DEFORMATION AND FRACTURE BEHAVIOR OF EUTECTIC ALUMINUM-SILICON CASTING ALLOYS

by

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Abstract

The effect of silicon particle size and morphology on the deformation and fracture behavior of a binary aluminum silicon casting alloy was studied. Castings of eutectic composition were solidified either slowly without modifiers, or quickly with strontium modification to produce two different as-cast microstructures. Solution treatment of the castings was performed for various lengths of time at 540°C to further differentiate the structures. Quantitative image analysis was used to describe the size and shape of the silicon particles. Samples were tested in both tension and compression.

Particle sizes were in a region where, neither continuum plasticity nor dislocation based models can fully predict the effect of the silicon phase on the deformation behavior of the alloy. The deformation and fracture behavior were dependent on the size and shape of the silicon particles. Damage, in the form of particle cracking, degraded the strength of the alloy and reduced ductility. This was compared to a simple analytical model which describes the effect of damage on flow in a composite material.
Table of Contents

ABSTRACT ........................................................................................................ ii

TABLE OF CONTENTS .................................................................................. iii

LIST OF TABLES .............................................................................................. v

LIST OF FIGURES .......................................................................................... vi

ACKNOWLEDGMENTS .................................................................................... x

1. INTRODUCTION ......................................................................................... 1

2. LITERATURE REVIEW ................................................................................ 4
   2.1 AL-Si CASTING ALLOYS ....................................................................... 4
       2.1.1 Effect of Solidification Rate on Microstructure ....................... 6
       2.1.2 Eutectic Modification .................................................................. 7
       2.1.3 Heat Treatments .......................................................................... 10
       2.1.4 Mechanical Properties ............................................................... 11
   2.2 STRENGTHENING MECHANISMS IN TWO PHASE MATERIALS .......... 14
       2.2.1 Strengthening Due to Large Second Phase Particles .................. 15
       2.2.2 Strengthening Due to Small Second Phase Particles ................. 18
       2.2.3 Particles in Between Ranges ...................................................... 26
   2.3 DAMAGE IN TWO PHASE MATERIALS ............................................... 27
       2.3.1 Types of Damage ....................................................................... 28
       2.3.2 Effect on Flow ............................................................................ 30
       2.3.3 Modeling of Damage ................................................................. 31
   2.4 SUMMARY ............................................................................................ 32

3. SCOPE AND OBJECTIVES ......................................................................... 33

4. EXPERIMENTAL METHODS ..................................................................... 35
   4.1 CASTING ............................................................................................. 35
       4.1.1 Alloy Preparation ....................................................................... 35
       4.1.2 Melting of the Alloy ..................................................................... 37
       4.1.3 Slow Casting Technique .............................................................. 39
       4.1.4 Fast Casting Technique ............................................................... 40
   4.2 SPECIMEN PREPARATION .................................................................. 41
       4.2.1 Sectioning of the Casting ............................................................. 41
       4.2.2 Specimen Dimensions ............................................................... 43
   4.3 HEAT TREATMENT OF THE SPECIMENS ....................................... 44
   4.4 METALLOGRAPHY ............................................................................... 45
       4.4.1 Grinding and Polishing Procedure .............................................. 46
       4.4.2 Etching ....................................................................................... 47
LIST OF TABLES

TABLE 2-1 Properties of aluminum, silicon and aluminum-silicon alloys ........14
TABLE 4-1 Chemical composition of the master alloys used to fabricate the
    castings. All values listed are in wt.%. ..............................................36
TABLE 4-2 Heat treatment schedule for 6 fast or slow castings resulting in 18
    pairs of compression and tensile specimens ..................................45
TABLE 5-1 Image analysis results ..........................................................62
TABLE 5-2 Average engineering properties determined from tensile tests ....73
TABLE A-1 Individual results of image analysis ......................................129
TABLE A-2 Porosity measurements for each casting ...............................130
TABLE A-3 Individual results of tensile testing .....................................131
List of Figures

FIGURE 2-1 TYPICAL MICROSTRUCTURE OF AN AS-CAST EUTECTIC AL-SI ALLOY SHOWING THE EUTECTIC ALUMINUM (WHITE) AND SILICON (DARK) PHASES. ...............................5
FIGURE 2-2 MICROSTRUCTURES OF AN ALUMINUM-5% SILICON ALLOY (A) SAND CAST (B) PERMANENT MOLD CAST AND (C) DIE CAST. THE DENDRITE CELL SIZE AND CONSTITUENT PARTICLE SIZE DECREASE WITH INCREASING COOLING RATE (I.E. FROM A TO C). (x500) ..........................................................7
FIGURE 2-3 OPTICAL PHOTOGRAPHS SHOWING (A) UNMODIFIED AND (B) MODIFIED EUTECTIC SILICON PHASE. .................................................................8
FIGURE 2-4 SEM PHOTOGRAPHS SHOWING (A) UNMODIFIED AND (B) MODIFIED EUTECTIC SILICON PHASE. .................................................................8
FIGURE 2-5 SCHEMATIC DIAGRAM SHOWING THE BREAK-UP, SPHERODIZATION AND COARSENING OF THE EUTECTIC SILICON PHASE IN (A) UNMODIFIED AND (B) MODIFIED AL-SI ALLOYS. .................................................................11
FIGURE 2-6 DENDRITE CELL SIZE AS A FUNCTION OF SOLIDIFICATION RATE FOR VARIOUS ALUMINUM ALLOYS. .................................................................12
FIGURE 2-7 ULTIMATE TENSILE STRENGTH, YIELD STRENGTH AND PERCENT ELONGATION TO FAILURE AS A FUNCTION OF DENDRITE CELL SIZE FOR AN A356-T62 ALLOY. .................................12
FIGURE 2-8 AXIAL STRESS DISTRIBUTION ALONG FIBER LENGTH WHEN FIBER LENGTH IS (A) EQUAL TO AND (B) GREATER THAN CRITICAL LENGTH FOR REINFORCEMENT. .................................................................17
FIGURE 2-9 STRENGTHENING DUE TO LOAD TRANSFER AS A FUNCTION OF ASPECT RATIO AND VOLUME FRACTION OF ELLIPSOIDAL REINFORCING PARTICLES .................................................................17
FIGURE 2-10 SCHEMATIC OF OROWAN DISLOCATION LOOPING AROUND PARTICLES. .................................................................21
FIGURE 2-11 SCHEMATIC SHOWING THE GENERATION OF GEOMETRICALLY NECESSARY DISLOCATIONS THROUGH THE INTRODUCTION OF A STRAIN GRADIENT .................................................................21
FIGURE 2-12 GEOMETRICALLY NECESSARY AND STATISTICALLY STORED DISLOCATION DENSITY AS A FUNCTION OF SHEAR STRAIN. .................................................................24
FIGURE 2-13 GEOMETRICALLY NECESSARY PRISMATIC LOOP ARRAYS AT AL2O3 PARTICLES IN A COPPER CRYSTAL. .................................................................24
FIGURE 2-14 SCHEMATIC REPRESENTATION OF THE EFFECT OF PARTICLE SIZE ON THE RELATIVE DISLOCATION STRENGTHENING EFFECT. .................................................................26
FIGURE 2-15 SCHEMATIC SHOWING THE RANGE OF APPLICABILITY FOR CONTINUUM PLASTICITY AND DISLOCATION BASED MODELS AND THE LOCATION OF SOME TYPES OF TWO PHASE MATERIALS .................................................................27
FIGURE 2-16 SCHEMATIC ILLUSTRATION OF VOID GROWTH AND COALESCENCE NEAR STIFF REINFORCEMENTS LEADING TO FINAL FAILURE IN A COMPOSITE MATERIAL .................................................................28
FIGURE 4-1 THE RESISTANCE FURNACE USED IN THE CASTING OF THE ALLOYS ...........................................................................................................38
FIGURE 4-2 SCHEMATIC OF THE GRAPHITE MOLD USED IN THE SLOW CASTING METHOD ..............................................................................................39
FIGURE 4-3 SCHEMATIC OF THE FAST CASTING TECHNIQUE .................................................................................................................................41
FIGURE 4-4 LONGITUDINAL SECTION TAKEN FROM EACH CASTING FOR EXAMINATION PURPOSES ...........................................................................................................42
FIGURE 4-5 SCHEMATIC SHOWING HOW AN ACCEPTED CASTING WAS SECTIONED FOR TAKING MECHANICAL TEST SPECIMENS ...........................................................................................................43
FIGURE 5-20 COMPRESSIVE STRESS-STRAIN CURVES FOR SLOW CAST, UNMODIFIED MATERIAL .......................................................... 76
FIGURE 5-21 COMPRESSIVE STRESS-STRAIN CURVES FOR FAST CAST, MODIFIED MATERIAL 76
FIGURE 5-22 TENSILE VERSUS COMPRESSIVE STRESS STRAIN CURVES FOR A SLOW CAST, 1 HOUR SPECIMEN .......................................................... 78
FIGURE 5-23 TENSILE VERSUS COMPRESSIVE STRESS STRAIN CURVES FOR A SLOW CAST, 72 HOUR SPECIMEN .......................................................... 78
FIGURE 5-24 TENSILE VERSUS COMPRESSIVE STRESS STRAIN CURVES FOR A FAST CAST, 1 HOUR SPECIMEN .......................................................... 79
FIGURE 5-25 TENSILE VERSUS COMPRESSIVE STRESS STRAIN CURVES FOR A FAST CAST, 72 HOUR SPECIMEN .......................................................... 79
FIGURE 5-26 DIFFERENCE BETWEEN COMPRESSIVE AND TENSILE TRUE STRESS AS A FUNCTION OF TRUE STRAIN IN SLOW CAST, 1 HOUR SAMPLES ................. 80
FIGURE 5-27 DIFFERENCE BETWEEN COMPRESSIVE AND TENSILE TRUE STRESS AS A FUNCTION OF TRUE STRAIN IN SLOW CAST, 72 HOUR SAMPLES ......................... 80
FIGURE 5-28 DIFFERENCE BETWEEN COMPRESSIVE AND TENSILE TRUE STRESS AS A FUNCTION OF TRUE STRAIN IN FAST CAST, 1 HOUR SAMPLES .................................................. 81
FIGURE 5-29 DIFFERENCE BETWEEN COMPRESSIVE AND TENSILE TRUE STRESS AS A FUNCTION OF TRUE STRAIN IN FAST CAST, 72 HOUR SAMPLES .................................................. 81
FIGURE 5-30 EXAMPLE OF INDENTATION MEASUREMENT INDUCED CRACKING USED TO ESTIMATE SILICON FRACTURE TOUGHNESS ............................................. 84
FIGURE 5-31 CRACKED SILICON PARTICLES FROM SLOW CAST, 1 HOUR COMPRESSION SAMPLE .......................................................... 84
FIGURE 5-32 PROBABILITY OF PARTICLE CRACKING AS A FUNCTION OF ORIENTATION TO THE TENSILE AXIS IN COMPRESSION TESTS, ASSUMING THE TENSILE AXIS IS AT 0° AND THE COMPRESSION AXIS IS AT 90° .................................................. 85
FIGURE 5-33 (A) CROSS-SECTIONAL OPTICAL AND (B) NORMAL SEM PHOTOGRAPHS OF THE FRACTURE SURFACE FOR A SLOW CAST, 1 HOUR SAMPLE .................................................. 87
FIGURE 5-34 (A) CROSS-SECTIONAL OPTICAL AND (B) NORMAL SEM PHOTOGRAPHS OF THE FRACTURE SURFACE FOR A SLOW CAST, 72 HOUR SAMPLE .................................................. 88
FIGURE 5-35 (A) CROSS-SECTIONAL OPTICAL AND (B) NORMAL SEM PHOTOGRAPHS OF THE FRACTURE SURFACE FOR A FAST CAST, 1 HOUR SAMPLE FROM THE BOTTOM OF A CASTING .......................................................... 89
FIGURE 5-36 (A) CROSS-SECTIONAL OPTICAL AND (B) NORMAL SEM PHOTOGRAPHS OF THE FRACTURE SURFACE FOR A FAST CAST, 1 HOUR SAMPLE FROM THE TOP OF A CASTING .......................................................... 90
FIGURE 5-37 (A) CROSS-SECTIONAL OPTICAL AND (B) NORMAL SEM PHOTOGRAPHS OF THE FRACTURE SURFACE FOR A FAST CAST, 72 HOUR SAMPLE FROM THE BOTTOM OF A CASTING .......................................................... 91
FIGURE 5-38 (A) CROSS-SECTIONAL OPTICAL AND (B) NORMAL SEM PHOTOGRAPHS OF THE FRACTURE SURFACE FOR A FAST CAST, 72 HOUR SAMPLE FROM THE TOP OF A CASTING .......................................................... 92
FIGURE 5-39 DUCTILE FRACTURE SURFACE FROM A FAST CAST, 1 HOUR SPECIMEN USED FOR DIMPLE SIZE MEASUREMENTS .......................................................... 94
FIGURE 5-40  Ductile fracture surface from a fast cast, 72 hour specimen used for dimple size measurements. .................................................................94
FIGURE 6-1  As-cast structure in a slow cast, unmodified casting. ..................98
FIGURE 6-2  As-cast structure in a fast cast, modified casting. .......................98
FIGURE 6-3  Work hardening rate as a function of strain in fast cast 1 hour and
72 hour specimens. ..................................................................................102
FIGURE 6-4  Method to determine the particle spacing for the slow cast (A) 1
hour and (B) 72 hour materials..................................................................105
FIGURE 6-5  True stress and work hardening rate as a function of true strain,
showing the satisfaction of the Considere criterion for slow cast 72 hour
specimens in tension and compression......................................................109
FIGURE 6-6  High magnification view of fracture surface in slow cast, 1 hour
specimen.....................................................................................................112
FIGURE 6-7  High magnification view of fracture surface in slow cast, 72 hour
specimen.....................................................................................................112
FIGURE 6-8  High magnification view of fracture surface in fast cast, 1 hour
specimen.....................................................................................................113
FIGURE 6-9  High magnification view of fracture surface in fast cast, 72 hour
specimen.....................................................................................................113
FIGURE 6-10  Some particle cracking observed away from the fracture surface
in a fast cast, 72 hour specimen. .................................................................114
FIGURE 6-11  Experimental fast, 72 hour tensile and compressive stress-strain
curves compared with model curves for zero and full particle cracking.117
FIGURE 6-12  Experimental fast, 1 hour tensile and compressive stress-strain
curves compared with model curves for zero and full particle cracking.117
FIGURE 6-13  Model stress-strain curves for matrix, fully intact and fully
cracked composite and experimental curve for fast cast, 72 hour
compression sample..................................................................................119
FIGURE 6-14  Particle cracking in fast cast, 1 hour specimen near the top of the
casting at (A) regular region of casting and (B) in region of long plates
due to rosette. .........................................................................................123
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1. Introduction

Aluminum casting alloys are important engineering materials. They have excellent specific strength and stiffness (i.e. strength or stiffness to density ratio) and good corrosion resistance, which makes them attractive for many applications. In particular, aluminum castings have experienced tremendous growth in the automotive industry. The weight savings made possible by replacing parts made of steel or cast iron with aluminum castings are important to auto manufacturers. This is due to the pressure to reduce fuel consumption, by means of reduced vehicle weight or otherwise, which has increased over the years.

Traditionally, aluminum castings have been used only in non-load bearing parts such as oil pans or cylinderheads. This is due, in part, to the relatively inconsistent and unpredictable mechanical properties of castings. However, with mounting pressure to reduce vehicle weight further, cast aluminum parts are being considered for safety critical load bearing components. In fact, it has been predicted that the largest sector of growth for aluminum castings in the automotive industry in the next 10 years will be for suspension components.¹ One vehicle, the Lincoln Mark VIII, has already made this switch, by using an A356-T61 casting for their lower rear suspension control arm.² However, failure of these safety-critical types of parts is not acceptable, and for aluminum castings to gain widespread acceptance, better control and consistency of the mechanical properties must be obtained. To this end, it would be useful to have a predictive model, which links measurable processing conditions to the final mechanical
properties of the casting. The bridge between processing conditions and the mechanical properties lies within the microstructure.

