EFFECT OF MINOR ELEMENTS ON CASTABILITY, MICROSTRUCTURE AND MECHANICAL PROPERTIES OF RECYCLED ALUMINIUM ALLOYS

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Effect of Minor Elements on Castability, Microstructure and Mechanical Properties of Recycled Aluminium Alloys

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DECLARATION

This thesis is my original work and has not been presented for a degree in any other University

Signature……………………………… Date…………………………..

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This thesis has been submitted for examination with our approval as University Supervisors.

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Signature……………………………… Date:………………….

Mr. Thomas O. Mbuya
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DEDICATION

To my parents,

Mary and Mose Ondieki

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LIST OF ABBREVIATIONS

BHN	Brinell Hardness Number
CH	Cylinder heads
LSM	London & Scandinavian Metallurgical Co.
OM	Optical Microscopy
PS	Pistons
SEM	Scanning Electron Microscope
TEM	Transmission Electron Microscopy
WLS	Wheels
UON	University of Nairobi

NOMENCLATURE

β	Beta phase
α-Al	Aluminium matrix
Al	Aluminium
Be	Beryllium
Ca	Calcium
Cr	Chromium
Cu	Copper
Fe	Iron
K	Potassium
Mg	Magnesium
Mn	Manganese
Na	Sodium
Ni	Nickel
P	Phosphorus
Pb	Lead
PS	Pistons
P_g	Sum of the pressure of dissolved gases (MPa)
P_s	Sum of pressure due to shrinkage (MPa)
P_h	Sum of pressure due to metallostatic head (MPa)
P_{s-t}	Surface tension (N)
P_{atm}	Atmospheric pressure (MPa)

ρ_{exp}	Measured density (g/cm^3)
ρ_{th}	Theoretical density (g/cm^3)
Sb	Antimony
Si	Silicon
Sn	Tin
Sr	Strontium
Ti	Titanium
T_p	Pouring temperature (˚C)
UTS	Ultimate Tensile Strength (MPa)
V	Vanadium
Zn	Zinc
Zr	Zirconium

ABSTRACT

As part of the wider effort to improve the performance of cast aluminium alloys, this research set out to investigate the effect of minor elements (Fe, Mn, Cr, Sr, Sb and Ti-B based grain refiners) on castability (fluidity and porosity formation), microstructure and mechanical properties (UTS, percentage elongation, impact strength and Brinell hardness) of recycled aluminium alloys. A survey of the local aluminium foundries was also conducted to establish the prevailing aluminium casting practices such as recycling methods, casting design, melt treatment and quality control aspects, and whether these practices are effective in making quality castings.

Fluidity tests were conducted using CO_2/sodium silicate bonded spiral sand moulds. Fluidity was indicated by the length the molten metal ran before being stopped by solidification. Specimens for mechanical tests and porosity analysis were prepared from castings obtained from a permanent mould. For tensile tests, specimens were prepared from the sound section in accordance to the ASTM B108-82b while those for impact tests were prepared in accordance to ASTM E-23 standards. Hardness test specimens were polished using fine SiC paper to remove any machining marks on the surface before indentation. Fracture surface specimens were prepared in accordance to ASTM E-8M standards to help reveal features contributing to failure during tensile tests. Metallographic specimens were also polished using SiC papers with a fine polish being performed on grinding wheels using 0.25µm diamond paste, and to aid in identification of the phases present in the microstructure, the samples were deep etched in a solution of 10% NaOH in distilled water. A T6 heat treatment procedure was performed on specimens for mechanical tests to enable make comparison with as cast specimens. Porosity analysis was performed using density measurement method. To collect data for the survey, a questionnaire was designed with test items being administered both orally and in some cases foundrymen were requested to fill them.

Results from fluidity tests indicated that fluidity increased by 21% when Fe was raised to the critical content of 0.48% for LM25. A combination of 0.3%Mn or 0.6%Cr with 0.6%Fe in LM25, resulted in a fluidity increase of 13% and 8%, respectively compared to the base alloy, but a combination of 0.6%Fe, 0.3%Mn and 0.2%Cr decreased the fluidity by 9%. A 34% increase in fluidity was observed when the Fe content in LM27 was raised from 0.41% to the critical level of 0.6%Fe with further increase when Mn was raised to 0.3%Mn. Addition of 0.015%Sr and 0.02%Sr increased the fluidity of LM25 and LM27 by 9% and 21% respectively. Furthermore, a 0.28% Al-5Ti-1B grain refiner addition decreased the fluidity of LM 25 and LM27 by 2% and 19% respectively.

The results from porosity analysis indicated that the volume percent porosity in LM13-type base alloy was 1.6%. With addition of Sr at levels of 0.02% and 0.05% the volume percent porosity increased to 2.9% and 2.6% respectively. When individual additions of 0.53%Mn and 1.06%Cr were made to the alloy, the volume percent porosity reduced to 0.6% and 0.4% respectively. It was further noted that a

combined addition of 0.3%Mn together with 0.2%Cr reduced the volume percent porosity of this alloy to 1.0%. Other reductions in porosity were observed when 0.53%Mn was added in combination with each of 0.02%Sr, 0.05%Sr and 0.2%Sb. Addition of 0.02%Sr together with 0.28%Al-5Ti-1B grain refiner slightly reduced the volume percent porosity of the base alloy to 1.4%.

Microstructural results showed that the as cast microstructure of LM13-type alloy consisted of a structure with coarse Si particles, large AlCuNi phases in addition to Al_2Cu phases on the α-Al matrix. Addition of 0.02%Sr to the alloy led to partial modification of the eutectic Si particles while addition of 0.05%Sr led to full modification of the eutectic Si. Separate additions of 0.53%Mn and 1.06%Cr transformed the flake-like AlCuNi structures to Chinese script morphology. It was also noted that the effect of heat treatment was spheroidization and coarsening of the Si particles. At the same time, there was fragmentation of intermetallic particles.

Results from UTS and % elongation revealed that with addition of Sr at levels of 0.02% and 0.05% the average values of UTS increased by 4% and 5% respectively in as cast while it increased by 6% and 8% in T6. Moreover, with the 0.02%Sr and 0.05%Sr addition the % elongation increased by 22% and 17% in as cast and by 24% for each level of Sr in T6. When individual additions of 0.53%Mn and 1.06%Cr were made, improvements of 13% and 18% in the UTS were recorded in as cast condition while in T6 condition the improvements were 15% and 16%. With these additions,

the % elongation of the alloy increased by 35% and 61% in as cast and by 38% and 43% in T6.

The results from hardness and impact tests indicated that addition of Sr improved the Brinell hardness and impact strength of the LM13-type alloy in both as cast and T6 condition. On the other hand, improvements in these properties in as cast and in T6 condition were recorded when individual additions of 0.53%Mn and 1.06%Cr were made to the alloy. Furthermore, improvements in the Brinell hardness and impact strength of the alloy were recorded when each of 0.2%Sb, 0.02%Sr and 0.05% Sr was combined with 0.53%Mn both in as cast and in T6 condition.

Results from the survey demonstrated that nearly all the foundries visited operate at about 40% capacity utilization. In addition, control of process parameters was limited, with methods being non competitive and quality control aspects being hardly adequate. Use of alloying elements (additives) as a means to improve properties of castings was seldom utilized, leave alone knowledge of existence of such additives. Out of 45 companies visited 98% were not using additives to control the properties of castings. This in turn resulted in production of low quality and unreliable castings. Noting that some companies visited supply their products to the international market, the majority would find themselves disadvantaged if it turns out that their products need to meet certain standards that call for use of additives. It was therefore recommended that collaborative effort be made between research institutions and industry as well as the government as one of the basic steps towards improvement of the casting practice and enhancement of capacity utilization in local foundries.

Overall the results indicate that use of minor alloying elements can have significant effects on castability, microstructure and mechanical properties of recycled Al alloys. Addition of Fe to the critical Fe level to LM25 and LM27 type alloys gave higher fluidity results compared to Fe levels below and above this level. It was also established that modification due to 0.02%Sr addition was not the same as that due to 0.05%Sr addition. A 0.02%Sr addition caused partial modification while 0.05% caused full modification. Addition of high levels of Cr (not applied by most researchers in the past for fear of sludge formation) to LM13 type alloy led to very superior mechanical properties of the alloy. Micrographs from samples with Cr addition did not show formation of sludge. A survey of the local foundries in the major towns of Nairobi, Mombasa and Nakuru indicated that the recycling methods, casting design, melt treatment and quality control used in these foundries were hardly adequate to ensure production of premium castings.

CHAPTER ONE

1.0 INTRODUCTION

As a key trend manufacturers are changing material for engine blocks, pistons and others from cast iron to Al alloys so as to significantly reduce the overall weight of automobiles and aircrafts. This has made Al alloys to be an attractive engineering material compared to other materials. With the escalation oil prices and environmental concerns Al alloys remain the preferred material in the automotive and aerospace industries. The section that follows introduces the cast Al alloys that will be of concern in this study.

1.1 Cast Aluminium Alloys

Aluminiun scrap is the common raw material in the commercial aluminium foundry industry. This scrap is obtainable from a number of sources including the automotive and aerospace industries. In the local foundries, automobiles are the greatest source of aluminium scrap where it is obtained from a wide range of components such as pistons, cylinder heads, engine blocks, oil sumps, wheels and manifolds. These components are usually manufactured from several Al-Si based cast alloys such as the A356/A357 (Al7Si0.25-0.7Mg) alloys which are popularly used for making wheels and to a less extent cylinder heads and blocks [Brown (1999); Hatch (1987)]. The 319-type alloys (Al6Si3Cu) are used for making parts such as cylinder heads and blocks, manifolds and engine mountings while the 380 (Al9Si3Cu)/383 (Al10Si3Cu) die casting alloys are popular for making gear and rear axle housings and other general engineering parts and the 332 (Al9Si3Cu1Mg1Zn) or 336 (Al12Si1Cu1Mg) are common for making pistons.

Al-Si foundry alloys are the most commonly used casting alloys particularly in the automotive and aerospace industries owing to their excellent casting characteristics and good mechanical properties than their competitors [Mbuya et al. (2007)]. At the same time, these alloys demonstrate fewer tendencies to shrinkage, hot tearing and porosity defects than most of the other cast alloys such as Al-Cu alloys [Gruzleski and Closset (1990)]. Furthermore, these alloys have excellent fluidity characteristics. The number of Al-Si casting alloys used for various applications is large and their general composition is as shown in Table 1.1 [Mbuya et al. (2003)].

Table 1.1: Composition limits for aluminium silicon alloys [Mbuya et al. (2003)].

Element	Composition range, Wt-%
Si	5-25
Cu	0-5.0
Mg*	0-2.0
Zn	0-3.8
Fe	≤3.0
Mn, Cr, Co, Mo, Ni†, Be, Zr	≤3.0 (Total)
Na, Sr	≤0.02 (Total)
P	≤0.01
Sn	0-0.25
Pb	0-0.35
Ti	0-0.25

*The composition limit of Mg is normally 1.0%.

†Compositions of Ni alone can be up to 2.5% in certain alloys e.g. for 336 alloy.

(Composition is in weight percent throughout the report)

The effects of major elements in the specific Al-Si based cast alloys are generally well known and documented. On the other hand, there is relatively little organized information on the effects and interactions of minor elements. An element is considered major or minor depending on the significance attached to its influence on the castability and/or properties of the specific alloy. For instance, Si and Mg are major elements in 356/357 type alloys and the rest are minor while Si and Cu are the major elements in 319 and 380 type alloys.

Silicon improves castability by reducing solidification shrinkage thus minimizing shrinkage porosity. It increases fluidity due to its high latent heat of fusion and reduces the propensity to hot tearing by reducing the alloy's thermal expansion coefficient [Caceres et al. (1997)]. Silicon also strengthens the alloy by a combination of dispersion, solid solution and precipitation hardening [Caceres et al. (1997)]. Copper and Magnesium on the other hand are usually added to increase strength mainly by precipitation hardening, but reduce ductility [Mbuya (2003)]. Copper also increases hot tearing and porosity formation [Mbuya (2003)].

Studies on the effect of alloying and impurity elements on castability and mechanical properties are necessary to optimize alloy chemistry and obtain improved properties of cast aluminium products. Understanding the effect of minor and impurity elements is particularly important when recycled material is used as impurity levels are usually higher.

Several contributions have so far been made towards this subject. Mondolfo (1976) reviewed the influence of elements on the properties of various aluminium alloy systems. Other recent reviews on the effect of elements include those of Taylor (1995), Mbuya (2006) and Wang et al. (1995).

The presence of Si leads to the formation of an Al-Si eutectic with coarse acicular shaped Si plates which limit the ductility of the alloy. To modify these particles to a fine fibrous structure and improve mechanical properties, certain elements such Sr, Na, and Ca are usually added; a phenomenon referred to as modification. Sb also refines the Si particles to a fine lamellar structure. Modification by Na addition is quick, but has poor recovery, its fumes are toxic, fades quickly, has inconsistent results and can lead to overmodification [Apelian et al. (1984)]. The use of Sr has therefore become more popular because it is non-toxic, has good recovery and has a semi-permanent modification effect. It is also less likely to lead to overmodification [Apelian et al. (1984)]. Ca and Sb have less modification efficiency and Sb can form toxic gases. The beneficial effects of chemical modification can, however, be masked by the tendency to increase porosity formation [Gruzleski and Closset (1990)]. Several explanations have been given as to why this happens, but none seems to be conclusive.

Iron is always present in commercial aluminium alloys and has consistently emerged as the main impurity element and perhaps the most detrimental to castability (mainly porosity) and the mechanical properties of these alloys as documented by a number of reviews [Couture (1981); Crepeau (1995); Mbuya et al. (2003)]. It is usually added inadvertently through the use of equipment that contain iron during melting and casting and/or through the use of scrap charges containing iron and/or rust. Iron is also added intentionally in some alloys, the most common being aluminium-copper-nickel group, to which it is added to increase high temperature strength. Iron is also added to die-casting alloys to mitigate die soldering. It has also been observed that Fe grain refines aluminium alloys besides aiding in the efficacy of Al-5Ti-1B grain refiners [Mbuya et al. (2003)]. For most aluminium alloys, however, the deleterious effects of Fe outweigh its benefits and efforts are made to keep its levels as low as is economically possible.

Iron has a low solid solubility in aluminium and segregates during solidification forming intermetallic compounds with Al and other elements such as Si, Mn, Cr, Be and Mg. Under certain conditions of processing and alloy composition, Fe can combine with other elements such as Mn and Cr to form primary intermetallics (sludge). The detrimental effects of iron are usually associated with the formation of intermetallic compounds. The most common of these compounds are the α-Al_8Fe_2Si and π-$Al_8FeMg_3Si_6$ with a Chinese script shape and the β-Al_5FeSi which is the most detrimental due to its needle-like shape. When Be, Mn, Co, Mo and Cr are present, then less deleterious Chinese script compounds tend to form in place of the β-Al_5FeSi phase. This is the basis of their neutralizing effect. The effect of Fe can also be ameliorated by addition of small quantities of other elements such as Sr and rare earth elements (La, Ce and Nd individually or in combination as mischmetal) [Mbuya et al. (2003)].

A fine equiaxed grain structure is commonly developed through grain refinement and is known to confer beneficial effects to both mechanical and processing properties of Al-Si alloys. It imparts a high yield strength, high toughness, good extrudability and uniform distribution of second phases and micro-porosity on a fine scale. Furthermore, it also improves machinability, surface finish, resistance to hot tearing and various other desirable properties [Murty et al. (2002)]. The effect of grain refiners on some castability concepts such as fluidity and feedability is however a bit confusing as there have been contradicting reports in the literature especially on fluidity (Mbuya, 2006).

1.2 Problem Statement

There is relatively little organized information on the effects and interactions of minor elements on castability, microstructure and mechanical properties of aluminium foundry alloys and yet such effects can be profound. This is a major weakness in the casting industry especially because it is largely dependent on scrap recycling which results in secondary alloys with higher impurity levels than the primary alloys. The problem is exacerbated by the fact that different alloys with different compositions are often used to produce similar parts depending on the product model and the manufacturer's requirements. For instance different alloys are used to produce automotive engine cylinder heads depending on the vehicle model and changing preference of the manufacturers. During recycling most of these parts are mixed together without regard to their differences in composition, as sorting of these parts may not be commercially viable. Efforts are then made to correct the composition of the resulting alloy on line. This practice also has some economic limitations. Furthermore, certain elements are either difficult and/or expensive to remove (e.g. iron and magnesium). The need for characterizing the effect of minor elements has generated extensive research interest in the area and has in part motivated the current study.

The need to improve the performance of the local aluminium casting industry is also a major driving force behind this study. It is easily observed that Kenya provides a market for aluminium castings especially automobile spare parts such as wheels and pistons. However, it is rare to find a locally manufactured premium aluminium casting in the market. It is even more unfortunate to note that most of the aluminium scrap available locally ends up being exported without any value addition. It is important to find out why local foundries are not effectively involved in aluminium casting.

1.3.Objectives

The objectives of this study are;

1.) To investigate the effect of some minor elements (Fe, Mn, Cr, Sr, Sb and Ti-B based grain refiners) on castability (fluidity and porosity formation), microstructure, and mechanical properties (tensile, impact and hardness) of cast aluminium alloys obtained from recycled automobile parts.

2.) To carry out a survey of the local foundries to establish the prevailing aluminium casting practices such as recycling methods, casting design, melt treatment and quality control aspects, and whether these practices are effective in making quality castings. Suggestions for improvement would be given where necessary.

1.4 Thesis Structure

This thesis contains six chapters of which this is the first. Chapter two contains a literature review which is divided into three main sections covering castability (fluidity and porosity formation), microstructure and mechanical properties of cast Al-Si alloys. Chapter three contains details of all the experimental methods used in the course of this project and the results and discussion of the experimental work are presented in chapter four. Chapter five discusses the findings of the survey and chapter six gives conclusions and recommendations for future work.

CHAPTER TWO

2.0 LITERATURE REVIEW

2.1 Introduction

The industrial demand for light components has made Al-Si casting gain a major market share in the automotive, marine and aircraft industries. With the increasing oil prices and stringent CO_2 emission standards, Al-Si alloys will continue to give a competitive edge over other engineering materials. Production of thin walled components for particular applications has been on the rise. This requires that the fluidity, which is an aspect of castability for these alloys, be adequate to fill the mould cavities. Reports indicate that use of alloying elements in Al-Si alloys can have significant influence on the fluidity of these alloys. It is therefore necessary to have a thorough understanding of the extent to which these alloying elements can affect the fluidity of Al-Si alloys with a view to improve their performance. At the same time, it will also be necessary to review element effect on porosity formation, microstructure and mechanical properties of Al-Si alloys. Though an effort will be made to confine the literature review to Al-Si foundry alloys, deviations to other Al alloy systems such as Al-Cu may be apparent. The next section reviews the effect of various elements on fluidity of cast Al alloys.

2.2 Fluidity of Cast Aluminium Alloys

Fluidity in casting science is the characteristic length that the molten alloy can flow before it is stopped by solidification [Mbuya (2006)]. It is used to refer to the capability of molten metal to fill the mould cavity. Empirically, it is defined as the length (measured, for instance in millimeters) that the molten metal flows in a standard channel with a small constant cross sectional area as it solidifies. For good castability, fluidity must be sufficient to enable filling of the mould, but not too high to cause mould penetration in sand moulds, flashing at mould joints or surface roughness in die and permanent mould castings. Good fluidity is important in casting thin sections and complex castings. Liquid metal flow is also salutary during later stages of solidification when feeding occurs to compensate for shrinkage. Some relationship between fluidity and feeding is expected since both depend on the ability of the metal to flow during solidification [Loper and Prucha (1990)]. Both fluidity and feeding involve an understanding of the factors affecting fluid flow for a given alloy.

It is generally acceptable that fluid flow depends not only on the composition of the alloy but also on mould properties such as its dimensions, geometry, material, metal-mould temperature, mould permeability and venting capacity of the given mould/gating system, moisture content, bond strength and gas evolution. Casting parameters also influence fluid flow and these include pressure head, mould vibration, pouring temperature and degree of superheat. Since fluid flow depends upon several processing factors, fluidity tests are highly process sensitive and fluidity results reported at fixed pouring temperature will be different from those reported at fixed melt superheat.

The earliest documented scientific investigation on the factors controlling fluidity is that of Portevin and Bastien (1932). They developed the early fluidity spiral test which allowed them to quantify the important parameters contributing to the fluidity of cast iron. It was not until early 1950s that similar investigations were initiated into the study of the fluidity of aluminium alloys [Mollard et al. (1987)]. To date, studies on the subject continue to attract interest because of the continued need to produce thin aluminium castings of complex shapes.

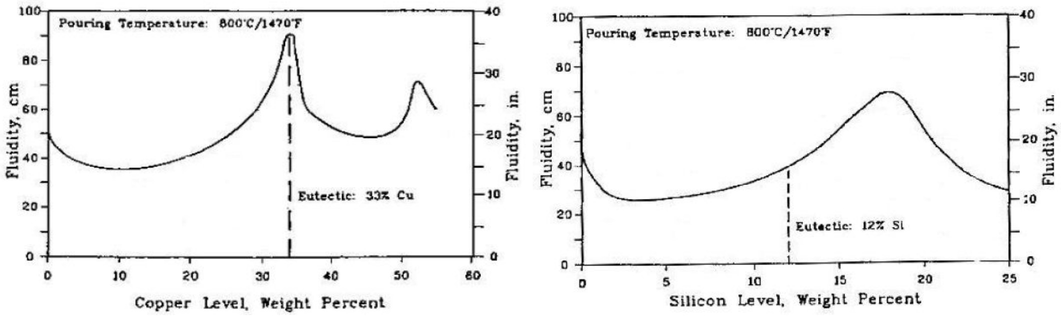
Presently the available data on fluidity of cast aluminium alloys is very confusing, yet such data is important in optimizing the castability of these alloys [Mbuya (2006)]. Reviews by Mollard et al. (1987), Loper (1992), Taylor (1995) and Mbuya (2006) on the fluidity of cast aluminum alloys are remarkable. It is evident from these reviews that although the effects of mould properties and casting parameters on fluidity of cast aluminium alloys seem clear, those due to alloy composition are not. Differing opinions on how elements affect fluidity of cast aluminium alloys have been reported as will be discussed in the section that follows.

2.2.1 Effect of Alloy Composition on Fluidity

Aluminium alloy composition is one of the factors that affect fluidity [Campbell (2003); Mbuya (2006)]. This occurs because compositional variations can vary the viscosity, surface tension, freezing range and solidification mode of the alloys

[Mbuya (2006)]. Changes in fluidity originating from composition induced changes in viscosity and surface tension may be less significant than variations due to freezing range and solidification mode. However, viscosity and surface tension may have a much greater influence on interdendritic feedability and the nucleation and growth of porosity [Mollard et al. (1987)].

The fluidity trends of the binary aluminium alloys (e.g. Al-Cu, Al-Si) with respect to major alloying elements are well understood (see Figure 2.1).



(a) (b)

Figure 2.1: Fluidity curves for (a) a typical binary alloy system (Al-Cu) and (b) the anomalous but important Al-Si alloy system [Mollard (1987)].

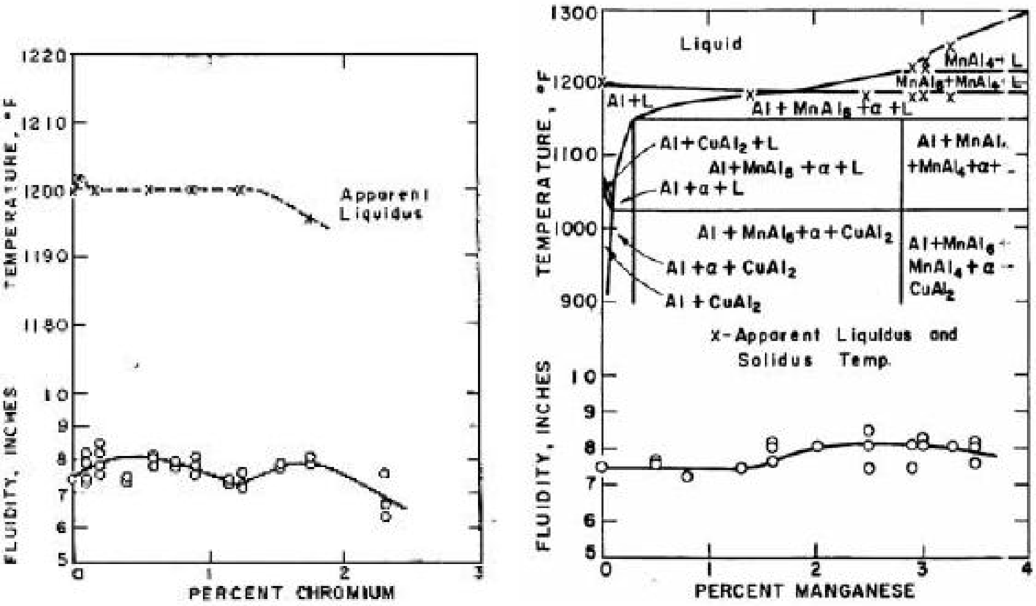
It is noted from Figure 2.1 that the fluidity of aluminium foundry alloys generally decreases with increase in the major alloying element until a minimum is reached (close to maximum freezing range composition). Fluidity then increases to a maximum. In the case of Al-Cu (Figure 2.1a), maximum fluidity corresponds to the eutectic composition with a planar solidification front. As the freezing range widens again into the hypereutectic region, fluidity decreases. The commercially important

Al-Si alloy system, however, displays a slight exception to the above rule as shown in Figure 2.1b. Maximum fluidity in this system is shifted to the hypereutectic composition of about 18%Si while minimum fluidity occurs at around 3-4% silicon content. The unexpected increase in fluidity beyond the eutectic point has famously been attributed to the higher latent heat of fusion of Si compared to Al. Increasing Si content increases the heat generated during freezing and delays solidification as more heat will have to be extracted from the alloy.

Understanding the effect of minor alloying elements is important since commercial aluminium casting alloys contain various minor elements in addition to their major elements. While scarce data is available on the influence of minor alloying elements on fluidity in commercial aluminium casting alloys, observations based on plant experience show that such minor elements somewhat reduce the fluidity [Mollard et al. (1987)]. It is suggested that minor elements contribute to the breakdown of the smooth solidification front that exists between solid and liquid in pure metals. This results in a ‘mushy’ freezing alloy whose flow stops earlier than that of a ‘planar front’ freezing of pure metal [Mollard et al. (1987)].

Studies by Kaufmann et al. (2005) indicate that addition of silicon into the melt of commercial aluminium alloys improves fluidity and reduces shrinkage, thereby giving good castability. Gowri and Samuel (1994) studied the A380 die casting alloy and observed that although a high silicon level of 9% imparted relatively good fluidity, increase in either copper or magnesium contents improved the fluidity even further. However, they reported a reduction in fluidity with Fe addition. In contrast, Pan and Hu (1998) reported a decrease in fluidity with Cu addition, while others [Moustafa et al. (2002) and Fasoyinu (1994)] reported reductions with Mg addition. Other reports [Rooy (1985); Wang et al. (1995)] indicate that Fe deteriorates fluidity. Wang et al. (1995) reported a decrease in the fluidity of molten Al-Si alloy with increase in Fe without appreciable change in surface tension. They attributed the reduction in fluidity to the increase in the amount of insoluble Fe-bearing phases that form in the alloy. Increasing Zn has also been shown to deteriorate fluidity [Kim and Loper (1995); Gowri and Samuel (1994)]. These observations however, contradict those of Pfeiffer and Sabath (1986), who observed that the fluidity increased as the total combined concentration of Fe, Mn, and Zn in an Al-8Si-3Cu alloy was increased.

Sheshradi and Ramachandran (1965) studied the effect of elements on the fluidity of pure Al and found that addition of Ti, Fe, Zr, Cr, Mn and Cu decreases the fluidity. No work has so far been done on the effect of Cr on the fluidity of Al-Si alloys. However, Flemings et al. (1961) conducted studies on the effect of Cr and Mn on the fluidity of Al-4.5Cu alloy and their results are as shown in Figure 2.2 below.



(a) (b)

Figure 2.2: Effect of (a) Cr and (b) Mn on Liquidus Temperature (top) and Fluidity (bottom) of Al-4.5Cu Alloy [Flemings et al. (1961)]

Addition of modifiers such as Sr and Na to Al-Si alloys is a common industrial practice; with reports indicating that modifiers generally decrease fluidity. Kotte (1985) reported that both Sr and Na modification reduce fluidity but Na causes more significant reduction. Decrease in fluidity as a result of Sr and Na addition is attributed to decrease in viscosity during the later stages of solidification. A long solidification range as a result of addition of Sr and Na has also been reported to account for decreased fluidity. Lang (1972) reported that fluidity deteriorates with Na addition. Other reports [Mbuya (2006)] indicate that higher additions of Sr (0.031 to 0.054%Sr) to aluminium foundry alloys considerably increase fluidity of Fe containing alloys. Moreover, Pan and Hu (1997) reported a 20 to 30% increase in fluidity of hypoeutectic alloys containing 5.4%, 7.1%, 8.9% and 11.6% Si with additions of both Sr (0.021 to 0.028%) and Ti (0.22 to 0.29%) as shown in figure 2.3 below. They attributed this increase to higher latent heat of fusion when Sr and Ti are used jointly. Argo and Gruzleski (1988) studied the effect of Sr and Na on fluidity of A356, A319, and A413 and observed no significant changes in fluidity. Modification by Mg has been found to give better fluidity than unmodified alloy and this has been attributed to formation of Mg-phases with high heat of fusion which delays the solidification of the alloy [Di Sabatino (2005)]. Additions of Cerium and Yttrium have been observed to result in increased fluidity and this is attributed to low levels of oxides generated when these additions are made [Taylor (1995)]. Venkateswaran et al. (1986) reported increase in fluidity of eutectic alloy with Sb addition either individually or in combination with Ti. Sb levels between 0.00 to 0.14% decreases fluidity due to the cancellation of its effect by Sr at such low levels.

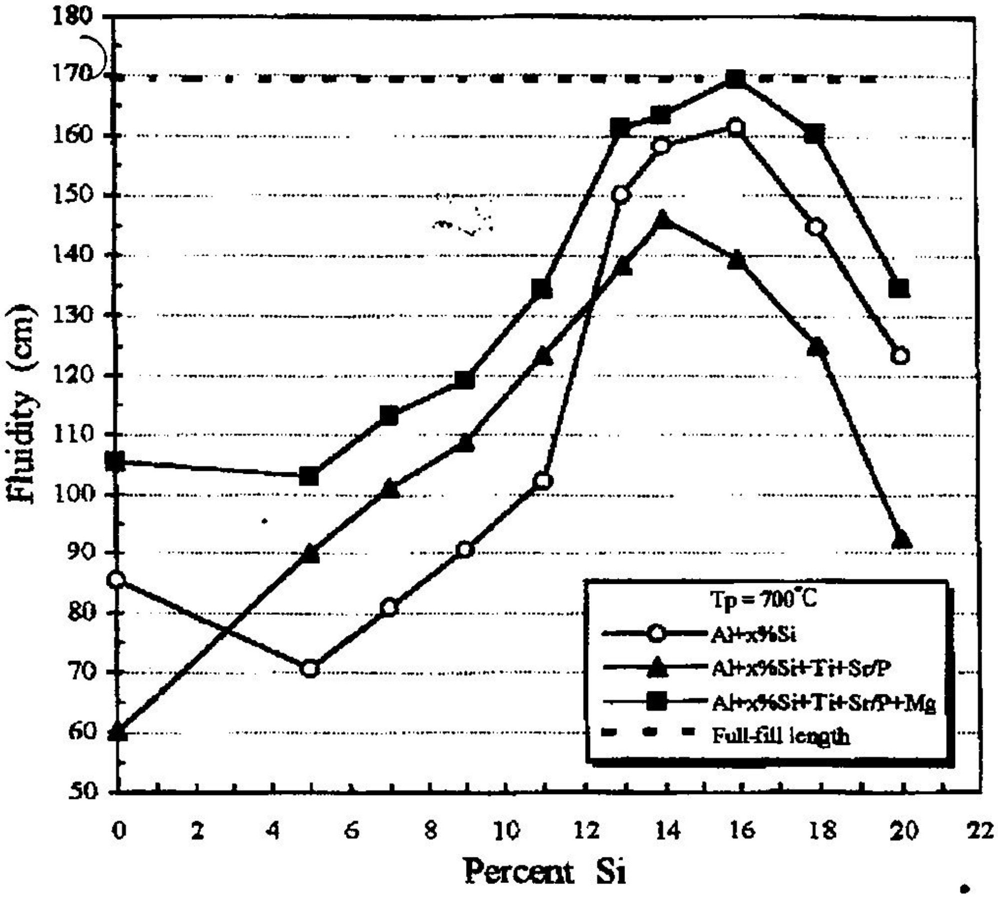


Figure 2.3: Effect of Sr and Ti on fluidity of Al-Si alloys [Pan and Hu (1997)]

Addition of grain refiners may result in delayed dendrite coherency and higher fluidity [Chai (1994)]. The dendrite coherency point is the point at which a more or less continuous three-dimensional dendrite network is formed in the semi-solid material [Dahle et al. (1996)]. Delays to coherency may tend to increase time for mass flowing without formation of a solid network of grains. This may in turn increase fluidity. Lang (1972) reported an increase in fluidity when grain refining agents are present because of the reduced possibility of growing of large dendrites. He further indicated a significant increase in fluidity with boron additions in the range of 0.04% to 0.07% to Al-Si alloys tested with a bar die casting. Kwon et al. (2003) studied the effect of grain refinement on A356 alloy and reported that whereas 0.03%Ti added as Al-5Ti-1B grain refiner appreciably improved the fluidity at the lowest temperature of 700°C, addition of 0.2%Ti had no appreciable effect on the alloy. Other authors argue that grain refinement reduces fluidity.

Gruzleski and Closset (1990) suggested that early nucleation caused by grain refinement results in a slurry flow from the moment of pouring. Since slurries flow with more difficulty than simple liquids, the fluidity should be reduced with grain refinement. Di Sabatino and Arnberg (2005) reported that fine particles are more effective in stopping a flowing stream than equivalent coarse particles. Due to this, one should expect fluidity to decrease with grain refinement. Other reports also indicate that addition of grain refiners containing Ti leads to reduction in the fluidity of Al-Si casting alloys [Loper (1992); Loper and Prucha (1990)]. Mollard et al. (1987) reported a reduction in the fluidity of an Al-4.5Cu alloy when 0.15%Ti was added using a vacuum fluidity test apparatus. Dahle et al. (1996) conducted studies on fluidity by successive addition of Al-5Ti-1B grain refiner to Al-7Si-Mg and Al-11Si-Mg alloys using a sand moulded spiral. They observed that fluidity decreased with grain refinement below 0.12%Ti while it increased with additions above 0.12%Ti. Tiryakioglu et al. (1994) found no effect on fluidity in A356 and 319 alloys, upon addition of 0.04% Ti as Al-5Ti-1B in a sand moulded spiral.

Results on elemental effect on fluidity as held by different investigators can be summarized as shown in table 2.1.

Table 2.1: Summary of effect of elements on fluidity as reported by various authors

Element	Effect on Fluidity	Reference	Comment
Si (Base alloying element in Al-Si alloys)	Increases for Si levels above about 4% to about 17 %. Decreases otherwise.	Kaufman et al. (2005); Gowri and Samuel (1994); Lang (1972)	Si increases latent heat of fusion
Fe + Mn + Zn	Increase	Pfeiffer and Sabath (1986)	Mn forms phases less detrimental to fluidity
Fe	Decrease	Rooy (1985); Gowri and Samuel (1994)	Fe forms Fe-bearing phases that deteriorate fluidity
Cu	Decrease	Pan and Hu (1988)	Cu leads to formation of Cu-phases
Mg	Decrease	Moustafa et al. (2002); Fasoyinu (1994)	Formation of detrimental phases
Mg	Increase	Di Sabatino (2005)	Forms Mg-phases with high latent heat of fusion.
Zn	Decrease	Kim and Loper; Gowri and Samuel (1994)	Forms phases that make flow difficult
Sr	Increase	Mbuya (2006)	Modification and neutralization effect

Previous Table 2.1 continued showing summary of effect of elements on fluidity as reported by various authors

Sr	Decrease	Argo and Gruzleski (1988); Lang (1972); Kotte (1985)	Increases solidification range
Sr + Al-5Ti-1B	Increase	Pan and Hu (1997)	Higher latent heat of fusion
Na	Decrease	Kotte (1985); Lang (1972)	Decreases viscosity during later stages of solidification
Al-5Ti-1B	Increase	Chai (1994); Lang (1972); Kwon et al. (2003)	Delays dendrite coherency point. Reduces possibility of growing large dendrites.
Al-5Ti-1B	Decrease	Gruzleski and Closset (1990); Di Sabatino and Arnberg (2005);	Causes a slurry flow that retards fluidity
Al-5Ti-1B (0.04%)	No effect	Tiryakioglu et al. (1994)	Formation of fine particles
Cr	Decrease	Flemings et al. (1961)	Sludge formation
Mn	Increase	Flemings et al. (1961)	Fe-correction
Sb	Increase	Venkateswaran et al. (1986)	Low levels of oxides generated
Sb (0 to 0.14%)	Decrease	Venkateswaran et al. (1986)	Cancellation of its effect by inherent Sr in alloy at such low levels

2.2.2 Effect of Melt Cleanliness on Fluidity

The presence of inclusions (including non-metallics and primary intermetallics) within the melt is reported to decrease casting fluidity [Wang and Apelian (1993); Loper (1992)]. Groteke (1985) observed substantial reductions in fluidity of A319 alloy when the melt was 'dirty' (i.e. heavily contaminated with entrained oxides). Oxide skins on the melt surfaces increase the apparent surface tension of the melt and it is estimated that the apparent surface tension of a melt with a heavy oxide skin may be up to three times greater than the actual surface tension of an oxide free melt [Loper (1992)]. Running lengths obtained with the inert gas low pressure casting process are typically double those obtained by the conventional process, and up to four times those obtained using gravity die cast methods [Mollard et al. (1987). The improvements have been directly related to the reduction of surface oxide skin obtained via use of inert gas usually argon.

2.2.3 Effect of Melt Superheat on Fluidity

Melt superheat is defined as the difference between the liquidus temperature (composition dependent) and the pouring temperature [Taylor (1995)]. Melt superheat is a key factor in influencing the fluidity of commercial aluminium alloys [Di Sabatino and Arnberg (2005)]. Many authors including Flemings et al. (1961), Tuttle et al. (1989) and Kolsgaard (1993) have reported that for a given alloy composition, the fluidity increases linearly with increasing melt temperature. Kolsgaard (1993) further reported that an increase of 1°C in the melt temperature gives an increase of

1% in the fluidity length of Al-7Si-0.6Mg alloy reinforced with 10 to 30% SiC particles. This finding is also supported by the recent work by Di Sabatino et al. (2005). From these studies, it can be concluded that a higher degree of melt superheat will be necessary when casting thin sections because a greater degree of superheat increases fluidity and hence greater running lengths.

2.3 Porosity Formation

2.3.1 Introduction

Porosity is cited as one of the most common and undesirable defects in aluminium castings and has been a major concern in aluminium foundries for the past fifty years [Sigworth and Wang (1993)]. It is the term used to indicate the voids and cavities that form within the casting during solidification. Presence of porosity in castings leads to poor surface finish and decline in mechanical properties such as tensile strength, ductility and fatigue life. Roy et al. (1996) noted that porosity in castings occur because of rejection of gas (mainly hydrogen) from the liquid during solidification (gas porosity) and/or inability of liquid metal to feed through the interdendritic regions to compensate for volumetric shrinkage associated with solidification (shrinkage porosity). It is to be noted further that hydrogen is the only gas capable of dissolving significantly in molten aluminium but upon solidification, the solubility of hydrogen is greatly reduced leading to porosity formation.

2.3.2 Mechanisms and Effect of alloying elements on Porosity Formation

Numerous attempts to explain the mechanisms responsible for porosity formation have been advanced. Unfortunately however, no single explanation seems to command universal support making the concept of porosity formation to remain elusive. While some authors believe that altering the solidification sequence of the alloy or changing the nucleation and growth characteristics can cause porosity formation in aluminium casting alloys; others [Manas and Makhlouf (2002)] report that decreased permeability arising from physical obstruction of interdendritic feeding paths contribute to porosity formation. A few more others have noted that a combination of all these mechanisms account for porosity formation [Otte et al. (1999)].

Evidence that changes in solidification characteristics cause porosity has been reported by Lee et al. (1990). The authors noted that porosity content decreases with increasing solidification rate. Backerud et al. (1990) also reported that the solidification sequence becomes more complex when the Fe content increases from 0.13% to 0.47% in Al-Si-Cu-Mg-Fe 380-type alloy. On the other hand, Taylor et al. (1999a-c) proposed that porosity in Al-Si alloys can significantly be influenced by the solidification sequence of Fe-bearing phases. In addition, they observed that porosity was at a minimum at the critical Fe content for each of the alloys investigated and attributed this behavior to changes in interdendritic permeability caused by different solidification sequences.

Reports on the nucleation and growth theories indicate that the pore nuclei must exceed a particular critical radius before growth can proceed [Taylor (1995)]. Pressure imbalance within the solidifying casting has been noted to play an important role in determining the critical pore nuclei radius at which porosity forms and grows [McDonald et al. (2004)]. In this regard, it is reported that if the sum of the pressure of dissolved gases P_g, in the melt and that due to shrinkage, P_s, is greater than the sum of pressure due to metallostatic head, P_h, surface tension, P_{s-t}, and atmospheric pressure, P_{atm}, then the pore nuclei can reach the desired critical radius and grow leading to porosity formation. This requirement can be represented mathematically as shown in equation (2.1).

$$P_g + P_s > P_{atm} + P_h + P_{s-t} \tag{2.1}$$

Dinnis et al. (2004b) indicated that impurity and alloying elements may alter the surface tension and therefore, change the term P_{s-t}. Those elements that reduce the surface tension would make pore formation easier and lead to increased levels of porosity. On the other hand, hydrogen solubility may be altered by impurity and alloying elements which will in turn change P_g. If P_g increases, chances of pore formation will be increased leading to high levels of porosity.

Taylor (1995), Easton and StJohn (2000) and Mohanty et al. (1993) observed that pores in castings start nucleating on heterogeneous sites which are inherent in almost all molten metals. The reason why pores commence at heterogeneous sites has been attributed to their provision of low energy nucleation sites where the critical pore radius can be achieved leading to porosity formation [Taylor (1995)]. Dinnis et al. (2004b) observed that alloying elements increase the inclusion content in the melt which form the most important category of heterogeneous sites in the melt. Unfortunately, even when the best corrective measures are undertaken, some inclusions still find their way into the melt and serve as pore nucleation sites [Mohanty et al. (1993)] which finally leads to generation of high levels of porosity.

In certain cases, addition of inclusions can be intentional such as in metal matrix composites. The presence of such substrates greatly enhances porosity formation in aluminium casting alloys. Taylor (1995) indicated that deliberate addition of oxides skimmed from the melt leads to generation of pores. However, when these inclusions are removed by filtration, the porosity level reduces significantly [Mohanty et al. (1993)]. Though widely accepted that pores will nucleate at low energy inclusion sites where the critical pore radii can form, the relationship between element levels and inclusions is not well understood. Furthermore, it is important to understand the effect of elements on some of the parameters in equation (2.1) such as P_g, P_s and P_{s-t} and hence their contribution to porosity formation in cast aluminium alloys.

2.3.2.1 Role of Modifiers

Much research has demonstrated that modifier addition to Al-Si alloys increase the amount, size and distribution of porosity. In addition, the pore morphologies change from irregular and interconnected to more rounded and isolated pores as a result of modification. Theories explaining the effect of modification on porosity formation include those that are hydrogen related, surface tension related and feeding related [McDonald et al (2004)]. In hydrogen related effects, it has been noted that Sr-modified alloys increase hydrogen pick up which in turn affect the term P_g in equation (2.1) and hence porosity. In order to determine the role of modification in hydrogen gas pick up, Denton and Spittle (1985) carried out a detailed investigation of the solidification characteristics of both Sr- and Na- modified LM6-type alloys. Their studies indicated that modification increases the susceptibility of the alloys to pick up more hydrogen into the melt. Bien et al. (2000) also reported that Sr addition markedly increases the hydrogen content in Al-Si melts and accelerates the gassing rate of the melt. They further observed that in unmodified melts, the hydrogen level decreases with time until it reaches a stable level as shown in Figure 2.4.

Other studies however, have indicated that hydrogen pick up does not increase with Sr modification. Gruzleski et al. (1986) undertook hydrogen measurement using the Telegas instrument in Sr modified A356 alloy and reported that Sr modification does not in any way result in more melt-dissolved hydrogen in this alloy. Shahani (1985) also reported that modified samples had relatively lower hydrogen content although pores were formed more frequently in these samples. This implies that the increased porosity in Sr-modified alloys must be due to some reason other than increased hydrogen pick up.

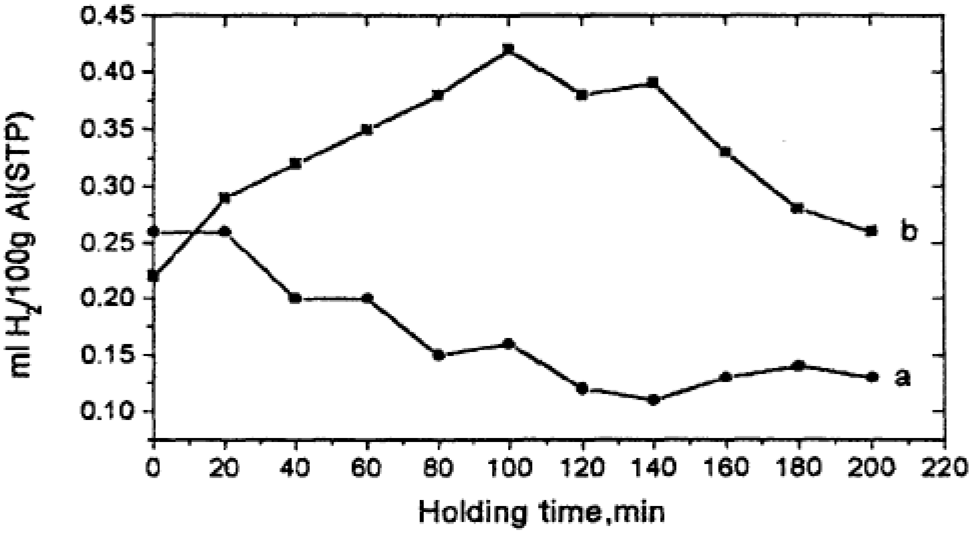


Figure 2.4 : Hydrogen content against holding time in unmodified and modified Al-10Si melts at a holding temperature of 750˚C, (a) unmodified, (b) modified with addition of 0.05%Sr at 720˚C [Bien et al. (2000)]

An explanation as to how modification affects surface tension and hence porosity has been given. It is to be noted that addition of Sr reduces the surface tension of Al-Si alloys and hence affect the term P_{s-t} in equation (2.1) leading to porosity formation. McDonald et al. (2004) however, indicated that the 19% decrease that occurs due to addition of 100 ppm Sr is insufficient to account for any significant changes in porosity characteristics. In fact, mathematical models have demonstrated that surface tension reductions of even 50% for the same level of Sr are inadequate to cause changes in porosity formation characteristics in Al-Si casting alloys. One is therefore tempted to conclude that reduction in surface tension and hence P_{s-t} due to modification is not responsible for increased porosity.

In terms of feeding related effects, it is to be noted that pressure drop due to restricted flow could affect the term P_s in equation (2.1) and cause changes in porosity characteristics in Al-Si casting alloys. The changes in P_s however, due to modifier addition are too small to have significant changes in porosity formation. McDonald et al. (2004) further noted that addition of either Sr or Na has an effect in increasing the freezing range of Al-Si casting alloys. Studies by Argo and Gruzleski (1988) indicated that longer freezing ranges do increase porosity levels in modified alloys. However, it is noted that increases in freezing range due to modification are insufficient to explain the higher levels of porosity associated with modification.

Recently, a theory that seeks to explain increased porosity formation due to modification based on the solidification mode of the Al-Si eutectic has become popular. It is reported that the Al-Si eutectic can solidify in three different modes as shown in Figure 2.5. The different eutectic solidification modes have different influences on the permeability of the mushy zone and interdendritic fluid flow which will either reduce or increase porosity levels. According to Dahle et al. (2000), directional eutectic growth from the walls to the center of the casting (as in Na-, Ca- or Y-modified alloys) provides the most open interdendritic feeding channels with very high permeability. This solidification mode also reinforces the surface of the casting quite early during solidification compared to the other two eutectic solidification modes. It is therefore likely that porosity will concentrate in the regions that solidify last and surface defects such as hot tearing or surface shrinkage/slumping will be minimized due to surface reinforcement during the early stages of eutectic growth. Large concentrated shrinkage pores are therefore expected in the hot spot regions if poorly fed. With good feeding such that the hot spot is shifted from the casting to the riser, this can result in very sound castings.

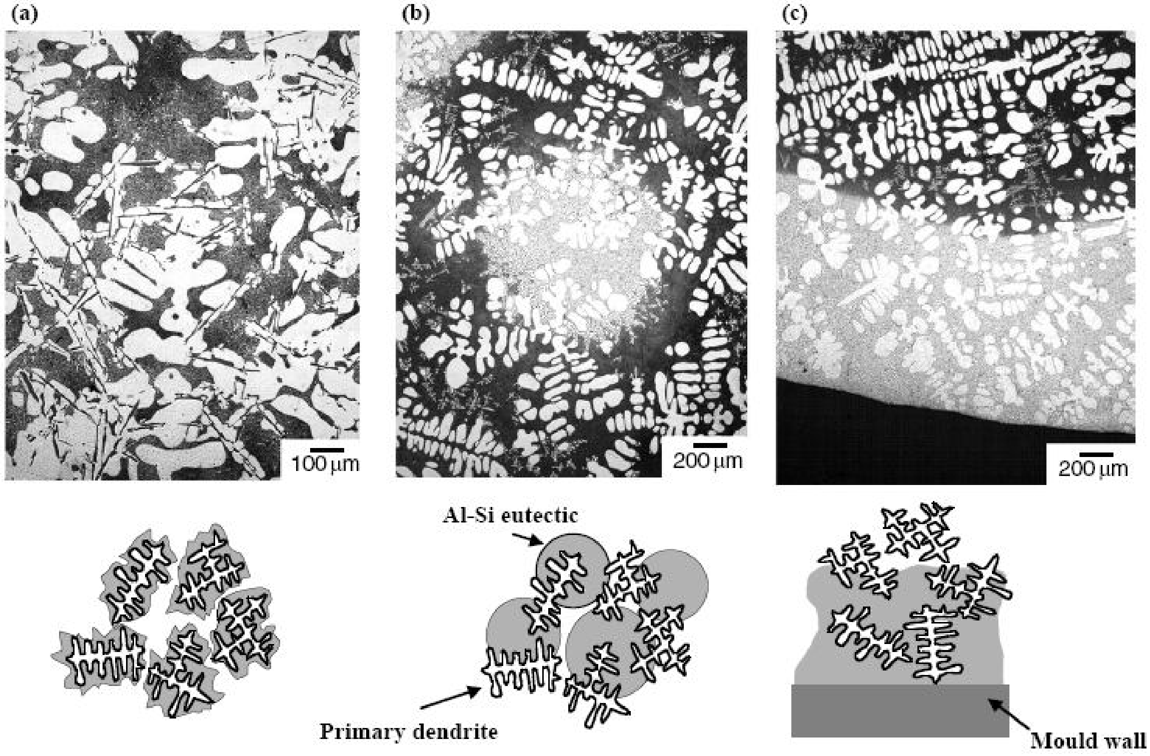


Figure 2.5: Eutectic growth modes. (a) Nucleation on or adjacent to primary aluminium dendrites, (b) independent heterogeneous nucleation of eutectic grains in interdendritic spaces and (c) growth of the eutectic solidification front opposite to the thermal gradient. [Dahle et al. (2002)]

On the other hand, the nucleation and growth of the eutectic either directly on the primary aluminium dendrites (as in unmodified, high-purity binary Sr-modified or very high Sr-modified alloys) or independently within interdendritic spaces (as in Sr-, Sb-, Ba- or Yb-modified alloys) will result in an increase in the specific surface area of the mushy zone and thereby reduce the feeding efficiency and increase porosity formation [Dahle et al; (2000)]. The first case is likely to result in isolated pools of large shrinkage pores while the latter will tend to result in smaller pores distributed throughout the casting.

