3.	2.7398	1.26	Ŷ Ŷ	0.81
<u>,</u> 3	35.8475	0.14	10.	3.	2.5030	1.26	XX	1.15
4	38.0700	0.08	697.	5.	2.3618	85.81	X	3.39
5	- 38.1775	0.12	812.	5.	2.3554	100.00	x x	6.03
6	44.3375	0.10	697.	6.	2.0414	85.81	X -	6.03
7	44.4650	0.08	480.	6.	2.0409	59.05	X	1.32
8	46.9400	0.10 504	142.	5.	1.9341	17.43	хх	2.34
9	54.1075	0.40	10.	з.	1.6936	1.26	хх	0.89
10	55.7925	0.32	72.	2.	1.6464	8.90	хх	8.51
11	64.7325	0.12	154.	з.	1.4389	18.93	X	3.63
12	64.9525	0.12	66.	з.	1.4381	8.08	′ X	0.95
13	68.7575	0.40	12.	з.	1.3642	1.42_	хх	1.95
14	76.0150	0.16	35.	4.	1.2510	4.29	X X	0.85
15	77.8750	0.16	396.	4.	1.2257	48.75	x	10.23
16	78.1325	0.08	219.	4.	1.2253	26.97	X	1.05
17	82.0550	0.08	79.	4.	1.1735	9.75	X	1.38
18	82.4123	0.16	29.	4.	1.1722	3.59	X	0.87
20	B7.0820	00	30.	<u></u>		4.43		1.74
20	99.0000	0.1A	76	7.	1 01/0	1.2/	× ×	1.45
22	99,1050	0.08	44.	5.	1.0147	5.34	Ŷ	ດ ດາ
23	106.3800	0.24	14.	· 3.	0.9621	1.69	x x	1.45
24	111.6900	0,12 ~	83.	4.	0.9309	10.20	X	1.95
25	112.1450	0.20	36.	4.	0.9306	4.43	X	2.09
26	113.7050	0.10	30.	4.	0.9200	3.72	x	0.95
27	116.1950	0.12	58.	5.	0.9074	7.11	X	1.45
28	116.6800	0.16	25.	5.	0.9072	3.08	x	0.89
29	121.6250	0.06	17.	4.	0.9823	2.07	Х	0.78
30	127.1525	0.24	13.	4.	0,48602	1.60	x	0.79
31	137.1925	0.06	213.	6.	0.8274	26.24	Ot	0.93

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b) growth rate 1.0 cm/h

Figure IIA continued



110.0

20°

120.0

130.0

140.0

1.62 1.28 0.98 0.72 0.50

0.32 0.18 0.08

0.02

1.62 1.28 0.98 0.72 0.50 0.32 0.18 0.08 9.02

80.0

90.0

Intensity



100.0

Figure IIA continued

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d) sand cast

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Figure IIA continued ن**،** ک

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## ELECTRICAL CONDUCTIVITY DATA OBTAINED FROM TWO-PHASE ALLOYS DURING THE SOLUTION TREATMENT