It is known that the properties of aluminum casting alloys are dependent on the microstructure. One microstructural component, common to almost all aluminum casting alloys, is silicon. The silicon, which is added to improve the castability (i.e. improve fluidity and feeding characteristics) of aluminum, is present as a brittle second phase in the aluminum alloy matrix. The size and morphology of this phase can be widely varied by changing solidification rates, heat treatments and through the implementation of modifying agents such as sodium or strontium. The ease with which the silicon phase can be altered, makes this system very attractive for studies of the effect of reinforcement size and shape on mechanical properties in two phase materials.

Strengthening in two phase materials is dependent on the shape and size of the stiff, reinforcing phase. Large particles (generally larger than 10 μm), such as those found in metal matrix composites (MMCs) strengthen the material through load sharing with the matrix. The overall effectiveness of a particle in taking up load is dependent on its shape. Elongated particles are much more efficient at load transfer than equiaxed particles. Models to predict this strengthening are grounded in continuum elasticity and plasticity. Small particles (generally smaller than 0.1 μm) such as those found in precipitation hardenable aluminum alloys, on the other hand, increase the matrix strength through dislocation mechanisms, such as Orowan looping or the generation of geometrically necessary dislocations. Strengthening is increased as the spacing between particles decreases.
There exists a range of particle sizes (0.1 to 10 μm) where neither continuum plasticity based models nor dislocation based models accurately predict the strengthening effect of the reinforcement. It is important, therefore, to examine this regime to further the understanding of length scales in two-phase materials.

In the present study, the size and morphology of the silicon phase in a binary aluminum-silicon alloy of eutectic composition will be altered by changing casting conditions, heat treatments and through the use of modifying agents. The samples will then be tested in tension and compression to determine the effects of particle size, morphology and damage on the deformation and fracture behavior of the material.
2. Literature Review

In this chapter, a review of the relevant literature on aluminum-silicon casting alloys, and deformation of two-phase materials will be presented. In particular, processing conditions affecting the silicon phase, strengthening due to second phase particles, and damage of second phase particles will be reviewed.

2.1 Al-Si Casting Alloys

Of all aluminum casting alloys, those which contain silicon as the major alloy addition (designated as 300 and 400 series alloys by the U.S. Aluminum Association) are by far the most common, accounting for 85% to 90% of all cast aluminum parts produced. These alloys are popular because of their excellent castability as a result of the presence of silicon. Silicon is added to foundry alloys (usually between 4% and 20% by weight) to improve their fluidity (the ability of liquid metal to flow readily in a mold and fill thin sections) and feeding characteristics (the ability of liquid metal to be drawn into the interstices between dendrites during casting) and to increase resistance to cracking during casting.\textsuperscript{4,5,6}

When binary Al-Si alloys solidify, the resulting microstructure is a combination of a primary phase, either silicon (in hypereutectic alloys) or aluminum (in hypoeutectic alloys) and/or a eutectic mixture of these two elements. The eutectic, which occurs at approximately 12.0 wt.% silicon, does not solidify in a "normal" lamellar or rod-like fashion because of the anisotropic growth characteristics of silicon, relative to
aluminum. The silicon phase, which nucleates with random orientation, will grow with a random orientation to the growth interface because of its crystallographically preferred growth direction. This random orientation with respect to the growth interface leads to constant termination and nucleation of silicon crystals which are growing in competition with one another. Consequently, the eutectic phase is considered discontinuous or anomalous and is comprised of randomly oriented, acicular plates of essentially pure silicon in a continuous matrix of aluminum (as shown in Figure 2.1), with a low concentration of silicon in solid solution (maximum of 1.65 wt. % at 577°C). These angular plates of silicon are brittle and may reduce the ductility of the alloy. As a result, much work has been done in researching methods to alter the shape and size of the eutectic silicon phase and thereby improve the mechanical properties.

Figure 2-1 Typical microstructure of an as-cast eutectic Al-Si alloy showing the eutectic aluminum (white) and silicon (dark) phases.
2.1.1 Effect of Solidification Rate on Microstructure

The rate at which a casting solidifies is largely determined by the choice of casting process and there are three main techniques that are used to cast aluminum alloys. Sand castings, permanent mold castings and die castings account for approximately 98% of all cast aluminum parts. An important difference among these three processes is the rate of solidification. Sand castings cool at a rate of approximately 0.1-0.5°C/s, permanent mold castings at a rate of 0.3-1.0°C/s and die castings at a rate of 50-500°C/s. Figure 2.2 shows how the choice of process and, consequently, the solidification rate affects the resultant microstructure of an aluminum-5% silicon alloy. Increasing solidification rate results in a more refined microstructure, as seen by comparing the dendrite cell size and Si particle size of the die cast specimen to that of the sand cast specimen in Figure 2.2. The faster growth front velocity associated with high cooling rates, allows less time for diffusion of solute (i.e. silicon). Consequently, the silicon is distributed over shorter distances, resulting in finer particles at higher cooling rates. It is important to note that while changing the solidification rate alters the scale of the eutectic silicon phase, the morphology of the particles remains essentially unchanged.
Figure 2-2 Microstructures of an aluminum-5% silicon alloy (a) sand cast (b) permanent mold cast and (c) die cast. The dendrite cell size and constituent particle size decrease with increasing cooling rate (i.e. from a to c).4 (x500)

2.1.2 Eutectic Modification

Modification is the refinement and change in morphology, from acicular to fibrous, of the eutectic silicon phase by means of chemical additions in aluminum-silicon casting alloys. It has been the subject of extensive research and is reviewed in works by Gruzleski and Closset3, Sigworth9, and Bäckerud, et al.10

The most common chemical modifying agents are sodium and strontium. Additions on the order of 0.01 to 0.1 weight percent of these elements to hypoeutectic and eutectic Al-Si melts can cause dramatic changes in the microstructure as shown in Figure 2.3. The structure is refined but, more importantly, the morphology of the eutectic silicon phase is now more rounded and fibrous than acicular. The SEM photographs of Figure 2.4 show the three dimensional “seaweed” structure in a modified alloy. This
Figure 2-3  Optical photographs showing (a) unmodified and (b) modified eutectic silicon phase.

Figure 2-4  SEM photographs showing (a) unmodified and (b) modified eutectic silicon phase.
change in morphology has been shown to improve mechanical properties in Al-Si alloys, particularly sand castings, where the unmodified structure is very coarse and acicular.\textsuperscript{11,12} However, it has also been observed that additions of modifying agents may also increase the incidence of microporosity.\textsuperscript{3}

The exact mechanism by which the small concentration of modifying agent changes the morphology of the eutectic silicon is not completely understood. However, the Impurity Induced Twinning (IIT) theory put forth by Lu and Hellawell\textsuperscript{13} has recently gained significant acceptance. They observed that the eutectic silicon in modified alloys has a twin density several orders of magnitude higher than found in unmodified alloys. They proposed that atoms of the modifying agent adsorb onto the growth front of the silicon phase and cause twins to form. The intersection of numerous twins increases the surface roughness of the silicon and these imperfections act as nucleation sites for new branches to form.

Higher solidification rates assist the modification process. That is, the faster solidification occurs, the less modifier that is required to produce a fully modified structure. In fact, it is possible to produce a "modified" structure without chemical additions through extremely fast cooling. This is known as quench modification. Optically, a chemically modified microstructure and a quench modified microstructure appear similar. However, electron microscopy reveals that the quench modified structure is similar to the unmodified structure in that the silicon contains very little or no
twinning. It is, in fact, just an exceedingly fine form of the unmodified eutectic caused by extreme solidification rates.\textsuperscript{13}

2.1.3 Heat Treatments

Like wrought alloys, aluminum casting alloys can be considered either heat-treatable or non-heat treatable. Additions of elements such as magnesium or copper, which have decreasing solubility in aluminum with decreasing temperature, and can therefore be used for precipitation hardening, make an alloy heat treatable. Binary Al-Si alloys are not considered heat treatable because only a small amount of silicon is soluble in aluminum (1.65 wt.% maximum) and any silicon that may reprecipitate causes very little hardening.\textsuperscript{5}

The main impetus for heat treating a part is to increase the strength of the casting. However, it has been shown that exposure to solutionizing temperatures (on the order of 540\degree C), particularly for prolonged periods of time, will cause spherodization and eventually coarsening of the eutectic silicon particles.\textsuperscript{6,12,14,15} This effect occurs in both unmodified and modified alloys although, the coarsening is generally faster in modified alloys, due to the finer initial microstructure. The process is shown schematically in Figure 2.5. This alteration of the silicon phase affects the particle size, shape and distribution in the alloy and consequently can alter the mechanical properties. The instability of eutectic phases and particle coarsening has been extensively reviewed by Martin and Doherty.\textsuperscript{16}
2.1.4 Mechanical Properties

A large amount of research has focused on relating the microstructure to the mechanical properties of aluminum casting alloys. In the seminal work of Spear and Gardner, the overall refinement of the microstructure in various aluminum alloys cast over a range of solidification rates was evaluated by measuring the dendrite cell size and was found to decrease with increasing solidification rates, as shown previously in Section 2.1.1. The tensile properties were then related to the measured cell size. For an A356-T2 alloy, it was found that both the ultimate tensile strength and elongation to failure decreased with increasing dendrite cell size (i.e. decreasing cooling rate) as shown in Figures 2.6 and 2.7. To a first approximation, the yield strength of the alloy was unaffected by the dendrite cell size.
Figure 2-6 Dendrite cell size as a function of solidification rate for various aluminum alloys.\textsuperscript{23}

Figure 2-7 Ultimate tensile strength, yield strength and percent elongation to failure as a function of dendrite cell size for an A356-T62 alloy.\textsuperscript{23}
Since 1963, there have been a number of papers examining the relationship of microstructural features to the mechanical properties in aluminum casting alloys, including a recent review of Spear and Gardner's work by Cáceres and Wang. However, with the advent of quantitative image analysis, the focus has generally been on relating the mechanical properties to characteristic features of the silicon phase, such as particle size, aspect ratio, or shape factor. This gives a more localized measure of the refinement of the alloy and allows the role of silicon particles in flow and fracture of the alloy to be more closely examined. Additionally, the shape and size of the eutectic silicon is being varied not only by changing the casting conditions, but through modification and/or heat treating.

The most rigorous of these studies were those performed by Paray and Gruzleski. They varied solidification rates, heat treatment schedules and levels of modification in a 7% Si (356) alloy. Through image analysis, they measured various characteristics (e.g. particle area, diameter, aspect ratio and shape factor) of the silicon phase and then established empirical relationships between these measurements and the mechanical properties of the casting. In general, they found that size-based parameters gave better correlation to the tensile results than shape-based parameters. While these relationships may be useful for the specific alloy, they are empirical in nature and provide little insight into how the silicon phase affects the properties of the alloy.

For an understanding of deformation behavior, it is important to know some of the important properties of the constituent elements in these alloys. These are presented in Table 2.1 for pure silicon, pure (99.5%) aluminum and an aluminum-silicon casting alloy
A413.2 which is of eutectic composition and negligible impurity. Some of the properties have been estimated. For example, the fracture toughness of the A413.2 alloy is estimated to be between 25 and 28 MPa√m but, reported values for A356 are in the 16 to 18 MPa√m range.\textsuperscript{25,26} It is expected that increasing the silicon content will decrease the fracture toughness of an alloy, since the fracture toughness of silicon itself is so low. Considering that A356 has 7% silicon while the A413.2 has 12% silicon, it seems somewhat unrealistic that the A413.2 would have such a high fracture toughness.

| Table 2-1 Properties of aluminum, silicon and aluminum-silicon alloys\textsuperscript{27} |
|---------------------------------|------------------|------------------|------------------|
| Property                        | Cast Aluminum (99.5% Al) | Silicon          | Al-Si Alloy A413.2 (Al-12%Si) |
| Elastic (Young’s) Modulus (GPa) | 69-69.5            | 107-113          | 71-71.5          |
| Shear Modulus (GPa)             | 26.5-26.6          | 42*-44           | 26.5*-26.6       |
| Yield Strength (MPa)            | 10                | -                | -                |
| Tensile Strength (MPa)          | 80-83              | 600*-1000        | 170-200          |
| Fracture Toughness (MPa√m)      | 30-35              | 1*-2             | 25*-28           |

N.B. * indicates that the value has been estimated

2.2 Strengthening Mechanisms in Two Phase Materials

Aluminum-silicon casting alloys are essentially two-phase materials comprised of a hard, brittle phase embedded in a ductile metal matrix. It is therefore relevant to review strengthening mechanisms in two-phase materials.

Metal based two phase materials encompass a wide range of engineering materials. These include precipitation hardenable alloys, dual phase steels, aluminum
casting alloys and metal matrix composites (MMCs). In each of these materials, the second phase is introduced to improve the properties of the parent, be it through a precipitation hardening sequence, injection into a melt or otherwise. Common goals are to increase the strength, stiffness and, in the case of MMCs, the wear resistance beyond that of the parent material. Both precipitation hardening and the addition of a ceramic reinforcing phase can achieve these goals but, the mechanisms by which they are achieved are as different as the size of their constituent second phase. Precipitates are only a fraction of a micron in size while the particulate reinforcement of MMCs is usually greater than 10 μm in size.

2.2.1 Strengthening Due to Large Second Phase Particles

Recently, there has been a large interest in advanced MMCs. In these systems, the matrix is usually aluminum or titanium and the reinforcing phase is typically a hard, brittle ceramic such as alumina or silicon carbide present in volume fractions of 5 to 20%. Reinforcement is usually in the form of discontinuous particles because of the low cost and ease of production but, continuous fibre composites are also produced. Reinforcing particles may range in size from 5 μm to as large as 250 μm but, are typically in the 10 to 30 μm range.28

The presence of strong, stiff ceramic particles gives the composite high strength through load sharing with the matrix. The load is transferred from the ductile matrix to the stiffer particle by means of shear stresses developed at the particle-matrix interface during loading. This process has been successfully modeled by the shear lag model (for
continuous and short fibre reinforcement), Eshelby models and finite element (FEM) models based on continuum mechanics. The strength of the composite as modeled by the shear lag model (for composites with short fibres aligned parallel to the axis of loading) is given by equation 2.1:

\[
\sigma_c = \sigma_f \left(1 - \frac{\ell_c}{2\ell}\right)V_f + \sigma'_m (1 - V_f)
\]

where \(\sigma_c\) is the strength of the composite, \(\sigma_f\) is the strength of the fibre, \(\sigma'_m\) is the stress acting on the matrix phase at the strain to fracture of the fibre, \(\varepsilon_p\), \(V_f\) is the volume fraction of the fibre reinforcement, \(\ell\) is the average length of the fibres and \(\ell_c\) is the critical fibre length required to achieve maximum loading potential in the fibre. It can be seen from equation 2.1 that increasing the length (i.e. size) of the reinforcement results in an increase in the strength of the composite. Figure 2.8 shows how the stress distribution in a fibre oriented to the axis of loading changes with length. An analytical technique, known as the Eshelby method, is also efficient at predicting the load sharing in MMCs. The advantage of Eshelby models is that they can be used for spheres or ellipsoidal particles. However, all of these models are only successful in predicting strengthening in systems where the particles are greater than about 10 \(\mu\)m in size. An excellent review of load transfer and Eshelby models is given by Clyne and Withers.
Figure 2-8  Axial stress distribution along fiber length when fiber length is (a) equal to and (b) greater than critical length for reinforcement.

Figure 2-9  Strengthening due to load transfer as a function of aspect ratio and volume fraction of ellipsoidal reinforcing particles.
The effectiveness of a given volume fraction of reinforcement on the strength and stiffness of the composite is dependent on its shape. Figure 2.9 shows schematically how the ratio of composite to matrix yield strength changes with aspect ratio for various volume fractions of ellipsoidal reinforcing particles. Low aspect ratio particles (e.g. spheres where $a/b=1$) are clearly much less efficient strengtheners than high aspect ratio particles regardless of their orientation to the axis of loading. Increasing the volume fraction of reinforcement also improves the strengthening effect because there more load transfer is involved.