2.3.2.2 Role of Grain refinement

Although grain refinement is done with a view to control the grain structure of cast aluminium alloys, it has been observed to be beneficial to castability [Dave et al. (2002)]. Easton and StJohn (2000) indicated that one of the benefits attributable to grain refinement is reduced porosity. However, they caution that the practice of grain refinement must be done with prudence because there are some casting configurations such as the spoke-rim junction of permanent mould-cast wheels where addition of extra grain refiner causes an increase in localized porosity. Argo and Gruzleski (1988) observed that, in addition to promotion of a fine equiaxed structure, grain refinement also leads to a finer dispersion and reduction in the amount of porosity. Studies on grain refinement by Taylor et al. (1999) that investigated the effect of Ti-B based grain refiners on modified and unmodified Al-Si alloys indicated that in unmodified alloy, grain refiner addition fragmented the large central pore into a few moderate sized pores. This observation is similar to that of Easton and StJohn (2000) who reported that addition of grain refiners cause the pore size to decrease and porosity to be redistributed throughout the casting rather than being concentrated in a particular region. In the modified alloy, the grain refiner addition resulted in a multiplication of pore numbers and a reduction in average pore size within the dispersed porosity region. Figure 2.6 illustrates this observation.

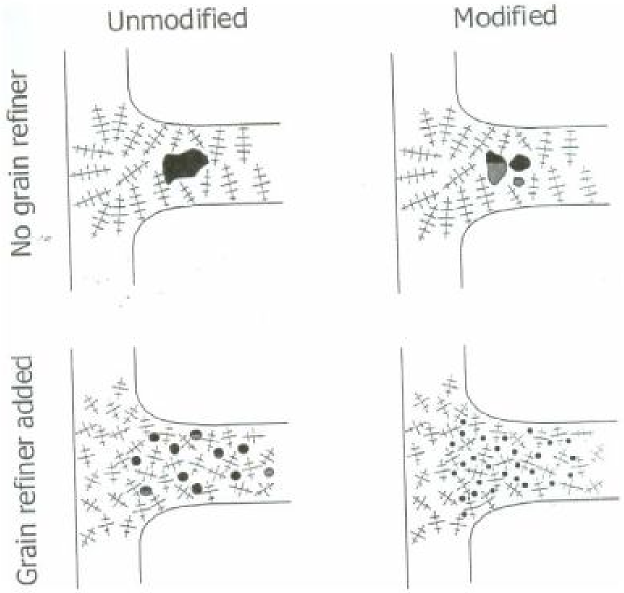


Figure 2.6: Approximate pore distributions and relative pore sizes for unmodified and modified Al-Si alloys upon addition of Ti-B based grain refiners [Taylor et al. (1999)].

2.3.2.3 Role of Fe and Fe-Correctors

Iron is an impurity element commonly found in Al-Si alloys and is highly soluble in molten aluminium but sparingly soluble in solid aluminium. Depending on the alloy chemistry, Fe can solidify to form different intermetallic phases such as β-Al_5FeSi, α-$Al_{15}Fe_3Si_2$ and δ-Al_4FeSi_2 [Crepeau (1995)]. The presence of β-Al_5FeSi leads to porosity formation due to the probable role the phase plays in blocking the interdendritic channels, making fluid flow difficult [Roy et al. (1996); Otte et al. (1999); Mbuya et al. (2003)]. Villeneuve et al. (2001) reported that precipitation of the β-platelets/needles favours the formation of massive shrinkage cavities due to difficulty encountered by the liquid metal to flow through the intercepting branched β-platelets especially during the later stages of solidification. Other authors [Dinnis et al. (2003); Dinnis et al. (2004a); Wang et al. (2003)] reported that the β-phases leads to rapid deterioration of interdendritic permeability because they are effective nucleation sites for the eutectic Si. This can only be possible if the Fe level is high because it is at high levels of Fe that the β-Al_5FeSi phases are likely to form prior to the Al-Si eutectic. Other studies [Samuel et al. (2000)] indicate that the amount of β-Al_5FeSi increase with phosphorus addition to Al-6.5Si-3.5Cu alloy and propose that P forms AlP particles which act as nucleation sites for β-Al_5FeSi platelets. The consequence of this observation is that the eutectic Si and β-Al_5FeSi platelets share a common nucleation site, AlP. The formation of β-phase platelets prior to eutectic Si at high Fe levels will dramatically reduce the number of AlP particles available for nucleation of Al-Si eutectic grains making the latter overly large blocking the flow paths and leading to porosity formation.

Many authors support the fact that porosity levels increase with Fe in Al-Si alloys. Otte et al. (1999) reported that the total porosity in Al-9Si-3Cu increased with iron content by approximately 30% across the range of iron content examined. Wang et al. (2003) observed that increasing the Fe content from 0.13% to 0.47% significantly increases the amount and size of Fe-rich intermetallic phases and this in turn increases the level of porosity in the microstructure. Dinnis et al. (2004a) conducted a detailed experimental work in which they varied Fe levels in order to determine the effect of this variation on porosity formation. Figure 2.7 below provides a summary of their results. It is evident from the study that porosity levels increase with increase in iron content.

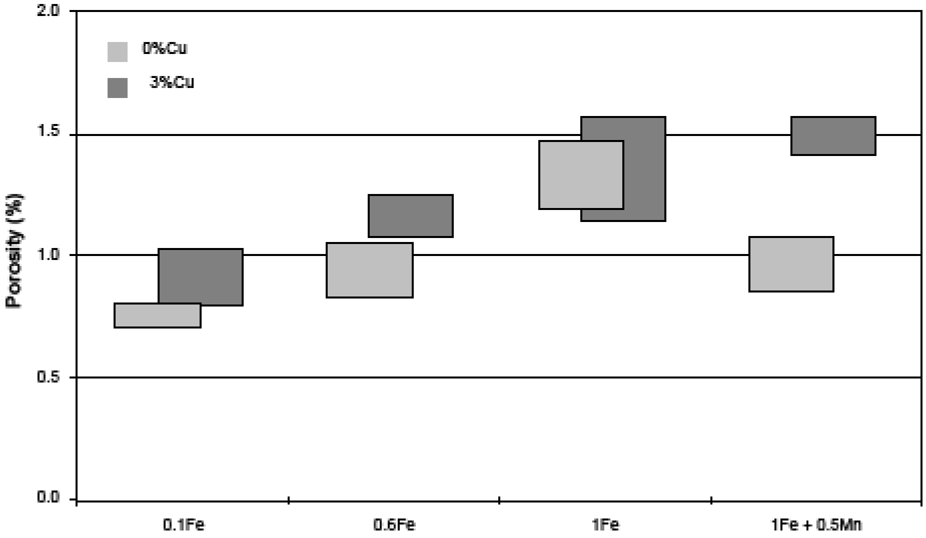


Figure 2.7: Increase in porosity levels with increasing iron content [Dinnis et al. (2004a)].

The porosity reduction due to Mn addition to the alloy with 1.0%Fe (Figure 2.7) indicates that use of Fe-correctors can impart significant changes on porosity characteristics of Al-Si alloys. The reduction in porosity levels due to addition of Mn such as observed by Dinnis et al. (2004a) was previously attributed to the relative compactness of α-$Al_{15}(Fe,Mn)_3Si_2$ particles that forms compared to β-Al_5FeSi. Recent studies however, indicate that addition of Mn to Fe containing aluminium alloys reduce porosity not because of the relative compactness of α-$Al_{15}(Fe,Mn)_3Si_2$ particles but rather because these particles do not share a common nucleant particle, AlP, with Al-Si eutectic [Dinnis et al. (2006)]. Instead, the α-$Al_{15}(Fe,Mn)_3Si_2$ particles nucleate on substrates other than AlP particles. This allows more AlP particles to be available for nucleation of Al-Si eutectic grains making them smaller in size. Consequently, fluid flow through interdendritic paths becomes less difficult resulting in reduced porosity. Taylor (2004) in his review indicated that addition of Fe-correcting elements especially Mn in certain Mn/Fe ratios is a widespread practice in Al-Si casting alloys containing high Fe levels that lead to reduction in porosity. It is expected that addition of Cr individually or in combination with Mn may have a similar effect.

2.4 Microstructure of Al-Si Casting Alloys

2.4.1 Introduction

The mechanical and physical properties of cast Al-Si alloys are influenced by casting defects such as porosity and oxide films, and various microstructural features such as grain size, dendrite arm spacing, eutectic Si morphology and Fe intermetallic phases. Microstructural features are controlled by the chemical composition, solidification rate and heat treatment. The increasing demand for higher material properties such as fatigue strength, tensile strength, ductility and corrosion resistance require close microstructural control through tighter specification of composition, casting practice and heat treatment. However, secondary alloys generally have higher levels of impurity elements that can be difficult to control. It is therefore necessary to have a thorough understanding of the influence that these elements impart on the microstructure of these alloys and subsequently their performance in service. The effects of some common minor and impurity elements on the microstructure of Al-Si alloys are briefly reviewed here.

2.4.2 Typical Al-Si Microstructure

Al-Si alloys can either be hypoeutectic with the Si content below the eutectic composition (11.7%Si) [Kral et al. (2006)], eutectic with Si content around the eutectic or hypereutectic with silicon above the eutectic. Typical microstructures of these three categories are shown in Figures 2.8 to 2.10. It is noted from Figure 2.8 that the microstructure of the hypoeutectic alloys in as-cast condition consists of primary α-Al solid solution with interdendritic regions of Al-Si eutectic. In addition, various intermetallic compounds of different types of Fe-, Cu-and Mg- bearing phases are commonly found in the microstructure of these alloys. Figures 2.9 and 2.10 show the microstructural features found in hypereutectic and eutectic alloys respectively and as expected, they consist mainly the eutectic colonies and fine primary Al dendrites.

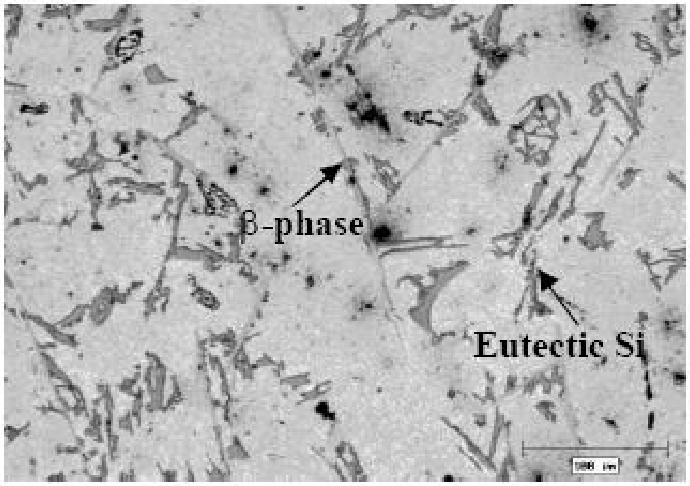


Figure 2.8: Typical microstructure of sand cast Al-7Si-0.3Mg-0.7%Fe hypoeutectic alloy in the as-cast condition [Sreeja et al. (2006)]

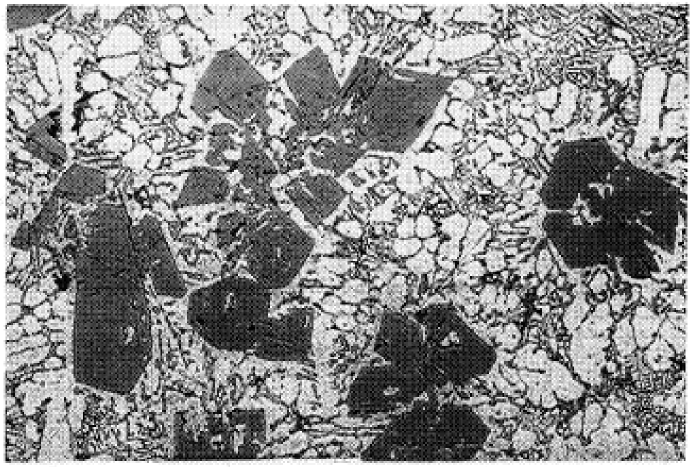


Figure 2.9: Microstructure of hypereutectic Al-Si alloy remelted and cast without any melt treatment [Mohanty el al. (1993)]

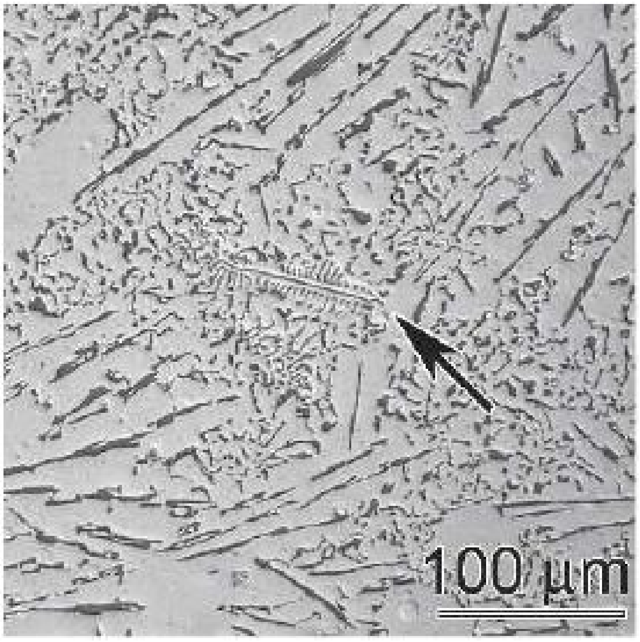
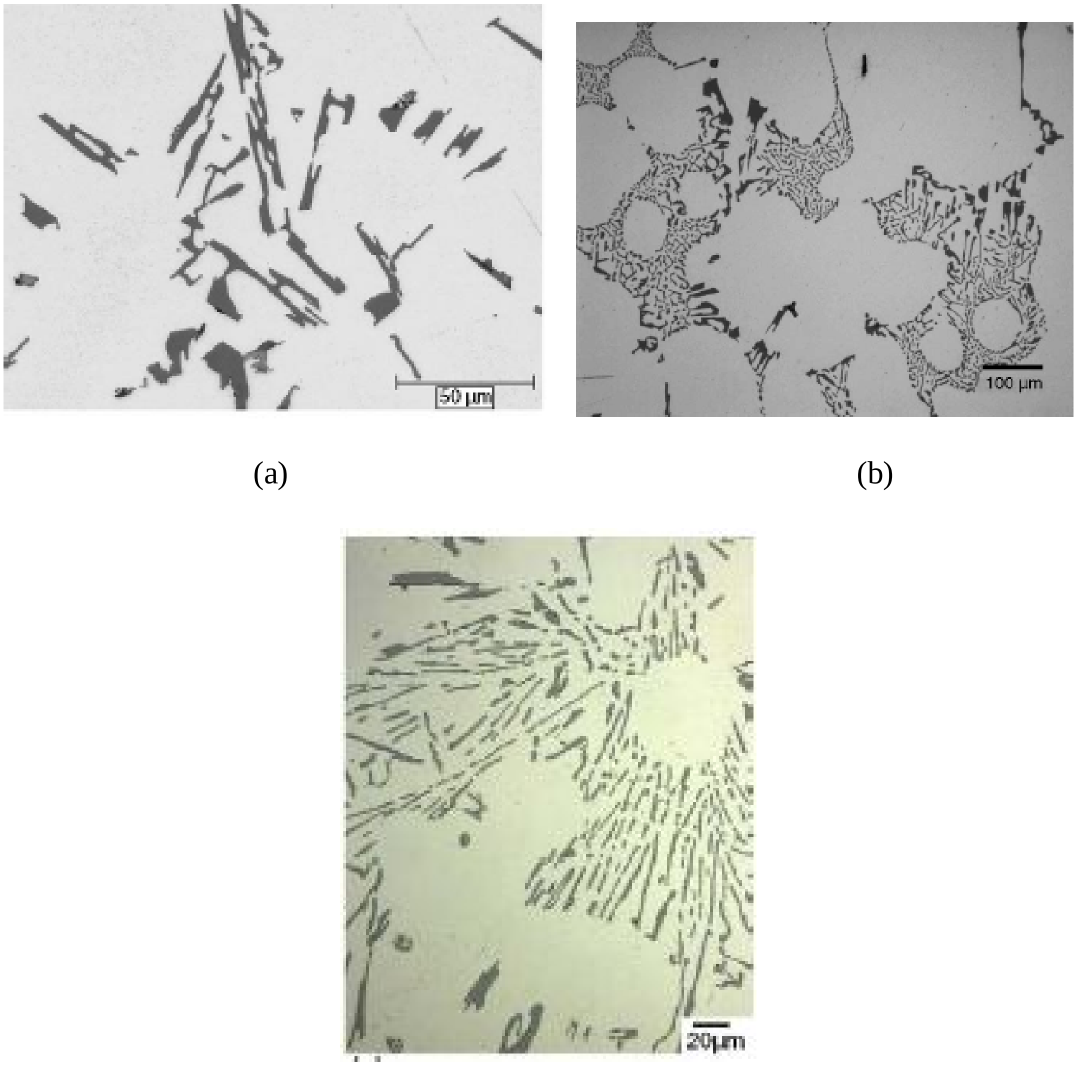


Figure 2.10: Typical microstructure of a eutectic alloy [Kral et al. (2006)]

2.4.3 Effect of Al-Si Eutectic Modification and Primary Si Refinement on Microstructure

2.4.3.1 Al-Si Eutectic Modification

In unmodified condition the eutectic Si assumes a coarse platelet morphology which deteriorates the mechanical properties of cast Al-Si alloys. Studies aimed at enhancing the morphology of the eutectic Si particle and improving the mechanical properties of the alloys indicate that addition of minor alloying elements called modifiers can change the coarse and large needles of eutectic Si into a fine well rounded form. Work by Li et al. (2004) to improve the performance of AA319-type alloys indicated that addition of 0.022%Sr led to modification of the eutectic Si morphology to fibrous form and caused the segregation of the blocky $CuAl_2$ phases which upon dissolution during T6 heat treatment enhances properties. Other elements such as Na and Sb have been used for chemical modification in addition to Sr. Shabestari and Shahri (2004) noted that addition of Sb or Sr to an A356 alloy modifies the eutectic Si and imparts significant improvements to mechanical properties of the alloy. Furthermore, the authors noted that modification plays a significant role in the kinetics of spheroidisation of the Si particles during heat treatment. Micrographs indicating unmodified, Sr- and Sb- modified alloys are shown in Figure 2.11.



(a) (b)

(c)

Figure 2.11 Micrographs showing (a) unmodified A356 Al-Si alloy [Shahrooz and Reza (2006)] (b) 0.027%Sr-modified Al-7Si-0.5Mg-0.9Cu alloy [Dons et al. (2005)] (c) 2400 ppm Sb-modified Al-10Si hypoeutectic alloy [Dahle et al. (2005)]

2.4.3.2 Primary Silicon Refinement

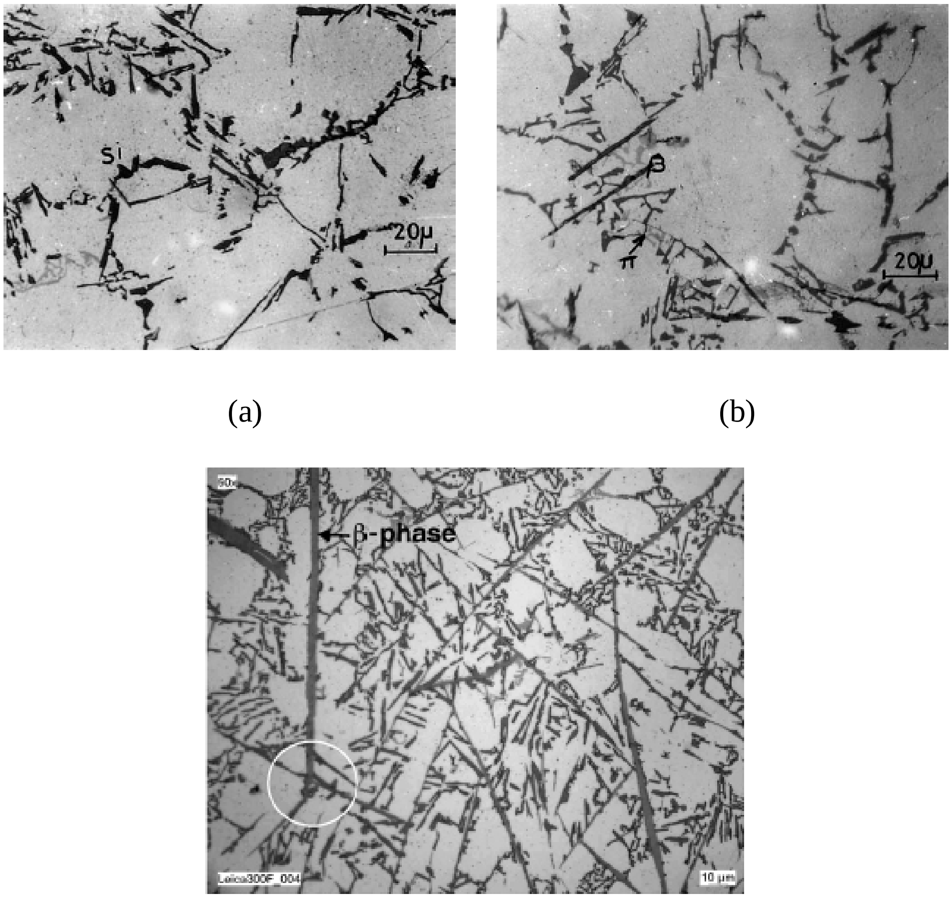
Refinement of primary Si through addition of alloying elements has become popular due to its simplicity. Phosphorus is commonly used for this purpose and the reason for its application is attributed to its tendency to form finely dispersed AlP particles which act as nucleation sites for the Si particles. Tanihata et al. (2008) noted that if P additions are made to the Al-Si piston alloys, significant refinement of the primary Si will be realized leading to significant improvement in the fatigue performance of the alloy. They however noted that the wear resistance of the alloy will be deteriorated. Recent studies by Kim (2003) have demonstrated that Ca additions to B390 type alloys causes refinement of the primary Si in addition to modification of the eutectic Si. At the same time, improvement of mechanical properties such as tensile, elongation and impact strength were reported.

2.4.4 Effect of Iron

Iron is a common impurity element in cast aluminium that leads to formation of Fe-rich intermetallic phases that are detrimental to mechanical properties, especially ductility. The most detrimental of these phases is the platelet shaped β-Al_5FeSi and less deleterious ones include the Chinese script α-Al_8Fe_2Si and π-Al_8Mg_3Fe phases. The β-phases are reported to be the most detrimental phases due to their needle/platelet morphology.

2.4.4.1 Effect of Fe Level

Increasing the Fe content in Al casting alloys increases the size and volume fraction of the β-Al_5FeSi platelets. In addition, the shape of the β-phases becomes more needle-like at high Fe levels [Sreeja et al. (2006)]. This trend can easily be observed in micrographs in Figure 2.12.



(a) (b)

(c)

Figure 2.12: As cast microstructures of LM25 alloy containing (a) 0.2%Fe and (b) 0.6%Fe [Ravi et al. (2002)] (c) Micrograph showing permanent mould cast Al-7Si-0.3Mg alloy with 0.8% Fe [Sreeja et al. (2007)].

2.4.4.2 Neutralizing Fe Effects

Since removal of iron from the melt is not commercially viable, reduction of its detrimental effects is done through neutralization. This can be achieved by adding small amounts of elements such as Mn, Cr, Be, Mo, Co [Ravi et al. (2002)] and, more recently, Sr. These elements are known to transform the plate-like β-phase to less the detrimental Chinese script or polygonal α-phases and improve the mechanical performance of the alloys. Manganese is widely used for this purpose and is usually added at the ratio of Fe to Mn of 2. For Cr, a ratio of Fe to Cr of 3 is optimal in transforming the harmful plate-like β-phases to less detrimental phases in A413 alloy containing 1.12%Fe as recently reported by Mahta et al. (2005). It is to be noted that Sr levels for neutralization purpose has been reported to be higher than those usually added for Si modification. Neutralization of the negative effects of Fe can also be achieved through increasing the cooling rates or by melt superheat. The various methods used to neutralize Fe effects are discussed hereunder.

2.4.4.2.1 Effect of Cooling Rate

Moustafa (2009) noted that cooling rates have a direct impact on the equilibrium kinetics and quantities of Fe-phases present in the microstructure. Accordingly, he observed that if the cooling rate is low it will favour the precipitation of the β-phases. Additionally, Crepeau (1995) noted that at low cooling rates of 0.1˚C/s formation of the β-phases is favoured while at high cooling rates of 20˚C/s formation of the β-phases is inhibited. By increasing the cooling rate, Vorren et al. (1994) further noted that the negative influences of Fe can be effectively reduced. At the same time, Narayanan et al. (1995) reported that increasing the cooling rate refines the microstructure of the Fe-intermetallics and a high cooling will be preferable for both the 319- and 413-type alloys used in their studies.

2.4.4.2.2 Effect of Manganese

Manganese is widely used as an effective method through which the detrimental influences of Fe to these alloys can be reduced because of its low cost and availability as reported by Cao and Campbell (2006). The authors however, regret that the amount of Mn required to neutralize Fe has not been well established. The common practice has been to use an Fe to Mn ratio of 2. A different neutralization formula has been suggested for both sand and permanent mould casting as reported by Cao and Campbell (2006) and this is Mn = 2(%Fe-0.5). Studies by Narayanan et al. (1995) indicated that Mn addition to Fe-containing alloys in the ratio Mn/Fe of 0.5 played a positive role in combining with the Fe to form a Chinese script structure instead of one with the plate like morphology in the 319 and 413-type alloys. Hwang et al. (2008) reported that while Mn additions cause the replacement of the β-phase with the α-phase, it may however enhance the formation of coarse polyhedral or star-like particles commonly referred to as sludge. Furthermore, Narayanan et al. (1994) indicated that Mn additions are associated with the problem of hot tearing. Therefore, addition of Mn should be restricted only to amounts necessary to convert the β-phase unless it is being added as an alloying element to impart other effects such as high temperature strength.

In a study of Al-13Si with 0.5-1.5%Fe carried out by Muralli et al. (1994), it was observed that addition of 0.6% Mn led to production of a compact phase that is less detrimental to strength and ductility. Bonsack (1942) indicated that addition of 0.5%Mn to Al alloys transformed the β-phases into Chinese scripts that negated the bad influences of Fe on physical and mechanical properties of Al-Si alloys. Zednic (1948) reported that addition of 0.41%Mn or 0.44%Co to Al-9.1Si-0.22Mg-0.4Fe alloy produced Chinese scripts instead of the β-platelets. Addition of Co was however found to have a greater effect than Mn as indicated by the higher ductility records. Crepeau (1995) in his review noted that addition of 0.5%Mn to Al-Si alloys suppresses the β-phase when the Fe level is less than 1.5%. Furthermore, addition of 0.3%Mn counters the coarsening effect of Ti on β-phase platelets in Al-13Si-0.44Fe. Cho et al. (2006) in studying the effect of alloy addition on high temperature properties of Al-Si(CuNiMg) cast alloys commonly used for making automotive pistons reported that addition of up to 0.5%Mn leads to formation of a number of intermetallic phases such as α-Al(MnFe)Si as well as Al_6(MnFe). This was accompanied by significant improvement in high temperature properties. Figure 2.13 illustrates the formation of Chinese script morphology as a result of addition of 0.85%Mn to a 319-type alloy.

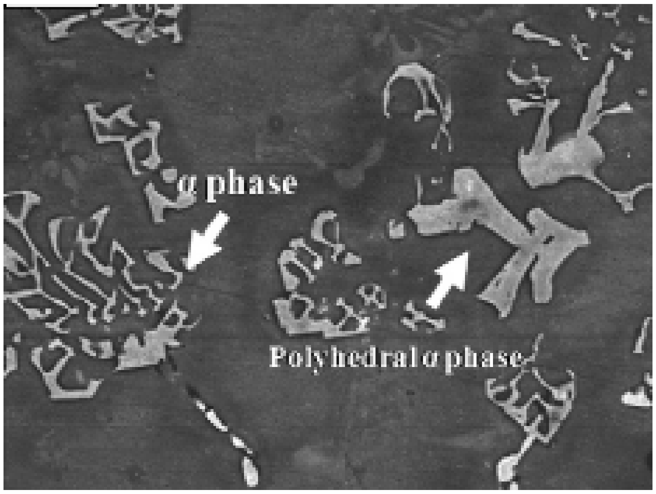


Figure 2.13: Formation of Chinese script morphology in a 319-type alloy after addition of 0.85%Mn [Hwang et al. (2008)]

2.4.4.2.3 Effect of Chromium

Use of Cr to counteract the negative effects of Fe has been applied. Narayanan et al. (1995)] noted that addition of Cr to Al-Si alloys alters the morphology of the deleterious β-phases and enhances precipitation of phases that are less detrimental. Crepeau (1995) noted that addition of Cr to variants of Al-7Si-0.3Mg alloys causes the coarse β-Al_5FeSi platelets to be replaced by the Chinese script phases identified as α-$Al_{13}(Fe,Cr)_4S_4$ phase with improved properties. Furthermore, Crepeau reported that addition of Cr at levels of between 0.2 - 0.6% prevents the embrittlement of Al-13Si alloys for Fe levels that are greater than 1%. Hatch (1990) reported that addition of Cr to the A356 alloy containing an Fe level of about 0.5% causes replacement of the coarse β-phases. Gustafsson et al. (1986) observed that Cr addition to Al-7Si-0.3Mg alloy with 0.5%Fe transformed the morphology of the β-phases into a Chinese script morphology similar to that which is well established for additions of Mn to casting alloys of this type. Recently, Mahta et al. (2005) demonstrated that addition of Cr to A413-type alloy leads to transformation of the β-phases to less harmful phases.

2.4.4.2.4 Effect of Strontium

Sr addition to Al-Si casting alloys has long been used to modify the eutectic Si with a concomitant improvement in properties. Recently however, this addition has been shown to modify the morphology of the needle Fe-rich phases by inhibiting their coarsening effect. The level of Sr required for this purpose has been noted to be higher than that used for modification of the Al-Si eutectic. Reported cases where higher levels of Sr cause the β-phase transformation include that of Samuel et al. (1996) who indicated that addition of 300ppm Sr to 319-type alloy accelerated dissolution of the β-phases through fragmentation of the long intercepting needles. In addition, Villeneuve et al. (2001) also noted that addition of 300 ppm Sr to 319-type alloy causes fragmentation of the β-phases into the small pieces, which are expected to have less detrimental influence on properties. Shabestari et al. (2002) further noted that addition of higher values of Sr of about 0.1% to an A413-type alloy with 3%Mn and 2.51%Fe leads to the fragmentation of β-phases and forms star-like phases. Mulazimoglu et al. (1996) however, observed that addition of Sr as low as 30ppm to 6201 alloy with 0.003%Mn and 0.29%Fe can stabilize the α-phase with very little β-phases appearing in the microstructure.

Other contributions on the role of Sr in modifying the β-phases have been made by Shan-Noon et al. (1999) who indicated that addition of 0.031%Sr to Al-11Si casting alloys refines the needle Fe-rich phases. Furthermore, Cho et al. (2008) reported that addition of 220ppm Sr to Al-10Si-1.1Fe alloys causes refinement of Fe-rich β-phases. In addition, the authors noted that Sr addition suppresses the branching of β-Al_5FeSi by poisoning the preferential nucleation sites and also by partial dissolution of β-Al_5FeSi platelets resulting in fragmentation of the β-phases. Ashtari et al. (2004) reported that addition of 0.014%Sr individually or in combination with 0.31%Mn to an Al-Si-Cu-Fe alloy successfully modified the branched β-phases into Chinese script or spherical morphologies.

2.4.4.2.5 Effect of Superheat

By subjecting Al-Si alloy to high temperature, an action commonly known as melt superheat, the morphology of the β-phases can be changed to a less detrimental one. A number of contributions on the effect of superheat on β-phases have been made. Mondolfo et al. (1976) reported that melt superheating reduces the size of Fe-intermetallic compounds. Xiufang et al. (1992) further indicated that the morphology of Fe compounds change from long needle like form to rosettes and to spheroidised shape as the melt temperature increases from 840 to 920˚C. On the other hand, Crepeau (1995) noted that if the melt superheat is increased up to 500˚C above the melting point, the Fe-rich intermetallic particles in the casting become finer.

2.4.4.2.6 Effect of Other Fe Neutralizers

Use of other neutralizers such as Be, Ca and Co have been reported. Figure 2.14 shows that addition of Be and Mn leads to formation of the Chinese script morphology instead of the β-phases. On the other hand, Ca addition in the range of 0.02 to 0.05% has been reported to transform the β-phase into the α-phase in Al-(0.45-0.9)Mg-(0.2-0.6)Si alloy, thereby improving the surface characteristics of extrusions [Sreeja et al. (2006)]. Murali et al. (1994) noted that trace additions of Co serves to tie up the Fe-phases to form new phases with Chinese script shapes.

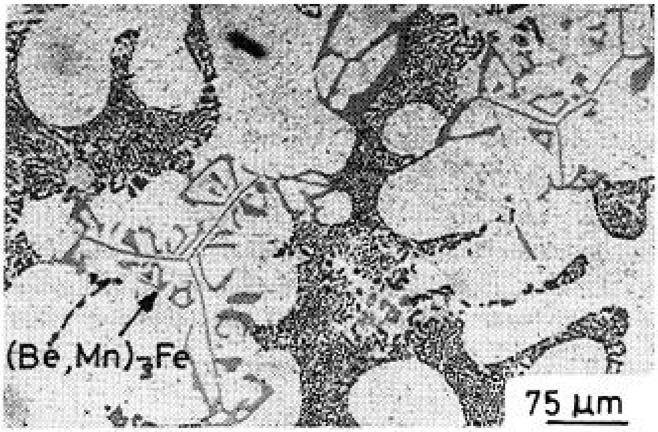


Figure 2.14: Microstructures of sand cast modified Al-7S1-0.3Mg -0.76%Fe alloy with trace additions of Mn and Be [Murali et al. (1997)]

2.4.5 Effect of Grain Refinement

Addition of grain refiners to aluminium casting alloys is a common practice in many commercial foundries and is reported to impart significant influences on the microstructure of Al-Si alloys. Originally, grain refinement was done with a view to suppress columnar grain growth in castings and promote formation of a fine equiaxed structure. Later, Gruzleski and Closset (1990) noted that a grain refined cast part is more homogeneous with better casting soundness and increased mechanical properties than a non grain refined one. In addition, the authors observed that grain refinement confers greater resistance to hot tearing. Barresi et al. (2000) gave further evidence that grain refinement provides other beneficial effects to aluminium foundry alloys which include reduced porosity and improved feeding. Couper et al. (2005) reported that addition of low levels of Ti in form of Al-5Ti-1B to A356.2-type alloy is beneficial to properties. However, higher levels of Ti results in coarse Al-Ti-Si particles that are linked to reduction of tensile elongation during mechanical testing. Liao and Sun (2004) demonstrated that addition of B in the range of 0.012 to 0.036 added as Al-1B to Al-11.6Si-0.4Mg casting alloys modified with Sr gives strong refinement of the dendritic α-phase. However, addition of higher levels of B (0.44%) degenerates the shape and size of the dendrites probably due to the poisoning effect that Sr and B could have. Some reports [Easton and StJohn (2000)] do indicate that grain refinement can lead to increased porosity depending on the amount added and the type of casting.

2.4.6 Typical Microstructure of Piston alloys

The typical microstructure of an Al-Si piston alloy, an example of the type which is studied in the present work is shown in Figure 2.15. It is to be noted that new generation high temperature performing piston alloys have been designed to contain a high content of nickel and copper. Moffat (2007) indicated that when Ni and Cu are added jointly they combine to form different types of phases including Al_3Ni, $Al_3(NiCu)_2$ and Al_7Cu_4Ni. It is evident from the microstructure that in addition to the AlCuNi phases, the silicon phases also do form. Furthermore, other Fe-bearing phases such as AlFeNi and AlFeCu also form in the alloy as reported by Moffat (2007) and are especially harmful to fatigue properties. Moffat (2007) reported that AlFeNi act as crack initiation sites leading to low fatigue properties.

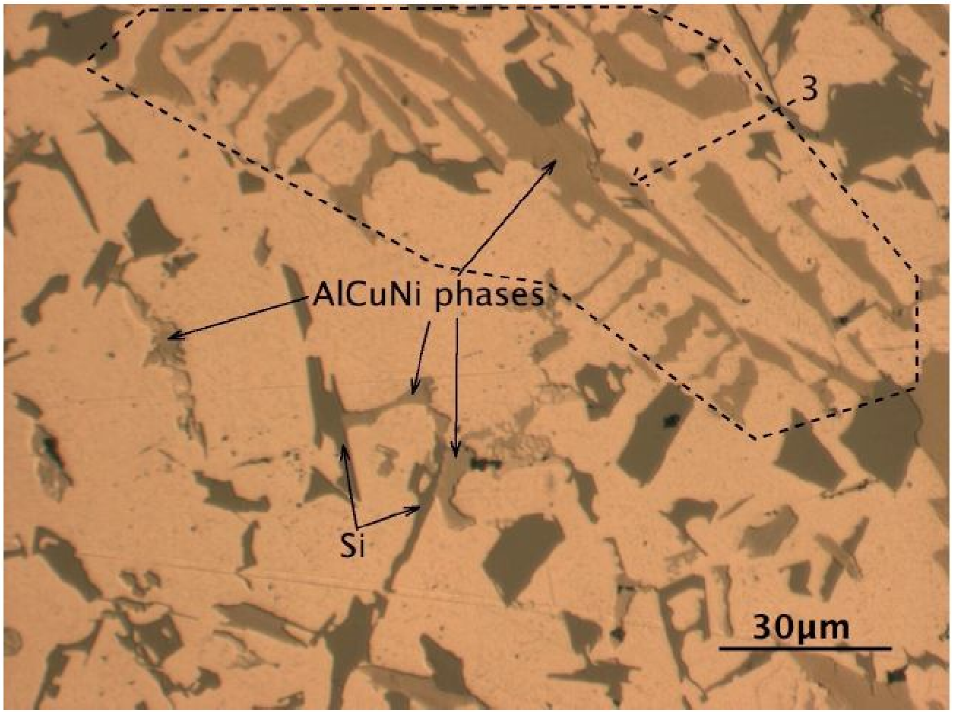


Figure 2.15: Typical microstructure of an Al-Si piston alloy with AlCuNi phases [Moffat (2007)]

2.5 Mechanical Properties

2.5.1 Introduction

In engineering applications, many components including automobile wheels, engine blocks, cylinder heads and pistons are subjected to various loadings which tend to shorten their service life and limit the mechanical performance. It is to be noted further that the quality and reliability of cast aluminium components are largely threatened by casting defects such as porosity, inclusions, and oxide films that may be present in their structure. In order to eliminate these defects and improve the mechanical performance of the alloys, close control of casting parameters such as mould design, pouring temperatures, use of chills to achieve directional solidification and melt treatment procedures have been employed. In addition, alloying elements and heat treatments have been utilized to counter undesirable influences of Fe and eutectic Si and improve properties. Unfortunately, there is considerable scatter in literature as to how elements additions affect properties. Furthermore, available data may be difficult to interpret. It is therefore necessary that an in-depth understanding of the subject be attempted so as make the data on mechanical properties reliable.

2.5.2 Alloy Effect on Mechanical Properties

Variation of alloy composition during manufacture of cast aluminium components may impart significant changes in mechanical properties of these alloys [Murali et al. (1994)]. In the following sections the effect of alloying elements on mechanical properties are reviewed.

2.5.2.1 Effect of Modifiers

The shape, size and distribution of eutectic Si particles can impart significant influences on the mechanical properties of cast aluminium alloys. Wang et al. (2003) indicated that optimal properties of aluminium foundry alloys can be obtained when the eutectic Si particles are small, spherical and evenly distributed. On the other hand, when eutectic Si particles are present in the form of large flakes, the mechanical properties of the commercial aluminium alloys are adversely affected. In order to mitigate the negative influences of the coarse eutectic Si flakes and optimize the mechanical properties of aluminium foundry alloys, modification is done using such elements as Sr and Na. The improvements that come with Sr- and Na- modification can however be negated by increase in porosity. Use of Sb as a modifier has been employed with the advantage of not only decreasing porosity, but also causing significant improvement in mechanical properties. In addition, there is considerable evidence that Ca modifies the eutectic Si with the advantage of decreased microporosity in the castings [Abdollahi and Gruzleski (1998)].

Closset (1988) reported that modification improved unnotched impact strength of both as-cast and heat treated A356 and 413 alloys. For the A356 alloy and in as-cast condition, the impact value increased from 9.5 J/cm^2 for unmodified structure (0%Sr) to 18.7 J/cm^2 for a modified structure (0.016%Sr). Jenabali et al. (2004) conducted studies on the effect of modifiers (Sr and Sb) on mechanical properties of A356 aluminium alloy and observed that the impact energy of unmodified, 0.013%Sr modified and 0.1%Sb modified samples were in the sequence of 6.2, 10.4 and 7.5 J/cm^2 for unnotched specimens. Shivkumar et al. (1994) reported that refinement of the eutectic structure through addition of Sr-containing master alloys improves the impact properties of A356-T6 castings. They further noted that Sr modification can be used to effectively lower the heat treatment times leading to a significant reduction in the overall cost of the finished component. The traditional method of preparation of specimens for impact testing was to provide for a notch. Recent research has however, recommended unnotched specimens since it is believed that the technique, the size and shape of specimen used for impact testing will have great bearing on the validity of the results obtained. The accuracy of the measurements from unnotched specimens is popular because it emphasizes on the microstructural effects. Taking note that, for brittle materials, the impact strength is very low it can be decreased further by the presence of a notch. Besides, if a notch is present, the absorbed energy may be more dependent on the notch geometry than on the microstructure. Paray et al. (2000) observed that for Al-6.8Si-0.3Mg in T6 condition, the values for U-notched, V-notched and unnotched samples were 7.26 J/cm^2, 5.10 J/cm^2 and 22.16 J/cm^2 respectively. This makes data from unnotched specimens more reliable.

2.5.2.2 Effect of Fe

Iron leads to formation of Fe-bearing intermetallic compounds the most common being α-Al_8 Fe_2Si, β-Al_5FeSi and π-$Al_8Mg_3FeSi_6$ as earlier mentioned. The presence of Fe-bearing intermetallic compounds increases hardness but decreases most of the other mechanical properties such as ductility [Mbuya et al. (2003)]. Ravi et al. (2002) reported that the yield strength, UTS and percent elongation in T6 condition decreased with increase in Fe content from 0.2% to 0.6% as shown in Figure 2.16 below. Kim et al. (1996) in studies to determine the effect of Fe and Ca on properties of AC2B-T6 casting alloy observed that the percent elongation of this alloy consistently decreased with increase in Fe as shown in Figure 2.17. Couture (1981) indicated that the ultimate tensile strength and elongation of Al-5Cu alloy decreases from 370-395 MPa to 260-285 MPa and from 12-13% to 3-4% respectively, when the Fe content is increased from 0.25 to 0.6%.

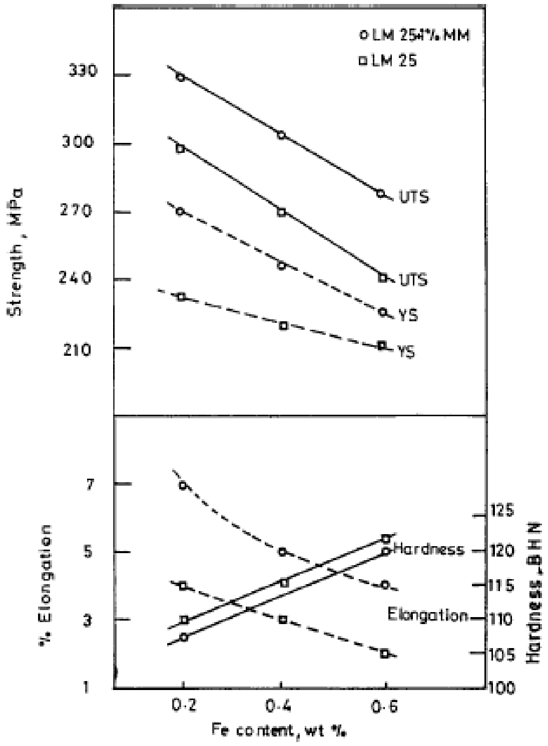


Figure 2.16: Mechanical properties of the LM 25 alloys in T6 condition containing Fe and Mischmetal [Ravi et al. (2002)]

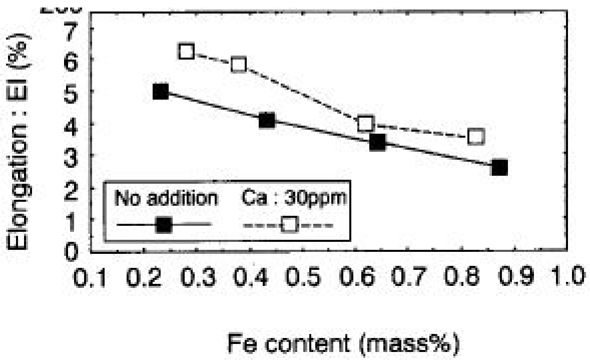


Figure 2.17: Effects of Fe and Ca on the percent elongation of AC2B-T6 aluminium casting alloys [Kim et al (1996)]

Although it is widely accepted that high Fe content confers inferior mechanical properties to Al-Si alloys due to formation of Fe-rich phases, different explanations have been advanced for this observation. Li et al. (2004) indicated that reduction in properties of Al-Si alloys containing Fe is due to the brittleness of the Fe-rich phases. Cao and Campbell (2003) suggested that diminished mechanical properties could be more likely due to the presence of pre-cracked central axis of the phases because of the presence of folded oxide films. The oxide film is believed to have two sides: the dry inner surface and the wetted outer surface. The wetted outer surfaces of the oxide film act as preferred substrates for nucleation and growth of β-phases. It could therefore be that the gap between the two dry sides of oxide films constitutes the observed cracks and deteriorated properties either in the intermetallics or in the matrix because of incomplete bonding across the plane. In addition, Cao and Campbell (2006) reported that Fe-rich phases are detrimental to mechanical properties because they act as stress raisers and are points of weak coherence with the Al matrix that are greatly vulnerable to failure.

Some early experimental work on charpy impact test in Fe-containing alloys include that of Vorren et al. (1994) who conducted studies on the effect of Fe on fracture toughness of aluminium foundry alloys. They found out that an increase in iron content from 0.15% to 0.3%Fe in as-cast Al-7Si-0.3Mg alloy resulted in a reduction of 20-25% in charpy impact energy while a further increase to 0.4%Fe resulted in no decrease. Murali et al. (1992) varied the iron content in the range of 0.2-0.6% and observed a significant and continuous decrease in charpy impact energy of Al-7Si-0.6Mg-T6 alloy. Ma et al. (2003) observed that the impact strength exponentially decreases as the β-Al_5FeSi content increases as shown in the Figure 2.18.

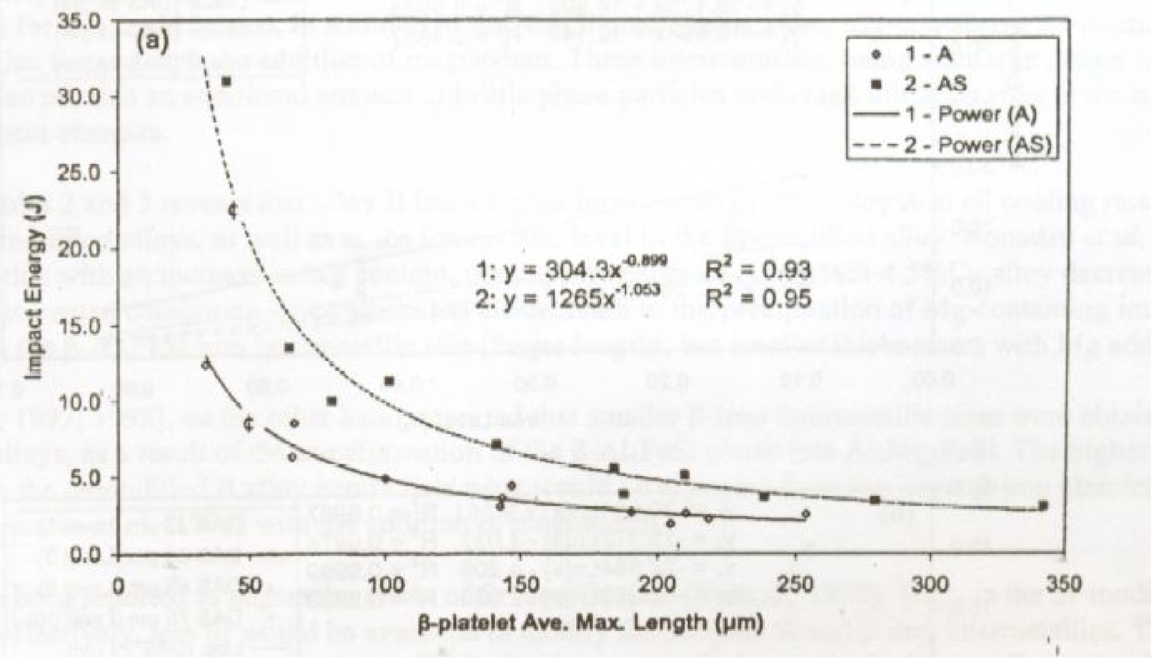


Figure 2.18: Impact energy versus β-platelets average maximum length [Ma et al. (2003)]

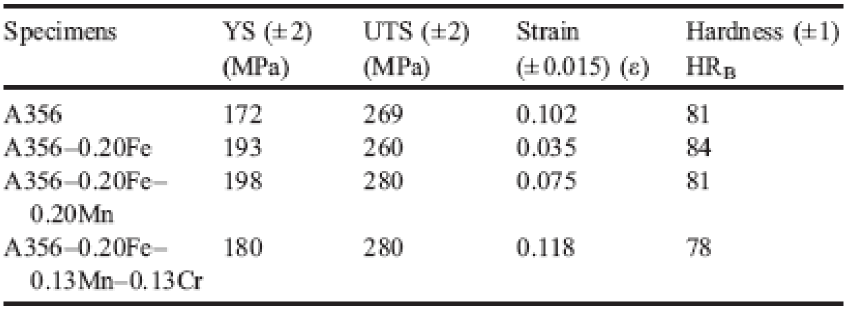
Whereas extensive literature has been published on the effect of Fe in aluminium foundry alloys, lack of agreement on the degree of influence on mechanical properties has been reported [Mbuya et al. (2003)]. Some researchers have reported that Fe increases tensile and yield strength of aluminium foundry alloys while others have observed a decrease in both tensile and yield strength. Grand (1964) reported that addition of more than 1%Fe to Al-5Zn-Mg alloy leads to decrease in tensile properties and impact strength. Badia (1966) studied the effect of composition on properties and reported that room temperature tensile and elongation of Al-11.5Si-5Ni alloy drastically reduced when the Fe content increased from 0.1 to 1.3%. Wang et al. (2003) indicated that increasing the Fe intermetallics of A356 alloy results in reduction of the UTS by about 8%. The yield strength was unaffected. Ravi et al. (2002) reported that the general effect of increasing Fe from 0.1 to 0.9% in sand cast heat treated Al-7Si-0.3Mg alloy was to decrease the tensile strength and slightly increase the yield strength and Brinell hardness. It is however, in principle agreeable that Fe is detrimental except in high pressure die casting (HPDC) where it is employed to mitigate the die soldering problem. Other beneficial effects of Fe include increased elevated temperature strength, wear resistance and chip breaking during machining [Bangyikhan (2005)].

