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Mechanical properties and machining of aluminum-silicon alloys modified by bismuth or tin

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MECHANICAL PROPERTIES AND MACHINING OF Al-Si ALLOYS MODIFIED BY Bi OR Sn

by

Peisheng Chen

A Thesis
Submitted to the Faculty of Graduate Studies through Engineering Materials
in Partial Fulfillment of the Requirements for the Degree of Master of Applied Science at the University of Windsor

Windsor, Ontario, Canada
2012
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MECHANICAL PROPERTIES AND MACHINING OF Al-Si ALLOYS MODIFIED BY Bi OR Sn

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10 December 2012
DECLARATION OF CO-AUTHORSHIP/PREVIOUS PUBLICATION

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I hereby declare that this thesis does not incorporate material that is result of joint research. In all cases, the key ideas, primary contributions, experimental designs, data analysis and interpretation, were performed by the author and Dr. A.T. Alpas as advisor.

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<td>Chapter 4</td>
<td>Effect of Bismuth on the Tensile Properties and Dry Machining Performance of Al-12.7 wt% Si Alloy, Ceramic Transactions, 207, 2009, 215-223.</td>
<td>Published</td>
</tr>
<tr>
<td>Chapter 6</td>
<td>Mechanical Properties and Machinability of an Aluminum-16 wt.% Silicon Alloy Modified by 0.5 and 1.0 wt.% Bismuth, Proceedings of the 114th AFS Metalcasting Congress, March 20-23, 2010 Orlando, USA</td>
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ABSTRACT

The dry machining performance, microstructure and mechanical properties of hypoeutectic and hypereutectic Al-Si alloys modified with different amounts of Bi and Sn, and cast at different cooling rates, were studied. The measured cutting and thrust forces decreased with the addition of Bi and Sn. These elements caused intense shear localization, and promoted the formation of segmented chips, thereby improving the dry machining performance. Also Bi and Sn melted and thus acted as lubricants during dry turning. However, the mechanical properties decreased with the addition of Sn, although Sn had no effect on the Si morphology. Bi had no effect on the Si morphology when cast at high cooling rates. The optimum amount of Bi addition was found to be 0.5 % and this alloy cast under a high cooling rates of 26 °C/s, improved the overall machining performance without compromising the mechanical properties.
DEDICATION

TO MY FAMILY
ACKNOWLEDGEMENTS

I would like to express my sincere gratitude to my advisors Dr. A. T. Alpas for his supervision, encouragement, support and patience during my graduate studies. I would like to profusely acknowledge the assistance of Drs. M.J. Lukitsch and Y. -T. Cheng of the GM Global Research for their valuable suggestions and encouragement.

I also wish to extend my thanks to my committee members: Dr. H. Hu and Dr. H. Wu for their invaluable discussions, suggestions and time. Sincere thanks to Mr. P. Seguin for his invaluable help with the construction and calibration of the force measurement system. Help and training provided by Mr. Robinson is gratefully acknowledged. Special thanks to the Technical Assistance Shop crew for providing assistance with the set-up and for coordinating the tests. Dr. Meng-Burany is acknowledged for the TEM characterization of chips.

I am grateful to all past and current researchers of the NSERC/General Motors of Canada Industrial Research Chair in Tribology of Lightweight Materials for their help and friendship. Special thanks to my fellow researchers Mr. S. Bhowmick, Mr. F. Sen, Dr. M. Safiei for their support and friendship.

Financial support provided by NSERC (Nature Science and Engineering Research Council of Canada), General Motors of Canada Limited through an Industrial Research Chair Program under Dr. A. T. Alpas is greatly appreciated.
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LIST OF ABBREVIATIONS, NOMENCLATURE

PDZ Primary Deformation Zone
SDZ Secondary Deformation Zone
TDZ Tertiary Deformation Zone
BUE Built-Up Edge
PCD Poly Crystalline Diamond
SEM Scanning Electron Microscope
TEM Transmission Electron Microscope
EDS Electron Dispersive Spectroscopy
COF Coefficient of Friction
MQL Minimum Quantity Lubrication
UTS Ultimate Tensile Strength
YS Yield Strength
ECM Electrochemical Machining
EDM Electrodischarge Machining
UM Ultrasonic Machining
SADS Secondary Dendritic Arm Spacing
HSS High speed steel
PCBN Polycrystalline cubic boron nitride
WC Tungsten carbide
VSI mode Vertical scanning interferometry mode
EDX Energy dispersive X-ray spectrometer
DLC Diamond-like carbon

α Rake angle
β Friction angle between chip and tool
\( V_c \) Cutting speed
\( f \) Feed per revolution or feed rate
\( t_c \) Chip thickness
\( D \) Depth of cut
\( w \) Width of chip
\( \phi \) Shear angle
\( \mu_e \) Coefficient of friction
\( F_{friction} \) Friction force between rake face of tool and chip
\( F_r \) Force exerted by the tool on the workpiece
\( F_c \) Cutting force
\( F_t \) Thrust force
\( F_s \) Shear force along shear plane
\( N_{f} \) Normal force acting perpendicular to shear plane
\( F_{p} \) Force parallel to rake face of tool
\( N_{p} \) Force perpendicular to rake face
\( N_{f} \) Normal force acting perpendicular to rake face
\( r \) Cutting ratio: defined as ratio of feed to chip thickness
\( \tau_s \) Shear stress on the shear plane
\( \sigma_s \) Normal stress on the shear plane
\( \tau_f \) Shear stress on the chip in contact with the rake face
\( w \) Width of chip
\( \gamma \) Shear strain in cutting
\( \dot{\gamma} \) Strain rate in cutting
\( d_s \) Thickness of shear band
\( \gamma_s \) Shear displacement in shear band
\( \dot{\varepsilon} \) Equivalent plastic strain
\( \theta \) Deformation angle
<table>
<thead>
<tr>
<th>Symbol</th>
<th>Description</th>
</tr>
</thead>
<tbody>
<tr>
<td>$d_s$</td>
<td>Displacement between the shear bands formed</td>
</tr>
<tr>
<td>$t_s$</td>
<td>Thickness of shear bands</td>
</tr>
<tr>
<td>$H$</td>
<td>Vickers hardness</td>
</tr>
<tr>
<td>$\sigma$</td>
<td>Flow stress</td>
</tr>
<tr>
<td>$C$</td>
<td>Material constant, equal to 3</td>
</tr>
<tr>
<td>$Q$</td>
<td>Volume removed per revolution of workpiece</td>
</tr>
<tr>
<td>$u_s$</td>
<td>Specific cutting power</td>
</tr>
<tr>
<td>$W$</td>
<td>Work of plastic deformation per unit volume of material</td>
</tr>
<tr>
<td>$\rho$</td>
<td>Density</td>
</tr>
<tr>
<td>$C_p$</td>
<td>Specific heat capacity of material</td>
</tr>
<tr>
<td>$K$</td>
<td>Thermal diffusivity of the material</td>
</tr>
<tr>
<td>$T_{amb}$</td>
<td>Ambient temperature</td>
</tr>
<tr>
<td>$F_{sys}$</td>
<td>Force on force measurement system</td>
</tr>
<tr>
<td>$M$</td>
<td>Total mass of weights</td>
</tr>
<tr>
<td>$m$</td>
<td>Mass of weight hanger pan</td>
</tr>
<tr>
<td>$g$</td>
<td>Acceleration due to gravity</td>
</tr>
<tr>
<td>$\sigma_f$</td>
<td>Normal stress on the rake face</td>
</tr>
<tr>
<td>$\sigma_0$</td>
<td>Yield stress</td>
</tr>
<tr>
<td>$R$</td>
<td>Material removal rate during cutting</td>
</tr>
<tr>
<td>$U$</td>
<td>Power consumed per unit time during cutting</td>
</tr>
<tr>
<td>$\gamma_{gl}$</td>
<td>The grain-to-particle interface</td>
</tr>
</tbody>
</table>
CHAPTER 1  Introduction

1.1. Focus of the Thesis

Lightweight materials have become crucially important in automotive technology [1, 2] as the result of an ever growing demand for improved product performance and fuel economy through mass reduction. Al-Si castings constitute 85% to 90% of the total aluminum cast parts produced. They are the material of choice for a variety of automotive components because of their high strength-to-mass ratio, superior castability, corrosion resistance, recycling potential and mechanical properties [1, 3-6].

Most Al-Si components require machining before they can be used in assembly. Current government and consumer attention on the environmental sustainability of products and processes has motivated manufacturers to reduce the volume of their waste streams, particularly the metal cutting fluids used as coolants and lubricants. The dry machining of Al-Si alloys may become an environmentally sustainable alternative to the conventional flooded cutting of castings. The primary reasons for the shift to dry or near dry machining are: (i) concerns over the contamination of the environment that results from the use and disposal of large quantities of coolants and lubricants, (ii) expenses generated by the use of these cutting fluids, and (iii) the health hazards experienced by machining operators. The dry machining of Al alloys, however desirable, remains challenging because the absence of cutting fluids allows Al chips to adhere to the tool surfaces [7-9], which causes premature tool failure.

Tool life has a strong economic impact on production operations. For example, studies show that a Michigan transmission plant could save upwards of a million dollars if their tool life could be improved by 20% [10]. Cutting tool failures have been
traditionally classified as [11]: adhesive wear, abrasive wear, diffusion wear, microchipping, plastic deformation, induced wear, or delamination—often with more than one wear type occurring simultaneously. Some of the factors that would improve tool life include a decrease in cutting forces, a reduction of build-up-edge (BUE), and a low coefficient of friction (COF) and/or temperature between the tool rake face and the chips. It is also important to consider the factors affecting a workpiece’s properties and microstructures. The composition, melt treatment, and solidification rate during casting all play roles in determining the microstructure.

The resulting properties of cast Al-Si alloys are determined by the effect of additives, namely Al grain refiners and silicon modifying elements [6]. The presence of primary and eutectic Si particles; which are known to impart high wear resistance, high stiffness and low thermal expansion; plays an important role in the restriction of metal flow during machining [6]. Large, unevenly distributed primary Si particles in hypereutectic Al-Si alloys cause greater tool wear than smaller, more uniformly distributed primary phases through melt treatment using silicon modifying elements [6, 11]. Al-Si alloys used in the automotive industry are cast at different cooling rates and display different properties, microstructures, and machining performance. For example, engine blocks manufactured from hypereutectic A390 can be cast and solidified at a slow cooling rate using sand casting, or at a high cooling rate using die casting [12]. A transmission body manufactured from A383 is cast and solidified at a high cooling rate using die casting, while the liners in engine blocks manufactured from Al-25Si are cast and solidified at a high cooling rate using spray casting [13]. Al-Si alloys with high cooling rates have refined microstructures [6].
The microstructures of Al-Si alloys can also be refined through melt treatment [6] using Al grain refiners, such as Al-Ti-B. Al-Sr is used for hypoeutectic and eutectic Al-Si alloys and Al-P for hypereutectic Al-Si alloys to achieve silicon particle size and morphology modification. Low-melting-point elements, such as Bi, Sn, Pb, and Cd all have an influence on the microstructure and properties of Al-Si alloys. Some studies have been conducted on the machinability of alloys with low-melting-point elements [14, 15]. These studies revealed that the dry drilling performance of a cast Al alloy B319 can be improved through the addition of low-melting-point elements, namely Bi, Sn, and Pb. However, there are no systematic studies on the role of low-melting-point elements in the microstructure, mechanical properties and machining performance of Al-Si alloys. In this perspective, this research focuses on the following points: (i) Understanding the interactions between low melting point additives, such as Bi, Sn, Si, and Mg as well as Sr, P additives; (ii) Understanding the effects that low melting point additives have on the microstructure of Al-Si alloys at different cooling rates during casting; (iii) Studying the mechanical properties of the resulting castings; (iv) Conducting machining tests to assess the performance of the new alloys based on the measurements of cutting forces, temperature, chip morphology, and the distribution of low melting point additives; (v) Analyzing chip formation mechanisms and the effects that the resulting castings have on tool life; and (vi) Optimizing dry turning performance and properties with composition, microstructure, and casting parameters.

1.2. Organization of the Thesis

This thesis is composed of seven chapters. Following the introduction, Chapter 2 presents a review of the literature on orthogonal turning, the properties improvement of
cast Al-Si alloys, the machining of Al alloys, and Al alloys with low melting point elements. Chapter 3 describes the experimental methods and procedures, including the raw materials, melt treatment, cooling rate and casting procedure, metallography and mechanical tests, dry turning system, tool condition, and methods for measuring force and temperature. Chapter 4 focuses on alloy design and microstructure control and the preparation of Al-Si castings for the following tests. Chapter 5 presents the mechanical properties and dry turning performance of Al 319.2 hypoeutectic Al-Si alloys modified by Bi or Sn with different cooling rates. Chapter 6 presents and compares the properties and dry turning performance of Al 390 hypereutectic Al-Si alloys modified by Bi and cast at two different cooling rates. The properties and mechanical performance of Al 390 alloys with different concentrations of Bi are also investigated in Chapter 6. Chapter 7 summarizes the present work.

1.3. Thesis Objective

The objective of this work was to assess the effect of Bi and Sn on the microstructure, mechanical properties and dry turning performance of Al-Si alloys cast at different cooling rates. Optimizing the composition and casting parameters of hypoeutectic and hypereutectic Al-Si alloys were through the morphology of Si particles, the distribution of Bi and Sn phases and the tensile test results. Dry turning performance of the alloy, which was judged by analysis of factors such as forces generated during the process and the power expended during machining, interfacial temperature between the tool rake face and chip, quality of surface generated and the geometry of chip produced, was studied. The melting and shear deformation of Bi and Sn in chips were also analyzed through SEM-BSI and EDS maps.
CHAPTER 2  Literature Review

The dry machining of advanced materials is being promoted in the manufacturing industry, especially the automobile industry, where Al alloys are of interest because of the significant mass reductions achievable over traditional cast iron and steel. There are two main driving forces to promote dry machining: (i) improved working conditions for machine operators for health and safety by eliminating the use of metal cutting fluids, and (ii) the resulting potential economic benefits. It is well known that the dry machining of aluminum alloys is a challenging task since in the absence of cutting fluids, aluminum chips adhere to the tool surface and cause premature tool failure.

This chapter reviews the state of the art in the machining of aluminum alloys. Orthogonal machining processes in terms of cutting forces, stresses and temperature distribution are reviewed in Section 2.1. Section 2.2 discusses the existing literature on the melt treatment and microstructural control of aluminum alloys that are used in castings where machining is important. Section 2.3 focuses on the existing literature on the dry and near-dry machining of aluminum alloys, while Section 2.4 specifically examines the role of low melting point elements, such as Bi and Sn, on the microstructure and properties of aluminum alloys. Section 2.5 is a summary of this literature survey.

Briefly, machining is the most widespread metal-shaping process in the manufacturing industry. It is a severe and localized plastic process, usually coupled with thermo-chemical events that convert most of the mechanical work into heat due to friction between the workpiece, tool and chips, and the plastic deformation during the formation of chips [11, 16-18]. Machining research has three study areas as follows: (i)
cutting tools, including cutting tool materials and failure mechanisms, (ii) the properties of workpiece materials, and (iii) operation parameters, namely speed and feed rate [19].

Machinability has traditionally been used to define some level of machining performance, with a focus on the workpiece. For aluminum alloys, the machining performance can be categorized into five classes on the basis of chip length and surface characteristics of the final product, namely A, B, C, D and E, in increasing order of chip length and decreasing order of the surface finish produced [10, 20]. Workpieces with different compositions, microstructures and properties require different cutting tools, and their machining parameters must be optimized. Cast aluminum alloys have been found to show a drastic decrease in cutting forces if PCD tools are used [10, 20-21]. The machinability of Al-Si alloys modified by Sr are improved, however Sr may cause undesirable porosity [22]. Bi can counteract the undesirable effect of Sr during casting [23-25]. It is therefore important to study the casting and subsequent machining processes together by investigating the effects of the casting process parameters on the machinability of Al-Si alloys [26].

2.1. Orthogonal Turning

2.1.1 Different Chip Formation

Parts manufactured by casting, forming and other shaping processes often require to be machined to its final shape and dimensions with one or more of the processes listed in Table 2.1 [27]. Turning, milling and drilling are the three most common types of machining operations. In the turning process, a wedge-shaped cutting tool that consists of two surfaces intersecting to form the straight cutting edge is used. The surface along which the chip flows is termed the rake face, and the surface that contacts the new or
machined workpiece surface is known as the flank face. The turning process is an example of orthogonal cutting, which is a two-dimensional process; it possesses a relatively simple cutting geometry. Therefore, it has been widely used in theoretical and experimental work [11, 27]. Turning is the process studied in this research, and thus, the principles of this operation and the factors that determine the orthogonal turning performance will be explained in detail.

Table 2.1 Range of material-removal processes

<table>
<thead>
<tr>
<th>Cutting</th>
<th>Abrasive</th>
<th>Nontraditional</th>
</tr>
</thead>
<tbody>
<tr>
<td>Circular</td>
<td>Various</td>
<td>Bonded</td>
</tr>
<tr>
<td>Turning</td>
<td>Milling</td>
<td>Grinding</td>
</tr>
<tr>
<td>Boring</td>
<td>Planning</td>
<td>Honing</td>
</tr>
<tr>
<td>Drilling</td>
<td>Shaping</td>
<td>Coated abrasives</td>
</tr>
<tr>
<td></td>
<td></td>
<td>Lapping</td>
</tr>
<tr>
<td>Drilling</td>
<td>Broaching</td>
<td>Polishing</td>
</tr>
<tr>
<td></td>
<td></td>
<td>Sawing</td>
</tr>
<tr>
<td></td>
<td></td>
<td>Filing</td>
</tr>
<tr>
<td>Gear forming</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Gear generating</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

The important process variables of orthogonal turning are the cutting speed (generally expressed in units of m/s), and feed (generally expressed in units of mm/rev). Correspondingly, the force F exerted by a cutting tool has three mutually perpendicular components as shown in Fig. 2.1 (a). The coordinate system is oriented along the directions of primary and feed motions, i.e. the total force is resolved into the perpendicular projection of these two directions [27] into the cutting force and feed force denoted by symbols \( F_c \) and \( F_f \). Moreover, there is a back force \( F_p \) tending to push the tool away from the workpiece in a radial direction perpendicular to the working plane. This force is usually ignored for orthogonal cutting. As can be seen in Fig. 2.1 (b), in
orthogonal cutting, the entire force system lies in a single plane and the appropriate force components can be easily calculated by drawing the Merchant circle diagram [11, 27-28], the details of which will be discussed in Section 2.1.2.

Continuous chips are long and ribbon-like, generally produced during the machining of soft, ductile materials such as 6061 Al and copper [29, 30]. These long chips may lead to operational hazards as they may get tangled around the tool or workpiece. High speeds, small feeds and small depths of cut are conditions that favour the generation of continuous chips. Discontinuous chips are fragmented chips formed by the chip breaking into small segments due to cyclic fracturing, generally encountered during the machining of brittle materials such as cast iron and some Al-Si alloys [31].

2.1.1 Continuous Chip Formation

Various theories [17, 32-38] have been proposed to explain the formation of continuous chips during orthogonal cutting process and are described below.

The model proposed by Piispanen and Merchant [17, 32-33] considers the shear angle formed during an orthogonal cutting process to assume a value such that the work expended for the process will be the smallest. It assumes that the chip forms across a straight single shear plane as shown in Fig 2.2.

Fig. 2.3 shows the crystal structure of the metal as circles drawn on the sides of a stack of cards, elongated in the direction of the major axis of the ellipses produced, which is considerably different from the direction of shear, making an angle $\psi$ with it, and this was seen to be in conformance with experimental results.

Angles $\psi$, $\phi$ and $\alpha$, as shown in Fig. 2.3, are related as:

$$\cot \psi = \cot \phi + \tan(\phi - \alpha)$$

(2.1)
where $\phi$ is the shear angle and $\alpha$ is the true rake angle of the cutting tool.

The card-like elements, each having a finite thickness of $\Delta x$, are displaced by the cutting tool through a distance of $\Delta s$ with respect to its neighbour during the formation of the chip. The shearing strain $\varepsilon$ can thus be found as follows:

$$\varepsilon = \frac{\Delta s}{\Delta x} \quad (2.2)$$

$$\varepsilon = \cot \psi \quad (2.3)$$

Oxley and Welsh [34] proposed a parallel-sided shear zone model to predict the variations in shear angle and cutting forces with variations in cutting speed and feed rate. In their model, the constant shear angle leads to constancy of shear speed across the chip thickness, as the material shears on a straight plane and constant distance under a constant cutting speed.

Dewhurst proposed a model shown in Fig. 2.4 based on the assumption that the shear plane is made up of several curved surfaces or planes that are mutually parallel to each other [35]. The curved shear planes lead to curved chips as they dictate a non-uniform shear on the shear plane during orthogonal cutting.

Von Turkovich [36] used a dislocation mechanics approach to develop a theory for high strain rate processes in machining, and postulated that stress saturation occurs above a specific strain generated in the workpiece material. Experimental studies performed by Ramalingam and Black [37] examined the machining problem from a physical or metallurgical view point, using scanning and transmission electron microscopy of the plastically deformed zone at the root of the chip to shed light on the role of the microstructure of the workpiece material (including the role of stacking fault energy and dislocation density) on the machinability. They suggested that a dynamic
equilibrium state is established between the work hardening and recovery processes during continuous chip formation. Another feature of the deformation microstructure developed during machining is the formation of localized shear bands.

The shear zone model proposed by Hitomi and Okushima [38] is based on the existence of a large transitional zone AOB for plastic deformation and is shown in Fig. 2.5. They started with the assumption that the workpiece material is a rigid plastic solid, and that the boundary surfaces of the flow or shear zone are planes emanating from the cutting edge, as seen in Fig. 2.5. The stress in the flow zone is in the yield state. This model predicts the shear strain to be zero at the lower boundary of the shear zone, and maximum at the upper boundary of the shear zone. The shear stress $\tau_\infty$ can be computed by the formula:

$$\tau_\infty = R \sin[\beta/(ab_1)]$$

(2.4)

where $\beta$ is the friction angle of the tool face, $a$ is the contact length between the chip and the tool, $b_1$ is the width of the tool, and $R$ is the resultant force on the tool. The concept of a flow region in this model can also be applied to discontinuous chip formation, which eliminates the conventional dilemma that fracture occurs abruptly on a particular single shear plane and is associated with extremely large strains.

2.1.1.2 Discontinuous Chip Formation

The transition from continuous to discontinuous chip formation depends on the thermo-physical properties and the metallurgical state of the workpiece material, as well as on the cutting process [39-45].

According to Drucker [39], chips become more segmented (discontinuous) with decreasing cutting speeds and increasing feeds. Low rake angles, rigidity of tool, high
feed rates and the presence of inhomogeneities in the workpiece material like second phase particles, lead to the formation of discontinuous chips in metals. For a completely homogeneous material, the transition from continuous to discontinuous chips would be abrupt. However, it would be rather gradual for a non-homogeneous material. At lower speeds, fractures produced in the chips are almost complete and deep, leading to the formation of segmented chips [28].

Cook [40] explained the steps in the formation of discontinuous chips based on the single shear plane concept. The chip formation process starts at a high shear angle, which is followed by the sliding of the chip onto the rake face, which in turn leads to an increase in the frictional force deterring the motion of the newly formed chip that causes it to bulge, and there is a simultaneous decrease in the shear angle. The strain along the shear plane increases constantly while the bulging occurs, until it reaches the point of ductile shear fracture, as shown in Fig 2.6. This fracture leads to the formation of segments similar to saw-teeth. This process continues cyclically leading to discontinuous or serrated chip formation.

Cowie, et al. [41] presented a review of the several flow localization processes taking place during cutting processes. The adiabatic shear deformation theory of discontinuous chip formation suggests that saw-toothed chips are formed primarily due to a catastrophic thermoplastic instability in the material being cut [42], which is characterized by a decrease in the flow stress due to thermal softening that is associated with an increase in strain, which more than offsets the effects of the associated strain hardening [7]. It was found that during the orthogonal cutting experiments on titanium alloys [43-45], extremely high temperatures developed in them are due to the
substantially high cutting pressure and extremely low thermal properties, which cause the heat to be accumulated in a thin layer at the chip/tool contact.

2.1.2 Forces in Orthogonal Turning

There are different theories and mechanical models for continuous and discontinuous chip formation. The behaviour of materials during cutting processes in discontinuous chip formation is almost the same as continuous chip formation except that fracture occurs when the shear strain at the end boundary line reaches the breaking limit repeatedly as the cutting proceeds [36].

The geometrical resolution of the total cutting force is usually done by assuming that the chip is a solid body in stable mechanical equilibrium under the action of the forces exerted on it at the rake face and at the shear plane as shown in Fig 2.1 (b), as suggested by Merchant [11, 16] on the basis of the following assumptions [11]: (i) the chip is continuous with no built up edge, (ii) the tool tip is perfectly sharp, which leads to absolutely no contact between the machined surface and the flank of the tool, (iii) a perfectly planar shear plane, along which shearing occurs as the tool tip pushes into the material, extending from the tool tip to the chip root, (iv) the depth of cut is assumed to be very large with respect to the feed of cut, thus reducing the problem to one of plane strain, and (v) shear stress, strain and strain rates are all assumed to be uniform.

The relationship between the various forces determined by using these force diagrams is as follows:

\[ F_i = F_c \cos \phi - F_s \sin \phi \]  
\[ N_s = F_c \sin \phi + F_t \cos \phi \]  
\[ F_f = F_c \sin \alpha + F_t \cos \alpha \]
\[ N_f = F_c \cos \alpha - F_t \sin \alpha \]  \hspace{1cm} (2.8)

where \( F_s \) is the shear force on the shear plane, \( F_c \) is the cutting force, \( F_t \) is the thrust force, \( N_s \) is the normal force on the shear plane, \( F_f \) is the shear force on rake face, \( N_f \) is the normal force on the rake face, \( \phi \) is the shear angle, and \( \alpha \) is the true rake angle of the cutting tool.

The coefficient of friction (\( \mu \)) at the tool face can be computed as:

\[ \mu = \frac{F_f}{N_f} = \frac{F_c \sin \alpha + F_t \cos \alpha}{F_c \cos \alpha - F_t \sin \alpha} = \tan \beta \]  \hspace{1cm} (2.9)

where \( \beta \) is the friction angle.

The shear angle in Eqn. 2.5 is determined from a section of the chip by microscopic observations or by using the cutting ratio \( r \), which is the ratio of the feed \( f \) (distance moved by the tool-bit along the workpiece revolution axis, per revolution of the workpiece) to the chip thickness \( t_c \) (wall thickness of the tube being machined off) as is shown below [10]:

\[ r = \frac{f}{t_c} = \frac{AB \sin \phi}{AB \cos (\phi - \alpha)} \]  \hspace{1cm} (2.10)

and

\[ \phi = \tan^{-1} \left( \frac{r \cos \alpha}{1 - r \sin \alpha} \right) \]  \hspace{1cm} (2.11)

where \( AB \) equals the length of the shear plane, and \( \alpha \) is the rake angle of the tool.
2.1.3 Stresses on the Shear Plane and the Rake Face

The card model or shear plane model [10] is useful to explain orthogonal cutting, and presents a simplified method for predicting the forces on the shear plane and rake face of an orthogonally cut sample. It assumes that the distribution of the shear and normal forces on the shear plane and on the rake face is uniform. From this assumption, the shear stress on the shear plane $\tau_s$ can be calculated as follows:

$$\tau_s = \frac{\text{Shear force on shear plane}}{\text{Area of shear plane}}$$  \hspace{1cm} (2.12)

From Eqn. 2.12, we can derive that,

$$\tau_s = \frac{F_c \cos \phi - F_s \sin \phi}{\frac{wf}{\sin \phi}}$$ \hspace{1cm} (2.13)

where $f$ is the uncut chip thickness that is equal to the feed rate, and $w$ is the width of the chip that equals the depth of cut in the turning process.