2.2.2 Strengthening Due to Small Second Phase Particles

The small particle size scale includes the precipitation hardenable alloys, such as 2000 or 6000 series wrought aluminum alloys. In these systems, the strengthening precipitates are typically 1 to 100 nm (0.001 to 0.1 μm) in size and are present in small fractions on the order of 1 to 5% by volume. These small particles are generally classified as either shearable or non-shearable and strengthen the material by impeding the motion of dislocations. Shearable precipitates are generally coherent with the matrix and small in size and merely increase the force required to move dislocations through the precipitate (i.e. shear the precipitate). An example of a shearable region of precipitates would be Guinier-Preston (G.P.) zones formed early in an aging process. Much larger or incoherent particles are considered non-shearable because dislocations bypass the particle by moving around it rather than shearing through it. Non-shearable particles induce a
mechanism known as Orowan looping. The interaction of dislocations with particles has been extensively reviewed by Brown and Ham.\textsuperscript{31}

Orowan looping involves the bowing of a dislocation line, under the resolved shear stress on the slip plane, around non-shearable particles separated by a distance, $L$. A rough approximation of the particle separation distance, $L$ can be calculated by equation 2.2\textsuperscript{32}

$$L = \left( \frac{\pi d^2}{4f} \right)^{\frac{1}{2}} \quad \text{eqn. 2.2}$$

which is an approximation for a square array of fibre reinforcement where $d$ is the average fibre diameter and $f$ is the volume fraction of the reinforcing phase. Eventually the dislocation line will wrap all the way around the particles and leave a dislocation loop around the particles, while the dislocation line moves on. The passage of successive dislocation lines will result in more loops being created around the initial loop. Orowan looping is shown schematically in Figure 2.10. The effective strength increase as a result of the stress required for Orowan looping is given by equation 2.3\textsuperscript{32}

$$\Delta\sigma = \alpha \frac{\mu b}{L} \quad \text{eqn. 2.3}$$

where $\Delta\sigma$ is the increase in strength, $\mu$ is the elastic shear modulus of the matrix, $b$ is the Burgers vector of the matrix, $L$ is the mean particle spacing and $\alpha$ is a geometric constant of the order 1 depending on the nature of the dislocations.

The presence of small, non-deformable second phase particles is also responsible for the generation of geometrically necessary dislocations, a concept put forth by
Cottrell showed that to accommodate a strain gradient in any system, dislocations must be generated to satisfy geometric considerations. This is well illustrated by considering a single crystal beam of length, $l$ and thickness, $t$ being bent to a radius of curvature $r$ as shown in Figure 2.11. The top surface of the crystal experiences tensile deformation while the bottom surface of the crystal undergoes compressive deformation. The length of the tensile (upper) surface of the crystal has increased from $l$ to $l + \delta l$ (where $l = \pi \theta$ and $\delta l = \theta / 2$) while the bottom surface has been compressed to $l - \delta l$ due to bending. From this, it can be seen that there is a gradient of strain in the crystal. If the lengths of the upper and lower surfaces of the crystal are then divided by the spacing between atoms in the closest packed direction, $b$ (which is the magnitude of the Burgers vector) the number of atomic planes on each surface can be determined. By doing so, it can be seen that there are more atomic planes on the tensile surface than there are on the compressive surface. The only way to accommodate this difference in number of atomic planes between surfaces is through the introduction of edge dislocations into the crystal. Thus, it can be seen that a strain gradient will generate dislocations to satisfy geometric requirements.
Figure 2-10 Schematic of Orowan dislocation looping around particles.  

Figure 2-11 Schematic showing the generation of geometrically necessary dislocations through the introduction of a strain gradient.
Since second phase particles are usually subject to an elastic misfit strain with the matrix, geometrically necessary dislocations must be present in two phase systems. The Orowan loops left at particles during deformation as described above are a simple example of geometrically necessary dislocations generated due to the gradient in strain between the particle and the matrix. As the strain gradient increases with increasing stress on the system, the number of Orowan loops (i.e. geometrically necessary dislocations) also increases. This is expressed in equation 2.4\(^{35}\) which is used to calculate the density of geometrically necessary dislocations, \(\rho^G\)

\[
\rho^G = \left(\frac{1}{\lambda_G}\right) \frac{4\gamma}{b}
\]

\text{eqn. 2.4}

where \(\gamma\) is the shear strain, \(b\) is the matrix Burgers vector and \(\lambda_G\) is the geometric slip distance. For plates, \(\lambda_G\) is the distance \(L\), usually assumed to be the spacing between plates, while for equiaxed particles it is given by \(r/f\) where \(r\) is the particle radius and \(f\) is the particle volume fraction.

Geometrically necessary dislocations are present in addition to statistically stored dislocations which are generated due to processing and deformation of the alloy. Since the work hardening rate in alloys is directly related to the rate of increase of the dislocation density, the increase in the total dislocation density generation rate due to the generation of geometrically necessary dislocations will result in an increase in work hardening of the material. However, the effect is relatively minor unless, the geometric slip distance, \(\lambda_G\) is very small (i.e. for small particles and high volume fractions). This is shown schematically in Figure 2.12 where short slip distances result in a high density of
geometrically necessary dislocations that dominate over the statistically stored dislocations, particularly at small strains. Fleck, et al.\textsuperscript{36} note that at 10\% strain, the density of geometrically necessary dislocation is dominant over statistically stored dislocations when the slip distance is less than 50 μm in single crystals and less than 20 μm in polycrystals. The data for the density of statistically stored dislocations shown in Figure 5.12 is based on experiments with copper.

It is important to note that generation of dislocations and the resulting work-hardening eventually becomes limited through competition with processes which serve to relieve the stress within and around the reinforcement. These processes may be catastrophic such as particle cracking and interfacial failure (see Section 2.3 for more details) or non-catastrophic. An example of non-catastrophic relaxation would be through dislocation motion and rearrangement such as the punching out of prismatic loops surrounding the reinforcement as shown in Figure 2.13. This effectively redistributes the dislocations around the reinforcement to lower the stress in the particle. Cross slip or secondary slip may also be induced to reduce particle stress resulting in more complicated dislocation networks around the reinforcement. An excellent treatment of these non-catastrophic relaxation mechanisms for both equiaxed and plate-like particles is given by Ashby.\textsuperscript{35}
Figure 2-12  Geometrically necessary and statistically stored dislocation density as a function of shear strain.\textsuperscript{36}

Figure 2-13  Geometrically necessary prismatic loop arrays at Al\textsubscript{2}O\textsubscript{3} particles in a copper crystal.\textsuperscript{35}
Note that both the generation of geometrically necessary dislocations and the stress increase due to Orowan looping are increased with decreasing particle spacing. That is, for a given particle volume fraction, finer particles will be more effective in strengthening the system. For example, if we take two systems of hard, particulate reinforcement in an aluminum matrix, both with 5% reinforcement by volume but, one system with an average particle diameter of 0.1 microns and the other with an average particle diameter of only 0.01 micron, there will be a considerable difference in strengthening. By using equation 2.2, it can be determined that the average particle spacing in a system with 5% reinforcement, 0.1 microns in size will be 0.4 μm while for a 0.01 micron average size, the spacing will be only 0.04 μm. Using equation 2.3 these spacings translate to a strength increase of approximately 18.9 MPa (assuming μ=26,500 MPa and b=0.286 nm for aluminum and α=1) in the larger particle system and 189 MPa for the finer particle system. Similarly, these same two systems have geometric slip distances of 2 microns for the large particles and 0.2 microns for the small particles. The system with the smaller reinforcement will obviously have a greater contribution to strength due to geometrically necessary dislocations. This effect of particle size on the efficiency of strengthening is shown generically in Figure 2.14.
Figure 2-14  Schematic representation of the effect of particle size on the relative dislocation strengthening effect.\textsuperscript{28}

2.2.3 Particles in Between Ranges

As shown above, there are effective theories to explain the strengthening due to the presence of sufficiently large (>10 \( \mu \text{m} \)) or sufficiently small (<0.1 \( \mu \text{m} \)) particles in two phase materials. However, both continuum plasticity and dislocation based models underestimate the strengthening that occurs outside of their range of applicability. Consequently, there exists a range of particle size (as shown in Figure 2.15), from 0.1 to 10 microns where both continuum and dislocation effects may interact and for which no satisfactory theory exists to explain their observed behavior. This range of particle size is of interest and has been identified as an important area of study in a number of articles.\textsuperscript{28,32,37}
2.3 Damage in Two Phase Materials

While the presence of large, non-deforming, second phase particles may enhance the strength of a composite, there is usually an associated drop in ductility due to the accumulation of damage in the material as a result of the reinforcement. Damage is defined by Vedani and Gariboldi\textsuperscript{38} as “the process of initiation of microcracks and cavities which results in a progressive material strength and stiffness deterioration.” In other words, once damage is initiated, voids associated with the damage may quickly grow and coalesce leading to final ductile fracture of the material, as shown schematically in Figure 2.16. The void nucleation can occur in three general ways, as described in the next section.

---

Figure 2-15 Schematic showing the range of applicability for continuum plasticity and dislocation based models and the location of some types of two phase materials.
2.3.1 Types of Damage

There are three basic types of damage in a two phase material; reinforcement cracking, reinforcement-matrix interface debonding and void formation in the matrix surrounding the reinforcement. It is important to note that void formation may occur because of particle cracking or interfacial decohesion. However, if the particle stiffness and interfacial strength are sufficiently high, void nucleation may occur in the matrix simply due to high stress concentrations near the reinforcement.\cite{39}
Interfacial decohesion occurs in systems where the bond between matrix and reinforcement is very low. When the reinforcement separates from the matrix there is a resulting loss in the level of load transfer to that particle. The interfacial strength is dependent on variables such as particle coatings, heat treatments, chemical reaction layers and environmental attack. In two phase eutectic systems, such as Al-Si, the interfacial bond can be considered to be very strong due to the nature of the solidification process and consequently, interfacial decohesion is not problematic.\textsuperscript{16} Due to chemical equilibrium between the reinforcement and matrix, the interface is also very resistant to chemical degradation or reaction.

Particle cracking occurs when the stress level in the reinforcement exceeds its fracture strength. When the particle fractures, the load carried by the particle is significantly reduced, thereby reducing the overall load carrying capacity of the composite. The splitting of the particle also induces a void into the matrix. Since most second phase particles used for strengthening are brittle in nature, cracking is expected to occur over a distribution of stresses described by Weibull statistics. An extensive introduction to Weibull statistics is given by Watchman.\textsuperscript{40} Essentially, larger particles have a higher probability of failure at a given stress due to a increased probability of the presence of a critically sized flaw. This is expressed mathematically by equation 2.5\textsuperscript{41}, taken from Caceres and Griffiths\textsuperscript{41} work describing the cracking of silicon in Al-Si-Mg casting alloys.

\[
p = 1 - \exp \left[ -\frac{V}{V_o} \left( \frac{\sigma_p}{\sigma_o} \right)^m \right] \text{ eqn. 2.5}
\]
P in equation 2.5 is the probability of cracking in a particle of volume, \( V = (1/6)\pi d^2 \). \( V_0, \sigma_0 \)
and \( m \) are all constants determined for a specific volume of reinforcing phase and \( \sigma_p \) is the stress on the particle. Consequently, the statistical nature of the strength of a brittle phase introduces another length scale dependency into the system. That is, the system properties and behavior are partially determined by particle cracking which is, in turn, determined by the size of the particles.

2.3.2 Effect on Flow

When damage occurs, be it particle cracking, interfacial decohesion or matrix failure, there is a reduction in area of load bearing material. Consequently, damage results in a decrease in the flow stress of the material and eventually leads to failure. This can be expressed as shown in equation 2.6\(^{42}\)

\[
\sigma = \sigma(1 - D) \quad \text{eqn. 2-6}
\]

where \( \sigma \) is the stress in the composite at a given strain, \( \sigma \) is the stress in a completely undamaged material at the same strain and \( D \) is a damage parameter which ranges from 0 for a completely intact composite to 1 at fracture. The damage parameter has been defined by Lemaitre\(^{42}\) as

\[
D = 1 - \frac{E_{pl}}{E_i} \quad \text{eqn. 2-7}
\]

where \( E_i \) is the initial elastic modulus of the material and \( E_{pl} \) is the modulus at a given plastic strain. Knowing that the level of damage will change with strain, equation 2.6 can be differentiated to express the apparent work hardening rate as shown in equation 2.8\(^{42}\).
\[
\frac{d\sigma}{d\varepsilon} = \frac{\partial \sigma}{\partial \varepsilon} (1 - D) - \sigma \frac{\partial D}{\partial \varepsilon}
\]

eqn. 2-8

It is readily apparent that increasing either the level of damage and the rate of damage evolution \((\partial D/\partial \varepsilon)\) will result in a decrease in the work hardening rate. Equations 2.6 and 2.8 reinforce the idea that the presence of damage will result in a decrease in the flow stress of a composite. Consequently, it should be noted that testing in tension, which tends to open flaws that may be present in the reinforcement, should result in a lower flow stress than testing in compression, where existing flaws tend not to be opened and damage in the form of particle cracking is much less likely to occur.

2.3.3 Modeling of Damage

Considering its effect on both the stress level and rate of work hardening, it is important to incorporate damage into any model predicting the mechanical response of a two phase material. A number of models have been developed to this end.\textsuperscript{41,43,44,45} The basis for each model range from Weibull statistics\textsuperscript{41} to a self-consistent method which considers damaged and undamaged regions as distinct phases.\textsuperscript{44} However, each model is reasonably successful in predicting the effect of damage on the mechanical behavior of two-phase systems and correlates well with experimental data, where provided.
2.4 Summary

In the present study, the microstructure of aluminum-silicon alloys of eutectic composition will be varied using the methods described above, such that a range of silicon particle shapes and sizes is produced. Most silicon particles are expected to lie within the range of 0.1 to 10 μm in size so that, both continuum plasticity and dislocation effects might occur. These alloys will then be tested in tension and compression to assess the overall strengthening due to the silicon phase, and to examine the effect of damage on the flow and fracture of the material.
3. Scope and Objectives

Aluminum-silicon based casting alloys are an important class of engineering materials, particularly in the automotive industry. The low density, good corrosion resistance and general ease of fabrication of these materials, make them attractive for a wide range of parts. However, their inconsistent and unpredictable mechanical properties have limited their use to mostly non-critical applications, such as oil-pans, intake manifolds and cylinderheads. For automakers to consider replacing safety-critical steel components, such as suspension arms, with aluminum castings they must be able to obtain more consistent mechanical properties.

It has been shown the mechanical properties of aluminum-silicon based alloys are strongly dependent on the microstructure of the casting which is, in turn, dependent on the thermal and chemical production variables. Small changes in any of the production steps may therefore, have an effect on the performance of the casting. Rather than mechanically test every casting to see if it meets minimum performance requirements, it would be more efficient to be able to predict, through computer modeling, the mechanical properties resulting from measured casting conditions. For this to be possible, the links must be established between processing conditions, casting microstructure and mechanical properties.

By producing a series of castings under varying conditions, the present study will attempt to examine the link between a prominent feature of the microstructure, the silicon
phase, and deformation and fracture behavior in binary eutectic aluminum silicon casting alloys.

The objectives of the present study are as follows:

1. To produce binary eutectic aluminum-silicon castings with a wide range of silicon particle shapes and sizes (particularly in the 0.1 to 10 \( \mu \text{m} \) range) by varying cooling rates, heat treatments and the addition of chemical modifiers.

2. To characterize the microstructural features of the silicon phase by means of image analysis.

3. To test casting specimens in tension and compression to examine the effects of particle size, shape, distribution and overall composite damage (in the form of particle cracking) on the flow and fracture of the material.