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Appendix III

	Alloy		Cond	luctivit	:y (pct.	IACS) a	at Solu	tion Tr	eatme	ntTim	e (h)	
	Name †	0	ł	1	- 1 <del>1</del>	2	3	5	8	12 ,	16	24
	12A	49.8	54.0	56.8	57.5	54.7	51.0	49.3	48.5	49.3	49.5	50.2
	13A	45.2	51.5	53.3	53.8	52.0	48.8	48.1	47.3	47.4	48.0	48.4
	14A	44.0	48.4	52.3	52.9	51.9	48.6	47.4	47,0	47.3	47.8	47.7
	15A	41.7	46.4	50.8	51.3	50.3	47.9	46.7	45.3	46.4	47.2	46.9
	16A	40.8	43.6	47.3	47.8	47.4	45.8	45.9	44.3	45.7	45.7	45 6
	17A	39.2	40.0	40.6	40.1	40.4	40.6	41.4	42.0	43.3	43.2	43.2
	12B	50.5	54.8	58.0	58.5	56 0	51.2	49.3	49.5	49.5	49.3	50.1
	13B	47.4	51.7	52.9	54.6	53.3	48.2	47.3	46.9	47.1	47.2	47.4
	14B	46.6	51.0	53.4	53.6	52.8	48.2	47.2	47.0	47.0	47.3	47.3
	15B	46.3	49.6	52.0	52.3	518	49.2	46.9	46.6	46.4	46.6	46 6
	16B	44.2	46.0	49.6	50.0	49.6	46.9	46.0	45.6	45.8	45 7	45.4
	17B	43.5	44.8	45.Q`	45.3	45.2	45.0	45.3	44.5	44.0	44.1	44.2
	6M,1	47.5	53.6	54.5	55.8	54.3	49.0	47.0	47 <sub>,</sub> 1	47.3	47.4	47.3
	6M2	44.1	50.7	51.6	52.1	50.8	46.6	46.0	46.2	46.4	46.6	47.0
	6M3	40.3	46.3	48.1	47.4	46 4	43.8	43.7	44.1	44.5	44 6	44 7
	6M4	38.2	42.5	44.1	45.4	43.9	40.9	42.0	42.4	42.5	43.0	43.3
æ,	6M5	35.0	37.8	39.1	39.6	38.9	38.0	38.5	39.4	40.4	40.7	41.0
	6M6	32.7	33.5	34.6	34.9	34.5	34.7	37.8	38.8	39.6	40.0	40.4
	6M1B	47.8	55.8	-56.3	57.4	55.9	49 4	46.8	46.9	47.3	47.3	47.3
	6M2B	44.1	56 <i>.</i> 7	56.9	57.8	56.4	50.7	49.4	48.2	48.9	49.0	48.7
	<b>6</b> M3B	42.8	51.6	52.1	52.6	51.5	47.3	45.6	45.2	45.8	45.8	45.8
	6M4B	42.5	47.3	48.0	48.7	47.5	44.5	43.5	42.4	44.0	43.7	44.0
,	6M5B	42.3	45.0	46.1	46.8	46.1	45.1	44.1	42.8	43.0	43.1	43.0
	6М6В	42 <sup>.2</sup>	42.8	43.8	44.7	44.4	43.9	43.6	42.8	42.8	42.7	42.4

† See Table 3.4 for chemical composition

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Appendix III. continued

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Alloy	Conductivity (pct. IACS) at Solution Treatment Time (h)										
†	0	$\frac{1}{2}$	1	11	2	3	5	8	12	16	24
7M1	45.2	50.1	53.8	54.6	52.4	45.9	44.9	44.3	45.1	44.9	45.0
7M2	42.1	47.3	50.1	50.4	49.4	42.8	42.8	42.1	43.6	43.б	43.8
7M3	38.8	42.7	46.5	46.3	44.9	40.9	41.1	41.1	42.5	42.3	42.3
7M4	35.7	38.4	40.9	41.3	39.7	36.0	38.4	37.8	39.8	40.0	40.2
7M5	33.6	35.2	36.0	37.8	36.0 <sup>、</sup>	34.6	36.1	36.1	37.7	38.0	38.0
7M6	32.1	33.5	34.4	33.6	33.9	35.2	35.2	35.7	36.3	36.0	36.3
7M1B	45.5	51.6	54.0	54.1	52.7	45.2	45.2	44.8	45.0	45.0	45.0
7M2B	42.6	47.8	51.5	51.9	50.9	43.8	44.0	42.4	43.4	43.6	43.6
7Ӎ3В	41.3	45.3	47.5	48.0	46.4	41.7	41.8	41.8	42.2	41.9	42_3
7M4B	40.4	43.7	45.8	46.6	44.8	40.2	41.1	40.3	40.7	41.1	41.0
7M5B	40.2	41.7	43.0	44.4	42.8	40.0	40.4 <sup>°</sup>	40.0	40.4	40.1	40.0
7M6B	40.0	40.7	42.0	40.6	39.4	39.8	39.2	39.5	39.2	39.5	39.2
8M1	45.2	48.8	48.9	49.1	50.0	45.6	45.1	45.1	¢45.2	45.3	45.2
8M2	41.6	45.6	46.0	46.4	46.6	42.8	42.3	42.4	_43.1	43.2	43.0
8M3	38.6	42.0	42.5	42.8	43.0	40.4	40.0	40.4	40.8	41.0	40.8
8M4	36.1	40.0	40.4	41.4	40.4	38.7	37.6	38.8	39.2	39.8	39.8
8M5	31.6	32.3	32.6	32.8	31.8	29.8	30.4	30.4	31.8	31.9	31.7
8M6	30.1	32.4	32.6	32.8	31.3	31.7	32.1	32.2	33.7	33.5	33.6
8M1B	45.8	51.0	51.2	52.5	52.0	47.2	46.6	46.4	46.3	46.4	46.4
8M2B	43.7	48.7	48.6	48.6	48.5	45.5	45.1	44.8	44.5	44.4	44.6
8M3B	40.3	43.7	43.6	44.0	43.8	41.4	41.0	41.1	41.6	41.6	41.5
8M4B	39.4	42.5	42.8	<sup>,</sup> 43.0	43.3	41.0	40.8	41.0	40.8	40.9	40.8
8M5B	35.7	37.5	37.9	38.6	38.5	36.4	36.8	37.0	37.8	37.5	37.5
8M6B	33.6	33.8	33.7	34.4	34.5	33.6	34.2	34.1	34.7	34.6	34.6