Normal stress on the shear plane $\sigma_s$ can similarly be computed as follows:

$$\sigma_s = \frac{\text{Normal force on shear plane}}{\text{Area of shear plane}}$$ \hspace{1cm} (2.14)

From Eqn. 2.14, we can derive that,

$$\sigma_s = \frac{F_c \sin \phi + F_s \cos \phi}{\frac{wf}{\sin \phi}}$$ \hspace{1cm} (2.15)

Similarly, the shear stress $\tau_f$ on the chip in contact with the rake face is equal to:

$$\tau_f = \frac{F_c \sin \alpha + N_s \cos \alpha}{wl}$$ \hspace{1cm} (2.16)
And the normal stress on the rake face, $\sigma_f$, can be written as:

$$\sigma_f = \frac{F_c \cos \alpha - N_c \sin \alpha}{wl}$$  \hspace{1cm} (2.17)

Where $l$ is the length of the sliding contact.

### 2.1.4 Cutting Power Requirements

In metal cutting, the chip formation zone [27] can be considered to consist of three characteristic zones I-III, which can be schematically described as shown in Fig. 2.7. According to the authors, the primary deformation zone I (PDZ) is distinguished by the area $OAB$ and is characterized by a number of slip lines including the $OFA$ line (the entry boundary) on which plastic deformation begins and the $OCQB$ line (the exit boundary). Then, the material adjacent to the tool/chip interface of OD length is subsequently deformed in depth $\Delta_1$ due to intensive interfacial friction. The $OCD$ region is called the secondary deformation zone II (SDZ). Additionally, the tertiary deformation zone III (TDZ) is localized below the cutting edge in depth $\Delta_2$. The total work done by the cutting tool to remove the metal is dissipated in the following forms for a sharp tool: (1) as work done in PDZ, (2) as work done to overcome chip/tool interracial friction in SDZ, and (3) as work done to generate a new surface in cutting, which is often neglected. The energy required to perform primary and feed motions can be calculated using **Eqns 2.18 and 2.19**:

$$E_c = \int_0^l F_c V_c \, dt$$  \hspace{1cm} (2.18)

$$E_f = \int_0^l F_f V_f \, dt$$  \hspace{1cm} (2.19)

where $F_c$ is the cutting force, $V_c$ is the cutting speed defined as the speed of the lathe rotation in a turning operation, and $V_f$ is the feed rate defined as the distance
perpendicular to the cutting direction between the surface machined in the previous tool pass and the surface exposed in the current tool pass. Total energy during cutting will be consumed in several ways [11]: (i) as shear deformation energy on the shear plane, or the deformation energy within the deformation zones, (ii) as frictional energy on the tool surface, (iii) as surface energy due to the formation of a new surface, (iv) as momentum energy due to the momentum change as the metal crosses the shear plane.

2.1.5 Temperature Distribution in Orthogonal Cutting

The chip and rake face interfacial temperatures play a major role in the performance of machining processes. Temperatures rise mainly in the primary shear zone due to the large amount of plastic deformation. The primary shear zone temperature is of importance for its influence on the flow stress. The temperature on the tool face also plays a major role in determining the size and stability of the built-up edge (BUE) [11].

In machining processes, complex temperature fields are present, but the maximum temperature is observed at the rake face near the middle of the tool/chip interface, as the chip experiences friction when sliding against the rake face of the tool in the secondary shear zone [11].

There are several experimental techniques for measuring the temperature along the face of a cutting tool, including the use of the tool-work thermocouple technique based on the electromotive force, measurement of infrared radiations emitted from the cutting zone, and coating specimens with thermo-sensitive paints [11]. The tool-work thermocouple technique estimates the temperature over the entire contact area between the chip and tool. Leading values could be obtained if a BUE or oxide layers are present. Various methods have been suggested for the estimation of the temperature distribution
in an orthogonal cutting process. As a first approximation, the following are assumed [11]: (i) all of the energy expended at the shear zone and along the tool face is converted to thermal energy, (ii) the energy at the shear zone and tool face is concentrated on a plane surface and is uniformly distributed, and (iii) no thermal energy is lost to the environment during the chip forming process.

A method to predict the rise in temperature \( \Delta T \) in the primary deformation zone is:

\[
\Delta T = \frac{\zeta}{\rho C_p} \int_0^{\varphi_{\text{ini}}} (\sigma) d\tilde{\varepsilon}
\]

where \( \sigma \) is the flow stress, \( \zeta \) is the fraction of the deformation energy converted to heat, \( \rho \) is the density of material, and \( C_p \) is the specific heat capacity of the material. Elmadagli and Alpas [11] used this method to estimate the temperature distribution in the primary deformation zone during the orthogonal cutting of Al 1100 and found a considerable rise in the temperature near the tool tip.

To use Eqn. 2.20, the equivalent plastic strain and equivalent flow stress \( \sigma \) have to be determined. Ni et al. [29, 30, 48], Pratibha [31], and Song [49] have done substantial work towards exploring the deformation of aluminum alloys during orthogonal cutting. The strain distribution in the deformed material ahead of the cutting tool tip was estimated using a metallographic method based on the observation of shear angles. Local flow stress values were estimated using microhardness measurements.

Zhang and Alpas [19] used the orientation change of the deformed grain boundaries on the cross-sectional plane to estimate the values of the shear angle \( \phi \) at different locations in front of the tool tip during orthogonal machining. The position of each individual grain boundary seen in Fig. 2.8 was determined and plotted to the actual
scale as shown in Fig. 2.9. The displacement fields were quantified by measuring the orientation changes due to the bending of the flow lines entering the primary shear zone. The values of the shear angles $\phi$, were computed from the slopes of the flow lines bending in the direction of plastic deformation using Eqn. 2.21:

$$\phi = \tan^{-1}\left(\frac{\Delta y}{\Delta x}\right)$$  \hfill (2.21)

A schematic diagram that shows how $\phi$ is defined is given in Fig. 2.10. Fig. 2.11 shows the distribution of the shear angles in the material ahead of the tool tip during orthogonal cutting of 6061 Al. The equivalent strains plotted in Fig. 2.12 that were developed in the material ahead of the tool tip were computed using Eqn. 2.22 from the shear angle distribution diagram shown in Fig. 2.11. The same method was also used by Elmadagli, et al. [29] to estimate the strains during the dry machining of commercial-purity copper and Al 1100, as shown in Fig. 2.13. The relationship between the equivalent plastic strain $\bar{\varepsilon}$ and the shear angle $\theta$ ($\theta = 90 - \phi$) is:

$$\bar{\varepsilon} = \frac{\sqrt{3}}{3} \tan \theta$$  \hfill (2.22)

Indentation hardness measurements done on a material can be used to find the flow stress of the material in front of the tool tip [18, 19].

$$H = C \sigma$$  \hfill (2.23)

where $H$ is the hardness value from the Vickers Hardness test (in MPa), $\sigma$ is the equivalent flow stress in compression, and $C=3.0$ is a constant.

Elmadagli and Alpas [30] have used microhardness measurements taken along regular intervals lying along the points of intersection of a grid superimposed on the
material ahead of the tool tip (Fig. 2.14) to characterize local plastic deformation taking place in Al 1100 during orthogonal cutting. To obtain an estimate of the stresses present in the matrix at these locations, flow stress values were calculated by using:

$$\sigma = H / C$$  \hspace{1cm} (2.24)

Finally, temperature increments can be calculated using the equivalent plastic strain and equivalent flow stress $\sigma$; the temperate distribution map showing local increments in temperature in the region ahead of the tool tip for an orthogonally cut Al 1100 sample is shown in Fig.2.15 [29]. The temperature increase ($\Delta T$) due to the conversion of deformation energy into heat within a unit volume of 1100 Al, with density $\rho = 2.71$ g/cm$^3$ and specific heat capacity $c = 904$ J/kg °C, can be estimated using:

$$\Delta T = \frac{\beta}{\rho C} \int_{\varepsilon}^{\varepsilon_{s}} \left(\frac{\varepsilon - (\sigma_s - \sigma_0) \exp\left(-\frac{\varepsilon}{\varepsilon_c}\right)}{\varepsilon_c}\right) d\varepsilon$$  \hspace{1cm} (2.25)

where $\beta$ is the fraction of plastic work converted into heat. It is generally assumed that $\beta = 0.95$. $\sigma_s = 299.0$ MPa is the saturation stress or the stress at which the work hardening rate becomes zero, $\sigma_0 = 138.8$ MPa is the flow strength of the material, and $\varepsilon_c = 1.37$ is a constant. They noted that the temperature steadily rises as one moves from the chip root to the tool tip, along the shear plane (Fig.2.15).

Shaw and Loewen [11] presented a method to estimate the temperature in the shear zone:

$$T_{sc} = \frac{\sigma_j}{\rho C_p} \left[ \frac{1}{1 + 1.328 \frac{k_y}{v_c d}} \right] + T_{amb}$$  \hspace{1cm} (2.26)
where \( \tau \) is the shear stress, \( \gamma \) is the shear strain, \( \rho \) is the density, \( C_p \) is the specific heat, \( k \) is the thermal diffusivity of the material, \( T_{\text{amb}} \) is the ambient temperature, \( v_c \) is the cutting speed and \( d \) is the depth of cut for the workpiece material. The temperature rise at the shear plane varies directly with the shear energy per unit volume going into the chips, and inversely with the volume specific heat of the workpiece (\( \rho C_p \)).

The mean temperature of the chip surface along the tool face (\( T_{\text{Tool}} \)) will be the sum of the mean shear-plane temperature (\( T_{SZ} \)) and the mean temperature rise due to friction (\( \Delta T_F \)) [11]:

\[
T_{\text{Tool}} = T_{SZ} + \Delta T_F = T_{SZ} + \frac{0.377 (R_2 q_2) a}{K_2 \sqrt{L_2}}
\]  

(2.27)

Fig 2.16 [11] is a comparison of the calculated average shear-plane temperature with the temperatures measured by the tool-chip thermocouple method for Al 2024. This model gives acceptable quantitative results.

Song [49] applied Eqn. 2.26 to estimate the temperatures in the shear zone for orthogonally turned aluminum alloy A380, to find that the heat produced during segmented or discontinuous chip formation is concentrated in the shear zones, with a negligible rise in the temperatures in the regions separating the shear zones.

2.1.6 Strain and Stress Distribution in Discontinuous Chips

2.1.6.1 Strain Measurements And Distribution

Song [49] investigated the dry machining of aluminum alloy A380, which produces discontinuous chips during orthogonal machining. The second phase particles, present in the area in front of the tool tip in the workpiece, had fractured and had aligned themselves along the direction of deformation and the flow of the material being sheared
away during the process (Fig. 2.17). The value of the shear angle (Φ) at each point within the workpiece was used to estimate the direction of the plastic flow at that point. By following the direction of the particle flow, the values of local equivalent strains $\varepsilon'''$ in the material ahead of the tool tip were estimated using following equation:

$$\frac{-\varepsilon''}{\varepsilon} = \frac{\sqrt{3}}{3} \tan \theta$$

(2.28)

The shear strain $\gamma_c$ in the shear band can be estimated by:

$$\gamma_c = \frac{d_s}{t_s}$$

(2.29)

Where $d_s$ is the shear displacement within the shear band, and $t_s$ is the thickness of the shear band (Fig. 2.18). The average length of $d_s$ was measured and found to be 480 µm, and the width of the shear band estimated from the optical and SEM micrographs was 40 µm. Therefore, the estimated shear strain in the shear band was 12.

The metallographically determined strain distribution in the material outside the shear band (using Eqn. 2.28) is shown in Fig. 2.19. Unlike the case of pure Cu and Al, which form continuous chips, no secondary deformation zone was found near the rake face. The strain in the chip was confined to the shear band. Within the segments between the shear bands, very little deformation occurred. Above the cutting line, the value of the strain increased from 0.2 at the tool tip to a value of 1.5 near the shear band. The strain in the primary deformation zone was in the range of 0.1 to 0.8.

2.1.6.2 Flow Stress Measurements And Distributions

Microhardness measurements were done in order to determine the local flow stress values in the workpiece ahead of the tool tip. Then, the flow stress was calculated by using Eqn. 2.24. Fig. 2.20 shows the stress distribution in the workpiece at the tool
tip. In the primary deformation zone, the stress gradually increased from a value of 220 MPa to 285 MPa above the cutting line. A decreased value of 246 MPa was observed in the shear band area, which indicates that thermal softening occurred in this area.

### 2.1.7 Surface Roughness, Tool Wear and Adhesion

Surface quality of the machined surface is one of the most important concerns in a machining operation. Surface roughness measurements and optically observed features on the finished surface give important information about the quality of the surface produced and hence about the machinability of the given alloy under the conditions used for the process. The surface roughness produced in turning depends on the feed rate, tool geometry, tool wear, and the material characteristics of the workpiece and the tool; it can be reduced by decreasing the feed rate and the depth of cut, increasing the cutting speed and the rake angle, and improving the workpiece material by adding free-machining additives or increasing the hardness [28].

Cutting tool wear may be classified as follows: adhesive wear, abrasive wear, diffusion wear, fatigue, delamination wear, microchipping, gross fracture, and plastic deformation [11]. The most frequently employed tool materials are: high speed steel (HSS), cobalt-enriched high speed steel (HSS-Co), sintered tungsten carbide (WC), polycrystalline cubic boron nitride (PCBN), polycrystalline diamond (PCD), and single-crystal natural diamond [28]. Diamond is of special interest because its properties are well suited for cutting-tool applications, such as extremely high hardness, high thermal conductivity, and low sliding friction.

Dry machining is possible for some aluminum alloys and cast irons. New coating materials can play a role in making dry machining practically feasible [11]. The dry
tapping of 319 Al was performed using diamond-like carbon (DLC) coated and uncoated HSS taps. HSS-dry tapping caused immediate tool failure within less than 20 holes due to aluminum adhesion, resulting in high forward and backward torques. DLC-dry tapping improved tool life and exhibited smaller torques. The low coefficients of friction (COFs) of DLC coatings against 319 Al limited the temperature increase during DLC-dry tapping to 75 °C. The low COF of DLC against aluminum was responsible for preventing BUE formation, and was thus instrumental in improving thread quality. The use of minimum quantity of lubrication (MQL) reduced the tapping temperature to 55 °C.

The assessment of the cutting performance of DLC coated HSS drills on 319 Al alloy, using 30 ml/h of distilled water (H2O-MQL), indicated that H2O-MQL, in conjunction with NH-DLC coatings, reduced the average torque and thrust force compared to dry cutting, and was comparable to conventional flooded drilling. DLC coatings were chosen because they showed low COF and low aluminum adhesion [50].

The best-known analytical model of tool wear rates [28] predict the wear rates due to adhesive, abrasive and diffusion wear for a carbide tool when cutting steel. The incremental volume \( dv \) of material worn away at a given point of the tool can be calculated by using:

\[
dv = C_1 \cdot q \exp \left[ -\frac{C_2}{\theta} \right] dL_s
\]

where \( dL_s \) is the incremental distance slid, \( q \) is the normal stress at the point in question, \( \theta \) is the interfacial temperature, and \( C_1 \) and \( C_2 \) are empirical constants. \( C_1 \) is proportional to the tool hardness [28]. This model is useful because it illustrates the importance of the tool hardness (through \( C_1 \)) and the interfacial temperature on the wear rate.
2.1.8 Machinable Alloys

Some alloys have secondary phases or inclusions that act as solid lubricants and thereby improve their machining performance. Gray cast iron normally gives best results when machined dry. The constituent graphite acts as a solid lubricant, as well as a diffusion barrier for carbon moving from the tool to the chip [11].

There are alloying elements that are added to steel to specifically to increase the machinability at conventional and high speeds [51]. These include lead, sulfur, manganese, sulfides, phosphorous, calcium, and bismuth. Typically, these additives, which are often used in combination, result in insoluble inclusions in the matrix. In addition to the content, the size, shape and distribution of the inclusions also affect the machinability. These inclusions cause the metal matrix to deform more easily and facilitate crack propagation, resulting in reduced cutting forces, enhanced chip breakability, and improved surface finish. The resulting grades are designated variously as free machining, free cutting, or enhanced machining steels [28]. For example, the sulfur is often added in the form of manganese sulfide, MnS, a solid lubricant, which forms inclusions in the matrix [52]. During cutting, MnS coats and lubricates the rake face of the tool, which reduces the friction, tool-chip temperatures, and tools wear rates. The inclusions also enhance chip breaking. However, depending on the amount of MnS, other mechanical properties of the steel such as corrosion resistance, ductility, toughness, formability, and weldability may be negatively impacted [28].

To avoid crater formation, a stationary or low-speed protective layer is essential to prevent the major tool-face shear energy from reaching the tool face. The secondary shear zone should be capable of acting not only as a thermal barrier but also as a
diffusion barrier. A highly oriented relatively low-melting point, layered structure of an inorganic inclusion (oxide, sulfide, phosphide, etc.), with the layers running parallel to the tool face, should be useful from all points of view [11].

2.2. Improvement of the Properties of Cast Al-Si Alloys

2.2.1 Introduction to Al-Si alloys

Al-Si castings constitute 85% to 90% of the total aluminum cast parts produced. They are widely used for engineering applications, especially in the transportation industries [53]. For example, aluminum castings have been applied to various automobile parts and constitute about 40% of wheels, brackets, brake components, suspensions (control arms, supports), steering components (air bag supports, steering shafts, knuckles, housings, wheels), instrument panels, and engine blocks [1]. Their high strength-to-weight ratio, good formability, good corrosion resistance and recycling potential make these alloys an ideal choice for light-weight components [1].

Al-Si alloys are classified into three categories based on their silicon percentage: (i) eutectic, (ii) hypoeutectic, and (iii) hypereutectic. Fig. 2.21 illustrates the binary phase diagram and the typical microstructures for these three cast Al-Si alloys. Si particles appear as platelets or needle-like particles in hypoeutectic Al-Si alloys A 380 and A 319, while hypereutectic A 390 contains both the block-like primary Si phase particles and platelet or needle-like eutectic Si particles.

The very low solubility of silicon in aluminum means that Al-Si alloys contain virtually pure α-aluminum with silicon as either a primary (in hypereutectic alloy) or eutectic (in hypoeutectic or eutectic alloys) phase, depending on whether the silicon percentage is greater or less than the eutectic point (12.6% Si), on the cooling rate, and
on the concentration of modifiers added. Results of ambient temperature mechanical tests demonstrate an increase in matrix microhardness and 0.2 % yield stress, and a decrease in the ductility with an increase in silicon content in aluminum (Table 2.2). Samples were solutionized for 1 hour at 529 °C, quenched in cold water and aged at 160 °C for various intervals of time before testing.

Grain refinement of the α-Al phase in Al–Si alloys, carried out by the addition of elements such as Ti and B, leads to a marked increase in the mechanical properties of these alloys. The yield stress and toughness are increased, and a favourable uniform distribution of the secondary phases in the microstructure is produced. In adequately modified structures, the eutectic silicon with an acicular morphology is converted into thin fibers, thus contributing to an increase in the ultimate tensile strength (UTS) and the elongation [4, 54-59].

Table 2.2 Properties of aged alloys with varying Si content [56].

<table>
<thead>
<tr>
<th>Material</th>
<th>0.2% YS (MPa)</th>
<th>UTS (MPa)</th>
<th>Ductility (%)</th>
<th>Microhardness (HV)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al-7Si</td>
<td>55.3±2.1</td>
<td>141.7±2.1</td>
<td>12.2±0.5</td>
<td>38.5±1.2</td>
</tr>
<tr>
<td>Al-10Si</td>
<td>75.4±1.6</td>
<td>154.7±3.4</td>
<td>10.3±0.8</td>
<td>39.2±0.4</td>
</tr>
<tr>
<td>Al-19Si</td>
<td>80.8±3.2</td>
<td>129.6±8.7</td>
<td>2.3±1.9</td>
<td>43.4±2.1</td>
</tr>
<tr>
<td>Al-19Si (Extruded)</td>
<td>82.7±3.1</td>
<td>189.0±12.1</td>
<td>21.4±8.8</td>
<td>59.2±0.5</td>
</tr>
</tbody>
</table>

The microstructure and properties of the Al-Si alloys are determined by (i) alloying elements (such as Mg, Cu) and impurity elements (such as Fe, Na), and (ii) melt treatment, solidification and heat treatment process. A multi-component Al–Si alloy (14.81 % Si, 2.06 % Cu, 1.58 % Ni, 0.92 % Mg, 0.33 % Fe, 0.02 % Mn by wt.) allows many complex intermetallic phases to form. Table 2.3 lists their elastic moduli and hardness. Si phase has the highest hardness.
Table 2.3 Hardness and elastic moduli of the different phases in Al-Si cast alloys by depth-sensing nanoindentation method [60]

<table>
<thead>
<tr>
<th>Phase</th>
<th>H (GPa)</th>
<th>E (GPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Si</td>
<td>11.13</td>
<td>147.59</td>
</tr>
<tr>
<td>AlFeMnSi</td>
<td>10.82</td>
<td>175.32</td>
</tr>
<tr>
<td>Al$_2$Cu$_4$Ni</td>
<td>9.25</td>
<td>163.66</td>
</tr>
<tr>
<td>Al$_2$Cu$_2$Mg$_8$Si$_6$</td>
<td>6.51</td>
<td>117.76</td>
</tr>
<tr>
<td>Al$_2$Cu</td>
<td>5.77</td>
<td>109.73</td>
</tr>
<tr>
<td>Al</td>
<td>1.45</td>
<td>84.6</td>
</tr>
</tbody>
</table>

The hypoeutectic (5.75 % Si by weight) and hypereutectic (16 % Si) have been chosen in this research to investigate the dry machining performance of Al-Si alloys.

2.2.2 Role of Alloying Elements

**Si:** The Al-Si binary system shown in Fig. 2.21 (a) forms a eutectic at 12.6 wt.% silicon at 577 °C. The latent heat evolved during the eutectic precipitation of Si from Al-Si alloys is nearly three times that of aluminum. Addition of silicon improves the fluidity and reduces shrinkage, and so the castability of aluminum alloys is improved. The silicon phase has the highest microhardness (Table 2.4) [61] and imparts superior properties. The presence of primary and eutectic Si particles, which are known for their high wear resistance, high stiffness and low thermal expansion, play an important role in restricting metal flow during machining.

Table 2.4 Microhardness testing results of the phases of an Al-11%Si alloy [61]

<table>
<thead>
<tr>
<th>Phases</th>
<th>α-Al</th>
<th>Si Phase</th>
<th>Al$_{15}$(Fe, Mn)$_3$Si$_2$</th>
<th>CuAl$_2$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Vickers (10g)</td>
<td>63.5±4.5</td>
<td>755±96.7</td>
<td>455.3±42.8</td>
<td>241.6±30.3</td>
</tr>
</tbody>
</table>

**Mg:** Magnesium reacts with the silicon during heat treatment to form a strengthening compound, magnesium-silicide (Mg$_2$Si), which precipitates out and resides in the grain boundaries of the alloy (Fig. 2.22). However, depending on the intended application for the alloy, the strength requirements may vary, which may accordingly
necessitate a change in the magnesium content. Mg additions of up to 0.4 wt.% to the Al-11.5% Si alloy increase the yield strength by 94 % and decrease the ductility by 40 %. Al₃Mg₆Cu₂Si₆ and Al₈FeMg₃Si₆ intermetallic forms (Mg > 0.65 %) are difficult to dissolve during solution treatments at 540 °C [62-64].

Mg forms intermetalics with Sn or Bi when its concentration is over 0.35 % by weight. TEM studies indicate that free Si and Mg₂Sn were present in as-cast structures of aluminum alloy 319, but Mg₂Sn alone is present in T4 and T7 heat treated samples. Mg₂Sn is formed in the as-quenched (T4) samples and shows high tendency to form Mg₂Sn, due to the rapid diffusivity of tin in the aluminum matrix [65].

**Fe, Mn:** Iron is a tramp element contained in aluminum, which is produced from bauxite ores that often contain ferric oxide. The presence of Al₅FeSi considerably decreases ductility and fracture toughness, and increases shrinkage porosity [66-67]. The formation of Al₅FeSi is minimized by rapid solidification [68] or the addition of transition elements such as Mn. When the iron content of the alloy is equal to or greater than 0.4 %, it is necessary that the weight ratio of manganese to iron be in the range of 1.2 to 1.75, and preferably below 1.5. When iron is present in an amount less than 0.4 %, the weight ratio of manganese to iron is in the suitable range of 0.6 to 1.2, provided that the manganese content of the aluminum alloy is at least 0.2 % by weight. Therefore, manganese is added to the Al-Si alloy to promote the formation of Al₁₅(Mn, Fe)₃Si₂ (Table 2.5), which has a compact morphology (indicated in **Fig. 2.21 (b), (c) and (d)**) and does not initiate cracks in cast Al-Si alloys to the same extent as Al₅FeSi does. For most casting methods, it is preferred that the iron content not exceed 0.8 % by weight of
the alloy. Al₅FeSi (platelet) and Al₈FeMg₃Si₆ intermetallics cannot be modified by heat treatment. Mn will also form MnAl₆ or Al₁₅(Fe, Mn)₃Si₂.

Chinese script-like phases, or compact polygonal particles of Al₁₂(Mn, FeCu)₃Si₂ decrease the machinability of 396 Al alloys, whereas, increasing the Fe-content from 0.5 % to 1 % promotes the formation of Al₅(Mn,Fe)Si in a monoclinic structure and produces a distinct improvement in the machinability of 319 alloys [26].

**Table 2.5** Sequence of phase precipitation in hypo-eutectic Al-Si-alloys [55]

<table>
<thead>
<tr>
<th>T (°C)</th>
<th>Phase precipitated</th>
<th>Suffix</th>
</tr>
</thead>
<tbody>
<tr>
<td>650</td>
<td>Primary Al₁₅(Mn, Fe)₃Si₂ (sludge)</td>
<td>Pre-dendritic</td>
</tr>
<tr>
<td>600</td>
<td>Aluminum dendrites</td>
<td>Dendritic</td>
</tr>
<tr>
<td></td>
<td>and Al₁₅(Mn, Fe)₃Si₂</td>
<td></td>
</tr>
<tr>
<td></td>
<td>and/or Al₅MnFeSi</td>
<td></td>
</tr>
<tr>
<td>550</td>
<td>Eutectic Al+Si</td>
<td>Eutectic</td>
</tr>
<tr>
<td></td>
<td>and Al₅FeSi</td>
<td></td>
</tr>
<tr>
<td></td>
<td>Mg₂Si</td>
<td>Co-eutectic</td>
</tr>
<tr>
<td>500</td>
<td>Al₂Cu and more complex phases</td>
<td>Post-eutectic</td>
</tr>
</tbody>
</table>

**Cu:** Copper behaves like magnesium and is often added as alloying elements to increase the strength and hardness of Al-Si alloys. It is believed that copper enhances the age-hardening effects of the alloy by precipitating fine particles of copper-aluminide (CuAl₂) homogeneously throughout the alloy [69]. Cu in the form of aluminide (Al₂Cu) improves the tensile strength, machinability and thermal conductivity at the expense of ductility.

**Ni:** The addition of nickel improves elevated temperature properties by forming stable intermetallics [70]. It is often present in commercial aluminum alloys and can be tolerated in amounts up to about 2 % by weight.

**Sn (0.2%):** Si tends to react with the magnesium to form Mg₂Sn. At low levels of magnesium (0.01 %), tin causes deterioration in the material properties; it decreases the
hardness in the as-cast condition. Minor additions of Sn, up to 0.1 % wt., increase the peak hardness of the alloys due to the enhanced formation of interdendritic structures. Elevated levels of Sn (~0.04 wt.%) and Pb (~0.03 wt.%) are found to lower the Brinell Hardness (HB) of the bulkhead after heat treatment. Therefore Sn and Pb levels must be kept below 0.00050 wt.% and 0.02 wt.%, respectively, to satisfy the bulkhead HB requirements of a V6 engine [71].

**Bi:** It has been determined that the presence of bismuth within the A 319 alloy enhances the lubricity of the alloy by essentially remaining as elemental bismuth within the alloy [14]. However, the amount of bismuth must also be limited since it tends to react with magnesium to form Mg₃Bi₂, which reduces the strengthening potential of the alloy. Therefore, it is desirable to limit the bismuth content to 1-2% (by wt.) within the preferred alloy [14]. This range of bismuth provides the enhanced lubricity while minimizing the loss in strength of the alloy.