4. To compare the results of the experiments with a model from the literature describing the effect of particle size, shape, distribution and damage on mechanical properties in two-phase materials.
4. Experimental Methods

In the present study, a number of experimental procedures were used to examine the link between the microstructure and mechanical properties of cast Al-Si alloys of eutectic composition. The eutectic alloy was cast under different conditions and then machined into metallographic and mechanical test specimens. These specimens were heat treated, metallographically examined (both optically and through image analysis) and then mechanically tested in tension or compression. The following sections describe these procedures in detail.

4.1 Casting

In order to obtain the broadest range of shapes and sizes of the silicon phase, it was necessary to develop two different casting techniques. A slow cooling rate technique was used to attempt to produce a material with a relatively large, acicular eutectic silicon phase. A smaller, more refined phase was produced through a faster cooling rate technique, in conjunction with chemical modification. Solidification in all castings was conducted directionally, to concentrate macroshrinkage at the top of the casting and to minimize the occurrence of porosity.

4.1.1 Alloy Preparation

For a binary eutectic Al-Si alloy, 11.7-12.6 wt.% Si is required. This range reflects the uncertainty of the exact composition of the binary eutectic noted in the
literature review. Initially, the value of 12.6 wt.% Si was used but, was found to result occasionally in the presence of primary silicon particles. Consequently, most castings were made with 12.0 wt.% Si, which did not produce primary silicon. For the alloys that required modification, the presence of 0.045 wt.% Sr was found to be effective for full modification.

All castings were produced from commercial purity aluminum (~99.8 wt.% Al) and a Al-Si master alloy (36.5 wt.% Si). For castings that required chemical modification, a Al-Sr master alloy (3.5 wt.% Sr) was used. The exact chemical composition of these alloys is listed below in Table 4.1. The listed impurities resulted in a total impurity content of less than 0.25 wt.% in each eutectic casting, the most concentrated of which was iron at 0.126 wt.%.

<table>
<thead>
<tr>
<th></th>
<th>Si</th>
<th>Sr</th>
<th>Fe</th>
<th>Cu</th>
<th>V</th>
<th>Zn</th>
<th>Ti</th>
<th>Ga</th>
<th>P</th>
<th>Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>Commercial</td>
<td>0.03-0.05</td>
<td>-</td>
<td>0.05</td>
<td>0.002</td>
<td>0.01</td>
<td>&lt;0.01</td>
<td>-</td>
<td>&lt;0.01</td>
<td>-</td>
<td>bal.</td>
</tr>
<tr>
<td>Purity Al</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Si Master Alloy</td>
<td>36.5</td>
<td>-</td>
<td>0.26</td>
<td>-</td>
<td>0.01</td>
<td>0.02</td>
<td>0.02</td>
<td>0.02</td>
<td>&lt;0.01</td>
<td>bal.</td>
</tr>
<tr>
<td>Sr Master Alloy</td>
<td>0.05</td>
<td>3.5</td>
<td>0.15</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>&lt;0.01</td>
<td>bal.</td>
</tr>
</tbody>
</table>

Table 4-1 Chemical composition of the master alloys used to fabricate the castings. All values listed are in wt.%. 

The exact amount of each master alloy required for a given casting composition and weight was calculated using an Excel worksheet. The master alloys were cut on a bandsaw into small pieces, no larger than approximately 1 cm³, and the required amounts
were weighed out on a digital scale accurate to 0.01 grams. The constituents were mixed together and placed in the appropriate crucible for melting.

4.1.2 Melting of the Alloy

All of the castings were melted in a graphite crucible placed in the vertical tubular, resistance-heated furnace shown in Figure 4.1. Graphite was chosen as the crucible material because it is non-soluble in aluminum, has a high thermal conductivity and it is reasonably easy to machine. The cylindrical crucible walls were wrapped in Fibrefax (ceramic fibre paper) insulation to help prevent oxidation of the graphite at the high temperatures required for melting. The crucible sat on a 2.5 inch high alumina support tube to keep the melt near the center of the furnace. This minimized the effect of thermal gradients near the end of the furnace. The exact specifications for the crucibles were dependent on whether the slow or fast casting technique was being used and are described in the respective sections below.

The temperature of the furnace was controlled by a Tempstar I Microprocessor Based Temperature Controller which monitored the furnace temperature through a K-type thermocouple. To protect the exposed coils of the furnace, a quartz tube 75 mm in diameter (just smaller than the furnace diameter) was put into the heating section. Heat loss through the top of the furnace was reduced by placing a steel cap, backed with Fibrefax, on top of the quartz tube. Two holes were drilled into this cap, one of which was used to insert thermocouples to measure the melt temperature. Compressed argon
was blown through the other hole to create an inert shroud over the melt, thereby reducing the amount of oxidation to both the metal and the graphite crucible.

Even though the eutectic temperature in this system is approximately 577°C, the melts were taken up to between 770°C and 800°C and held there for 10 to 15 minutes to ensure complete melting of the silicon and, in the case of the modified alloys, the strontium intermetallics. Thorough mixing in the liquid was obtained by plunging the melt vigorously with a graphite plunger. This was done when the alloy first melted and again, shortly before solidification was initiated.

Temperatures in the melt were measured with a K-type thermocouple placed in the melt inside a protective alumina tube. Preheating the ceramic tube was essential to prevent thermal shock fracture when inserting it into the melt. These temperature measurements were used to determine when to initiate solidification.

Figure 4-1 The resistance furnace used in the casting of the alloys.
4.1.3 Slow Casting Technique

In producing the slow cast material, the melting crucible was also used as the mold for solidification of the casting. A schematic drawing of the crucible is shown in Figure 4.2. After the alloy had melted, stirred and held at high temperature for a suitable length of time, the furnace set point was reduced to 600°C, which is above the eutectic temperature. When the temperature at the bottom of the melt reached the desired 600°C, a cylindrical water-cooled copper chill was raised on a laboratory jack into the furnace until contact with the bottom of the graphite mold was made. This chill served to draw heat from the bottom of the casting and caused the solidification front to advance vertically upwards. The time required for complete solidification to occur in this manner was approximately 20 minutes.

Figure 4-2 Schematic of the graphite mold used in the slow casting method.
4.1.4 Fast Casting Technique

To achieve a fine microstructure, a much higher cooling rate than that achieved through solidification within the furnace was necessary. As a result, solidification had to occur outside of the furnace. To this end, the graphite crucible used to melt the alloy had a 3/4 inch hole drilled in the center of the bottom. An oversized graphite rod was used to plug this hole and prevent the alloy from pouring out during melting. A mold of the exact configuration used in the slow cast technique was placed on a water-cooled copper block directly below the bottom opening of the tube furnace. The alloy was melted and stirred as described previously and then the furnace power was reduced. When the melt reached a temperature in the 670°C to 680°C range (a superheat of approximately 100°C), it was again stirred and then the plug was pulled from the hole. The molten alloy poured into the mold below and directional solidification occurred vertically. This technique is shown schematically in Figure 4.3. The time required for solidification in this manner was less than 2 minutes.

This method of casting generally produced a gradient in microstructural features (very fine particles to large, coarse particles) from top to bottom, due to a gradient in cooling rates. As a result, chemical modification was employed in the form of a strontium addition to lessen this effect. The strontium alloy was added at the same time as the commercial purity aluminum and the silicon master alloy.
4.2 Specimen Preparation

After the completion of a casting, the casting was examined to see if it would be suitable for testing. The overall microstructural features and the degree of porosity and any shrinkage had to be assessed before it would be accepted. If it was found to be acceptable, mechanical test specimens would be machined from the casting. This section describes this procedure.

4.2.1 Sectioning of the Casting

Each 375 gram casting measured approximately 70 mm in height and 51 mm (2 inches) in diameter. To examine the suitability of the casting, a longitudinal section was taken from one side of the casting as shown in Figure 4.4. The flat face of this section
was ground and polished (as detailed in section 4.4.1). It was then examined optically on a Unimet inverted light microscope. Causes for rejection of a casting included the following; the presence of primary silicon, large number of intermetallics, extreme gradients in structural features through the height of the casting, regions of undermodification in modified castings, melt contamination (e.g. insulation, broken ceramic tubes) and a large degree of porosity.

Figure 4-4 Longitudinal section taken from each casting for examination purposes

Once a casting passed the initial examination it was machined into mechanical test specimens. The casting was cut into three transverse planes of equal thickness as shown in Figure 4.5. More sections were not taken because of the presence of shrinkage cavities and increasing porosity near the top of the casting. From each of these transverse planes, one tensile and one compressive sample were taken. The remaining material from each section was left for metallography, if required.
Figure 4-5  Schematic showing how an accepted casting was sectioned for taking mechanical test specimens.

4.2.2 Specimen Dimensions

The tensile and compressive specimens for each level of the casting were taken in plane (i.e. perpendicular to the direction of solidification). The tensile specimens were non-standard rounds, with a reduced area length of 20 mm. The compressive specimens were barrels with the recommended height to diameter ratio of 1.4:1. The exact dimensions for both tensile and compressive specimens are shown in Figure 4.6.
**4.3 Heat Treatment of the Specimens**

To examine the effects of changes in the particle size and/or shape, heat treatments were employed to alter the morphology of the silicon phase. Specimens were subjected to one of two heat treatments: 1 hour or 72 hours at 540°C followed by an immediate water quench. The difference in time at 540°C does not affect the condition of the matrix after quenching, so that the only difference between specimens is the reinforcing phase. To ensure that maximum randomness between specimens was
obtained, the heat treatment schedule shown in Table 4.2 was employed for the final castings. Note that there were 6 “fast cast” and 6 “slow cast” castings to start. There were 10 pairs (i.e. one tensile and one compressive sample) of 1 hour specimens and only 8 pairs of 72 hour specimens for both slow and fast cast sets, as differences between specimens were expected to be greater in the 1 hour samples.

<table>
<thead>
<tr>
<th>Relative Position</th>
<th>Casting # 1</th>
<th>Casting # 2</th>
<th>Casting # 3</th>
<th>Casting # 4</th>
<th>Casting # 5</th>
<th>Casting # 6</th>
</tr>
</thead>
<tbody>
<tr>
<td>Top</td>
<td>72 hours</td>
<td>1 hour</td>
<td>1 hour</td>
<td>72 hours</td>
<td>1 hour</td>
<td>1 hour</td>
</tr>
<tr>
<td>Middle</td>
<td>72 hours</td>
<td>1 hour</td>
<td>72 hours</td>
<td>72 hours</td>
<td>1 hour</td>
<td>72 hours</td>
</tr>
<tr>
<td>Bottom</td>
<td>1 hour</td>
<td>72 hours</td>
<td>1 hour</td>
<td>1 hour</td>
<td>72 hours</td>
<td>1 hour</td>
</tr>
</tbody>
</table>

Table 4-2 Heat treatment schedule for 6 fast or slow castings resulting in 18 pairs of compression and tensile specimens

The 72 hour specimens were placed in an air furnace at 540°C for the duration of their heat treatment. For the 1 hour heat treatment, specimens were placed into a stirred molten salt bath at 540°C. This was done to minimize the effect of the transient heat up of the specimen to temperature as found in the air furnace. After heat treating, the specimens were immediately quenched into room temperature water.

4.4 Metallography

In the present study, metallography played an important role. It was necessary to be able to examine the microstructural features and relate them to the mechanical performance of each specimen. Specimens were examined at various stages during the course of the study. As discussed previously (in Section 4.2), the castings were examined
in the as-cast state to determine their suitability for testing. Measurements of the porosity via image analysis were taken from these longitudinal sections. Fractured tensile specimens were ground parallel to the axis of loading and polished at the centerline of the reduced section. In this instance, both the undeformed (in the buttonhead of the tensile sample) and deformed and/or damaged (in the reduced section) heat treated microstructure could be examined and evaluated. The fracture path could also be examined with this specimen. The fracture surfaces were also investigated in the scanning electron microscope (SEM). The following sections discuss the procedures used to prepare the specimens for optical and SEM microscopy and some of the measurements made through image analysis.

4.4.1 Grinding and Polishing Procedure

All of the samples were ground and polished by hand on a Buehler Ecomet IV Polisher/Grinder according to the following schedule:

- 180 grit SiC paper
- 320 grit SiC paper
- 500 grit SiC paper
- 1000 grit SiC paper
- 6 μm diamond suspension on Texmet 1000 cloth
- 1 μm diamond suspension on Texmet 1000 cloth
- 0.05 μm colloidal SiO₂ suspension on Chemomet cloth

The time spent at each stage was sample dependent. It is important to note that all samples were cold mounted in an epoxy resin for ease of handling, with the exception of the longitudinal sections used to assess the castings. These specimens were too large for mounting.
4.4.2 Etching

When it was considered desirable to increase the amount of contrast between the two constituent phases of a sample, a chemical etchant was used. It was found that immersion in Keller and Dix etch (1% hydrofluoric acid, 1.5% hydrochloric acid, 2.5% nitric acid and 95% water) for 10-20 seconds followed by rinsing in warm water provided good contrast between the aluminum and silicon phases. Modified Murkami’s etch was used to reveal the as-cast macrostructure of a casting. The sample was immersed in the etch for 5 minutes and then rinsed with warm water.

4.4.3 Image Analysis Techniques

Quantitative metallography was done on a C•Imaging Systems image analyzer. Silicon particle parameters which were measured were: maximum length, maximum breadth, equivalent circle diameter and aspect ratio. The first three parameters give a measure of the scale of the particles involved, while the aspect ratio describes the shape of the particles. The formula for calculating the equivalent circle diameter and aspect ratio are given by Equations 4.1 and 4.2.

\[ E.C.D. = \sqrt{\frac{4 \times \text{ParticleArea}}{\pi}} \quad \text{eqn. 4.1} \]

\[ A.R. = \frac{\text{MaximumLength}}{\text{MaximumBreadth}} \quad \text{eqn. 4.2} \]

If a particle is assumed to be ellipsoidal, an aspect ratio of 1 indicates perfect circle (i.e. an ellipse of equal axis length). The average area fraction of the silicon phase in the material was also measured.
4.4.4 Deep Etching

It is useful to examine the three-dimensional nature of the second phase particles in a two-phase system. To this end, deep etching, which preferentially removes the aluminum matrix and leaves only the silicon phase behind, was conducted. Samples were immersed in a solution of 5% hydrochloric acid and 10% hydrofluoric acid for 60 to 75 minutes. The silicon phase was then be examined in the SEM.

4.5 Mechanical Testing

All of the mechanical testing was performed on an MTS/Instron servo-hydraulic testing machine. The setup for both tensile and compression tests are described below. Data (time, load, position, and strain) was acquired directly on a personal computer at a global rate of 0.1 kHz resulting in approximately 1000 to 3000 data points per sample, depending on the strain to failure. For each sample, a true stress-true strain curve was calculated from the accumulated data. The equations of interest to generate these curves are shown in equations 4.3 to 4.6 as follows:

\[ \varepsilon_{\text{eng}} = \frac{\Delta \ell}{\ell} \]  
\[ \text{eqn. 4.3} \]

\[ \varepsilon_{\text{true}} = \ln(1 + \varepsilon_{\text{eng}}) \]  
\[ \text{eqn. 4.4} \]

\[ \sigma_{\text{eng}} = \frac{F}{A} \]  
\[ \text{eqn. 4.5} \]

\[ \sigma_{\text{true}} = \sigma_{\text{eng}} (1 + \varepsilon_{\text{eng}}) \]  
\[ \text{eqn. 4.6} \]
where $\varepsilon_{\text{eng}}$ and $\varepsilon_{\text{true}}$ are the engineering and true strains, $\sigma_{\text{eng}}$ and $\sigma_{\text{true}}$ are the engineering and true stresses, $\Delta \ell$ is the change in specimen length as measured by the extensometer, $\ell$ is the initial length of the specimen (within the gage length), $F$ is the measured load (in kN) and $A$ is the cross-sectional area of the specimen in mm$^2$.

Sources of error include sample dimension measurement and some machine compliance. The expected accuracy of the sample measurement is ±0.005 mm.