2.5.2.3 Effect of Neutralization

Gustafsson et al. (1986) reported that Cr addition to Al-7Si-0.3Mg alloy with 0.5% Fe altered the morphology of β-plates into the less detrimental Chinese script morphology with subsequent improvement in tensile ductility and fracture toughness. Crepeau (1995) noted in his review on effect of Fe on Al-Si casting alloys that addition of 0.2-0.3%Cr to Al-5Si-1.5Cu-0.5Mg increased the elongation from 1.7 to 3.8% when the Fe level was 0.4%. For the same alloy, the author noted that addition of 0.4%Cr increased the elongation from 0.8 to 2.6% when the Fe level was 0.75%.

Couture (1981) reported that in a 4.5% Cu alloy in which the Fe content varied from 0.73 to 1.06%, the strength and elongation increased from 250 MPa and 5% respectively to 275 MPa and 5.5% in the presence of 0.55%Mn. Bonsack (1942) observed that addition of 0.5% Mn to Al-13Si-1.5Fe alloy transformed the β-plates into Chinese script that negated the detrimental effect of Fe on mechanical properties. Zednik (1948) reported that addition of 0.44% Co or 0.41% Mn to Al-9.1Si-0.22Mg-0.4Fe alloy produced Chinese script phases instead of β-plates with high recorded values of ductility. However, Co addition resulted in higher ductility than Mn addition. Kim et al. (2006) studied the effect of alloying elements on the mechanical properties of the A356 type alloy and observed that addition of 0.20%Fe lowers the UTS and strain and attributed this to the occurrence of the acicular β-Al_5FeSi during solidification and aging heat treatment. Addition of Mn either individually or in combination with Cr improves the mechanical properties of the alloys as shown in Table 2.2.

Table 2.2: Mechanical properties of A356 alloy with and without alloying elements [Kim et al. (2006)]



Specimens	YS (±2) (MPa)	UTS (±2) (MPa)	Strain (±0.015) (ε)	Hardness (±1) HR_B
A356	172	269	0.102	81
A356–0.20Fe	193	260	0.035	84
A356–0.20Fe–0.20Mn	198	280	0.075	81
A356–0.20Fe–0.13Mn–0.13Cr	180	280	0.118	78

In Al-7Si-0.3Mg alloys, Be addition transforms the β-plates into Chinese script and improves properties [Murali et al. (1994)]. Sreeja et al. (2007) noted that small additions of Be to Al-Si-Mg-0.8Fe alloys significantly improve the mechanical properties of the alloys due to modification of the eutectic Si as well as morphological change from the β-plates into Chinese script form. In addition, the authors noted that Be preferentially oxidizes forming BeO on the surface of the casting. This in turn reduces Mg loss making it available for formation of the required volume of Mg_2Si for strengthening the α-Al matrix. Furthermore, Be addition reduces oxidation rate of the melt leading to reduced porosities (oxide inclusions account for increase in porosity as reported by McDonald et al., 2004). The observed improvements in properties due to Be may in part be attributed to its contribution in reducing Mg losses as well as its reduction in melt oxidation. Beryllium is however, carcinogenic and most foundrymen are reluctant to use it.

Ca and Sr have recently been employed in refinement of the β-plates, an effort that has led to significant improvement in the mechanical performance of Al-Si alloys containing Fe [Sreeja et al. (2007)]. This is besides their original purpose of modifying the eutectic Si. Ma et al. (2003) noted that addition of Sr can be employed to neutralize the detrimental effect of β-plate like intermetallics in 319- and 356-type alloys and improve the impact strength even at high Fe levels. The summary of results on element effect on UTS and impact strength from studies by Sreeja et al. (2007) is shown in Figure 2.19. It is noted that the best combination of properties are obtained in combined additions of Be + Mn, Ca + Mn and Sr + Mn.

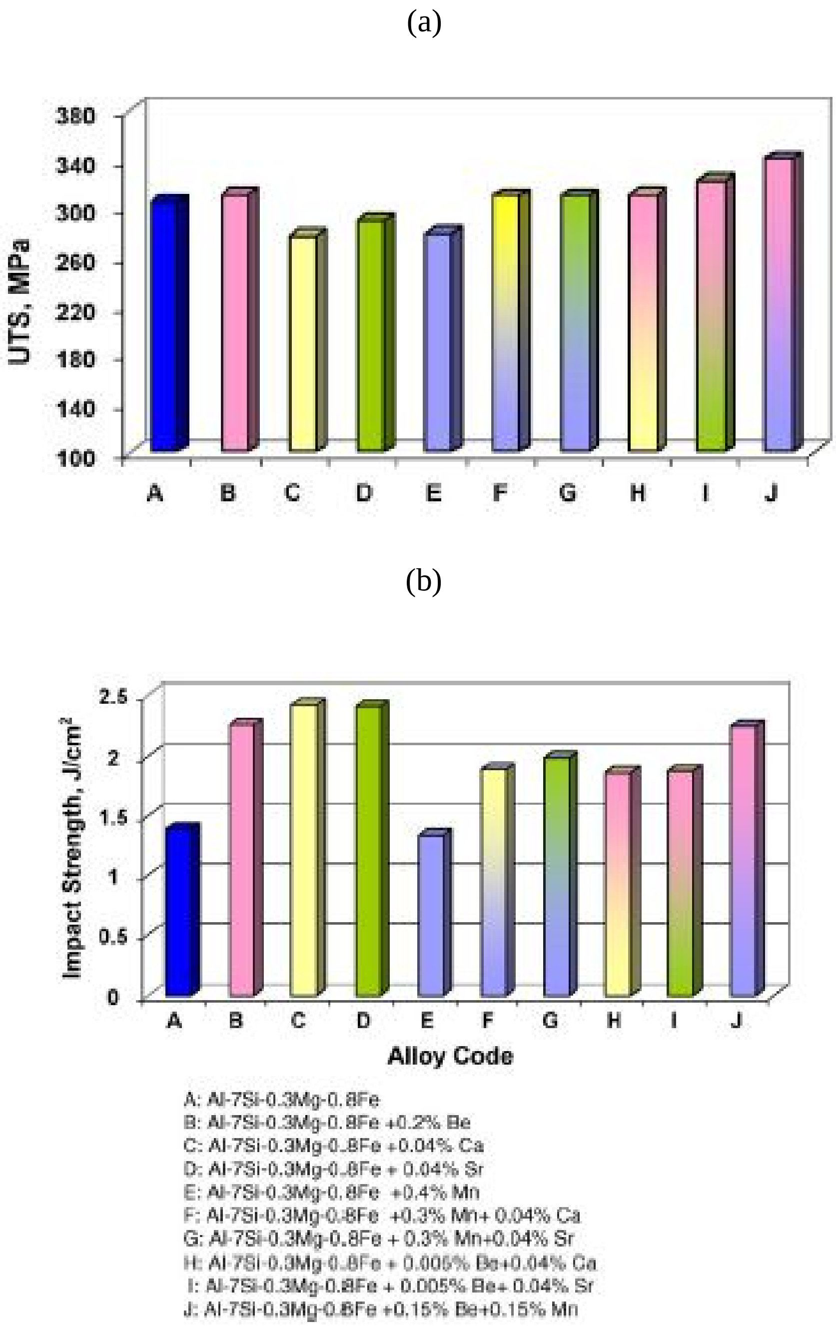


Figure 2.19: Effect of Be, Mn, Ca and Sr individually and in combination on Al-7Si-0.3Mg-0.8Fe alloy (a) UTS and (b) Impact strength [Sreeja et al. (2007)]

2.5.2.4 Effect of Grain Refinement and Si

Couper et al. (2005) noted that addition of TiB_2 grain refiner improved the elongation of A356.2 foundry alloy without causing detrimental effect on yield stress or UTS. Furthermore, Smith et al. (2004) reported that grain refinement by TiB_2 (added as Al-Ti-B) increased the ductility of the alloy up to 4% at shorter aging times. After aging for 4 hours however, the authors noted that the ductility decreased to about 1%. Si has also been reported to have significant influences on the mechanical properties of Al-Si alloys. Paray et al. (2000) noted that increasing the amount of Si to the eutectic composition results in formation of a large volume fraction of eutectic which leads to improved fluidity and feedability but decreased ductility. In addition, Tsukuda et al. (1978) noted that the charpy impact value rapidly decreased at Si content of 6% to 8% or more when the amount of Si was varied from 1% to 13% in an Al-Si-0.15Sb alloy.

2.6 Summary

The use of Al alloys in the manufacture of automotive, aerospace and marine components currently experience a great demand but at the same time the alloys suffer unforeseen problems arising from castability, microstructure and mechanical properties. The presence of Fe-rich intermetallic phases particularly the brittle β-Al_5FeSi platelets in the microstructure of Al-Si alloys causes significant reductions in mechanical properties, especially ductility and tensile strength. Furthermore, existence of β-platelets in the microstructure of these alloys causes dramatic decrease in interdendritic permeability leading to formation of shrinkage pores. Porosities constitute a major casting defect that deteriorates properties in these alloys. Other defects like misruns do occur due to insufficient fluidity leading to high rates of casting rejects.

Finding creative solutions in order to achieve high success in the design of high performance complex Al castings for long term service has become a major challenge for Engineers as well as the casting community. Efforts to reduce casting defects and have sound castings with improved mechanical performance have continued to be achieved through melt cleanliness and design of proper filling techniques.

Use of alloying elements has been noted as an effective and promising approach for reduction of defects and improvement of properties. Unfortunately however, the available data on effect of alloying elements is not only meager in some cases but also confusing and contradictory. The present research therefore, was an attempt to investigate the effect of minor alloying elements on the castability, microstructure and mechanical properties of Al-Si alloys as part of the wider effort to improve the performance of these alloys. In addition, a survey of the local foundries was conducted to find out the casting practices and quality control measures adopted by the foundrymen to ensure production of premium quality castings.

CHAPTER THREE

3.0 METHODOLOGY

In this chapter the methodology pertinent in studying three cast Al alloys namely automobile wheels, cylinder heads and pistons is presented. First the procedure of melting separately the three categories of scrap and preparation of samples for chemical analysis is described. This is followed by experimental details on how fluidity and porosity measurements were done. Details of microstructural characterization and mechanical tests (tensile, ductility, impact and Brinell hardness) are also given.

3.1 Alloy Preparation

The alloys used in this study were obtained by separately melting aluminum wheels, pistons and engine cylinder head scrap in a 70 kg capacity oil fired graphite crucible furnace located in Jomo Kenyatta University of Agriculture and Technology. Each scrap category was charged into the furnace and melted under a cover flux to a temperature of 730˚C. The melt was skimmed off just before pouring and then transferred directly into 4 kg capacity ingot moulds fabricated from mild steel sheets. The melt was then allowed to cool before the ingots were removed. Samples for chemical composition analysis were prepared by pouring small samples into a copper mold which was immediately quenched in cold water to avoid compositional segregation. The chilled samples were sent to LSM in the UK for chemical analysis using inductively coupled plasma spectroscopy and the result is given in Table 3.1. In the table, the alloy obtained from wheels is coded as WLS. As expected, its chemical composition shows that it is equivalent to the British LM 25 alloy, the US 357 alloy or the Japanese AC4CH. The alloy obtained from cylinder heads is coded as CH and is equivalent to LM 27 or AC4B as expected. However, alloy CH has a slightly low Cu content (1.38) compared to specified Cu limits for LM 27 (1.5 to 2.5 - all compositions are in weight percent throughout). The CH alloy also has a slightly higher Si content (8.78) than the specified limit for LM 27 (6 to 8). The alloy obtained from pistons was coded as PS and as expected, its chemical composition shows that it is equivalent to the British LM 13 alloy, or the Japanese AC8A. However, the alloy has a slightly low Mg content (0.78) compared to the specified limits for LM 13 (0.8 to 1.50). The alloy also has a slightly higher Fe content (1.06) than the specified limit for LM 13 (1.00 max).

Table 3.1. The chemical composition (in wt %) of the alloys used in this study.

Alloy		Si	Mg	Mn	Fe	Pb	Cr	Ti	Sn	Cu	Zn	Ni
WLS		6.76	0.55	0.04	0.14	0.01	0.03	0.18	0.01	0.02	0.02	<0.01
Equivalent Alloys	LM25	6.50-7.50	0.20-0.45	0.30 max	0.50 max	0.10 max	-	0.20 max	0.05 max	0.10 max	0.10 max	0.10 max
	357	7.0	0.55	-	<0.15	-	-	-	-	<0.05	<0.05	-
	AC4CH	6.5-7.5	0.20-0.40	0.10 max	0.20 max	0.05 max	0.05 max	0.20 max	0.05 max	0.10 max	0.10 max	0.05 max
CH		8.79	0.17	0.24	0.41	0.01	0.01	0.01	0.01	1.38	0.08	0.02
Equivalent Alloys	LM27	6.0-8.0	0.30 max	0.20-0.60	0.80 max	0.20 max	-	0.20 max	0.10 max	1.5-2.5	1.0 max	0.30 max
	AC4B	7.0-10.0	0.50 max	0.50 max	1.0 max	0.20 max	0.20 max	0.20 max	0.10 max	1.0-1.5	1.0 max	0.35 max
PS		10.60	0.78	0.08	1.06	0.01	0.03	0.06	0.02	1.36	0.09	1.08
Equivalent alloys	LM 13	10.0-12.0	0.80-1.5	0.50 max	1.0 max	0.10 max	-	0.20 max	0.10 max	0.70-1.50	0.50 max	1.5 max
	AC8A	11.0-13.0	0.70-1.3	0.15 max	0.80 max	0.05 max	-	0.20 max	0.05 max	0.8-1.3	0.15 max	0.80-1.50

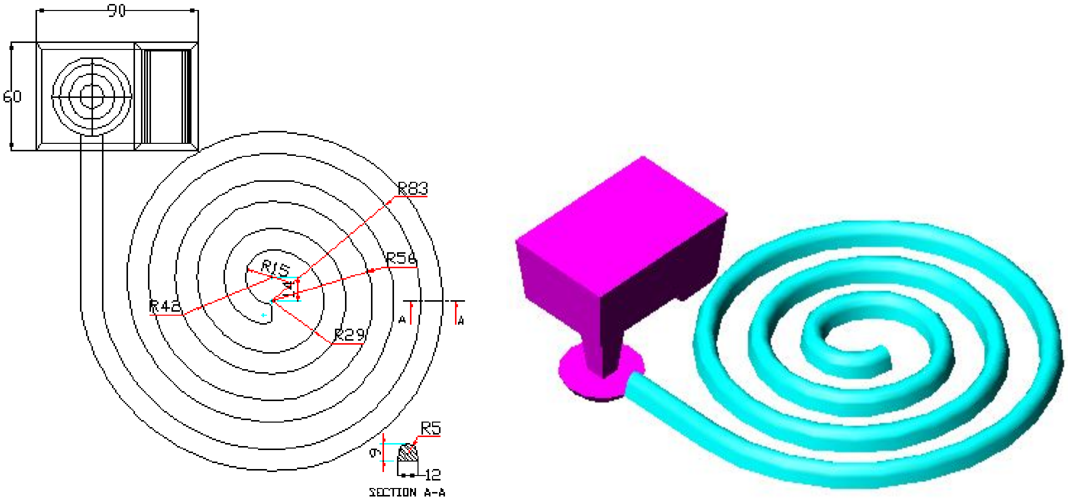
Note: The chemical compositions for WLS, CH and PS were carried out by LSM of UK.

WLS had Sb and P levels of <0.005% each, CH had Sb level of 0.05% and P level of <0.01%. B and Sr levels were <0.01% and PS had a Sr content of <0.01% and a K level of 0.04%.

3.2 Fluidity Measurement

3.2.1 Fabrication of Spiral Patterns

First a permanent spiral mould was designed and machined from cast iron. Aluminium scrap was then melted in an electric box furnace and cast into the mould to produce the required spiral pattern that was subsequently used in the preparation of the spiral sand moulds. The spiral pattern geometry was as shown in Figure 3.1.

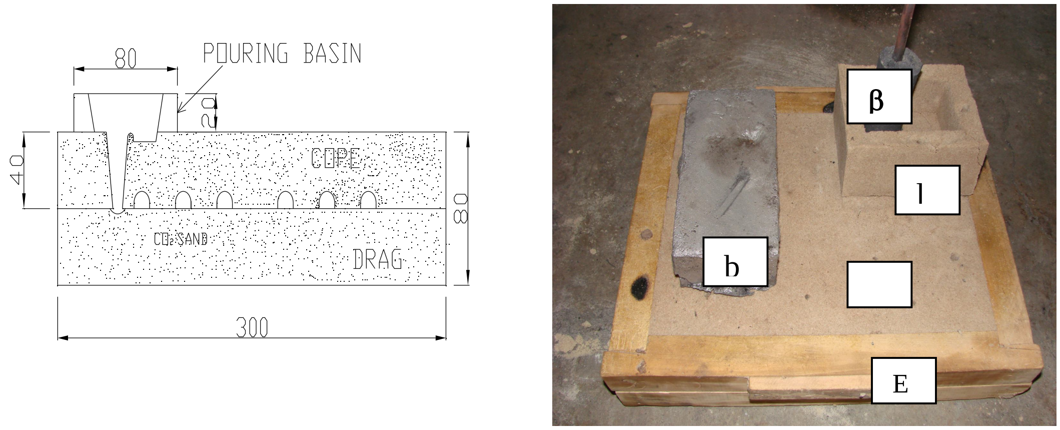


(a) (b)

Figure 3.1: (a) The dimensional details of the spiral fluidity mold (in mm) and (b) a 3-D rendering of the spiral pattern.

3.2.2 Preparation of the Spiral Sand Moulds

Moulds were prepared from CO_2/sodium silicate bonded silica sand. The moulds consisted of a cope with a spiral cavity and a flat drag in wooden flasks measuring 300x280x40mm. The filling system consisted of an offset weir rectangular pouring basin that was used in conjunction with a sand moulded stopper during pouring. The sprue was tapered and rectangular and a small hemispherical shaped well was incorporated at its sprue base. Figure 3.2 show a schematic sketch of the mould assembly.



(a) (b)

Figure 3.2: (a) The geometry of the spiral sand mould and (b) the mould assembly. The main parts shown in (b) are: A-stopper rod, B-pouring basin, C-clamping block, D-sand mold, E-wooden flask.

3.2.3 Element Additions

In this study, the effect of Fe, Mn, Cr, grain refinement by Ti-B based grain refiner (as Al-5%Ti-1%B master alloy) and eutectic Si modification by Sr (as Al-10Sr master alloy) on the fluidity of LM25 and LM27 alloys was investigated. The grain refiner and Sr modifier were available in the form of metallic rods while Fe, Mn and Cr were available as Altab Iron 75%, Altab Manganese 75% and Altab Chrome 75% briquettes. The levels of the various elements investigated are shown in Table 3.2.

Table 3.2: Various levels of the elements investigated for LM25- and LM27-type alloys

LM25 base alloy + any additions	**LM27 base alloy + any additions**
LM25	LM27
LM25 + 0.015%Sr	LM27+ 0.02%Sr
LM25 + 0.28%Al-5Ti-1B	LM27 + 0.28%Al-5Ti-1B
LM25 with 0.4% Fe	LM27 + 0.02%Sr +0.28%Al-5Ti-1B
LM25 with 0.48%Fe	LM27 with 0.6%Fe
LM25 with 0.6%Fe	LM27 with 1.01%Fe
LM25 with 0.6%Fe + 0.3%Mn	LM27 with 0.6%Fe +0.3%Mn
LM25 with 0.6%Fe + 0.6%Cr	-
LM25 with 0.6%Fe + 0.3%Mn + 0.2%Cr	-

Note: LM25 base alloy contained 0.14%Fe, 0.04%Mn, and 0.03% Cr, while LM27 base alloy contained 0.41%Fe, 0.24%Mn, and the values indicated represent the total % weight after addition.

3.2.4 Melting and Pouring

The ingots were charged into a graphite crucible and melted in an electric muffle furnace and held at between 720 to 750^{o}C during alloying. The melt was then allowed to settle for about 30 minutes and then poured into two spiral moulds at 720^{o}C for each of the tests. After each pour, the melt was returned to the furnace to allow its temperature to be maintained at 720^{o}C for each pour. A stopper was used during pouring to allow the melt to fill the pouring basin after which it was removed to allow flow into the spiral cavity. After solidification and cooling, the sand moulds were collapsed, the fluidity spirals retrieved and their lengths measured. The same procedure was repeated for each test. For the base alloy, pouring was done in the same manner except that no additions were made to the melt. In this experiment, melt temperature was strictly maintained at 720^{o}C in the furnace just before pouring for each test. An equivalent time gap was maintained for each test from the point of opening the furnace to withdrawing the stopper to allow metal flow to the spirals. A few tests in which the pourer was either too first or too slow were disregarded. This method of controlling the pouring temperature was adopted instead of the conventional technique of temperature measurement at the pouring basin because of unavailability of a temperature acquisition system. The method was found to be effective and in hindsight simulated a practical way of controlling temperature in a commercial foundry.

3.3 Porosity Measurement

To obtain specimens for porosity measurement, a permanent mould was designed and fabricated in a manner described below. The addition rates of the various elements used in the study is also given. The castings from the permanent mould were also used for preparation of specimens for microstructure and mechanical tests.

3.3.1 Fabrication of the Permanent Mould

A permanent mould was designed and machined from cast iron with two sections: one having a hot spot to encourage formation of shrinkage pores and the other well fed to minimize porosity formation. The ingot samples were then melted in an electric muffle furnace and cast into the permanent mould to produce the required castings that were subsequently used to investigate porosity characteristics, microstructural features, hardness, tensile and impact tests. Details of the permanent mould used in this study are given in Figures 3.3 (a), (b) and (c).

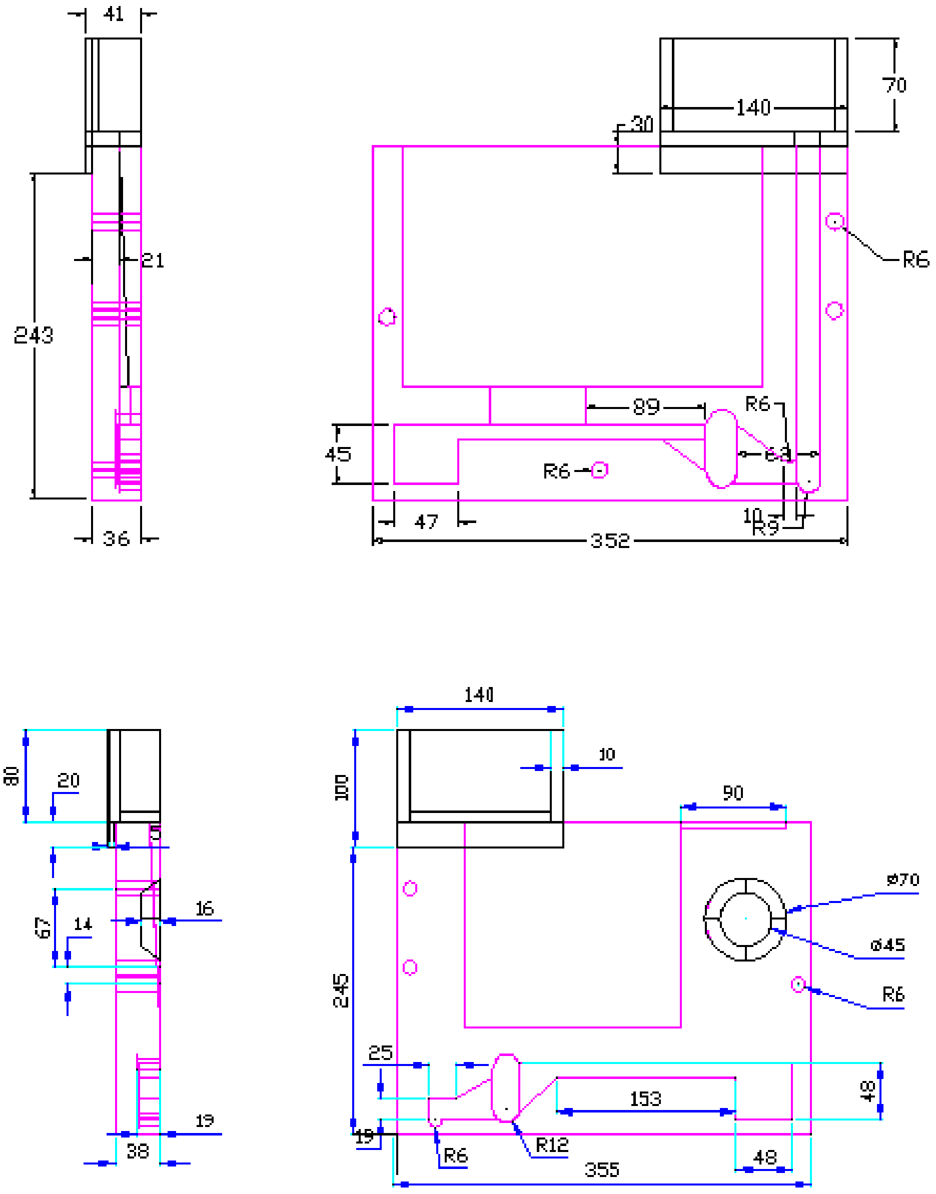


Figure 3.3 (a) Dimensional details (in mm) of both halves of the permanent mould.

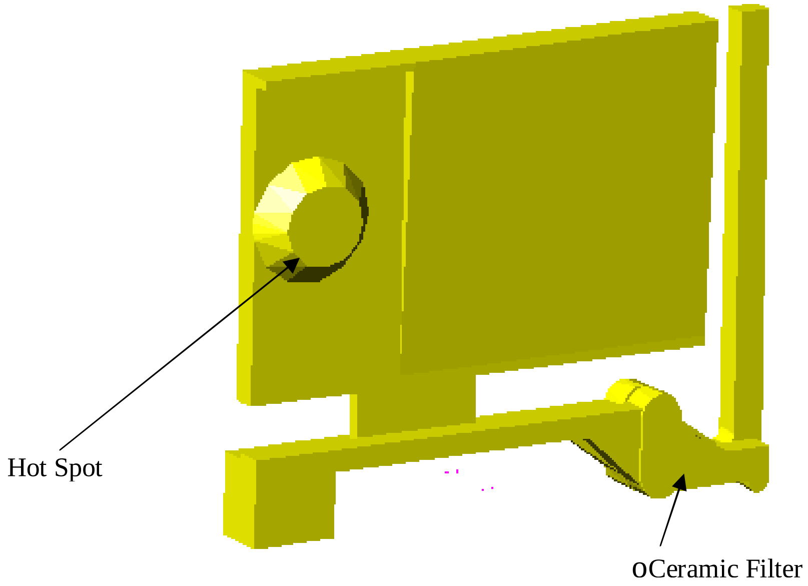


Figure 3.3 (b): The 3-D rendering of the casting

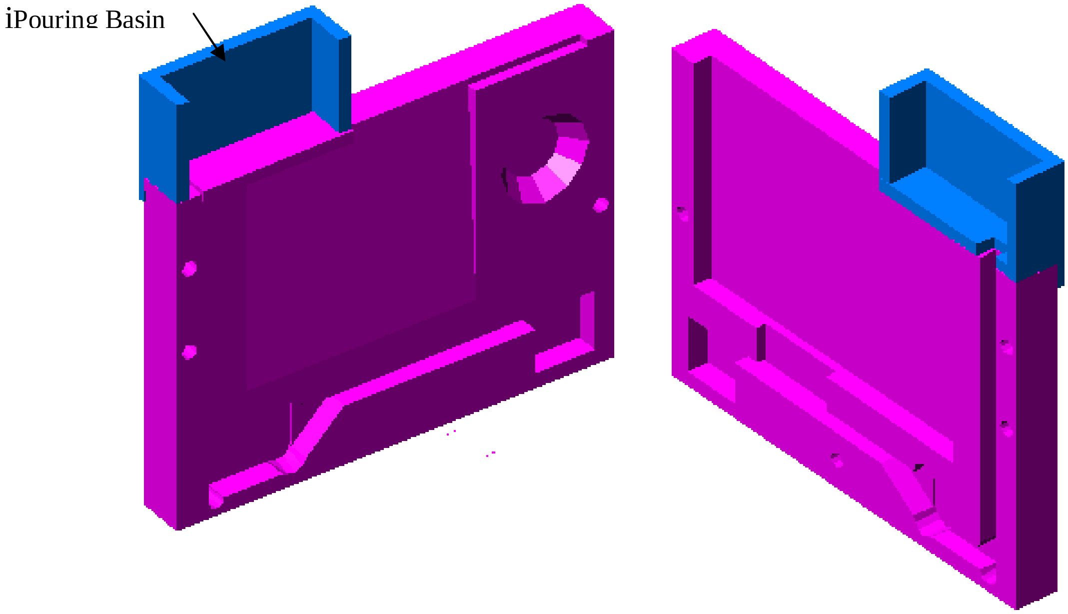


Figure 3.3 (c) A 3-D rendering of the permanent mould.

3.3.2 Element Additions

In this study, the effect of Mn, Cr, grain refinement by Ti-B based grain refiner (added as Al-5%Ti-1%B master alloy) and eutectic Si modification by Sr (added as Al-10Sr master alloy) and Sb on porosity formation of alloy PS in Table 3.1 was investigated. In addition, the effect of these elements on microstructure and mechanical properties of the alloy was investigated as will be discussed later on. The grain refiner and Sr modifier were available in the form of metallic rods while Mn and Cr were provided as Manganese 75% and Altab Chrome 75% briquettes. Sb was added as a pure metal in the form of granules. The levels of the various elements investigated are shown in Table 3.3.

Table 3.3. Various levels of the elements investigated for LM13-type alloy.

LM13 base alloy + any addition
LM13
LM13 + 0.02%Sr
LM13 + 0.05%Sr
LM13 + 0.02%Sr + 0.28%Al-5Ti-1B
LM13 with 0.53%Mn
LM13 with 1.06%Cr
LM13 with 0.3%Mn and 0.2%Cr
LM13 +0.02%Sr with 0.53%Mn
LM13 + 0.05%Sr with 0.53%Mn
LM13 + 0.2%Sb with 0.53%Mn

Note: LM13 base alloy contained 0.08%Mn, and 0.03% Cr and the values indicated represent the total % weight after addition.

3.3.3 Melting and Pouring

The 4kg ingots were charged into a graphite crucible and melted in an electric muffle furnace. The melt was held at a temperature of between 720 to 750˚C under a cover flux and then skimmed before adding the elements. In order to dissolve the additives, stirring of the melt was done gently. The melt was degassed using nitrogen gas for 15 minutes and after holding the melt for about 20 minutes, it was poured at 740^{o}C into a 470^{o}C preheated wedge shaped permanent mould shown in Figure 3.3(c) above. For the base alloy, pouring was done in the same manner except that no additions were made to the melt. A ceramic foam filter was used in the running system to serve twofold functions; to screen out inclusions and remove oxide films that might find their way into the casting and to make the flow as quiescent as possible. Studies by Campbell (2003) indicate that the action of ceramic filters is to eliminate surface turbulence in the flow of the liquid and help in reduction of oxide films and hence casting defects especially porosity. After solidification and cooling, the permanent mould was disassembled and the casting retrieved. The section with the hot spot was then cut off and used for porosity evaluation while the remaining section was used for mechanical tests as will be discussed later.

3.3.4 Preparation of Specimens for Analysis of Porosity Characteristics

Specimens for porosity characteristics analysis were obtained from segments from the central region of the casting with the hot spot [See Figure 3.3(b)] after having removed and discarded the edge sections. This was done because the edges of the casting tend to have higher defect concentration and also cool faster and hence may influence the trend in properties (in this case porosity characteristics) as suggested by Mbuya et al. (2006). Each segment of the alloy for porosity analysis was weighed to obtain the mass in grams using an electronic weighing machine. The displaced volume was obtained by dipping each segment in an overflow can filled with water and a measuring cylinder was used to collect displaced water and indicate its volume in cm^3. With the values of mass (in g) and displaced volume (in cm^3) for each segment, experimental densities were calculated using equation (3.1). The theoretical density on the other hand was obtained from a sample from the bottom of the casting and after being heated to about 400˚C it was forged to close the pores contained in it. The mass and displaced volume of water for this sample was measured and density calculated as usual. This was considered theoretical density. The volume percentage porosity in each segment of the alloy was evaluated by comparing the experimental density and theoretical alloy density in accordance to equation (3.2).

$$\rho = \frac{Mass(g)}{Volume(cm^3)} \tag{3.1}$$

Where; ρ = Density

$$\%Porosity = 100\frac{\rho_{th} - \rho_{\exp}}{\rho_{th}} \tag{3.2}$$

Where;

ρ_{th} = Theoretical density = (mass of forged sample/displaced volume) = (300/112) =2.6786 g/cm^3

ρ_{exp} = Measured density

3.4 Microstructure Characterization

3.4.1 Sectioning of Test Specimens and Heat Treatment

3.4.1.1 Sectioning of Specimens for Testing

The edges of the sound section of the casting were discarded and the remaining portion sectioned as shown in Figure 3.4 for analysis of microstructure and mechanical properties of the alloys. A small portion from the top part of the casting marked microscopy was used to prepare specimens for microstructural examination while the rest of the top part was used to prepare tensile test specimens. The middle part of the casing was used for Brinell hardness tests and the bottom part for impact tests. As seen from the schematic representation of the casting, the left hand side was used in as cast condition while the right hand side was heat treated to a T6 condition whose parameters are described below.

PT6
H/TREATED

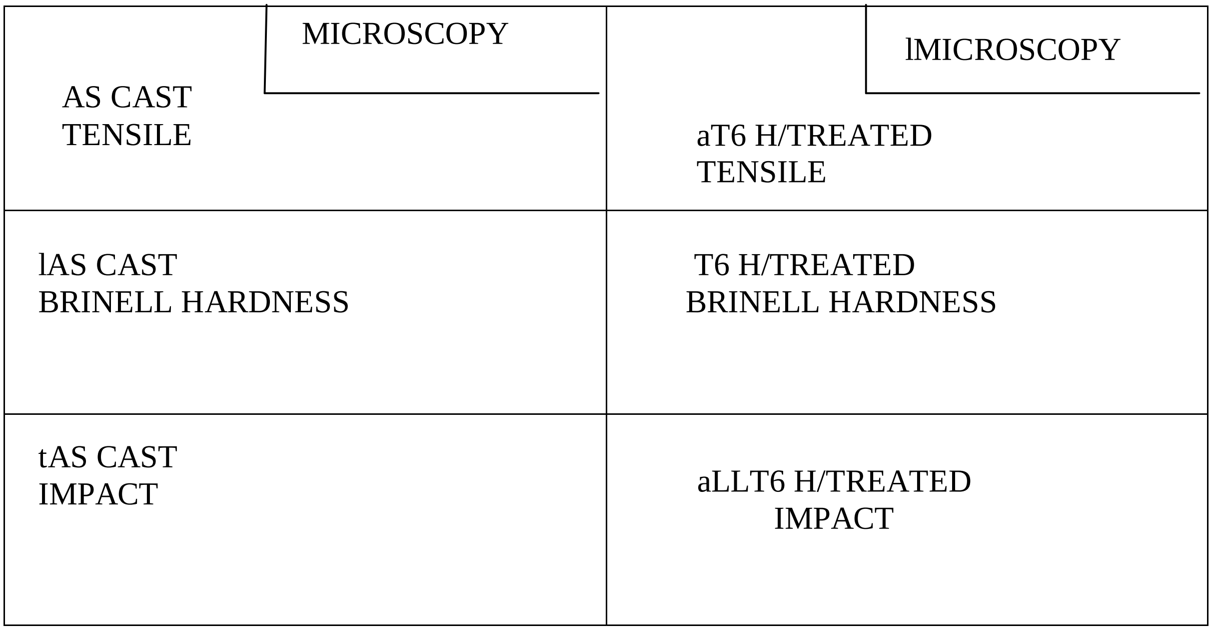


Figure 3.4: Schematic showing how the castings were sectioned for different tests

3.4.1.2 Heat Treatment Parameters

In the T6 heat treatment procedure, the alloys were subjected to solution heat treatment, quenching and artificial aging. Solution heat treatment involved heating the alloys to a temperature of 495˚C for 8 hours during which time maximum solute strengthening elements such as Cu and Mg were taken into solution to obtain a nearly homogeneous solid solution. Quenching was done in hot water at a temperature of about 80˚C before artificial aging was done for 8 hours at a temperature of 190˚C to achieve precipitation hardening necessary for improvement of properties.

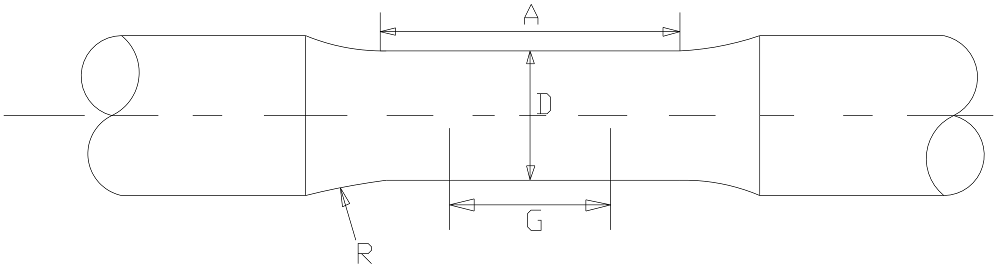
3.4.2 Preparation of Specimens for Microscopy

Specimens for microstructure analysis were polished down to the 600 grit SiC papers. A fine polish on grinding wheels using 0.25µm diamond paste was performed, and to aid in identification of the phases present in the microstructure, the samples were deep etched in a solution of 10% NaOH in distilled water and then washed using a soap-ethanol solution mixture. The samples were dried using forced air to avoid carrying over any contamination that could affect the results. This etching procedure was capable of giving several intermetallic phases a characteristic colour. The union ME-3295 optical microscope with an attached camera was used to obtain the micrographs.

3.5 Mechanical Testing

3.5.1 Tensile Testing

As-cast and T6 tensile test specimens were prepared according to ASTM B108-82b with a gauge diameter of 6.35 mm and gauge length of 25.4 mm. The Samuel Denison Tensile testing machine located at the Ministry of Public Works was used to pull the specimens at a strain rate of 1.5mm/min. Three specimens were used for each alloy giving three data points from which an average value was obtained. The dimensional details of the specimen used for tensile testing are given in Figure 3.5. After breaking the tensile test specimens, the two pieces were reassembled and the new gauge length obtained. The % elongation to fracture was measured as a measure of ductility.



G: Gage length: 25mm

D: Diameter: 6.35mm

R: Radius of fillet, min: 4.76mm

A: Length of reduced section, min: 31.75mm

Total length of specimen: 72mm

Figure 3.5: Dimensional details of Tensile specimen

3.5.2 Fractography

Flat tensile test specimens were prepared according to the ASTM E-8M standard with a gauge length of 25 mm and a rectangular cross section of 6 x 6 mm and the Samuel Denison Tensile Testing Machine located at the Ministry of Public Works was used to pull the specimens at a loading rate of 1.5mm/min. Upon fracture, the two surfaces were reassembled and mounted using bakelite resin and polished down to 600 grit size using SiC papers. A fine polish on grinding wheel using 0.25μm diamond paste was done. To aid identification of the phases present and responsible for fracture failure, the samples were deep etched in a solution of 10% NaOH in distilled water and then washed using a soap-ethanol solution mixture. The samples were dried using forced air to avoid carrying over any contamination that could affect the results. The union ME-3295 optical microscope with an attached camera was used to obtain the micrographs.

3.5.3 Impact Testing

Impact testing was performed on unnotched samples measuring 55mm x 10mm x 10mm in accordance to ASTM E-23. To remove any machining marks, the surfaces were polished with a fine SiC paper. The samples were tested using a conventional Torsee's charpy impact testing machine located at Jomo Kenyatta University of agriculture and Technology and whose characteristics were as follows; hammer weight 25.71kg, length of hammer arm 0.75m and hammer lift angle was 142.5˚.

3.5.4 Brinell Hardness Testing

The specimens used for hardness tests were polished using fine SiC paper so as to remove any machining marks and permit measurement of the diameters to ASTM specified accuracy of 0.05mm. Indentations were made using a 10 mm steel ball indenter with a load of 500kgf for a dwell time of 30 seconds. Each Brinell hardness value (BHN) obtained was an average of at least 3 readings taken from randomly distributed impressions made on each test surface of the specimen. The Torsee's Brinell Hardness Testing Machine that was used for the hardness measurements is located at Jomo Kenyatta University of Agriculture and Technology

3.6 The Survey

Three major towns; Nairobi, Mombasa and Nakuru were chosen for the survey because of the cluster of the foundry enterprises in these towns. In addition, the sugar belts were also visited. A total of 45 foundries were visited in which questionnaires and in-depth interviews with the foundrymen were held to elicit information about the prevailing Al practices such as the recycling methods, casting design, melt treatment and quality control aspects. The information obtained from the survey was used to establish whether these practices are effective in making premium quality castings.

CHAPTER FOUR

4.0 EFFECT OF MINOR ELEMENTS ON CASTABILITY, MICROSTRUCTURE AND MECHANICAL PROPERTIES OF RECYCLED ALUMINIUM ALLOYS: RESULTS AND DISCUSSION

4.1 Fluidity – An Aspect of Castability

4.1.1 Results

Tables 4.1 and 4.2 show the fluidity test results for LM25 and LM27 alloys respectively. Both tables indicate that addition of 0.015%Sr and 0.02%Sr increases the fluidity of LM25 and LM27 by 9% and 21% respectively. Furthermore, a 0.28% Al-5Ti-1B grain refiner addition decreases the fluidity of LM 25 and LM27 by 2% and 19% respectively. Table 4.1 shows that no change in fluidity occurs when the Fe level is increased from 0.14 to 0.4% in LM25, but it increases by 21% when Fe is raised to the critical content of 0.48%. Furthermore, a decrease of 32% results when the Fe level is increased to 0.6%. A combination of 0.3%Mn or 0.6%Cr with 0.6%Fe in LM25, results in a fluidity increase of 13% and 8%, respectively compared to the base alloy, but a combination of 0.6%Fe, 0.3%Mn and 0.2%Cr decreases the fluidity by 9%. On the other hand, Table 4.2 shows that a 34% increase in fluidity occurs when the Fe content in LM27 is raised from 0.41% to the critical level of 0.6%Fe with a further increase when Mn is raised to 0.3%Mn. Increasing the Fe content to 1% in LM27 leads to a drop in fluidity of 9%. A combined addition of 0.02%Sr and

0.28%Al-5Ti-1B decreases the fluidity of LM27 by 8%. The average trends in fluidity with the various levels of the elements tested are shown in Figures 4.1 and 4.2 for LM25 and LM27 respectively. The error bars indicate the scatter in the results.

Table 4.1: Fluidity test results for the LM25 alloy.

LM25 base alloy + any addition	Measured lengths of flow of molten metal (mm)			Std deviation of the sample s (mm)	Resulting effect on fluidity*
	(i)	(ii)	Average		
LM25	312	325	319	7	-
LM25 + 0.015%Sr	340	353	347	7	9% Increase
LM25 + 0.28%Al-5Ti-1B	300	325	313	13	2% Decrease
LM25 with 0.4% Fe	315	323	319	4	No change
LM25 with 0.48%Fe	362	408	385	23	21% Increase
LM25 with 0.6%Fe	198	237	217	20	32% Decrease
LM25 with 0.6%Fe + 0.3%Mn	355	365	360	5	13% Increase
LM25 with 0.6%Fe + 0.6%Cr	340	350	345	5	8% Increase
LM25 with 0.6%Fe + 0.3%Mn + 0.2%Cr	285	293	289	4	9% Decrease

*Percentage increase or decrease is based on average length of flow of molten metal for LM25

Table 4.2: Fluidity test results for the LM27 alloy.

LM27 base alloy + any additions	Measured lengths of flow of molten metal (mm)			Std deviation of the sample s (mm)	Resulting effect on fluidity*
	(i)	(ii)	Average		
LM27	355	360	358	3	-
LM27 + 0.02%Sr	412	455	434	22	21% Increase
LM27 + 0.28%Al-5Ti-1B	280	297	289	9	19% Decrease
LM27 + 0.02%Sr +0.28%Al-5Ti-1B	310	350	330	20	8% Decrease
LM27 with 0.6%Fe	438	520	479	41	34% Increase
LM27 with 1.01%Fe	320	330	325	5	9% Decrease
LM27 with 0.6%Fe +0.3%Mn	545	560	553	8	54% Increase

*Percentage increase or decrease is based on average length of flow of molten metal for LM27

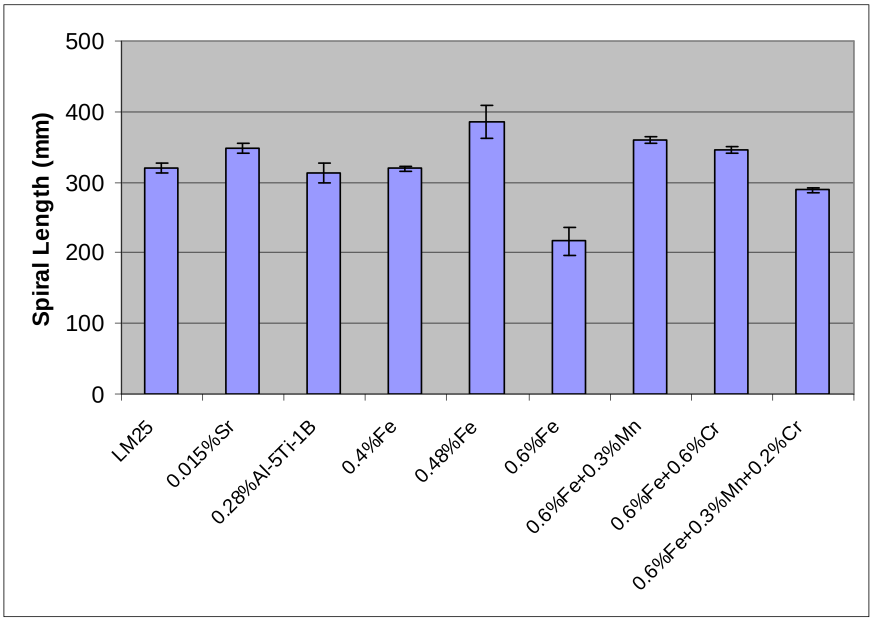


Figure 4.1: Fluidity trends in LM25 with various levels of the elements studied.

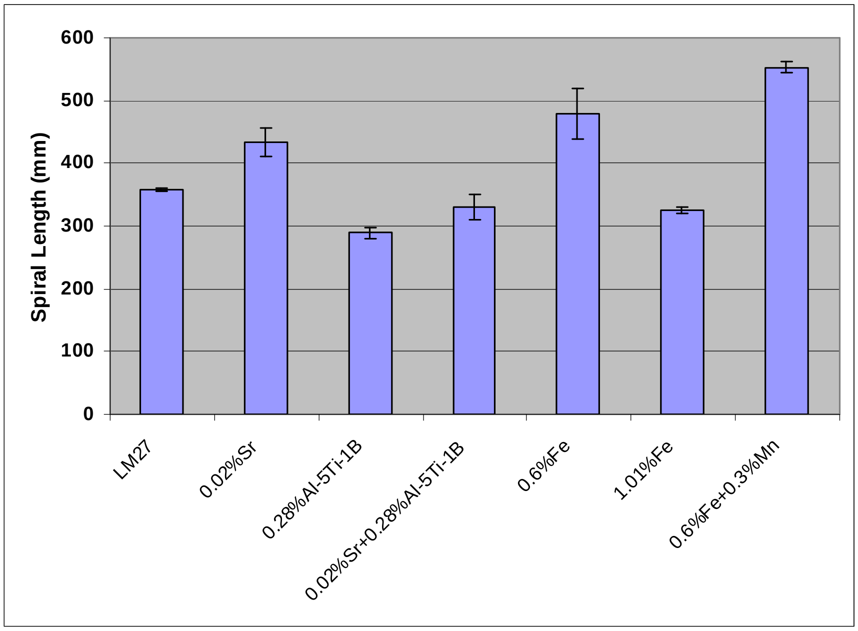


Figure 4.2: Fluidity trends in LM27 with various levels of the elements studied

4.1.2 Discussion

Tables 4.1 and 4.2 show results of the running lengths of the alloys tested with different element addition rates. It is evident that minor changes in chemical composition affected the fluidity of the alloys investigated. Whereas some elements enhance the fluidity of the alloys, others indicate a decrease.