### 2.2.3 Melt Treatments and High Cooling Rates

#### 2.2.3.1 Grain Refinement

There are many methods of grain refinement available, including agitation of the solidifying melt via mechanical or electromagnetic means, and the chemical or constitutional undercooling of the melt [72]. The common practice is by the addition of Al-Ti or Al-Ti-B master alloys as grain refiners. Grain refinement reduces the hot tearing tendency, and leads to a finer distribution of porosity. Aluminum alloys are grain-refined by the addition of typically 0.02 to 0.15 % Ti, or Ti-B mixtures in the range of 0.01-0.03 % Ti, and 0.01 % B. The microstructure of a commercial Al-Ti/B master alloy contains intermetallic phase TiAl₃, TiB₂ or (Ti, Al)B₂ which are known to be the effective nucleants for aluminum [55, 73].
The mechanism of grain refinement is achieved because Al-Ti master alloys contain numerous titanium aluminide crystals. When these are added to aluminum or to aluminum-based alloys, they act both to nucleate aluminum solid grains and to slow down the initial growth rate of aluminum crystals after nucleation. The resulting increased ratio of nucleation rate to growth rate gives a fine-grained casting. For Al-Ti-B master alloy, the (Al, Ti)B$_2$ phase assists the grain refinement either by acting directly as a nucleant, or by nucleating TiAl$_3$ crystals during cooling [55, 73].

2.2.3.2 Modification of Hypoeutectic and Eutectic Al-Si Alloys

The morphology of silicon plays an important role on the mechanical properties of finished products. The coarse silicon plates, which are of the unmodified acicular silicon structure, act as internal stress raisers in the microstructure and provide easy paths for fracture. Therefore, the eutectic silicon structure is modified to a fine fibrous structure by rapid cooling, by the addition of a chemical modifier, or by superheating the melt to ~850°C [74]. With modification, the structure becomes finer and the silicon more rounded, both of which contribute to somewhat higher values of ultimate strength and greatly increased ductility. Any process which reduces the size of the brittle phase particles or increases their separation will improve the properties of the alloy. Therefore, the common practice is to modify the as-cast flake or acicular silicon morphology by the addition of certain modifiers or by special heat treatment.

Four variables determine the microstructure of the modified Al-Si alloys that will form [73]: (i) type and amount of the modifier used, (ii) silicon content of the alloy, (iii) impurities present in the melt, and (iv) cooling rate.
The amount of each element required depends on the alloy composition—higher silicon contents require larger amounts of the modifying agent. Typical retained sodium levels are in the range of 0.005-0.01 %. Strontium, in amounts of 0.02 %, are sufficient to modify a 7 % Si alloy, such as the A 356, but up to 0.04 % is needed for a eutectic alloy such as the A 413. Alloys that are easy to modify have low phosphorus contents. The effect of Bi-Sr interaction on the Al-Si eutectic reaction using thermal analysis has been examined [55]. When the Sr/Bi ratios is below 0.2, the Si structure remains unmodified and the eutectic temperature remains high, whereas at ratios between 0.2 and 0.45, the eutectic temperature drops down and the modification of the Si structure increases dramatically. Quenching experiments have been carried out on cast alloys. Silicon nucleates and grows ahead of the solidification front, greatly increasing the effective area of the front, in the conditions used in this experiment. When Na was added, the solidification front of Si was straighter, reducing its area. As the heat transferred out of the casting was unchanged, the velocity of forward motion of the front was increased. This increase in speed, equivalent to a chilling effect, accounted for the finer structure of the eutectic [74, 75].

According to the widely accepted impurity-induced twinning theory [73], Na or Sr is absorbed on the growing Si crystal surfaces, and thus, the crystal growth is restricted. This leads to the forced twinning of the Si crystal, enhanced branching, and a fibrous microstructure. In the presence of a chemical modifier, both the twinning frequency and the angle of branching increase. The modifying elements, such as Sr, must have an affinity to silicon, which will facilitate its absorption on to the surface of the silicon crystals that are growing in an aluminum melt (Fig. 2.23).
2.2.3.3 Modification of Hypereutectic Al-Si Alloys

The liquidus of the hypereutectic Al-Si system is very sharp, so that the casting temperatures and the solidification ranges of the alloys increase with increasing the silicon content. This factor limits the maximum silicon content of the hypereutectic alloys to 20 wt.%, above which it is difficult to obtain sound castings. The primary silicon in hypereutectic alloys is usually coarse (Fig. 2.24) and imparts poor mechanical properties to these alloys. A non-uniform distribution of the primary silicon particles in cast Al-Si alloys creates large areas in the microstructure that are free of silicon particles, which cause the initiation of local wear of the aluminum matrix [73]. Large and unevenly distributed particles cause greater tool wear than do smaller and more uniformly distributed primary phases, and they depreciate the machinability of the alloy [73]. Thus, fine and evenly distributed primary silicon is an essential feature of cast hypereutectic alloys.

Phosphorus is deliberately added to the melt because it reacts with aluminum to form AlP particles which nucleate primary silicon [73]. These result in a fine dispersion of primary particles (Fig. 2.25), an effect that is quite the opposite of that in hypoeutectic alloys, where the silicon is coarsened by phosphorus addition. The melt can be modified through the addition of master alloy Cu–15 wt.% P (0.02 wt.% P in the melt) at near 800 °C, to refine the primary silicon crystals. The mechanism of silicon modification is that phosphorus reacts with the liquid aluminum to form aluminum phosphide, AlP, which has a crystal structure very similar to that of silicon, and thus acts as an effective heterogeneous nucleant[73].
High cooling rates also can refine the silicon particle size. Fig. 2.26 presents the cooling rate and a micrograph of the Al–20 % Si alloy test sample solidified at different cooling rates [76]. Si particles were refined to ~10 μm at a high cooling rate of 83 K/s.

2.3. Machining of Aluminum Alloys

Increasing environmental concerns over the use and disposal of coolants and lubricants, and their high costs, mandate the reduction of coolant and lubricants used during machining. The increasing thrust towards dry machining has seen aluminum alloys frequently being attempted to be machined without lubricants, however, their high thermal conductivity leads to unwanted deformation of the surface produced and the increased adhesion of workpiece material to the tool [10]. This makes the surface characteristics that are obtained without use of lubrication a continued subject of concern for the industry.

2.3.1 Machining of Wrought Aluminum Alloys

Ni et al [29] studied the microstructure generated during the orthogonal cutting of Al 1100. Samples were examined under TEM. A schematic diagram summarizing the major aspects of the evolution of deformation microstructures in the material ahead of the tool tip during the orthogonal cutting of 1100 Al samples is shown in Fig. 2.27. The small elongated subgrains, resulting from the initial grain structure fragmenting into smaller units at relatively lower strains, are present at the lower boundary of the primary deformation zone, with the boundaries between these units accommodating the lattice misorientation accompanying the process. Elongated subgrains, with interiors that are almost dislocation free, and smaller equiaxed grains, are formed as the strains are further increased. Evidence for recrystallization in the zone was also found in the form of very
small grains (Fig. 2.28). The chip microstructure was also investigated to reveal elongated grains and a high volume fraction of small equiaxed grains, with the same aspect ratio as those formed in the primary deformation zone, which showed that there was no further refinement in grain size after passing through the primary deformation zone, even as the plastic strains increased in magnitude from around 0.8 in primary deformation zone to 2.3 in the chip. Based on the microstructural evidence, it was concluded that dynamic recrystallization started as strains rose to about 1.0 and was completed in the secondary deformation zone, where evidence of extensive grain growth was found.

Machining of aluminum alloys under lubricated conditions normally generates cutting forces below 1200 N [10, 46]. The high thermal conductivity of aluminum alloys and their low melting temperatures are important concerns during the machining of these alloys, as they may lead to the plastic deformation of the material being machined, due to the heat generated during the process [10]. They may also lead to high residual stresses being introduced in the component due to the non-uniform heating of the different parts of the component being machined.

A wrought alloy A 2011 containing Bi and Pb is a commonly used free machining grade alloy. The alloy A 6262 is the easiest to machine in the 6xxx series, because it contains Bi that has been added specifically to improve machinability [28]. A 6061 and A 1100 produce very long strings of ribbon-like chips during machining [10, 46], which pose a serious hazard during the process, as they have a tendency to get tangled around the cutting tool and machinery. As the tangled chip does not break off easily on its own when the process is underway, it leads to an unwanted increment in the power expended during the process.
Zhang and Alpas [19], and Elmadagli and Alpas [29, 30] estimated the equivalent strains from the orientation change of the extrusion lines or flow lines during the orthogonal cutting of 6061 Al and 1100 Al, and they found that the strains are non-linearly distributed in the primary deformation zone and the adjoining areas, which proves that the single shear strain values predicted by Shaw [11] for fixed rake angle and cutting velocity is an over-simplification of the problem at hand. They calculated the equivalent strain along the distance from the cutting line, with the largest strains occurring in the material at the rake face, 300 µm above the tool tip, in the 6061 Al alloy during an orthogonal cutting process.

List et al. [77] used a WC-Co tool for the orthogonal cutting of the aluminum alloy 2024-T351 under dry conditions. The formation of a BUE always creates a new geometry in which a high contact pressure leads to adhesion through the interlocking of asperities. S´anchez et al. [78] studied BUE formation on the tool edge as well as the development of built-up layer (BUL) on the tool rake face during the dry turning of AA2024 and AA7050 alloys. They concluded that once the BUL initially develops, the aluminum accumulated on the tool surface reduces its early hardness and raises its thermal conductivity, thus causing a decrease in the temperature achieved in the subsequent states of the cutting process. This prevents the initial melting of the bulk alloy and increases the BUL thickness.

Ni et al. [29] used a metallographic technique to study the strain distribution in the deformed material ahead of the cutting tool tip and in the primary deformation zone (PDZ) during the orthogonal cutting of 1100 Al. The machined chips consisted of a mixture of elongated sub-grains and recrystallized equiaxed grains, while the
microstructure of the secondary deformation zone (SDZ) contained large equiaxed grains. It was concluded that dynamic recrystallization started as strains rose to about 1.0 and was completed in the secondary deformation zone where evidence of extensive grain growth was found.

Song et al. [49] observed the deformation microstructures of the material ahead of the tool tip during the orthogonal cutting of 1100 Al under feed rates of 0.25 and 0.80 mm/rev. They developed a stress-strain relationship using a Voce-type exponential equation. A saturation flow stress of 302 MPa was obtained for 1100 Al.

2.3.2 Dry Machining of Cast Aluminum Alloys

Hypoeutectic alloys generally contain less than 12 % silicon and are easy to machine at low speeds, as the silicon present in them is in the eutectic phase, and primary silicon is absent. However, machining may be accompanied by occurrence of BUE in some cases [10, 46].

Hypereutectic aluminum alloys have a silicon content of more than 12 %. They have large primary silicon particles, which lead to abrasive wear of the tool used for machining, with the degree of abrasion rising with an increase in the silicon content in the alloy [10]. In addition to the cutting forces being high for these alloys, they cannot be machined using carbide tools, which are more economical than diamond-based PCD tools, because the larger silicon particles lead to faster tool wear.

Machinability is influenced by the amount of alloying elements (strontium, iron, magnesium, copper, bismuth, indium, lead and/or tin) and heat treatments; some observations are [73]:


(i). Sr exerts positive effects on the machining properties because of the refined Si particles produced.

(ii). Tool life is a strong function of the iron content. Tool wear is directly proportional to the silicon content of the alloy, the silicon grain size, and the loading conditions on the tool [73].

(iii). A low Mg content (Mg-intermetallics) yields the longest tool life.

(iv). Hardness has a pronounced effect on the formation of BUE for the Al-Si alloys. Heat treatments that increase the hardness also reduce the built-up-edge on the cutting tool [79, 80].

(v). Additions of small but effective amounts of one or more of bismuth, indium, lead and/or tin to an aluminum casting alloy markedly improve the dry machinability of a casting made from the modified alloy [15].

(vi). The secondary dendrite arm spacing (SDAS) is an important microstructural feature of Al 319 that has a significant effect on the cutting forces. For traditional materials such as cast iron, grain size is usually the key microstructural factor that affects the machinability of the material [73]. But one study concludes that not grain size but SDAS is the dominant microstructural feature in terms of machinability for cast Aluminum alloys like Al 319 and Al 356 [81], and the SDAS is related to the solidification rate.

2.3.2.1 Stress-Strain Distributions in the Workpiece

Song [49] investigated the dry machining of the aluminum alloy A 380, which produced discontinuous chips during orthogonal machining. The equivalent plastic strain of 1.5 near the shear bands, the shear strain value of 12 in the shear band, and the flow
stress of 285 MPa can be regarded as initial failure prediction values. No secondary deformation zone was found. During the dry machining of an aluminum alloy, it was suggested that the tool wear was mainly caused by the formation of an adhesive aluminum layer and built-up-edge [77], both of which greatly affect the quality of the machined surface.

Pratibha [31] studied the dry machining performance of cast 319 Al. Plastic strain and flow stress estimations in the region ahead of the tool-tip were high, specifically 1.8 and 369 MPa respectively, in shear zone. Experimental observations indicated that the 319 Al alloy possesses a good degree of machinability under dry conditions.

2.3.2.2 Effects of Tool Coatings

Bhowmick [8] studied the performance of hydrogenated and non-hydrogenated diamond-like carbon (DLC) tool coatings during the dry drilling of 319 Al. Dry drilling performed by the H-DLC coated drill produced consistently lower average torques and thrust forces when compared to NH-DLC drills. Other indicators, such as the mass of adhered aluminum and the number of torque spikes, were less significant during drilling with H-DLC drills. DLC coated drills are superior to HSS drills when drilling 319 Al under ambient conditions without using metal removal fluids. Under the same conditions, among the DLC coatings, hydrogenated DLC had advantages over the non-hydrogenated DLC coatings in terms of keeping the cutting forces low and limiting adhesion.

Basavakumar et al. [82] studied the influence of melt treatments (grain refinement and modification) and turning inserts (un-coated, PVD and polished CVD diamond-coated) on the cutting force and surface integrity when turning Al-12Si and Al-12Si-3Cu cast alloys under dry conditions using a lathe machine. They found that the combined
addition of a grain refiner and a modifier (Al–1Ti–3B-Sr) to the Al–12Si–3Cu cast alloy, lowered the cutting forces and improved the surface finish, in contrast to untreated alloys. They also found that the performance of the polished CVD diamond-coated insert (compared to PVD and un-coated inserts) resulted in lower cutting forces and workpiece surface roughness. Also, the tendency for material build-up at the cutting edge was reduced when a polished CVD diamond-coated insert was used.

2.3.2.3 Near-Dry Machining

A few results based on the examination of the near-dry machining of aluminum alloys have been reported and these are summarized below.

Klocke et al. [83] studied the effect of a minimum quantity of lubricant with a flow rate of 10 ml/h on the drilling of cast aluminum alloys (AlSi9Cu3). In terms of feed force, torque and hole diameter, there were no significant differences. The smallest variations were found in the cases of TiAlCN + Me-C:H and TiAlCN + MoS2 coatings. The holes displayed better surface roughness with MQL (mean peak-to-valley height of 15-35 µm) when compared to dry cutting (mean peak-to-valley height of 25-65 µm).

Bardetsky et al. [84] concluded that dry machining produced the highest degree of aluminum adhesion with a significant amount of the workpiece material adhering to the flank, clearance and rake faces during the milling of 319 Al. On the other hand, in the case of MQL (10 ml/h) milling, moderate adhesion (discontinuous areas) was found on the flank, rake and clearance surfaces. The highest resultant cutting force was found during dry machining due to the significant aluminum adhesion, while tool wear was highest in dry machining due to the absence lubrication.
Kishawy et al. [85] investigated the effects of the type of coolant application (flooded lubrication, minimum quantity of lubrication) on the tool wear, chip morphology, surface quality and cutting forces during the milling of A356. The authors stressed that the MQL technology (flow rate of 30 ml/h) could be a viable alternative to flooded lubrication. Due to the presence of silicon in the A356 alloy, the primary mechanism of wear was abrasive, which was found to occur at the tool tip region. Adhesive wear due to material adhesion was found on the flank and rake faces of the cutting tool. The lowest flank wear was observed in MQL cutting.

Bhowmick [86] studied the effect of minimum quantity lubrication (MQL) on the cutting performance of uncoated and DLC-coated HSS tools. He found that, like the H-DLC coated drill, lower torques and thrust forces were observed with triglyceride lubrication when compared to dry and water conditions. The formation of BUE at the cutting edge of the drill and the transfer of aluminum to the drill flutes significantly reduced under triglyceride lubrication conditions.

2.4. Effect of Low Melting Point Elements on the Machining of Al Alloys

2.4.1 Selection of Free Machining Constituents in Aluminum Alloys

Free machining Al alloys typically include elements such as lead, tin, and bismuth (Table 2.6) for improved machinability by promoting the formation of shorter chips. These problems have been studied since the early 1950's. The presence of a low melting eutectic constituent in the microstructure of an aluminum alloy seems to be a viable way to form small chips. In fact, machining of AA6262 (0.4-0.8 wt % Si, 0.8-1.2 wt % Mg, 0.15-0.4 wt % Cu, 0.4-0.7 wt % Pb, 0.4-0.7 wt % Bi and the balance of Al) is improved by the addition of a Pb-Bi eutectic [28].
Although the addition of low melting point elements to aluminum alloys can improve their machinability, this causes hot shortness by the low melting metals [28]. A hot tear is one of the most serious defects that a casting can suffer. Low-melting second phases could weaken grain boundaries by depressing the solidus temperature of the metal and prolonging liquid film life. Long freezing range alloys are more susceptible to hot tearing due to the fact that they tend to remain in the semi-solid state for a considerable period after pouring, and thereby permit any tensile/shear stresses arising due to contraction restrains to propagate cracks/tears. Hot tearing causes intergranular cracks: the coarser the grain size, the more the tendency to hot tear.

The shape of the grain boundary particles (Fig. 2.29) is largely controlled by the relative surface energies of the grain-to-grain interface itself, $\gamma_{gg}$, and the grain-to-particle interface, $\gamma_{gl}$ [87]. The balance of forces is:

$$\gamma_{gg} = \gamma_{gl} \cos \theta$$

(2.31)

For most values of the equilibrium dihedral angle $2\theta$, the grain boundary liquid assumes compact shapes.

If a low melting point element is to be a suitable chip-breaking additive, it should meet the following requirements:

(i) The solubility of the additive in the solid matrix must be low, and must cause granular micro-segregation in the aluminum alloy.

(ii) It should not form a high melting point intermetallic compound with the base metal nor with the other important alloying element.

(iii) The liquid inclusions must not wet the grain boundaries (their dihedral angle should be larger than $60^\circ$, otherwise a continuous network of liquid would be
present along the grain edges and the alloy would be severely hot-short). For example, in Al-Pb alloys, the liquid Pb phase wets the grain boundaries during the failure process.

In Al-Cd alloys, the liquid Cd at the grain boundaries does not wet and spread, but remains as compact globules. These alloys therefore fail by a more ductile type of fracture. However, Cd cannot be added as an alloying element because of its toxic properties. Also, indium is more expensive than the other low melting point elements, and therefore, bismuth and/or tin are the preferred additives over Pb, In and Cd.

**Table 2.6 Properties of bismuth, indium, lead and tin**

<table>
<thead>
<tr>
<th>Element</th>
<th>Atomic number</th>
<th>Atomic weight</th>
<th>Melting point °C</th>
<th>Boiling point °C</th>
<th>Specific gravity</th>
</tr>
</thead>
<tbody>
<tr>
<td>Bi</td>
<td>83</td>
<td>208.98</td>
<td>271.3</td>
<td>1560</td>
<td>9.747</td>
</tr>
<tr>
<td>In</td>
<td>49</td>
<td>114.82</td>
<td>156.61</td>
<td>2000</td>
<td>7.31</td>
</tr>
<tr>
<td>Pb</td>
<td>82</td>
<td>207.19</td>
<td>327.5</td>
<td>1744</td>
<td>11.35</td>
</tr>
<tr>
<td>Sn</td>
<td>50</td>
<td>118.69</td>
<td>231.89</td>
<td>2270</td>
<td>7.31</td>
</tr>
<tr>
<td>Cd</td>
<td>48</td>
<td>112.40</td>
<td>320.9</td>
<td>765</td>
<td>8.65</td>
</tr>
</tbody>
</table>

Under suitable machining conditions, it has been suggested that the Bi, Sn and Pb act as internal lubricants by forming a viscous layer, and also as a diffusion barrier between the tool and the chip, thereby prolonging tool life and improving productivity. Their globules do not adversely affect the strength or the hardness (Table 2.7) [15] of the casting but enable surfaces of the casting to be machined without the use of a coolant or a lubricating machining fluid.

**Table 2.7 Properties of the B319 alloy with varying amounts of bismuth** [14].

<table>
<thead>
<tr>
<th></th>
<th>Microhardness (Brinell)</th>
<th>Tool Life (no. of holes)</th>
<th>Power (Kw)</th>
<th>Torque (Nm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Conventional B319</td>
<td>74 to 80</td>
<td>12</td>
<td>2.8</td>
<td>2.6</td>
</tr>
<tr>
<td>B319 + 0.2% Bi</td>
<td>667</td>
<td>&gt;5000</td>
<td>2.8</td>
<td>2.0</td>
</tr>
<tr>
<td>B319 + 0.5% Bi</td>
<td>74</td>
<td>&gt;5000</td>
<td>1.9</td>
<td>1.5</td>
</tr>
<tr>
<td>B319 + 1.0% Bi</td>
<td>5000</td>
<td>&gt;5000</td>
<td>1.8</td>
<td>1.2</td>
</tr>
</tbody>
</table>
2.4.2 Al-Sn System

Tin has been used as a solid lubricant in automotive engine bearing alloys. Tin smear over the bearing surface and forms a protective film, and thus protects the aluminum bearing surface from direct contact with the iron shaft surface. It also provides the alloy with a soft phase and gives the alloy good embeddability, conformability and compatibility.

According to the aluminum-tin equilibrium phase diagram (Fig. 2.30), the tin solubility in solid solution at 900 K is 0.1 %, while at the eutectic temperature it reduces to 0.005-0.07 %. Tin with aluminum forms a eutectic system at a low eutectic temperature of 501-502 K and there is a strong deviation of the eutectic point towards tin (99.5 % Sn). Problems in the solidification of Al-Sn alloys arise from their large freezing range, which together with the wide density difference between the two components, greatly increase the possibility of tin segregation during solidification. It has therefore been difficult to introduce and uniformly disperse tin in aluminum to the desired extent by conventional melting and casting techniques. Tin additions above 1 % significantly affect the castability of these alloys negatively, due to the large increase in the alloy solidification range.

All parameters causing grain refinement, as well as the decrease of inter-dendritic distances, will also result in producing a more even distribution of tin. Grain refinement can reduce the hot tearing tendency of alloys. Alloys cast in metal moulds with higher cooling rates possess better mechanical properties than those in resin-coated sand moulds.
2.4.3 Al-Bi System

The aluminum-bismuth equilibrium phase diagram (Fig. 2.31) contains a monotectic reaction followed by a domain of liquid immiscibility. Thus, there is a monotectic reaction with a region of immiscibility below 1310 K which extends down to 930 K from 0.002-84.0 at.% Bi, and a eutectic reaction which extends from 0-100 at.% Bi at 543 K in the phase diagram. Thus, a hyper-monotectic alloy liquid is separated into globules of bismuth dispersed in a matrix of aluminum, upon cooling from a temperature above the critical line bounding the immiscible domain. The Bi-Sn system has a eutectic reaction at a low temperature of 411.5 K (Fig. 2.32).

The solidification of Bi particles takes place over a broad range of temperatures. It has been indicated that the average size of the Bi-rich particles increases with an increase in the Bi content. Higher casting rates result in immiscible alloys with Bi particles dispersed in the microstructure.

Bismuth decreases the surface tension because it is surface-active, and is therefore found at the grain boundaries [88]. The solidification of alloys from a system with a region of two-liquid immiscibility can be divided into three categories depending on the alloy composition: (a) alloys with gross composition below the monotectic one, (b) alloys of exact monotectic and (c) alloys of hypermonotectic composition. The solidification of Al alloys with 0.5 wt.% Sn or 0.5 wt.% Bi belong to the solidification of category of the first type, and can be treated as the classical case of miscible alloys during solidification processes [89].
2.5. Summary of the Literature Survey

The literature survey above indicates that the mechanical models for continuous and discontinuous chip formation are different. The models for predicting temperature and temperature distributions are also different. The force relationship during orthogonal turning is usually based on the simple shear plane model. The bulk of the energy evolved in metal cutting is consumed along the shear plane.

The composition, melt treatment, and cooling rate during solidification have a strong effect on the size and morphology of Si and other intermetallics during Al-Si casting. Cu, Bi, and Sn improve the machining performance, while Si, Fe, Mg, and Mn deteriorate machining performance because of the high hardness values of the intermetallics formed, and their high melting points. Grain refinement, modification and high cooling rates can refine Si particles and also the other intermetallic phases in Al-Si alloys, thereby improving their machining performance.

Al-Si alloys of type 319 possess a good degree of machinability under dry turning conditions, and form discontinuous chips. The dry turning of cast Al-Si alloys may be possible with the new DLC coated tool at low cutting speeds, but the chips that are formed during turning tend to adhere to the tool surface and cause BUE. Bismuth and tin are preferred additives for improving the dry machining performance of Al-Si alloys. Under suitable machining conditions, it has been suggested that the Bi and Sn act as internal lubricants by forming a viscous layer and they also act as a diffusion barrier between the tool and the chip. Understanding the effects of low melting point additives on the microstructure, the mechanical properties, and the dry machining performance of
Al-Si alloys, cooled at different rates during casting, is necessary to optimize their dry turning performance and casting parameters.
Fig. 2.1 Force resolution in the cutting zone for orthogonal cutting.
(a) Geometrical force resolution in the cutting zone for orthogonal cutting with a continuous-type chip, (b) physical force resolution in the cutting zone for orthogonal cutting with a continuous-type chip [27].
Fig. 2.2 Merchant and Piispanen idealized ‘Stack of Cards’ Model. Illustrating mechanism of formation of continuous chip and the resulting deformation, where $\Delta s$ is the distance between the card and its neighbour, $\Delta x$ is the thickness of the card, $\phi$ is the shear angle [32].

Fig. 2.3 Deformation texture of the chip preferred orientation of grains. $\phi$ is the shear angle and $\psi$ is the angle between the directions of shear and elongation of the crystal structure [32].
Fig. 2.4 Schematic outlining curled chip model for orthogonal cutting. Proposed by Dewhurst, where $\phi$ is the shear angle and $\Delta s$ is the distance with respect to the neighbour [35].

Fig. 2.5 Shear Zone Model proposed by Hitomi and Okushima. AOB is the large transitional zone for plastic deformation, $\beta$ is the friction angle of the tool face, $a$ is the contact length between the chip and the tool, and R is the resultant force on the tool [38].
Fig. 2.6 Formation of discontinuous chips when cutting brittle material. The numbers beneath successive stages are frame numbers from high speed films [40].

Fig. 2.7 Schematic diagram of sectioned orthogonally cut sample showing important deformation zones. $\Delta_1$ is the deformation depth, and $\Delta_2$ the distance between the cutting edge and the deformation zone III [27].
Fig. 2.8 Optical micrograph of a section through a machined chip of 6061 Al attached to the work piece. Cutting speed=0.6 m s$^{-1}$, feed= 0.30 mm, rake angle=−5° [19].

![Optical micrograph of a section through a machined chip of 6061 Al attached to the work piece.](image)

Fig. 2.9 A computer-generated image of the cross-section of the machined work piece with the chip still attached. The image was obtained by determining the location of each point on the deformation lines and these were plotted to actual scale shown on the diagram [19].

![Diagram of a computer-generated image of a cross-section of a machined work piece.](image)
Fig. 2.10 Schematic diagram showing the method to measure the shear angles from the slope of the deformation lines [19].