4.5.1 Tensile Tests

For the tensile tests, a 5 kN load cell was used. The specimens were held in a split-half grip that screwed into the cross-head as shown in Figure 4.7. For measuring strain, an extensometer was attached to the reduced section of the sample by means of elastics and knife edges. Spacers were used to increase the gage length of the extensometer to 19 mm so that the majority of the 20 mm reduced section was accounted for. The specimens were deformed at a nominal rate of 0.01 mm/s (corresponding to a nominal strain rate of $0.5 \times 10^{-3} \text{ s}^{-1}$) until failure. One half of each fractured specimen was mounted and ground and polished (as described in section 4.4.1) for optical photography and image analysis. The other half was examined in the SEM.
4.5.2 Compression Tests

Due to higher loads required for deformation of compression samples, a 50 kN load cell was used. However, the compressive samples are much shorter than the tensile specimens and an extensometer could not be attached to the specimen and still fit between the platens. As a result, the extensometer was attached to the fixed and moving platens and the strain was measured indirectly. This setup is shown schematically in Figure 4.8. The compression samples were deformed at a rate of 0.007 mm/s (again, corresponding to a strain rate of $0.5 \times 10^{-3} \text{s}^{-1}$).

In compressive testing, it is very important to minimize friction between the sample and the platen. If friction is present, it will cause the ends of the sample in contact
with the platens to stick, resulting in non-uniform deformation. Consequently, a 0.06 mm thick sheet of Teflon sprayed with molybdenum disulfide lubricant was placed between the ends of the samples and the platens. This minimized the friction and no barreling (i.e. non-uniform deformation) occurred. However, the deformation of the Teflon sheet and the compliance of the platens had to be accounted for in the overall stress-strain curve.

The deformation of the lubricant and the machine as a function of load was determined by compressing a hardened steel sample elastically with the lubricant in place and measuring the displacement with a strain gauge. The difference between the theoretical displacement level (based on a known Young's modulus) in the steel and the measured displacement level at a given load was then measured. The difference in displacement was attributed to the deformation of the lubricant and the platens of the machine. This results of this test are shown in Figure 4.9. From this curve a displacement difference of approximately 0.40 mm was found at a load of 5 kN. To account for this displacement in the 14 mm high compression barrel tests, a strain correction of approximately 0.029 (=0.4 mm/14 mm) was subtracted from the true stress-true strain curve. This correction put the elastic region of the compression curves in line with the elastic region of the tensile curves.
Figure 4-8 Setup for compression testing, showing placement of extensometer.

Figure 4-9 Results of deformation test to determine strain correction due to lubricant deformation and machine compliance.
5. Results

This chapter details the experimental results of this study. The microstructure of the castings are described qualitatively and quantitatively. The results from mechanical tests are summarized and some typical fracture surfaces are presented.

5.1 Microstructure Results

5.1.1 Qualitative Description of the Microstructure

The microstructures of the castings produced in the present study were evaluated by two different techniques. Optical photographs were taken of polished plane sections from each casting and heat treatment condition (for a description of the casting techniques and heat treatments see Sections 4.1.1 and 4.1.3 respectively). These photographs are shown in Figures 5.1a to 5.6a. These same samples were then deep etched to remove the aluminum matrix and reveal the three-dimensional structure of the silicon phase. Photographs of the deep-etched samples were taken on the scanning electron microscope (SEM) as shown in Figures 5.1b to 5.6b. Note how the silicon phase morphology is, in some cases, more complicated than indicated by the two-dimensional optical photographs.
Figure 5-1  (a) Optical and (b) SEM photograph of slow cast, unmodified microstructure in as-cast state. (400x)
Figure 5-2  (a) Optical and (b) SEM photograph of slow cast, unmodified microstructure after 1 hour at 540°C. (400x)
Figure 5-3  (a) Optical and (b) SEM photograph of slow cast, unmodified microstructure after 72 hours at 540°C. (400x)
Figure 5-4 (a) Optical and (b) SEM photograph of fast cast, modified microstructure in as-cast state. (1000x)
Figure 5-5  (a) Optical and (b) SEM photograph of fast cast, modified microstructure after 1 hour at 540°C. (1000x)
Figure 5-6  (a) Optical and (b) SEM photograph of fast cast, modified microstructure after 72 hours at 540°C. (1000x)
By comparing Figures 5.1 and 5.4, it is readily apparent that the scale of the as-cast silicon phase is much smaller in the fast cast material than it is in the slow cast material in the as-cast state. Deep etching reveals a more striking difference. The silicon in the unmodified, slow cast material is very flat, angular and platelike. On the other hand, the silicon in the modified, fast casting displays a fibrous, "seaweed-like" appearance. Note also the primary aluminum dendrite cells seen in Figures 5.4 and 5.5. These are the result of a shift in the eutectic point towards higher silicon concentrations, caused by the fast solidification rate and the presence of the modifying agent. This effect has been described by Gruzleski and Closset.3

After being held for 1 hour at 540°C, changes occurred in the fast cast, modified specimen. There was a noticeable coarsening of the silicon phase and the deep etched photograph shows an almost complete break-up of the seaweed structure. However, one hour at 540°C has done little to alter the morphology of the silicon phase in the slow cast material.

Increasing the time at 540°C to 72 hours allows for more noticeable changes to occur. In the slow cast material, the thin plates have coarsened and mostly broken up as shown in the SEM photograph. The particles appear to be shorter and wider, indicating a reduced aspect ratio. In the fast cast material the fine, branched network of the as-cast silicon has completely degenerated into a discontinuous spacing of coarser, almost spherical particles.

In addition to the aluminum matrix and the silicon phase, there are other microstructural features which were observed in the present castings. In both the fast and
slow cast material, long needles of iron-rich intermetallics could be found. These were postulated to be FeSiAl₅ (β-phase) intermetallics based on descriptions by Wang, Cáceres and Griffiths⁴⁶ for a low magnesium aluminum-silicon (7%) casting alloy. However, with an effective iron content of only 0.12 weight percent in the melt from impurities in the master alloys, this phase constituted only a minor fraction of the microstructure. While Wang, Cáceres and Griffiths⁴⁶ observed that the β phase tended to undergo dissolution when exposed to solutionizing temperatures (540°C) for prolonged periods of time, this was not readily observed in the present study. No other intermetallics or other phases were observed. Some porosity was visible in the as-cast condition and tended to increase as distance from the chill increased.

5.1.2 Quantitative Description of the Microstructure

Image analysis results describing the eutectic silicon particles are summarized in Table 5.1 for both fast cast, modified and slow cast, unmodified materials in the as-cast state and after 1 hour and 72 hours at 540°C. Image analysis allows for the quantification of the size and shape of the silicon phase. Relevant measurements include the maximum length, maximum breadth, aspect ratio, equivalent circle diameter and area fraction of the silicon phase. Descriptions of these measurements are found in section 4.4.3.
Table 5-1 Image analysis results

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Slow Cast, Unmodified</th>
<th>Fast Cast, Modified</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>As-Cast</td>
<td>1 hour at 540°C</td>
</tr>
<tr>
<td></td>
<td>1 hour at 540°C</td>
<td>72 hours at 540°C</td>
</tr>
<tr>
<td>Average</td>
<td>As-Cast</td>
<td>1 hour at 540°C</td>
</tr>
<tr>
<td>(Std. Dev)</td>
<td>540°C</td>
<td>540°C</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Maximum</td>
<td>10.1 (9.4)</td>
<td>10.4 (8.6)</td>
</tr>
<tr>
<td>Length</td>
<td></td>
<td>11.8 (10.5)</td>
</tr>
<tr>
<td>(µm)</td>
<td>540°C</td>
<td>1 hour at 540°C</td>
</tr>
<tr>
<td></td>
<td>2.2 (1.2)</td>
<td>2.6 (1.4)</td>
</tr>
<tr>
<td></td>
<td>540°C</td>
<td>72 hours at 540°C</td>
</tr>
<tr>
<td></td>
<td>3.9 (2.2)</td>
<td></td>
</tr>
<tr>
<td>Maximum</td>
<td>2.4 (1.6)</td>
<td>3.2 (2.0)</td>
</tr>
<tr>
<td>Breadth</td>
<td></td>
<td>2.8 (1.9)</td>
</tr>
<tr>
<td>(µm)</td>
<td>540°C</td>
<td>1 hour at 540°C</td>
</tr>
<tr>
<td></td>
<td>1.0 (0.5)</td>
<td>1.5 (0.6)</td>
</tr>
<tr>
<td></td>
<td>540°C</td>
<td>72 hours at 540°C</td>
</tr>
<tr>
<td></td>
<td>2.4 (1.4)</td>
<td></td>
</tr>
<tr>
<td>Aspect</td>
<td>4.2 (2.9)</td>
<td>2.2 (0.8)</td>
</tr>
<tr>
<td>Ratio</td>
<td></td>
<td>4.4 (3.1)</td>
</tr>
<tr>
<td></td>
<td>540°C</td>
<td>1 hour at 540°C</td>
</tr>
<tr>
<td></td>
<td>1.8 (0.7)</td>
<td></td>
</tr>
<tr>
<td></td>
<td>540°C</td>
<td>72 hours at 540°C</td>
</tr>
<tr>
<td></td>
<td>1.7 (0.6)</td>
<td></td>
</tr>
<tr>
<td>Equivalent</td>
<td>6.8 (3.4)</td>
<td>9.2 (4.5)</td>
</tr>
<tr>
<td>Circle</td>
<td></td>
<td>8.2 (4.7)</td>
</tr>
<tr>
<td>Diameter</td>
<td>540°C</td>
<td>1 hour at 540°C</td>
</tr>
<tr>
<td>(µm)</td>
<td>1.9 (1.7)</td>
<td>2.5 (2.1)</td>
</tr>
<tr>
<td></td>
<td>540°C</td>
<td>72 hours at 540°C</td>
</tr>
<tr>
<td></td>
<td>4.5 (4.0)</td>
<td></td>
</tr>
<tr>
<td>Area</td>
<td>13.2 (1.8)</td>
<td>11.4 (2.2)</td>
</tr>
<tr>
<td>Fraction</td>
<td></td>
<td>14.4 (1.6)</td>
</tr>
<tr>
<td>(%)</td>
<td>540°C</td>
<td>1 hour at 540°C</td>
</tr>
<tr>
<td></td>
<td>9.8 (1.1)</td>
<td></td>
</tr>
<tr>
<td></td>
<td>540°C</td>
<td>72 hours at 540°C</td>
</tr>
<tr>
<td></td>
<td>11.2 (1.3)</td>
<td></td>
</tr>
</tbody>
</table>

It is apparent from this data that a faster cooling rate and the presence of a modifying agent produced a much different structure in the fast cast material than in the slow cast material. The silicon particles in the slow as-cast material average over 10 microns in length and almost 2.5 microns in width, while the fast as-cast particles average just over 2 microns in length and 1 micron in width. The elongated nature of the slow cast particles is also represented by an average aspect ratio almost twice as high as in the fast cast material.

The effect of heat treatment time on the eutectic silicon particle characteristics can also be seen in Table 5.1. In the slow cast material, one hour at 540°C results in statistically insignificant increases in the average particle length, breadth and aspect ratio. However, 72 hours at 540°C causes the average aspect ratio to drop while the equivalent
circle diameter and maximum breadth continues to rise. This points to spherodization of the particles and continued coarsening. Figures 5.7 to 5.9 show the change in slow cast particle characteristic dimensions (length, width and aspect ratio) with heat treatment. Note the widening of the distribution of sizes as the heat treatment time increases. This is most likely due to break-up (i.e. spherodization) and coarsening of the plate structure.

In the case of the fast cast material, increasing time at 540°C causes increases the particle dimensions and a relatively insignificant drop in the aspect ratio. The standard deviation in particle dimensions also increases, indicating a widening of the particle size distribution. This is shown in Figures 5.10 to 5.12 where the distributions of particle length, width and aspect ratio are shown for the fast cast material in the three states of heat treatment (as-cast, 1 hour and 72 hours at 540°C). Note also from Table 5.1 that the percentage increase in the particle diameter ratio is much greater for a given time at temperature in the fast cast material than in the slow castings, indicating faster coarsening. That is, smaller particles coarsen faster than larger ones. This is to be expected, as the coarsening of a sphere is inversely proportional to the square of its radius (i.e. \( \frac{dr}{dt} \propto \frac{1}{r^2} \)).

The results presented for the area fraction of particles in Table 5.1 give an estimate of the volume fraction of the silicon phase. However, there is a reasonable degree of error associated with these measurements. Surface scratches cannot be differentiated from the silicon phase and may skew the statistics. Also, there is some difficulty resolving the very fine particles in the fast as-cast and 1 hour specimens at the relatively low magnification required for area fraction measurements.
Figure 5-7  Effect of heat treatment on particle length distribution in slow cast material.

Figure 5-8  Effect of heat treatment on particle width distribution in slow cast material.
Figure 5-9 Effect of heat treatment on particle aspect ratio distribution in slow cast material.
Figure 5-10 Effect of heat treatment on particle length distribution in fast cast material.

Figure 5-11 Effect of heat treatment on particle width distribution in fast cast material.
Figure 5-12 Effect of heat treatment on particle aspect ratio distribution in fast cast material.
The values listed in Table 5.1 are averages for the given condition and are comprised of samples from various positions in each casting. The image analysis measurements from the individual samples can be found in Appendix A. No discernible relationship between the particle characteristics and the position in the casting could be found in the slow castings. In the fast castings, however, there was a gradient in terms of the microstructure. Particles closest to the chill at the bottom of the casting tended to be smaller and more circular (i.e. lower aspect ratio) than those found near the top of the casting because the effective cooling rate is slower as the distance from the chill increases. The difference in average particle dimensions from bottom to top may also be due to the presence of regions similar to rosettes found in cast irons near the top of the castings. Rosettes are regions of small particles, surrounded by flakes or plates radiating outward from the center of the fine region as shown in Figure 5.13. They are usually found in thin sections or castings with high solidification rates.

The castings produced in the course of this study were considered to be relatively sound. Image analysis on the as-cast longitudinal section of each casting revealed an average pore area fraction of 0.21% in both the slow and fast cast materials. (Porosity measurements for each individual casting can be found in Appendix A.) A larger number of smaller pores were found in the fast castings as compared to the slow castings. The average pore size in the slow castings (0.00728 mm²) was approximately 50% larger than in the fast castings.
Figure 5-13 "Rosette" feature typically found in fast cast specimens away from the chill.

5.2 Mechanical Testing Results

5.2.1 Tensile Results

The true stress-true strain curves for the tensile tests of slow cast and fast cast samples are shown in Figures 5.14 and 5.15, respectively. Figures 5.16 and 5.17 compare the fast and slow cast samples after 1 hour and 72 hours at 540°C. These tensile curves show that for either heat treatment, at a given strain the flow stress is higher in the slow cast specimens than in the fast cast specimens. Note also, the serrated flow in the tensile
True Stress-True Strain Curves for Slow Cast, Unmodified Tensile Specimens Heat Treated for Various Times at 540°C
N.B. x denotes final fracture

Figure 5-14 Tensile stress-strain curves for slow cast, unmodified material.

True Stress-True Strain Curves for Fast Cast, Modified Tensile Specimens Heat Treated for Various Times at 540°C
N.B. x denotes final fracture

Figure 5-15 Tensile stress-strain curves for fast cast, modified material.
Figure 5-16 Tensile stress-strain curves for slow and fast cast materials, solution treated for 1 hour at 540°C.

Figure 5-17 Tensile stress-strain curves for slow and fast cast materials, solution treated for 72 hours at 540°C.
curves. The occurrence of these serrations was investigated by running tensile tests on extra Al-Si samples cast at a moderate solidification rate. Tested in the as-cast state, these samples did not exhibit the serrated flow. However, samples from the same casting which were heated to 540°C and then either water quenched, warm oil quenched or air cooled, displayed the same type of serrated flow observed in Figures 5.14 to 5.17. The magnitude of the serrations was proportional to the severity of the quench step before testing. Consequently, the serrations were attributed to dynamic strain aging most likely as a result of silicon in solid solution interacting with dislocation motion. However, the magnitude of this effect is minor compared to the differences arising from heat treatment differences.