The results of this investigation show that the effect of Fe on fluidity is not straight forward. It is generally expected that increasing the Fe level should lead to a continued reduction in fluidity. However, it is observed that increasing the Fe level from 0.14 to 0.4% in LM25 did not change the fluidity while raising it to 0.48% led to a significant increase in fluidity. This Fe level was chosen to coincide with the critical Fe content for the alloy at which a shift in the solidification sequence of second phases occurs. A much higher level of Fe of 0.6%, which is above the critical Fe level, resulted in a significant reduction in fluidity. The same trend is seen with LM27 alloy in which the fluidity increased significantly when Fe was increased from 0.41 to 0.6%, the critical level for the alloy. Higher Fe levels resulted in a significant drop in fluidity. Taylor et al. (1999a-c) first proposed that the castability of Al-Si alloys can be significantly influenced by the solidification sequence of Fe-bearing phases. They observed that porosity was at a minimum at the critical Fe content for each of the alloys investigated and attributed this behavior based on changes in interdendritic permeability caused by different solidification sequences. In Al-Si alloys, the phase diagram predicts three basic solidification sequences depending on the Fe level:

$$<Fe_{cr}: Al \rightarrow Al\text{-}Si \rightarrow Al\text{-}Si\text{-}Al_5FeSi \qquad (4.1)$$

$$=Fe_{cr}: Al \rightarrow Al\text{-}Si\text{-}Al_5FeSi \qquad (4.2)$$

$$>Fe_{cr}: Al \rightarrow Al\text{-}Al_5FeSi \rightarrow Al\text{-}Si\text{-}Al_5FeSi \qquad (4.3)$$

Where, $Fe_{cr,}$ is the critical Fe content, which results in the solidification sequence proceeding from primary Al directly to Al-Si-Al_5FeSi ternary eutectic. Many small β-Al_5FeSi platelets form at Fe_{cr} allowing the nucleation of many small eutectic grains that result in the most permeable developing interdendritic structure. Below the critical Fe content, the eutectic Si nucleates prior to the formation of β-Al_5FeSi because the Al-Si eutectic occurs before the Al-Si-Al_5FeSi ternary eutectic. This results in large Al-Si grains that work synergistically with β-Al_5FeSi phases to physically block fluid flow. Above the critical level, the β-Al_5FeSi phases form first and later nucleate Al-Si eutectic grains forming a complex network of large second phases that also obstruct fluid flow. However, Dinnis et al. (2006) reported recently that increasing Fe leads to β-Al_5FeSi formation that poisons the nucleation of eutectic grains that lead to large eutectic grains and hence increased obstruction.

This philosophy can easily be extended to explain the fluidity results observed in this work. It is proposed in this study that at the critical Fe content, the solidifying mass of aluminum dendrites and second phases will delay in interlocking to form a coherent network (delayed coherency) and thus allow more time for fluid flow and hence higher fluidity. An Fe level above or below the critical content will lead to early network coherency and hence reduced fluidity. The fluidity is expected to reduce even more at much higher Fe levels because primary Fe-phases are likely to form resulting in an even earlier premature coherency of solidifying crystals. Dinnis et al. (2006) reported that addition of Mn to Fe containing alloys inhibits the poisoning effect of β-Al_5FeSi phases on the Al-Si eutectic nucleation. The result is many small eutectic grains and a possible delay in the formation of coherent network and hence a subsequent increase in fluidity as observed in this work. It is possible that Cr works in same manner, explaining the increase in fluidity with Cr addition to Fe containing alloys. However, experimental evidence is required to confirm such a hypothesis. The reduction in fluidity with a combined addition of Mn and Cr is likely to be due to formation of large primary Fe-bearing phases (sludge) that may have negated any beneficial effect of individual additions.

Dahle et al. (1997) have shown that Sr addition to Al-7Si-Mg alloy delays coherency to higher solid fractions. It also displaces the eutectic composition to higher Si contents and reduces the eutectic temperature, thereby increasing the solidification range. Furthermore, Sr has been reported [Emadi et al. (1993)] to reduce the surface tension of molten A356 by about 19%. It is generally believed that these factors lead to increases in fluidity and thus could explain the increase in fluidity with Sr addition in this work as observed in Tables 4.1 and 4.2 and Figures 4.1 and 4.2. This is line with observations by Villeneuve et al. (2001).

The drop in fluidity with grain refinement is in line with reports by Dahle and Arnberg (1996) and Mollard et al (1987). This however contradicts the general view that grain refinement delays dendrite coherency and is thus expected to increase fluidity as reported by others [Lang (1972); Kwon and Lee (2003)]. This seems to reinforce the suggestion by Gruzleski and Closset (1990) that early nucleation caused by grain refinement results in a slurry flow from the moment of pouring and that since slurries flow with more difficulty than simple liquids, the fluidity should be reduced with grain refinement. Furthermore, Di Sabatino and Arnberg (2005) have also reported that fine particles are more effective in stopping a flowing stream than equivalent coarse particles. Similar reductions in the fluidity of Al-Si casting alloys with addition of grain refiners containing Ti have been reported by Loper (1992) and Loper and Prucha (1990). The slight reduction in fluidity observed after a combined addition of Sr and grain refiner can be attributed to a balance between the competing influences of each addition.

4.2 Porosity – An Aspect of Castability

4.2.1 Results

The influence of individual and combined additions of Sb, Mn, Cr, Sr and Al-5Ti-1B grain refiner on porosity formation of a high Fe secondary LM13-type piston alloy was investigated and the results are recorded in Table 4.3. It was observed that the volume percent porosity in the base alloy was 1.6%. With addition of Sr at levels of 0.02% and 0.05%, the volume percent porosity increased to 2.9% and 2.6% respectively. On the other hand, when additions of 0.53%Mn and 1.06%Cr were made to the alloy, the volume percent porosity reduced to 0.6% and 0.4% respectively. It was further noted that a combined addition of 0.3%Mn together with 0.2%Cr reduced the volume percent porosity of this alloy to 1.0% while addition of 0.2%Sb decreased it to 1.2%. Other reductions in porosity of 1.1%, 1.2% and 0.8% were observed when 0.53%Mn was added in combination with 0.02%Sr, 0.05%Sr and 0.2%Sb respectively. On the other hand, addition of 0.02%Sr together with 0.28%Al-5Ti-1B grain refiner slightly reduced the volume percent porosity of the base alloy to 1.4%. The trends in volume percent porosity with the various levels of the elements tested are shown in Figure 4.3 for the LM13-type alloy.

Table 4.3: Volume percent porosity for an LM13-type alloy with element additions

Sample	Mass (g)	Volume (cm^3)	P_{exp}=m⁄v (g/cm^3)	Porosity (vol pct)*
LM13 + None	658.93	250	2.6357	1.6%
LM13 + 0.02%Sr	665.64	256	2.6002	2.9
LM13 + 0.05%Sr	655.15	251	2.6102	2.6
LM13 + 0.2%Sb	661.57	250	2.6463	1.2
LM13 + 0.53%Mn	652.50	245	2.6633	0.6
LM13 + 1.06%Cr	656.08	246	2.6669	0.4
LM13 + 0.3%Mn + 0.2%Cr	663.03	250	2.6521	1.0
LM13 + 0.02%Sr + 0.53%Mn	659.88	249	2.6501	1.1
LM13 + 0.05%Sr + 0.53%Mn	658.71	249	2.6454	1.2
LM13 + 0.2%Sb + 0.53%Mn	648.68	244	2.6585	0.8
LM13 + 0.02%Sr + 0.28% Al-5Ti-1B	660.45	250	2.6418	1.4

*The volume percent porosity on last column of table was calculated based on equation 3.2

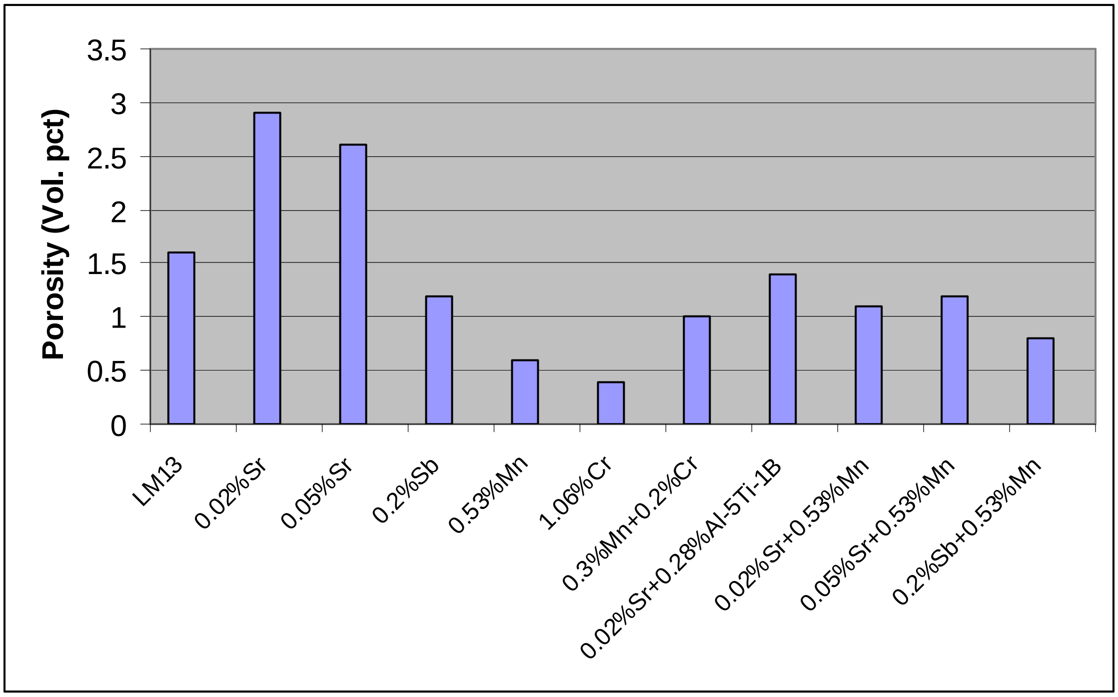


Figure 4.3: Volume percent porosity trends in LM13-type alloy with various levels of elements studied as calculated from density measurements

4.2.2 Discussion

It was observed from the results in Table 4.3 and Figure 4.3 that the high Fe level in the LM13 base alloy led to formation of high porosity levels of 1.6%. The reason behind the high level of porosity level in this alloy can be explained based on long flake-like phases observed in the Figure 4.4(a). These phases intersect and reduce the size of interdendritic feeding paths. As the interdendritic flow channels became narrow, feeding to compensate for solidification shrinkage became more difficult and this made it easier for porosity to form as recorded in Figure 4.3. Several authors have made similar observation. Wang et al. (2003) for instance observed that increasing the

Fe level from 0.13% to 0.47% in A356 alloys increases porosity from 0.4% to 1.5%. Furthermore, Samuel et al. (2001) noted that as the Fe content increased from 0.37% to 1.4% in 319-type alloys the percentage porosity also increased. The explanation for increased porosity due to Fe addition has been given by these authors based on the formation β-phases that intersect and reduce the size of interdendritic feeding paths. This limits feeding and increases the driving force for pores to form.

Addition of Sr at levels of 0.02% and 0.05% to the LM13 base alloy leads to a further increase in porosity. Despite the great improvements in UTS and ductility as discussed elsewhere in this report, porosity formation becomes the greatest undoing for this beneficial effect.

It was noted that Sr addition to the LM13 alloy affected the evolution of microstructure during solidification by changing the size and morphology of the phases (Figures 4.5 and 4.6), the freezing range and nucleation temperature of phases. In this manner the term, $P_{s,}$ in equation (2.1), was affected making feeding to compensate for volumetric contraction during solidification more difficult. To be noted also is the role that Sr addition plays in decreasing the surface tension of the melt which reduces the energy barrier for pore formation and hence high porosity levels. Dinnis et al. (2004b) reported that Sr addition causes strontium oxides to occur easily and this accounts for porosity formation in Sr modified alloys. Argo and Gruzleski (1988) also noted that interdendritic feeding becomes more difficult as a result of modified alloys having a longer mushy zone due to the depression of the eutectic temperature. It is therefore apparent in the context of the present study that changes in surface tension, freezing range, nucleation potency and Sr oxide formation worked synergistically to lead to higher levels of porosity in the alloys that are Sr modified.

In addition, differences in eutectic solidification modes could have taken centre stage in controlling evolution and permeability characteristics of the solidifying microstructure and hence porosity formation. Dinnis et al. (2004b) reported that different eutectic nucleation and growth modes have different effects on feeding and porosity formation characteristics because the eutectic solid is distributed differently throughout the dendritic network. In Sr modified alloys the eutectic grains nucleate and grow independently of the primary Al dendrites with a relatively smooth solidification front. The independent nucleation of the eutectic grains leads to a decrease in permeability and feeding efficiency because the eutectic grains obstruct the larger preferred feed paths. The independent nucleation of the eutectic grains may have another effect on porosity through increased buildup of hydrogen at the growth front. For these reasons Sr modified alloys demonstrated high porosity levels in the present study.

The results also show that addition of Mn and Cr either individually or in combination leads to significant reduction in porosity formation. Dinnis et al (2004a) observed a decrease in porosity in Al-9Si-0.5Mg-3Cu alloy containing 1.0%Fe upon addition of 0.5%Mn and explained this observation on the basis of increase in the number of eutectic nucleation events. They indicated that Mn addition to the alloy increased the number of eutectic nucleation events in Fe containing alloys which in turn influence the size and distribution of the eutectic and intermetallic phases and more importantly the permeability of the solidifying structure and the level of porosity. The observed decrease in porosity levels in the present study as a result of Mn and Cr additions can be explained based on the formation of a permeable structure as a result of decreased grain size of the eutectic as the number of nucleation events for the eutectic is increased. The combined addition of Sr and Mn indicate that Mn tends to reduce the level of porosity, but the presence of Sr masks this reduction as it makes the solidifying structure less permeable as earlier explained. This is why from the results it observed that the decrease in porosity was not significant when these additions were combined. The same rationale can be used to explain results due to combined addition of Sr and Al-5Ti-1B.

4.3 Microstructure

4.3.1 Results

Results from the effect of individual and combined additions of Sb, Mn, Cr, Sr and Al-5Ti-1B grain refiner on the microstructure of a high Fe secondary LM13 type piston alloy indicate that alloying elements impart significant influences on the microstructural features of the alloy. Without additions, the LM13 base alloy consists mainly of a structure with coarse Si particles, large Cu and Ni containing phases in addition to Al_2Cu phases. When 0.02%Sr or 0.05%Sr is added to the LM13 alloy, the acicular Si particles are modified to a fibrous form but the level of modification can either be partial or full depending on the amount of Sr. Furthermore, a 0.2%Sb addition causes refinement of the Si particles to a lamellar structure. When individual additions of 0.53%Mn and 1.06%Cr were made to the alloy, the large flake-like Cu and Ni bearing phases changed to Chinese script phases. A similar observation was made when 0.3%Mn was combined with 0.2%Cr. Moreover, when 0.53%Mn was added in combination with 0.02%Sr, 0.05%Sr and 0.2%Sb changes occurred to both the Si particles as well as the AlCuNi phases. With addition of 0.02%Sr and 0.28%Al-5Ti-1B the morphology of Si particles changed to fibrous form. It is interesting to note that after T6 heat treatment the Si particles are seen to spheroidize while at the same time AlCuNi phases fragment and the Al_2Cu phases dissolve. These changes in turn improve the mechanical performance of the alloy.

4.3.2 Discussion

The optical micrographs of an LM13-type base alloy in as cast condition and T6 heat treated condition are shown in Figure 4.4a and 4.4b. It is observed that the as cast microstructure of the base alloy consisted of coarse acicular Si particles and large Cu and Ni containing phases with complex morphology on the α-Al matrix. In addition, a network of small interconnected particles identified as Al_2Cu phases were observed in the microstructure. Moffat (2007) had noted that Fe-containing phases like AlFeNi and AlFeCu may be present in this type of alloy; but their influence on mechanical properties is not yet established. The presence of the coarse acicular Si particles in addition to the large Cu and Ni containing phases played a significant role in deterioration of mechanical properties of the alloy as will be discussed later in section 4.4. After T6 heat treatment (Figure 4.4b) some spheroidization of Si particles was observed. Furthermore, the Cu and Ni containing phases fragmented into smaller intermetallics. Moreover, the Al_2Cu phases seemed to have dissolved when a heat treatment schedule was performed. The changes brought by heat treatment on the microstructure were noted to contribute significantly to the improvement of mechanical properties as will be seen later.

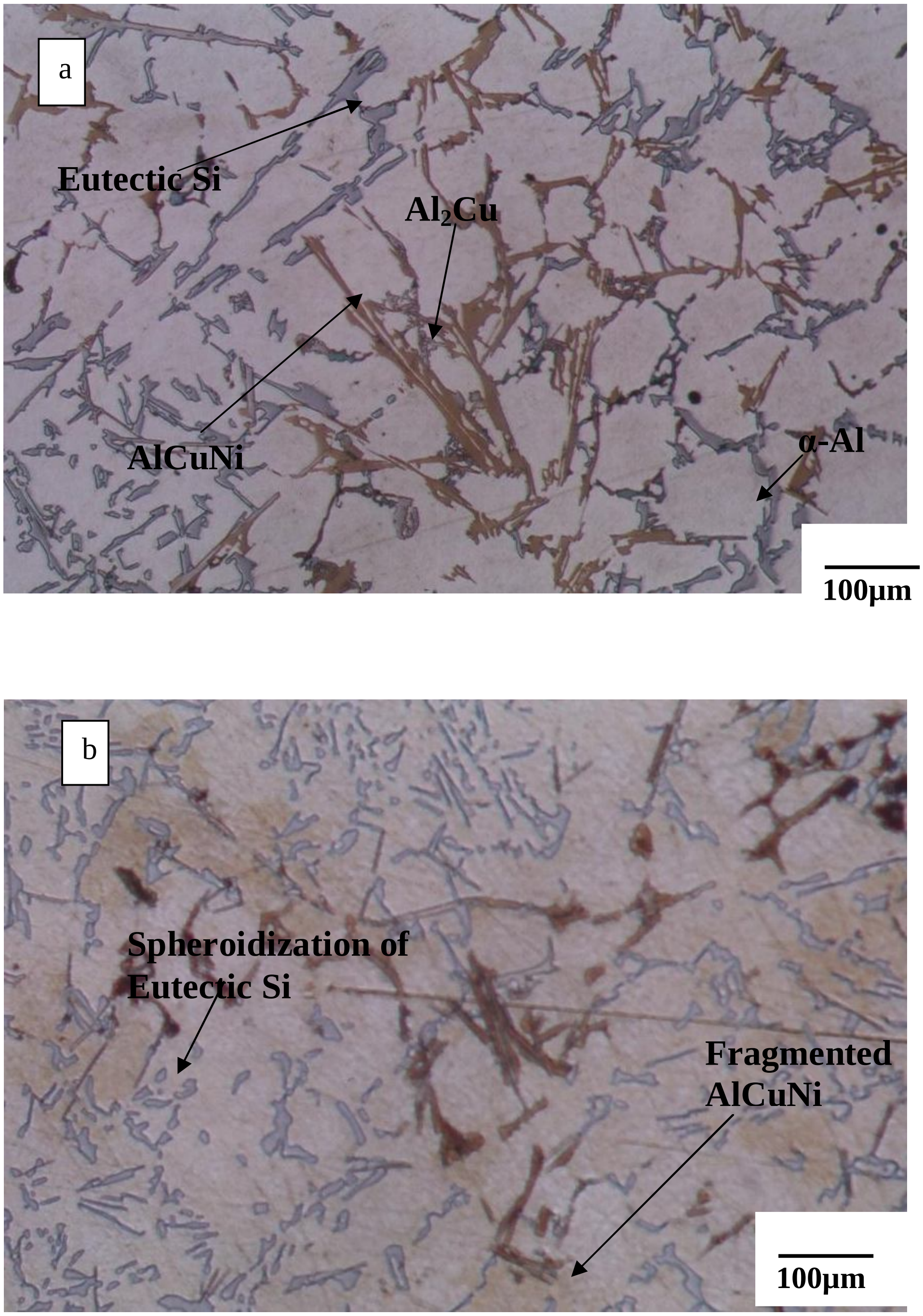


Figure 4.4: (a) As cast microstructure of LM13 base alloy showing large acicular Si particles and large Cu and Ni containing phases with complex morphology (b) T6 version of (a) showing some spheroidization of Si particles as well as fewer and smaller intermetallics

Addition of Sr at levels of 0.02% and 0.05% to the LM13 alloy in Figures 4.5 and 4.6 is seen to substantially reduce the size of the eutectic Si particles in addition to changing the morphology from acicular to fibrous form. It is however noted that the level of modification due to addition of 0.02%Sr is not the same as that due to addition of 0.05%Sr. With a Sr level of 0.02% only partial modification of the eutectic Si occurs as evidenced by some sections of the sample in Figure 4.5a having a coarse eutectic Si with acicular morphology. Furthermore, Figure 4.11a provides further evidence that addition of 0.02% Sr with 0.53%Mn caused partial modification with some eutectic Si phases remaining coarse and acicular. However, with a Sr level of 0.05%Sr full modification of the eutectic was obtained (Figure 4.6). Reports show that addition of higher levels of Sr can also be beneficial in causing fragmentation of the Fe-phases that may be present in the microstructure. Samuel et al. (1996) for instance indicated that addition of 300ppm Sr accelerated dissolution of the β phases through fragmentation of the long intercepting needles. In the present study, addition of 0.05%Sr to the alloy is noted to have caused fragmentation of long flake-like structures as demonstrated in Figure 4.6(a).

When T6 heat treatment was done on the Sr-modified specimens there was a significant effect on the size, shape and distribution of the modified eutectic Si particles. A comparison between the as cast and heat treated modified samples indicates that the eutectic Si particles undergoes fragmentation and spheroidization in addition to coarsening (Figure 4.5b). Furthermore, an increase in inter particle spacing of the eutectic Si particles is seen to occur (Figure 4.6b). The changes in microstructural features noted to occur due to Sr-modification and heat treatment may have accounted for the improvement of properties (percent elongation and UTS).

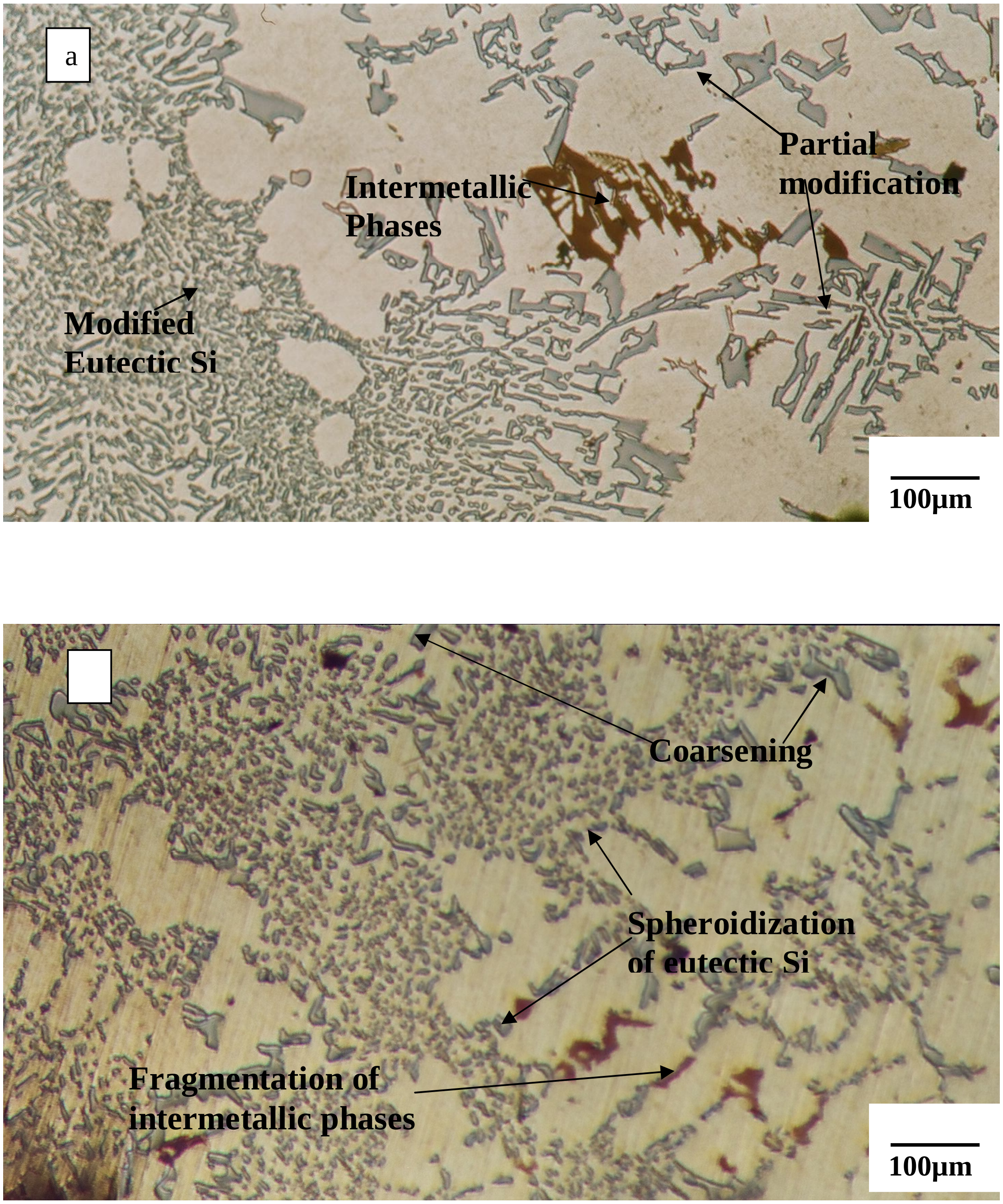


Figure 4.5: **(a) As cast microstructure of 0.02%Sr modified LM13 alloy showing modified and partially modified Si particles (b) T6 version of (a) showing some spheroidization and coarsening of Si particles and fragmentation of intermetallic phases**

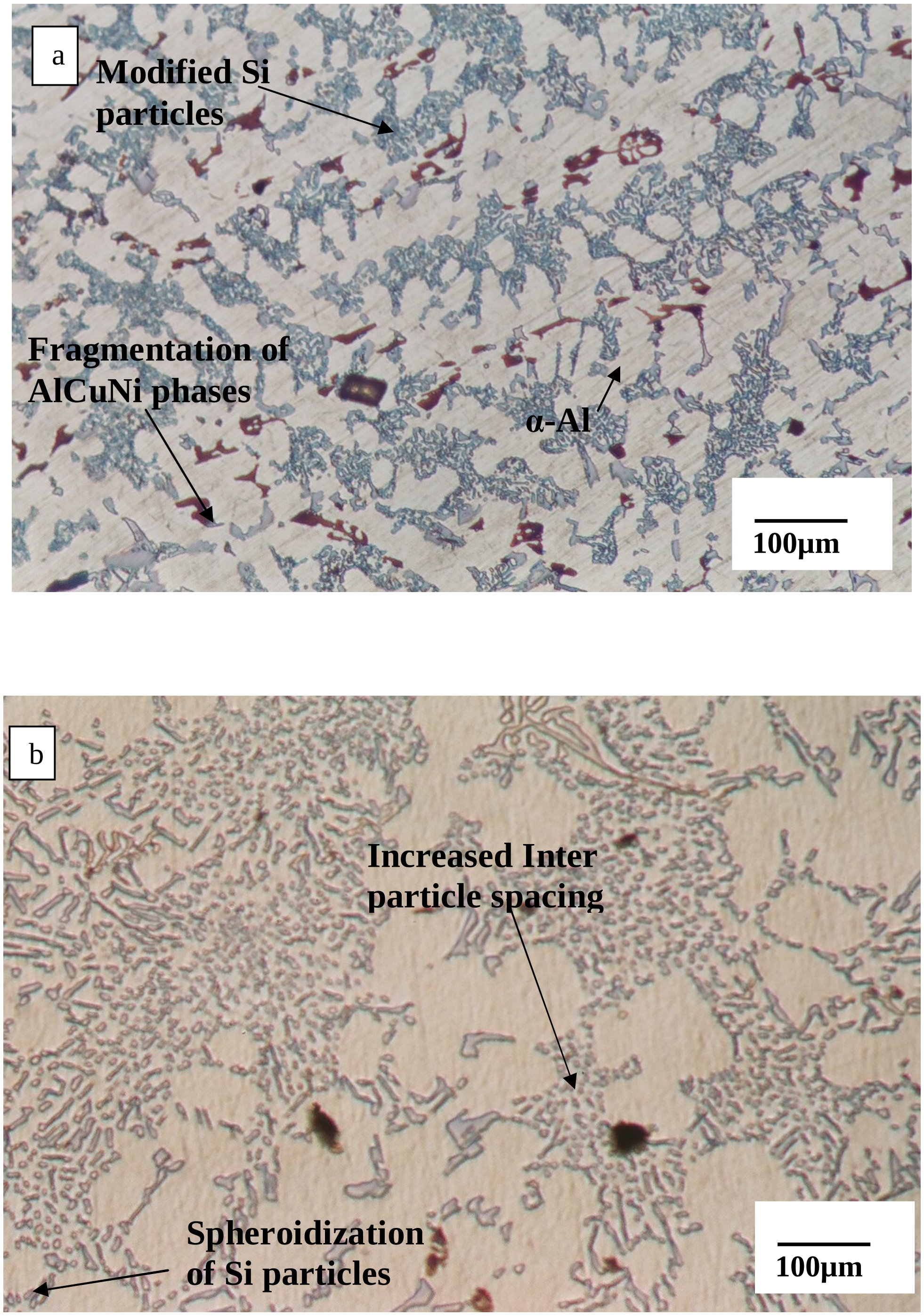


Figure 4.6: (a) As cast microstructure of LM13 alloy with 0.05%Sr showing modified Si particles with fragmented Cu and Ni containing phases (b) T6 version of (a) showing some spheroidization and coarsening of Si particles and increased inter particle spacing

It is noted from Figures 4.7a and b that addition of 0.2%Sb does not change the eutectic Si to fibrous form but rather causes a refinement of the eutectic Si. This observation agrees with that of Caceres et al (1997) who noted that addition of Sb to Al-Si alloys reduces the size of the eutectic flakes with improved results in mechanical properties. Furthermore, Jenabali et al. (2004) noted that addition of 0.1% Sb to A356 alloy changes the unmodified acicular morphology to a lamellar form. Upon heat treatment, there is notable spheroidization of the Si particles in addition to fragmentation of the intermetallic phases. This in turn affects properties as will be later discussed.

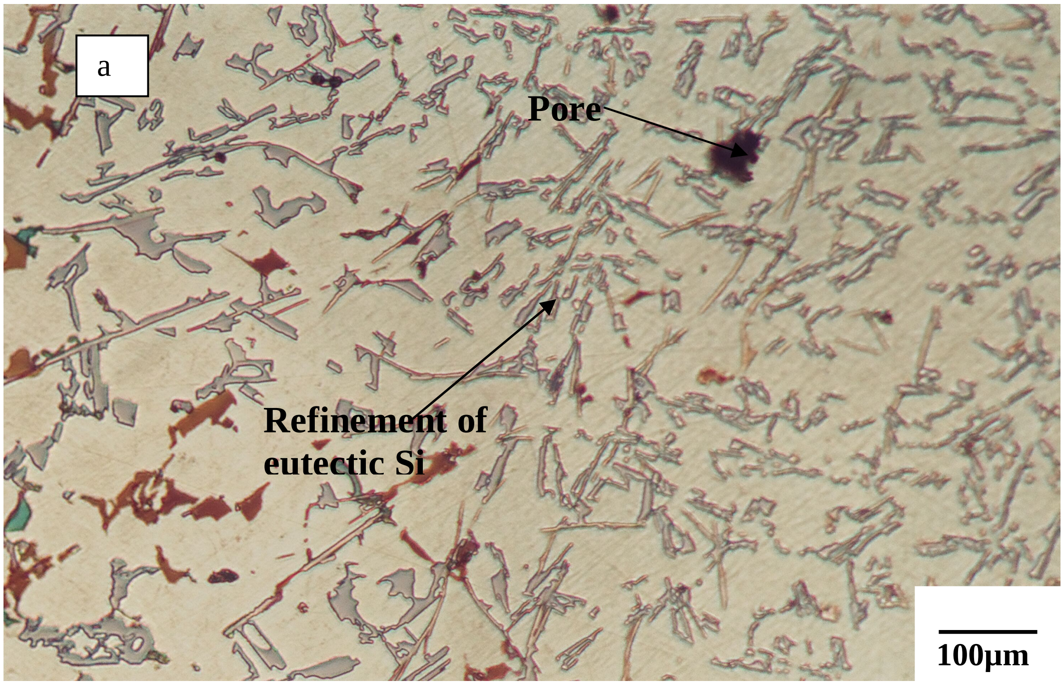


Figure 4.7: (a) As cast microstructure of LM13 alloy with 0.2%Sb showing refined Si particles and a pore

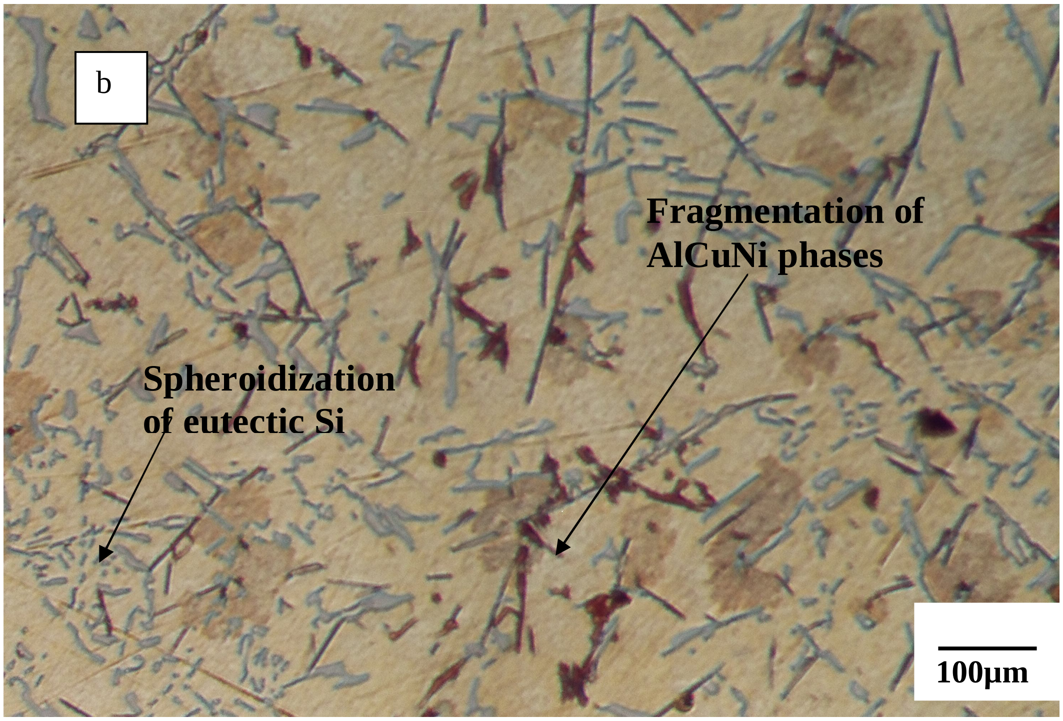
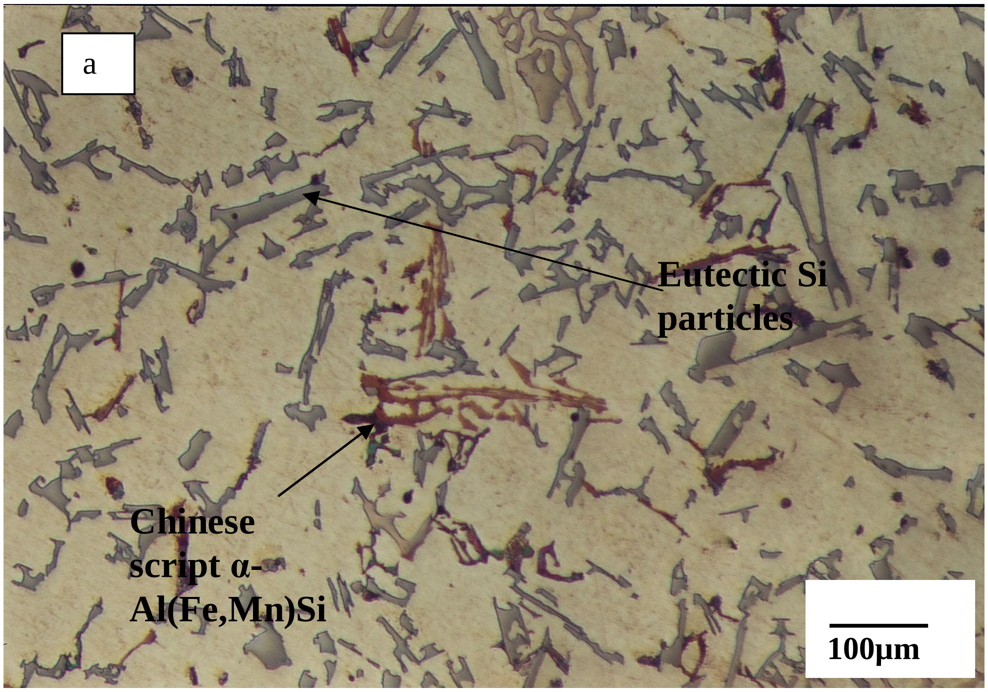


Figure 4.7: (b) T6 version of (a) showing some spheroidization of Si particles as well as fragmentation of intermetallics

According to the manufacturing practice, it is established that casting products from Al piston alloys suffer low mechanical properties attributable to the coarse microstructural features present in the alloy. The presence of the long flake-like structures identified as AlCuNi by Moffat (2007) in the microstructure of the piston alloy account for deteriorated mechanical properties as will be discussed in section 4.4. Efforts are made to mitigate these negative influences through alloying elements. The present study indicated that addition of Mn and Cr to the LM13 piston alloy led to transformation of the AlCuNi phases in Figure 4.4a into Chinese script phases (Figures 4.8 and 4.9). Cho et al. (2006) noted that addition of Mn to piston alloys changed the morphology and type of intermetallic phases from their long flake-like form to Chinese script morphology. The authors identified the Chinese script intermetallic phases as α-Al(Mn,Fe)Si when up to 0.5%Mn was added. Cr addition to the piston alloy was also observed to lead to formation of compounds which were identified as AlCrSi [Cho et al (2006)]. Mulazimoglu et al. (1996) noted that formation of phases with Chinese script morphology confer beneficial effects to properties based on their compact shape. (2006). A combined addition of Mn and Cr (Figure 4.10) may have led to a mixture of these two phases but experimental identification of these phases will be necessary to clarify this claim.

A Mn level of 0.53% was chosen to coincide with the Mn/Fe ratio of 0.5 which has been shown to cause complete neutralization effect on Fe containing Al-Si alloys [Couture (1981)]. This ratio however, did not lead to complete transformation of the flake-like structures as some still remained unchanged as shown in Figure 4.8b. It is therefore suggested that this Mn/Fe ratio of 0.5 recommended for complete neutralization of Fe effects does not necessarily lead to complete transformation of the AlCuNi phases from flake-like morphology to Chinese script morphology. Furthermore, it should be noted that because of the high levels of Ni and Cu in Al piston alloys the beta phases did not form and therefore Mn and Cr additions did not result in conventional neutralization. Hence use of the ratio of Mn/Fe of 0.5 in breaking up the entire flake-like structure to Chinese script morphology was inadequate. Noting further that even the amount of Mn needed for conventional neutralization of Fe is not well established, it may be recommended that more experimental studies be done to shed light to the optimum Mn/Fe necessary for this purpose.

For Cr additions, a ratio of Fe/Cr of 1.0 was applied in the present study and it was found to be more effective in transformation of the AlCuNi phases from the flake like morphology to Chinese script morphology (Figure 4.9). It was noted further that the effect of T6 heat treatment procedure on alloys with Mn and Cr additions was to spheroidize, coarsen and increase the inter particle spacing of the eutectic Si particles (Figure 4.10b). In addition, the T6 schedule served to fragment the intermetallic phases (Figure 4.9b). Moreover, dissolution of the Al_2Cu phases in Figure 4.9a took place as a result of the T6 heat treatment schedule as noted in Figure 4.9b.



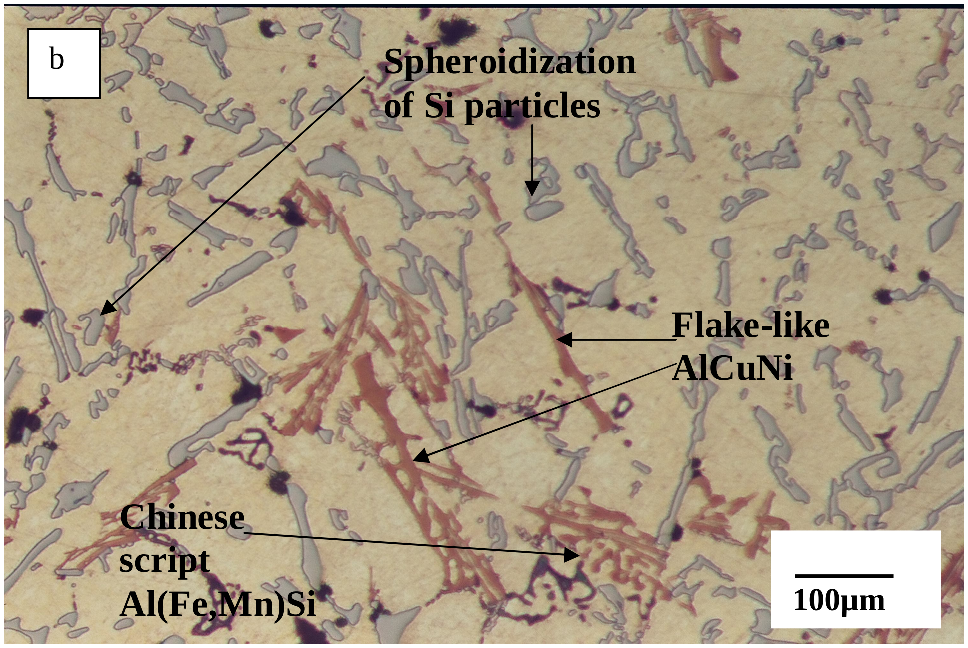


Figure 4.8: (a) As cast microstructure of LM13 alloy with 0.53%Mn showing large acicular Si particles and Chinese script Al(Fe,Mn)Si (b) T6 version of (a) showing some spheroidization of Si particles as well as large Cu and Ni containing phases

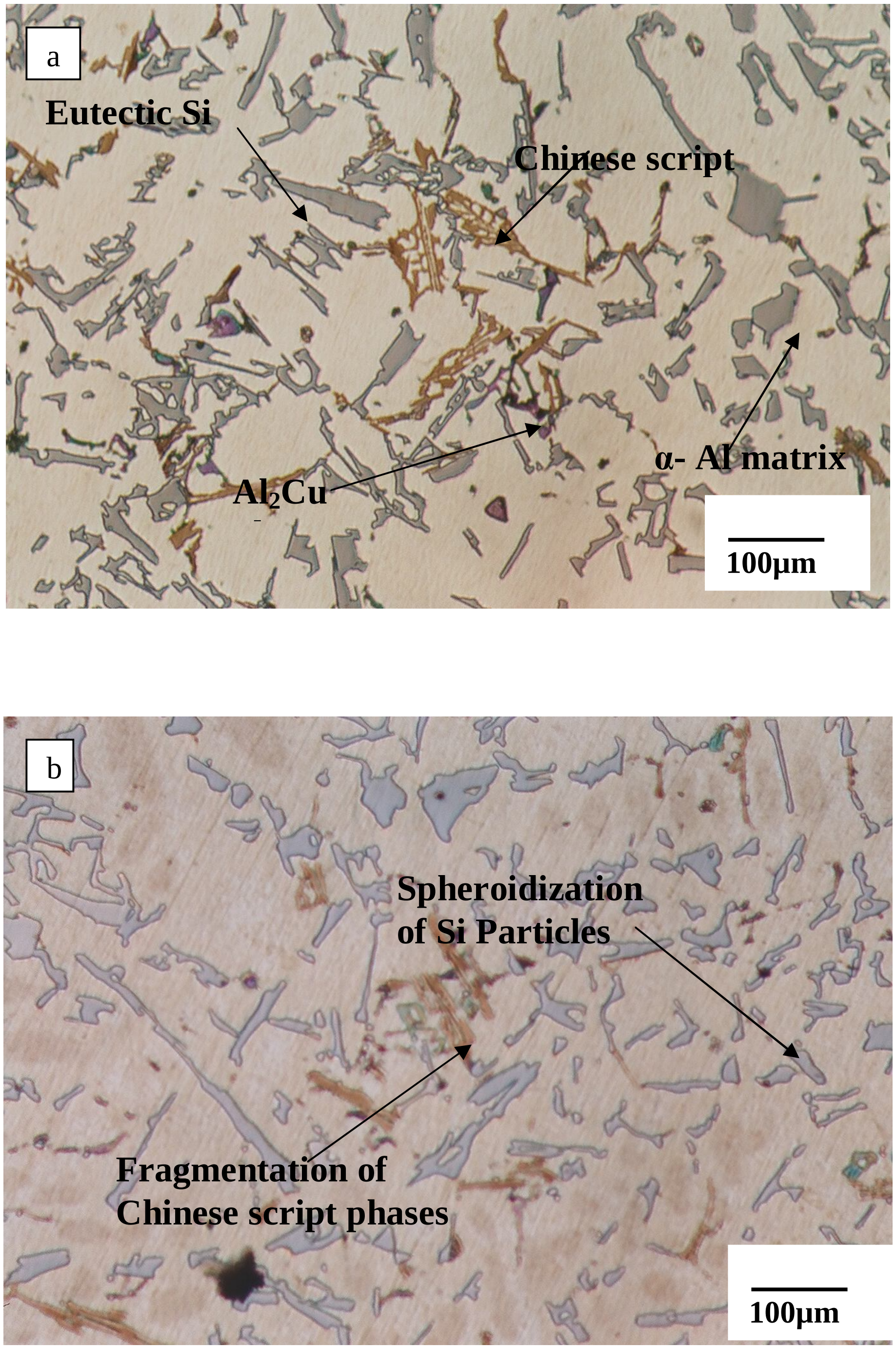
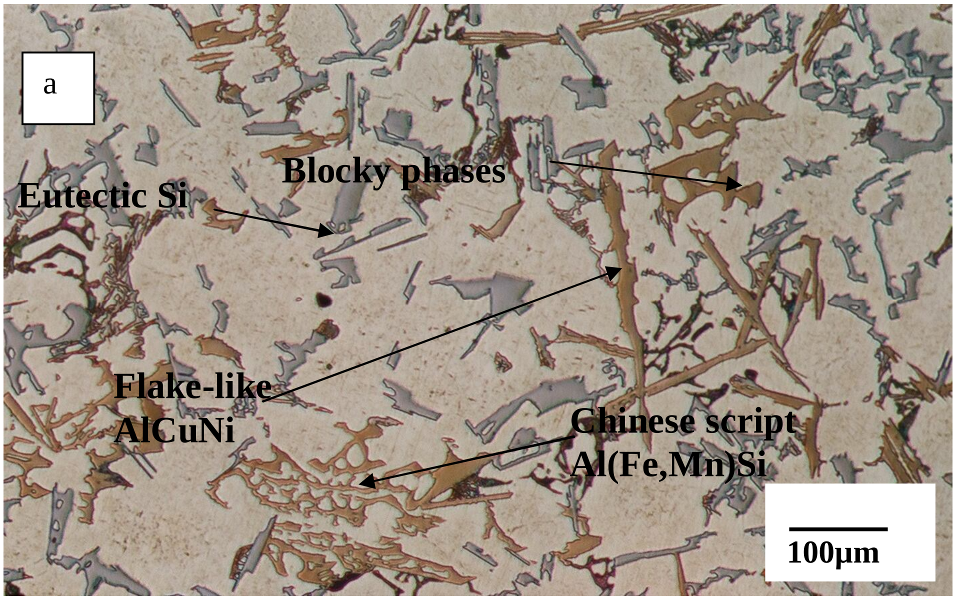


Figure 4.9: (a) As cast microstructure of LM13 alloy with 1.06%Cr showing large acicular Si particles, Al_2Cu phases and Chinese script phases (b) T6 version of (a) showing some spheroidization of Si particles as well as fragmentation and dissolution of some intermetallic phases



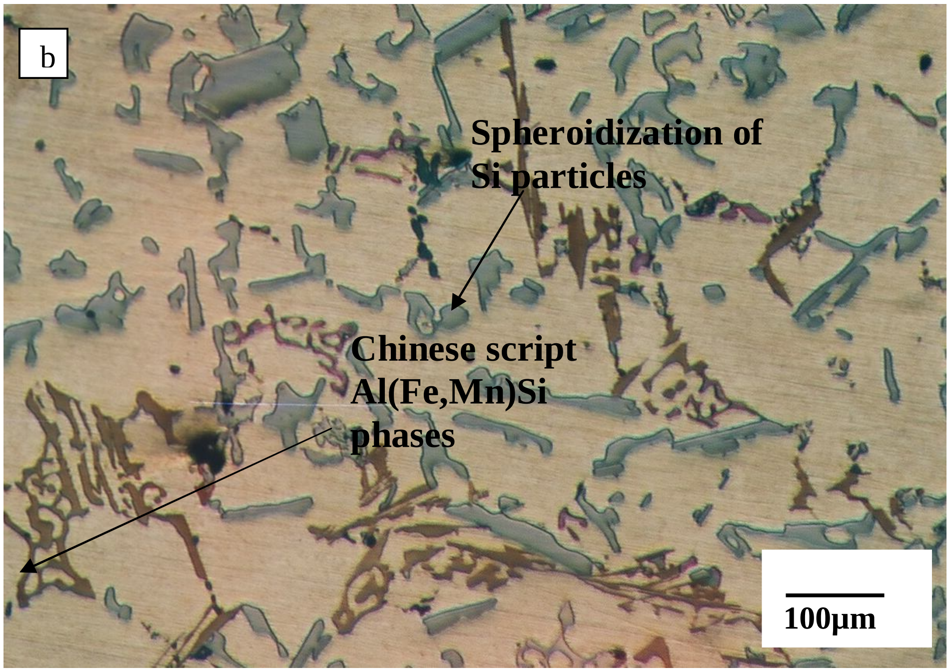


Figure 4.10: (a) As cast microstructure of LM13 alloy with 0.3%Mn and 0.2%Cr showing large acicular Si particles, Chinese script Al(Fe,Mn)Si, flake-like AlCuNi and blocky phases (b) T6 version of (a) showing some spheroidization of Si particles as well as fragmentation of Chinese script Al(Fe,Mn)Si

With a high Fe level of 1.06% in the LM 13 base alloy, it is expected that the number of complex intermetallic compounds containing Fe could form consistent with observations made by Gowri and Samuel (1994) that insoluble Fe complexes and their volume percent usually increase with increase in Fe content. However, in the piston alloy used in this study the AlCuNi intermetallics with long flake-like structures were the predominant phases as shown in Figure 4.4a. The Ni and Cu that combine to form this phase have been found to play a significant role in improving high temperature performance of piston alloys as reported by Moffat (2007). Moreover, Moffat (2007) noted that other Fe-containing phases such as AlFeNi and AlFeCu could also form in this alloy. It can be surmised that in the acicular form, the AlCuNi phase in addition to the Fe-bearing phases can inhibit enhanced properties. This phase must therefore be changed into a less detrimental form as reported by Cho et al. (2006) through addition of alloying elements. In this section of the study, the size and shape of the AlCuNi phases was controlled through combined addition of Mn and Sr (Figure 4.11 and 4.12). In addition, Sb was also combined with Mn and the effect on microstructural features assessed (Figure 4.13).