Fig. 2.11 Shear angle distribution diagram for an orthogonally cut 6061 Al sample. Showing the magnitude of shear angle at different locations in the material in front of tool tip [19].
Fig. 2.12 Strain distribution diagram in orthogonally cut 6061 Al [19].
Fig. 2.13 Strain distribution diagram in Al 1100 in the region ahead of the tool tip [29].
Fig. 2.14 Variation of microhardness of Aluminum 1100 ahead of the tool tip [29].
Fig. 2.15 Temperate (°C) distribution map showing local increment in the region ahead of the tool tip for an orthogonally cut Al 1100 sample [29].
Fig. 2.16 Temperature comparison of calculated with measured for 2024 Al (a) calculated average shear-plane temperature, (b) measured by tool-chip thermocouple method using WC tool, this gives acceptable quantitative results [11].
Fig. 2.17 Schematic drawing of equivalent plastic strain measurement [49]

Fig. 2.18 Schematic drawing of the shear displacement within the shear band. Shear band $d_s$ and the thickness of the shear band $t_s$ [49]
Fig. 2.19 Strain distribution of A380 (solution treated) ahead of the tool tip [49]

Fig. 2.20 Stress distribution (in MPa) of A380 (solution treated) ahead of the tool tip [49]
Eutectic Si Al\textsubscript{15}(Fe,Mn)\textsubscript{3}Si\textsubscript{2}

Al\textsubscript{5}Mg\textsubscript{8}Cu\textsubscript{2}Si\textsubscript{6}

Al\textsubscript{2}Cu

(a)

(b)
Fig. 2.21 Al-Si binary phase diagram and typical microstructures
(a) Al-Si binary phase diagram [54], (b) hypoeutectic 319, (c) near eutectic 380, and (d) hypereutectic B 390 alloy [55].
Fig. 2.22 Liqidus surface at the aluminum-rich corner of Al-Mg-Si ternary phase diagram
The pseudo-binary line for Al-Mg$_2$Si is shown [62].

Fig. 2.23 Absorption of impurity atoms on growth steps of a silicon crystal causes twinning to occur [73]
Fig. 2.24 Silicon particles in a sample of alloy 390 (~15%Si). Unetched, X56 [73].

Fig. 2.25 Silicon particles in a sample of alloy 390 (~15%Si) after addition of 0.03%P. Unetched, X56 [73].

Cooling curves of (a) ~ (d)
Fig. 2.26 The cooling rate and micrograph of the Al–20% Si alloys solidified at different CR
(a) CR of 4.9 °C/s. (b) CR of 14.7 °C/s. (c) CR of 23.9 °C/s. (d) 52.7 °C/s. (e) of 82.9 °C/s.
Average chemical composition of the hypereutectic Al–20% Si alloy (wt%): 20.0Si, 3.0Cu,
0.5Mg, 0.1Zn, 0.5Fe, 0.1Mn, 0.1Ni, 0.001Ti, 0.01P [76]
Fig. 2.27 Evolution of microstructure in Al 1100 during its orthogonal machining
(a) Schematic diagram; (b) Sequence of grain refinement events taking place in the material ahead of the tool tip [29].
Fig. 2.28 TEM micrographs of machined chips of Al 1100
(a) showing elongated subgrains and small equiaxed grains formed during machining; (b) Higher
magnification TEM micrograph showing nano-scale equiaxed grains [29].
Fig. 2.29 The shapes of the liquid phase at the grain boundary as a function of dihedral angle [87]

Fig. 2.30 Al-Sn phase diagram [88]
Fig. 2.31 Al-Bi phase diagram
Inserts show Al-rich and Bi-rich sides, respectively [88].
Fig. 2.32 Bi-Sn phase diagram [88]
CHAPTER 3    Experimental Details

3.1. Introduction

In this chapter, the experimental methods that were used in this work are outlined, namely casting, machining, metallographic analysis and mechanical testing. Cast Al-Si alloys (hypoeutectic and hypereutectic), which have extended applications in automotive components, were cast and cooled under different rates. Dry turning experiments that were performed on these different melts are described here in detail. The temperature generated between the chips and tool rake face, the chip morphology, the cutting and thrust forces during dry turning were systematically measured. The mechanical properties of the alloys were also experimentally determined.

The casting facility and casting process of the Al-Si alloys that were employed in this work are discussed in Section 3.2. The methods used to perform metallographic analyses and mechanical property tests are discussed in Sections 3.3 and 3.4, respectively. Cutting parameters, and the cutting tools and samples used for the dry turning tests are described in Section 3.5. The methods used to measure the cutting forces and the thrust forces in the dry turning tests are presented in Section 3.6. The method of measuring the temperature during dry turning is described in Section 3.7. Analytical methods employed in the characterization of the chip morphology and cross-sections of chips, the surface roughness and the TEM observations of the chips are described in Sections 3.8-3.10.
3.2. Casting Process

3.2.1 The Casting Facility

The casting mould and crucible used in the experiments are shown in Fig. 3.1. A permanent steel mould \((120 \times 120 \times 25 \text{ mm}^3)\) was made using two blocks of carbon steel, and a chilled copper mould \((120 \times 25 \times 25 \text{ mm}^3)\) was made by machining an electrolytic copper block \((99.9 \% \text{ Cu}, \text{C11000 grade})\). The crucible used for melting the aluminum was a ladle liner with lip purchased from Engineered Ceramics Company of 4 kg capacity. The crucible holder was made of stainless steel and was used to hold as well as pour the melt during casting.

Figure 3.2 shows the general layout of the casting facility. The furnaces used were Lindberg/Blue M and CF56822C (maximum temperature \(1200 \degree \text{C}, 2.6 \text{ kW}, 240 \text{ V}\)). Two furnaces were used together, one for preheating the steel mould, and the other for melting the aluminum alloy, with an Ar gas environment to avoid hydrogen absorption into the molten aluminum. Ar gas was injected into the furnace through a 6 mm diameter stainless steel pipe.

3.2.2 Raw Materials

Table 3.1 lists the composition of the raw materials used in this work. Melts were prepared by using 319.2 primary Al alloy (supplied by ALCAN Inc.). Al – 50 \% Si, Al – 25 \% Mn and Al – 20 \% Ni aluminum-based master alloys (in waffle shape), shown in Fig. 3.3, were used to increase the Si, Mn, and Ni contents. Fig. 3.4 shows the shape of the extruded rods of Al – 10 \% Sr alloy that were used for the hypoeutectic Al-Si alloy modification. Extruded rods of Al – 5 \% Ti – 1 \% B were used for grain refinement. A Cu – 8 \% P alloy was used for the hypereutectic Al-Si alloy modification. The
microstructures of the grain refiner and modifiers (purchased from Milward Alloys, Inc.), namely Al – 5 % Ti – 1 % B, Al – 10 % Sr and Cu – 8 % P are shown in Fig. 3.5. Al₃Ti and TiB₂ were present in the Al – 5 % Ti – 1 % B alloy, Al₄Sr in the Al – 10 % Sr alloy, and Cu₃P in the small shot Cu – 8 % P alloy (Table 3.1).

Table 3.1 Composition of raw materials

<table>
<thead>
<tr>
<th></th>
<th>Si</th>
<th>Cu</th>
<th>Fe</th>
<th>Mg</th>
<th>Zn</th>
<th>Ni</th>
<th>Mn</th>
<th>Ti</th>
<th>P</th>
<th>B</th>
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<td>319.2</td>
<td>5.75</td>
<td>3.09</td>
<td>0.17</td>
<td>0.02</td>
<td>0.02</td>
<td>0.01</td>
<td>0.01</td>
<td>0.1</td>
<td>0.01</td>
<td></td>
</tr>
<tr>
<td>Al-50%Si</td>
<td>48</td>
<td>0.19</td>
<td></td>
<td>0.02</td>
<td>0.02</td>
<td>0.01</td>
<td>0.01</td>
<td>0.1</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Al - 25%Mn</td>
<td>0.04</td>
<td>0.11</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>24.7</td>
</tr>
<tr>
<td>Al - 20%Ni</td>
<td>0.01</td>
<td>0.1</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>18.8</td>
</tr>
<tr>
<td>Al - 5%Ti - 1%B</td>
<td>0.05</td>
<td>0.11</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>4.8</td>
<td>1.2</td>
</tr>
<tr>
<td>Al - 10%Sr</td>
<td>0.04</td>
<td>0.08</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>Sr=9.3</td>
<td></td>
</tr>
<tr>
<td>Cu-8%P</td>
<td>82</td>
<td>0.001</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>8.0</td>
</tr>
<tr>
<td>Bi</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>&gt;99.999</td>
<td></td>
</tr>
<tr>
<td>Sn</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>&gt;99.999</td>
<td></td>
</tr>
</tbody>
</table>

3.2.3 Melt Treatment, Casting Process and Heat Treatment

The melting and casting process was performed in five steps as follows:

(i) Coating and baking the crucible, mould and tools

(ii) Melting and alloying under Ar gas protection,

(iii) Grain refining and modification,

(iv) Holding the molten alloy for 15 minutes,

(v) Casting into different moulds.

The crucibles, casting moulds and tools were coated with a BN coating (purchased from ZYP Coatings, Inc.), in order to protect them from erosion by the molten metal, and also to prevent any contamination of the molten alloys. They were baked to 300 °C in the furnace to evaporate any moisture present. The melt processes showed differences in modification for the hypereutectic and hypoeutectic Al-Si alloys.
For the hypoeutectic Al-Si alloys, melts were prepared using the commercial 319.2 aluminum alloy. Three different bismuth concentrations of 0.0, 0.5, and 1.0 wt.% were used. Strontium, within a composition range of 0.05 and 0.1 %, was added via an Al-10 % Sr master alloy. After melting and holding for 20 minutes at 750 °C in an Ar gas environment, the grain refiner (Al – 5 % Ti – 1 % B) was added to the melt in the range of 0.01 ~ 0.03 % Ti and ~ 0.01 % B along with a modifier, Al – 10 % Sr, to the fully molten alloy for the preparation of the eutectic alloys. The procedure used to prepare the hypoeutectic Al-Si alloy with 0.5 % Bi, solidified at a cooling rate of 9.4 °C/s, is shown in Fig. 3.6.

For the hypereutectic Al-Si alloys, the base hypereutectic alloy had a composition of Al-16 % Si. Melts were prepared using a 319.2 primary Al alloy. An Al-50 % Si master alloy was used to increase the Si concentration to 16 %, while an Al-20 % Ni alloy was used to obtain 0.5 % Ni in the final composition. A grain refiner (Al – 5 %Ti – 1 %B) and an 8PCu alloy (Cu-8 % P) were added at 770 °C to the melt of the hypereutectic alloys. Samples with two different Bi concentrations of 0.5 and 1.0 % were cast. P (60 ppm) was added to refine the primary Si particles via a Cu-P master alloy. Pure Bi (99.999 % Bi) was added to the fully melted alloy at 780 °C. Cu-8 % P in the form of fine shots were broadcast over the melt (0.003 %~0.015 % P in the final alloys) as wide as possible, to form aluminum phosphide, AlP, which has a crystal structure very similar to that of silicon, and also acts an effective heterogeneous nucleant [61]. After modification, the Al 16 % Si alloy structure becomes finer, and the silicon becomes more rounded, both of which contribute to increasing the ultimate tensile strength and ductility.
(see section 4.7). The procedure used to prepare the hypereutectic Al-Si alloy with 0.5 % Bi, solidified at a cooling rate of 26 °C/s, is shown in Fig. 3.7.

The molten aluminum was poured to the steel casting mould and the copper mould, after being mixed in the bath manually, and holding for 15 minutes. This holding duration was necessary to allow the inclusions to drop to the bottom of melt, and make the final samples free of these inclusions, which are often ceramic particles. These inclusions may adversely affect the machinability.

The molten alloy was then poured into the steel and copper moulds at 750 °C. To obtain two different cooling rates, the permanent steel mould was preheated to 400 °C to obtain a cooling rate of 9.4 °C/s and the copper mould was maintained at room temperature to obtain a cooling rate of 26 °C/s. The copper mould had a smaller cavity for the molten aluminum, and a thicker wall section. Higher solidification rates were achieved by using the chilled copper mould, to assist in the modification process. This process resulted in a noticeably finer microstructure, which can improve the machining properties of the alloy.

A 0.5 mm diameter K-type (Chromel-Alumel) thermocouple was placed 2 mm away from the wall of the permanent steel and copper moulds, to measure the cooling rate of the casting during the solidification process.

Castings were solution-treated at 480 °C for 8 hours, quenched in water, and then aged at 175 °C for 5 hours to remove micro and macro-segregation of the alloying elements.
3.3. Experimental Determination of the Mechanical Properties

3.3.1 Hardness Measurements

A Brinell hardness tester was used to measure the bulk hardness of the Al-Si alloys with a ball of 10 mm diameter, and a load of 500 kg. A microhardness tester, Buehler Micromet II® (Model 1600-9000), with a square-base diamond pyramid indenter (136 ° tip angle), was used to measure the aluminum matrix hardness. The indentation load applied was 10 g. The reported hardness values are the average of 10 measurements each.

3.3.2 Tensile Testing

Tensile tests were performed according to the procedure described in the ASTM Standard B557M-07, “Standard Test Methods for Tension Testing Wrought and Cast Aluminum- and Magnesium-Alloy Products [Metric]”, using a computer-controlled Instron 8500 testing machine. The testing machine was equipped with a data acquisition system as shown in Fig. 3.8. A strain rate of $2.2 \times 10^{-4} \text{ s}^{-1}$ was used and the tests were conducted at ambient temperature. The values reported are the averages of three tests. The tensile test specimens, whose dimensions are shown in Fig. 3.9, were machined from heat-treated cast ingots. A strain gauge extensometer was attached to the tensile test bar to measure elongation. The averages of yield strength (YS), tensile strength (TS) and elongation, together with their standard deviations were calculated.

The tensile fracture surface, the cross section along the tensile load direction and the size and distribution of Bi in the matrix were also observed and examined using a scanning electron microscope (SEM), equipped with an energy dispersive spectrometer (EDS).
3.3.3 Compression Testing

High temperature compression tests, following the procedure described in the ASTM Standard E209-00(2005), “Standard Practice for Compression Tests of Metallic Materials at Elevated Temperatures with Conventional or Rapid Heating Rates and Strain Rates”, were performed using an Instron 8500 testing system at a strain rate of $2.2 \times 10^{-3} \, \text{s}^{-1}$. The values reported are the average of the results of three tests. The compressive tests were performed at $240 \pm 5 \, ^\circ\text{C}$. The test coupons ($\varnothing 10 \times 22 \, \text{mm}$) were machined from the heat-treated cast ingots. The average values of compressive strength and strain were calculated.

3.4. Metallographic Analyses

Samples for microstructural examination were prepared by conventional grinding and polishing techniques. The samples were wet ground with 180, 240, 400, 600, 1200, and 2400 P SiC emery papers successively, on a rotating polishing machine. After the final grinding, the samples were polished using 3 and 1 $\mu$m diamond suspensions, and the final polishing was performed using a 0.1 $\mu$m diamond suspension. Etching for microstructural observations was performed by immersing the samples into a Graff-Sargent solution, which consisted of 84 ml H$_2$O, 15 ml HNO$_3$, 0.5 ml HF, and 3 g CrO$_3$, for 5 seconds. Quantitative metallographic measurements of the silicon particle size were conducted using an optical microscope, Axiovert 25, equipped with an image analysis software. The quantitative microstructural measurements included the measurement of the maximum silicon particle length and width. The particle length was determined by measuring the maximum length of each particle parallel to the maximum length, and the particle width was determined by measuring the widest distance across each particle in a
direction perpendicular to the direction of the maximum length. The aspect ratio is defined as the average particle length divided by average particle width, and the sphericity is defined by:

\[
Sphericity = \frac{4 \times \pi \times \text{area}}{\text{perimeter}^2}
\] (3.1)

The sphericity values range between zero and one, in which a value of one corresponds to a perfect circle. The area density of Si particles is defined by the percentage of the total area of Si particles divided by the total area analyzed. The area density obtained was the average of 8 to 10 measurements for each alloy.

3.5. Selection of Cutting Conditions

3.5.1 Selection of the Cutting Speed and the Feed Per Revolution

Samples for the orthogonal cutting tests were machined into hollow tubes, with a 25.4 mm outer diameter and a wall thickness of 3.0 mm. Cutting tests were performed on a lathe equipped with a rapid-action brake. Two cutting speeds, 370 rpm (0.42 m/s) and 800 rpm (0.90 m/s), were used with the same feed rate of 0.25 mm/rev. No cutting fluid was introduced into the system during machining.

Turning speeds normally used for the machining of aluminum alloys are less than 1250 rpm with typical speeds being approximately 800 rpm. The two speeds that were chosen for this study, 370 rpm and 800 rpm, are well within the range commonly used for these alloys [10, 11].

A good surface finish, that is desired in the final machined product, and the rigidity of the machine, are the main determinants of the feed per revolution used in a given machining process [10, 11]. The final finishing operations are done at light feeds, generally in the range of 0.05 mm to 0.15 mm. The initial rough turning feeds range from
0.15 mm to 0.65 mm for aluminum alloys. A turning feed of 0.25 mm was chosen in this work.

3.5.2 The Cutting Tool and the Turning Lathe

All dry orthogonal cutting tests were performed on a Harrison M300 lathe, which is a compact and reliable centre lathe, and is easy to operate (Fig. 3.10). It has a long, foot-operated and powerful spindle brake, with electrical disengagement for interrupted cutting. The machine is powered by a fan-cooled 2.2 kW, 3-phase, 1500 rpm motor. The spindle rotation speed can be chosen from the twelve different speeds available that range from 40 to 2500 rpm.

The cutting tool used for the cutting experiments was a polycrystalline diamond (PCD) insert VCMW 332FP (purchased from Sandvik Coromant), with a rake angle of zero degrees, as shown in Fig. 3.11(a). The polycrystalline diamond insert is the hardest of all the tool materials in use, possesses excellent resistance to wear, and has high dimensional stability. It also shows comparatively lower friction during turning, has high thermal conductivity and possesses very high fracture strength [21]. During the cutting process, the Al-Si casting generated discontinuous chips without the formation of a built-up edge at the tool tip.

The WC cutting tool (purchased from Sandvik Coromant) was used for the experiments that required temperature measurement, and had a rake angle of zero degree. A 0.5 mm diameter K-type (Chromel-Alumel) thermocouple was inserted into the blind hole, as shown in Fig. 3.11(b).
3.6. Force Measurement System

Cutting and thrust forces are the two important forces exerted by the tool on the workpiece during dry orthogonal turning. The different components of the force sensors and the data acquisition system used in this work are shown in Fig. 3.12 (a). Fig. 3.12(b) is a schematic showing the direction of the cutting and thrust forces on the tube-shaped sample that has been used for this study. A two-axis force sensor system for measuring the cutting and thrust forces was designed and constructed as part of this study. The entire force measurement system is shown in Fig. 3.12(c). The tool holder, a rectangular block of steel on which the cutting insert was mounted, was modified by cutting slots in it. The narrowed-down section, as shown in Fig. 3.13, was a rectangular block with a dimension of 1 cm x 1 cm. The purpose of this modification was to increase the strain sensitivity of the holder as it was being subjected to the cutting and thrust forces during cutting. Strain gauges were mounted on the modified narrow cross section, and were connected to a strain sensor attached to the tool-holder mounting. The data acquisition system comprised of a strain sensor, a wireless transmitter system (V-Link 2.4 GHz Wireless Voltage Node), an analog base station (MicroStrain Micro TxRx wireless base station with analog outputs), and a computer with a data acquisition software, ‘Agile Link’ (from MicroStrain). The cutting and thrust forces generated during machining were measured simultaneously with a data sweeping frequency of 2000 Hz.

Cutting and thrust forces were calibrated by the stepped loading and unloading of predetermined weights on the system under near-static conditions, which was followed by fitting an equation which defined the relationship between the applied load and the output recorded. The relationships between the cutting force and thrust force applied and
the corresponding output readings on the data acquisition system were found to be linear, and the equations for defining them have been shown in Eqn.3.2 and Eqn.3.3 below:

Cutting Force Equation is:

\[ F_c = -0.657a + 1340 \text{ (in units of Newton)} \]  \hspace{1cm} (3.2)

Thrust Force Equation is:

\[ F_t = 0.739b - 1507 \text{ (in units of Newton)} \] \hspace{1cm} (3.3)

3.7. Temperature Measurements during Turning

The contact surface temperature between the chips and the tool rake face was continuously measured and recorded by a multifunctional data acquisition system connected to a personal computer during turning, when the WC cutting tool was used. The sampling rate was 30 Hz for each channel. The contact surface temperature was measured using a 0.5 mm diameter K-type (Chromel-Alumel) thermocouple, which was inserted in the WC cutting tool 0.2 mm under the rake face, through a 0.5 mm diameter blind hole made by electrical discharge machining (EDM).

3.8. Characterization of Chip Morphology and the Cross-sections of the Chips

The length and shape of the chips produced from the machining of the alloys with different concentrations of Bi or Sn, and cast with different cooling rates during solidification, were compared. The shapes of the chips were recorded by optical photography. The free and side surfaces, and the cross-section of chips were observed using a scanning electron microscope (JEOL JSM- 5800LV SEM).

The distribution of Bi in the cross-section of chips was examined using a scanning electron microscope (SEM) equipped with an energy dispersive spectrometer (EDS).
Compositional EDS maps of the cross-sections of chips were obtained to determine the bismuth distribution and deformation.

3.9. Measurement of Surface Roughness

Surface roughness of the different Al-Si alloys, machined under the different conditions, was measured after the turning experiments, to quantitatively analyze the finish of the surface produced. A WYKO NT 1100 Optical Surface Profilometer, capable of non-contact three-dimensional surface metrology, was used to produce high quality three-dimensional surface profiles of the surfaces under study. Unfiltered white light, reflected from a reference mirror, was combined with the light reflected from the sample to create interference fringes in the vertical scanning interferometry (VSI) mode. The surface roughness value that was measured is expressed in terms of the Centre Line Average or the Ra value.

3.10. TEM Observation of the Chips

The samples that were observed under transmission electron microscopy (TEM) were subjected to large strain deformation during the dry turning process: Turning tests were performed on a lathe equipped with a rapid action brake. A cutting speed of 370 rpm (0.42 m/s) was used with a feed rate of 0.25 mm/rev. No cutting fluid was introduced into the system during machining. A diamond tip cutter with a 0° rake angle was used as the cutting tool insert. Samples for TEM observation were taken from the center of the chips, and parallel to the tool rake face. The shear strain of the TEM samples during dry turning was 0.3, measured by the local equivalent strain method [31].

Thin films for TEM were ground using a series of SiC abrasive papers to a thickness of 40 μm. These foils were further dimpled, and then thinned by ion beam
milling with an incidence angle of $3 - 5^\circ$. The foils were observed using a JEOL 2010 F transmission electron microscope (TEM) equipped with an Oxford (INCA 250) energy dispersive X-ray spectrometer (EDX). The machine was operated at 200 kV.
Fig. 3.1 Casting moulds and crucible.
(a) permanent steel mould (120×120×25 mm³), (b) chilled copper mould (120×25×25 mm³), and (c) the crucible with its holder.
Fig3.2 The general layout of casting facility
**Fig. 3.3** Waffle shape of aluminum-based master alloys (Al - 50% Si, Al - 25% Mn and Al - 20% Ni).

**Fig. 3.4** Shape of master alloys extruded rod Al - 10%Sr, extruded rod Al - 5%Ti - 1%B and small shot Cu - 8%P.
Fig. 3.5 Microstructures of the grain refiner and modifiers
(a) Al - 5%Ti - 1%B, (b) Al - 10%Sr, and (c) small shot of Cu - 8%P.

Fig. 3.6 Procedure of hypoeutectic Al-Si alloy with Bi cast at a cooling rate of 9.4 or 26 °C/s.
Fig. 3.7 Procedure hypereutectic Al-Si alloy with Bi cast at a cooling rate of 9.4 or 26 °C/s.
Fig. 3.8 Computer controlled Instron 8500 testing machine

Fig. 3.9 Drawing of tensile test specimens (ASTM Standard, B 557M – 07)
Fig. 3.10 Harrison M300 lath used for the orthogonal cutting experiments.

Fig. 3.11 Polycrystalline diamond (PCD) and WC cutting tool insert. (a) Polycrystalline diamond (PCD) insert (VCMW 332FP) with a rake angle of zero degree, and (b) WC cutting tool for the contact surface temperature measurement using a 0.5mm diameter K-type (Chromel-Alumel) thermocouple inserted in the a blind hole.
Fig. 3.12 The set-up of force measurement system
(a) Schematic diagram showing the different components of the force sensor and data acquisition system, (b) The direction of cutting and thrust forces, and (c) Actual arrangement showing the set-up with force measurement system mounted on the lath, cutting tool mounted on the tool holder and workpiece mounted on the lath spindle.
Fig. 3.13 Schematic showing step-wise modification of the tool-holder with two-axis force sensor construction for measurement of forces during orthogonal turning experiments [31].
CHAPTER 4  Alloy Design and Microstructural Control

4.1. Introduction

This chapter describes the microstructures of Al-Si alloys with different concentrations of low melting point additives (Bi and Sn) produced at different cooling rates. Hypoeutectic and hypereutectic alloys were made in order to understand the following points:

(1) The effect of low melting point additives (Bi or Sn) on the microstructure of Al-Si alloys,

(2) The role of cooling rates on the microstructure of Al-Si alloys.

The concentrations of the low melting point additives (Bi or Sn) were determined (Section 4.2), along with those of the other alloying elements in the Al-Si alloys. Liquidus temperatures of the different Al-Si alloys were calculated to understand the solidification process (Section 4.3), and the cooling curves were obtained during solidification using different casting moulds (Section 4.4). Sections 4.5 and 4.6 present analyses of the interactions between the low melting point additives (Bi or Sn) and the Si refiners, particularly the effect of Sr in the hypoeutectic alloys on the silicon microstructure and morphology under different cooling rates. The interactions between Bi and the modifying elements, such as P, in the hypereutectic Al-Si alloys are analyzed in Section 4.7, along with their effects on the microstructure and cooling rates of the 390 Al alloy.

4.2. Compositional Design

Silicon particles have the highest microhardness amongst all the second-phase particles in Al-Si alloys. Silicon added is to improve castability, fluidity and reduce
shrinkage [55]. The effects of the low melting point additives including Bi and Sn on the microstructure of the Al-Si alloys were studied separately for each type of alloy, hypoeutectic and hypereutectic. The cast alloys consisted of the following Si concentrations: 5.75 % Si (hypoeutectic) and 16 % Si (hypereutectic) Al-Si alloys.

To maintain the mechanical strength of the alloy, it is important to consider that a good chip-breaking additive (Bi, Sn) should neither form a high melting point intermetallic compound with the base metal nor with any other important alloying element. However, according to the magnesium-bismuth and magnesium-tin equilibrium systems shown in Figs. 4.1 and 4.2, the binary phase diagrams contain a series of intermetallic compounds, such as Mg₃Bi₂ in the Mg-Bi system and Mg₂Sn in the Mg-Sn system. When used as an additive, the concentration of Sn or Bi must be limited. It is desirable to limit the bismuth concentration to about 1-2 % [14]. The experiments in this work show that when Bi and Sn are added at 0.5 and 1.0 % respectively, the strength of the alloy is not affected.

Mg forms intermetallics with Bi when its concentration is over 0.6 %, based on Fig. 4.1. Mg also starts to form intermetallics with Sn at 0.107 % Sn [54], seen in Fig. 4.2. TEM investigation has indicated that free Sn and Mg₂Sn [90] is present in as-cast structures of aluminum alloy type 319, but Mg₂Sn alone is present in the T4 and T7 heat-treated samples of this alloy. Mg₂Sn is also formed in the as-quenched (T4) samples and the heat treatment shows a high tendency for the formation of Mg₂Sn due to the rapid diffusivity of Sn in the aluminum matrix [90]. To avoid the formation of intermetallics, the composition Mg was limited to a maximum of 0.1 % in all the alloys in this work.
The selected compositions of the hypoeutectic and hypereutectic Al-Si alloys are listed in Table 4.1.

4.2.1 Composition of the Hypoeutectic Alloys

In hypoeutectic alloys, the required amount of the modifier element (Sr) depends on the alloy composition; higher silicon contents require more of the modifying agent. Strontium added in amounts of 0.02% is sufficient to modify a 7% Si alloy such as the type 356 alloy [55]. Alloys that are easy to modify have low phosphorus contents. To find the relationship between Sr-Bi ratio and the eutectic temperature, the morphology of the Si in the alloys after casting was observed. It was noted that when the Sr-Bi ratio was between 0.2 and 0.45, the eutectic temperature dropped, indicating the modification level of the Si structure increased dramatically [24].