† See Table 3.4 for chemical composition

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APPENDIX IV

# COMPARISON OF ELECTRICAL CONDUCTIVITY AND RESIDUAL RESISTIVITY DATA

The measured room temperature conductivity values (pct. IACS) are converted to resistivities using the following relationship:

$$\rho(\mu\Omega cm) = 1.72.4 \left( \frac{1}{\sigma(\% IACS)} \right)$$
(IVA.1)

Figure 4.3 (see section 4.3.1) is then replotted by replacing conductivity as resistivity in Figure IVA.1. The resistivity of both quenched and air cooled alloys increases linearly with increasing silicon level up to 1.30 wt. pct. The linearity of the curve shows that room temperature resistivity behaviour of Al-Si solid solution alloys obeys Nordheim's Rule (see section 2.3.1). The rate of the increase of resistivity calculated from the slope of the curve is 0.70  $\mu\Omega$ cm per wt. pct. Si as obtained from the conductivity curve presented in Figure 4.3.

**Resistivity of solid solution alloys at room temperature is given as:** 

 $\rho_{T, 297K} = \rho_{th, 297K} + \rho_{O, 297K}$ (IVA.2)

since  $p_{0,297K} = p_{4.2K}$ 

Thermal resisitivity at room.temperature is then:

 $\rho_{th, 297K} = \rho_{T, 297K} - \rho_{4.2K}$ (IVA.3)

and the ratio of impurity and thermal resistivities becomes

$$\frac{\rho_{4\,2K}}{\rho_{th,\,297K}} = \frac{\rho_{4\,2K}}{\rho_{T,\,297K} - \rho_{4\,2K}}$$
(IVA.4)

$$\frac{\rho_{4.2K}}{\rho_{T, 297K} - \rho_{4.2K}} = \frac{1}{\left(\frac{\rho_{T, 297K}}{\rho_{4.2K}} - 1\right)}$$
 (IVA.5)

where

$$\frac{T.297K}{\rho_{4.2K}} = RRR$$
 (residual resistivity ratio)

In Figure IVA.2, 1000/(RRR-1) is plotted against silicon concentration. It also changes linearly with Si level as does the room temperature resistivity data presented in Figure IVA.1. This is clear evidence of similar resistivity behaviour of AI-Si solid solution alloys at room and low temperature (4.2K)

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as expected, eventhough Figure 4.3 (conductivity) and Figure 4.4 (RRR) in chapter 4 indicate a difference in behaviour (e.g. non-linear relation between RRR and Si concentration).

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Figure IVA.1: Room temperature electrical resistivity of Al-Si and Al-Si-Sr solid solution alloys.



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