It is noted from Figures 4.11 to 4.13 that combined additions of Mn with Sr or Sb converted the flake like AlCuNi phases to Chinese script morphology or blocky phases which can in turn be expected to impart significant improvements on the mechanical properties of the alloys. Heat treatment schedule on the alloys with combined additions of Mn with Sr or Sb (Figures 4.11b, 4.12b and 4.13b) indicate that the eutectic Si gets spheroidized and coarsened. In addition the inter particle spacing between the eutectic Si particles is increased.

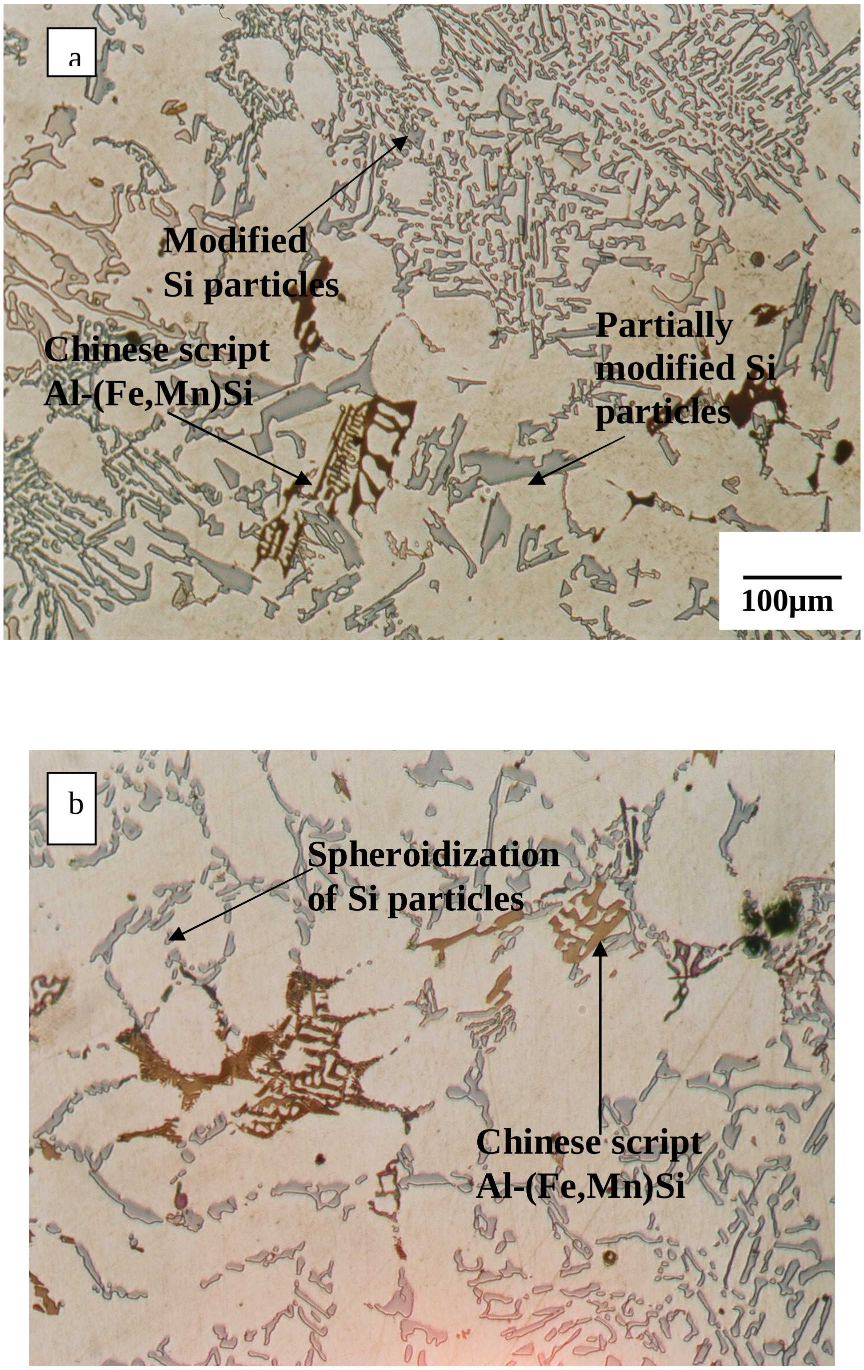
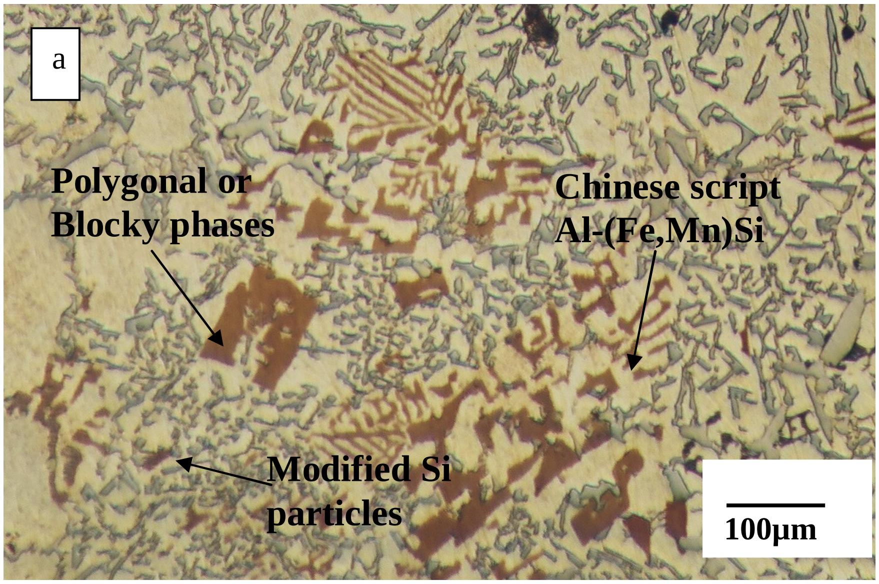


Figure 4.11: (a) As cast microstructure of LM13 alloy with 0.53%Mn and 0.02%Sr showing modified and partially modified Si particles and Chinese script Al[illegible] version of (a) showing some spheroidization of Si particles as well as the Chinese script Al(Fe,Mn)Si



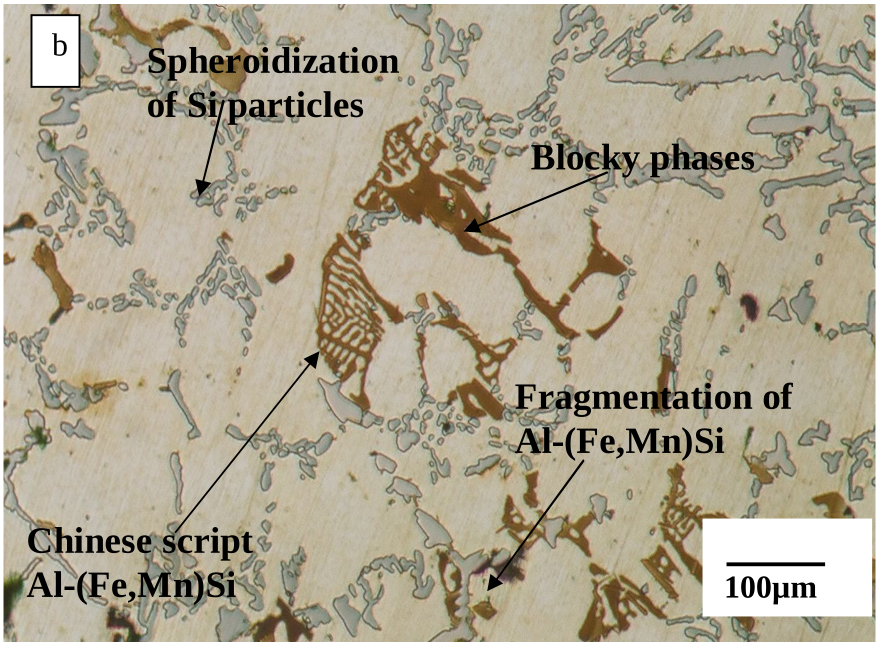


Figure 4.12: (a) As cast microstructure of LM13 alloy with 0.53%Mn and 0.05%Sr showing modified Si particles, the Chinese script Al(Fe,Mn)Si and blocky phases (b) T6 version of (a) showing some spheroidization of Si particles as well as fragmentation of Chinese script Al(Fe,Mn)Si

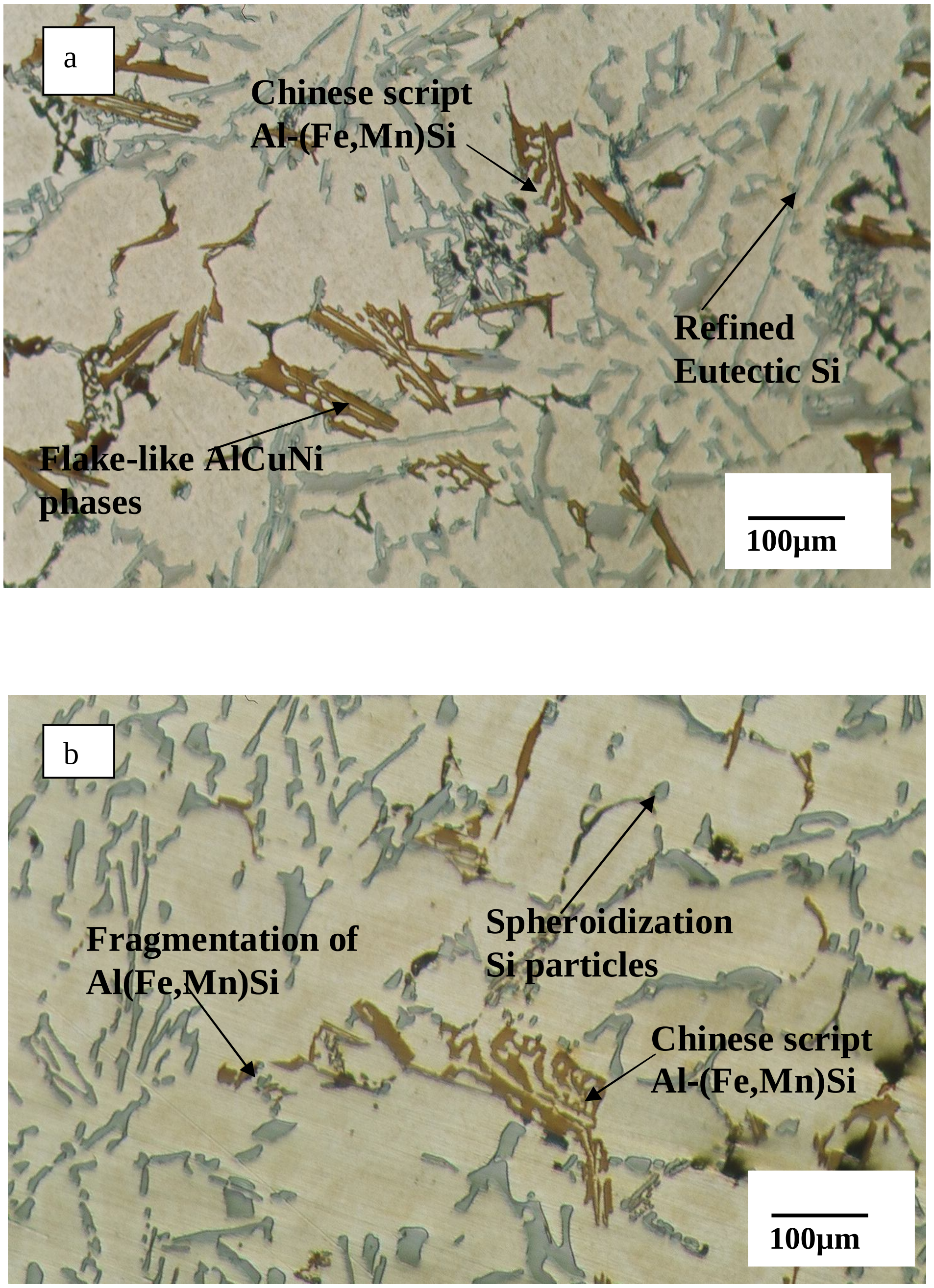


Figure 4.13: (a) As cast microstructure of LM13 alloy with 0.53%Mn and 0.2%Sb showing refined Si particles, flake-like AlCuNi phases, Chinese script Al(Fe,Mn)Si (b) T6 version of (a) showing some spheroidization of Si particles as well as fragmentation of Chinese script Al(Fe,Mn)Si

The micrographs showing microstructural effect of combined addition of 0.28%Al-5Ti-1B grain refiner and 0.02%Sr are represented in Figure 4.14. It is noted that these additions have the effect of changing the Al-Si eutectic from the acicular morphology to fibrous form in similar manner as addition of 0.02%Sr does. It is further observed that the additions have the effect of fragmentation of the flake like AlCuNi phases. The fragmentation seems to follow a line meaning that the long continuous flake like structure was broken down in the fragmentation process. Upon heat treatment the modified eutectic Si gets spheroidized and coarsened as shown in Figure 4.14(b). In addition, the inter particle spacing between the eutectic Si particles gets increased due to heat treatment schedule.

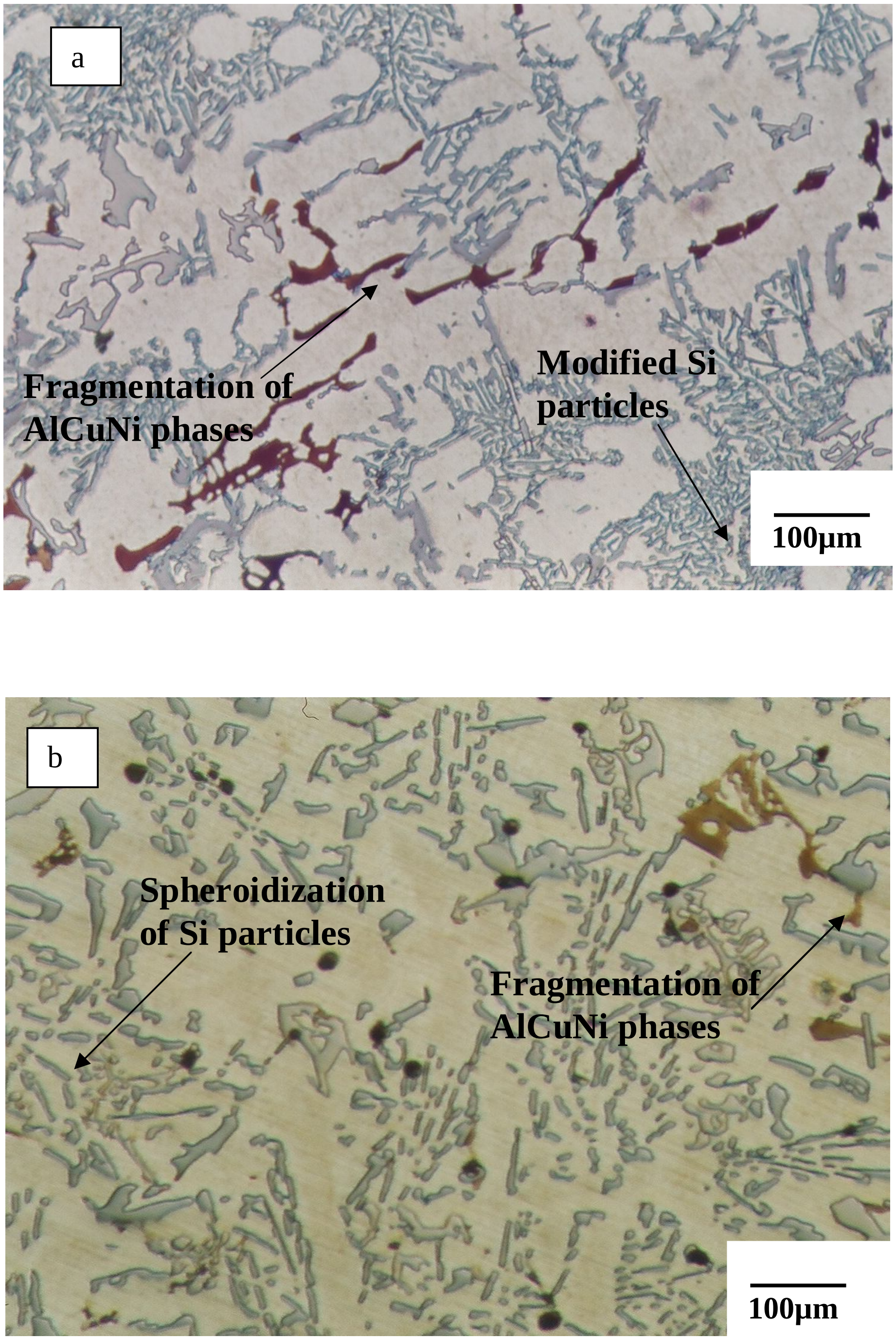


Figure 4.14: (a) As cast microstructure of LM13 alloy with 0.02%Sr and 0.28%Al-5Ti-1B showing modified Si particles and fragmented flake-like AlCuNi phases (b) T6 version of (a) showing some spheroidization of Si particles as well as fragmentation of AlCuNi phases

4.4 Mechanical Properties

4.4.1 Results

Table 4.4 shows the variation of UTS, percent elongation, Brinell hardness and impact energy values of LM13 with minor element additions. It is noted that the average values of the UTS, elongation, Brinell hardness and impact energy of the LM13 base alloy were 140 MPa, 2.3%, 78 and 3.43J, respectively in as cast condition while in T6 condition the values were 223 MPa, 2.1%, 83 and 4.15J. With addition of Sr at levels of 0.02% and 0.05% the average values of UTS increased by 4% and 5% respectively in as cast while it increased by 6% and 8% in T6. Moreover, with the 0.02%Sr and 0.05%Sr addition the % elongation increased by 22% and 17% in as cast and by 24% for each level of Sr in T6. When individual additions of 0.53%Mn and 1.06%Cr were made, improvements of 13% and 18% in the UTS were recorded in as cast condition while in T6 condition the improvements were 15% and 16%. With these additions, the % elongation of the alloy increased by 35% and 61% in as cast and by 38% and 43% in T6. A combined addition of 0.3%Mn and 0.2%Cr was noted to increase the UTS and % elongation of this alloy by 12% and 30% in as cast and by 12% and 33% in T6. With addition of 0.2% Sb the UTS and % elongation of the alloy increased by 4% and 30% respectively in as cast condition and by 10% and 17% in T6 condition. It was noted that combining 0.02%Sr with 0.28% Al-5Ti-1B decreases the UTS by 3% in as cast condition. However, % elongation in as cast increased by 17%. On the other hand, the T6 counterpart of this alloy indicated a 3% increase in UTS and a 9% increase in % elongation. When each of 0.02%Sr, 0.05%Sr and

0.2%Sb were combined with 0.53%Mn the UTS values reduced by 3%, 1% and 3% respectively in as cast condition while in T6 condition, the values increased by 1%, 2% and 3% respectively.

The average trends in UTS and % elongation values with various levels of the elements tested are shown in Figures 4.15 and 4.18 respectively. The error bars indicate the scatter in the results. On the other hand, the maximum and minimum trends in UTS values with various element additions are shown in Figures 4.16 and 4.17 respectively while the maximum and minimum trends in elongation to fracture values are shown in Figures 4.19 and 4.20 respectively.

A summary of the percentage changes in Brinell hardness and impact strength with element additions is given in Table 4.5. Additionally, the trends in Brinell hardness and impact strength with various element additions are shown in Figure 4.21 and 4.22. Furthermore, fracture profiles from flat tensile test specimens were analyzed to determine if failure was controlled by casting defects or by particle failure via particle fracture fracture or particle-matrix interface debonding. The nature of the facture profile was also analyzed for interdendritic and/or intradendritic failure modes. The results are shown in Figures 4.25 to 4.34.

Table 4.4: Variation of UTS, percent elongation, Brinell hardness and impact energy values of LM13 with minor element additions

LM13 + any addition	Condition	Average UTS (MPa)	Std Deviation s	Average % elongation	Std Deviation s	BHN	Std Deviation s	Impact Energy (J)	% Change in UTS*
None	As-Cast	140	2	2.3	0.4	78	2	3.43	-
	H/Treated	223	2	2.1	0.2	83	2	4.15	-
0.02%Sr	As-Cast	146	11	2.8	0.1	80	2	3.67	+4.0
	H/Treated	237	17	2.6	0.1	85	3	4.40	+6.0
0.05%Sr	As-Cast	147	3	2.7	0.4	81	2	3.91	+5.0
	H/Treated	241	4	2.6	0.0	87	2	4.65	+8.0
0.53%Mn	As-Cast	158	2	3.1	0.2	83	2	5.73	+13
	H/Treated	257	2	2.9	0.1	92	2	6.19	+15
0.2%Cr + 0.3%Mn	As-Cast	157	8	3.0	0.2	84	2	3.67	+12
	H/Treated	250	1	2.8	0.1	94	2	4.90	+12
1.06%Cr	As-Cast	165	8	3.7	0.6	85	4	6.72	+18
	H/Treated	259	8	3.0	0.2	96	2	7.26	+16
0.02%Sr + 0.28%Al-5Ti-1B	As-Cast	136	5	2.7	0.1	75	2	3.20	-3.0
	H/Treated	229	1	2.5	0.1	84	2	3.67	+3.0
0.2%Sb	As-Cast	146	5	3.0	0.7	78	2	4.15	+4.0
	H/Treated	246	7	2.7	0.1	88	1	4.90	+10
0.02%Sr + 0.53%Mn	As-Cast	136	3	2.6	0.9	79	2	4.40	-3.0
	H/Treated	225	6	2.3	0.2	86	2	4.65	+1.0
0.05%Sr + 0.53%Mn	As-Cast	138	2	3.3	0.5	80	2	4.40	-1.0
	H/Treated	227	2	2.4	0.0	86	3	4.65	+2.0
0.2%Sb + 0.53%Mn	As-Cast	136	2	2.7	0.5	80	2	4.4	-3.0
	H/Treated	230	5	2.5	0.1	90	2	4.90	+3.0

***** The percentage increase or decrease in UTS is based on the average UTS for the base alloy, LM13 in as cast and T6 heat treated samples

Table 4.5: Summary of the percentage changes in Brinell hardness and impact strength with element additions

LM13 + any addition	Condition	Percentage elongation				% change in el.*	BHN	% change in BHN*	Impact Energy (J)	% change in impact*
		(i)	(ii)	(iii)	Average					
None	As-Cast	2.0	2.0	2.8	2.3	-	78	-	3.43	-
	H/Treated	2.4	2.0	2.0	2.1	-	83	-	4.15	-
0.02%Sr	As-Cast	2.8	2.7	2.9	2.8	22	80	3	3.67	7
	H/Treated	2.7	2.5	2.6	2.6	24	85	2	4.40	6
0.05%Sr	As-Cast	2.4	2.5	3.2	2.7	15	81	4	3.91	14
	H/Treated	2.6	2.6	2.6	2.6	24	87	5	4.65	12
0.53%Mn	As-Cast	3.2	3.2	2.8	3.1	35	83	6	5.73	67
	H/Treated	2.8	3.0	2.9	2.9	38	92	11	6.19	49
0.2%Cr + 0.3%Mn	As-Cast	3.2	2.8	2.9	3.0	30	84	8	3.67	7
	H/Treated	2.8	2.8	2.7	2.8	33	94	13	4.90	18
1.06%Cr	As-Cast	4.0	2.7	4.0	3.7	61	85	9	6.72	96
	H/Treated	2.8	3.2	3.0	3.0	43	96	16	7.26	75
0.02%Sr + 0.28%Al-5Ti-1B	As-Cast	2.7	2.8	2.7	2.7	17	75	-4	3.20	-7
	H/Treated	2.6	2.4	2.6	2.5	19	84	1	3.67	-12
0.2%Sb	As-Cast	2.2	4.0	2.8	3.0	30	78	0	4.15	21
	H/Treated	2.7	2.6	2.7	2.7	29	88	6	4.90	18
0.02%Sr + 0.53%Mn	As-Cast	4.0	2.0	2.0	2.6	13	79	1	4.40	28
	H/Treated	2.4	2.4	2.0	2.3	10	86	4	4.65	12
0.05%Sr + 0.53%Mn	As-Cast	4.0	2.8	3.2	3.3	43	80	3	4.40	28
	H/Treated	2.4	2.4	2.4	2.4	14	86	4	4.65	12
0.2%Sb + 0.53%Mn	As-Cast	2.8	3.2	2.0	2.7	17	80	3	4.4	28
	H/Treated	2.4	2.6	2.4	2.5	19	90	8	4.90	18

***** The percentage increase or decrease in percent elongation, Brinell hardness and impact strength is based on the average values the properties for the base alloy, LM13 in as cast and T6 heat treated samples

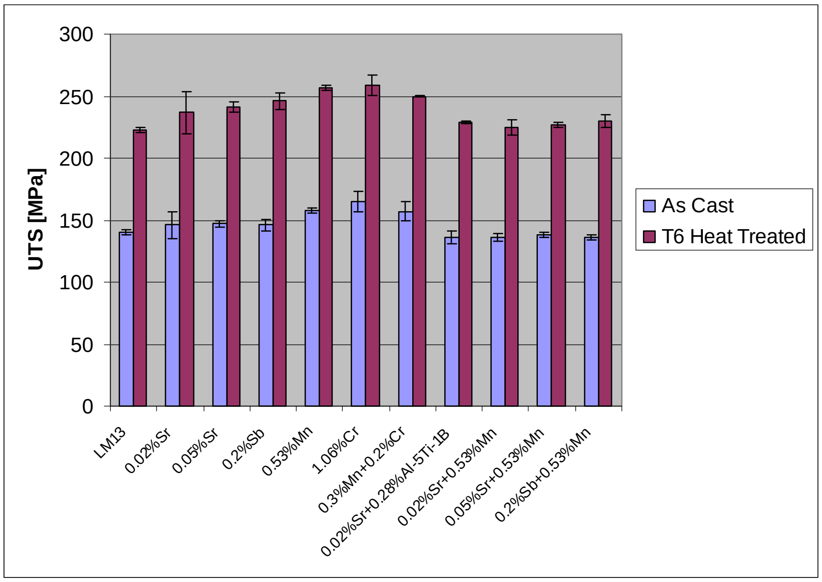


Figure 4.15: Average UTS values for LM13 type alloy as a function of element additions before and after T6 heat treatment with 1.06% Cr addition giving the highest UTS

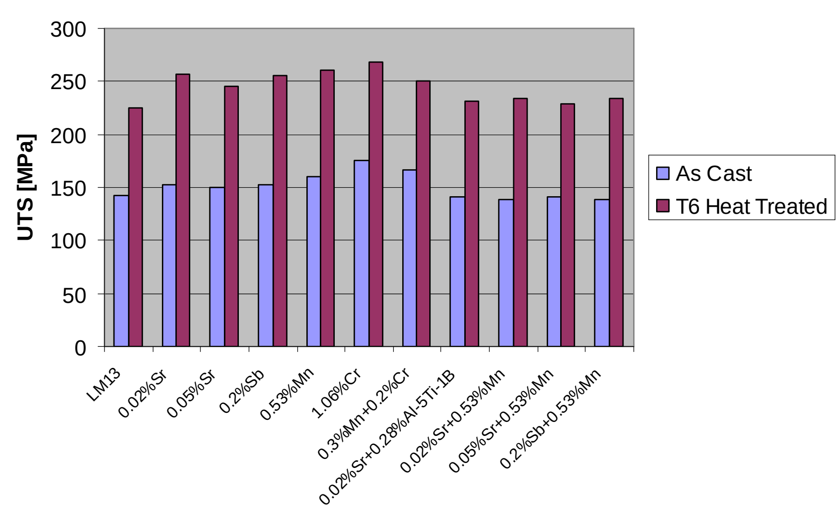


Figure 4.16: Maximum UTS values for LM13 type alloy as a function of element additions before and after T6 heat treatment

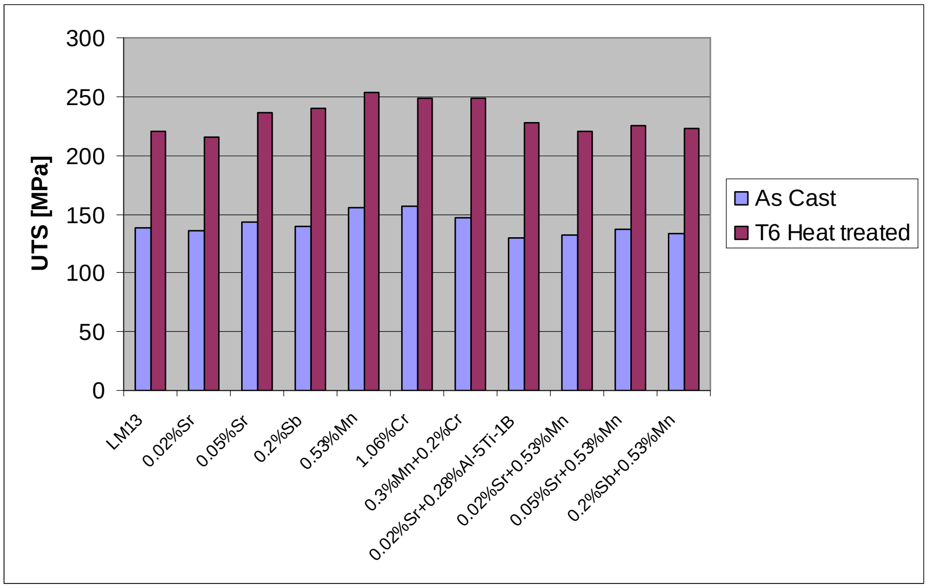


Figure 4.17: Minimum UTS values for LM13 type alloy as a function of element additions before and after T6 heat treatment

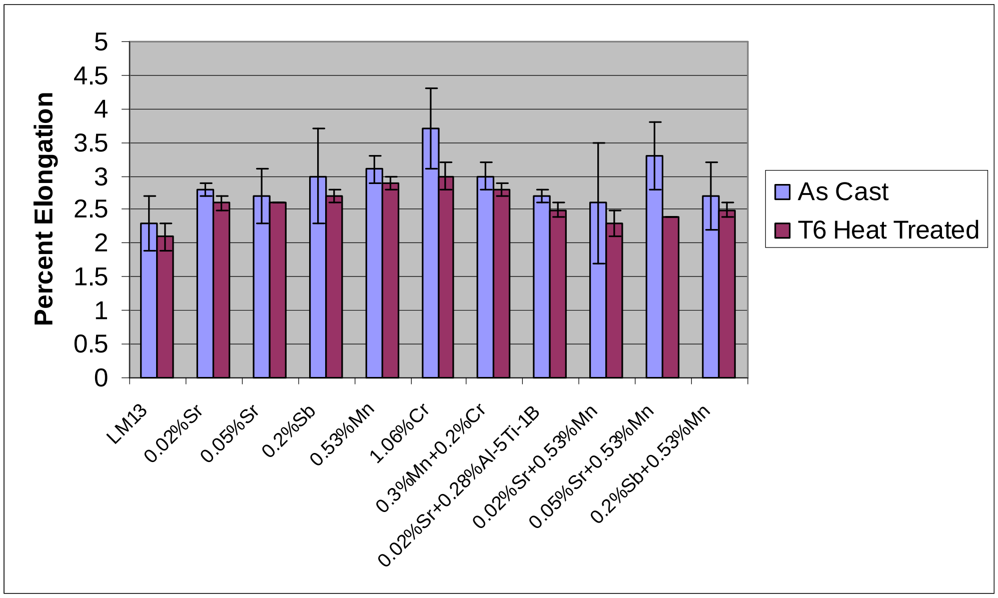


Figure 4.18: Average % elongation of LM13 type Alloy before and after T6 heat treatment with addition of 1.06%Cr giving the highest elongation

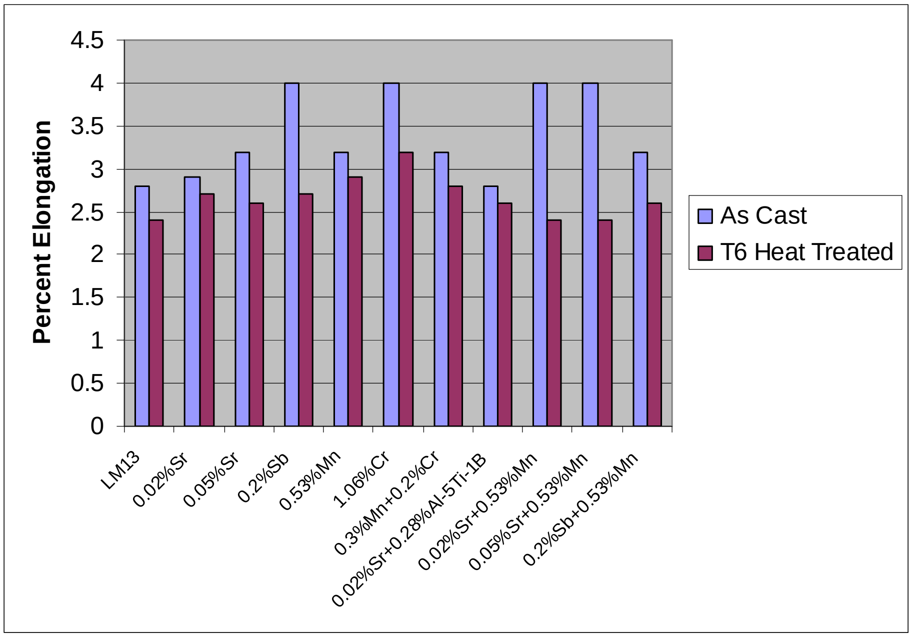


Figure 4.19: Maximum % elongation for LM13 type alloy as a function of element additions before and after T6 heat treatment

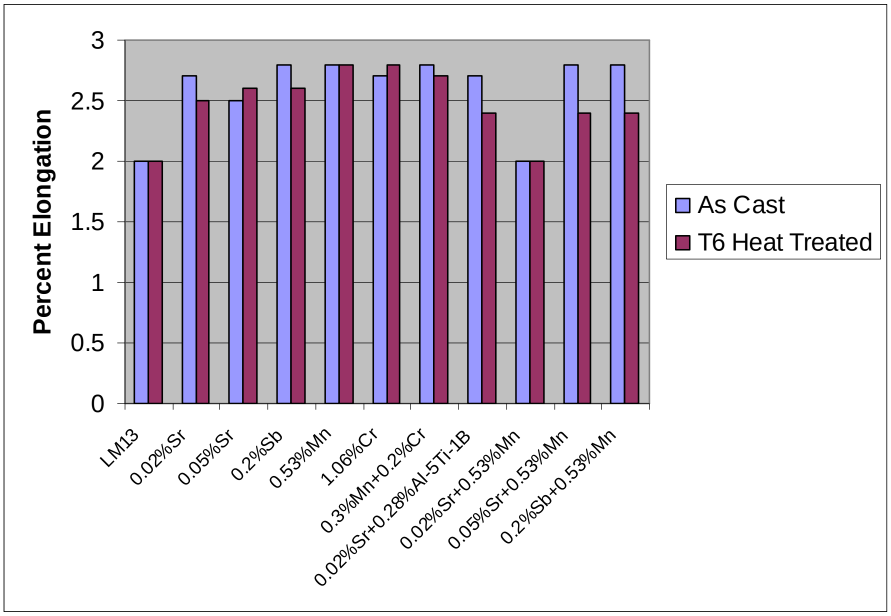


Figure 4.20: Minimum % elongation of LM 13 type alloy as a function of element additions before and after heat treatment

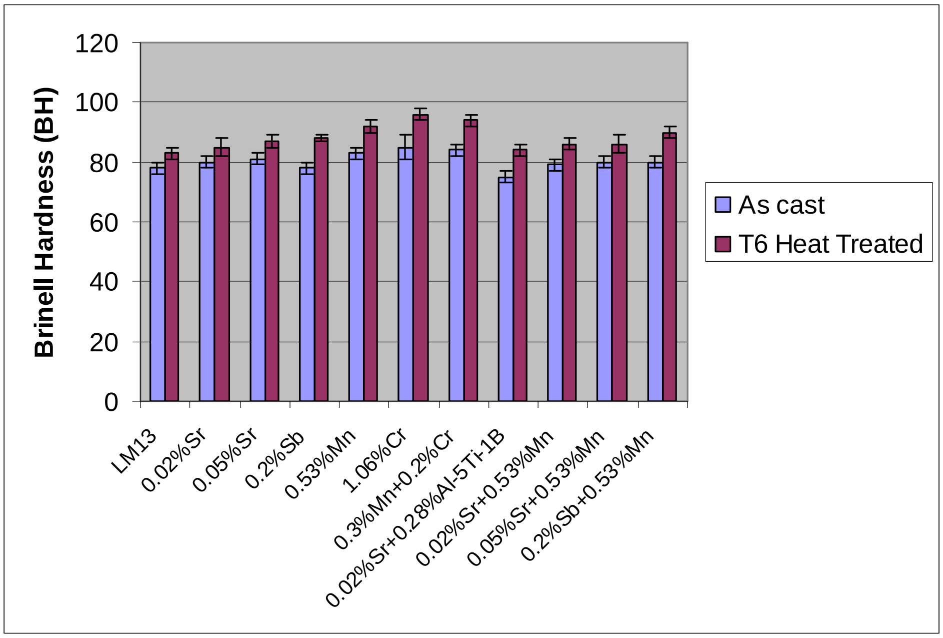


Figure 4.21: Average Brinell hardness values as a function of element additions before and after T6 heat treatment with 1.06%Cr addition giving the highest hardness

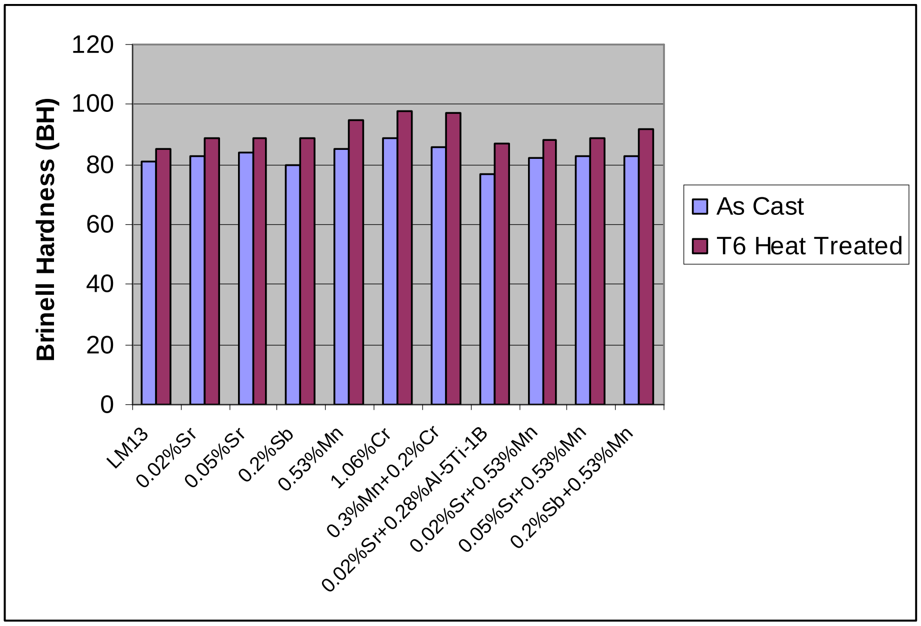


Figure 4.22: Maximum Brinell hardness values as a function of element additions before and after T6 heat treatment

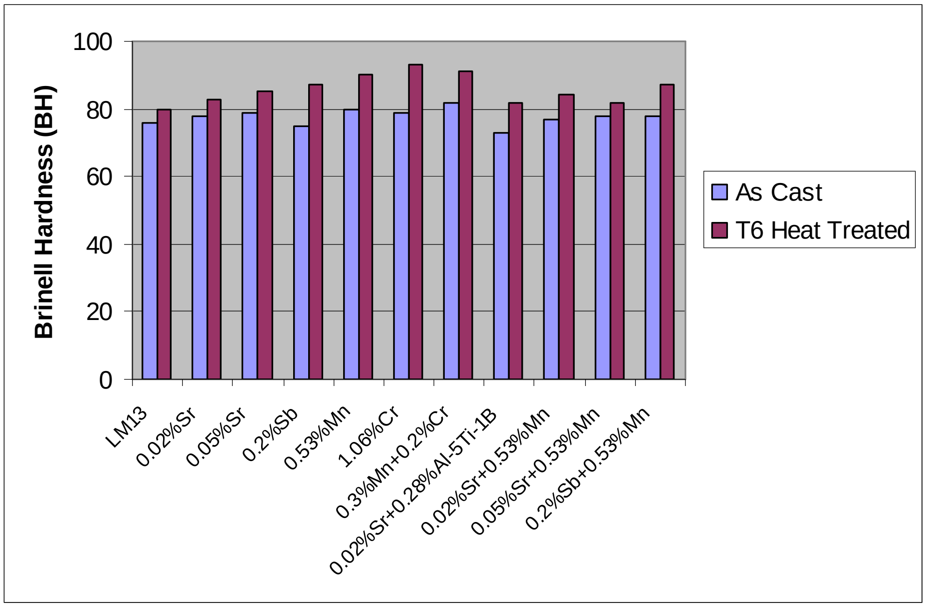


Figure 4.23: Minimum Brinell hardness values as a function of element additions before and after T6 heat treatment

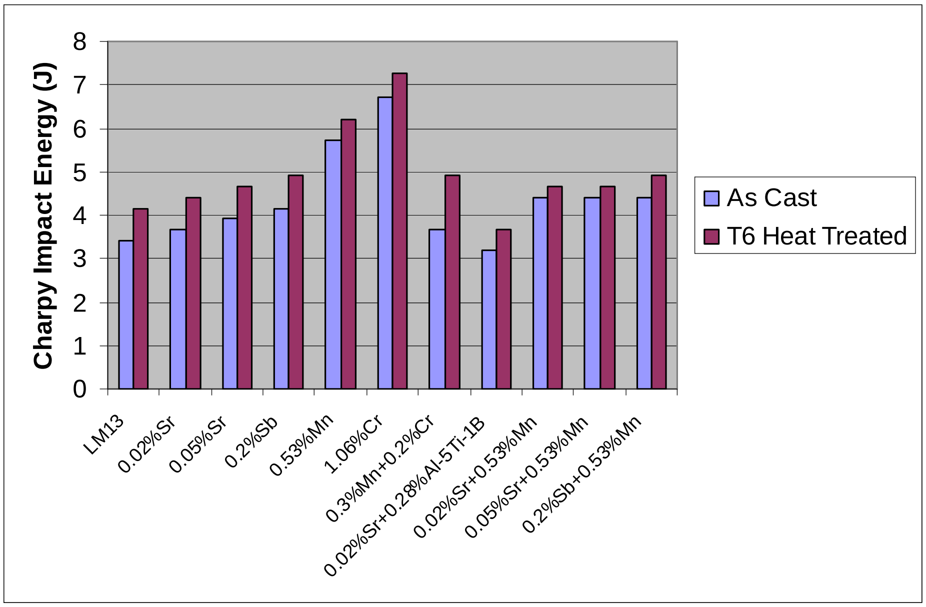


Figure 4.24: Charpy impact energy for LM13-type alloy as a function of element additions before and after T6 heat treatment.