4.2.2 Composition of the Hypereutectic Alloys

Coarse primary silicon particles in hypereutectic Al-Si alloys impart poor mechanical properties to them. Phosphorus is deliberately added because it reacts with the aluminum to form AlP particles, which nucleate primary silicon [73]. This results in a refined dispersion of primary Si particles, which is quite opposite to hypoeutectic alloys in which the eutectic silicon is coarsened by the addition of phosphorus [73]. In this work, the melt was modified by the addition of a Cu–15% P alloy (0.02% P in the melt) to refine the primary silicon crystals. High solidification rates assisted the modification process and led to a noticeably finer microstructure.
Table 4.1 Chemical compositions of the hypoeutectic and hypereutectic Al-Si alloys and their calculated liquidus temperatures ($T_{liq}$)

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Si</th>
<th>Cu</th>
<th>Fe</th>
<th>Mg</th>
<th>Mn</th>
<th>Ni</th>
<th>Zn</th>
<th>Ti</th>
<th>Sr</th>
<th>Bi</th>
<th>Sn</th>
<th>$T_{liq}$ (K)</th>
</tr>
</thead>
<tbody>
<tr>
<td>hypo</td>
<td>5.75</td>
<td>3.09</td>
<td>0.17</td>
<td>0.02</td>
<td>0.01</td>
<td>0.02</td>
<td>0.10</td>
<td>0.02</td>
<td>0.00</td>
<td>890.40</td>
<td></td>
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<td></td>
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<td></td>
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<td></td>
<td></td>
<td>0.50</td>
<td>887.29</td>
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<td></td>
<td>1.00</td>
<td>884.19</td>
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<td></td>
<td></td>
<td>0.50</td>
<td>887.75</td>
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<td>1.00</td>
<td>885.19</td>
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<td></td>
<td>0.50</td>
<td>0.25</td>
<td>885.94</td>
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<tr>
<td>hyper</td>
<td>16.00</td>
<td>2.60</td>
<td>0.28</td>
<td>0.10</td>
<td>0.60</td>
<td>0.50</td>
<td>0.07</td>
<td>0.07</td>
<td>0.00</td>
<td>907.34</td>
<td></td>
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</tr>
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<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>0.50</td>
<td>908.85</td>
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<td></td>
<td></td>
<td></td>
<td>1.00</td>
<td>910.37</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

4.3. Calculation of the Liquidus Temperature

The calculation of liquidus temperature is based on the new “element-equivalency concept” [71]. The analysis of two liquidus lines of the binary systems, Al–Si and Al–Xi, shows that the “equivalent effect” on the liquidus temperature of the aluminum alloy can be obtained by using “equivalent” concentrations of Si and Xi alloying ($Si_{EQ}^{Xi}$) or impurity elements.

Mathematically, the liquidus line of most binary Al–Xi phase diagrams can be approximated by a second-order polynomial equation varying with the concentration of element Xi. Taking into consideration the whole temperature range between the melting temperature of pure aluminum and the eutectic temperature of the binary alloy, the following relationship (Equation 4.1) can be established between $Si_{EQ}^{Xi}$ and the concentration of the element Xi. Some major, minor and other elements, including grain refiners, and silicon modifiers have an effect on the $Si_{EQ}^{Xi}$ [71]. Their effect has been expressed in the polynomial form as,

$$Si_{EQ}^{Xi} = a_0^{Xi} + b_0^{Xi} \cdot x_i + c_0^{Xi} \cdot x_i^2$$  \hspace{1cm} (4.1)
Table 4.2 shows a summary of the effect of the alloying elements in the various binary Al-Xi alloys [71], which represent the most common major and minor elements of the 3XX hypoeutectic and hypereutectic aluminum alloys.

**Table 4.2** Polynomial coefficients in Equation 4.1 for the various binary Al-Xi alloys

<table>
<thead>
<tr>
<th>Al-Xi alloy</th>
<th>b₀</th>
<th>c₀</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al–Cu</td>
<td>0.350</td>
<td>-0.027</td>
</tr>
<tr>
<td>Al–Mg</td>
<td>0.0258</td>
<td>-0.0088</td>
</tr>
<tr>
<td>Al–Mn</td>
<td>0.8221</td>
<td>-0.0349</td>
</tr>
<tr>
<td>Al–Fe</td>
<td>0.6495</td>
<td>0.0003</td>
</tr>
<tr>
<td>Al–Zn</td>
<td>0.1227</td>
<td>-0.0002</td>
</tr>
<tr>
<td>Al–Sn</td>
<td>0.7849</td>
<td>-0.0313</td>
</tr>
<tr>
<td>Al–Bi</td>
<td>0.9076</td>
<td>-0.0092</td>
</tr>
<tr>
<td>Al–Pb</td>
<td>0.859</td>
<td>0.0296</td>
</tr>
<tr>
<td>Al–Ca</td>
<td>0.0594</td>
<td>0.00685</td>
</tr>
<tr>
<td>Al–Sb</td>
<td>0.8255</td>
<td>-0.0327</td>
</tr>
<tr>
<td>Al–Ni</td>
<td>0.5644</td>
<td>-0.0285</td>
</tr>
<tr>
<td>Al–Sr</td>
<td>0.7854</td>
<td>-0.0157</td>
</tr>
<tr>
<td>Al–Ti</td>
<td>-0.8159</td>
<td>0.009927</td>
</tr>
<tr>
<td>Al–B</td>
<td>-0.9977</td>
<td>0.00007506</td>
</tr>
</tbody>
</table>

Note: a₀=0 for the elements presented in this table.

The equation of the liquidus line for the hypereutectic Al–Si binary phase diagram (Fig. 2.21) (for Si between 12.3 and 35 % by weight) can be mathematically expressed in the second-order polynomial form. The liquidus temperature for the multi-component hypoeutectic and hypereutectic 3XX aluminum alloys can be obtained using a second-degree polynomial as follows [71]:

\[
T_{LQ}^{Al–Si–ΣX_i} = 660.452 - 6.110 \Sigma S_i E_Q^X - 0.057(ΣS_i E_Q^X)^2
\]  (4.2)

\[
T_{LQ}^{Al–Si–ΣX_i} = 389.79 + 15.855S_i - 0.0561S_i^2 + 3.14 ΣS_i E_Q^X + 0.057(ΣS_i E_Q^X)^2
\]  (4.3)

The calculated liquidus temperatures for the multi-component hypoeutectic and hypereutectic Al-Si alloys are listed in Table 4.1. The liquidus temperature of the
5.75 % Si alloy is between 884.2 K and 890.4 K, and the liquidus temperature of the 16.0 % Si alloy is between 907.3 K and 910.4 K.

4.4. Cooling Rates of Different Alloys

Figure 4.3 shows the cooling curves of Al-5.75Si-0.5Bi castings solidified at cooling rates of 9.4 °C/s and 26 °C/s. Under a cooling rate of 9.4 °C/s, a solidification plateau of the primary Al dendrite formation was observed. The eutectic reaction occurs at 575 °C when the cast is cooling at a rate of 9.4 °C/s, and at 558.7 °C when cast is cooling at a rate of 26 °C/s.

Figure 4.4 (a) shows the cooling curves of Al-16Si-0.5Bi castings solidified at two different cooling rates of 9.4 °C/s and 26 °C/s. Under the cooling rate of 9.4 °C/s, a solidification plateau representing the formation of primary Si phases was observed at 628.3 °C (Fig. 4.4 (b)). Under the cooling rate of 26 °C/s, the primary Si phase formed at 620.5 °C without a clear solidification plateau. The growth of primary Si at the higher cooling rate is suppressed. The temperatures at which the initial eutectic Si formation occurred were 562.1 °C and 554.4 °C for cooling rates of 9.4 °C/s and 26 °C/s respectively.

4.5. Interaction of Sr and Bi in Hypoeutectic Alloys Cast under Different Cooling Rates

The microstructures of hypoeutectic Al-Si alloys were refined by Sr modification. However, the nature of the modification was changed with the addition of Bi.

4.5.1 Microstructures Developed under a Cooling Rate of 9.4 °C/s

The microstructures of the 319.2 base alloy (5.75 % Si) containing different Bi concentrations and cast at a cooling rate of 9.4 °C/s are shown in Fig. 4.5. Aluminum
dendrites and eutectic silicon particles were present in the microstructure (Fig. 4.5 (a)). The microstructures of the 319.2 alloy with 0.5 % and 1.0 % bismuth additions, shown in Figs. 4.5 (b) and (c), exhibited different morphologies of the eutectic silicon phases. The average diameter, length and width of the eutectic Si in the 319.2 alloys with different Bi concentrations are shown in Table 4.3. The average length of the silicon particles was 6.2 ± 3.3 µm in the Al-Si alloy with 1 % Bi, 4.9 ± 2.0 µm in the alloy with 0.5 % Bi, and 2.4 ± 1.3 µm in the alloy without Bi.

**Table 4.3** Average diameter and aspect ratio of the eutectic Si particles in Al 319.2 modified by Bi cast at a cooling rate of 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Diameter (µm)</th>
<th>Length (µm)</th>
<th>Sphericity</th>
</tr>
</thead>
<tbody>
<tr>
<td>319</td>
<td>1.6 ± 0.6</td>
<td>2.4 ± 1.3</td>
<td>0.68 ± 0.11</td>
</tr>
<tr>
<td>0.5Bi</td>
<td>2.2 ± 1.0</td>
<td>4.9 ± 2.0</td>
<td>0.60 ± 0.14</td>
</tr>
<tr>
<td>1.0Bi</td>
<td>2.7 ± 1.6</td>
<td>6.2 ± 3.3</td>
<td>0.54 ± 0.17</td>
</tr>
</tbody>
</table>

**Table 4.4** Hardness of Al 319.2 modified by Bi cast at 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Brinell Hardness (HB)</th>
<th>HV</th>
</tr>
</thead>
<tbody>
<tr>
<td>319</td>
<td>73.4±4.1</td>
<td>78±3.6</td>
</tr>
<tr>
<td>0.5Bi</td>
<td>64.5±3.8</td>
<td>71±4.0</td>
</tr>
<tr>
<td>1.0Bi</td>
<td>62.6±3.9</td>
<td>67±3.8</td>
</tr>
</tbody>
</table>

The hardness values of the 319.2 alloy that was cast with a cooling rate of 9.4 °C/s is listed in Table 4.4. The average hardness was 62.6 ± 3.3 in the Al-Si alloy with 1.0 wt. % Bi, 64 ± 2.0 in the alloy with 0.5 % Bi, and 73.4 ± 1.3 in the alloy without Bi.

**4.5.2 Microstructures Developed under a Cooling Rate of 26 °C/s**

Figure 4.6 shows the microstructures of the 319.2 base alloy with 0.5 and 1.0 % Bi additions, cast with a cooling rate of 26 °C/s. The eutectic Si particles in the 319.2 alloy with 1.0 % Bi that was cast with a high cooling rate are similar to those in the base alloy. The average diameter and aspect ratio of the eutectic Si in the 319 alloys with
different amounts of Bi additions at a cooling rate of 26 °C/s are listed in Table 4.5, and their hardness values are listed in Table 4.6.

The average length of the silicon particles is 3.6 µm in the Al-Si alloy with 1.0 % Bi, 3.0 µm in the alloy with 0.5 % Bi, and 2.1 µm in the alloy without Bi. The Brinell hardness value of the Al-Si alloy with 1.0 % Bi is 76, of the alloy with 0.5 % Bi is 78.7, and of the alloy without bismuth is 81.4.

**Table 4.5** Average diameter and aspect ratio of the eutectic Si particles in Al 319.2 modified by Bi cast at 26 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Diameter (µm)</th>
<th>Length (µm)</th>
<th>Sphericity</th>
</tr>
</thead>
<tbody>
<tr>
<td>319</td>
<td>1.4 ± 1.0</td>
<td>2.1 ± 1.0</td>
<td>0.68 ± 0.1</td>
</tr>
<tr>
<td>0.5Bi</td>
<td>1.8 ± 1.1</td>
<td>3.0 ± 1.7</td>
<td>0.66 ± 0.1</td>
</tr>
<tr>
<td>1.0Bi</td>
<td>2.1 ± 1.2</td>
<td>3.6 ± 2.0</td>
<td>0.65 ± 0.11</td>
</tr>
</tbody>
</table>

**Table 4.6** Hardness of Al 319.2 modified by Bi cooled cast at 26 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Brinell Hardness (HB)</th>
<th>HV</th>
</tr>
</thead>
<tbody>
<tr>
<td>319</td>
<td>81.4±3.6</td>
<td>79±3.2</td>
</tr>
<tr>
<td>0.5Bi</td>
<td>78.7±3.5</td>
<td>76±3.3</td>
</tr>
<tr>
<td>1.0Bi</td>
<td>76±3.7</td>
<td>73±3.6</td>
</tr>
</tbody>
</table>

### 4.6. Interaction of Sr and Sn in Hypoeutectic Alloys Cast under Different Cooling Rates

#### 4.6.1 Microstructures Developed under a Cooling Rate of 9.4 °C/s

The microstructures of the 319.2 base alloys with 0.5 and 1.0 % Sn concentrations are shown in Fig. 4.7. Primary aluminum dendrites and eutectic silicon particles were present in the microstructure of the 319.2 base alloys, as seen in Fig. 4.5 (a). The average diameter, length and width of the eutectic Si in the Al-Si alloys with different Sn concentrations are given in Table 4.7. The average length of silicon particle was 2.9 ± 1.6 µm in the Al-Si alloy with 1.0 % Sn, 2.8 ± 1.5 µm in the alloy with 0.5 % Sn, and 2.4
± 1.3 µm in the alloy without Sn, which indicated that Sn had no strong effect on the silicon morphology.

**Table 4.7** Average diameters and aspect ratios of the eutectic Si particles in Al 319.2 modified by Sn cast at 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Diameter (µm)</th>
<th>Length (µm)</th>
<th>Sphericity</th>
</tr>
</thead>
<tbody>
<tr>
<td>319</td>
<td>1.6 ± 0.6</td>
<td>2.4 ± 1.3</td>
<td>0.68 ± 0.11</td>
</tr>
<tr>
<td>0.5Sn</td>
<td>1.9 ± 0.9</td>
<td>2.8 ± 1.5</td>
<td>0.69 ± 0.12</td>
</tr>
<tr>
<td>1.0Sn</td>
<td>1.9 ± 1.2</td>
<td>2.9 ± 1.6</td>
<td>0.69 ± 0.11</td>
</tr>
</tbody>
</table>

The hardness of 319 base alloys with different amounts of Sn added and cast with a cooling rate of 9.4 °C/s is shown in **Table 4.8**. The HB was 58 ± 1.6 in the Al-Si alloy with 1.0 % Sn, 60 ± 1.5 in the alloy with 0.5 % Sn, and 73.4 ± 1.3 in the alloy without Sn, which indicated that Sn influenced the hardness of alloys. However, Sn didn't influence the Si morphology.

**Table 4.8** Hardness of Al 319.2 modified by Sn cast at 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Brinell Hardness (HB)</th>
<th>HV</th>
</tr>
</thead>
<tbody>
<tr>
<td>319</td>
<td>73.4±4.1</td>
<td>78±3.6</td>
</tr>
<tr>
<td>0.5Sn</td>
<td>60±3.5</td>
<td>65±3.4</td>
</tr>
<tr>
<td>1.0Sn</td>
<td>58±3.6</td>
<td>60±3.5</td>
</tr>
</tbody>
</table>

4.6.2 Microstructures Developed under a Cooling Rate of 26 °C/s

The microstructures of the 319.2 base alloy cooled at a high rate, with 0.5 and 1.0 % Sn concentrations are shown in **Figs. 4.8 (a) and (b)**. Both Aluminum dendrites and eutectic silicon particles were present in the microstructure of the 319.2 base alloys, as seen in **Fig. 4.6 (a)**. The average diameter, length and width of the eutectic Si in Al-Si alloys with different Sn concentrations are given in **Table 4.9**. The average length of silicon particle was 2.9 ± 1.6 µm in the Al-Si alloy with 1.0 % Sn, 2.9 ± 1.4 µm in the alloy with 0.5 % Sn, and 2.1 ± 1.0 µm in the alloy without Sn, which indicated that Sn had no effect on the silicon morphology.
Table 4.9 Average diameters and aspect ratios of eutectic Si particles in Al 319.2 modified by Sn cast at 26 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Diameter (µm)</th>
<th>Length (µm)</th>
<th>Sphericity</th>
</tr>
</thead>
<tbody>
<tr>
<td>319</td>
<td>1.4 ± 1.0</td>
<td>2.1 ± 1.0</td>
<td>0.68 ± 0.1</td>
</tr>
<tr>
<td>0.5Sn</td>
<td>1.8 ± 1.1</td>
<td>2.9 ± 1.4</td>
<td>0.68 ± 0.1</td>
</tr>
<tr>
<td>1.0Sn</td>
<td>1.8 ± 1.1</td>
<td>2.9 ± 1.6</td>
<td>0.68 ± 0.1</td>
</tr>
</tbody>
</table>

The hardness of the 319 alloy with different amounts of Sn added, and cast at a cooling rate of 26 °C/s is shown in Table 4.10. The HB is 73.4 ± 1.6 in the Al-Si alloy with 1.0 wt. % Sn, 76 ± 1.5 in the alloy with 0.5 wt. % Sn, and 81.4 ± 1.3 in the alloy without Sn, which demonstrated that the addition of Sn affected the hardness of the alloys.

Table 4.10 Hardness values of Al 319.2 modified by Sn cast at 26 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Brinell Hardness (HB)</th>
<th>HV</th>
</tr>
</thead>
<tbody>
<tr>
<td>319</td>
<td>81.4±3.6</td>
<td>79±3.2</td>
</tr>
<tr>
<td>0.5Sn</td>
<td>76±3.8</td>
<td>69±3.5</td>
</tr>
<tr>
<td>1.0Sn</td>
<td>73.4±3.9</td>
<td>67±3.3</td>
</tr>
</tbody>
</table>

4.7. Interaction of P and Bi in Hypereutectic Alloys Cast with Different Cooling Rates

4.7.1 Microstructures Developed under a cooling rate of 9.4 °C/s

Figs. 4.9 (a), (b) and (c) show the microstructures of the Al 390 alloys with of 0.0, 0.5 and 1.0 % Bi additions, respectively. The alloys were cast at a cooling rate of 26 °C/s. Both the primary and the eutectic Si particles that were present in the microstructures were evenly distributed in the matrix. Table 4.12 lists the average diameters and aspect ratios of the primary Si particles in the Al-16Si alloys cast with a cooling rate of 9.4 °C/s, with different Bi additions. The particle size of the primary Si particles in the alloys modified with Bi did not vary significantly. The average diameter of primary Si particles in the casting that was solidified at 9.4 °C/s was 38 ± 23 µm in the Al-Si alloy with 1.0 % Bi, 33 ± 21 µm in the alloy with 0.5 % Bi and 31 ± 20 µm in the base alloy. As listed in
Table 4.13, the average eutectic Si particle diameter was $7.5 \pm 3.5$ µm in the alloy without Bi, and this decreased slightly to $5.8 \pm 3.0$ µm in the alloy with 1.0 % Bi. This suggests that Bi has little effect on the size of primary Si particles, while the eutectic Si particles are increasingly refined as the amount of Bi increases.

Table 4.12 Average diameters and aspect ratios of the primary Si particles in the Al 390 modified by Bi cast at 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Primary Si Diameter (µm)</th>
<th>Primary Si Aspect Ratio</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.0 wt.%Bi</td>
<td>38±23</td>
<td>1.83±0.82</td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>33±21</td>
<td>1.74±0.82</td>
</tr>
<tr>
<td>0 wt.%Bi</td>
<td>31±20</td>
<td>1.74±0.80</td>
</tr>
</tbody>
</table>

Table 4.14 lists the average Brinell (HB) and Vickers (HV) hardness values for the Al-16Si alloys that were cast at a cooling rate of 26 °C/s. The addition of Bi lead to a decrease in the alloy hardness. When cast at a cooling rate of 26 °C/s, the hardness of the base alloy was 94 ± 5.0, and the hardness decreased to 90 ± 5.3 for the alloy with 0.5 % Bi and 85 ± 3.6 for the alloy with 1.0 % Bi. The corresponding Vickers hardness values were 76 ± 2.8, 70 ± 6.1 and 62 ± 6.4, respectively.

Table 4.13 Average diameters and aspect ratios of the eutectic Si particles in the Al 390 modified by Bi cast at 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Eutectic Si Diameter (µm)</th>
<th>Eutectic Si Aspect Ratio</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.0 wt.%Bi</td>
<td>5.8±3.0</td>
<td>2.83±1.06</td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>6.1±3.3</td>
<td>2.93±0.93</td>
</tr>
<tr>
<td>0 wt.%Bi</td>
<td>7.5±3.5</td>
<td>2.95±1.01</td>
</tr>
</tbody>
</table>

Table 4.14 Hardness values of the Al 390 modified by Bi and cast at 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Brinell Hardness (HB)</th>
<th>HV</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.0 wt.%Bi</td>
<td>85±3.6</td>
<td>62±6.4</td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>90±5.3</td>
<td>70±6.1</td>
</tr>
<tr>
<td>0 wt.%Bi</td>
<td>94±5.0</td>
<td>76±2.8</td>
</tr>
</tbody>
</table>
4.7.2 Microstructures Developed under a Cooling Rate of 26 °C/s

4.7.2.1 Microstructural Evolution

Figs. 4.10 (a), (b) and (c) present the microstructures of Al-16Si alloys with 1.0 % Bi, 0.5 % Bi and no Bi added, and cast with a cooling rate of 26 °C/s. Both the primary and eutectic Si particles were evenly distributed in the matrix, like in the alloys solidified at a lower rate, but particles were smaller in size, as was expected.

The average primary Si particle diameter was 20 ± 10 µm in the Al-Si alloy with 1.0 % Bi, 19 ± 9 µm in the alloy with 0.5 % Bi and 18 ± 9 µm in the alloy without Bi, as listed in Table 4.15. The average eutectic Si particle diameter was 2.9 ± 1.8 µm in the alloy with 1.0 % Bi, and 3.8 ± 2.7 µm in the alloy without Bi, as listed in Table 4.16. This indicates that the role of Bi in the modification of eutectic and primary Si particles is not significant. However, Figs. 4.9 and 4.10 show that their microstructures are different when formed under different cooling rates. The following were observed:

(i) The percentage of the eutectic Si phase is twice as much when solidified alloys at a cooling rate of 26 °C/s, as compared to that when solidified at a cooling rate of 9.4 °C/s, because Si has had less time to diffuse, which contributes to the growth of the primary Si phase at these higher cooling rates. Less Si was consumed from the molten Al-Si solution during fast solidification.

(ii) From the cooling curves, it can be discerned that a high cooling rate resulted in the refinement of both primary and eutectic Si particles. The sizes of the Si particles of the alloys cast with a cooling rate of 26 °C/s shown in Figs. 4.10 (a), (b) and (c) are about 50 % smaller than those of the alloys cast with a cooling rate of 9.4 °C/s, as shown in Figs. 4.9 (a), (b) and (c).
The histograms, showing the distributions of primary Si particle size in the alloy with 0.5 % Bi in Fig. 4.11, indicate that the particle size distribution was in the range of 8-73 μm when cast with a cooling rate of 9.4 °C/s, and 8-49 μm when cast with a cooling rate of 26 °C/s. This means that a higher cooling rate narrows the distribution range of the primary Si particles, which corresponds to the standard error values quoted in Tables 4.16 and 4.19 for the primary Si particle diameter size.

4.7.2.2 Hardness Properties

Table 4.17 lists the average Brinell (HB) and Vickers (HV) hardness values for the Al-16Si alloys cast with a cooling rate of 26 °C/s. The HB of the alloys cast with a cooling rate of 26 °C/s decreased from 112 ± 4.1 for the alloy without Bi to 104 ± 4.8 for the alloy with 0.5 % Bi, and 95 ± 5.1 for the alloy with 1.0 % Bi. These values were higher than those of the alloys cast with a cooling rate of 9.4 °C/s, consistent with the inference that the microstructural refinement at higher cooling rate leads to higher hardness values.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Primary Si Diameter (μm)</th>
<th>Primary Si Aspect Ratio</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.0 wt.%Bi</td>
<td>20±10</td>
<td>1.80±0.81</td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>19±9</td>
<td>1.70±0.78</td>
</tr>
<tr>
<td>0 wt.%Bi</td>
<td>18±9</td>
<td>1.70±0.78</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Eutectic Si Diameter (μm)</th>
<th>Eutectic Si Aspect Ratio</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.0 wt.%Bi</td>
<td>2.9±1.8</td>
<td>1.84±0.71</td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>3.4±2.1</td>
<td>1.85±0.54</td>
</tr>
<tr>
<td>0 wt.%Bi</td>
<td>3.8±2.7</td>
<td>1.88±0.67</td>
</tr>
</tbody>
</table>
Table 4.17 Hardness of Al 390 modified by Bi and cast at 26 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Brinell Hardness (HB)</th>
<th>HV</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.0 wt.%Bi</td>
<td>95±5.1</td>
<td>72±2.4</td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>104±4.8</td>
<td>76±4.3</td>
</tr>
<tr>
<td>0 wt.%Bi</td>
<td>112±4.1</td>
<td>82±4.5</td>
</tr>
</tbody>
</table>

4.8. Summary of the Microstructural Evolution and Alloying Element Interactions

Based on the analysis of the experimental results, the following conclusions can be drawn:

(i) The microstructures and liquidus temperatures of the alloys are different for the hypoeutectic alloy (5.75 % Si in type 319.2) and the hypereutectic alloy (16 % Si in Al 390).

(ii) To maintain the hardness of the base alloy, the optimum concentration of Bi is 0.5 % Bi.

(iii) The calculated liquidus temperatures for the multi-component hypoeutectic and hypereutectic Al-Si alloys vary with the amounts of Si and the alloying elements.

(iv) The refining effect of Sr on the eutectic silicon morphology reduces with the increase in the Bi content, because the silicon morphology becomes coarser with an increase in the Bi content in hypoeutectic Al-Si alloys.

(v) For the hypereutectic Al-Si alloy, Bi has only a small effect on the size of the primary Si particles. Eutectic Si particles are increasingly refined as the amount of Bi is increased.

(vi) The microstructural evolution under different casting conditions shows that high rates of solidification can facilitate the development of more refined and more uniformly dispersed primary and eutectic Si particles. In this case, the short
growth times for Si particles at a high cooling rate, which is a non-equilibrium process taking place far away from the equilibrium solidification, determines the refinement of the microstructures. The quantitative results indicate that the cooling rate plays a significant role in the refinement of the microstructure.
Fig. 4.1 Mg-Bi binary phase diagram.
Fig. 4.2 Mg-Sn binary phase diagram.

Fig. 4.3 The cooling curves of Al 319.2 with 0.5% Bi
Fig. 4.4 The cooling curves of Al 390 with 0.5% Bi
**Fig. 4.5** Microstructures of Al 319.2 modified by Bi cast at the cooling rate of 9.4 °C/s
(a) 319.2, (b) 319.2 with 0.5 % Bi, (c) 319.2 with 1.0 % Bi.
Fig. 4.6 Microstructures of Al 319.2 modified by Bi cast at the cooling rate of 26 °C/s (a) 319.2, (b) 319.2 with 0.5 % Bi, (c) 319.2 with 1.0 % Bi.
**Fig. 4.7** Microstructures of Al 319.2 cast at the cooling rate of 9.4 °C/s modified by Sn (a) with 0.5 % Sn, (b) with 1.0 % Sn
Fig. 4.8 Microstructures of Al 319.2 cast at the cooling rate of 26 °C/s modified by Sn (a) with 0.5 % Sn, (b) with 1.0 % Sn.
Fig. 4.9 Microstructures of Al 390 modified by Bi cast at a cooling rate of 9.4 °C/s (a) with 1.0 % Bi, (b) with 0.5 % Bi, (c) 0 % Bi.
Fig. 4.10 Microstructures of Al 390 modified by Bi cast at a cooling rate of 26 °C/s
(a) with 1.0 % Bi, (b) with 0.5 % Bi, (c) 0.0 % Bi.
Fig. 4.11 Histograms of the primary Si size in Al 390 with 0.5 % Bi cast at different cooling rates. (a) 9.4 °C/s, and (b) 26 °C/s.
CHAPTER 5   Mechanical Properties and Dry Turning
Performances of 319.2 Al Modified by Bi and Sn

5.1. Introduction

This chapter presents the results of the mechanical property tests and the dry turning performance tests of the 319.2 Al-Si alloys modified with Bi and Sn. The main purpose of the experiments was to examine the variation of the mechanical properties of the alloys with different amounts of the Bi and Sn. The effects of Bi and Sn additions on the tensile strength and fracture were examined. The results of turning experiments have been analyzed, and the effects of Bi and Sn additions to this hypoeutectic Al-Si alloy have been derived by determining the following: (i) cutting and thrust forces, (ii) temperature generated between the workpiece and the tool rake face, (iii) microstructure of chip-root cross-sections, and (iv) chip morphology. The results of the turning experiments are discussed along with the results of the mechanical tests.