The important engineering data (yield stress, ultimate tensile strength and strain to failure) measured from the tensile curves are shown in Table 5.2. As can be seen from the stress-strain curves and Table 5.2, increasing the time at 540°C from 1 hour to 72 hours causes a drop in the average flow stress and an increase in the average strain to failure. The most brittle specimens were the slow castings, heat treated for 1 hour at 540°C and the most ductile were the fast cast samples, heat treated for 72 hours at 540°C. Both the fast and slow cast materials, heat treated for 72 hours at 540°C had nearly identical yield and ultimate tensile strengths but differed in strain to failure by almost 85%.
Table 5-2  Average Engineering Properties Determined from Tensile Tests

<table>
<thead>
<tr>
<th>Property</th>
<th>Slow Cast, Unmodified</th>
<th>Fast Cast, Modified</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>1 hour at 540°C</td>
<td>72 hours at 540°C</td>
</tr>
<tr>
<td>0.2 % Yield Stress (MPa)</td>
<td>67.8 (5.9)</td>
<td>75.9 (2.0)</td>
</tr>
<tr>
<td></td>
<td>63.1 (1.6)</td>
<td>63.1 (1.9)</td>
</tr>
<tr>
<td>Ultimate Tensile Strength (MPa)</td>
<td>159.5 (11.7)</td>
<td>177.0 (10.6)</td>
</tr>
<tr>
<td></td>
<td>152.0 (6.5)</td>
<td>152.8 (7.0)</td>
</tr>
<tr>
<td>Strain to Failure (%)</td>
<td>2.8 (0.8)</td>
<td>7.5 (3.6)</td>
</tr>
<tr>
<td></td>
<td>6.3 (2.8)</td>
<td>11.6 (3.4)</td>
</tr>
</tbody>
</table>

It is important to note that the values in Table 5.2 are averages for all samples in a given condition. However, there did exist some difference in average properties as a function of position in the casting. Figures 5.18 and 5.19 show the differences from bottom to top in average elongation to failure and ultimate tensile strength, respectively, for samples in all four testing conditions. Moving from the bottom of the casting (nearest the chill) towards the top, increased both the fracture strain and ultimate tensile strength in the slow cast material, while the opposite was true in the fast cast material, regardless of the heat treatment condition. The similarity between the trends in strain to failure and the UTS as a function of position for a given condition are to be expected, as all of these materials displayed little or no post-necking deformation. While there were differences between casting technique and heat treatment conditions, there was no discernible relationship between the yield strength and position in the casting.
Figure 5-18 Average strain to failure as a function of position in the casting for various casting and heat treatment conditions.

Figure 5-19 Average ultimate tensile strength as a function of position in the casting for various casting and heat treatment conditions.
5.2.2 Compressive Results

The true stress-true strain curves for the compression tests are presented in Figures 5.20 and 5.21 for the slow cast and fast cast materials, respectively. These specimens were not loaded to failure and, consequently, there is no determination of ultimate tensile strength or strain to failure. However, the same effect of heat treatment time on the average flow stress observed in the tensile tests is seen here. By increasing the time at 540°C from 1 hour to 72 hours, the average flow stress of the material drops, regardless of its as-cast structure. Similarly, for either heat treatment at a given strain the flow stress in the slow cast material is higher than in the fast castings. This was shown previously for the tensile samples.
Figure 5-20 Compressive stress-strain curves for slow cast, unmodified material.

Figure 5-21 Compressive stress-strain curves for fast cast, modified material.
5.2.3 Tension versus Compression

The tensile and compressive stress-strain curves for representative pairs from all four testing conditions are shown in Figures 5.22 to 5.25. Note that in Figures 5.22 (slow cast, 1 hour), 5.23 (slow cast, 72 hours) and 5.24 (fast cast, 1 hour) as the strain increases, so does the difference between compressive true stress and tensile true stress. However, the samples tested for Figure 5.25 (fast cast, 72 hours) did not show this effect.

The measured difference between compressive true stress and tensile true stress as a function of true strain for all slow cast 1 hour, slow cast 72 hours, fast cast 1 hour and fast cast 72 hour specimens are shown in Figures 5.26 through 5.29, respectively. The legends for each graph describe the casting which the samples came from (the letter) and their position in the casting (the number), where 1 represents the bottom of the casting (closest to the chill), 2 represents an intermediary position and 3 denotes the uppermost sample in the casting. Large differences between tension and compression were seen in the slow cast 1 and 72 hour specimens and they evolved over a relatively short range of strain, as shown by Figures 5.26 and 5.27. This increase in difference in stress was also seen in most of the fast cast, 1 hour samples, although over a longer range of stress, as shown in Figure 5.28. However, the two samples which came from the bottom of their respective castings (V1 and W1) showed very little difference between compression and tension, until just before the onset of failure. The majority of the fast cast, 72 hour specimens did not show significant difference between tension and compression (see Figure 5.29). However, two samples (U1 and T2) are exceptions to this trend. The T2 sample pair had an anomalously low tensile flow stress while the U1 sample pair had a
Figure 5-22 Tensile versus compressive stress strain curves for a slow cast, 1 hour specimen.

Figure 5-23 Tensile versus compressive stress strain curves for a slow cast, 72 hour specimen.
Figure 5-24 Tensile versus compressive stress strain curves for a fast cast, 1 hour specimen.

Figure 5-25 Tensile versus compressive stress strain curves for a fast cast, 72 hour specimen.
Figure 5-26 Difference between compressive and tensile true stress as a function of true strain in slow cast, 1 hour samples.

Figure 5-27 Difference between compressive and tensile true stress as a function of true strain in slow cast, 72 hour samples.
Figure 5-28 Difference between compressive and tensile true stress as a function of true strain in fast cast, 1 hour samples.

Figure 5-29 Difference between compressive and tensile true stress as a function of true strain in fast cast, 72 hour samples.
significantly high compressive flow stress. The deviation of these two individual curves from the norm, are responsible for the large difference between compressive and tensile stress shown in Figure 5.29.

This difference between tension and compression, where present, was attributed to increased silicon particle cracking, resulting in decreased load carrying capacity of the tensile specimens. Consequently, some simple measurements were made to estimate the strength of the silicon phase.

5.2.4 Cracking of Silicon Particles

Vicker's microhardness measurements were made on large silicon particles to estimate the yield strength of the silicon phase. Using a 25 gram load with a 10 second dwell time, the average measured hardness was $H_v=945$. Multiplying this value by $9.81/3 \text{ ms}^2$ (to convert the Vickers number into a proper stress unit) gives an approximate yield strength of 3090 MPa. This may be used as an upper bound for the strength of the silicon phase.

Some of the indentations during the microhardness tests caused cracks to form in the silicon extending out from the corners of the pyramid as shown in Figure 5.30. These cracks were measured using the micrometer on the microhardness machine, and used to estimate the fracture toughness, $K_C$, of the silicon. This approximation can be made using Equation 5.247:

$$K_C = d \left( \frac{E}{H} \right)^{\frac{1}{2}} \left( \frac{P}{a^{\frac{3}{2}}} \right)$$  \hspace{1cm} \text{eqn. 5.2}
where $E$ is the elastic modulus (estimated to be 112.65 MPa (Metals Handbook)), $H$ is the measured hardness value, $P$ is the indentor load, $a$ is the measured crack length and $d$ is an indentor geometry-dependent constant estimated to be 0.016. By utilizing Equation 5.2 an approximate fracture toughness in the range of 1.31 MPa√m was obtained. This relates well to reported values of 1 to 2 MPa√m.27

The fracture strength of silicon was also estimated by examining some cracked particles from the slow cast compression specimens. It is usually expected that compressive loading will not induce particle cracking because there is no force for opening of existing flaws. However, the expansion of the compression barrel in the radial direction induces tensile strains perpendicular to the compression axis which may cause particle cracking. While this effect was not observed in the fast cast specimens, where the particles tended to be spherical, it was seen in the long, angular plates of the slow cast specimens. Optical micrographs, such as the one shown in Figure 5.31, were taken of slow cast compression samples and the number of cracked and uncracked particles were counted and their angles to the tension axis (i.e. the compression axis is at 90°) were measured and recorded. Figure 5.32 shows the percentage of particles that fractured in a given orientation range for a slow cast 72 hour specimen. It can be seen that as the particle orientation changes from perpendicular to parallel to the compression axis the probability of fracture decreases.
Figure 5-30  Example of indentation measurement induced cracking used to estimate silicon fracture toughness.

Figure 5-31  Cracked silicon particles from slow cast, 1 hour compression sample.
Figure 5-32  Probability of particle cracking as a function of orientation to the tensile axis in compression tests, assuming the tensile axis is at 0° and the compression axis is at 90°.

The length and thickness of the cracked plates of silicon were also measured and an estimate of the silicon fracture strength, \( \sigma_f \), was made using equation 5.3 which is the result of a simple shear lag analysis for a flat plate:

\[
\sigma_f = \frac{2\tau_m(t + w)}{tw} \ell
\]

where \( \ell \) is the plate length (in the loading direction), \( w \) is the plate width, \( t \) is the plate thickness and \( \tau_m \) is the matrix shear strength taken to be 50 MPa. However, the plate width, \( w \) was not explicitly measured. Based on the observation of the deep etch samples (see Figure 5.2a) the plates were assumed to be approximately square, such that \( w = \ell \).
This yields an average silicon fracture strength of 1608.6 MPa in the slow cast 1 hour samples and 726.1 MPa in the 72 hour samples. Although, this is a somewhat crude approach, it allows some estimate to be placed on the in situ fracture strength. The fracture strength is expected to be somewhat lower in the 72 hour specimens as they are larger and consequently have a higher probability of containing a critically sized flaw.

5.2.5 Fracture Surfaces

Cross-sectional optical photographs and SEM photographs normal to the fracture surface of representative samples for each material condition are presented in Figures 5.33 through 5.38. A fractured slow cast specimen ($e_f=3.5\%$) heated for 1 hour at $540^\circ$C is presented in Figure 5.33. The SEM photograph reveals a number of very long, flat plates of silicon which have undergone cleavage. The optical cross-section of the fracture surface show a relatively angular fracture surface with minimal elevation change. Figure 5.34 shows the fracture surface of a slow cast specimen ($e_f=5.6\%$) heated for 72 hours at $540^\circ$C. The normal fracture surface shows fewer but, slightly larger cleaved plates of silicon than in the 1 hour specimen. The particle coarsening can also be observed in the cross-sectional view, as can an increase in amount of elevation changes.

The fracture surface of the fast cast 1 hour specimen ($e_f=10.8\%$) taken from the bottom of a casting shown in Figure 5.35 is much more dimpled and rough than the slow castings shown previously. This indicates a more ductile failure and is reflected in the higher strain to failure. Another fast cast 1 hour specimen ($e_f=6.1\%$), this time taken from the top position in the casting, is shown in Figure 5.36. Note the presence of rings
Figure 5-33  (a) Cross-sectional optical and (b) normal SEM photographs of the fracture surface for a slow cast, 1 hour sample.
(a) Cross-sectional optical and (b) normal SEM photographs of the fracture surface for a slow cast, 72 hour sample.

Figure 5-34
Figure 5-35  (a) Cross-sectional optical and (b) normal SEM photographs of the fracture surface for a fast cast, 1 hour sample from the bottom of a casting.
Figure 5-36 (a) Cross-sectional optical and (b) normal SEM photographs of the fracture surface for a fast cast, 1 hour sample from the top of a casting.
Figure 5-37 (a) Cross-sectional optical and (b) normal SEM photographs of the fracture surface for a fast cast, 72 hour sample from the bottom of a casting.
Figure 5-38 (a) Cross-sectional optical and (b) normal SEM photographs of the fracture surface for a fast cast, 72 hour sample from the top of a casting.
of brittle plates of silicon radiating out from finer, equiaxed particles on both the normal and cross-sectional views of the fracture surface. This is one of the “rosettes” mentioned previously in section 5.2.

A fractured fast cast specimen ($\varepsilon_f=13.0\%$) taken from the bottom of a casting and heated for 72 hours is shown in Figure 5.37. The surface is very dimpled in accordance with the high strain to failure. Another fast cast 72 hour sample ($\varepsilon_f=7.7\%$), this time taken from the top of the casting, is shown in Figure 5.38. This specimen shows a more mixed fracture surface with ductile dimpling and some small, cleaved plates of silicon visible on the SEM photograph. Note also the large cavity which appears to be from a pore. However, the scale of this cavity is not representative of the size of pores found in each casting through image analysis.

Higher magnification images of the more ductile fracture surfaces (i.e. the bottom specimens in the fast castings) were taken in the SEM. Examples are shown in Figures 5.39 and 5.40. From these images, the average dimple size was measured. In the fast cast 1 hour specimens, the average dimple size ranged from 3-4 $\mu$m in diameter compared with 7-11 $\mu$m for the 72 hour specimens. These values correspond reasonably well with the effective particle spacing calculated by using Equation 2.2. For calculation purposes, $f$ was estimated by the area fraction and $d$, by the equivalent circle diameter determined for the casting by image analysis and listed in Table 5.1. Substituting the appropriate parameters into equation 2.2 gives $L=6.9$ $\mu$m for the 1 hour specimens and 11.9 $\mu$m for the 72 hour specimens. Recall that equation 2.2 is an approximation for fibres rather than particles.
Figure 5-39  Ductile fracture surface from a fast cast, 1 hour specimen used for dimple size measurements.

Figure 5-40  Ductile fracture surface from a fast cast, 72 hour specimen used for dimple size measurements.
5.3 Summary

The microstructures arising from the four different conditions of casting and heat treatment were markedly different in terms of size and shape of the silicon phase. The results of the mechanical testing also show differences between the microstructures. Consequently, it is assumed that changes in the size and morphology of the silicon phase are important in determining the deformation and fracture behavior of Al-Si alloys. The reasons for these changes are discussed in the next chapter.
6. Discussion

This chapter presents a discussion of the plastic flow and fracture of cast aluminum-silicon eutectic alloys. The effects of size and morphology of the silicon phase on the tensile and compressive behavior are examined, with particular reference to strengthening, damage accumulation and fracture in the alloy. An interpretation of the experimental results is presented along with a comparison to a model for particle cracking.

6.1 Discussion of Microstructural Features

By changing casting conditions from a slow solidification rate without modifiers to a faster solidification rate with modification, a difference in microstructural size and morphology was created. This is reflected in the measured silicon particle characteristics for the fast and slow cast materials presented (see Table 5.1) in the results. However, by deep etching the aluminum matrix away to reveal the three-dimensional nature of the silicon phase, some shortcomings of the particle measurement technique were exposed. For example, the results of the image analysis describe the silicon phase in the slow as-cast material as being long, narrow rods. This is how they appear in the optical (two-dimensional) photographs (see Figures 5.1a to 5.3a). However, the true morphology of the silicon phase in these castings, as revealed by deep etching, is that of long, thin plates (see Figures 5.1b to 5.3b). There is a low probability of sectioning and polishing directly
in the thickness plane of any given plate of silicon. Similarly, the results from image analysis do not fully describe the seaweed structure of fast cast, modified material in the as-cast state (i.e. the continuous nature of the structure is lost). However, with increasing solution treatment time at 540°C the interconnected silicon plate structure of the slow cast material and the fibrous network of the fast cast material spherodized and started to coarsen. As the silicon approached a spherical morphology, the two-dimensional image analysis results became more representative of the actual three-dimensional structure. Nevertheless, the image analysis results for all stages do give some indication of the differences of scale between the fast and slow cast silicon phase.