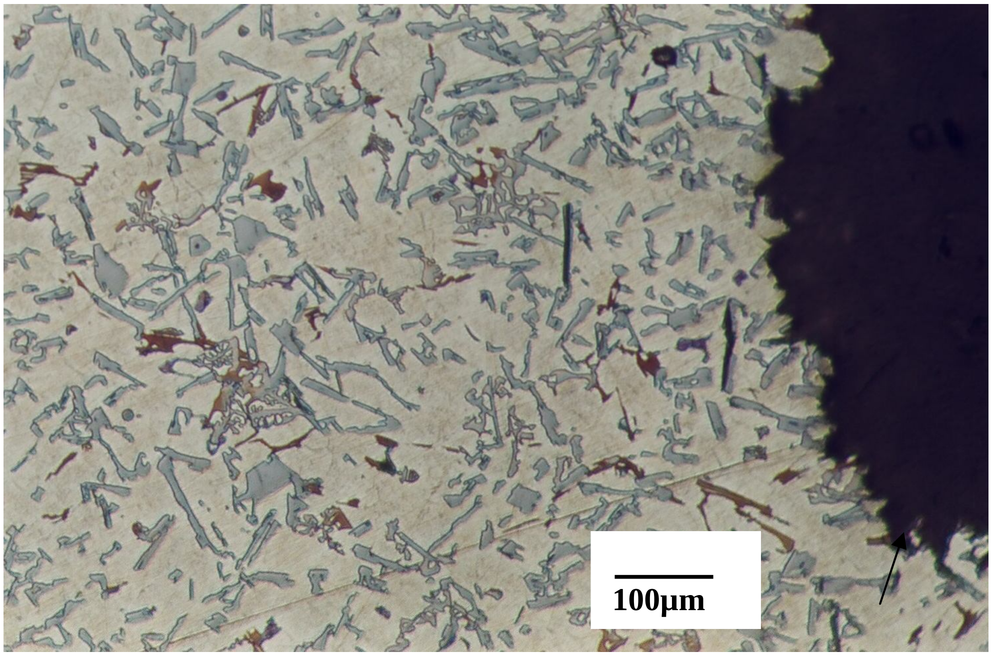


Figure 4.25: As cast microstructure showing near fracture profile of an LM13-type base alloy

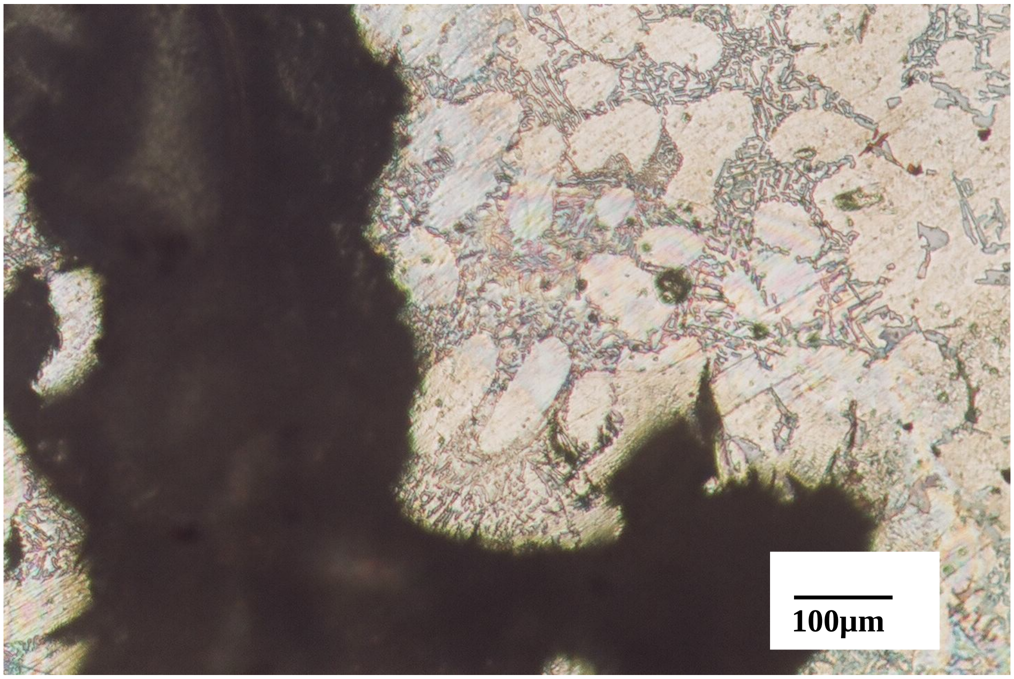


Figure 4.26: As cast microstructure showing near fracture profile of an LM13-type alloy with 0.02%Sr

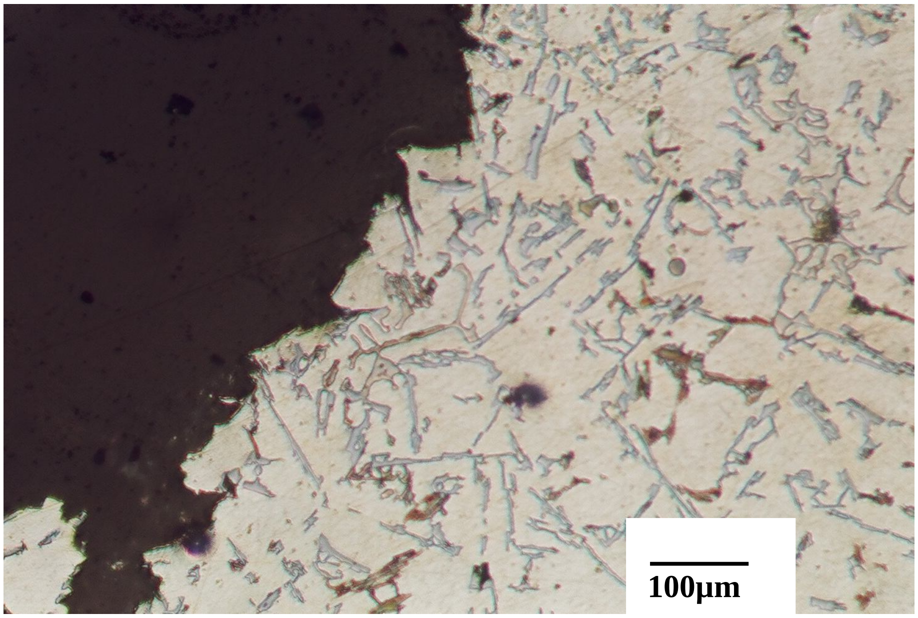


Figure 4.27: **As cast microstructure showing near fracture profile of an LM13-type alloy with 0.2%Sb**

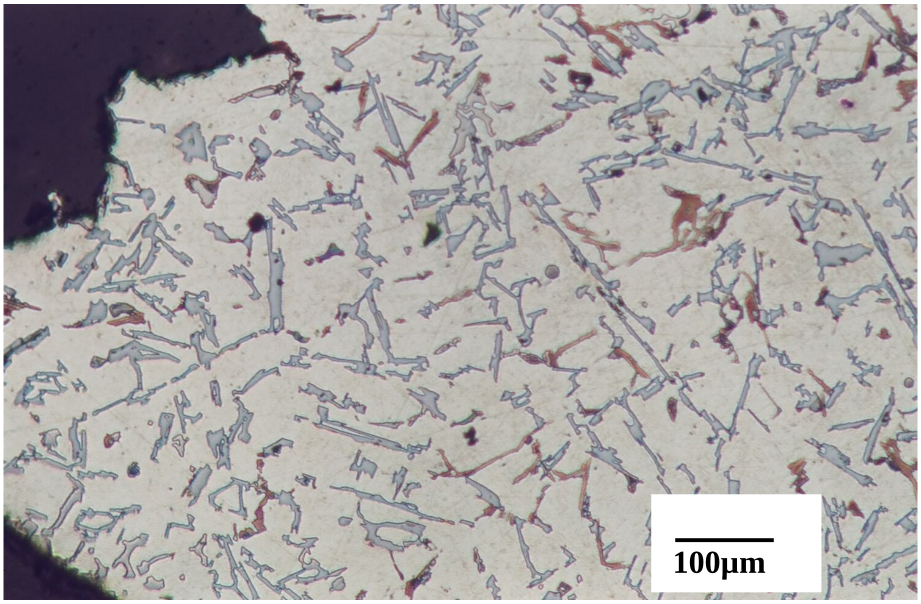


Figure 4.28: **As cast microstructure showing near fracture profile of an LM13-type alloy with 1.06%Cr**

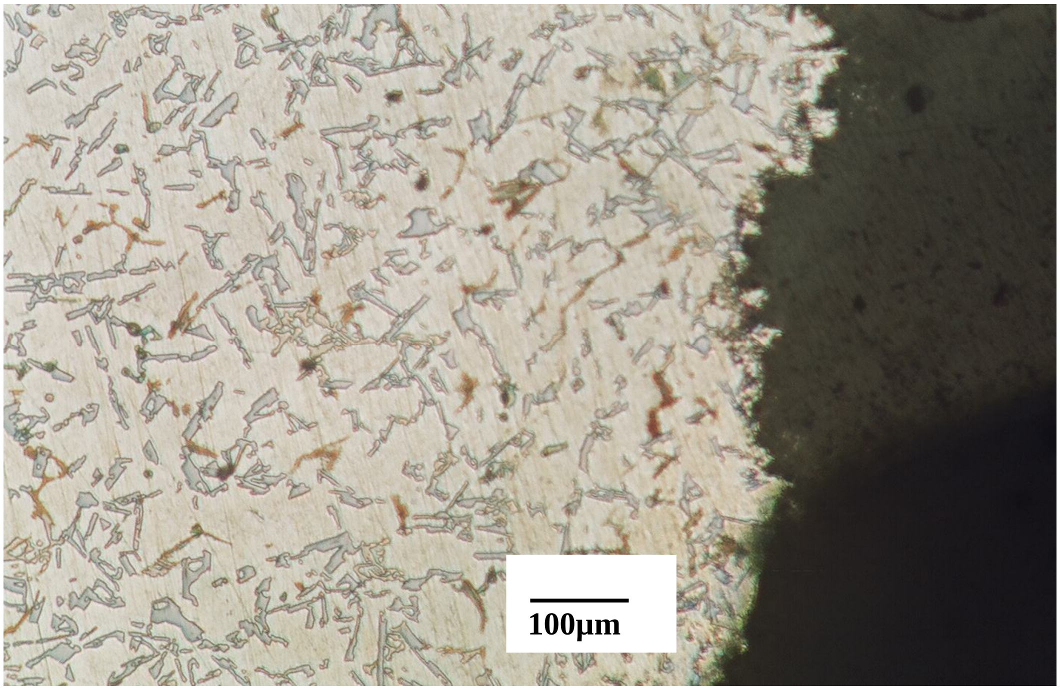


Figure 4.29: As cast microstructure showing near fracture profile of an LM13-type alloy with 0.2%Cr + 0.3%Mn

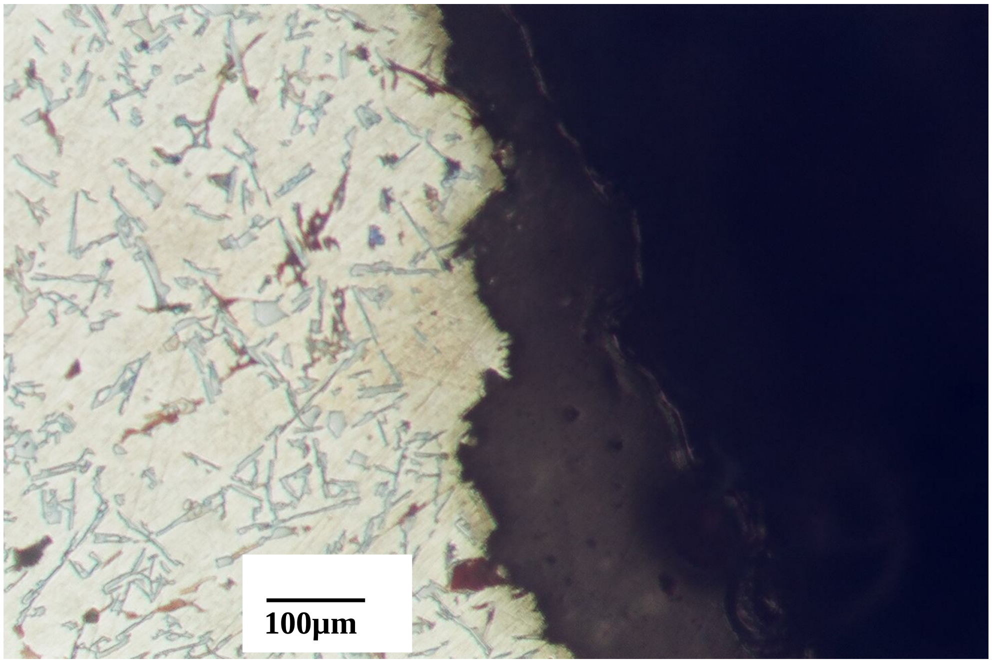


Figure 4.30: As cast microstructure showing near fracture profile of an LM13-type alloy with 0.53%Mn

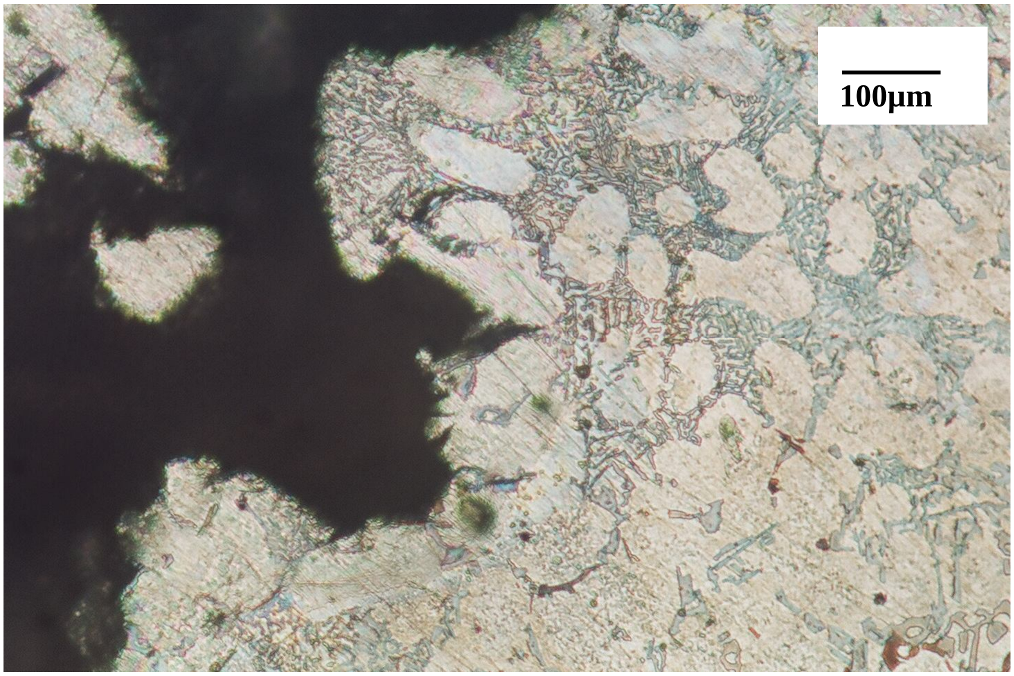


Figure 4.31: As cast microstructure showing near fracture profile of an LM13-type alloy with 0.02%Sr + 0.53%Mn

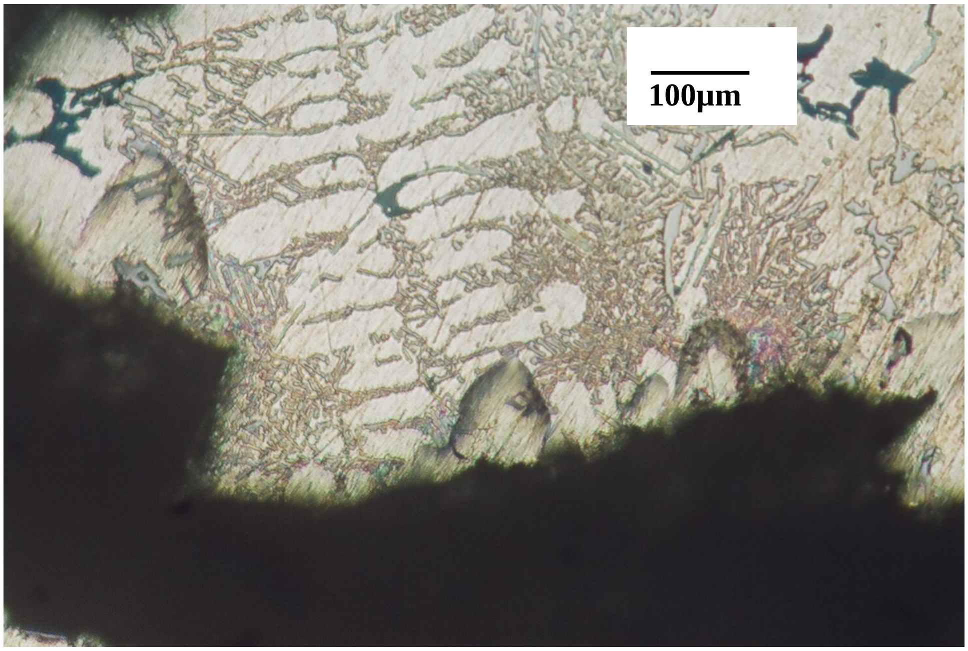


Figure 4.32: As cast microstructure showing near fracture profile of an LM13-type alloy with 0.05%Sr + 0.53%Mn

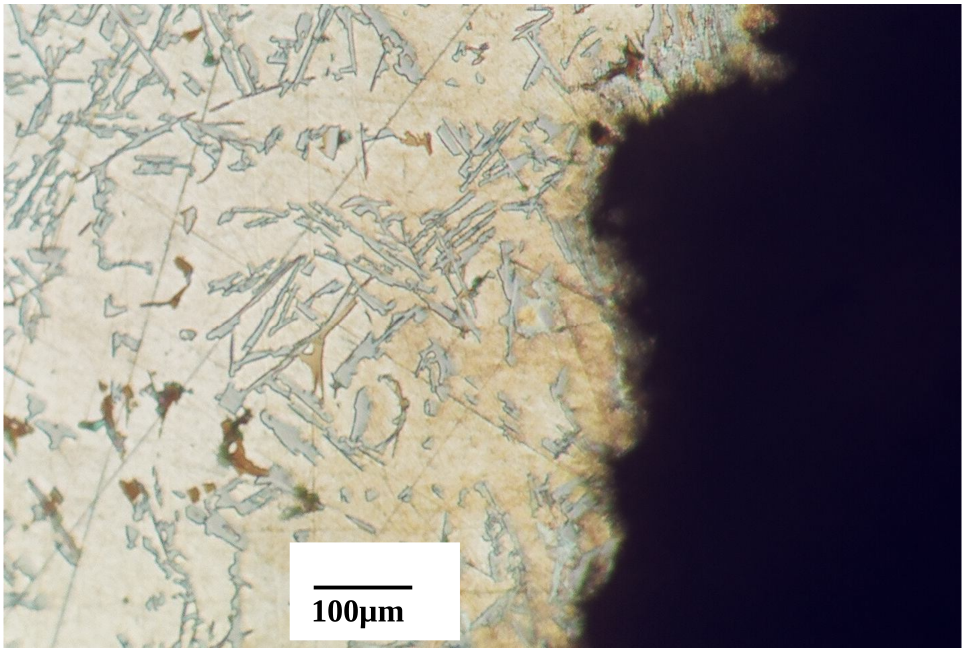


Figure 4.33: As cast microstructure showing near fracture profile of an LM13-type alloy with 0.2%Sb + 0.53%Mn

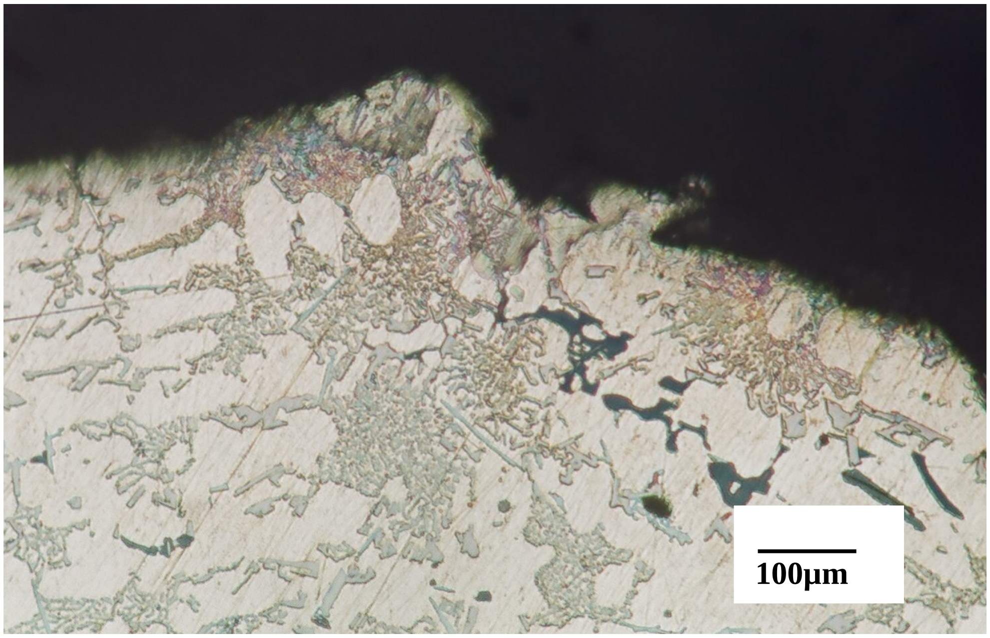


Figure 4.34: As cast microstructure showing near fracture profile of an LM13-type alloy with 0.02%Sr + 0.28%Al-5Ti-1B

4.4.2 Discussion

Looking at the results from Figures 4.15 to 4.20, it is noted that changes in alloy composition can cause significant effects on the UTS and % elongation of LM13-type cast aluminium alloy. UTS values of this alloy in T6 heat treatment condition are noted to be higher as compared to those in as cast condition. A recent study by Hwang et al. (2008) to determine Mn effect on properties of Al-Si-Cu casting alloys showed that UTS values in T6 condition were much higher than those in as cast condition due to the precipitation hardening of the T6 treated alloy. Similar results were obtained by Wang et al. (2004) who explained the enhancement of the UTS in T6 condition based on spheroidization of the eutectic Si particles and dissolution of the intermetallic compounds. In the present investigation, the higher UTS values in T6 condition can be explained based on the precipitation hardening in addition to spheroidization of the eutectic Si (Figure 4.4b) and dissolution of intermetallics compounds (Figure 4.9b). The hardening of the matrix as a result of T6 heat treatment, however, causes significant reductions in % elongations of the alloys as seen from Figures 4.18 and 4.19.

Since alloy variations have an impact on the levels of porosity (Figure 4.3) and intermetallic phases that form in the microstructure (Figure 4.4), it is believed these changes will in turn affect properties as noted from the trends in Figures 4.15 to 4.20. Murali et al. (1992) reported that when Fe is present in the base alloy or picked up during melting, it impacts negatively on the mechanical properties of Al-Si-Mg alloys probably due to formation of Fe-containing intermetallic phases. The Fe level of 1.06% contained in the base alloy used in the present study in addition to the 1.08%Ni and 1.36%Cu led to the precipitation of intermetallic phases during solidification identified AlCuNi. In their flake-like morphology, the AlCuNi phases (Figure 4.4a) in addition to the acicular structure of unmodified eutectic Si particles adversely affected the mechanical properties (UTS and % elongation) of the base alloy as indicated in Figures 4.15 to 4.20. In order to determine the role of microstructural features in deterioration of properties, near fracture specimens were polished and profiles examined using an optical microscope and the results are presented in Figures 4.25 to 4.34.

The microstructure of near fracture profile of the base alloy (Figure 4.25) indicates that cracks occurred in the eutectic Si crystals as well as in intermetallic phases. This is because these phases are known to be hard and brittle hence favoured sites for crack propagation. Similar observations have been reported by Pedersen and Arnberg (2001) who noted that the fracture profile of Al-Si-Mg alloys followed a path of cracked Si particles. At the same time, Wang et al. (2004) reported that the intermetallic compounds act as stress raisers especially at the edge tips of the softer α-Al matrix and therefore serve as points of weakness that diminish strength and ductility. Furthermore, Caceres et al. (2003) observed that the large Si particles and intermetallic phases will crack first before the small and round Sr- modified particles. Others [Mulazimoglu et al. (1996)] indicate that the Fe phases especially the β-phases are poorly bonded to the matrix and will therefore decohere under load and pull out from the surface and cause premature failure. It is to be noted that the base alloy (Figure 4.25) used in the current study with its coarse eutectic Si plates together with the long flake-like AlCuNi plates contributed to rapid cracking due to their brittle nature. Since these plates are more or less interconnected, the crack propagates rapidly through the structure causing failure at relatively low UTS values. Furthermore, the arrow in Figure 4.25 shows that decohesion along the AlCuNi phases contributed to premature failure further explaining why the values of the UTS were low in the base alloy.

Addition of minor elements that cause grain refinement of the primary α-Al grains, modification of the eutectic Si and neutralization of detrimental Fe effects recommended by authors including Samuel et al (2001) as a means to improve properties and hence performance of Al-Si alloys were utilized in the current study. It is noted from Table 4.4 and Figure 4.15 that modification of the eutectic Si using 0.02%Sr, 0.05%Sr and 0.2%Sb increased the UTS by 6%, 8% and 10% respectively in T6 condition. Furthermore, the microstructure exhibited by modified alloys (Figures 4.5 to 4.7) indicate that the Si plates are much more rounded and refined than in unmodified alloy (Figure 4.4) where they appear to have an acicular form with sharp corrugated edges. The observed improvements in UTS due to Sr- and Sb-modification are in line with the results of Jenabali et al. (2004) who reported that modification using 0.013%Sr and 0.1%Sb increased the UTS of A356-type alloys.

Additionally, Yie et al. (1999) recorded a 19.3% increase in UTS with 0.031%Sr addition to Al-11Si alloy and explained this increase based on the spheroidization of the eutectic Si. At the same time, the damage from the needle eutectic Si was reportedly diminished. Moreover, Gruzleski and Closset (1990) noted that the increase in properties due to modification is ascribed to the transformation of the eutectic Si structure from acicular to a fibrous form. Modification through Sr addition, however, has been reported [Nogita et al. (2001)] to be accompanied by increases in porosity levels therefore negating the beneficial effect of improved mechanical properties. Porosity characteristics analysis in the current study indicated that Sr addition increased porosity in the LM13 type alloy (Figure 4.3) further negating the beneficial effect of Sr-modification.

The micrographs from the fracture profiles of Sr- and Sb- modified alloys in Figures 4.26 and 4.27 indicate that cracks occurred in the modified or refined eutectic Si particles. Moreover, it is noted from Figure 4.34 that grain refiner (Al-5Ti-1B) and modifier (Sr) addition to the base alloy, gives a similar fracture profile as a Sr modified alloy. In order to explain the fracture mechanism of Sr- and Sb-modified alloys, it is reasonable to believe that cracking of the Si particles takes place in the same way as in the unmodified alloys but due to small particles. When the size of the eutectic Si particles reduces and fragmentation takes place due to modification or refinement, the crack has to propagate through the more ductile Al matrix which explains why UTS and % elongation values in modified alloys were comparatively higher than in unmodified alloy.

It is well established that Mn and Cr are effective elements that inhibit the formation of plate like β-phases. Though the base alloy in the current study had no β-phases, it is interesting to note that addition of 0.53%Mn led to partial fragmentation of the large flake-like AlCuNi phases shown in Figure 4.4a with a 15% improvement in UTS. Earlier investigations by Cho et al. (2006) revealed that addition of Mn to the piston alloy enhanced elevated temperature properties of the alloy. They attributed this to the formation of large quantities of intermetallics consisting of α-$Al_{12}(Mn,Fe,Cu,Ni,V)_3Si_2$ as well as fine dispersoids of $Al_6(Mn,Fe)$. The authors further noted that the intermetallics formed after Mn addition had Chinese script morphology. The beneficial effect of the Chinese script has been explained by Mulazimoglu et al. (1996) based on the compact shape and diffuse interface with Al-matrix. Additionally, Chinese scripts will present less internal stress concentration than do the sharp flake-like phases. At the same time, the script phase has a stronger interface bonding which results in enhanced strength and a diminished tendency to pull out the phases during loading. The improvements recorded with Mn addition in the present study can therefore be explained based on the formation of phases with Chinese script morphology (Figure 4.8).

The present study revealed that among the elements added to the LM13 base alloy, Cr gave the best results. This occurs after application of an Fe to Cr ratio of 1, a level that many will not use for fear of increased chances of sludge formation. From the micrographs, however, formation of sludge was not observed (Figure 4.9). Instead, addition of 1.06%Cr to the base alloy changed the entire flake-like AlCuNi phases to Chinese script morphology (Figure 4.9) with UTS values increasing by 18% and 16% in as cast and T6 condition respectively. Gustafsson et al. (1986) also reported that while addition of either Mn or Cr will be beneficial to Al-Si alloys in changing the harmful intermetallic phases to less harmful phases, Cr addition could be more beneficial than Mn. Based on this observation and those from the results, it is to be noted that the ability of Cr to convert the entire flake-like phases to Chinese script shape contributed to better properties than Mn which only effected partial conversion of the flake-like AlCuNi (Figure 4.8).

Though others have suggested that combined addition of Mn and Cr are better than individual additions, the current results suggest otherwise. It is noted from Table 4.4 that a combined addition of 0.3%Mn and 0.2%Cr increased the UTS of the alloy by 12% in both as cast and T6 condition. At the same time, the morphology of some of the flake-like intermetallics changed to Chinese script morphology while others did not (Figure 4.10). Reports on combined addition of Mn and Cr to Fe- containing Al-Si alloys including those of Couture (1981) and Murali et al., (1994) indicate that the effectiveness in enhancing the mechanical properties is based on the ability of the elements to change the harmful phases to less detrimental ones. The present improvements in the UTS due to combined addition of Cr and Mn can be explained based on this premise. However, these values are comparatively less than those obtained from Cr addition because the addition only caused partial conversion of the AlCuNi phases to Chinese script form (Figure 4.10).

The micrograph showing the near fracture profile of the alloy with Cr addition is shown in Figure 4.28. It is noted that cracking in this alloy occurred mainly through eutectic Si. In the case of individual addition of Mn and combined addition of Mn and Cr the fracture profiles in Figures 4.29 and 4.30 indicate that fracture occurred through the intermetallics and debonding of particle-matrix interface. Moreover, there was cracking through the eutectic Si particles in these alloys. The occurrence of fracture through intermetallics, eutectic Si and debonding of particle-matrix interface could have worked synergistically to deteriorate UTS giving further evidence why Cr addition gave better properties than individual addition of Mn or when combined with Cr.

Ashtari et al. (2004) reported that a combined addition of Sr and Mn results not only in the modification of Si phases but also in enhancement of the formation of the Chinese script phases. This will in turn give more superior properties due to the combined effect of modification and neutralization, but this was however, not the case in the current study. The UTS and elongation results from combined addition of

Sr and Mn are much lower than those obtained from 1.06%Cr addition. Several reasons can be attributed to this and are explained as follows. Ashtari et al. (2004) noted that for combined Sr and Mn addition to be more effective, the Fe level should be about 0.5%. In this study the Fe content was above the 0.5% level. On the other hand, Shabestari et al. (2002) suggested that addition of low levels of Sr (up to 0.05%) neutralizes the positive effect of Mn in reducing the harmful phases. At the same time, it is noted from Figure 4.3 that the combined addition of Sr and Mn did not cause significant reduction in porosity as compared to Cr addition. It is therefore true that these factors must have worked synergistically to deteriorate the properties when Sr and Mn additions were made to the alloy. In fact, the percentage increase in UTS due to combined addition of 0.02%Sr and 0.53%Mn was 1% while that due to 0.05%Sr and 0.53%Mn was 2% in T6 condition. From the fracture profiles in Figures 4.31 and 4.32 it is observed that cracking occurred in modified eutectic Si as well as through the intermetallic phases. Additionally, cracks are thought to have occurred through the pores. Though not apparent from the micrographs, it is speculated that as the crack progressed through the Si particles and the intermetallic phases, the presence of pores made it easier to propagate through the Al matrix thereby reducing the maximum obtainable UTS and % elongation values.

The results from hardness tests (Figures 4.21 to 4.23) indicate that the base alloy had a low Brinell hardness value. However, with addition of Sr at levels of 0.02% and 0.05% the Brinell hardness increased by 3% and 4% respectively in as cast condition and by 2% and 5% in T6 condition. The observed increase in hardness due to Sr addition compares well with the results of Garcia-Hinojosa et al. (2003) who recorded a 7% increase in Brinell hardness upon addition of 0.02%Sr to an Al-7Si-0.5Ni alloy. Treatment with Sr and the subsequent increase in Brinell hardness can be explained based on the following reasons. The presence of Sr is known to deactivate oxide films, making them unfavorable substrates for Fe-rich intermetallics. Having no favourable substrates on which to form, the intermetallics would not form. This makes the Fe to remain in supersaturated solution to strengthen the matrix by solid solution hardening. Furthermore, Sr addition increases the tendency to segregate certain phases such as Al_2Cu which then go into solid solution to strengthen the matrix. Based on the role that Sr plays in increasing the matrix strength, the hardness values are noted to increase accordingly as seen from the results in Table 4.4 and Figure 4.21.

When individual additions of 0.53%Mn and 1.06%Cr were made, improvements in Brinell hardness of 6% and 9% respectively were recorded in as cast condition while in T6 condition the improvements were 11% and 16%. It was further observed that a combined addition of 0.3%Mn and 0.2%Cr increased the hardness of this alloy by 8% in as cast condition and by 13% in T6 condition. Among the sources consulted, no studies on the effect of Mn and/or Cr on hardness have been conducted for comparison. However, the observed increase in Brinell hardness values due to addition of these elements can be explained based on the role these elements play in strengthening the matrix. As noted by Narayanan et al. (1995) some alloying elements do go into solid solution to form strengthening precipitates that lead to higher hardness values as observed from the results in the current study. Furthermore, it is believed that change of morphology from the deleterious flake-like morphology (Figure 4.4a) to the less harmful Chinese script morphology (Figure 4.9) offers a strong bonding with the matrix which results in enhanced strength and hence higher hardness values.

The hardness values in heat treated condition are observed to be generally higher compared to the hardness values in as cast condition. Earlier investigations [Li et al. (2006)] on the precipitation behaviour of Al-7Si-0.5Mg during heat treatment revealed that addition of Cu into the alloys leads to precipitation of hardening precipitates such as θ'-Al_2Cu phases that contribute to hardness and strength of the alloys. Moreover, the authors noted that the Cu present goes into solid solution to further form strengthening precipitates. The present results can also be explained based on formation of hardening precipitates such as θ'-Al_2Cu during heat treatment that caused the recorded values of hardness in heat treated condition to be higher compared to those in as cast condition. Furthermore, in heat treated condition Fe, Cu and Mg go into solid solution and in this way increase the hardness of the alloys. Moustafa (2009) however, reported that if the Fe-phases are not allowed sufficient time to dissolve during heat treatment, the hardness values for the as cast samples can be higher than those that are heat treated.

When each of 0.2%Sb, 0.02%Sr and 0.05% Sr was separately combined with 0.53%Mn, improvements in Brinell hardness of 3%, 1% and 3% respectively were recorded in as cast condition. In T6 condition however, the improvement was 8% when 0.2%Sb was combined with 0.53%Mn and 4% when each level of Sr was combined with 0.53%Mn. No improvement in hardness was recorded when 0.2%Sb was added to LM13 alloy in as cast condition. With a T6 heat treatment, Sb addition improved the hardness by 6%. There was a reduction in brinell hardness of 4% in as cast alloy when 0.02%Sr was combined with 0.28%Al-5Ti-1B. However, in T6 condition an increase of 1% was observed. Caceres et al. (2003) noted that when there are large intermetallic phases in addition to coarse eutectic Si particles there will be accelerated cracking of the particles which in turn limits the maximum achievable strength of Al-Si alloys. Moreover, the authors noted that the small and round Sr-modified Si particles do not crack as easily as the coarse ones. The improvements observed in the present study due to combined addition of modifiers (Sr and Sb) and Mn can therefore be explained on the basis of change of morphology from one that is coarse and harmful to one that is fine and less detrimental.

It is observed from Table 4.4 that alloying elements impart significant changes on the impact strength of the LM13-type alloy. This is consistent with the results of Paray et al. (2000) who noted that the impact strength of Al-Si foundry alloys depends not only on the alloy composition but also on the solidification conditions. Furthermore, it is well established that coarser microstructures will lower the impact energy of Al-Si alloys. Looking at the unmodified LM13 alloy (Figure 4.4a) it is noted that the Al-Si eutectic has a coarse morphology with sharp corrugated edges. In addition, the Cu and Ni containing phases (AlNiCu) have a large and flake-like structure. The low impact strength values for the base alloy (Figure 4.24) in as cast and T6 condition can therefore be attributed to the role the large microstructural features play in deteriorating the impact strength of the alloy.

Addition of modifying elements at different levels (0.02%Sr, 0.05%Sr and 0.2%Sb) is noted to confer a finer Al-Si eutectic to the alloy as seen from the micrographs in Figures 4.5 to 4.7. Furthermore, Table 4.5 shows that addition of 0.02%Sr and 0.05%Sr to LM13 alloy increased the impact strength by 7% and 14% respectively in as cast condition and by 6% and 12% in T6 condition. These results are in line with those of Paray et al. (2000) who observed an increase in impact energy when Sr-modification was undertaken together with a heat treatment schedule for 1 hour in A356 type alloy. On the other hand, addition of 0.2%Sb increased the impact energy by 21% in as cast condition and by 18% in T6 condition. In order to explain the increase in impact strength due to modification, Gruzleski and Closset (1990) noted that any process that reduces the size of the brittle eutectic Si particles or increases the inter particle spacing will improve the impact properties. In the current study, modification through Sr and Sb led to reduction of the size of the brittle eutectic Si particles. At the same time, solution treatment accomplished the spheroidization and separation of particles (Figures 4.5b to 4.7b) which explains why there was observed increase in impact strength in modified as cast and T6 heat treated alloys.

Previous studies on influence of elements on impact energy have concentrated on modification of Al-Si eutectic. In order to characterize the effect of other alloying elements, addition of Mn and Cr either individually or in combination and with modifiers was done and the results are as shown in Figure 4.24. These additions are noted to have remarkable improvements in impact energy. When separate additions of 1.06%Cr and 0.53%Mn were made to the alloy, the impact strength increased by 96% and 67% respectively in as cast condition and by 75% and 49% in T6 condition. At the same time, a combined addition of 0.2%Cr and 0.3%Mn increased the impact energy by 7% in as cast condition and by 18% in T6 condition. The increase in impact strength as a result of these additions can be explained based on the transformation of the AlCuNi phases from the coarse morphology to relatively finer Chinese script morphology (Figures 4.8 to 4.10). In the coarse morphology, these phases act as stress raisers that place a limit on the maximum achievable impact strength of the alloy.

When 0.53%Mn was combined with either 0.02%Sr or 0.05%Sr the impact energy was observed to increase by 28% in each of the as cast alloy and by 12% in each of the T6 alloys. Combining 0.2%Sb with 0.53%Mn on the other hand, increased the impact energy by 28% in as cast condition and by 18% in T6 condition. The increase in impact properties due to these additions can be explained based on the transformation of coarse microstructural features (Figure 4.4) to finer and more rounded ones (Figures 4.11 to 4.13). Whereas Mn acts to reduce the size of Cu and Ni containing phases, Sr and Sb refine the eutectic Si. Furthermore, the T6 heat treatment schedule accomplishes the spheroidization of Si particles and fragmentation of the intermetallic phases.

CHAPTER FIVE

5.0 SURVEY OF LOCAL FOUNDRIES

5.1 Introduction

While there has been tremendous growth of the foundry industry since independence, its contribution towards the national economy has been insignificant. The impact of local foundries at national level is yet to be felt, especially in the manufacturing sector. Although many foundries are aware of the enormity of the problems facing their industry, they are unable to communicate with each other on methods of how to resolve their problems. This difficulty in communication coupled with a great number of foundrymen using old technology has been a source of great weakness that has retarded development of local foundries. Even as the quality and technology of world foundries continue to advance, the question as to whether the local foundries are moving in tandem with the world advances arose. It was felt that an up to date survey of the local foundries was necessary, to find out the aluminium casting practices such as the recycling methods, casting design, melt treatment and quality control aspects, and whether these practices are effective in making premium quality castings.

The full list of the foundry shops involved in this survey is shown in Appendix I. The shops selected are distributed in Nairobi and its environs, Mombasa, Nakuru and in the sugar belts. In order to elicit information from the foundries, contact was done through a questionnaire, visit or both. The questionnaire used in this study is shown in Appendix II while the key questions selected for detailed analysis are presented in

Appendix III. The summary of the responses to key questions are shown in Appendix IV.

5.2 The Survey

During the survey it was observed that a number of components are cast in the local foundries. These components range from domestic to ornamental items and industrial products. Figures 5.1 to 5.6 show some of the items commonly cast in the local foundries while Figure 5.7 shows the distribution of cast products amongst the shops involved in the study.

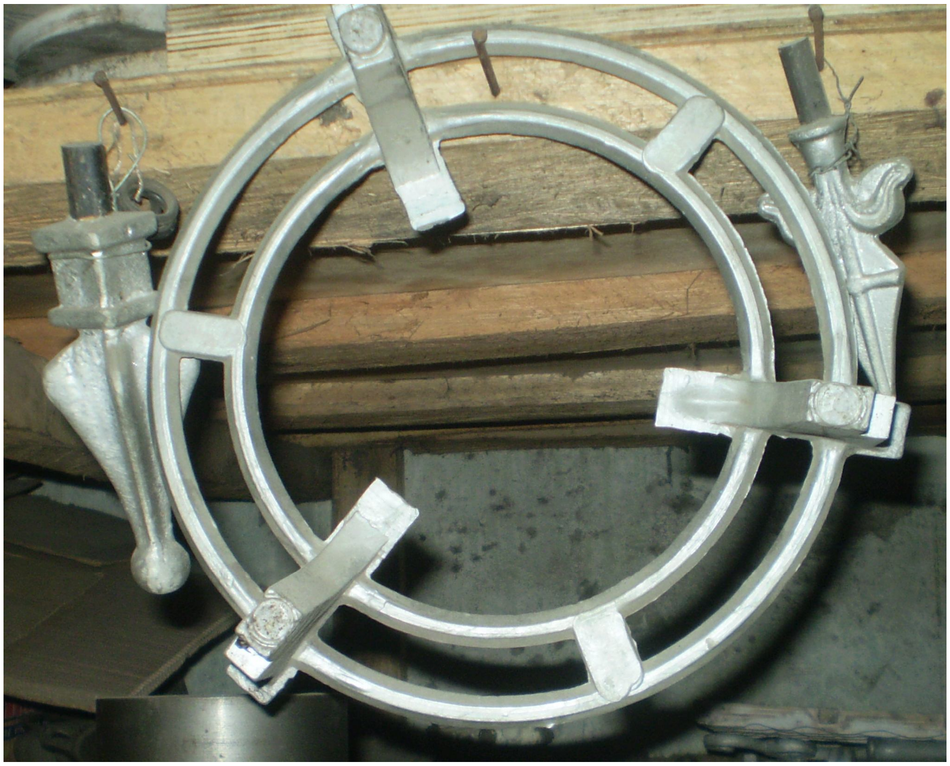


Figure 5.1: Meko jiko grill: Courtesy of Kariobangi Light Industries Ltd, Nairobi

Figure 5.2: Propeller: Courtesy of African Marines & General Engineering Co. Ltd, Mombasa

Figure 5.3: Decorative castings: Courtesy of Kens Metal Industries Ltd, Nairobi

Figure 5.4: Automobile connecting rods with a strainer in between: Courtesy of Kariobangi Light Industries Ltd, Nairobi

Figure 5.5: Decorative work: Courtesy of Dandora Foundry, Nairobi

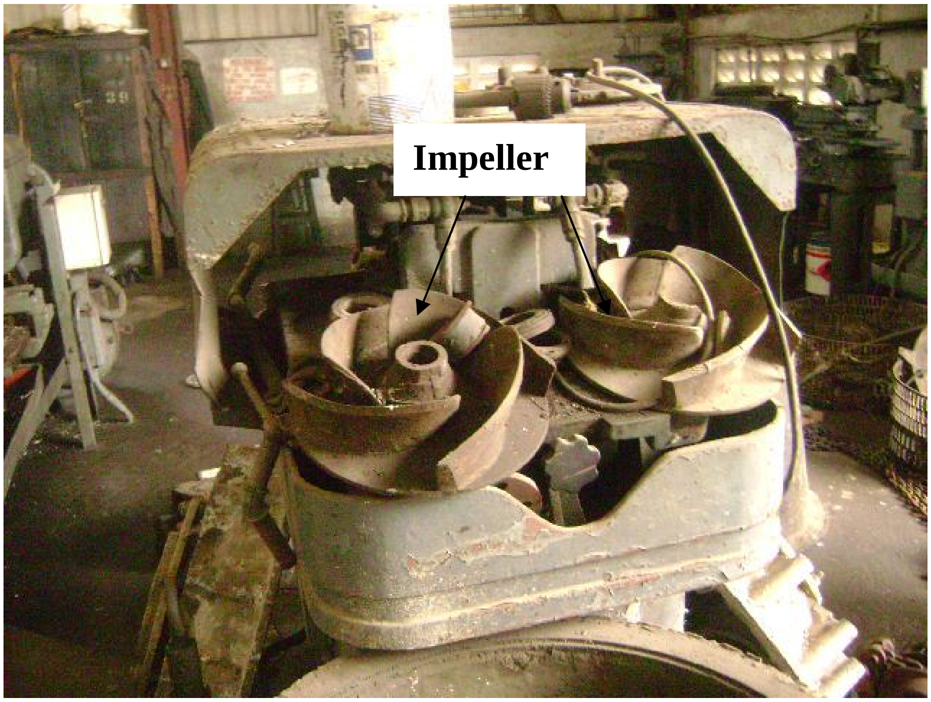


Figure 5.6: Impellers: Courtesy of African Marines & General Engineering Co. Ltd, Mombasa

The casting methods commonly employed in making the above products and the numbers of companies using them are shown in Figure 5.8. It is noted that sand casting is the most preferred method based on its simplicity and low cost implication. It is further noted from the castings above that a number of products like connecting rods, propellers and impellers require some degree of strength for optimal operation during service. To help improve the structural strength of such products, addition of alloying elements has been employed in overseas foundries with results indicating that through this effort more superior properties are achievable. Unfortunately, it is observed from Figure 5.9 that except for degassers, most of the other additives known for their contribution in improving the performance of aluminium alloys are not utilized by local founders. This makes it difficult to achieve reliable castings of premium quality.

It is to be noted further that use of technology known to eliminate historical casting problems such as entrainment of oxide films has not been embraced. Instead, the local casting community has continued to use old technology that neither controls alloy chemistry nor uses additives. Moreover, the methods used do not ensure quiescent melt handling, gas and inclusion measurement and inclusion removal. This has in turn impacted negatively on the quality of locally cast products. Additionally, use of filling systems believed to increase liquid turbulence during flow and hence defects were common. This has led to an increased number of casting rejects attributable to poor workmanship and lack of understanding of the fundamental quality control issues in casting science. These and other challenges have continued to impact negatively on the local casters and are discussed hereafter.

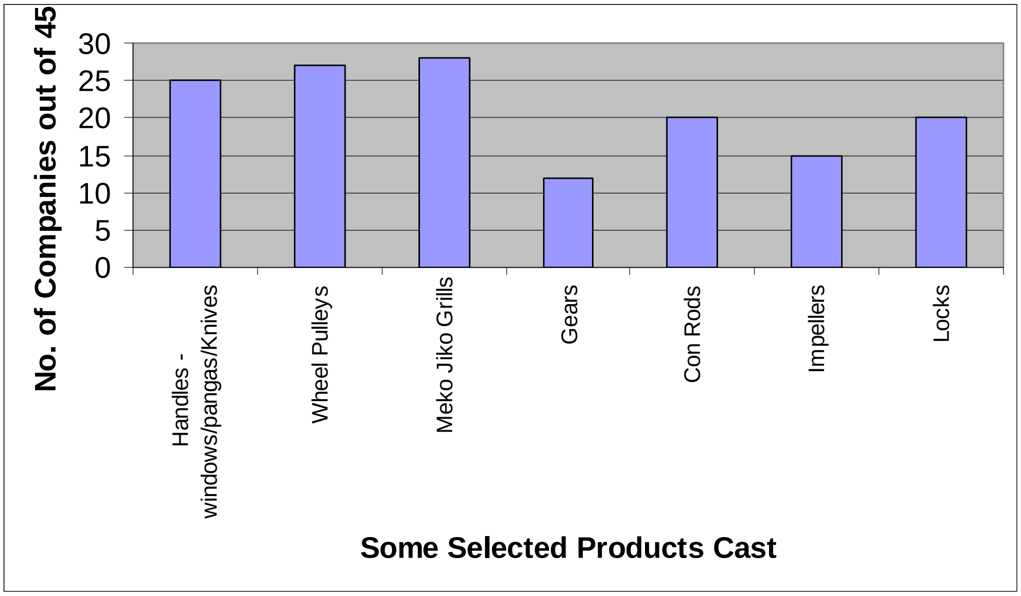


Figure 5.7: Distribution of cast products amongst the foundry shops visited

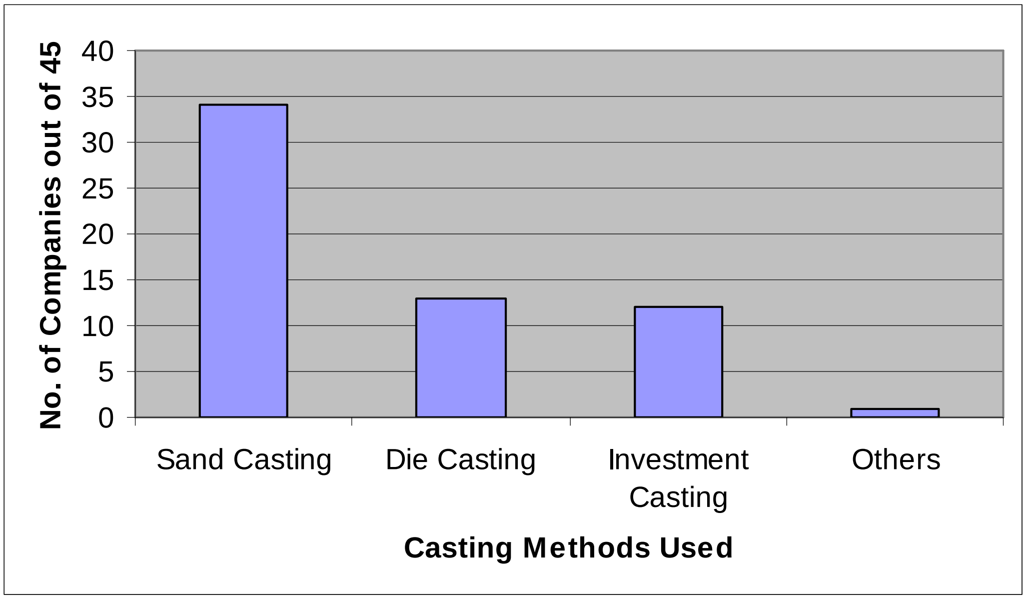


Figure 5.8: Casting methods commonly used in local foundries

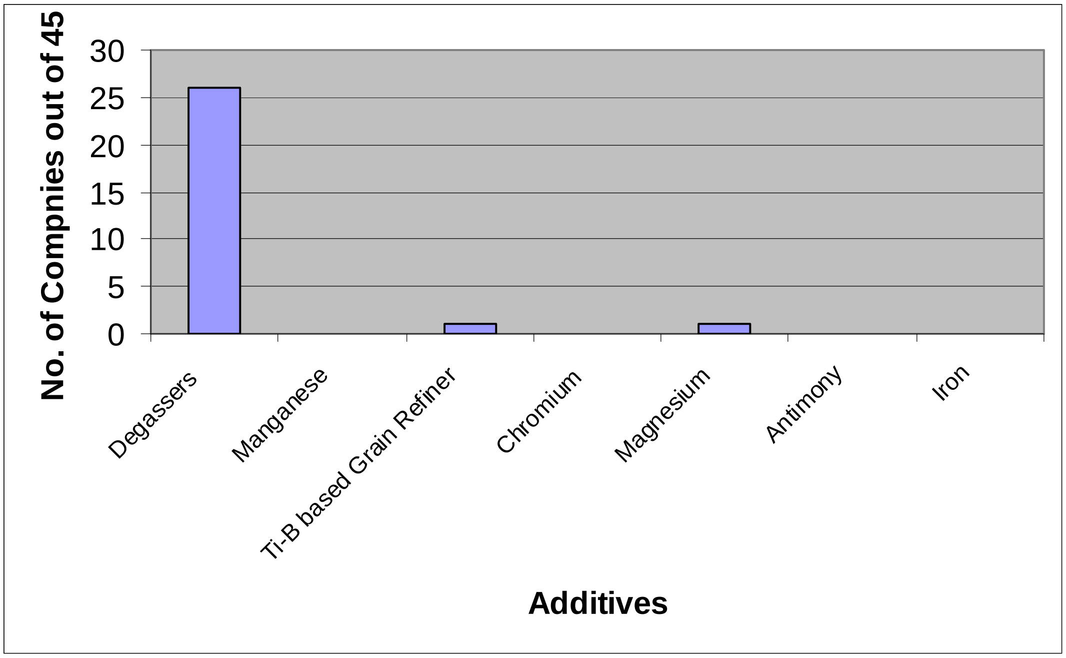


Figure 5.9: Commonly used additives in the local foundries

5.2.1 Challenges Facing Local Foundries in Producing Quality Products

The challenges faced by the local foundrymen have been a major setback to the development of the foundries and have moreover deterred manufacture of premium castings of reliable quality. A number of castings were found to present problems of leakage, premature failures and variable mechanical properties as reported by Campbell (2004). Furthermore, there were numerous cases of castings that suffered porosity, hot tears and cracks. Based on the observations made during the industrial visits and survey, as well as the information provided by the parties involved in the foundry activities, the magnitude as well as the effects of the problems has been analyzed and categorized. In accordance to the extent of the survey, the challenges identified have been categorized into two major categories, namely; technical, and non-technical.

5.2.2 Technical Challenges

This category focuses on the technical setbacks that undermine the quality of castings. These challenges are associated with the type and quality of raw materials, equipment and technical processes such as pattern making, mould making as well as quality control policy and manpower.

5.2.3 Inconsistent Quality of Scrap Metal and Sand

Scrap metal is the major raw material in the casting industry. It is readily available and has relatively low cost. The major source of scrap metal includes; motor vehicle junkyards, machining swarf and domestic wastes like sufurias. The scrap, on delivery to the foundry from vendors, is usually in the form of an assortment of different metals of varying chemical compositions. In this condition, it becomes difficult and costly to accurately sort it out into appropriate categories such as steel, aluminium, iron, bronze and also with respect to their individual chemical composition. This difficulty results in melting of chemically variant scrap which causes a serious problem in controlling the chemical composition of the casting metal. The different metals in the charge act as chemical impurities in the parent metal being cast thus making this practice a major setback to successful recycling of aluminium alloys. Depending on the quality of the desired castings, and effectiveness of the alloying and fluxing process, the resulting cast products would be of inferior quality due to inadequate control of the alloy chemistry. It was however noted that one of the major companies use specific scrap from their own machines which gives them a competitive edge in serving a market that is sensitive to quality.

It is to be noted further that there are no local companies that supply graded sand materials and this forces each foundry to prepare its own sand. Use of such sand that is neither graded nor mixed properly with clay, water and other additives can lead to difficulty in controlling sand casting process which also happens to be the most commonly used casting method (Figure 5.8). This in turn leads to production of poor castings of unreliable quality.

5.2.4 Pattern and Mould Making

The survey showed that only a few foundries are equipped with moulding machines. East African Foundry Works (K) Ltd, Kens Metal Industries Ltd and African Marines & General Engineering Co. Ltd are among the few foundries equipped with moulding machines. Hand moulding becomes the usual practice in most foundries, which leads to mould variability due to human element. The design and production of patterns especially those for unique and/or complex machine parts continue to pose an enormous challenge in most foundries. Due to lack of expertise and equipment for machining of intricate patterns, some foundries are forced to subcontract other organizations such as Kenya Industrial Research and Development Institute (KIRDI) and Numerical Machining Complex Ltd.

In making moulds, whether sand or permanent moulds, it is expected that the filling system be designed in a way that optimizes the properties and reliability of the castings. Recent studies by Rezvani et al. (1999) indicated that poor filling systems in aluminium casting alloys can lead to generation of oxide films in the casting and subsequent degradation of mechanical properties. Knowledge of proper mould design and filling system was found to be lacking in the local foundries as shown in Figure 5.10. Most casting operations for instance, use a conical pouring basin as shown in Figure 5.11 to funnel the melt into the mould cavity. As reported by Campbell (2004), the conical shape will act as an air pump that maximizes entrainment of defects. So as to minimize entrainment of defects into castings, it was noted that conical basins should be substituted with filling basins that reduce entrainment of defects such as bubbles and bifilms. An offset stepped basin is suggested in this case which according to recent studies at the University of Birmingham spearheaded by Prof. Campbell indicate beneficial results in reducing entrainment of defects hence solving many undesirable features of the conical basin.