The effects of the different cooling rates (9.4 °C/s and 26 °C/s) on the properties of the alloys have been discussed in Sections 5.2 and 5.3. Sec. 5.2 examines the microstructures and the tensile properties of the Al-Si alloys generated with these cooling rates. The results of the experiments performed on the alloys cast with a cooling rate of 9.4 °C/s, which include the distribution and size of Bi particles, tensile strength, ductility and fracture surface characteristics, are presented in the Sec. 5.2.1. The corresponding results for the alloys cast with a cooling rate of 26 °C/s are described in Sec. 5.2.2. Dry machining performances of the alloys with different concentrations of Bi and Sn cast with cooling rates of 9.4 °C/s and 26 °C/s and machined with two speeds of 0.42 m/s and 0.9 m/s are reported in Sec. 5.3 using (i) cutting and thrust forces, COF and specific
cutting energy, (ii) interfacial temperature, (iii) surface roughness, (iv) cross-sectional microstructure of the chips and (v) chip morphology. Analyses and conclusions are presented in Sec. 5.4.

5.2. Microstructure and Tensile Properties

5.2.1 Alloys Cast with a Cooling Rate of 9.4 °C/s

5.2.1.1 Distribution and size of the Bi particles

5.2.1.1.1 Alloy 319.2 with Bi modifications

The SEM-BEI (backscattered electron image) micrographs of the 319.2 Al-Si alloy with 0.5 % and 1.0 % Bi additions solidified at a cooling rate of 9.4 °C/s are shown in Figs. 5.1 (a) and (b). The particles that appear bright are pure Bi. They are present in the form of spherical particles. It was observed that the Bi particles formed mainly at the interface of eutectic Si particles, and occasionally at the interfaces of the intermetallic particles with Al matrix.

The average size of the Bi Particles in the microstructure of the 319.2 Al-Si alloy with 0.5 % and 1.0 % Bi additions were 4.0 ± 2 µm and 5.5 ± 1.5 µm, respectively, at the cooling rate of 9.4 °C/s, as observed by SEM-BEI. These Bi Particles in the 319.2 Al-Si alloy cooled at 9.4°C/s, couldn't be observed with optical microscopy. The differences in the microstructure are mainly in the morphology of eutectic Si particles observed by optical microscopy. The average length of silicon particles was 6.2 ± 3.3 µm in the Al-Si alloy with 1 % Bi, 4.9± 2.0 µm in the alloy with 0.5 % Bi, and 2.4 ± 1.3 µm in the alloy without Bi, as listed in Table 4.6.

5.2.1.1.2 Alloy 319.2 with Sn modifications

The SEM-BEI images of the 319.2 Al-Si alloy with 0.5 % and 1.0 % Sn solidified under a cooling rate of 9.4 °C/s are shown in Figs. 5.2 (a) and (b). The particles that
appear bright are pure Sn. Sn was not present in the form of spherical particles as Bi, because it tended to form a film on the boundaries of the Al grains [55] or at the interface of Al and Si. Comparing Figs. 5.2 (a) and (b), the amount of Sn at the Al-Si interface is slightly higher in the alloy containing 1.0 % Sn than in the alloy with 0.5 % Sn. It is difficult to quantify because the Sn phase is a very thin film. The grey phase seen in the microscopy images is the eutectic Si particles in fine spherical morphology, and Sn has no strong effect on this as Bi does in the alloys modified with it. The average length of silicon particles was 2.9 ± 1.6 µm in the Al-Si alloy with 1.0 % Sn, 2.8 ± 1.5 µm in the alloy with 0.5 % Sn, and 2.4 ± 1.3 µm in the alloy without Sn, as listed in Table 4.6.

The SEM-BEI image of the 319.2 Al-Si alloy with 0.5 % Bi and 0.25 % Sn solidified under a cooling rate of 9.4 °C/s is shown in Fig. 5.2 (c). The grey phase is the eutectic Si particles, which were only partly modified [55] as those that were present in the 319.2 Al-Si alloy with Bi modification. The morphology of Si particles in the alloy with 0.5 % Bi and 0.25 % Sn was similar to that seen in the alloy with 0.5 % Bi. On the other hand, the white phase, which is composed of Bi+Sn particles, had a morphology similar to that of the Sn particles in the 319.2 Al-Si alloy with Sn modification.

These Bi, Sn and Bi+Sn particles, as seen in Figs. 5.1 and 5.2, were in the pure state. Fig. 5.3 is the EDS spectrum of the Bi particles and Bi+Sn particles in Al alloys. These were also present in the eutectic Si/Al matrix interface. In general, the main difference observed between the alloys with Bi and the alloys with Sn was that the morphology of the Sn-containing phases was not of a fine spherical shape as those containing Bi, because Sn wets the Al grain boundaries.
In summary, both the additions of Bi and Sn to the 319.2 Al-Si alloy have effects on the microstructures:

(1) The morphology of Si particles became coarse with the increase of Bi addition
(2) The morphology of Si particles did not change with the increase in Sn content
(3) The morphology of the Si particles in the alloy with 0.5 % Bi and 0.25 % Sn was similar to that of the alloys containing 0.5 %Bi
(4) Sn particles were not present in spherical form, as the Bi particles were.
(5) The amount of Bi or Sn at the interface of Al-Si was slightly higher in the alloys with 1.0 % modification than that in alloys with 0.5 % modification

5.2.1.2 Tensile strength, ductility and fracture behaviour of alloy 319.2 with different Bi and Sn additions

5.2.1.2.1 Alloy 319.2 with the addition of Bi

The tensile stress-strain curves of the cast 319.2 Al-Si alloys with different Bi concentrations cast with a cooling rate of 9.4 °C/s are shown in Fig. 5.4. Tensile properties of the 319.2 Al-Si alloys with different amounts of Bi and cast with a cooling rate of 9.4 °C/s are listed in Table 5.1. The average tensile strength of the 319.2 base alloy cast with a cooling rate of 9.4 °C/s was 214.2 ± 1.8 MPa, which was decreased to 187.0 ± 1.7 MPa for the alloy with 0.5 % Bi, and to 181.9 ± 2.0 MPa for the alloy with 1.0 % Bi. The alloy with 0.5 % Bi has an elongation of 3.2 ± 0.6 %, which is close to that of the base alloys. The elongation decreased to 2.8 ± 0.6 % for alloy with 1.0 % Bi.
Table 5.1 Tensile properties of Al 319.2 modified by Bi cast at 9.4 °C/s

<table>
<thead>
<tr>
<th>Composition</th>
<th>Elongation</th>
<th>Tensile Strength (MPa)</th>
<th>YS (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Elongation (%)</td>
<td>Change (%)</td>
<td>Tensile Strength (MPa)</td>
</tr>
<tr>
<td>0 Bi</td>
<td>3.2±0.4</td>
<td>0</td>
<td>214.2±1.8</td>
</tr>
<tr>
<td>0.5 Bi</td>
<td>3.2±0.6</td>
<td>0</td>
<td>187.0±1.7</td>
</tr>
<tr>
<td>1.0 Bi</td>
<td>2.8±0.6</td>
<td>12</td>
<td>181.9±2.0</td>
</tr>
</tbody>
</table>

The change ratio of different properties is calculated by using:

\[
\text{ratio} = \frac{\text{property of material 1} - \text{property of base alloy}}{\text{property of base alloy}} \times 100
\]  

(5.1)

The tensile fracture surface of Al 319.2 is shown in Fig. 5.5 (a). Fine dimples were observed on the surface resulting from the modified eutectic Si particles. The tensile fracture surfaces of Al 319.2 with 0.5 and 1.0 % Bi in Figs 5.5 (b) and (c) show that there are Bi particles present on the tensile fracture surfaces. Bi particles are the light colored particles shown in the SEM-BEI images as confirmed by the EDS spectrum. Because the eutectic Si particles in the 319.2 alloy modified with Bi are longer than those in the 319.2 base alloy, there are more fractured eutectic Si particles on the tensile fracture surface in the 319.2 modified with Bi. This can be seen in the cross-sections of tensile samples in Fig. 5.6. The eutectic Si particles near the tensile fracture surface of the Al 319.2 alloy in Fig. 5.6 (a) are refined and smaller than those in the Al 319.2 alloy with 1.0 % Bi shown in Fig. 5.6 (b). The larger eutectic Si particles in the 319.2 alloy modified by Bi reduce its tensile strength.

5.2.1.2.2 Alloy 319.2 with the addition of Sn

Table 5.2 lists the tensile properties of the 319.2 Al-Si alloy with different amounts of Sn additions and cast with a cooling rate of 9.4 °C/s. The tensile strength of
the 319.2 base alloy that was cast with a cooling rate of 9.4 °C/s was 214.2 ± 1.8 MPa, which decreased to 171.0 ± 1.8 MPa with 0.5 % Sn addition, to 162.0 ± 1.8 MPa with 1.0 % Sn addition and 173.2 ± 1.9 MPa with 0.5 % Bi and 0.2 % Sn additions.

**Table 5.2** Tensile properties of Al 319.2 modified by Sn and cast at a cooling rate of 9.4 °C/s

<table>
<thead>
<tr>
<th>Composition</th>
<th>Elongation (%</th>
<th>Change (%)</th>
<th>Tensile Strength (MPa)</th>
<th>Change (%)</th>
<th>YS (MPa)</th>
<th>Change (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0 Sn</td>
<td>3.2±0.4</td>
<td></td>
<td>214.2±1.8</td>
<td>0</td>
<td>141±0.6</td>
<td></td>
</tr>
<tr>
<td>0.5 Sn</td>
<td>2.7±0.5</td>
<td>-15</td>
<td>171.0±1.8</td>
<td>-20.2</td>
<td>90±0.5</td>
<td>-36.2</td>
</tr>
<tr>
<td>1.0 Sn</td>
<td>2.0±0.5</td>
<td>-37</td>
<td>162.0±1.8</td>
<td>-24.4</td>
<td>90±0.7</td>
<td>-36.2</td>
</tr>
<tr>
<td>0.5Bi0.25Sn</td>
<td>3.1±0.5</td>
<td>-3</td>
<td>173.2±1.9</td>
<td>-19.1</td>
<td>90±0.7</td>
<td>-36.2</td>
</tr>
</tbody>
</table>

The tensile fracture surfaces of Al 319.2 with 0.5 and 1.0 % Sn shown in Figs. 5.8 (a) and (b), demonstrate the effect of Sn on the tensile fracture surface. The Sn particles are mostly seen on the surface of eutectic Si particles, which is confirmed by the EDS spectrum in Fig. 5.9. There is evidence of more eutectic Si particles on the tensile fracture surface in alloys with Sn addition. Some of the eutectic Si particles near the tensile fracture surface of the Al 319.2 alloy are surrounded by Sn in Fig. 5.10. This is because that Sn can wet the grain boundary of Al, which weakens the interfacial bonding. All these factors reduce the tensile properties for the alloys with Sn addition.

5.2.1.3 Summary

In summary, both the additions of Bi and Sn to the 319.2 Al-Si alloy have effect on the microstructure, tensile properties and fracture surfaces of the modified alloys cast with a cooling rate of 9.4 °C/s. The effects are as following:
The distribution of Bi and Sn particles in the microstructure occurred at the interface of the Al matrix with the eutectic Si particles. The amount of Bi or Sn at the interface increased with an increase in the Bi or Sn addition.

The morphology of Bi and Sn particles in the microstructure was different; Bi was in the form of spherical particles, and the Sn phase was in a film morphology.

The morphology of the silicon particles changed from refined to coarse with an increase in the Bi addition. However, Sn had no strong effect on the silicon morphology.

The tensile strength decreased with the addition of Bi and Sn. The tensile strength of the alloys with Bi modification was higher than that of the alloys with Sn modification.

The area fraction of Sn on the tensile fracture surfaces was higher than Bi with 1.0 % addition to the alloys, and some of the eutectic Si particles near the tensile fracture surface of the Al 319.2 were surrounded by Sn, which indicates that Sn weakens the Al-Si interface bonding.

The reason for the decrease in the tensile strength is different for Bi and Si modifications; Bi changed the morphology of Si particles, while Sn weakened the Al-Si interface bonding.
5.2.2 Alloys Cast with a Cooling Rate of 26.0 °C/s

5.2.2.1 Distribution and size of Bi particles

5.2.2.1.1 Alloy 319.2 modified with Bi

The SEM-BEI micrographs of the 319.2 Al-Si alloy with 0.5 % and 1.0 % Bi additions cast with a cooling rate of 26 °C/s are shown in Figs. 5.11 (a) and (b), in which the bright particles are the pure Bi. They were present in the form of spherical particles. It was observed that Bi particles were formed mainly at the interface of the eutectic Si particles and occasionally at the interfaces of the intermetallic particles.

The size of the Bi Particles in the microstructure of the 319.2 Al-Si alloy with 0.5 % and 1.0 % Bi additions were 2.4 ± 1.0 μm and 2.9 ±1.5 μm respectively, when cast with a cooling rate of 26 °C/s, observed by SEM-BEI. The differences in the microstructure were mainly due the morphology of eutectic Si particles observed by optical microscopy. The average length of silicon particle was 3.6 μm in the Al-Si alloy with 1.0 % Bi, 3.0 μm in the alloy with 0.5 % Bi, and 2.1 μm in the alloy without Bi as shown in Table 4.10. These are refined compared to those at the cooling rate of 9.4 °C/s

5.2.2.1.2 Alloy 319.2 modified with Sn

The SEM-BEI images of the 319.2 Al-Si alloys with 0.5 % and 1.0 % Sn cast with a cooling rate of 26 °C/s are shown in Figs. 5.12 (a) and (b). The phase appearing bright is pure Sn, which formed mainly at the interface of the Al-Si particles and occasionally at the interfaces of the intermetallic particles. However, Sn was not present in the form of spherical particles like the Bi particles, because it tended to form a film at the boundaries of the Al grains. Sn has no strong effect on the silicon morphology, as observed by optical microscopy, and the average length of the silicon particle was 2.9 ± 1.6 μm in the Al-Si alloy with 1.0 % Sn, 2.9 ± 1.4 μm in the alloy with 0.5 % Sn, and 2.1
± 1.0 µm in the alloy without Sn, which shows that Sn has no effect on the silicon morphology, as shown in Table 4.10. In summary,

1. The detrimental effect of Bi on the morphology of Si at lower cooling rates (9.4 °C/s) is greatly decreased by using a high cooling rate (26 °C/s), and the morphology of Si particles is similar to that of the 319.2 base alloy.

2. The addition of Sn to the 319.2 Al-Si alloys has no effect on the microstructures at high cooling rates.

3. The distribution of Bi and Sn is mainly at the Al-Si interface.

5.2.2.2 Tensile strength, ductility and fracture behavior of 319.2 with different Bi and Sn concentrations

5.2.2.2.1 Alloy 319.2 modified with Bi

The tensile stress-strain curves of the 319.2 Al-Si alloys with different Bi concentrations cast with a cooling rate of 26 °C/s are shown in Fig. 5.13. It was observed that the tensile strength decreased with Bi addition. Table 5.3 lists the tensile properties of the 319.2 Al-Si alloy with different amounts of Bi added and cast with a cooling rate of 26 °C/s. Accordingly, the tensile strength of the 319.2 base alloy when case with a cooling rate of 26 °C/s was 287.2 ± 2.3 MPa. This decreased to 286.2 ± 2.0 MPa for the alloy with 0.5 % Bi, and to 266.1 ± 1.9 MPa for the alloy with 1.0 % Bi.

The tensile fracture surface of Al 319.2 is shown in Fig. 5.14 (a). It is observed that there are fine dimples on the surface resulting from the refined eutectic Si particles. The tensile fracture surfaces of Al 319.2 with 0.5 and 1.0 % Bi in Figs. 5.14 (b) and (c) show that there are Bi particles present on the tensile fracture surface. Bi particles are the particles that appear brighter in the SEM-BEI images. Fig. 5.15 is the cross-sections of tensile samples. The morphology of the eutectic Si particles near the tensile fracture
surface of Al 319.2 in Fig. 5.15 (a) is similar to that in the Al 319.2 modified with Bi, shown in Fig. 5.15 (b) and (c).

In general, the 319.2 Al-Si alloy with 0.5 % Bi had good elongation and tensile strength. The tensile strength of the alloy with 0.5 % Bi cast with a cooling rate of 26 °C/s maintained the mechanical properties of base alloy.

Table 5.3 Tensile properties of Al 319.2 modified by Bi cast at 26 °C/s

<table>
<thead>
<tr>
<th>Composition</th>
<th>Elongation</th>
<th>Tensile Strength</th>
<th>YS</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Elongation (%)</td>
<td>Change (%)</td>
<td>Tensile Strength (MPa)</td>
</tr>
<tr>
<td>0 Bi</td>
<td>10.5±0.9</td>
<td>287.2±2.3</td>
<td>0</td>
</tr>
<tr>
<td>0.5 Bi</td>
<td>13.8±1.3</td>
<td>31.4</td>
<td>286.2±2.0</td>
</tr>
<tr>
<td>1.0 Bi</td>
<td>8.3±1.1</td>
<td>-21.0</td>
<td>266.1±1.9</td>
</tr>
</tbody>
</table>

5.2.2.2.2 Alloy 319.2 modified with Sn

The typical tensile stress-strain curves of the 319.2 Al-Si alloys with different Sn additions cast with a cooling rate of 26 °C/s are shown in Fig. 5.16. It was observed that the tensile strength decreased when Sn was added, compared to that of the 319.2 Al-Si base alloys.

Table 5.4 lists the tensile properties of the 319.2 Al-Si alloy with different amounts of Sn added and cast with a cooling rates of 26 °C/s. The tensile strength of the 319.2 base alloy was 287.2 ± 2.3 MPa, which decreased to 206.0 ± 2.0 MPa with 0.5 % Sn, to 200.3 ± 1.9 MPa with 1.0 % Sn, and to 260.0 ± 1.8 MPa with 0.5 % Bi and 0.2 % Sn.

The tensile fracture surfaces of the Al 319.2 with 0.5 and 1.0 % Sn is shown in Figs. 5.17 (a) and (b), which show the presence of Sn on the tensile fracture surface. The
Sn particles were mostly seen on the surface of the eutectic Si particles, which is confirmed by the EDS spectrum. There were more eutectic Si particles on the tensile fracture surface the in alloys with Sn modification, and these were surrounded by Sn particles. This is because that Sn can wet the grain boundary of Al, and when they appear on the eutectic Si particles, it weakens the interfacial bonding, thereby reducing the tensile strength.

**Table 5.4** Tensile properties of Al 319.2 modified by Sn cast at 26 °C/s

<table>
<thead>
<tr>
<th>Composition</th>
<th>Elongation (%)</th>
<th>Change (%)</th>
<th>Tensile Strength (MPa)</th>
<th>Change (%)</th>
<th>YS (MPa)</th>
<th>Change (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0 Sn</td>
<td>10.5±0.9</td>
<td></td>
<td>287.2±2.3</td>
<td>0</td>
<td>159±0.5</td>
<td></td>
</tr>
<tr>
<td>0.5 Sn</td>
<td>5.9±0.8</td>
<td>-43.8</td>
<td>206.0±2.0</td>
<td>-28.3</td>
<td>115±0.5</td>
<td>-27.7</td>
</tr>
<tr>
<td>1.0 Sn</td>
<td>5.4±0.9</td>
<td>-48.6</td>
<td>200.3±1.9</td>
<td>-30.3</td>
<td>110±0.6</td>
<td>-30.8</td>
</tr>
<tr>
<td>0.5Bi0.25Sn</td>
<td>5.2±0.9</td>
<td>-50.5</td>
<td>260.0±1.8</td>
<td>-9.5</td>
<td>149±0.6</td>
<td>-6.3</td>
</tr>
</tbody>
</table>

**Fig. 5.18** shows the cross-sections of the tensile samples. The morphology of eutectic Si particles near the tensile fracture surface of Al 319.2 is similar to that observed in the Al 319.2 alloy modified with Sn. However, Si particles are surrounded by a dark phase, which indicates the presence of Sn around Si particles.

5.2.2.3 Summary

In summary, both the additions of Bi and Sn to the 319.2 Al-Si alloys have an effect on the microstructures, tensile properties and fracture behaviour of the alloys cast with a cooling rate of 26 °C/s. They can be summarized as follows:

(1) The distribution of Bi and Sn particles in the microstructure was observed at the interface of the Al matrix with the eutectic Si particles.
(2) The morphology of Bi was in the form of spherical particles, while the Sn phase assumed a thin film morphology.

(3) The morphology of the silicon particles with Bi or Sn addition was similar to that of base alloy.

(4) Sn weakened the Al-Si interface bonding, and also decreased the tensile strength.

(5) The tensile strength was maintained with the 0.5 % Bi modification, resulting from the refinement of the Bi and Si particles when cast with the higher cooling rate.

5.2.3 Summary of the Effects of Bi and Sn on the Microstructure and Tensile Properties

The microstructure and tensile properties of the 319 Al alloy with different amounts of Bi or Sn concentration and cast with different cooling rates can be summarized as following:

(1) The morphology of the Bi and Sn particles in the 319 alloy were different. Bi was in the form of spherical particles, while Sn phase assumed a thin film morphology.

(2) The distribution of Bi and Sn occurred at the interface of the Al matrix with the eutectic Si particles. This is because both Bi and Sn are insoluble in Si and Al, causing them to form at the eutectic Si/Al interface.

(3) The morphology of the silicon particles was strongly dependant on both the cooling rate and the amount of Bi addition. The size of Si particles changed marginally under high cooling rates with Bi addition. The tensile strength of the alloy with 0.5 % Bi was nearly that of the base alloy under the high cooling rate, but decreased under the lower cooling rates.
(4) The morphology of the silicon particles depended on the cooling rate, and changed marginally with Sn addition. Sn didn't affect the modification behavior of Sr. The tensile strength of the alloys with Sn decreased because Sn weakens the Al-Si interface bonding.

5.3. Dry Machining Performance at 0.42 m/s and 0.9 m/s

The 319.2 alloys modified with different Bi and Sn additions, and cast with different solidification rates, were subjected to dry orthogonal cutting tests using the parameters detailed in the Chapter 3. The test methodology was the same for all the alloys cast with the two different cooling rates, 9.4°C/s (in Sec. 5.3.1) and 26°C/s (in Sec. 5.3.2).

5.3.1 Alloys Cast with a Cooling Rate of 9.4 °C/s

5.3.1.1 Cutting and thrust forces

Figure 5.19 shows the typical cutting force- and thrust force-time curves for the 319.2 base alloy cast with a cooling rate of 9.4 °C/s and dry turned with a speed of 0.9 m/s and feed rate of 0.25 mm/rev. It was observed that the change in force occurs at four periods from the beginning to the end of the test as follows:

(i) The force - time curves were horizontal before the dry turning began

(ii) The curves varied with time during dry turning

(iii) The forces dropped suddenly when the turning was stopped, but did not return to the initial value

(iv) At the end of the test, the forces returned to the initial value after removing the cutting tool. The average cutting and thrust forces reported are the average values of the relative forces in period (ii)
5.3.1.1.1 Alloy 319.2 with the addition of Bi

Two speeds were used to test 319.2 with Bi additions, 0.42 and 0.90 m/s, and the results of these are reported below:

(1) Turning with a speed of 0.42 m/s

The average cutting force at a speed of 0.42 m/s was 416 N for the base alloy, and 361 N for the alloy with 0.5 % Bi which was cast with a cooling rate of 9.4 °C/s, a decrease of 13 % as shown in Table 5.5 and Fig. 5.20 (a). The average cutting force was 346 N for the alloy with 1.0 % Bi, lower by 4% with an increase in modification from 0.5 % to 1.0 %. Bi. The decrease in the cutting force was less significant when Bi modification was further increased to 1.0 %.

The average thrust force generated during dry turning at 0.42 m/s followed the same trend with respect to the cutting force. The thrust force decreased with an increase in the Bi addition as demonstrated in Table 5.5 and Fig. 5.20 (b). The average thrust force for the 319.2 base alloy was 91 N, 69.3 N for the alloy with 0.5 % Bi, and 67 N for the alloy with 1.0 % Bi, when cast with a cooling rate of 9.4 °C/s.

Table 5.5 Average cutting forces and thrust forces generated during the dry turning of Al 319.2 modified by Bi cast at 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Cutting Force</th>
<th>Thrust Force</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Cutting Force (N)</td>
<td>Change (%)</td>
</tr>
<tr>
<td>319.2</td>
<td>416.1±20.8</td>
<td>0</td>
</tr>
<tr>
<td>0.5Bi</td>
<td>361±18.1</td>
<td>-13.2</td>
</tr>
<tr>
<td>1.0Bi</td>
<td>346±17.3</td>
<td>-16.8</td>
</tr>
<tr>
<td>319.2</td>
<td>403.1±20.2</td>
<td>0</td>
</tr>
<tr>
<td>0.5Bi</td>
<td>343.3±17.2</td>
<td>-14.8</td>
</tr>
<tr>
<td>1.0Bi</td>
<td>340.3±17.0</td>
<td>-15.6</td>
</tr>
</tbody>
</table>

(2) Turning with a speed of 0.9 m/s
The average cutting force at a turning speed of 0.9 m/s was 403 N for the 319.2 base alloy, and 343.3 N for the alloy with 0.5 % Bi, a decrease of 15 % as shown in Table 5.5 and Fig. 5.20 (a). The average cutting force was 340.3 N for the alloy with 1.0 % Bi. The decrease in cutting force was less significant with a further increase in the Bi amount from 0.5 % to 1.0 %, when cast with a cooling rate of 9.4 °C/s.

The average thrust forces generated during turning followed the same trend as the cutting force. The thrust force decreased with the increase in the Bi addition as demonstrated in Fig. 5.20 (b). When cast with a cooling rate of 9.4 °C/s, the average thrust force for the Al-Si base alloy was 95 N, 72 N for the alloy with 0.5 % Bi, and 68 N for the alloy with 1.0 % Bi.

(3) COF and specific cutting energy during dry turning

The coefficient of friction ($\mu$) at the tool face can be computed:

$$\mu = \frac{F_f}{N_f} = \frac{F_c \sin \alpha + F_i \cos \alpha}{F_c \cos \alpha - F_i \sin \alpha} = \tan \beta$$

(2.9)

Since $\alpha = 0$ for the PCD tool used in the study, Eqn.2.5 was reduced to:

$$\mu_e = \frac{F_t}{F_c}$$

(5.2)

which was used for estimating the coefficient of friction for the conditions studied.

The calculated COF for the 319.2 alloy with different amounts of Bi additions cast with a cooling rate of 9.4 °C/s are shown in Table 5.6. It was observed that the calculated COF decreased from 0.22 for the base alloy to 0.19 for the alloy with 1.0 % Bi.
at a turning speed of 0.42 m/s. COF decreased from 0.24 for the base alloy to 0.20 for the alloy with 1.0 % Bi at the turning speed of 0.9 m/s.

The specific cutting energy or the energy consumed per unit volume of the material removed during cutting was calculated for the different cutting conditions, using the average cutting and thrust forces for each condition, as shown below:

\[ U = V_c (F_c + F_t) \]  
(5.3)

\[ R = d . f . V_c \]  
(5.4)

\[ u_s = \left( \frac{U}{R} \right) = \left( \frac{F_c + F_t}{tf} \right) \]  
(5.5)

where \( U \) is total energy consumed per unit time or power, \( F_c \) is the cutting force, \( F_t \) equals the thrust force, ‘\( d \)’ is the depth of cut, ‘\( f \)’ equals the feed rate, ‘\( R \)’ is the material removal rate, and \( u_s \) is the specific power required for cutting. The feeding rate is 0.25 mm/rev (Sec. 3.5.1).