The grain sizes of the castings were not measured explicitly. However, Figures 6.1 and 6.2 show the as-cast structure of the slow and fast cast materials, respectively. The presence of a fine equiaxed chill zone is seen around the sides and bottom of the fast casting. The center of the casting is also equiaxed but, much coarser. The primary aluminum dendrites were easily visible to the naked eye, indicating a grain size on the order of a millimeter. Primary aluminum dendrites were uncommon in the slow cast material making it difficult to estimate the grain size but, it is to be expected to be larger than in the fast cast material because of the much slower cooling rate. Grain boundary effects are expected to be minimal due to the large spacing (on the order of a millimeter or greater) compared to the spacing of the silicon phase. This has been suggested by Gruzleski and Closset who noted that, while grain boundaries may affect hot tearing, they are of little significance in determining the room temperature mechanical properties for Al-Si casting alloys.
Figure 6-1  As-cast structure in a slow cast, unmodified casting.

Figure 6-2  As-cast structure in a fast cast, modified casting.
It is expected that the presence of the brittle intermetallics and porosity observed in the castings might degrade the mechanical properties and, particularly, the fracture behavior of the samples. However, the general shape, size and distribution of both the iron-rich intermetallic phase and the porosity appeared not to be affected by the heat treatment. Consequently, while their presence may affect the mechanical properties of a casting, they cannot be the major cause of the dramatic change in behavior with increasing heat treatment time, indicating that some other feature is dominating the behavior of the material. This assumption is supported by the observation that in low magnesium Al-Si casting alloys it is the cracking of the silicon particles, rather than cracking of the $\beta$ (FeSiAl$_5$) intermetallic phase, which controls fracture.\textsuperscript{46} Similarly, evidence of porosity was rarely seen on the fracture surfaces presented in Figures 5.33 to 5.38.

6.2 Mechanical Testing Results

Mechanical tests on the four different sample conditions (i.e. unmodified, slow cast and modified, fast cast solution treated for 1 or 72 hours at 540°C) produced quite different results. For the reasons discussed above, these changes in mechanical properties and fracture behavior are primarily attributed to a change in the size and morphology of the silicon particles. Recall from the literature review that the shape and size of brittle,
second phase particles may affect the material behavior in many ways, including the following:

1. Decreasing particle aspect ratio (i.e. change in morphology from elongated plates or fibres to spheres) reduces the level of load transfer to the reinforcement, resulting in a drop in overall material strength. (See Section 2.2.1.)

2. Decreasing particle size leads to eventual breakdown of continuum (i.e. load transfer) models. This results in an increase in strength due to dislocation mechanisms. A range of particle sizes exist where both continuum plasticity and dislocation effects may be present. (See section 2.2)

3. Decreasing particle size increases effective particle strength due to lower probability of finding flaws in the particle. (See Section 2.3.1)

In the following discussion of the mechanical test results, the general morphology of the silicon phase in the fast cast material is assumed to be spherical, which is a reasonable approximation according to the three-dimensional shape shown in Figures 5.5b and 5.6b. The silicon phase in the slow cast material however, is much more complicated and any approximation will be somewhat crude. For the purposes of this discussion, the silicon phase in the slow cast material is assumed to be present in the form of thin plates.

6.2.1 Strengthening Due to Silicon Particles

It was noted that there was a difference in average flow stress when comparing either the method of casting or the effect of solution treatment time for both tension and compression. For a given heat treatment, the slow cast materials exhibited a higher flow stress than the fast cast material (see Figures 5.16 and 5.17). And for a given casting
technique, the samples treated for 1 hour at 540°C had an average flow stress higher than for samples held for 72 hours at 540°C (see Figures 5.14 and 5.15). For clarity, the discussion of these results, will be divided into a series of four relevant comparisons between testing conditions (i.e. fast cast, 1 hour versus fast cast, 72 hours; fast cast, 1 hour versus slow cast, 1 hour; etc.). A summary of the trends will be made at the end of this section.

6.2.1.1 Fast Cast 1 Hour vs. Fast Cast 72 Hours

The significant drop in average flow stress upon increasing the solutionizing time from 1 hour to 72 hours in the fast cast material (see Figure 5.14) is attributed primarily to a decrease in the generation of geometrically necessary dislocations. The coarsening of the spherical particles caused an increase in the geometric slip distance, \( \lambda_G \), of more than 60%, from 12 \( \mu \text{m} \) to 20 \( \mu \text{m} \) (as determined from \( \lambda_G = \frac{r}{f} \), where \( r \) is the particle radius and \( f \) is the volume fraction of reinforcement). Both of these slip distances are short enough (\( \leq 20 \mu \text{m} \)) to create a density of geometrically necessary dislocations which can dominate flow behavior, particularly at small strains (see Figure 2.12). However, the increase in geometric slip distance caused by particle coarsening results in a drop in the geometrically necessary dislocation density of approximately 40% at a given shear strain. The decrease in generation of geometrically necessary dislocations with coarsening is also reflected in Figure 6.3 where the work hardening rates of 1 hour and 72 hour compression samples (where damage is negligible) are plotted together. Note the reduced work hardening rate in the 72 hour sample. The high initial work hardening rate in the
one hour specimens is responsible for the increase in yield stress observed. The two curves converge at higher strains indicating that the statistically stored dislocations are then dominating flow and hardening behavior rather than the geometrically necessary dislocations.

![Work Hardening Rate as a Function of Strain for Fast Cast Material Solution Treated for 1 or 72 Hours at 540°C](image)

**Figure 6-3** Work hardening rate as a function of strain in fast cast 1 hour and 72 hour specimens.

The other dislocation based strengthening mechanism introduced in the literature review, Orowan looping, has very little effect in this system. The strength increase, given by equation 2.3, is inversely proportional to the particle spacing, L, which changes from 7 μm after 1 hour at 540°C, to 12 μm after 72 hours. Assuming a matrix shear modulus of
26.5 GPa, a Burgers vector of 0.286 nm and α equal to 1, the strength increase due to Orowan looping for a particle spacing of 7 µm is only 1.09 MPa. This value drops to only 0.64 MPa for a spacing of 12 µm. Relatively speaking, this is only a minor contribution to the overall strength of the material, and the 0.45 MPa drop in Orowan strengthening is insignificant to the overall flow stress drop obtained by increasing the heat treatment time.

The relatively low aspect of the ratio silicon particles in these two materials (1.8 and 1.7 for the 1 hour and 72 hour groups, respectively) make them somewhat inefficient in strengthening the alloy by means of load transfer, as shown schematically in Figure 2.9. Nonetheless, load transfer still occurs in these materials. The drop in aspect ratio with increasing solutionizing time is only 7% and the resulting drop in load transfer efficiency is considered to be minor compared to the overall flow stress change.

6.2.1.2 Slow Cast 1 Hour vs. Slow Cast 72 Hours

As shown in Figure 5.14, a drop in flow stress with increasing solutionizing time also occurred in the slow cast materials. This drop is attributed to a reduction in both the load transfer capability and the generation of geometrically necessary dislocations.

The reasonably high aspect ratio (>3) plates of silicon present in these materials (as seen in Figures 5.2b and 5.3b) are quite efficient at strengthening the system through load transfer from the matrix. Again referring to Figure 2.9, the more elongated the particle, the more efficient it is in strengthening the composite material, regardless of its orientation to the axis of loading. Thus, the 25% decrease in aspect ratio with particle
spherodization is considered to be significantly, though not solely, responsible for the drop in flow stress.

Increasing the solutionizing time from 1 hour to 72 hours also caused an increase in the geometric slip distance. This slip distance could not be calculated from the image analysis results because of the platelike structure of the silicon. However, the geometric slip distance for plate reinforcement is assumed to be equal to the spacing between the plates. This value was estimated by averaging the length of the shortest straight line between a selected particle and all of the surrounding particles which could be reached, without touching an intermediary particle. This process is shown schematically in Figure 6.4. From these measurements, a geometric slip distance of 19μm was found for the 1 hour material and \( \lambda_G \) equaled 27 μm for the 72 hour material. The slip distance in the one hour material is considered to be small enough (\( \lesssim 20 \mu m \)) to create a dominant strengthening effect. However, the increase in slip distance due to particle coarsening after 72 hours, reduces the generation of geometrically necessary dislocations to a point where they may be quickly swamped by statistically stored dislocations at low strains. This reduction in dislocation strengthening is assumed to be partially responsible for the drop in average flow stress with increasing solutionizing time.
Figure 6-4 Method to determine the particle spacing for the slow cast (a) 1 hour and (b) 72 hour materials.
Because of the platelike morphology of the silicon phase in the slow cast material, an effective particle spacing, L, could not be calculated. However, it was shown in the previous section that the Orowan looping contribution to strength of the fast cast material was relatively insignificant. Knowing that the particle spacing is even greater in these materials, it can be assumed that Orowan strengthening is even lower and still insignificant.

### 6.2.1.3 Slow Cast vs. Fast Cast

The difference in flow stress between the two casting techniques, is less pronounced than the difference created by changing heat treatment times, as shown by Figures 5.16 and 5.17. Nonetheless, the average flow stress for the slow cast material is higher than that for the fast cast material after 1 or 72 hours at 540°C. This is likely due to enhanced load transfer in the slow cast material. The fine, spheroidal particles of the fast cast material cause a large density of geometrically necessary dislocations but, are reasonably inefficient at load transfer. On the other hand, the plates of the slow cast material are effective at taking up load from the matrix but, after 1 hour at 540°C are still closely spaced enough to generate a significant density of geometrically necessary dislocations. It is interesting to note that despite quite different particle sizes and morphologies, the fast and slow cast materials after 72 hours at 540°C have almost identical yield and ultimate tensile strengths. They do differ, however, in their average strain to failure by almost 85% (see Section 6.2.2. for more details).
6.2.1.4 Summary of Strengthening

It has been shown that both dislocation based and continuum plasticity strengthening mechanisms are operating in these materials, though it is difficult to determine the absolute magnitude of either. Both load transfer and the generation of geometrically necessary dislocations are hindered by spherodization and coarsening of the structure. The shape change imparted by break-up of the as-cast structure, reduces the effectiveness of the silicon in load transfer. Similarly, coarsening of the reinforcement phase increases the particle spacing (i.e. geometric slip distance) which reduces the generation of geometrically necessary dislocations. Thus, the flow of aluminum-silicon alloys can be said to be a function of the size and shape of the silicon particles.

6.2.2 Damage in the Samples

It was shown that damage evolution in tension(almost exclusively in the form of particle cracking) was equal to or greater than it was in compression for all specimens (see Figures 5.26 through 5.29). Damage generally accumulated slower in the fast cast materials than in the slow cast materials, particularly the fast cast 72 hour samples and the fast cast 1 hour samples from the bottom of castings. These specimens displayed little difference between tension and compression until just before the onset of failure. Damage was negligible in the fast cast compression samples but, some elongated particle cracking was observed in the slow cast compression samples.
The cracking or decohesion of silicon particles is a catastrophic stress relaxation process which competes against work hardening (recall equation 2.8) during plastic deformation.

\[
\frac{d\sigma}{d\varepsilon} = \frac{\partial\bar{\sigma}}{\partial \varepsilon} (1 - D) - \bar{\sigma} \frac{\partial D}{\partial \varepsilon} 
\]

eqn. 2-8

The effect of damage on the work hardening rate is shown in Figure 6.5, where the true stress and work hardening rate in both tension and compression versus true strain are plotted for a slow cast 72 hour specimen. The decreased work hardening rate in tension is caused by damage in the sample. Sample necking is assumed to occur when the work hardening rate is equal to the stress in the material as described by the Considere Criterion given by Equation 6.1.

\[
\frac{d\sigma_T}{d\varepsilon_T} = \sigma_T
\]

eqn. 6.1

The reduction of work hardening rate in the tensile sample due to damage causes this criterion to be met at a lower strain as can be seen in Figure 6.5. Noting that these materials exhibit very little or no post necking deformation, the damage causes a drop in the strain to failure.

As discussed in Section 2.3.2, the occurrence of this damage is expected to be less frequent in compression than in tension, because of the lack of a flaw opening force. This appears to be confirmed by the results shown in Figures 5.26 and 5.27 (slow cast 1 hour and 72 hours, respectively) and most of the samples in Figure 5.28 (fast cast, 1 hour). However, the fast cast, 1 hour specimens from the bottom of the castings and most of the fast cast 72 hours specimens (with the exception of two rather anomalous sample pairs)
Figure 6-5  True stress and work hardening rate as a function of true strain, showing the satisfaction of the Considere criterion for slow cast 72 hour specimens in tension and compression.
showed little damage (i.e. difference between compressive and tensile stress at a given strain) until just before failure. To investigate this difference further, it is useful to examine the particles in the fractured tensile specimens.

Figures 6.6 through 6.9 show the particles at the fracture surface for slow cast, 1 hour and 72 hour and fast cast, 1 hour and 72 hour samples respectively. The slow cast specimens in Figures 6.6 (1 hour) and 6.7 (72 hours) show particle cracking, particularly along the fracture surface. This is not surprising, as Weibull statistics suggest that the larger particles will fail (i.e. crack) before smaller ones because of the increased probability for the presence of a flaw of critical size. While the actual number of particles cracked is not very high, the large particle size makes these significant damage events. That is, fewer of these particles would have to crack for an equivalent drop in load carrying capacity, than in a system of fine particles such as the fast cast, 1 hour specimen shown in Figure 6.8.

Note in Figure 6.8, the extent of particle cracking just below the fracture surface. The evolution of this much damage occurs over a greater strain than occurs in the slow cast specimens, as evidenced by comparing Figures 5.26, 5.27 and 5.28 which show the difference in compressive and tensile stresses as a function of strain. It is proposed that the two bottom samples did not show significant damage until near failure, unlike the other samples, because of a smaller range of particle size. The samples further away from the chill in the fast castings were prone to formation of rosettes (as described in section 5.1.2) which inherently widens the distribution of particle sizes. It is assumed that the flakes of silicon which form the outer ring of a rosette would crack much sooner than the
fine particles of the rest of the sample because of their larger size. Consequently, samples
with a wide range of particle sizes will accumulate damage much sooner than a sample
with a narrow size range, when the means are the same. Once the critical stress for the
particle size is reached, damage accumulates very rapidly until failure.

An example of the fast cast 72 hour specimens which displayed negligible damage
evolution difference between tension and compression, is shown in Figure 6.9. Note, the
limited formation of cracks across the particles, especially compared to the fast cast 1
hour specimen (Figure 6.8). There is cracking within the intermetallic β phase needles
near this surface and some voids present near the particles, but little particle cracking.
Notice how the actual fracture surface tends to move around the particles rather than
through them as was observed in the slow cast specimens. The fact that these particles
exhibited less cracking than the smaller particles found in the fast cast one hour
specimens is somewhat surprising, given that Weibull statistics suggest that the larger
particles have a greater probability of failure. However, the increase in particle size with
increasing heat treatment time also brought about a drop in the composite stress as shown
in the previous section. This reduced stress meant that the particles, at a local level,
experienced a lower stress level. This allowed the particles to remain intact during
loading to failure. It is interesting to note that some particle cracking was observed away
from the fracture surface, as shown in Figure 6.10 though, it was not manifested in the
differences between the tensile and compressive curves.
Figure 6-6  High magnification view of fracture surface in slow cast, 1 hour specimen.

Figure 6-7  High magnification view of fracture surface in slow cast, 72 hour specimen.
Figure 6-8 High magnification view of fracture surface in fast cast, 1 hour specimen.

Figure 6-9 High magnification view of fracture surface in fast cast, 72 hour specimen.
Figure 6-10 Some particle cracking observed away from the fracture surface in a fast cast, 72 hour specimen.