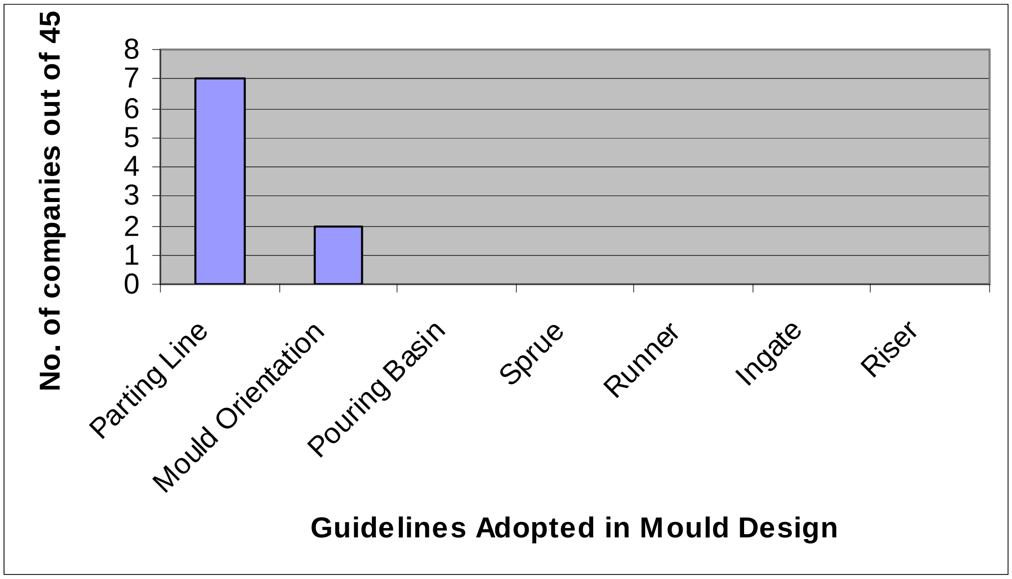


Figure 5.10: General understanding of mould design and filling system by local founders

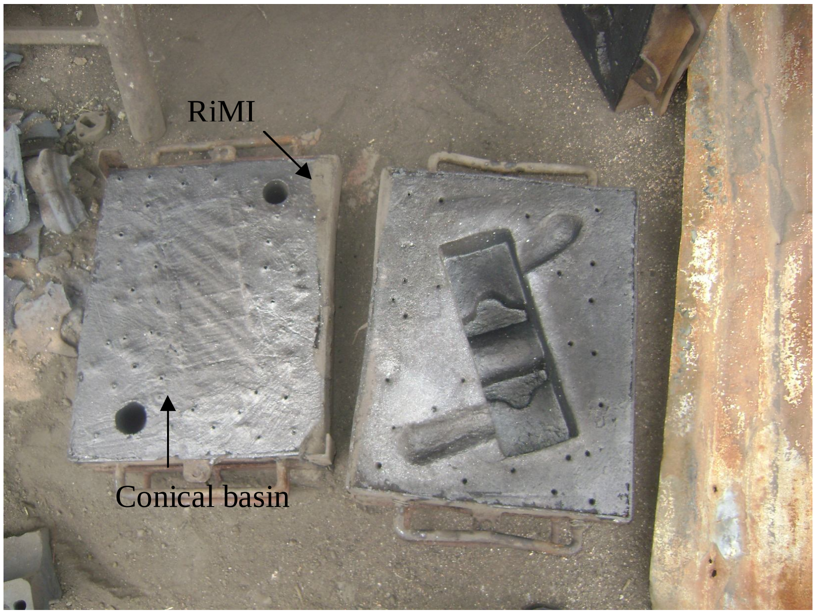


Figure 5.11: Sand mould with a conical basin for aluminium casting: Courtesy Hakika Engineering Workshop, Nakuru

5.2.5 Difficulties in Controlling Melt Temperature

It was noted that the furnaces commonly used for melting scrap in local foundries have no inbuilt temperature control mechanisms. This is besides lack of portable pyrometers for use for temperature measurement and control. This results in uncontrolled heating of the metal that occasionally results in loss of volatile alloying elements such as Mg. It also leads to ineffective fluxing. Furthermore, the melt superheat required for optimal flow of metal in the mould cavity for optimum solidification process is not properly controlled and this often results in casting defects such as surface porosity due to excessive gases being generated and trapped inside the cavity as a result of hydrogen pickup and thermal decomposition of binders. In addition, uncontrolled melt superheating may lead to formation of high levels of oxides as reported by Narayanan et al (1994). When oxides on the surface get entrained, it gets folded forming an inner dry side and the wetted outer side. Studies indicate that the dry side of the oxide will act as a crack and hence a source of failure. There is considerable evidence in literature [Campbell, 2003] that the wetted outer surface act as favoured substrates of some second phases; which further serve to deteriorate properties and reliability of castings. At the same time, difficulty in proper control of melt temperature in the furnace may lead to premature solidification especially when the right pouring temperature is not attained. This results in funneling under melted metal into the mould cavity causing a possible blockage of the flow. Figure 5.12 shows an example of a crucible furnace used in one of the foundries.

Figure 5.12: Furnace with no temperature control for melting aluminium scrap: Courtesy of Booth Extrusions Ltd, Thika

5.2.6 Melt Contamination

The quality of the melt before pouring is important in ensuring production of quality castings. In light of this, Dispinar (2005) reported that extensive efforts have been made to produce high quality metal from various alloy systems to meet the increasing demand for reliable mechanical behaviour. Dispinar identified three important features that define metal cleanliness and these are; removal of inclusions, reduction of dissolved gas and control of trace elements. Inclusions in aluminium alloys act as stress-raisers that cause premature failure of components. Furthermore, dissolved gas leads to porosity in castings that significantly affect mechanical properties of aluminium castings. In addition, presence of porosities makes the properties of castings unpredictable. Alloying elements can significantly impact on the solidification sequence of the melt as well as the properties of the final casting [Gowri and Samuel (1994)]. Whereas Dispinar's studies have benefited overseas foundries, the present survey showed that very little is done in local foundries to control the three factors that define melt cleanliness. This is partly due to lack of equipment for gas and inclusion measurement in addition to poor melt handling procedures. As a result, the quality of the melts from these foundries becomes compromised to a great extent. In fact, the local casting community hardly achieves clean melts of consistent and acceptable quality. This in turn makes castings from the foundries unreliable and of poor quality. A few foundries have however, recognized the significance of melt cleanliness and use degassers (Figure 5.9) to minimize the level of dissolved gas. A few others use fluxes to aid in removal of slag and other inclusions. Since castings employ liquid metal to give the shape of the object directly, the quality factor of the melt becomes crucial in production of castings of reliable quality.

The present survey further revealed that some level of contamination of the melt can be caused by the fuel used. In the case of the cupola furnace, some ash is produced from the coke during the melting process. This ash level is usually not predetermined before hand, and can be absorbed into the melt. This in turn interferes with the chemical composition of the melt especially when the ash content is in excess of 1.5% of the total weight of the coke. For crucible furnaces that utilize industrial oil as the fuel, improper loading of the furnace may allow some of the injected fuel to contaminate the melt thus interfering with the quality of the metal.

5.2.7 Quality Control

In order to produce reliable and sound castings of premium quality, local foundries must be equipped with modern facilities dedicated to moulding, casting, testing and inspection. This is because there are process variables that require measurement and inspection as casting progresses and availability of such equipment becomes crucial. Due to lack of vital facilities, castings produced in the foundries are hardly tested and are frequently hard, brittle, and suffer shrinkage porosities and blowholes and hence do not meet the required standards. Furthermore, many of the foundries visited had no quality control department. However, Booth Extrusions Ltd and East African Foundry Works (K) Ltd have a quality section committed to evaluation of the soundness of their products. It was further noted that the local foundries lack facilities for controlling metallurgical variables such as alloy composition and microstructure of cast compounds. Figure 5.13 shows how acute the shortage of equipment is to the companies. In fact, a spectrometer was only found in two firms (i.e. Booth Extrusions Ltd and Kalu Works Ltd).

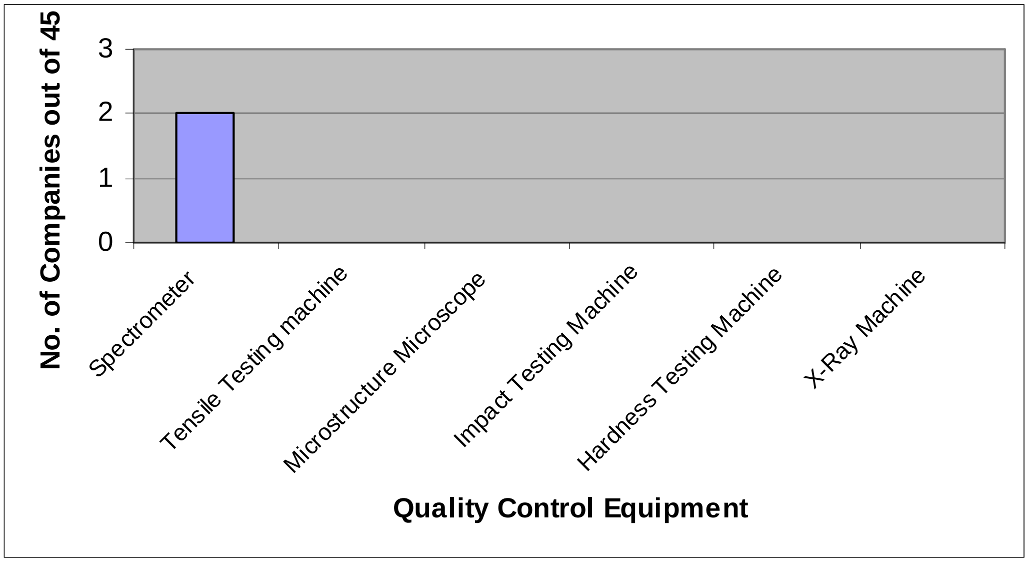


Figure 5.13: Quality Control Equipment available in the local foundries

Other facilities that were noted to be lacking include those that control mould variables such as sand composition, grain size and shape, clay type and content, moisture content, mould hardness and permeability. The problem of lack of facilities becomes more pronounced when it comes to the process of reconditioning of sand for further use after each casting and collapsing of the sand moulds. The skills and experience of the workers remains the only tool depended upon to achieve the required sand mixing and conditioning. Figures 5.14 to 5.16 show some of the casting defects that occur due to poor control of process parameters.

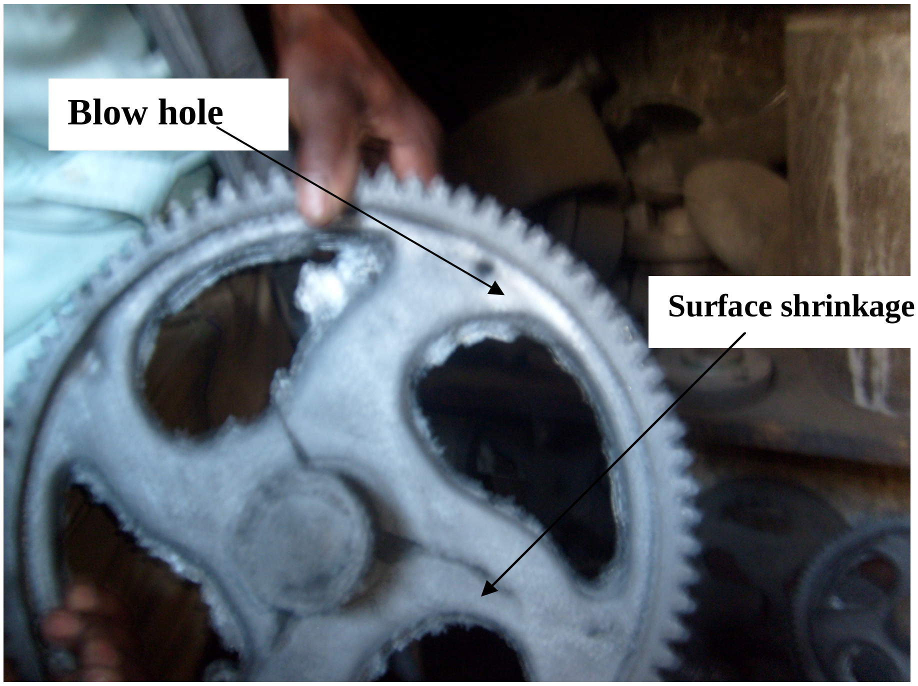


Figure 5.14: **Casting reject with a blow hole and surface shrinkage (cast iron): Courtesy of Premier Bag & Cordage Co. Ltd, Juja**

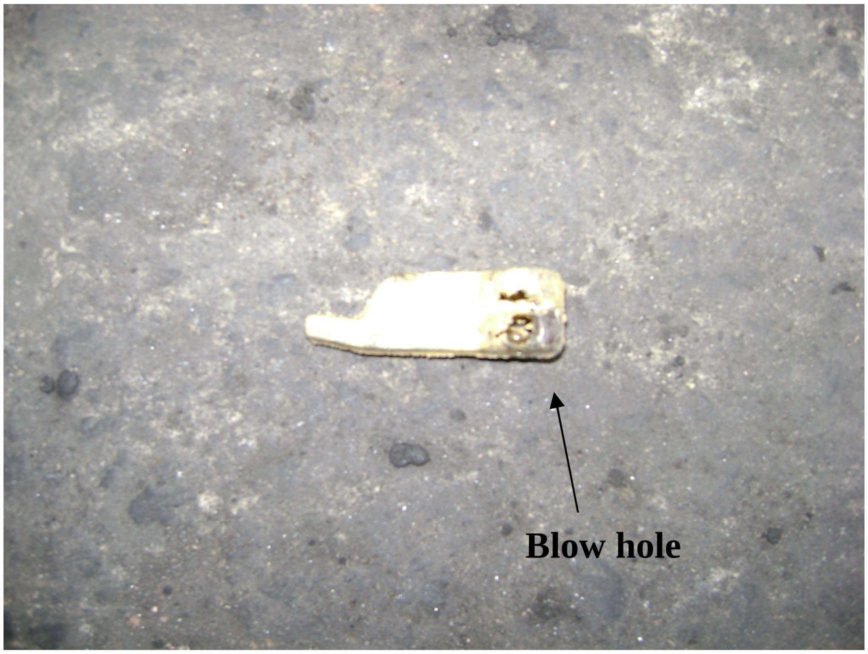


Figure 5.15: **Casting reject with a blow hole (brass): Courtesy of Kens Metal Industries Ltd, Nairobi**

Figure 5.16: Casting reject with a blow hole (aluminium): Courtesy of Kens Metal Industries Ltd, Nairobi

5.2.8 Inspection

Inspection was noted to be carried out only at the end of the casting process to check for any physical defects. This kind of approach of inspection of castings only checks for visible defects and does not take into account existence of internal defects. In addition, this inspection technique does not undertake to provide preventive measures to minimize future occurrence of the defects. So as to achieve castings of acceptable quality, inspection of internal defects will be necessary. This is in line with the studies by Fox and Campbell (2000) who noted that internal imperfections in aluminium alloy castings are potential sources of failure and if reliable castings are to be consistently produced by the foundry then efforts must be made to fully understand failure mechanisms associated with internal defects. For local foundries, this will require availing inspection facilities that permit examination of internal defects. In addition, process parameters must be closely monitored guided by research to ensure production of premium castings.

5.2.9 Labour

The study showed that the knowledge and experience of the technical and management staff in most foundries was inadequate. Moreover, the foundry personnel are not well trained to control process variables and also to cope with new developments in advancing foundry technology. Process variables are controlled using working experience and as such the quality levels are barely adequate and are only accepted because there is no alternative. Campbell (2003) noted that the potential of introducing defects into castings begins primarily during alloy preparation which lies in the hands of workers. Unless the foundry workers carry out alloy preparations well and pouring operations quiescently, great damage can occur to the melt leading to formation of oxide films that end up being incorporated into melt during transfer and mould filling. This in turn impacts negatively on the properties and reliability of castings.

It was regrettably observed that formal training in local institutions of learning on founding skills has not been intensive. The employer needing founding expertise is therefore expected to instill basic skills in moulding, pattern making and pouring procedures through experienced trainers who acquired their skills from others. This system of learning plays a significant role in skill formation as noted from Figure 5.17. Furthermore, it is observed from the diagram that the workers with College and University level of education are few. In addition, the foundry skills gained from such institutions are limited. It is therefore to be noted that an understanding of the fundamental issues in casting technology in terms of metal flow, solidification sequence and changes in metal structure due to alloy variation, causes of casting defects and knowledge on analysis of metal scrap will contribute significantly to the competitive advantage of the casting process in the local foundries. Instead of trial and error methods that most foundries rely on, the foundry workforce will now rely on both knowledge and expertise leading to significant reduction of the historical casting problems. In this way the competitive advantage enjoyed by the overseas companies will be realized. It is the lack of this understanding that has continued to make local foundries record very few industrial products as shown in Figure 5.18.

As the foundries seek to develop their workforce through inculcation of necessary foundry skills to produce quality products, it was felt that research and training institutions should also take recognition of the pivotal role they need to play in offering foundry training to learners taking courses in Foundry Technology and Material sciences.

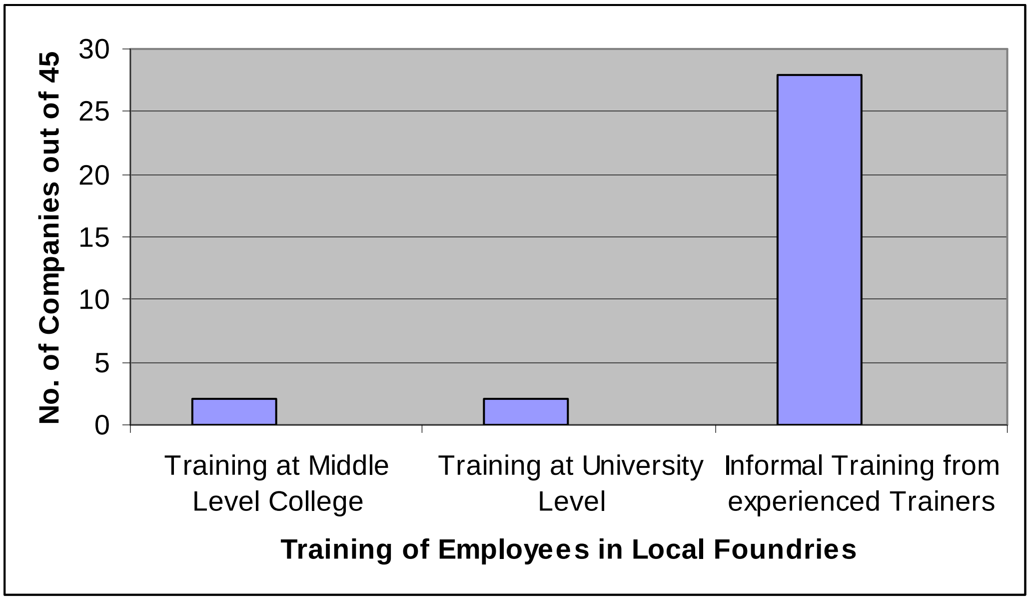


Figure 5.17: Training of employees in local foundries

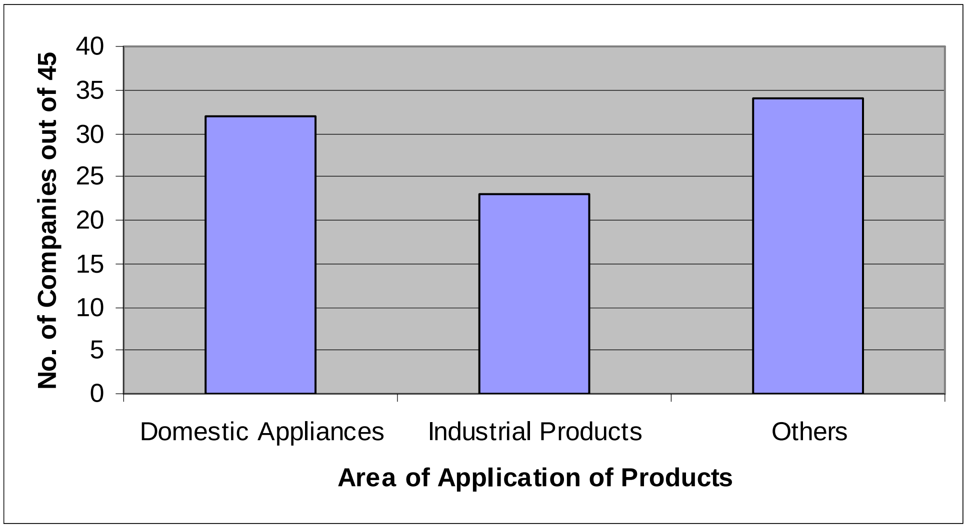


Figure: 5.18: Category of products cast in local foundries

5.3 Non Technical Challenges

5.3.1 Liberalization of Economy

Economic liberalization is a broad term that usually refers to fewer government regulations and restrictions in the economy in exchange for greater participation of private entities. In Kenya, it was felt that private management, whether local or foreign would inject the long craved for efficiency, increased productivity and economic growth that would trickle-down to benefit all and epecially the rural poor. It was further thought that economic liberalization would make business be conducted more professionally devoid of corruption and that there will be market freedom and fairness.

Unfortunately, with the opening up of our market, the country has become a supermarket of foreign goods which are not only cheap but also of poor quality. This has in turn killed the local industries, rendering many workers jobless. Greatly affected as a result of liberalization is the foundry industry. Players in the foundry industry have continued to complain and lament on the adverse effects of a liberalized economy that came with the removal of tariffs imposed on imported products. An influx into the local market of cheaper foundry products of poor quality has been recorded in the past years of operation of a liberalized economy. Consequently, the demand for locally cast products has drastically declined. Many firms have ended up compromising the quality of their cast products as they seek to cut down on production costs. In addition, many others have opted to pull out citing lack of market for their products. In fact, during the present survey, a number of foundries were found to have pulled out of foundry as a core business and these include; Settlers Engineering and Rift Valley Engineering in Nakuru, Rubani Engineers and Kenya Castings Ltd in Nairobi. Others including Satel Engineers indicated intention to pull out unless the government moves fast to protect the domestic industries.

5.3.2 Cost of Raw Materials

From the oral interview it was observed by a number of foundries that the scrap metal which is the major raw material has continued to get a bigger market share abroad. The local foundries are therefore faced with the problem of high cost of scrap as a result of exportation as seen from Figure 5.19. The demand for scrap outside Kenya has made the raw material to be insufficient to satisfy the local industry. The majority of the foundrymen are therefore forced to search vigorously for the raw material before they can produce castings. This accounts for instability of the prices of the products from the industry. It was further noted that the suitability of locally available sand in casting operations especially in shell moulding is in doubt because it is neither graded nor properly mixed with additives to address foundry needs. Ibitoye and Ilori (1998) observed that in order to produce defect free castings, moulding sand must possess proper refractory based particles such as silica as well as binders and additives of organic and inorganic materials. Most foundrymen engaged in shell moulding operations that require specially graded sand are forced to import the sand for their applications. It was also noted that the coke used in the local casting industry especially for coke fired furnaces is imported. Unfortunately, government duties and taxes levied on imported goods make them costly and unaffordable to most foundry entrepreneurs. The cost of electricity and oil (for oil fired furnaces) has greatly increased in the past years further making it unaffordable to most foundries. Overall, with increased cost of raw materials and energy, the local foundries continue to struggle to meet the high cost of production. To make the cost of production affordable many foundries are forced to compromise the quality of their products so as to remain in market.

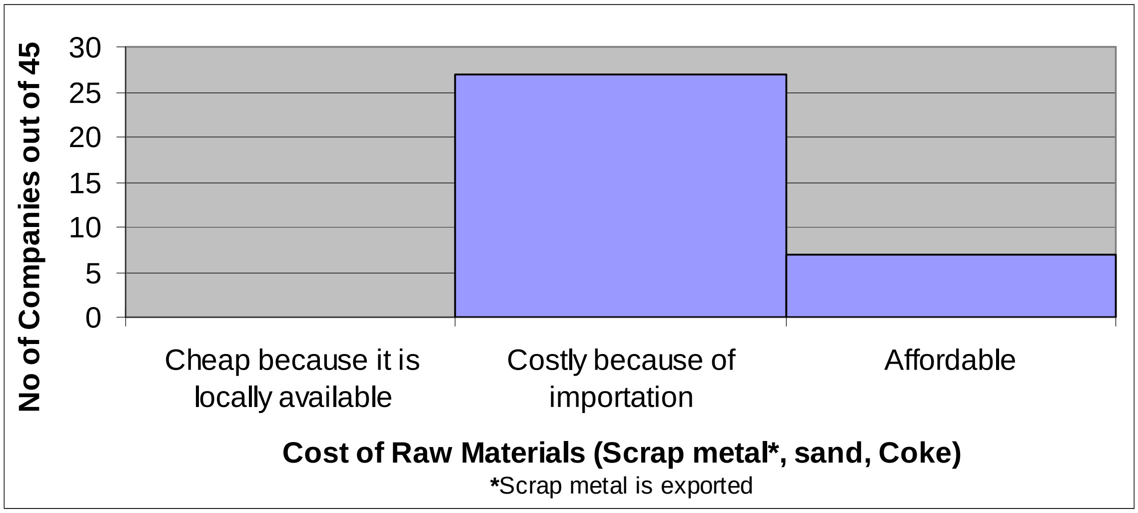


Figure 5.19: Foundry responses on escalating costs of raw materials

5.3.3 Cooperation

It was observed that there was no cooperation among the foundries; to share ideas and compare notes on how best to improve the operations in the trade. At the same time, it was noted that the employees were not keeping themselves abreast with recent technological advancement and discoveries through refresher courses, seminars and further training. This problem has a major influence on the quality of the casting products since the workforce play a significant role in production of quality castings. It is felt that the management should reconsider investing heavily on the training of the workforce and organizing seminars with other firms engaged in foundry operations.

It is recommended that institutions of higher learning and major foundry industries can be identified for training, dissemination of knowledge and research. Furthermore, overseas institutions can also be identified as sources of information and for collaboration. One notable contact that our University has is the London & Scandinavian Metallurgical Company in the UK renown for its production and supply of high quality alloying elements. In addition, they have a well equipped metallurgical lab with an Inductively Coupled Plasma (ICP) Spectrometer for analysis of the chemical composition of aluminium samples. The anticipated collaboration may further allow dispatch of expertise to train local foundry operators on selected skills and subjects of particular interest. Regional workshops can also be organized with a view to provide forum where foundry practitioners drawn from Kenya and other regions can meet and compare notes on new technological developments in the trade.

5.3.4 Quality Management

It is recognized that in any business undertaking in the world today, quality management concepts are increasingly being used to make the production process more efficient, while at the same time increasing the effectiveness of processes to meet the customer's expectations. In doing this, efforts have been made by production and manufacturing enterprises to understand the market, innovate products and processes while providing good customer service. For quality improvement to be successful every member of the organization should have a clear understanding of customer requirements. To maintain this understanding there has to be a constant interaction with the customer and measurement of product against his expectations should be a continuously ongoing process.

Unfortunately the local casting community has continued to use the traditional models, where each worker has a certain work quota to meet on a daily basis. The managers and other workforce are compelled to focus more on completing their work quota than on the quality of work done. Consequently, while completion of work quotas is addressed, the concern for customers and the organizational goals are overshadowed. This in turn, has impacted negatively on the quality of foundry products. In order to reap the benefits of quality management, the need to establish an organizational mission statement is inevitable. Furthermore, simple histograms and control charts could be applied to indicate trends of occurrence of defects in castings and failures of components with a view to prevent them from occurring in future. It was established that only a handful of local foundries including Booth Extrusions Ltd, East African Foundry Works (K) Ltd and Kens Metal Industries Ltd had put in place the aspect of quality management; an initiative that has resulted in achievement of some degree of quality in their products.

Due to failure on the part of most casting industries to put in place the aspect of quality, the casting practice has continued to be done without proper work ethics and mission statement. Among the workers interviewed, it was noted that the majority of them were only concerned with completion of tasks given them without regard of the quality of their work. It was further observed that there was no effective, coherent and sincere way of dissemination of information throughout the organizational structure and with other organizations pursuing the same interests. This served to aggravate the already worse situation of the local foundries.

5.3.5 Poor Work Conditions

It was generally noted that the working conditions especially for the workers involved in ramming, metal sorting, metal breaking were very poor. No safety gears such as gloves, helmets, safety shoes, masks and dust coats were provided. Poor pay was also confessed. Furthermore, the building structures and space provided for conducting foundry activities were found to be inappropriate. As a consequence of these poor working conditions, the morale of workers is stifled leading to poor quality castings.

CHAPTER SIX

6.0 CONCLUSIONS AND RECOMMENDATIONS

6.1 Conclusions

A study was carried out to determine the effect of minor elements (Fe, Mn, Cr, Sr, Sb and Ti-B based grain refiner) on castability (fluidity and porosity formation), microstructure and mechanical properties (UTS, percent elongation, hardness and impact strength) of LM13, LM25 and LM27-type alloys. In addition, a survey of the local foundries was conducted to establish the prevailing aluminium casting practices such as recycling methods, casting design, melt treatment and quality control measures adopted in these foundries to ensure production of premium quality products. Based on the results obtained and their analysis, the following conclusions are drawn.

6.2 Fluidity

There was no change in fluidity when the Fe level was increased from 0.14 to 0.4% in LM25, but it increased by 21% when Fe was raised to the critical content of 0.48%. Furthermore, a decrease of 32% resulted when the Fe level was increased to 0.6%. A combination of 0.3%Mn or 0.6%Cr with 0.6%Fe in LM25, resulted in a fluidity increase of 13 and 8%, respectively compared to the base alloy, but a combination of 0.6%Fe, 0.3%Mn and 0.2%Cr decreased the fluidity by 9%. A 34% increase in fluidity was observed when the Fe content in LM27 was raised from 0.41% to the critical level of 0.6%Fe with a further increase when Mn was raised to 0.3%Mn.

Increasing the Fe content to 1% in LM27 led to a drop in fluidity of 9%. Addition of 0.015%Sr and 0.02%Sr increased the fluidity of LM25 and LM27 by 9% and 21% respectively. Furthermore, a 0.28% Al-5Ti-1B grain refiner addition decreased the fluidity of LM 25 and LM27 by 2% and 19% respectively while a combined addition of 0.02%Sr and 0.28%Al-5Ti-1B decreased the fluidity of LM27 by 8%.

6.3 Porosity Formation

The volume percent porosity in the base alloy was 1.6%. With Sr addition at levels of 0.02% and 0.05%, the volume percent porosity increased to 2.9% and 2.6% respectively. On the other hand, when individual additions of 0.53%Mn and 1.06%Cr were made to the alloy, the volume percent porosity reduced to 0.6% and 0.4% respectively. It was further noted that a combined addition of 0.3%Mn together with 0.2%Cr reduced the volume percent porosity of this alloy to 1.0% while addition of 0.2%Sb decreased it to 1.2%. Other reductions in porosity of 1.1%, 1.2% and 0.8% were observed when 0.53%Mn was added in combination with 0.02%Sr, 0.05%Sr and 0.2%Sb respectively. Addition of 0.02%Sr together with 0.28%Al-5Ti-1B grain refiner slightly reduced the volume percent porosity of the base alloy to 1.4%.

6.4 Microstructure

Individual and combined additions of Sb, Mn, Cr, Sr and Al-5Ti-1B grain refiner on the microstructure of a high Fe secondary LM13 type piston alloy indicated that alloying elements impart significant influences on the microstructural features of the alloy. With no additions, the LM13 base alloy consists mainly of a structure with coarse Si particles, large Cu and Ni containing phases in addition to Al_2Cu phases. When 0.02%Sr or 0.05%Sr was added to the LM13 alloy, the acicular Si particles were modified to a fibrous form but the level of modification can either be partial or full depending on the amount of Sr. Furthermore, a 0.2%Sb addition causes refinement of the Si particles to a lamellar structure. When individual additions of 0.53%Mn and 1.06%Cr were made to the alloy, the large flake-like Cu and Ni bearing phases changed to Chinese script phases. A similar observation was made when 0.3%Mn was combined with 0.2%Cr. Moreover, when 0.53%Mn was added in combination with 0.02%Sr, 0.05%Sr and 0.2%Sb changes occurred to both the Si particles as well as the AlCuNi phases. With addition of 0.02%Sr and 0.28%Al-5Ti-1B the morphology of Si particles changed to fibrous form. It is interesting to note that after T6 heat treatment the Si particles are seen to spheroidize while at the same time AlCuNi phases fragment and the Al_2Cu phases dissolve. These changes in turn improve the mechanical performance of the alloy.

6.5 Mechanical Properties

6.5.1 UTS and Percent Elongation

The results in this section indicated that the average values of the UTS and % elongation of the LM13 base alloy were 140 MPa and 2.3% respectively in as cast condition while in T6 condition the values were 223 MPa and 2.1%. With addition of Sr at levels of 0.02% and 0.05% the average values of UTS increased by 4% and 5%

respectively in as cast while it increased by 6% and 8% in T6. Moreover, with the 0.02%Sr and 0.05%Sr addition the % elongation increased by 22% and 17% in as cast and by 24% for each level of Sr in T6. When individual additions of 0.53%Mn and 1.06%Cr were made, improvements of 13% and 18% in the UTS were recorded in as cast condition while in T6 condition the improvements were 15% and 16%. With these additions, the % elongation of the alloy increased by 35% and 61% in as cast and by 38% and 43% in T6. A combined addition of 0.3%Mn and 0.2%Cr was noted to increase the UTS and % elongation of this alloy by 12% and 30% in as cast and by 12% and 33% in T6. With addition of 0.2% Sb the UTS and % elongation of the alloy increased by 4% and 30% respectively in as cast condition and by 10% and 17% in T6 condition. It was noted that combining 0.02%Sr with 0.28% Al-5Ti-1B decreases the UTS by 3% in as cast condition. However, % elongation in as cast increased by 17%. On the other hand, the T6 counterpart of this alloy indicated a 3% increase in UTS and a 9% increase in % elongation. When each of 0.02%Sr, 0.05%Sr and 0.2%Sb were combined with 0.53%Mn the UTS values reduced by 3%, 1% and 3% respectively in as cast condition while in T6 condition, the values increased by 1%, 2% and 3% respectively.

6.5.2 Brinell hardness

The influence of individual and combined additions of Sb, Mn, Cr, Sr and Al-5Ti-1B grain refiner on the Brinell hardness of a high Fe secondary LM13 type piston alloy indicated that addition of 0.02%Sr and 0.05%Sr to the alloy increased the Brinell hardness by 3% and 4% respectively in as cast condition and by 2% and 5% in T6 condition. On the other hand, when 0.53%Mn and 1.06%Cr were separately added to the alloy, improvements in Brinell hardness of 6% and 9% respectively were recorded in as cast condition while in T6 condition the improvements were 11% and 16%. It was further observed that a combined addition of 0.3%Mn and 0.2%Cr increased the hardness of this alloy by 8% in as cast condition and by 13% in T6 condition. When each of 0.2%Sb, 0.02%Sr and 0.05% Sr was separately combined with 0.53%Mn, improvements in Brinell hardness of 3%, 1% and 3% respectively were recorded in as cast condition. In T6 condition however, the improvement was 8% when 0.2%Sb was combined with 0.53%Mn and 4% when each level of Sr (0.02% and 0.05%) was combined with 0.53%Mn. No improvement in hardness was recorded when 0.2%Sb was added to LM13 alloy in as cast condition. With a T6 heat treatment, Sb addition improved the hardness by 6%. There was a reduction in brinell hardness of 4% in as cast alloy when 0.02%Sr was combined with 0.28%Al-5Ti-1B. However, in T6 condition an increase of 1% was observed.

6.5.3 Impact Energy

It was observed that addition of 0.02%Sr and 0.05%Sr to the LM13 base alloy increased the impact strength by 7% and 14% respectively in as cast condition and by 6% and 12% in T6 condition. On the other hand, addition of 0.2%Sb increased the impact energy by 21% in as cast condition and by 18% in T6 condition. When separate additions of 1.06%Cr and 0.53%Mn were made to the alloy, the impact strength increased by 96% and 67% respectively in as cast condition and by 75% and 49% in T6 condition. At the same time, a combined addition of 0.2%Cr and 0.3%Mn increased the impact energy by 7% in as cast condition and by 18% in T6 condition. When 0.53%Mn was combined with either 0.02%Sr or 0.05%Sr the impact energy was observed to increase by 28% in each of the as cast alloy and by 12% in each of the T6 alloys. Combining 0.2%Sb with 0.53%Mn on the other hand, increased the impact energy by 28% in as cast condition and by 18% in T6 condition.

6.6 The Survey

Results from the survey indicated that nearly all the foundries visited operate at about 40% capacity utilization. In addition, control of process parameters was limited, with methods being non competitive and quality control aspects being hardly adequate. Use of alloying elements (additives) as a means to improve properties of castings was seldom utilized, leave alone knowledge of existence of such additives. Out of 45 companies visited 98% were not using additives to control the properties of castings. This in turn resulted in production of low quality and unreliable castings. Noting that some of the companies visited supply their products to the international market, the majority would find themselves disadvantaged if it turns out that their products need to meet certain standards that call for use of additives. It was therefore concluded that collaborative effort be made between research institutions and industry as well as the government as one of the basic steps towards improvement of the casting practice and enhancement of capacity utilization in local foundries.

6.7 Recommendations

On the basis of the results obtained from the study of the effect of minor elements on castability, microstructure and mechanical properties of recycled aluminium alloys and the survey of the local foundries, the following recommendations are made;

1. That the method of effective recycling of scrap obtained from scrap vendors is to sort the scrap into groups of similar components namely pistons, cylinder heads and wheels so that the resulting alloys can be closely equivalent to the commercial alloys used to make the components.
2. That depending on the application and properties required, minor alloying elements be used in commercial Al-Si alloys to enhance their mechanical performance.
3. Fatigue testing to determine the effect of minor elements on the fatigue performance of the alloys.
4. High temperature performance tests of the alloys would be important since most alloys like the LM13 and LM27 are usually operated at high temperatures of between 250 to 400˚C.
5. That the knowledge gained from this deliberate and detailed critical investigation of the state of foundry in Kenya is important and should be used as a basis to improve the current state of our local foundries to be able to compete favourably with other foundries in the world in producing premium quality products. Efforts must be stepped up to take the level of technology a

notch higher. The quality control measures adopted by local foundrymen must be attended to in order to produce premium high quality products.

REFERENCES

Abdollahi A. and Gruzleski J. E., "An Evaluation of Ca as Eutectic Modifier in A357 Alloy," *International Journal of Cast Metals Research,* Vol. 11, pp. 145-155, (1998).

Argo, D. and Gruzleski, J. E., "Porosity in Modified Aluminium Alloy Castings," *AFS Transactions,* Vol. 96, pp. 65 – 74, (1988).

ASTM Standards, Vol. 02.02. "Standard Methods of Testing Wrought and Cast Aluminium Alloy Products," 1989.

Apelian D., Sigworth G. K. and Whaler K. R., "Assessment of Grain Refining and Modification of Al-Si Foundry Alloys by Thermal Analysis," *AFS Transactions,* Vol. 92, pp. 297-307, (1984).

Ashtari P., Tezuka H. and Sato T., "Modification of Fe-Containing Intermetallic Compounds in an Al-Si-Cu-Fe Cast alloy Using Sr, Li, and Mn Additions," *Materials Forum,* Vol. 28, pp. 951-955, (2004).

Bangyikhan K., "Effect of Oxide Films, Fe-rich Phase, Porosity and their Interactions on Tensile Properties of Cast Al-Si-Mg alloys," PhD Thesis, Submitted to the School of Metallurgy and Materials, Faculty of Engineering of the University of Birmingham, Birmingham B15 2TT, UK, (2005).

Backerud L, Chai G. and Tamminen J., "Solidification Characteristics of Aluminium Alloys," Vol. 2, Foundry Alloys, Des Plaines, IL: *American Foundrymen's Society,* (1990).

Badia F. A., "Effect of Composition and Heat Treatment on the Tensile Properties of an Al-11.5Si-5Ni Casting Alloy," *Modern Castings*, Vol. 50, pp 104-115 (1966).

Barresi, J., Kerr, M. J., Wang, H. and Couper, M. J., "Effect of Mg, Iron and Cooling Rate on Mechanical Properties of Al-7Si-Mg Foundry Alloys" *AFS Transactions*, Vol. 117, pp. 563-570, (2000).

Bien X., Zhang Z. and Liu X., "Effect of Strontium Modification on Hydrogen Content and Porosity Shape of Al-Si Alloys," *Materials Science Forum*, Vols. 331-337, pp. 361-366, (2000).

Bonsack W., "Discussion on the Effect of Minor Alloying Elements on Aluminium Casting Alloys," *AST Bulletin*, Vol. 117, pp. 45-59, (1942).

Brown, J.R., (1999): "Foseco Non-ferrous Foundryman's Handbook", Butterworth-Heinemann Ltd, 11th Edition, Oxford, (1999).

Caceres C. H., Davidson C. J., Griffiths J. R., Hogan L. M. and Wang Q. G., "Hypoeutectic Al-Si-Mg Foundry Alloys," *Materials Forum*, Vol. 21, pp. 27-43, (1997).

Caceres C. H., Svensson I. L. and Taylor J. A., "Strength-Ductility Behaviour of Al-Si-Cu-Mg Casting Alloys in T6 Temper," *International Journal of Cast Metals Research*, Vol. 15, No. 5, pp. 531-543, (2003).

Campbell J., Castings, 2nd Edition, (Butterworths-Heinnman Ltd), (2003).

Campbell J., "The Properties of Cast Metals," Presented at 66th World Foundry Congress, Istanbul, Turkey, (2004).

Cao X. and Campbell J., "The Nucleation of Fe-Rich Phases on Oxide Films in Al-11.5Si-0.4Mg Cast Alloys," *Metallurgical and Materials Transactions A*, Vol. 34A, pp. 1409-1420, (2003).

Cao X. and Campbell J., "Morphology of β-Al_5FeSi Phase in Al-Si Cast Alloys," *Materials Transactions,* Vol. 47, N0. 5, pp. 1303-1312, (2006).

Chai G., "Dendrite Coherency During Equiaxed Solidification in Aluminium Alloys." PhD Thesis, Department of Structural Chemistry, Stockholm University, Sweden, (1994).

Cho Y. H., Lee H. C., Oh K. H. and Dahle A. K., "Effect of Strontium and Phosphorus on the Eutectic Al-Si Nucleation and Formation of β-Al_5FeSi in Hypoeutectic Al-Si Foundry Alloys," *Metallurgical and Materials Transactions A,* Vol. 39A, pp. 2435-2448, (2008).

Cho Y. H., Joo D. H., Kim C. H. and Lee H. C., "The Effect of Alloy Addition on the High Temperature Properties of Over-aged Al-Si(CuNiMg) Cast Alloys," *Materials Science Forum,* Vols. 519-521, pp. 461-466, (2006).

Closset B. M., "Modification and Quality of Low Pressure Al-Si Castings," AFS Transactions, Vol. 96, pp. 249-260, (1988).

Couper, M. J., Smith, C. L., Easton, M. A., Barresi, J. and Zhang, X., "The Effect of Ti Content on Microstructure, Castability and Properties of Foundry Alloys A356.2," *TMS Letters,* pp. 49-50, Vol. 2, Issue 2, (2005).

Couture, A. "Iron in Aluminium Casting Alloys- A Literature Survey," *AFS International Cast Metals Journal,* Vol. J6 9, pp. 9-17, (1981).

Crepeau P. N., "Effect of Iron in Al-Si Casting Alloys: A Critical Review," *AFS Transactions,* Vol. 103, pp. 361-366, (1995).

Dahle A. K., Tondel P. A., Paradies C. J. and Arnberg L., "Effect of Grain Refinement on the Fluidity of Two Commercial Al-Si Foundry Alloys," *Metallurgical and Materials Transactions A,* Vol. 27A, pp. 2305-2313, (1996).

Dahle, A.K. and Arnberg, L., "The Effect of Grain Refinement on the Fluidity of Aluminum Alloys," *Materials Science Forum*, Vols. 217-222, pp. 259-264 (1996).

Dahle, A.K., Arnberg, L. and Apelian, D., "Burst Feeding and its Role in Porosity Formation during Solidification of Al Foundry Alloys", *AFS Transactions*, Vol. 105, pp. 963-970 (1997).

Dahle, A.K., J.A. Taylor and D.A. Graham, "The Role of Eutectic Growth Mode in Porosity Formation in Al-Si Alloys", *Aluminum Transactions,* Vol. 3 (1), pp. 17-30, (2000).

Dahle, A.K., S.D. McDonald and K. Nogita, "Eutectic Solidification in Hypoeutectic Al-Si Alloys and its Effect on Porosity", Proc. 2nd Int. Aluminum Casting Technology Symposium, 7-9 October, Columbus, OH (ASM International, 2002), pp. 1-10.

Dahle A. K., Nogita K., McDonald S. D., Dinnis C. and Lu L., "Eutectic Modification and Microstructure Development in Al-Si alloys," *Materials Science and Engineering A,* Vols. 413-414, pp. 243-248, (2005) Dave, B., Couper, P., David, H. S. and Dahle, A. K., "A Comparison of Grain Refiner Master Alloys for the Foundry," *131st Annual Meeting and Exhibition, CAST*, pp. 1-7, (2002).

Denton, J.R. and J.A. Spittle, "Solidification and Susceptibility to Hydrogen Absorption of Al-Si Alloys Containing Strontium", Materials Science and Technology, 1, pp. 305-311, (1985).

Dinnis C. M., Taylor, J. A., Dahle A. K., "Neutralization of Iron Related Porosity in Al-Si Foundry Alloys," *Proceedings of the Light Metals Technology Conference*, pp. 1-4, (2003).

Dinnis C. M., Taylor J. A. and Dahle A. K., "Porosity Formation and Eutectic Growth of Al-Si-Cu-Mg Alloys," *Materials Forum,* Vol. 28, Edited by J. F. Nie, A. J. Morton and B. C. Muddle, pp. 1016-1021, (2004a).

Dinnis C. M., Otte M. O., Dahle A. K. and Taylor J. A., "The Influence of Strontium on Porosity Formation in Al-Si Alloys," *Metallurgical and Materials Transactions A*, Vol. 35A, pp. 3531-3541, (2004b).

Dinnis C. M., Taylor, J. A., Dahle A. K., "Interactions between Iron, Manganese and the Al-Si Eutectic in Hypoeutectic Al-Si alloys," *Metallurgical and Materials Transactions A,* Vol. 37A, pp.3282-3291, (2006).

Di Sabatino M., "Fluidity of Aluminium Foundry alloys," PhD Thesis Submitted to the Norwegian University of Science and Technology, (2005).

Di Sabatino, M. and Arnberg, L., “Effect of Grain Refinement and Dissolved Hydrogen on the Fluidity of A356 Alloy,” *International Journal of Cast Metals Research*, Vol. 18, No. 3, pp. 181-186, (2005).

Di Sabatino M., Syvertsen F., Arnberg L. and Nordmark A, “An Improved Method for Fluidity Measurements by Gravity Casting of Spirals in Sand Moulds,” *International Journal of Cast Metals Research,* Vol. 18, No. 1, pp. 59-62, (2005).

Dispinar D., “Determination of Metal Quality of Aluminium and its Alloys,” PhD Thesis Submitted to The University of Birmingham, (2005).

Dons A. L., Heiberg G., Voje J., Maeland J. S. Loland J. O. and Prestmo A., “On the Effect of Additions of Cu and Mg on the Ductility of Al-Si Foundry Alloys Cast with a Cooling Rate of Approximately 3 K/s,” *Materials Science and Engineering A,* Vols. 413-414, pp. 561-566, (2005).

Easton M. A. and StJohn D. H., “The Effect of Grain Refinement on the Formation of Casting Defects in Alloy A356 Casting,*” International Journal of Cast Metals Research.*, Vol. 12, pp. 393-408, (2000).

Emadi, D., Gruzleski, J.E. and Toguri, J.M., “The Effect of Na and Sr Modification on Surface Tension and Volumetric Shrinkage of A356 Alloy and their Influence on porosity Formation,” *Metallurgical Transactions B*, Vol. 24B, pp. 1053-1063 (1993).

Fasoyinu, F. A., “Effect of Mg Concentration on Drossing Tendency, Inclusion Formation, Casting Fluidity, Hydrogen Pickup and Mechanical Properties from Gravity Die-cast Alloy A390.0,” *AFS Transactions,* Vol. 102, pp.515-528, (1994).

Flemings M. C., Niyama E. and Taylor H. F., “Fluidity of Aluminium: An Experimental and Quantitative Evaluation,” *AFS Transactions,* Vol. 69, pp. 566-576, (1961).

Fox S. and Campbell J., “Visualization of Oxide Film Defects During Solidification of Aluminium Alloys,” *Scripta Materialia,* Vol. 43, pp. 881-886, (2000).

Garcia-Hinojosa J. A., Gonzalez C. R., Gonzalez G. M. and Houbaert Y., “Structure and Properties of Al-7Si-Ni and Al-7Si-Cu Cast Alloys Non-modified and Modified with Sr,” *Journal of Materials Processing,* Vols.143-144, pp. 306-310, (2003).

Gowri, S. and Samuel, F. H., “Effect of Alloying Elements on the Solidification Characteristics and Microstructure of Al-Si-Cu-Mg-Fe 380 Alloy,” *Metal Transactions A,* Vol. 25A, pp. 437 - 447, (1994).

Gustafsson G., Thorvaldsson T. and Dunlop G. L., “The Influence of Fe and Cr on the Microstructure of Cast Al-Si-mg Alloys,” *Metallurgical Transactions A,* Vol. 17A, pp. 45-52, (1986).

Grand L., “Influence of Some Impurities on the Quality of Aluminium Alloys, *Fonderie,* Vol. 217, pp 95-100, (1964).

Groteke, D. E., “Influence of SNIF Treatment on Characteristics of Aluminium Foundry Alloys,” *AFS Trans,* Vol. 93, pp. 953 - 960, (1985).

Gruzleski J., Handiak N. and Campbell H., “Hydrogen Measurement by Telegas in Strontium Treated A356 Melts,” *AFS Transactions,* pp. 147-154, (1986).

Gruzleski, J. E. and Closset, B., "The Treatment of Liquid Aluminum Silicon Alloys", *American Foundrymen Society, Des plaines, Illinois* (1990).

Hatch, J.E., "Aluminum: Properties and Physical Metallurgy", *ASM, Metals Park, Ohio* (1987).

Hatch J. E., “Aluminium Properties and Physical Metallurgy,” *American Society for Metals,* Metals Park, Ohio, (1990).

Hwang J. Y., Doty H. W. and Kaufman M. J., “Effect of Mn additions on the Microstructure and Mechanical Properties of Al-Si-Cu casting Alloys,” *Materials Science and Engineering A,* Vol. 488, pp. 496-504, (2008).

Ibitoye S. A. and Ilori M. O., “Indigeneous Metal Casting in Nigeria: Its technology, Attendant Problems, Business Prospects and Policy Implications,” *Tecnovan,* Vol. 18, No. 11, pp. 705-711, (1998)

Jenabali S. A. J, Dehghan A. and Malekjani S., “Effects of Optimum amount of Sr and Sb Modifiers on Tensile, Impact and Fatigue Properties of A356 Aluminium Alloy,” *Iranian Journal of Science and Technology, Transaction B,* Vol. 28, No. B2, pp. 225-231, (2004).

Kaufmann H., Fragner W. and Uggowitzer P. J., “Influence of Variations in Alloy Composition on Castability and Process Stability. Part 1: Gravity and Pressure Casting Processes,” *International Journal of Cast Metals Research,* Vol. 18 No. 5, pp. 273-278, (2005).