Furthermore, the specific cutting energy was calculated and is listed in Table 5.6. The specific cutting energy decreases from 676.6 MJm\(^{-3}\) for the 319.2 base alloy to 550.7 MJm\(^{-3}\) for the 319.2 alloy with 1.0 % Bi at the turning speed of 0.42 m/s, and it decreases from 664.1 MJm\(^{-3}\) for the 319.2 base alloy to 544.4 MJm\(^{-3}\) for the 319.2 alloy with 1.0 % Bi at the turning speed of 0.9 m/s, all cast with a cooling rate of 9.4 °C/s.
Table 5.6 COF and specific cutting energy of Al 319.2 modified by Bi cast at 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>COF</th>
<th>Change (%)</th>
<th>Specific cutting energy (MJ/m³)</th>
<th>Change (%)</th>
<th>Speed (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td>319.2</td>
<td>0.22</td>
<td></td>
<td>676.6</td>
<td></td>
<td>0.42</td>
</tr>
<tr>
<td>0.5Bi</td>
<td>0.19</td>
<td>-13.6</td>
<td>573.7</td>
<td>-15.2</td>
<td></td>
</tr>
<tr>
<td>1.0Bi</td>
<td>0.19</td>
<td>-13.6</td>
<td>550.7</td>
<td>-18.6</td>
<td></td>
</tr>
<tr>
<td>319.2</td>
<td>0.24</td>
<td>0.0</td>
<td>664.1</td>
<td>0.0</td>
<td>0.9</td>
</tr>
<tr>
<td>0.5Bi</td>
<td>0.21</td>
<td>-12.5</td>
<td>553.7</td>
<td>-16.6</td>
<td></td>
</tr>
<tr>
<td>1.0Bi</td>
<td>0.20</td>
<td>-16.7</td>
<td>544.4</td>
<td>-18.0</td>
<td></td>
</tr>
</tbody>
</table>

In summary, the average forces generated during the turning of the 319.2 alloys varied with the amount of Bi modification and the cutting speed. Both the cutting force and the thrust force decreased with Bi addition, as did the specific cutting energy and COF, which indicate that Bi improved the machining performance. The cutting force decreased with an increase in the cutting speed. However, the thrust force decreased with a decrease in the cutting speed. This caused the COF to decrease with a decrease in the cutting speed.

5.3.1.1.2 Alloy 319.2 modified with Sn

The same two cutting speeds were used to test the dry turning behavior of the 319.2 alloy modified with Sn additions:

(1) Turning with a speed of 0.42 m/s

It is demonstrated in Table 5.7 and Fig. 5.21 (a) that the average cutting force of 319.2 alloys was 416.1 N for the base alloy that was cast with a cooling rate of 9.4 °C/s, which decreased to 350 N for the alloy with 0.5 % Sn, and to 347 N for the alloy with 1.0 % Sn. The decrease in cutting force was insignificant with a further increase in the amount of Sn from 0.5 % to 1.0 %.
As listed in Table 5.7 and Fig. 5.21 (b) the average thrust force generated during dry turning at 0.42 m/s followed the same trend as the cutting force, that is, the thrust force decreased with the increase in Sn addition. The average thrust force for the 319.2 base alloy was 91.4 N, which decreased to 75 N for the alloy with 0.5 % Sn, to 69 N for the alloy with 1.0 % Sn cast at the cooling rate of 9.4 °C/s.

(2) Turning with a speed of 0.9 m/s

As shown in Table 5.8 and Fig. 5.21(a), the average cutting force at this speed was 403.1 N for the 319.2 base alloy, 338.4 N for the alloy with 0.5 % Sn, and 337 N for the alloy with 1.0 % Sn. The decrease in cutting force was insignificant with a further increase in the Sn amount from 0.5 % to 1.0 %.

From Table 5.8 and Fig. 5.21(b) it can be seen that the average thrust force generated during cutting followed the same trend as the cutting force, that is, the thrust force decreased with an increase in Sn addition. The average thrust force for the Al-Si base alloy was 95 N, which decreased to 77 N for the alloy with 0.5 % Sn, and to 72 N for the alloy with 1.0 % Sn.

### Table 5.7 Average cutting and thrust forces of Al 319.2 modified by Sn cast at 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Cutting force</th>
<th>Thrust force</th>
<th>Speed (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Cutting force</td>
<td>Change</td>
<td>Thrust force</td>
</tr>
<tr>
<td>319.2</td>
<td>416.1±20.8</td>
<td>-15.9</td>
<td>91.4±4.6</td>
</tr>
<tr>
<td>0.5Sn</td>
<td>350.0±17.5</td>
<td>-16.6</td>
<td>75.0±3.8</td>
</tr>
<tr>
<td>1.0Sn</td>
<td>347.0±17.4</td>
<td>-16.4</td>
<td>69.0±3.5</td>
</tr>
<tr>
<td>0.5Bi0.25Sn</td>
<td>297.3±14.9</td>
<td>-28.6</td>
<td>56.8±2.8</td>
</tr>
<tr>
<td>319.2</td>
<td>403.1±20.2</td>
<td>-16.1</td>
<td>95.0±4.8</td>
</tr>
<tr>
<td>0.5Sn</td>
<td>338.4±16.9</td>
<td>-16.4</td>
<td>77.0±3.9</td>
</tr>
<tr>
<td>1.0Sn</td>
<td>337.0±16.9</td>
<td>-16.4</td>
<td>72.0±3.6</td>
</tr>
<tr>
<td>0.5Bi0.25Sn</td>
<td>289.8±14.5</td>
<td>-28.1</td>
<td>56.2±2.8</td>
</tr>
</tbody>
</table>
(3) Calculated COF and specific cutting energy during dry turning

Table 5.8 shows that calculated COF of the 319.2 alloy modified with different amounts of Sn and cast with a cooling rate of 9.4 °C/s decreased from 0.22 for the base alloy to 0.20 for the alloy with 1.0 % Sn when machined with a turning speed of 0.42 m/s, and it decreased from 0.24 for the base alloy to 0.21 for the alloy with 1.0 % Bi when machined with a turning speed of 0.9 m/s.

The specific cutting energy that was calculated is listed in Table 5.8. It was observed that the specific cutting energy decreased from 676.6 MJm⁻³ for the 319.2 base alloy to 554.7 MJm⁻³ for the 319.2 alloy modified with 1.0 % Sn when machine with a turning speed of 0.42 m/s. When turned with a speed of 0.9 m/s, it decreased from 664.1 MJm⁻³ for the 319.2 base alloy to 545.3 MJm⁻³ for the 319.2 alloy with 1.0 % Sn.

**Table 5.8** COF and specific cutting energy of Al 319.2 modified by Sn cast at 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>COF</th>
<th>COF Change (%)</th>
<th>Speed (m/s)</th>
<th>Specific cutting energy</th>
<th>Specific cutting energy Change (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>319.2</td>
<td>0.22</td>
<td></td>
<td>0.42</td>
<td>676.6</td>
<td></td>
</tr>
<tr>
<td>0.5Sn</td>
<td>0.21</td>
<td>-4.5</td>
<td></td>
<td>566.7</td>
<td>-16.2</td>
</tr>
<tr>
<td>1.0Sn</td>
<td>0.20</td>
<td>-9.1</td>
<td></td>
<td>554.7</td>
<td>-18.0</td>
</tr>
<tr>
<td>0.5Bi0.25Sn</td>
<td>0.19</td>
<td>-13.6</td>
<td></td>
<td>472.1</td>
<td>-30.2</td>
</tr>
<tr>
<td>319</td>
<td>0.24</td>
<td></td>
<td>0.9</td>
<td>664.1</td>
<td></td>
</tr>
<tr>
<td>0.5Sn</td>
<td>0.23</td>
<td>-4.2</td>
<td></td>
<td>553.9</td>
<td>-16.6</td>
</tr>
<tr>
<td>1.0Sn</td>
<td>0.21</td>
<td>-12.5</td>
<td></td>
<td>545.3</td>
<td>-17.9</td>
</tr>
<tr>
<td>0.5Bi0.25Sn</td>
<td>0.19</td>
<td>-20.8</td>
<td></td>
<td>461.3</td>
<td>-30.5</td>
</tr>
</tbody>
</table>

In summary, the average forces of dry machining of the 319.2 alloys varied with the addition of Sn and the cutting speed. Both the cutting force and thrust force decreased with increasing Sn addition, and the specific cutting energy and the COF also decreased with increasing Sn addition. These indicate that Sn could improve the machining
performance. The alloy with 0.5 % Bi and 0.25 % Sn has the lowest cutting and thrust forces, and the lowest specific cutting energy compared with either Bi or Sn addition (Tables 5.5 - 5.8), which indicates that the combined addition of Bi and Sn is more efficient in improving the machining performance.

5.3.1.2 Interfacial temperature

The contact surface temperature between the chips and the rake face of the WC cutting tool was measured using a 0.5 mm diameter K-type (Chromel-Alumel) thermocouple, which was inserted into the WC cutting tool, 0.2 mm under the rake face and through a 0.5 mm diameter blind hole made by EDM.

For the 319.2 alloys modified with Bi or Sn and cast with a cooling rate of 9.4 °C/s, the temperature-time curves recorded during dry turning are presented in Figs. 5.22 and 5.23. The interfacial temperature increased quickly at the beginning, and then reached an equilibrium temperature. The 319.2 base alloys turned at 0.42 m/s developed a higher temperature than the other alloys modified with Bi or Sn, and the temperature was also higher when the turning speed was 0.9 m/s than 0.42 m/s.

5.3.1.2.1 Alloy 319.2 modified with Bi

For the alloys cast with a cooling rate of 9.4 °C/s, and dry turning was performed with a speed of 0.42 m/s, Table 5.9 shows that the temperature that was developed between the chips and the tool rake face decreased from 249.7 °C for the 319.2 base alloy to 164.5 °C for the alloy with 0.5 % Bi, and to 130.3 °C for the alloy with 1.0 % Bi. Furthermore, when the dry turning speed was 0.9 m/s, the temperature decreased from 267.6 °C for the base alloy to 214.6 °C for the alloy with 0.5 % Bi, and to 154.0 °C for the alloy with 1.0 % Bi (Figs. 5.24). The low melting point of Bi and the distribution of the Bi particles within the alloys are the reason for the low temperature at tool rake face.
Table 5.9 Interfacial temperature between the chips and the rake face of the WC cutting tool for Al 319.2 modified by Bi cast at 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Temperature</th>
<th>Change (%)</th>
<th>Speed (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Temperature</td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>°C</td>
<td>%</td>
<td></td>
</tr>
<tr>
<td>319.2</td>
<td>249.7±12.5</td>
<td>-34.1</td>
<td>0.42</td>
</tr>
<tr>
<td>0.5Bi</td>
<td>164.5±8.2</td>
<td>-47.8</td>
<td></td>
</tr>
<tr>
<td>1.0Bi</td>
<td>130.3±6.5</td>
<td></td>
<td></td>
</tr>
<tr>
<td>319.2</td>
<td>267.6±13.4</td>
<td>-34.1</td>
<td>0.9</td>
</tr>
<tr>
<td>0.5Bi</td>
<td>214.6±10.7</td>
<td>-19.8</td>
<td></td>
</tr>
<tr>
<td>1.0Bi</td>
<td>154.0±7.7</td>
<td>-42.5</td>
<td></td>
</tr>
</tbody>
</table>

Table 5.10 Interfacial temperature between the chips and the rake face of the WC cutting tool for Al 319.2 modified by Sn cast at 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Temperature</th>
<th>Change (%)</th>
<th>Speed (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Temperature</td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>°C</td>
<td>%</td>
<td></td>
</tr>
<tr>
<td>319.2</td>
<td>249.7±12.5</td>
<td>-37.0</td>
<td>0.42</td>
</tr>
<tr>
<td>0.5Sn</td>
<td>157.3±7.9</td>
<td>-53.1</td>
<td></td>
</tr>
<tr>
<td>1.0Sn</td>
<td>117.1±5.9</td>
<td></td>
<td></td>
</tr>
<tr>
<td>0.5Bi-0.25Sn</td>
<td>138.0±6.9</td>
<td>-44.7</td>
<td></td>
</tr>
<tr>
<td>319.2</td>
<td>267.6±13.4</td>
<td>-31.5</td>
<td>0.9</td>
</tr>
<tr>
<td>0.5Sn</td>
<td>183.3±9.2</td>
<td>-49.7</td>
<td></td>
</tr>
<tr>
<td>1.0Sn</td>
<td>134.7±6.7</td>
<td></td>
<td></td>
</tr>
<tr>
<td>0.5Bi-0.25Sn</td>
<td>159.0±8.0</td>
<td>-40.6</td>
<td></td>
</tr>
</tbody>
</table>

5.3.1.2.2 Alloy 319.2 modified with Sn

For alloys cast with a cooling rate of 9.4 °C/s, and dry machined with a turning speed of 0.42 m/s, Table 5.10 shows that the temperature was 249.7 °C for the base alloy, 157.3 °C for the alloy with 0.5 % Sn, and 117.1 °C for the alloy with 1.0 % Sn. Furthermore, when the dry turning speed was 0.9 m/s, the temperature developed was
267.6 °C for the base alloy, 183.3 °C for the alloy with 0.5 % Sn, and 134.7 °C for the alloy with 1.0 % Sn (Figs. 5.25).

It is also observed from Figs. 5.24 and 5.25, and Tables 5.9 and 5.10 that, when the 319.2 base alloy was cast with a cooling rate of 9.4 °C/s, the temperature generated at the interface between the chips and the rake face of the WC cutting tool decreased with the addition of Bi or Sn during dry turning, and the temperature of the interface was lower in alloys with Sn than those with Bi.

5.3.1.3 Surface roughness

The surface finish of the 319.2 base alloy and the alloy with 1.0 % Bi solidified at a cooling rate of 9.4 °C/s, observed by SEM, are shown in Figs. 5.26 (a) and (b). The machined surfaces of the alloys with Bi had better finish.

The average surface roughness of the 319.2 alloys with different amounts of Bi (Table 5.11) and Sn (Table 5.12) addition were very similar, and were in the range of 0.808 µm to 1.033 µm (Rₐ) when machined with a turning speed of 0.42 m/s, and in the range of 0.877 µm to 1.114 µm for a turning speed of 0.9 m/s. The finished surfaces of the 319.2 alloys with Bi or Sn addition have no significant difference.

The average surface roughness of the 319.2 alloys with different amounts of Bi (Table 5.11) and Sn (Table 5.12) addition were very similar, and were in the range of 0.808 µm to 1.033 µm (Rₐ) when machined with a turning speed of 0.42 m/s, and in the range of 0.877 µm to 1.114 µm for a turning speed of 0.9 m/s. The finished surfaces of the 319.2 alloys with Bi or Sn addition have no significant difference.
Table 5.11 Average surface roughness of Al 319.2 modified by Bi and cast at a cooling rate of 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Surface roughness</th>
<th>Speed (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Surface Roughness (μm)</td>
<td>Change (%)</td>
</tr>
<tr>
<td>319.2</td>
<td>1.033±0.323</td>
<td>0.42</td>
</tr>
<tr>
<td>0.5Bi</td>
<td>0.833±0.278</td>
<td>-19.4</td>
</tr>
<tr>
<td>1.0Bi</td>
<td>0.808±0.277</td>
<td>-21.8</td>
</tr>
<tr>
<td>319.2</td>
<td>1.114±0.331</td>
<td>0.9</td>
</tr>
<tr>
<td>0.5Bi</td>
<td>0.955±0.271</td>
<td>-14.3</td>
</tr>
<tr>
<td>1.0Bi</td>
<td>0.877±0.270</td>
<td>-21.3</td>
</tr>
</tbody>
</table>

Table 5.12 Average surface roughness of Al 319.2 modified by Sn and cast at a cooling rate of 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Surface roughness</th>
<th>Speed (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Surface Roughness (μm)</td>
<td>Change (%)</td>
</tr>
<tr>
<td>319.2</td>
<td>1.033±0.323</td>
<td>0.42</td>
</tr>
<tr>
<td>0.5Sn</td>
<td>0.899±0.243</td>
<td>-13.0</td>
</tr>
<tr>
<td>1.0Sn</td>
<td>0.812±0.270</td>
<td>-21.4</td>
</tr>
<tr>
<td>0.5Bi-0.25Sn</td>
<td>0.841±0.266</td>
<td>-18.6</td>
</tr>
<tr>
<td>319</td>
<td>1.114±0.331</td>
<td>0.9</td>
</tr>
<tr>
<td>0.5Sn</td>
<td>0.922±0.248</td>
<td>-17.2</td>
</tr>
<tr>
<td>1.0Sn</td>
<td>0.897±0.261</td>
<td>-19.5</td>
</tr>
<tr>
<td>0.5Bi-0.25Sn</td>
<td>0.790±0.260</td>
<td>-29.1</td>
</tr>
</tbody>
</table>

5.3.1.4 Cross-sectional microstructure of the chip-root

The microstructure through the chip-root cross-sections (CRCS) show evidence of shear deformation during dry turning. The position of the CRCS is indicated at the centre of the chip as shown in Fig. 5.27. The chips were collected for study by interrupting the cutting process.
5.3.1.4.1 Alloy 319.2 modified with Bi

**Figure 5.28** shows the cross-sectional optical micrographs of the chips of the alloys modified with 0.5 % Bi that were turned at a speed of 0.42 m/s. The segments are highly serrated, almost separated. In the SEM–BEI image in Fig. 5.28 (b) the area near the tool tip shows elongated Bi stringers that resulted from the large shear plastic deformation.

**Figure 5.29** shows the cross-sectional micrographs of chip root taken by interrupting the cutting process of the 319 alloy modified with 1.0 % Bi at 0.9 m/s. The optical micrograph shown in **Fig. 5.29 (a)** shows that the segment is separated by micro-cracks. The average length of chip segments was 340 ± 10 µm for the 319 alloy modified with 1.0 % Bi, and 450 ± 20 µm for the 319 base alloy (**Fig. 5.30**). The EDS maps of the tool tip areas in **Fig. 5.29 (b)** show that the Bi particles had elongated into stingers, indicating that the Bi particles melted during the cutting process, which were also present on the surface facing the tool rake face. The COF between the chips and tool rake face was decreased possibly due to the Bi melting and acting as a lubricant. These promoted the segmentation, and is seen in the side surface images of the chips formed from the alloys with Bi addition (**Fig 5.31**).

The free surface of chips observed by SEM (**Fig. 5.32**) shows that chips formed from the 319.2 alloy with 0.5 % Bi (cast with cooling rate of 9.4 °C/s) machined using a diamond tool at 0.42 m/s are discontinuous type of chips.

5.3.1.4.2 Alloy 319.2 modified with Sn

The side surfaces of the 319.2 base alloy, the alloy with 0.5 % Sn, and the alloy with 1.0 % Sn, cast with a cooling rate of 9.4 °C/s using a diamond cutting tool are shown in **Figs. 5.33**. It is shown that all the segments are separated from each other. These chips
are discontinuous type of chips. Figure 5.34 is the free surface of the chip in Figs. 5.33 (b).

Figure 5.35 (a) presents the cross-sectional optical micrographs of the chips formed from the alloys with 1.0 % Sn machined at 0.9 m/s, in which the segments are separated. The area near the tool tip was observed by SEM–BEI image (Figure 5.35 (b)), which shows melted Sn stringers resulting from the large shear plastic deformation during cutting.

In summary, all of these alloys generated discontinuous-type chips, with most of the short chip segments loosely attached to each other. The cross-section of the chip formed from the 319.2 base alloy exhibits cracks; however, their occurrence is 20% less than that formed from the 1.0 % Bi or Sn alloy, and the chip has a less segmented appearance. The average length of the chip segments decreased with the addition of Bi or Sn in the 319 base alloy (Fig. 5.30).

5.3.1.5 Chip morphology

The chip morphology of the 319.2 alloys, modified with different concentrations of Bi and Sn, and cut at 0.42 m/s and 0.9 m/s using a diamond tool, is shown in Figs. 5.36, 5.37 and 5.38, and listed in Tables 5.13 and 5.14. The length of chips for the 319.2 base alloy in Fig. 5.36 (a) cut at 0.42 m/s is longer than that of the 319 base alloy with 1.0 % Bi as demonstrated in Fig. 5.36 (b). The length of chips for the 319 base alloy in Fig. 5.36 (c) cut at 0.9 m/s is longer than that of 319 base alloy cut at 0.42 m/s as shown in Fig. 5.36 (a). However, the length of chips for the 319.2 alloy with 1.0 % Bi cut at 0.9 m/s shown in Fig. 5.36 (d) is not increased when compared with that cut at 0.42 m/s as shown in Fig. 5.36 (b).
The length of chips for the 319.2 alloy with 1.0 % Sn cut at 0.42 m/s in Fig. 5.37 (a) and 0.9 m/s shown in Fig. 5.37 (b) followed the same trend as that of the alloys with Bi addition. The length of chips was shown in Fig. 5.38 for comparing.

Table 5.13 Length of the chips formed from Al 319.2 modified by Bi and cast at 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Length</th>
<th>Speed (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Length (mm)</td>
<td>Change (%)</td>
</tr>
<tr>
<td>319.2</td>
<td>21.0±1.1</td>
<td></td>
</tr>
<tr>
<td>0.5Bi</td>
<td>11.0±0.6</td>
<td>-47.6</td>
</tr>
<tr>
<td>1.0Bi</td>
<td>9.0±0.5</td>
<td>-57.1</td>
</tr>
</tbody>
</table>

Table 5.14 Length of the chips formed from Al 319.2 modified by Sn and cast at 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Length</th>
<th>Speed (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Length (mm)</td>
<td>Change (%)</td>
</tr>
<tr>
<td>319</td>
<td>21.0±1.1</td>
<td></td>
</tr>
<tr>
<td>0.5Sn</td>
<td>11.0±0.6</td>
<td>-47.6</td>
</tr>
<tr>
<td>1.0Sn</td>
<td>9.0±0.5</td>
<td>-57.1</td>
</tr>
<tr>
<td>0.5Bi-0.25Sn</td>
<td>9.0±0.5</td>
<td>-57.1</td>
</tr>
</tbody>
</table>

The chip morphology was strongly influenced by the tool material. Fig. 5.39 is the chip morphology formed by using a WC cutting tool for alloys with different amounts of Bi and Sn modification and cast with a cooling rate of 9.4 °C/s. The chips are of the continuous type for the 319 base alloy cut at 0.42 m/s and 0.9 m/s, as demonstrated by
Figs. 5.39 (a) and (b), however, the chips are discontinuous for alloys with Bi modification, as shown in Fig. 5.39 (c) and for alloys with Sn modification, as shown in Fig. 5.39 (d). It was observed that higher cutting speeds generated longer chips for the base alloy, but the chip length decreased in alloys with Bi and Sn modification.

5.3.1.6 Summary

The average cutting and thrust forces and the temperature generated at the tool rake face during dry turning of the 319.2 alloys with different amounts of Bi and Sn modifications cast with a cooling rate of 9.4 °C/s, were measured at turning speeds of 0.42 m/s and 0.9 m/s. The corresponding COF and specific cutting energy were calculated. The average surface roughness, microstructure of the chip-root cross-section, and the morphology of the chips were also studied. The machining characteristics are summarized as following:

1. Both the cutting force and the thrust force decrease with the addition of Bi and Sn. Sn decreases the machining forces more than Bi.
2. The specific cutting energy follows the same trend as the cutting force.
3. The temperature generated at the interface between the chip and the rake face of WC cutting tool decreases with the addition of Bi and Sn during dry turning. Sn is more effective in decreasing the interface temperature than Bi.
4. COF also decreases with the addition of Bi and Sn.
5. The alloy with 0.5 % Bi and 0.25 % Sn had the lowest cutting and thrust forces and the lowest specific cutting energy of all the alloys studied, which indicates that the combined modification by Bi and Sn is more effective in improving the machining performance.
6. The length of the chips formed from the 319 alloys modified with different amounts of Bi and Sn addition were lower than that of the base alloys. All chips were of the discontinuous type.

7. The elongated Bi and Sn stringers resulting from the large shear plastic deformation promote the segmentation of the chips.

8. The addition of Bi and Sn improved the machining performance of the 319 alloys

**5.3.2 Alloys Cast with a Cooling Rate of 26.0°C/s**

5.3.2.1 Cutting and thrust forces

5.3.2.1.1 Alloy 319.2 modified with Bi

Two turning speeds were used to test the 319.2 alloy modified with Bi:

(1) Turning with a speed of 0.42 m/s

The average cutting force for the 319.2 alloy at a cutting speed of 0.42 m/s was 524.4 N for the base alloy, 447.0 N for the alloy with 0.5% Bi which was cast with a cooling rate of 26 °C/s, a decrease of 15% ([Table 5.15, Fig. 5.40(a)]. The average cutting force was 429.1N for the alloy with 1.0% Bi, a decrease of 4% for an increase from 0.5% to 1.0%. Bi. The decrease in cutting force was less significant with a further increase in the Bi amount, from 0.5% to 1.0%.

The average thrust force generated during dry turning at 0.42 m/s decreased with an increase in modification. The thrust force decreased with an increase in Bi addition as shown in Table 5.15 and Fig. 5.40(b). The average thrust force for the 319.2 base alloy was 112.7 N, 96.1N for the alloy with 0.5% Bi, and 86.6 N for the alloy with 1.0% Bi, cast with a cooling rate of 26 °C/s during casting.
Table 5.15 Average cutting and thrust forces of Al 319.2 modified by Bi cast at 26 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Cutting force</th>
<th>Thrust force</th>
<th>Speed (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Cutting force</td>
<td>Change</td>
<td>Thrust force</td>
</tr>
<tr>
<td>319.2</td>
<td>524.4±26.2</td>
<td></td>
<td>112.7±5.6</td>
</tr>
<tr>
<td>0.5Bi</td>
<td>447.0±22.4</td>
<td>-14.8</td>
<td>96.1±4.8</td>
</tr>
<tr>
<td>1.0Bi</td>
<td>429.1±21.5</td>
<td>-18.2</td>
<td>86.6±4.3</td>
</tr>
<tr>
<td>319.2</td>
<td>502.7±25.1</td>
<td></td>
<td>121.0±6.1</td>
</tr>
<tr>
<td>0.5Bi</td>
<td>434.7±21.7</td>
<td>-13.5</td>
<td>99.0±5</td>
</tr>
<tr>
<td>1.0Bi</td>
<td>427.5±21.4</td>
<td>-15.0</td>
<td>89.0±4.5</td>
</tr>
</tbody>
</table>

Table 5.16 COF and specific cutting energy of Al 319.2 alloys modified by Bi cast at 26 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>COF</th>
<th>Specific cutting energy</th>
<th>Speed (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>COF</td>
<td>Change (%)</td>
<td>Specific cutting energy (MJm⁻³)</td>
</tr>
<tr>
<td>319.2</td>
<td>0.21</td>
<td>0</td>
<td>849.5</td>
</tr>
<tr>
<td>0.5Bi</td>
<td>0.21</td>
<td>0.0</td>
<td>724.1</td>
</tr>
<tr>
<td>1.0Bi</td>
<td>0.20</td>
<td>-4.8</td>
<td>687.6</td>
</tr>
<tr>
<td>319.2</td>
<td>0.24</td>
<td>0.0</td>
<td>831.6</td>
</tr>
<tr>
<td>0.5Bi</td>
<td>0.23</td>
<td>-4.2</td>
<td>711.6</td>
</tr>
<tr>
<td>1.0Bi</td>
<td>0.21</td>
<td>-12.5</td>
<td>688.7</td>
</tr>
</tbody>
</table>

The calculated COF and specific cutting energy of the 319.2 alloy with different amounts of Bi addition that were cast with a cooling rate of 26 °C/s are listed in Table 5.16. The COF decreased from 0.21 for the base alloy to 0.20 for the 319.2 alloy with 1.0 % Bi. The specific cutting energy decreased from 849.5 MJm⁻³ for the 319.2 base alloy to 724.1 MJm⁻³ for the 319.2 alloy with 0.5 % Bi, and to 687.6 MJm⁻³ for the 319.2 alloy with 1.0 % Bi.

(2) Turning with a speed of 0.9 m/s
The average cutting force at 0.9 m/s was 502.7 N for the 319.2 base alloy, and 434.7 N for the alloy with 0.5 % Bi, a decrease of 13.5 % as shown in Table 5.15 and Fig. 5.40 (a). The average cutting force was 427.5 N for the alloy with 1.0 % Bi when cast with a cooling rate of 26 °C/s, a decrease of 1.7% with an increase in modification from 0.5 % to 1.0 %. Bi, The decrease in the cutting force is less significant with a further increase in the Bi amount, from 0.5 % to 1.0 %.