6.2.2.1 Comparison of Damage With Particle Cracking Model

Brockenbrough and Zok\textsuperscript{43} developed a simple analytical model from a series of finite element method calculations to show the effect of particle cracking on the flow response of metal matrix composites reinforced with spherical particles. They assumed the matrix uniaxial response to follow the Ramberg-Osgood relation given by equation 6.2\textsuperscript{43}
\[
\frac{\varepsilon}{\varepsilon_*} = \frac{\sigma}{\sigma_*} + \alpha \left( \frac{\sigma}{\sigma_*} \right)^{\frac{1}{N}}
\]

eqn. 6.2

where \(\varepsilon\) and \(\sigma\) are the axial strain and stress, \(\alpha\) is a numerical constant assumed to be \(3/7\), \(\sigma_*\) is a reference stress (considered to be the matrix yield stress), \(N\) is the hardening exponent, and \(\varepsilon_* = \sigma_* / E_m\) where \(E_m\) is the elastic modulus of the matrix. They fit their results to yield Equations 6.3 for a system with no cracked particles.

\[
\sigma_{N,0}(\varepsilon) = \sigma(\varepsilon) \left[ 1 + (N) \tan \left( \frac{3\pi f}{4} \right) + (N) f^2 \right]
\]

eqn. 6.3

where \(\alpha(N)\) and \(\beta(N)\) are empirically determined coefficients linearly related to the hardening exponent, \(N\) as follows:

\[
\alpha(N) = 0.15 + 0.913N \quad \text{eqn. 6.3a}
\]

\[
\beta(N) = 2.86 + 21.4N \quad \text{eqn. 6.3b}
\]

In equation 6.3 \(\sigma_{N,0}(\varepsilon)\) is the composite stress for a material with hardening coefficient, \(N\) and a volume fraction of reinforcement, \(f\), of which no particles are cracked. The stress-strain response of the work-hardening matrix in the absence of reinforcement, is given by \(\sigma(\varepsilon)\) and can be determined from equation 6.2.

Similarly, the equation from Brockenbrough and Zok\(^{43}\) for a composite in which all the particles are cracked, is given by Equation 6.4:

\[
\sigma_{N,1} = \sigma(\varepsilon) \left[ 1 - \gamma(N)f - \xi(N)f^2 \right]
\]

eqn. 6.4

where the empirically determined constants are given by:

\[
\gamma(N) = 0.916 - 1.61N \quad \text{eqn. 6.4a}
\]

\[
\xi(N) = 0.01 + 0.828N \quad \text{eqn. 6.4b}
\]
Equations 6.3 and 6.4 can then be used to plot the upper and lower bounds for the stress-strain response for a composite composed of a matrix of work hardening exponent, \( N \) and a reinforcing phase of spherical particles present in a volume fraction of \( f \).

These equations were applied to the results from the fast cast 1 hour and 72 hour specimens. The value of \( f \) for each case was taken as the area fraction measurement from image analysis results in Table 5.1. The reference stress, \( \sigma_* \), was taken to be 42 MPa which is similar to that used by Kiser, et al.\(^{48} \) in their study of particle cracking in Al-Si alloys. The value of \( N \) was then varied for equation 6.2 (no cracked particles) to fit the highest strength compressive result for the group of samples being examined. The strain hardening exponent was found to be 0.23 for the fast cast, 72 hour specimen and 0.28 for the 1 hour specimens. This increase in strain hardening exponent is attributed to the increase in generation of geometrically necessary dislocations described above. The bounding curves (i.e. fully intact and all particles cracked) are plotted with all tensile and compressive results for fast cast 72 hours and 1 hour in Figures 6.11 and 6.12 respectively. In Figure 6.11 all of the measured curves fall between the calculated bounds of zero and full particle cracking. This is to be expected, as very little particle cracking was observed in these specimens. In Figure 6.12, the experimental curves initially follow the calculated bounds. However, they eventually fall below the model line for total particle cracking. This, of course, is not realistic as many of the particles visible in Figure 6.9 are still intact. This is simply a limitation of the power law used to generate the curve in this model. Relaxation processes are also competing with the initially high work hardening of the alloy. The model of Brockenbrough and Zok\(^{43} \) is
Figure 6-11 Experimental fast, 72 hour tensile and compressive stress-strain curves compared with model curves for zero and full particle cracking.

Figure 6-12 Experimental fast, 1 hour tensile and compressive stress-strain curves compared with model curves for zero and full particle cracking.
based on continuum plasticity and does not account explicitly for relaxation of dislocation build-up. Still, this model is useful in showing the general effect of particle cracking and indirectly highlighting the change in work hardening rate between the fast cast 1 hour and 72 hours specimens. This model could not be applied to the slow cast material, as the empirical calculations are based on spherical particles. However, it is expected that the trends observed would be the same.

Figure 6.13 again shows the model curves for zero and full particle cracking in the fast cast, 72 hour material. The matrix stress-strain curve as determined from equation 6.2 with the same hardening exponent ($N=0.23$) used for the model curves is also shown along with an experimental compressive curve. At a given strain, the stress in the composite (i.e. the experimental curve) and the stress in the matrix can be found. By using these values in Equation 6.5 an estimate of the stress borne by the silicon phase can be made.

$$\sigma_{Al-Si} = f\sigma_{Si} + (1-f)\sigma_{Al}$$

In equation 6.5, $f$ is the volume fraction of reinforcement, $\sigma_{Si}$ is the stress borne by the silicon phase, $\sigma_{Al}$ is the stress in the aluminum matrix and $\sigma_{Al-Si}$ is the overall stress in the sample. At 10% true strain, the composite stress is 176 MPa and the matrix stress is only 164.5 MPa. Assuming a silicon reinforcement volume fraction of 0.1123, yields a stress in the silicon phase of 267 MPa. This is obviously much less than the calculated yield strength (3090 MPa) and even the estimated fracture strength (726 to 1608 MPa) which further illustrates the brittle and statistical nature of the behavior of the silicon phase.
Figure 6-13  Model stress-strain curves for matrix, fully intact and fully cracked composite and experimental curve for fast cast, 72 hour compression sample.
6.2.3 Fracture

It was noted that the elongation to failure in the slow cast material was considerably lower than in the fast material. Similarly, spherodization and coarsening of the sample improved the ductility of castings produced using either technique. The fracture surfaces of the slow cast materials, tended to be reasonably brittle in appearance, with many large, cleaved plates of silicon visible. This compared to the ductile, dimpled surface of the fast cast specimens.

Examining the fracture surfaces from Chapter 5 and the high magnification views of particles shown in the previous section (Figures 6.6 and 6.7), it is noted that the fracture path in the slow cast materials tended to follow the long plates of silicon, generally those oriented perpendicular to the tensile axis. It is proposed that link-up of these cracks led to final failure of the specimens. However, by increasing the mean free distance between the plates through the spherodization and particle coarsening imparted by increasing solution treatment time, a less continuous brittle crack path was available for fracture. This, helps to explain the increase in average elongation to failure and the increase in elevation changes in the fracture surface observed with the increasing heat treatment time. The lower strength of the material after 72 hours also improves the ductility.

The fast cast specimens displayed much more dimpled fracture surfaces than the slow cast specimens. This would indicate ductile failure by microvoid nucleation, growth and coalescence, as described in the literature review. It is interesting to note that two of
the fast cast 1 hour specimens (both from the bottom of their casting) had relatively high strains to failure compared to the rest of the samples. This is proposed to be the result of a more uniform structure, as discussed in the preceding section. The large brittle plates of silicon associated with the rosettes higher in the casting were likely damaged early in the loading, thereby inducing large voids into the sample. Figures 6.14 (a) and (b) show two different regions of the fracture surface for a fast cast, 1 hour sample which contained rosettes. The majority of the fracture surface appeared like Figure 6.14 (a) while (b) shows the cracking along a large plate in the rosette. The bottom level castings were not subject to this damage and remained relatively intact until later in the loading cycle.

Increasing the solution treatment time from 1 to 72 hours also improved ductility in the fast cast materials. The particles are larger after coarsening making the probability of the presence of a flaw of critical size greater but, the associated drop in the stress on the system negated this effect somewhat. This is also reflected in the increased particle spacing and the corresponding increased in measured dimple size.

To summarize, the size and the morphology of the silicon particles play an important role in the fracture behavior of these materials. Samples with large, elongated plates are very prone to brittle fracture because of the high stress associated with load transfer and also an increased probability of the presence of a critically sized flaw. Damage is thus, more likely to occur and induce voids and cracks into the sample. The plate structure also provides an almost continuous brittle crack path for failure to occur. Spherical reinforcement is less likely to be damaged because of low load carrying capacity and the reduced stress concentrations associated with the shape. Also, spherical
particles do not provide a long, brittle crack path for final fracture to occur, instead requiring a crack to pass a greater distance through the ductile aluminum matrix. This is the primary reason why the fast cast 72 hour specimens had a strain to failure nearly 85% higher than their slow cast counterparts, despite having nearly identical strength properties.
Figure 6-14  Particle cracking in fast cast, 1 hour specimen near the top of the casting at (a) regular region of casting and (b) in region of long plates due to rosette.
7. Summary and Conclusions

The effect of silicon particle size and morphology on the flow and fracture of cast aluminum silicon alloys of eutectic composition was examined in this work and led to the following observations.

The castings contained silicon particle sizes in the region where neither continuum plasticity nor dislocation based models can fully predict deformation behavior. It was observed that both continuum and dislocation effects were affecting the behavior of the materials.

The flow behavior was dependent on the length scale of the silicon phase which is, in turn, dependent on the size and shape of the silicon particles. Larger, acicular plates of relatively high aspect ratio, such as those found in the slow cast materials, were more effective in terms of strengthening the alloys through load transfer than the more refined particles in the fast cast materials. The distribution of fine, spherical particles produced through modification and a fast cooling rate, were more effective in strengthening the material through dislocation mechanisms such as the generation of geometrically necessary dislocations. However, both continuum and dislocation effects were occurring in both the fast and slow cast materials.

Increasing the length scale (i.e. interparticle spacing and size of particles) in the system by means of particle spherodization and coarsening, resulting from prolonged solution treatments at 540°C, caused a drop in the flow stress and an increase in elongation to failure for both the fast and slow cast materials. The spherodization
reduced the average particle aspect ratio, thereby lessening the effectiveness of the particles for load sharing with the matrix, particularly for the slow cast materials. The coarsening of the particles also caused an increase in the average geometric slip distance which reduces the effectiveness of strengthening due to dislocation mechanisms.

Damage, mostly by means of particle cracking, was observed to occur with greater frequency in the tensile samples than in the compressive samples. Damage evolved much slower in systems with smaller, spherical particle reinforcement than those with large, angular particle reinforcement. Damage was also suppressed when exposed to lower stresses in the system. Damage in the form of particle cracking caused a decrease in both the elongation to failure and the work hardening rate of the material. General trends in flow behavior affected by damage agree with the predictions of a simple model of particle cracking.

Fracture in slow cast, unmodified material was mostly brittle. Large cleaved plates of silicon were a prominent fracture surface feature. In the fast cast, modified materials, fracture was more ductile, exemplified by a dimpled fracture surface. This indicates failure by means microvoid nucleation (possibly by means of particle cracking) growth and coalescence. The increase in average particle spacing brought about by increased solutionizing time also increased the mean free path between brittle particles, which helped to increase the ductility of both fast and slow cast materials. This is buoyed by the fact that the stresses in the system were also lower.
8. Future Work

The aluminum-silicon system studied in the present work is a model system for examining the effects of changes in length scale on the flow and fracture of two-phase materials. Further experiments are suggested wherein the smallest silicon particle size is decreased even further with extremely rapid cooling conditions such as splat cooling. This would allow for examination of the lower end of the range (~0.1 μm) of particle sizes which display both dislocation and continuum plasticity effects in strengthening the system. The length scale might also be changed by varying the volume fraction of the silicon reinforcement. It would also be of interest to test samples in a near as-cast state, with a very short solutionizing treatment to normalize the matrix but, not long enough to impart spherodization or coarsening. This would allow for a more continuous silicon phase, particularly in the modified alloys (which had the seaweed structure), which may also affect the fracture properties.

The results also show that damage plays a role in the mechanical behavior and fracture of these alloys. It would be of interest to strain these materials to induce damage without sample failure, recrystallize the matrix through a short heat treatment and then re-test the material, to examine the effect of pre-existing damage on the mechanical behavior. By delaying the onset of damage through the introduction of damage in a previous processing step, it may be possible to increase the ductility of the sample.

Finally, in order to appreciate the magnitude of the strengthening effects due to the silicon phase in aluminum-silicon casting alloys, it is suggested that similar tests be
run on alloys with additions of magnesium so that the strength of the matrix may also be varied. This would be of more commercial importance because of the limited application of binary aluminum-silicon casting alloys.
Appendix A

This appendix presents the individual results of the image analysis, porosity measurements and tensile testing in three tables.
### Table A-1: Individual Results of Image Analysis

<table>
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<th>Position</th>
<th>MAX_LENGTH</th>
<th>MAX_BREADTH</th>
<th>ASPECT_RATIO</th>
<th>AREA(TOT)</th>
<th>EQUIV CIRCLE DIAMETER</th>
<th>AREA FRACTION</th>
</tr>
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<td></td>
<td>(microns)</td>
<td>(microns)</td>
<td>(microns²)</td>
<td>(microns)</td>
<td>(microns)</td>
<td>(%)</td>
</tr>
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<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
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<td></td>
<td></td>
<td></td>
<td></td>
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<td>F1T V1 Bottom</td>
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<td>1.107835</td>
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<td>2.904914</td>
<td>1.92318761</td>
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<td>1.829</td>
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<tr>
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<td>1.74845</td>
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<td>4.55228475</td>
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<tr>
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<td>2.578307</td>
<td>1.711165</td>
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<td>4.164527037</td>
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<td><strong>SLOW, UNMODIFIED AS CAST</strong></td>
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<td></td>
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<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
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<td>2.877442</td>
<td>4.208467</td>
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<tr>
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<td>2.904541</td>
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<td>67.360454</td>
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<td>12.305279</td>
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<td><strong>AVERAGE</strong></td>
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<td>2.817</td>
<td>4.447</td>
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<td></td>
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<td><strong>Position</strong></td>
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<td>3.399</td>
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Table A-2 Porosity Measurements for Each Casting

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<tr>
<th>Porosity measurements</th>
<th>Casting</th>
<th>Number of Pore</th>
<th>Pore Density</th>
<th>Area Fraction (%)</th>
<th>Average Pore Size (mm(^2))</th>
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<tr>
<td>Slow</td>
<td>SUE_G</td>
<td>54</td>
<td>0.270</td>
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<td>SUE_H</td>
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### Table A-3 Individual Results of Tensile Testing

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<thead>
<tr>
<th>Grouping</th>
<th>Position</th>
<th>Name</th>
<th>Yield Strength (MPa)</th>
<th>Ultimate Tensile Strength (MPa)</th>
<th>Strain to Failure (%)</th>
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</thead>
<tbody>
<tr>
<td>Slow Cast, 1 hour</td>
<td>Bottom</td>
<td>S1T_L1</td>
<td>60.8</td>
<td>140.1</td>
<td>1.7</td>
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<tr>
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<td>Bottom</td>
<td>S1T_J1</td>
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<td>157.3</td>
<td>2.3</td>
</tr>
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<td></td>
<td>Middle</td>
<td>S1T_H2</td>
<td>70.9</td>
<td>154.0</td>
<td>2.2</td>
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<td></td>
<td>Top</td>
<td>S1T_H3</td>
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<td>170.1</td>
<td>3.6</td>
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<tr>
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<td>Top</td>
<td>S1T_J3</td>
<td>62.0</td>
<td>164.6</td>
<td>3.3</td>
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<td></td>
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<td>S1T_L3</td>
<td>76.2</td>
<td>171.1</td>
<td>3.5</td>
</tr>
<tr>
<td>Slow Cast, 72 hours</td>
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<td>S72T_H1</td>
<td>64.0</td>
<td>139.8</td>
<td>3.2</td>
</tr>
<tr>
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<td>Middle</td>
<td>S72T_J2</td>
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<td>150.3</td>
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<td>156.3</td>
<td>5.0</td>
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<td>10.3</td>
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<td>S72T_G3</td>
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<td>156.5</td>
<td>10.2</td>
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<td>5.6</td>
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<td>F1T_V1</td>
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<td>10.8</td>
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<tr>
<td></td>
<td>Middle</td>
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<td>76.5</td>
<td>168.9</td>
<td>4.8</td>
</tr>
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<td>4.4</td>
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References


7. Chalmers, B., Principles of Solidification, John Wiley and Sons, New York (19??).


Ashby, M.F. et al., Cambridge Materials Selector V. 2.02, Granata Design Ltd., 1994.