Kim, J. M. and Loper, C. R., Jr., “Effect of Solidification Mechanism on Fluidity of Al-Si Casting Alloys,” *AFS Transactions,* Vol. 103, pp. 521-529, (1995).

Kim H. J., Kobayashi T. and Ito T., “Effects of Fe and Ca on Usual and Impact Fatigue Characteristics of AC2B-T6 Aluminium Casting Alloys,” *Materials Science Forum,* Vols. 217-222, pp. 1395-1400, (1996).

Kim H. J. “Effect of Calcium on Primary Silicon Particle Size in Hypereutectic Al-Si Alloys,” *Materials Science and Technology,* Vol. 19, pp. 915-918, (2003).

Kim H. Y., Han S. W. and Lee H. M. “The influence of Mn and Cr on the Tensile Properties of A356-0.20Fe alloy,” *Materials Letters,* Vol. 60, pp 1880-1883, (2006).

Kolsgaard A., “Casting Characteristics of SiC Particulate Reinforced Al-7Si-Mg Aluminium Matrix Composites,” PhD Thesis, Norwegian University of Science and Technology (NTNU), (1993).

Kotte B., “Strontium Modification Gives Critical Melt Control,” *Modern Casting,* pp. 33-35, (1985).

Kral M. V., Nakashima P. N. H. and Mitchell D. R. G., “Electron Microscope Studies of Al-Fe-Si Intermetallics in an Al-11Si Alloy,” *Metallurgical and Materials Transactions A,* Vol. 37A, pp. 1987-1997, (2006).

Kwon Young-Dong and Lee Z. H., "The Effect of Grain Refining and Oxide Inclusion on the Fluidity of Al-4.5Cu-0.6Mn and A356 Alloys," *Materials Science and Engineering A,* Vol. 360, pp. 372-376, (2003).

Lang G., "Giessereigenschaften und oberflachenspannung Von Aluminium und Binaren Aluminiumlegierungen, Part I," *Aluminium,* Vol. 48, No. 10, pp. 664-672, (1972).

Lee Y. W., Chang E. and Chieu C. F., "Modelling of Feeding Behaviour of Solidifying Al-7Si-0.3Mg Alloy Plate Casting," *Metallurgical Transactions B,* Vol. 21B, pp. 715-727, (1990).

Li Z., Samuel A. M. Ravindran C, Valtierra S. and Doty H.W., "Parameters Controlling the Performance of AA319-type Alloys, Part I: Tensile Properties," *Materials Science and Engineering A 367,* pp. 96-110, (2004).

Li J. Y., Brusethaug S. and Olsen A., "Influence of Cu on Mechanical Properties and Precipitation Behaviour of Al-7Si-0.5Mg Alloy During Aging Treatment," *Acta Materialia,* Vol. 54, pp. 99-103, (2006).

Liao H. and Sun G., "Influence of Boron on the Microstructure and Mechanical Properties of Al-11.6Si-0.4Mg Casting Alloy Modified with Strontium," *Materials Science and Technology,* Vol. 20, pp. 521-527, (2004).

Loper, C. R., Jr, "Fluidity of Aluminium Silicon Casting alloys," *AFS Transactions,* Vol. 100, pp. 533 – 538, (1992).

Loper, C. R., Jr and Prucha, T. E, "Feed Metal Transfer in Al-Cu-Si alloys," *AFS Transactions,* Vol. 98, pp.845 - 853, (1990).

Ma, Z., Samuel, A. M., Samuel, F. H., Doty, H. W. and Valtierra, S., "Effect of Fe Content and Cooling Rate on the Impact Toughness of Cast 319 and 356 Aluminium Alloys," *AFS Transactions*, pp. 255-265, (2003).

Mahta M., Emamy M., Daman A., Keyvan A. and Campbell J., "Precipitation of Fe-Rich Intermetallics in Cr-and Co-Modified A413 Alloy," *International Journal of Cast Metals Research,* Vol. 18, No. 2, pp. 73-79, (2005).

Manas D. and Makhlouf M., "Effect of Key Alloying Elements on the Feeding Characteristics of Aluminium Silicon Casting Alloys," *Journal of Light Metals*, pp. 251-265, (2002).

Mbuya, T. O., Odera, B. O. and Ng'ang'a, S. P., "Influence of Iron on Castability and Properties of Aluminium Silicon Alloys: A Literature Review," *International Journal of Cast Metals Research*, Vol. 16 (5), pp. 451–465, (2003).

Mbuya, T.O., (2003): "Cast Aluminium Scrap Recycling: Influence of Alloy Chemistry and Process Variables on the Structure and Mechanical Properties," MSc. Thesis, Department of Mechanical Engineering, University of Nairobi, (2003).

Mbuya, T. O., "Element Effects on the Fluidity of Cast Al-Si Alloys," *AFS Transactions,* Vol. 114, pp.163 - 180, (2006).

Mbuya T. O., Oduori M. F., Rading G. O. and Wekesa M. S., "Effect of Runner Design on Mechanical Properties of Permanent Mould Aluminium Castings," *International Journal of Cast Metals Research,* Vol. 19, No. 6, pp. 357-360, (2006).

Mbuya, T.O, Odera, B.O. Ng'ang'a, S.P. and Oduori, M.F., (2007): "Effect of Some Casting Parameters on the Microstructure and Mechanical Properties of Recycled Aluminium Castings of Various Automobile Components", *The Kenya Journal of Mechanical Engineering,* Vol. 3, No. 1, pp. 29-43, (2007).

McDonald S. D., Dahle A. K., Taylor J. A. and StJohn D. H., "Modification-Related Porosity Formation in Hypoeutectic Aluminium-Silicon Alloys," *Metallurgical and Materials Transactions B,* Vol. 25B, pp. 1097-1106, (2004).

Moffat A. J., "Micromechanistic Analysis of Fatigue in Aluminium Silicon Casting Alloys," PhD Thesis submitted to the University of Southampton, Faculty of Engineering, Science & Mathematics, School of Engineering, (2007).

Mohanty P. S., Samuel F. H. and Gruzleski J. E., "Mechanism of Heterogeneous Nucleation of Pores in Metals and Alloys," *Metallurgical Transactions A,* Vol. 24A, pp. 1845-1856, (1993).

Mollard, F. R., Flemmings, M. C. and Niyama, E. F., "Understanding the Aluminium Fluidity: The key to Advanced Cast Products," *AFS Transactions,* Vol. 95, pp. 647-652, (1987).

Mulazimoglu M. H., Zaluska A., Gruzleski J. E. and Paray F., “Electron Microscope Study of Al-Fe-Si Intermetallics in 6201 Aluminium Alloy,” *Metallurgical and Materials Transactions A,* Vol. 27A, pp. 929-936, (1996).

Mondolfo, L. F., “Aluminium Alloys: Structure and Properties,” *London Butterworths,* pp. 282 - 285, 534 - 536, 759 - 774, (1976).

Moustafa, M. A., Samuel F. H., Doty, H. W. and Valtiera, S., “Effect of Mg and Cu Additions on the Microstructural Characteristics and Tensile Properties of Sr-Modified Al-Si Eutectic Alloys,” *International Journal of Cast Metals Research,* Vol.14, pp. 235-253 (2002).

Moustafa M. A., “Effect of iron Content on the Formation of β-Al_5FeSi and Porosity in Al-Si eutectic Alloys,” *Journal of Materials Processing Technology,* Vol. 209, pp. 605-610, (2009).

Murali S., Raman K. S. and Murthy K. S. S., “Effect of Magnesium, Iron (impurity) and Solidification rates on the Fracture toughness of Al-7Si-0.3Mg Casting Alloy,” *Materials Science and Engineering,* Vol. A151, pp. 1-10, (1992).

Murali, S., Raman, K. S., and Murthy, K. S. S., “Morphological Studies on β-$FeSiAl_5$ Phase in Al-7Si-0.3Mg Alloy with Trace addition of Be, Mn, Cr and Co,” *Materials characterization,* Vol. 33, pp. 99-112, (1994).

Murali S., Arvind T. S., Raman K. S. and Murthy K. S. S., “Fatigue Properties of Sand Cast, Stircast and Extruded Al-7Si-0.3Mg Alloy with Trace additions of Be and Mn,” *Materials Transactions, JIM,* Vol. 38, No. 2, pp. 28-36, (1997).

Murty B.S., Kori S. A. and Chakraborty M., "Grain Refinement of Aluminium and its Alloys by Heterogeneous Nucleation and Alloying," *International Materials Reviews*, Vol. 47 No.1, pp. 3 - 29, (2002).

Narayanan L. A., Samuel F. H. and Gruzleski J. E., "Crystallization Behaviour of Iron-Containing Intermetallic Compounds in 319 Aluminium Alloy," *Metallurgical and Materials transactions A*, Vol. 25A, pp. 1761-1773, (1994).

Narayanan L. A., Samuel F. H. and Gruzleski J. E., "Dissolution of Iron Intermetallics in Al-Si alloys Through Non-equilibrium Heat Treatment," *Metallurgical and Materials Transactions A*, Vol. 26A, pp. 2161-2174, (1995).

Nogita K. and Dahle A. K., "Determination of Eutectic Solidification Mode in Sr-Modified Hypoeutectic Al-Si Alloys by ESBD," *Materials Transactions*, Vol. 42, No. 2, pp. 207-214, (2001).

Nogita K., McDonald S. D. Zindel J. W. and Dahle A. K., "Eutectic Solidification Mode in Sodium Modified Al-7Si-3.5Cu-0.2Mg Casting Alloys," *Materials Transactions*, Vol. 42, No. 9, pp. 1981-1986, (2001)

Otte M. O., McDonald S. D., Taylor J. A., StJohn D. H. and Schneider W., "Controlling Porosity-Related Casting Rejects: Understanding the Role of Iron in Al-Si Alloys," *AFS Transactions*, Vol. 112, pp. 471-478, (1999).

Pan, E. N. and Hu, J. F., "Study on Fluidity of Al-Si Alloys," *AFS Transactions*, Vol. 105, pp. 413-418, (1997).

Pan, E. N. and Hu, J. F., “Fluidity of Al-Si-(Cu) Alloys as Affected by Alloy Solidification Modes,” *AFS Transactions*, Vol.106, pp. 609-617 (1998).

Paray F., Kulunk B. and Gruzleski J. E., “Impact properties of Al-Si Foundry Alloys,” *International Journal of Cast Metals Research*, Vol. 13, pp. 17-37, (2000).

Pedersen L and Arnberg L., “The Effect of Solution Heat Treatment and Quenching Rates on Mechanical Properties and Microstructures in AlSiMg Foundry Alloys,” *Metallurgical and Materials Transactions A,* Vol. 32A, pp. 525-532, (2001).

Pfeiffer W. D. and Sabath G., “Effects of Zinc, Iron and Manganese on the Processing and Mechanical Properties of an Al-8Si-3Cu Alloy,’ *Banyaz. Kohasz. Lapok.*, Vol. 37(2), pp. 39-40 (1986). (MITS Translation NF 168)

Portevin, A. and Bastien, P., “Facteurs Principaux de la Coulabilite des Metaux Purs,” *Comptes rendeus,* Vol. 194, pp.599, (1932).

Ravi M., Pillai U. T. S., Pai B. C., Damodaran A. D. and Dwarakadasa E. S., “The Effect of Mischmetal Addition on the Structure and Mechanical Properties of a Cast Al-7Si-0.3Mg Alloy Containing Excess Iron (up to 0.6 Pct),” *Metallurgical and Materials Transactions A,* vol. 33A, pp. 391-400, (2002).

Rezvani M., Yang X. and Campbell J., “Effect of Ingate Design on Strength and Reliability of Al. Castings,” *AFS Transactions,* pp. 181-188, (1999).

Rooy, E. L., “Aluminium Scrap Recycling and its Impact on the Metal Casting Industry,” *AFS Transactions*, Vol. 93, pp. 935 - 938, (1985).

Roy N., Samuel A. M. and Samuel F. H., “Porosity Formation in Al-9Si-3Cu Alloy Systems: Metallographic Observations,” *Metallurgical and Materials Transactions A*, Vol. 27A, pp. 413-429, (1996).

Samuel A. M., Samuel F. H and Doty H. W., “Observation on the Formation of β-Al_5FeSi phase in 319-Type Al-Si alloys,” *Journal of Materials Science*, Vol. 31, pp. 5529-5539, (1996).

Samuel A. M., Pennors A., Villeneuve C., Samuel F. H., Doty H. W. and Valtierra S., *International Journal of Cast Metals Research*, Vol. 13, pp. 231-253, (2000).

Samuel A. M., Samuel F. H., Villeneuve C., Doty H. W. and Valtierra S., “Effect of Trace Elements on β-Al_5FeSi Characteristics, Porosity and Tensile Properties of Al-Si-Cu (319) Cast Alloys,” *International Journal of Cast Metals Research*, Vol. 14, pp. 97-120, (2001).

Shahani H., “Effect of Hydrogen on the Shrinkage Porosity of Aluminium Copper and Aluminium Silicon Alloys,” *Scandinavian Journal of Metallurgy*, Vol. 14, pp. 306-312, (1985).

Sheshradi M. R. and Ramachandran A., Modern Casting, pp. 110-122, (1965) - Cited.

Shabestari S. G., Mahmudi M., Emamy M. and Campbell J., “Effect of Mn and Sr on Intermetallics in Fe-Rich Eutectic Al-Si Alloy,” *International Journal of cast Metals Research*, Vol. 15, pp.17-24, (2002).

Shabestari S. G. and Shahri F., "Influence of Modification, Solidification Conditions and Heat Treatment on the Microstructure and Mechanical Properties of A356 Aluminium Alloy," *Journal of Materials Science,* Vol. 39, pp. 2023-2032, (2004).

Shan-Noon Y., Sheng-Long L., Yu-Huan L and Jing-Chien L., "Mechanical Properties of Al-11%Si Casting Alloys Containing Trace Be and Sr," *Materials Transactions, JIM,* Vol. 40, pp. 294-300, (1999).

Shahrooz N. and Reza Ghomashchi., "Effects of Modification during Conventional and Semi-solid Metal Processing of A356 Al-Si Alloy," *Materials Science and Engineering A,* Vol. 415, pp. 273-285, (2006)

Shivkumar S., Wang L. and Keller C., "Impact Properties of A356-T6 Alloys," *Journal of Materials Engineering and Performance,* Vol. 3(1), pp. 83-90, (1994).

Sigworth K. Geoffrey and Wang Chengming, "Mechanism of Porosity Formation During Solidification: A theoretical Analysis," *Metallurgical Transactions B,* Vol. 24B, pp. 349-364, (1993).

Smith C. L., Easton M. A., Nie J. F., Zhang X. and Couper M. J., "Effect of Ti Content on Mechanical Properties of an Al-7Si-0.35Mg Alloy, *Materials Forum,* Vol. 28, pp. 1222-1228, (2004).

Sreeja Kumari S.S., Pillai R.M., Nogita K., Dahle A.K., and Pai B.C., "Influence of Calcium on the Microstructure and Properties of an Al-7Si-0.3Mg-xFe Alloy," *Metallurgical and Materials Transactions A*, Volume 37A, 2581-2591, (2006).

Sreeja, K. S. S., Pillai R.M., Rajan T. P. D., Pai B. C., "Effect of Individual and Combined Additions of Be, Mn, Cr and Sr on the Solidification Behaviour, Structure and Mechanical properties of Al-7Si-0.3Mg-0.8Fe Alloy," *Materials Science and Engineering A*, Vol. 460-461, pp. 561- 573, (2007).

Tahinata, A., Sato, N., Kutsumata, H. and T. Shiraishi, "Piston for Internal Combustion Engine," *US Patent*, No. US 7398754 B2, (2008).

Taylor, J. A., "Metal Related Castability Effects in Aluminium Foundry Alloys," *Cast Metals*, Vol.8 (4), pp. 225 – 252, (1995).

Taylor J. A.,Graham D. A. and Easton M. A., "The Redistribution of Shrinkage Porosity in Eutectic Al-Si Alloy by the Addition of Ti-B Grain Refiner," *AFS Transactions,* Vol. 107, pp. 189-196, (1999).

Taylor, J.A., Schaeffer, G.B. and StJohn, D.H., "The Role of Iron in the Formation of Porosity in Al-Si-Cu-Based Casting Alloys: Part I. Initial Experimental Observations" *Metallurgical and Materials Transactions A,* Vol. 30A, pp. 1643-1650, (1999a).

Taylor, J.A., Schaeffer, G.B. and StJohn, D.H., “The Role of Iron in the Formation of Porosity in Al-Si-Cu-Based Casting Alloys: Part II. A Phase-Diagram Approach” *Metallurgical and Materials Transactions A,* Vol. 30A, pp. 1651-1655, (1999b).

Taylor, J.A., Schaeffer, G.B. and StJohn, D.H., “The Role of Iron in the Formation of Porosity in Al-Si-Cu-Based Casting Alloys: Part III. A Microstructural Model” *Metallurgical and Materials Transactions A,* Vol. 30A, pp.1657-1662, (1999c).

Taylor J. A. “The effect of Iron in Al-Si Casting Alloys,” *Materials Forum,*35th Australian Foundry Institute National Conference, Adelaide Convention Centre, Adelaide South Australia, 31st Oct-3rd Nov 2004.

Tiryakioglu M., Askeland D. R. and Ramsay C. W., “Fluidity of 319 and A356: An Experimental Design Approach,”*AFS Transactions,* Vol. 102, pp. 17-25, (1994).

Tsukuda M., Masoyuki H, Suzuki T. and Koike S., “The Effects of Si, Mg, Fe on the Mechanical Properties of Al-Si-Mg Alloys for Castings,” *J. Jpn. Inst. Light Metals,* Vol. 28 (3), pp. 109-115, (1978).

Tuttle, B. L., Keslinke A., Twarog, D. and Daniels, E, “Influence of Antimony on A356 Sr-Modified Aluminium Melts” *AFS Trans,* Vol. 97, pp. 889 – 902, (May 1989).

Venkateswaran, S., Mallya, R.M. and Sheshadri, M.R., “Effect of Trace Elements on the Fluidity of Eutectic Al-Si Alloys Using the Vacuum Suction Technique,” *AFS Transactions*, Vol 94. pp. 701-708, (1986).

Villeneuve C., Samuel A. M., Samuel F. H., Doty H. W. and Valtierra S., “Role of Trace Elements in Enhancing the Performance of 319 Aluminium Foundry Alloys,” *AFS Transactions,* Vol. 109, pp 287-300, (2001).

Vorren O., Evensen J. E. and Pedersen T. B., “Microstructure and Mechanical Properties of Al-Si(Mg) Casting Alloys,” *AFS Transactions,* Vol. 92, pp. 459-466, (1994).

Wang, L., Makhlouf, M. and Apelian, D, “Aluminum Die Casting Alloys: Alloy Composition, Microstructure, and Properties-Performance Relationships” *International Materials Review,* Vol. 40 (6), pp. 225 – 252, (1995).

Wang, L. and Apelian D, “Review: Aluminium Die Casting Alloys, Appendix A of A Study of Alloy-Microstructure-Performance Interaction. Phase 1-A Low Resolution Study of 380 Alloy,” *Aluminium Casting Research Laboratory, Worcester Polytechnic Institute, Massachusetts,* (1993).

Wang Q. C., Jones P. E. and Osborne M., “Effect of Iron on the Microstructure and Mechanical Properties of an Al-7Si-0.4Mg Casting Alloy,” *Advanced Materials Engineering,* pp. 1-9, (2003).

Wang P.-S, Lee S.-L, Yang C.-Y and Lin J.-C, “Effect of Beryllium and Non Equilibrium Heat Treatment on Mechanical Properties of B319.0 Alloy with 1.0%Fe,” *Materials Science and Technology,* Vol. 20, pp. 539-545, (2004).

Xiufang B., Guohua Z., Shengxu Z. and Jiaji M., *Cast Metals*, Vol. 5, No. 1, pp. 39-41, (1992) - Cited.

Yie S. N., Lee S. L., Lin Y. H. and Lin J. C., “Mechanical Properties of Al-11Si casting Alloys Containing Trace Be and Sr,” *Materials Transactions, JIM,* Vol. 40, No. 4 pp. 294-300, (1999).

Zednik V., “The Influence of Chemical Composition and Quality of Raw Materials on the Physical Properties of Silumin Gamma,” *Metallurgia,* Vol. 37, pp. 253-257, (1948).

APPENDICES

Appendix I- Foundries involved in the Survey

	COMPANY NAME	CONTACT ADDRESS
1	East African Foundry Works (K) Ltd	Baba Dogo Road, Ruaraka, P.O. Box 48624-00100-G.P.O. Nairobi. Tel: +254- 20-8563604/5/6/7, Fax: +254-20-8561263/8561507 Email:eafw@alloysteel.com
2	Kens Metal Industries Ltd	Off Enterprise Rd, P.O. Box 18583-00500, Tel: 020-651164/651167/550594 Fax: 020-651168/558730 Email: kensmetal@africaonline.co.ke Nairobi
3	Satel Engineers Ltd	Baba Dogo Road, Ruaraka, P. O. Box 45858, Nairobi. Tel: 020-851367/8

4	Alloy Steel Castings Ltd	Baba Dogo Road, Ruaraka, P.O. Box 48624-00100-G.P.O. Nairobi. Tel: +254- 20-8563604/5/6/7, Fax: +254-20-8561263/8561507 Email:eafw@alloysteel.com
5	Baumann Engineering Ltd	Kampala Road, P.O. Box 30092-00100, Tel: 020-557022/557874 Fax: 542823, Nairobi.
6	Jua Kali Casting Metal Engineering Ltd	P.O. Box 77292, Tel: 020- 791297, Nairobi
7	Rubani Engineers	Enterprise Road, P.O. Box 40439, Tel 020-558920/554965, Nairobi.
8	Kenya Castings Ltd	Likoni Road, P.O. Box 74654, Nairobi. Tel: 020- 555065/542747/8
9	East African Hydraulics & Metals Industries (K) Ltd	Lunga Lunga Road, P.O. Box 30730, Tel: 020-559273 Nairobi.

10	Steel Foundry (K) Ltd	Naivasha Road, P.O. Box 33-00902, Tel: 066-2034 Kikuyu
11	KANEMCO Metal Co. Ltd	Kitui Road, P.O. Box 49535, Nairobi. Tel: 020-555155
12	Joska Metals, Kariobangi Light Industry Ltd, Owned by: Gachie Ndungu Joseph	Tel: 0733986083, P. O. Box 20549-00200, Nairobi. Email: joskametal@yahoo.com
13	Kariobangi Light Industry, Owned by: Caleb	Tel: 0710230776 Nairobi
14	Small Foundry in Industrial area, Owned by: Mutuku	Tel: 0721404683, Nairobi
15	Small Foundry in Industrial area, Owned by: Mwangi	Tel: 0722176779 Nairobi
16	Small Foundry in Industrial area, Owned by: Patel	Tel: 0720-633670 Nairobi
17	Small Foundry in Dandora, Owned by: Moses Njoroge	P. O. Box 7557-00200, Tel: 0733775459, Nairobi.

18	Small Foundry in Dandora, Owned by: David Ouma Ogutu	P. O. Box 42047, Nairobi. Tel: 0723955129
19	Small Foundry Industry in Kariobangi, Owned by: Emmanuel Okusimba Were	Tel: 0727748199 Nairobi
20	Beta Engineering Company Ltd	End of Commercial Street, Nairobi. Tel: 020-535814/535815
21	Morris & Company (2004) Ltd	Lunga Lunga Road, Off Mogadishu Road, P. O. Box 59307-00200, Nairobi.
22	Technosteel Industries Ltd	Gilgil Rd, Off Enterprise Rd Tel: 020-555096/556354/558836. Nairobi
23	Wali Mohamed Hamid Ltd	Dar-es-Salaam Rd, Opp Post Office, Industrial Area, Tel: 020-558075, Nairobi
24	Brass Engineers & Allied Foundry	Busia Rd, Nairobi
25	Metal Cast Ltd	Lusaka Rd, Tel; 020-542548, Nairobi

26	African Marine & General Engineering Co. Ltd	Mbaraki Creek, Tagana Rd, P.O. Box 90462-80100, Tel: 041-2221651/2221652/2221653 Email: afmarine@africaonline.co.ke Fax: 0412313168, Mombasa.
27	Narcol Aluminium Rolling Mills Ltd	Mikindani, P.O. Box 80872-80100, Tel: 041-2226317/2225077/3433712 Email: narcol@africaonline.co.ke Mombasa.
28	Steelmakers Ltd	Liwatoni Rd, Ganjoni, Mombasa, Tel: 041-223356/221486/221448
29	Steel Africa Ltd	Pemba Rd, Changamwe, P.O. Box 81827-80100, Tel: 041-2315612/2312599/2314422 Mombasa

30	Kaluworks Ltd	Mwangeka Rd, P.O. Box 90421-80100, Mombasa. Fax: 33074 Kikambala, Tel: 041-33280/33234
31	Mabati Rolling Mills Ltd	P. O. Box 68-80113, Mariakani, Mombasa Tel: 041-33630/33020/33012/33049 Fax: 041-33445/33432 Email: mabati@mabati.com
32	Nakuru Blanket Industries Ltd	Nakuru-Nairobi Rd, P.O. Box 22-20100, Nakuru
33	Hakika Engineering Workshop	Biashara Rd, Tel: 051-221370 Nakuru
34	Settlers Engineering Works	Industrial Area, Timber Mill Rd, Nakuru, Tel: 051-2211082/2214069/2214687, P. O. Box 284-20100, Nakuru
35	Rift Valley Engineering Works	George Morara Ave, Nakuru, Tel: 051-2211693, P. O. Box 306-20100, Nakuru.

36	Fiber Glass Industry	Nakuru
37	Maina’s Foundry	Nakuru
38	Booth Extrusions Ltd	Kenyatta Highway Thika, P. O. Box 1574-01000, Thika. Tel: 067-22936/22937/22962/30159
39	Premier Bag & Cordage Industry Ltd	Private Bag, Code: 00232, Ruiru. Tel: 067-52172/52393/52358.
40	Mumias Sugar Co. Ltd	Private Bag, Code:50102 Mumias. Tel: 056-641620/641621
41	Chemelil Sugar Co. Ltd	P.O. Box 177-40107, Muhoroni.Tel: 057-51527/51534/
42	South Nyanza Sugar Co. Ltd	P.O Box 107-40405, Sare Awendo, Tel: 059-43621/43622 Email: sonysugar@yahoo.com
43	Metallica Engineering Industries Ltd	Industrial Area, Nairobi
44	Rollmills Ltd	Industrial Area, Nairobi
45	Ramna’s Foundry	Industrial Area, Nairobi

*Other source of information is through the former Superintendent of the Foundry Section of the former Kenya Railways Corporation, now part of Numerical Machining Complex Ltd.

Appendix II-The Questionnaire

Preamble

Since 2001, the Department of Mechanical Engineering has been working on aspects of foundry practice with a view to improving the casting skills and the quality of castings. In case of skills, trainees and students have been exposed to moulding methods using green sand, CO_2/sodium silicate bonded silica sand and plaster of Paris. Use of molasses, bentonite and sodium silicate as binders has been emphasized. Furthermore, invaluable skills on artistic casting using investment casting method have been gained. For non-metals, rubber technique has been introduced. The moulding investigations have involved good practice in gating and risering designs and moderate studies in simulation of solidification modeling.

The foundry shop has concentrated on aluminium scrap and melting using oil fired furnace with a capacity of 76 kg of aluminium and maximum temperatures of 1000°C. Alongside the training, the shop has data on moulding materials mainly sand - in terms of grain size, strength characteristics, hardness and compactibility. The data is compared with the standards used elsewhere. The formula for risering and gating have been developed and tested. Recently, there has been focus on the metal itself and the effect thereof on the quality of the casting. Studies have been done on the scrap composition as obtained from cylinder heads, pistons, engine blocks and wheels. The investigation on the effect of additives to recycled aluminium is currently being undertaken. In particular, the effect of additions of minor elements (to the molten aluminium) on the microstructure, mechanical properties, fluidity and porosity formation is one of the information anticipated.

We have been working in collaboration with the local foundries. In overseas, we have been in touch with research institutions and suppliers of additives and other foundry related services. Compliance with ISO standards is a prerequisite for production of high quality products. The importance of this study is to ensure that our foundries have a competitive edge so that they can be subcontracted by the international community.

The purpose of this questionnaire is to enable the foundry practitioners learn more on the extent of application of the anticipated results. Collaboration between research institutions and local foundries undoubtedly lead to production of premium high quality and competitive parts in the respective local and overseas markets.

The questions contained herein are presented in two sections and have been structured in a way that crucial information on areas considered fundamental to improving the performance of local foundries shall be extracted. It is believed that through this effort good casting design, mould design and melt quality control (which includes chemistry control using additives, quiescent melt handling, gas and inclusion measurement and inclusion removal) shall be arrived at. Our local foundries will then be encouraged to take advantage of the available expertise and research findings to make premium castings that can compete favourably in the international market. In answering the questions contained in this questionnaire the respondent is not duty bound to respond to all test items. Finally, the information that shall be obtained will be treated in confidence and will not be in a form to identify the organization.

GENERAL INFORMATION ABOUT THE ORGANIZATION FOR SOME SELECTED FOUNDRIES AND ENTREPRENEURS

Name: ____

Location: ____

Tel No: ____

Fax: ____

Email: ____

Contact Person and postal address: _______________________________________

____

GUIDED QUESTIONS

PART I

Product Details

What kind of products do you cast?

1 Wheel pulleys [ ]

2 Meko jiko grills [ ]

3 Impellers [ ]

4 Hacksaw blade handles [ ]

5 Window handles [ ]

6 Locks [ ]

7 Knife and Panga handles [ ]

8 Gears [ ]

9 Others [ ] Give details ________________________

Which of the following casting methods do you employ in making your products?

1 Sand Casting [ ]

2 Die casting [ ]

3 Investment casting [ ]

4 Others [ ] Give details ___________________

Why did your firm prefer the casting method(s) you are using now?

1 No need for advanced technology [ ]

2 Cheap [ ]

3 No skilled manpower [ ]

4 Others [ ]

Which areas do your products find application?

1 Domestic appliances [ ]

2 Industrial components [ ]

3 Others [ ] Give details ____________________

What type of material do you use for your casting operations?

1 Ferrous material [ ]

2 Non-ferrous material [ ] Give details_____________________

Competitor Information

Please give details on your prices and that of your competitors.

Type of Product	Your cost per kg (Ksh.)	Competitor's cost per Kg (Ksh.)
Cast Iron		
Aluminium alloy		
Brass		
Bronze		
Others		

Do your products compete with those from overseas? Yes [ ] No [ ]

Please give details____

Capacity Utilization

1. Does your firm produces on order? Yes [ ] No [ ]

2. Please give details on how frequent you get orders from customers.
    - Daily [ ]

- Weekly [ ]
- Monthly [ ]
- Others [ ] Give details______________________

3. In what quantities on average do you get your orders?
    - Small quantities [ ]
    - Large quantities [ ]
    - Others [ ] Give details____________________________

4. Please indicate whether your firm uses different melting equipment for each product or whether the same facilities are used to cast all aluminum cast products.
    1 Different melting equipment [ ]
    2 Same melting equipment [ ]
    3 Others [ ] Give details____________________________

5. Please indicate whether your firm uses different moulding equipment for each product, or whether the same facilities are used to cast all aluminium cast products.
    1 Different moulding equipment [ ]
    2 Same moulding equipment [ ]
    3 Others [ ] Give details ________________________________

6. If the equipment is shared in casting the products, how do you organize your shift schedules among the different products?
    1 Produce one product at a time [ ]
    2 Produce products alternatively [ ]
    3 Others [ ] Give details ___________________________

7. Do you have plans to expand capacity? Yes [ ] No [ ].

8. If yes, please give details on the specific areas that need to be expanded and for what purposes.

Area to be expanded	Purpose for expansion
Sand casting	
High pressure die casting	
Investment casting	
Others	

Aluminium Scrap recycling.

Recycling of resources has become popular for various reasons as articulated hereunder;

a) Conservation of materials
b) Conservation of energy
c) Protection of environment

Studies have shown that aluminium recycling saves up to 95% of the fuel energy that could be used in making an equivalent amount of wrought aluminium. Aluminium recycling will therefore continue to attract attention from both research institutions and casting industry because of these reasons. Besides, recycled aluminium commands good attributes such as ability to give high strength to weight ratio, better castability, better surface finish, better wear resistance, better corrosion resistance and better welding characteristics than their competitors. This research forms part of a wider effort to develop methods and procedures of improving cast aluminium alloys in order to produce high quality products.

1. Does your firm use aluminium scrap? Yes [ ] No [ ]

2. What is the source of the aluminium scrap that you use in your firm?
    - Car Dealers [ ]
    - Scrap Vendors [ ]
    - Machine shops [ ]

3. What guides you to select scrap from your suppliers?
    1 General application [ ]
    2 Specific application [ ]
    3 Others [ ] Give details ______________________________

4. Do you give pre-selection information to the supplier? Yes [ ] No [ ]

5. What specifications do you require your suppliers to adhere to in order to ensure that good scrap is supplied?

   1 Cleanliness [ ]
   2 Composition [ ]
   3 Pre-selection details [ ]
   4 Others [ ] Give details ________________________________

6. What would you consider good quality scrap?

   1 One with several components [ ]
   2 One that is clean [ ]
   3 One that is composed of small pieces [ ]
   4 One with no swarf [ ]
   5 Others [ ] Give details________________________

7. How do you sort out your scrap?

   1 Visual inspection [ ]
   2 Use of magnets to remove ferromagnetic materials [ ]
   3 Conducting chemical analysis and grouping accordingly [ ]
   4 Others [ ] Give details____________________________

8. Do you separate cast aluminium alloys from wrought aluminium? Yes [ ] No [ ]

9. How do you separate cast from wrought alloys?
    1 Manually [ ]
    2 Others [ ] Give details______________________________

10. What automobile/household components do you usually encounter in the scrap you use?
    1 Pistons [ ]
    2 Cylinder heads [ ]
    3 Oil sumps [ ]
    4 Manifolds [ ]
    5 Electrical components such as alternators [ ]
    6 Household items such as sufurias [ ]
    7 Others [ ] Give details ______________________________

11. Do you usually clean the scrap before charging to remove extraneous items such as grease, oil or dirt? Yes [ ] No [ ].
12. Please indicate the method you use to clean scrap.
    1 Using Hot water [ ]
    2 Using Paraffin [ ]

3 Using Methylated spirit [ ]

4 Others [ ] Give details_________________________

13. How do you remove other impurities (paint, surface coatings and metallic impurities e.g. iron) that may adulterate the quality of your scrap?

1 Paint - Through heating [ ]

2 Through solvents [ ]

3 Metallic impurities - Through magnetic separators [ ]

4 Others [ ] Give details ___________________________

14. Please indicate the typical composition of the Al scrap you usually use.

Element	Al	Si	Mg	Mn	Fe	P	Cr	Ti	B	Cu	Zn
%Wt Composition											

15. How often do you carry out chemical composition analysis of your scrap?

1 Regularly [ ]

2 Occasionally [ ]

16. Before charging scrap into the furnace, what precautionary measures do you take to avoid any possibility of occurrence of accidents?

1 Check for air pockets in the scrap [ ]

2 Check for any explosives/chemicals [ ]

Melting Procedures.

1. Indicate the type of equipment your firm utilizes to facilitate the melting of aluminum alloys.
    - Oil fired crucible [ ]
    - Induction crucible [ ]
    - Charcoal fired crucible [ ]
    - Coke fired crucible [ ]
    - Others [ ] Give details______________________________

2. Does your firm have a procedure for melting and treating aluminium alloys in order to achieve high quality products? Yes [ ] No [ ]

3. Describe in detail the procedures followed by your firm in melting and treatment of aluminium alloys. Tick appropriately.
    1 Sorting [ ]
    2 Checking for air pockets [ ]
    3 Checking for presence of explosives or dangerous substances [ ]
    4 Chemical composition analysis [ ]
    5 Charging into crucible [ ]
    6 Adding fluxes [ ]
    7 Adding grain refiners [ ]
    8 Adding strontium or any other modifier [ ]

9 Adding Manganese, chromium, or any other neutralizer [ ]

10 Any other procedure ___________________________

4. Briefly explain why your firm has chosen to adopt the procedures and precautionary measures that you have mentioned to ensure a slag and dross free melt.

Reason ____

Precautionary measures

1. Checking for air pockets [ ]
2. Checking for presence of explosives [ ]
3. Use of additives [ ]
4 Give details on the additives_________________________________

5. Do you utilize fluxes during the melting process to remove inclusions and prevent oxidation? Yes [ ] No [ ]

6. List the various fluxes your firm uses, their functions and the suppliers.

Type of flux	Function	Supplier
Coveral II - a deoxidizer for aluminium alloys		
Navac		
Others		

7. Do you add grain refiners to your aluminium melts? Yes [ ] No [ ]

8. Outline the purpose of grain refiners addition
    1 Reduce the size of grains from columnar to fine equiaxed structure [ ]
    2 Improve the fluidity of the aluminium melt [ ]
    3 Improve mechanical properties such as tensile strength, impact energy [ ]

9. Do you employ degassers to remove hydrogen gas from the aluminium melt? Yes [ ] No [ ].

10. Please indicate the method you employ to remove gases from aluminium melts without use of degassers.
    1 Use of argon gas [ ]
    2 Use of nitrogen gas [ ]
    3 Others [ ] Give details ___________________________________

11. The presence of hydrogen in the aluminium melt has been reported to be a major contributing factor to porosity formation hence, the need to know its concentration in the melt. Do you use pressure test to determine the concentration levels of hydrogen and inclusions in your aluminium melt? Yes [ ] No [ ].

12. If no, then how do you determine hydrogen concentration levels?

    1 Through observation [ ]
    2 Any other [ ] Give details____________________________________

13. Eutectic modifiers are usually added to the aluminium alloy melt with a view to alter the morphology of eutectic silicon from acicular to fibrous form thereby improving the properties of the castings. The most commonly used modifiers are sodium (Na), strontium (Sr) and antimony (Sb). Do you add any eutectic modifier to the aluminum melt? Yes [ ] No [ ].

14. If yes, then describe the modifiers utilized.

    1 Na based modifiers [ ]
    2 Sr based modifiers [ ]
    3 Sb based Modifiers [ ]
    4 Others [ ] Give details _________________________________

15. At what temperatures do you normally carry out the treatment of the aluminium melts?

    1 650°C [ ]
    2 700°C [ ]
    3 750°C [ ]
    4 780°C [ ]
    5 Others [ ] Give details ____________________________

16. Do you carry out your heat treatment in the melting furnace, holding furnace or in the transfer ladles?

  1 In the melting furnace [ ]

  2 In the holding furnace [ ]

  3 In the transfer ladles [ ]

17. At what temperature do you pour the melt into the moulds?

  2 650°C [ ]

  3 700°C [ ]

  4 750°C [ ]

  5 780°C [ ]

  6 Others [ ] Give details ______________________________________

18. Indicate the precautionary measures you take to ensure that the melt does not pick more gas and dross during transfer process.

  1 Use of degassers [ ]

  2 Others [ ] Give details ____

19. Do you have access to a spectrometer? Yes [ ] No [ ]

20. If no, then briefly explain how you determine the chemical composition of

your melts.

1 Through experience [ ]
2 No composition analysis is done [ ]
3 Through guess work [ ]

Mould Design

1. List the general requirements that you consider to be major when designing moulds for your products.
    1 Mouldability of the pattern [ ]
    2 Position of parting line [ ]
    3 Which side is to be at the top and which one to be at the bottom [ ]
    4 Angularity of the fillets [ ]
    5 Size of the casting [ ]
    6 Feeding requirements and the places to be fed [ ]
    7 Metal to be cast and its behaviour when being cast [ ]
    8 Moulding media to be used i.e. sand or permanent mould [ ]
    9 Whether there will be localized cooling through use of chills [ ]

2. Briefly outline the design guidelines that you adopt in designing each of the following elements of the moulding systems.
    1. the position of the parting line [ ]
    2. the orientation of the mould [ ]

3. the pouring basin [ ]
4. the sprue [ ]
5. the well [ ]
6. the runner(s) [ ]
7. the ingates [ ]
8. the risers [ ]

3. Do you utilize chills in the mould design to enhance localized cooling? Yes [ ] No [ ].

4. If yes, do you use external or internal chills or both?
    1 External [ ]
    2 Internal [ ]
    3 Both [ ]

5. If you use chills, which material do you use and why?

Material	Reason
Steel Rods	
Cast iron blocks	
Any other	

6. Indicate the problems you may have encountered while using chills, and the

remedial steps you have taken to alleviate these problems.

Problems

1 Places to locate chills not accessible [ ]

2 Any other [ ] please give details ___________________________

Remedial steps

3 Creativity and innovativeness as to where to locate the chill could help [ ]

4 Any other [ ] Please give details _____________________________

7. What advantages can you attribute to chills in your opinion?

1 Localized solidification [ ]

2 Attainment of small equiaxed grain structure [ ]

3 Improved tensile properties [ ]

4 Improved impact strength [ ]

5 Attainment of sequential or directional solidification [ ]

6 Others [ ] Give details_________________________________

8. Do you use cooling fins for localized cooling to enhance sequential solidification of a casting in the mould? Yes [ ] No [ ].

9. If yes, outline the design procedures you follow in incorporating cooling fins in the mould system.
    1 Identify place that needs sequential solidification [ ]
    2 Make provision in the mould for the cooling fins [ ]
    3 Introduce the fin [ ]

10. What problems have you encountered in using cooling fins and how have you solved them?
    1 Problem - some materials used for fins may burn and introduce gases into the mould [ ]
    2 Others [ ] Give details ______________________________________
    3 Solution ____

11. What advantages can you attribute to cooling fins in your opinion?
    1 Ability to help achieve sequential solidification [ ]
    2 Achievement of improved properties of the casting [ ]

Post Casting Operations

1. Do you carry out any post-casting operations (heat treatment, annealing, etc) on your products before packing them for delivery? Yes [ ] No [ ].

2. If yes, then describe the operations.
    1 Heat treatment [ ]

2 Annealing [ ]

3 Any other [ ] Please give details ______________________

PART II

Quality Control

1. Do you carry out any quality control measures to ensure that your products are of good quality? Yes [ ] No [ ]

2. List some of the equipment you have to facilitate performance of tests.

   1 Spectrometer [ ]
   2 Tensile testing machine [ ]
   3 Optical Microscope with an attached camera [ ]
   4 X-ray testing machine [ ]
   5 Hardness testing machine [ ]
   6 Charpy impact testing machine [ ]
   7 Any other [ ] Please give details ______________________

3. Indicate the quality control procedures your firm follows in ensuring that good quality castings are delivered to the customers.

   1 Tensile Tests [ ]
   2 Macro inspection [ ]
   3 Micro inspection [ ]
   4 Impact tests [ ]

    5 Porosity tests [ ]
    6 ISO Certification [ ]

Co-operative Partnerships

1. Does your firm have any form of cooperation with other foundry firms? Yes [ ] No [ ].

2. If yes, the briefly explain the nature of the cooperation.
    1. Technical cooperation [ ]
    2. Economic cooperation [ ]
    3. Others [ ] Give details______________________________

3. If your firm is not in any form of cooperation with other foundry firms, would you be interested in being part of such cooperation was it to be established? Yes [ ] No [ ].

5. Please give reasons why you would like to be in such cooperation.
    1 To be able to meet and exchange views on how to improve the trade [ ]
    2 To share experiences [ ]
    3 Share skills [ ]

Problems

1. What problems does your firm generally face while striving to achieve its objectives?
    1. Knowledge of experienced personnel going to waste due to privatization and retrenchment [ ]
    2. Retirement of old people without tapping from their experience through knowledge transfer to new young people [ ]
    3. Acquisition of additives and material difficult from agents overseas [ ]
    4. Lack of knowledge on melt treatment procedures [ ]
    5. Little or no training is done at intermediate level colleges and training in the university is inadequate [ ]

2. How do you go about solving these problems?
    1 Recruitment of new young people to tap from the experience of old people [ ]
    2 Introduce more vigorous training in middle level colleges [ ]
    3 Material and additive acquisition through local agents [ ]
    4 Any other [ ] Give details ___________________________________

Comments

1. How would you like other foundry firms, the Universities and Government to participate in improving the foundry sector in this country?
    - Universities ____
    - Government ____
    - Other foundry firms _____________________________________

2. What other comments do you have?

____

____

____

____

____

____

____

____

____

____

Appendix III- Key Questions Selected for Detailed Analysis

(1) What kind of products do you cast?

1 Wheel pulleys [ ]

2 Meko jiko grills [ ]

3 Impellers [ ]

4 Hacksaw blades [ ]

5 Window handles [ ]

6 Locks [ ]

7 Knife and Panga handles [ ]

8 Gears [ ]

9 Others [ ] Give details ______________________________

(2) Which of the following casting methods do you employ in making your products?

1 Sand casting [ ]

2 Die casting [ ]

3 Investment casting [ ]

4 Others [ ] Give details _________________________________

(3) Why did your firm prefer the casting method you are using now?

1 No need for advance technology [ ]

2 Cheap [ ]

3 No skilled manpower [ ]

4 Others [ ] Give details________________________________

(4) Which areas do your products find application?

1 Domestic appliances [ ]

2 Industrial applications [ ]

3 Others [ ]

(5) What type of material do you use for your casting operations?

1 Ferrous material [ ]

2 Non ferrous material [ ] Give details____________________________

(6) Do your products compete with those from overseas? Yes [ ] No [ ]

Please give details____

(7) What specifications do you require your suppliers to adhere to in order to ensure that good scrap is supplied?

1 Cleanliness [ ]

2 Composition [ ]

3 Pre-selection details as provided by foreman [ ]

4 Others [ ] Give details_____________________________________

(8) How do you remove other impurities (paint, surface coatings and metallic impurities e.g. iron) that may adulterate the quality of your scrap?

1 Paint-through heating [ ]

2 Use of solvents [ ]

3 Metallic impurities- through magnetic separators [ ]

4 Others [ ] Give details ____________________________________

(9) Please indicate the method you employ to remove gases from aluminium melts other than degasers?

1 Use of argon gas [ ]

2 Use of nitrogen [ ]

3 Others [ ] Give details _________________________________

(10) List some of the equipment you have to facilitate performance of quality tests.

1 Spectrometer [ ]

2 Tensile testing machine [ ]

3 Microscope with an attached camera [ ]

4 X-ray machine [ ]

5 Charpy impact testing machine [ ]

6 Hardness testing machine [ ]

7 Any other [ ] Please give details ________________________

(11) Briefly outline the design guidelines that you adopt in designing each of the following elements of the moulding systems.

1 The position of the parting line [ ]

2 The orientation of the mould [ ]

3 The pouring basin [ ]

4 The sprue [ ]

5 The well [ ]

6 The runners [ ]

7 The ingates [ ]

8 The risers [ ]

(12) Do you utilize chills in the design of moulds to enhance localized cooling? yes [ ] No [ ]

(13) What can you say about the cost of raw materials (scrap metal, sand and coke) used in your foundry?

1. Cheap because they are locally available [ ]
2. Costly because of finding market abroad or are being imported [ ]
3. Affordable [ ]

(14) Do you use control charts and histograms to show the number of defects you incur on daily basis? Yes [ ] No [ ]

Appendix IV: Foundry Responses to key Questions in Appendix III.

The companies marked in bold with * had either pulled out of foundry as a core business or did not allow the survey to be conducted in their premises.

Code	Q1	Q2	Q3	Q4	Q5	Q6	Q7	Q8	9	10	11	12	13	14
A0	1-9	1-3	1-4	1-3	1-2	Y	1,2	1,2	2	-	1	Y	3	Y
A1	1-9	1-3	1,4	1-3	1-2	Y	2	1	2	-	1	N	3	Y
A2	1-9	1-3	1,4	1-3	1-2	Y	2	1	2	-	1	Y	3	Y
A3	1-9	1-3	1,4	1-3	1-2	Y	2	1	2	-	1	N	3	N
A4*														
A5	1-9	1-3	1,4	1-3	1-2	Y	2	1	2	-	1	N	3	N
A6*														
A7*														
A8	1-9	1-3	1,4	1-3	1-2	Y	2	1	2	-	2	-	2	N
A9	1-9	1-3	1,4	1-3	1-2	Y	2	1	2	-	-	-	2	N
B0	1-9	1-3	1,4	1-3	1-2	Y	2	1	2	-	1	-	2	N
B1	1-9	1-3	1-3	1-3	2	N	1,2	1,2	-	-	-	Y	2	N
B2	1-9	1-3	1-3	1-3	2	N	1,2	1,2	-	-	-	-	2	N
B3	1-9	1	1,2	1-3	1-2	N	1,2	1	2	-	-	-	2	N
B4	1-9	1	1,2	1-3	1-2	N	1,2	1	2	-	-	-	2	N
B5	1-9	1	1,2	1-3	1-2	N	1,2	1	2	-	-	-	2	N
B6	1-9	1	1,2	1-3	1-2	N	1,2	1	2	-	-	-	2	N
B7	1-9	1	1,2	1-3	1-2	N	1,2	1	2	-	-	-	2	N
B8	1-9	1	1,2	1-3	1-2	N	1,2	1	2	-	-	-	2	N
B9	1-9	1	1,2	1-3	1-2	N	1,2	1	2	-	-	-	2	N

C0	1-9	1	1,2	1-3	1-2	N	1,2	1	2	-	-	-	2	N
C1	1-9	1	1,2	1-3	1-2	N	1,2	1	2	-	-	-	2	N
C2	1-9	1	1,2	1,3	1-2	N	1,2	1	2	-	-	-	2	N
C3	1-9	1	1,2	1-3	1-2	N	1,2	1	2	-	-	-	2	N
C4	9	1	1,2	1-3	1-2	N	1,2	1	2	-	-	-	2	N
C5	1-9	1	1,2	1-3	1-2	N	1,2	1	2	-	-	-	3	N
C6	1-9	1	1,2	1-3	1-2	N	1,2	1	2	-	-	-	2	N
C7	2-9	1	1,2	1-3	1-2	N	1,2	1	2	-	1	N	2	N
C8	1-9	1	1,2	1-3	1-2	N	1,2	1	2	-	-	-	2	N
C9	1-9	1	1,2	1-3	1-2	N	1,2	1	2	-	-	-	2	N
D0	1-9	1	1,2	1-3	1-2	N	1,2	1	2	-	-	-	3	N
D1	9	1	1,2	1-3	1-2	N	1,2	1	2	-	2	N	2	N
D2*														
D3*														
D4*														
D5*														
D6	9	1,4	2	3	2	Y	2	-	1	1	-	Y	2	N
D7	8-9	1	1-3	2	1,2	N	1	1	-	-	-	N	3	N
D8	1-9	1	1,2	1-3	1-2	N	1,2	1	-	-	1	N	2	N
D9	9	1-3	1,2	1-3	1-2	N	1,2	1	2	-	-	-	2	N
E0	9	1-3	1,2	1-3	1-2	N	1,2	1	2	-	-	-	2	N
E1*														
E2*														
E3*														
E4	9	1-3	1,2	1-3	1-2	N	1,2	1	2	-	-	-	2	N