The average thrust forces generated during cutting decreased linearly with the increase in Bi addition shown in Fig. 5.40(b). When cast with a cooling rate of 26 °C/s, the average thrust force for the Al-Si base alloy was 121.0 N, 99.0 N for the alloy with 0.5 % Bi, a decrease of 18 %. The average thrust forces for the alloy with 1.0 % Bi is 89.0 N, a decrease of 10% with an increase from 0.5 % to 1.0 %. Bi, The decrease in the cutting force is less significant with a further increase in the Bi amount, from 0.5 % to 1.0 %.

The COF decreases from 0.24 for the base alloy to 0.21 for the alloy with 1.0 % Bi machined with a turning speed of 0.9 m/s, a decrease of 12.5 %. The specific cutting energy decreases from 831.6 MJm⁻³ for the 319.2 base alloy to 688.7 MJm⁻³ for the 319.2 alloy with 1.0 % Bi machine with a turning speed of 0.9 m/s, when cast with a cooling rate of 26 °C/s.

5.3.2.1.2 Alloy 319.2 modified with Sn

Two speeds were also used to test the 319.2 alloy modified with Sn additions:

(1) Turning with a speed of 0.42 m/s

The average cutting force for the 319.2 alloy machine with a speed of 0.42 m/s was 524.4 N for the base alloy, and 380.0 N for the alloy with 0.5 % Sn, when cast with a
cooling rate of 26 °C/s [Table 5.17, Fig. 5.41 (a)], a decrease of 27.5 %. The average cutting force was 367.0 N for the alloy with 1.0 % Sn. The decrease in cutting force was less significant with a further increase in the Sn amount from 0.5 % to 1.0 %, which is 3.4 %.

The average thrust force generated followed the same trend as the cutting force, and the thrust force decreased with the increase in Sn addition [Table 5.17, Fig. 5.41 (b)]. The average thrust force for the 319.2 base alloy was 112.7 N, and 81.5 N for the alloy with 0.5 % Sn, a decrease of 27.7 %, and 77.1 N for the alloy with 1.0 % Sn. The decrease in the average thrust force was less significant with further increase in the Sn amount from 0.5 % to 1.0 %, which is 5.4 %.

The COF and specific cutting energy are listed in Table 5.18, which demonstrates that the COF maintained the same value of 0.21 for all the three alloys. The specific cutting energy is 849.5 MJm⁻³ for the 319.2 base alloy and 592.1 MJm⁻³ for the 319.2 alloy with 1.0 % Sn both machined at the turning speed of 0.42 m/s.

Table 5.17 Average cutting and thrust forces of Al 319.2 modified by Sn cast at 26 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Cutting force</th>
<th>Thrust force</th>
<th>Speed (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Cutting force</td>
<td>Change</td>
<td>Thrust force</td>
</tr>
<tr>
<td>319.2</td>
<td>524.4±26.2</td>
<td></td>
<td>112.7±5.6</td>
</tr>
<tr>
<td>0.5Sn</td>
<td>380.0±19.0</td>
<td>-27.5</td>
<td>81.5±4.1</td>
</tr>
<tr>
<td>1.0Sn</td>
<td>367.0±18.4</td>
<td>-30.0</td>
<td>77.1±3.9</td>
</tr>
<tr>
<td>0.5Bi-0.25Sn</td>
<td>346.1±17.3</td>
<td>-34.0</td>
<td>70.8±3.5</td>
</tr>
<tr>
<td>319.2</td>
<td>502.7±25.1</td>
<td></td>
<td>121.0±6.1</td>
</tr>
<tr>
<td>0.5Sn</td>
<td>362.0±18.1</td>
<td>-28.0</td>
<td>83.0±4.2</td>
</tr>
<tr>
<td>1.0Sn</td>
<td>353.0±17.7</td>
<td>-29.8</td>
<td>74.0±3.7</td>
</tr>
<tr>
<td>0.5Bi-0.25Sn</td>
<td>341.0±17.1</td>
<td>-32.2</td>
<td>70.7±17.1</td>
</tr>
</tbody>
</table>
### Table 5.18 COF and Specific cutting energy of Al 319.2 modified by Sn cast at 26 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>COF</th>
<th>Specific cutting energy</th>
<th>Speed (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>COF</td>
<td>Change</td>
<td>Specific cutting energy</td>
</tr>
<tr>
<td>319.2</td>
<td>0.21</td>
<td></td>
<td>849.5</td>
</tr>
<tr>
<td>0.5Sn</td>
<td>0.21</td>
<td>0.0</td>
<td>615.3</td>
</tr>
<tr>
<td>1.0Sn</td>
<td>0.21</td>
<td>0.0</td>
<td>592.1</td>
</tr>
<tr>
<td>0.5Bi-0.25Sn</td>
<td>0.20</td>
<td>-4.8</td>
<td>555.9</td>
</tr>
<tr>
<td>319.2</td>
<td>0.24</td>
<td></td>
<td>831.6</td>
</tr>
<tr>
<td>0.5Sn</td>
<td>0.23</td>
<td>-4.2</td>
<td>593.3</td>
</tr>
<tr>
<td>1.0Sn</td>
<td>0.21</td>
<td>-12.5</td>
<td>569.3</td>
</tr>
<tr>
<td>0.5Bi-0.25Sn</td>
<td>0.21</td>
<td>-12.5</td>
<td>549.2</td>
</tr>
</tbody>
</table>

(2) Turning with a speed of 0.9 m/s

The average cutting force decreases from 502.7 N for the 319.2 base alloy to 362.0 N for the alloy with 0.5 % Sn [Table 5.17 and Fig. 5.41 (a)], a decrease of 30.0 %. The average cutting force is 353.0 N for the alloy with 1.0 % Sn. The decrease in cutting force was less significant with further increase in the Sn amount, from 0.5 % Sn to 1.0 % Sn, a decrease of 2.5 %.

The average thrust forces generated during cutting follow the same trend as the cutting force, and the thrust force decreases with the increase of Sn addition [Table 5.17 and Fig. 5.41 (b)]. The average thrust force for the Al-Si base alloy is 95 N, which decreased to 77 N for the alloy with 0.5 % Sn, and to 72 N for the alloy with 1.0 % Sn.

**Table 5.18** demonstrates that the COF decreased from 0.24 for the 319.2 alloy, to 0.23 for the alloy with 0.5 % Sn and to 0.21 for the alloy with 1.0 % Sn. The specific cutting energy was 831.6 MJm⁻³ for the 319.2 base alloy, 593.3 MJm⁻³ for the 319.2 alloy with 0.5 % Sn, and 569.3 MJm⁻³ for the the alloy with 1.0 % Sn.
5.3.2.2 Interfacial temperature

Temperature at the interface between the chips and the rake face of the WC cutting tool varied with the amount of Bi or Sn addition during dry turning for the 319 alloy cast with a cooling rate of 26 °C/s.

The temperature-turning time curves recorded during the dry turning of the 319.2 alloys with Bi and Sn cast with a cooling rate of 26 °C/s are presented in Figs. 5.42 and 5.43, respectively. The interfacial temperature increased rapidly at the beginning, and then reached an equilibrium value. The 319.2 base alloys turned at 0.42 m/s had the highest temperature; the temperature decreased with 0.5 % Bi addition, and decreased further with 1.0 % Bi addition. The temperature was lowest for the alloy with 1.0 % Sn addition.

5.3.2.2.1 The 319.2 alloy modified with Bi

With the use of the WC cutting tool, the temperature for the 319.2 base alloy that was cast with a cooling rate of 26 °C/s, as shown in Fig. 5.44 and listed in Table 5.19 was 258.1 °C, 220.8°C for the alloy with 0.5 % Bi, and 177.9 °C for the alloy with 1.0 % Bi all machined with a turning speed of 0.42 m/s. However, the temperature was 271.0 °C for the base alloy, 237.4 °C for the alloy with 0.5 % Bi, and 207.4 °C for the alloy with 1.0 % Bi machine with a turning speed of 0.9 m/s.
Table 5.19 Interfacial temperature between chips and tool rake face for Al 319.2 modified by Bi cast at 26 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Temperature</th>
<th>Speed (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Temperature (°C)</td>
<td>Change (%)</td>
</tr>
<tr>
<td>319.2</td>
<td>258.1±12.9</td>
<td></td>
</tr>
<tr>
<td>0.5Bi</td>
<td>220.8±11.0</td>
<td>-14.5</td>
</tr>
<tr>
<td>1.0Bi</td>
<td>177.9±8.9</td>
<td>-31.1</td>
</tr>
<tr>
<td>319.2</td>
<td>271.0±14.3</td>
<td></td>
</tr>
<tr>
<td>0.5Bi</td>
<td>237.4±11.9</td>
<td>-17.0</td>
</tr>
<tr>
<td>1.0Bi</td>
<td>207.4±10.4</td>
<td>-27.5</td>
</tr>
</tbody>
</table>

5.3.2.2.2 Alloy 319.2 modified with Sn
At a turning speed of 0.42 m/s, the interfacial temperature developed in the chips of 319.2 alloy with different amounts of Sn addition and cast with a cooling rate of 26 °C/s is shown in Fig. 5.45 and listed in Table 5.20; 258.1 °C for the 319.2 base alloy, 159.1 °C for the alloy with 0.5 % Sn, and 121.6 °C for the alloy with 1.0 % Sn. When the turning speed was 0.9 m/s, the temperature was 271.0 °C for the base alloy, 211.0 °C for the alloy with 0.5 % Sn, and 155.0 °C for the alloy with 1.0 % Sn.

Table 5.20 Interfacial temperature between chips and the rake face for Al 319.2 modified by Sn cast at 26 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Temperature</th>
<th>Speed (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Temperature (°C)</td>
<td>Change (%)</td>
</tr>
<tr>
<td>319.2</td>
<td>258.1±12.9</td>
<td></td>
</tr>
<tr>
<td>0.5Sn</td>
<td>159.1±8.0</td>
<td>-38.4</td>
</tr>
<tr>
<td>1.0Sn</td>
<td>121.6±6.1</td>
<td>-52.9</td>
</tr>
<tr>
<td>0.5Bi-0.25Sn</td>
<td>142.2±7.1</td>
<td>-44.9</td>
</tr>
<tr>
<td>319.2</td>
<td>271.0±14.3</td>
<td></td>
</tr>
<tr>
<td>0.5Sn</td>
<td>211.0±10.6</td>
<td>-26.2</td>
</tr>
<tr>
<td>1.0Sn</td>
<td>155.0±7.8</td>
<td>-45.8</td>
</tr>
<tr>
<td>0.5Bi-0.25Sn</td>
<td>191.1±9.6</td>
<td>-33.2</td>
</tr>
</tbody>
</table>
5.3.2.3 Surface roughness

The average surface roughness of the 319.2 alloys with different amounts of Bi addition are listed in Table 5.21. The average surface roughness did not vary significantly and was in the range of 0.819 µm to 1.012 µm for the turning speed of 0.42 m/s, and in the range of 0.826 µm to 1.104 µm for the turning speed of 0.9 m/s.

Table 5.21 Average surface roughness of Al 319.2 modified by Bi and cast at 26 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Surface Roughness (µm)</th>
<th>Change (%)</th>
<th>Speed (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td>319.2</td>
<td>1.012±0.289</td>
<td></td>
<td>0.42</td>
</tr>
<tr>
<td>0.5Bi</td>
<td>0.821±0.241</td>
<td>-18.9</td>
<td></td>
</tr>
<tr>
<td>1.0Bi</td>
<td>0.819±0.243</td>
<td>-19.1</td>
<td></td>
</tr>
<tr>
<td>319.2</td>
<td>1.104±0.301</td>
<td></td>
<td>0.9</td>
</tr>
<tr>
<td>0.5Bi</td>
<td>0.993±0.238</td>
<td>-10.1</td>
<td></td>
</tr>
<tr>
<td>1.0Bi</td>
<td>0.826±0.229</td>
<td>-25.2</td>
<td></td>
</tr>
</tbody>
</table>

The average surface roughness of the 319.2 alloy with different amounts of Sn addition are listed in Table 5.22. The average surface roughness was in the range of 0.899 µm to 1.012 µm for the turning speed of 0.42 m/s, and in the range of 0.785 µm to 1.104 µm for the turning speed of 0.9 m/s.

Table 5.22 Average surface roughness of Al 319.2 modified by Sn and cast at 26 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Surface Roughness (µm)</th>
<th>Change (%)</th>
<th>Speed (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td>319.2</td>
<td>1.012±0.289</td>
<td></td>
<td>0.42</td>
</tr>
<tr>
<td>0.5Sn</td>
<td>0.921±0.240</td>
<td>-9.0</td>
<td></td>
</tr>
<tr>
<td>1.0Sn</td>
<td>0.899±0.240</td>
<td>-11.2</td>
<td></td>
</tr>
<tr>
<td>0.5Bi0.25Sn</td>
<td>0.902±0.238</td>
<td>-10.9</td>
<td></td>
</tr>
<tr>
<td>319.2</td>
<td>1.104±0.301</td>
<td></td>
<td>0.9</td>
</tr>
<tr>
<td>0.5Sn</td>
<td>0.887±0.233</td>
<td>-19.7</td>
<td></td>
</tr>
<tr>
<td>1.0Sn</td>
<td>0.785±0.235</td>
<td>-28.9</td>
<td></td>
</tr>
<tr>
<td>0.5Bi0.25Sn</td>
<td>0.833±0.236</td>
<td>-24.5</td>
<td></td>
</tr>
</tbody>
</table>
5.3.2.4 Cross-sectional microstructure at the chip root

The microstructure of chip root cross-section shows the shear deformation during dry turning. The position of chip-root cross section is indicated in the center of the chip in Fig. 5.28. The chip-root itself was sampled by interrupting the cutting process.

5.3.2.4.1 Alloy 319.2 modified with Bi

Figure 5.46 is the cross-sectional micrograph of the chip root taken by interrupting the cutting process of the 319 alloy with 1.0 % Bi turned at 0.42 m/s. Optical microscopy [Fig. 5.46 (a)] shows that the segments are separated by cracks. The EDS maps of the tool tip area [Fig. 5.46 (b)] show that Bi particles are elongated into stingers, and situated in the cracks.

5.3.2.4.2 Alloy 319.2 modified with Sn

The cross-sectional optical micrographs of the chips formed from the 319.2 base alloy [Fig. 5.47 (a)], and the alloy with 1.0 % Sn cut at 0.42 m/s [Fig. 5.47(b)] are presented. It shows that the segments are separated. The area near the tool tip was observed by SEM –BEI [Fig. 5.47(c)] and shows elongated Sn stringers resulting from the large shear plastic deformation.

In summary, these alloys generated discontinuous type of chips with short chip segments loosely attached to each other. The cross-section of the chip for the 319.2 base alloy exhibited cracks, but their number was less as compared to the 1.0 % Bi or the Sn alloy, and the chip was less segmented.

5.3.2.5 Chip morphology

Tables 5.23, 5.24 and Fig. 5.48 demonstrated the length of the chips for the 319.2 base alloy modified with different concentrations of Bi and Sn and cut at 0.42 m/s and 0.9 m/s using a diamond tool.
The length of the chips for the 319 base alloy (Fig. 5.49(a)) cut at 0.42 m/s was longer than that of the 319 base alloy with 1.0 % Sn (Fig. 5.49(b)). The length of the chips for the 319 base alloy (Fig. 5.49(c)) cut at 0.9 m/s was longer than that of the 319 base alloy cut at 0.42 m/s (Fig. 5.49(a)). The length of chips for the 319.2 alloy with 1.0 % Sn cut at 0.9 m/s (Fig. 5.49(d)) was also longer than that cut at 0.42 m/s (Fig. 5.49(b)). The reason is that the temperature was higher in the interface between the chips and the tool rake face.

**Table 5.23** Length of the chips for Al 319.2 modified with different amount of Bi cast at 26 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Length</th>
<th>Change (%)</th>
<th>Speed (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Length (mm)</td>
<td>Change (%)</td>
<td>0.42</td>
</tr>
<tr>
<td>319.2</td>
<td>70.0±3.5</td>
<td></td>
<td></td>
</tr>
<tr>
<td>0.5Bi</td>
<td>12.0±0.6</td>
<td>-82.9</td>
<td></td>
</tr>
<tr>
<td>1.0Bi</td>
<td>8.0±0.4</td>
<td>-88.6</td>
<td></td>
</tr>
<tr>
<td>319.2</td>
<td>400.0±20.0</td>
<td></td>
<td>0.9</td>
</tr>
<tr>
<td>0.5Bi</td>
<td>32.0±1.6</td>
<td>-92.0</td>
<td></td>
</tr>
<tr>
<td>1.0Bi</td>
<td>22.0±1.1</td>
<td>-94.5</td>
<td></td>
</tr>
</tbody>
</table>

**Table 5.24** Length of chips for Al 319.2 modified with different amount of Sn cast at 26 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Length</th>
<th>Change (%)</th>
<th>Speed (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Length (mm)</td>
<td>Change (%)</td>
<td>0.42</td>
</tr>
<tr>
<td>319.2</td>
<td>70±3.5</td>
<td></td>
<td></td>
</tr>
<tr>
<td>0.5Sn</td>
<td>12.0±0.6</td>
<td>-82.9</td>
<td></td>
</tr>
<tr>
<td>1.0Sn</td>
<td>9.0±0.5</td>
<td>-87.1</td>
<td></td>
</tr>
<tr>
<td>0.5Bi-0.25Sn</td>
<td>8.0±0.4</td>
<td>-88.6</td>
<td></td>
</tr>
<tr>
<td>319.2</td>
<td>400.0±20.0</td>
<td></td>
<td>0.9</td>
</tr>
<tr>
<td>0.5Sn</td>
<td>12.0±0.6</td>
<td>-97.0</td>
<td></td>
</tr>
<tr>
<td>1.0Sn</td>
<td>8.0±0.4</td>
<td>-98.0</td>
<td></td>
</tr>
<tr>
<td>0.5Bi-0.25Sn</td>
<td>9.0±0.5</td>
<td>-97.8</td>
<td></td>
</tr>
</tbody>
</table>
The length of the chips formed from the 319.2 alloy with 1.0 % Bi cut at 0.42 m/s and 0.9 m/s followed the same trend as that of the alloys with Sn addition. The chips formed from the 319 base alloy (Fig. 5.49(c)) cut at 0.9 m/s tended to be of the continuous type, while all others were of the discontinuous type.

Chip morphology is strongly influenced by the tool material. Fig. 5.50 is the chips morphology developed using the WC cutting tool for the alloys cast with a cooling rate of 26 °C/s. The chips are of the continuous type for the 319 base alloy cut both at 0.42 m/s and at 0.9 m/s (Fig. 5.50(a)), however, the chips are of the discontinuous type for alloys with Bi or Sn addition (Fig. 5.50(b)).

5.3.2.6 Summary

Average cutting and thrust forces of the 319.2 alloys with different amounts of Bi and Sn that were cast with a cooling rate of 26 °C/s, and the temperatures generated at the interface during cutting were measured at turning speeds of 0.42 m/s and 0.9 m/s. The corresponding COF and specific cutting energy were calculated. The average surface roughness, microstructure of chip-roots cross-section, and the morphology of chips were studied. In summary:

(1) Both cutting force and thrust force decreased with the addition of Bi and Sn. Sn is more effective in decreasing the forces than Bi.

(2) The specific cutting energy showed the same trend as the cutting force.

(3) The temperature generated at the interface between the chips and the rake face of the WC cutting tool decreased with the addition of Bi or Sn during dry turning. Sn was more effective in decreasing the interface temperature
than Bi. The elongated Bi and Sn stringers indicate that the Bi and Sn melted during turning.

(4) The COF also decreased with the addition of Bi and Sn.

(5) The length of chips for the 319 base alloys modified with different amounts of Bi or Sn addition decreased. All chips were of the discontinuous type.

(6) The elongated Bi or Sn stringers resulting from the large shear plastic deformation promoted the segmentation of the chips.

5.4. Discussion and Conclusions

5.4.1 Discussion

The effect of Bi and Sn additions on the microstructure, mechanical properties and dry turning performance of the 319.2 Al-Si alloys was examined. The dry turning performance was improved with modification, which was evaluated through cutting and thrust forces, the temperature generated between the chips and the tool rake face, microstructure of chip-root cross-section, and chip morphology. These improvements were directly determined by the microstructure and mechanical properties of the 319.2 Al-Si alloys with the addition of Bi and Sn, and the optimum amount of Bi or Sn was found to be 0.5 % cast with a high cooling rate.

(1) Effect of Bi on the microstructure and mechanical properties of 319.2 Al-Si alloys

Bi distribution in the 319.2 Al-Si alloys can only be observed with SEM-BEI (Fig. 5.1). The white particles in the SEM-BEI are pure Bi, and were formed mainly at the interface of the eutectic Si particles and some at the interfaces of the intermetallic particles. This is because Bi is insoluble in Si and Al, which cause it to associate with the eutectic Si/Al interface. Bi particles were present in the spherical form, because it did not
wet the Al grain boundary [87]. The Bi particle formation and segregation during solidification were similar to that discussed in Section 6.5. No Bi-Mg combined product was found in the alloys and the Mg concentration was 0.02 % (Table 4.1). The Bi phase itself had little effect on the mechanical properties of the 319.2 Al-Si alloys.

On the other hand, the refining effect of Sr on the eutectic silicon morphology decreased with the increase of Bi, because silicon morphology becomes coarser with the increase of Bi content in the 319.2 Al-Si alloys. The partly modified acicular silicon particles act as internal stress raisers in the microstructure and provide easy paths for fracture. According to the widely accepted impurity-induced twinning theory [73], Sr is absorbed onto the growing Si crystal surfaces and thus the crystal growth is restricted, which leads to the forced twinning of the Si crystal and results in enhanced branching and a fibrous microstructure; both the twinning frequency and the angle of branching increase. The modifying elements, such as Sr, must have an affinity to silicon, which will facilitate their absorption on the surface of silicon crystals growing in an aluminum melt (Fig. 2.23). However, with the segregated Bi appearing on the growing Si crystal surfaces, and occupying the position of Sr, the role of Sr decreases, and the silicon morphology becomes coarser. The mechanical properties of the Al-Si alloys are strongly influenced by the microstructural differences during solidification. The mechanical properties of the modified 319.2 Al-Si alloys decreased.

The microstructural evolution shows that very fast solidification conditions can facilitate the development of more refined, uniformly dispersed eutectic Si particles. In this case, short growth time for Si particles at a high cooling rate, a non-equilibrium process far away from the equilibrium solidification, determines the refinement of the
microstructures. The quantitative results indicate that the cooling rate plays a significant role in the refinement of the microstructure (Table 4.3, Table 4.5), because the average size of Si decreased. The tensile Properties of the 319 alloy with 0.5 % Bi cast with a cooling rate of 26 °C/s maintained the properties of the base alloy (Table 5.3).

(2) Effect of Bi on the dry turning performance of 319.2 Al-Si alloys

For 319.2 alloys modified with 0.5 % Bi, compared to the base alloy, cast with a cooling rate of 9.4 °C/s and turned at 0.42 m/s, the cutting and thrust forces decreased by 13.2 % and 24.2 %, the COF decreased by 13.6 %, the specific cutting energy decreased by 15.2 %, the average temperature at the interface between the chips and the rake face decreased from 249.7 °C to 164.5 °C, and the surface roughness decreased from 1.033 μm to 0.833 μm. The average length of chip segments decreased from 21 mm to 11 mm. Chips were of the discontinuous type. The variations in the average cutting and thrust forces, and the morphology of chips mirrored the role of the Bi addition; Bi decreased the cutting force and the tool rake temperature, promoted a discontinuous chip formation, and decreased the friction between the tool and the chips.

The temperature distribution in orthogonal cutting is summarized in Section 2.1.5. Temperate increment was calculated using with the equivalent plastic strain and equivalent flow stress $\sigma$, and the temperate distribution map showing the local increment in temperature in the region ahead of the tool tip for an orthogonally cut Al 1100 sample is shown in Fig.2.15 [29]. The maximum local increment was 244 °C, considering the room temperature to have been 25 °C. The theoretically calculated actual local temperate near the tool tip was 269 °C, which corresponds with the experimental results of the chip bulk temperature of 249.7 °C. From this point view, Bi particles melted during orthogonal
cutting. Figure 5.28 shows the cross-sectional optical micrographs of the chips for the alloys with 0.5% Bi for the turning speed of 0.42 m/s. The segments were almost separated. It was also observed from the SEM–BEI image shown in Fig. 5.28 (b) that the elongated Bi stringers in the area near the tool tip that the Bi particles melted, due to the large shear plastic deformation. Molten Bi acts as a lubricant between the chips and the tool rake face, and also makes the chips brittle, promoting segmentation of the chips.

(3) Effect of Sn on the microstructure and mechanical properties of 319.2 Al-Si alloys

The Sn distribution in the 319.2 Al-Si alloys could only be observed with SEM-BEI (Fig. 5.2). The white Sn in the SEM-BEI was formed mainly at the interface of eutectic Si particles, because Sn is insoluble in Si and Al, which causes it to form at the eutectic Si/Al interface. Sn is in the form of a thin film because it wets the Al grain boundary [87], which is different from the behavior of Bi. No Sn-Mg combined product was found in alloys and the Mg concentration was 0.02% (Table 4.1). Also, the silicon morphology showed no change with the increase of Sn content in the 319.2 Al-Si alloys. Sn also has no effect on the refining role of Sr on the eutectic silicon morphology. Tensile properties of the 319.2 alloys with Sn should have decreased. However, the Sn phase affected the mechanical properties of 319.2 Al-Si alloys, by decreasing the bonding of the Si interface.

(4) Effect of Sn on the dry turning performance of 319.2 Al-Si alloys

Compared with Bi, conclusions can be drawn as follows:

(1) Sn had a lower melting point temperature. Sn should be in the molten state in the shear band under the same conditions as Bi during orthogonal cutting.
(2) The 319 Al alloys modified with Sn had lower cutting forces and lower tool rake face temperature. Sn was in film morphology in the 319 Al alloys, because Sn wetted the grain boundary of Al. The direct result is that the 319 Al alloys become brittle after the melting of Sn.

(3) Sn is more effective in the improvement of the dry turning performance, but with a decrease in the mechanical properties.

(5) Mechanism of discontinuous chip formation

The curves of the cutting and thrust forces varying with turning time show that the cutting and thrust forces reach the maximum value, then quickly decrease because of the chip breaking (Fig. 5.51). The time duration from one maximum value to the next maximum value is 0.008 s for the 319 Al with 1.0 % Sn, however, the time duration increased to 0.012 s for the 319.2 base alloy. The shortening of the time duration of the cutting and thrust forces with turning time indicated that the chips fractured (or segmentation) more frequently in 319 Al with 1.0 % Sn.

During one cycle of increase and decrease in the cutting and thrust forces, one segment was fully developed as shown in the center segment in Fig. 5.52 (a); the shear angle in the last stage was $\Phi_2 = 26.5 ^\circ$. In the meantime, a new segment started to form with a shear angle of $\Phi_1 = 37 \sim 53 ^\circ$. In the middle stage of development of a segment, the shear angle was in the range of $\Phi = 26 \sim 37 ^\circ$ Fig. 5.52 (b). So the shear angle also varied with turning time as the cutting and thrust forces.

The mechanism of discontinuous chip formation

(1) Discontinuous chip formation is a shear fracture near the free surface.
(2) Plastic deformation resulting in the bulge, microcrack and gross cracking of the chips is triggered by the extrusion of the tool rake face.

(3) Cutting and thrust forces reach the maximum value, and then drop because of the chip breakage.

(4) Shear angle starts with a large value, then drops when the crack appears between the segments.

(5) Bi melts in the shear band and lubricates the tool rake face during turning, and promotes the crack formation.

(6) Temperature is the key factor, and low temperatures are with an improvement in the dry turning performance.

5.4.2 Conclusions

For the 319.2 base alloy with different Bi and Sn concentrations cast with different cooling rates and machined with different turning speeds, the dry machining characteristics can be summarized as follows:

(1) With the addition of Bi, the size of the Si particles only marginally changed at high cooling rates, but dramatically at low cooling rates. Bi counteracted the role of Sr modification.

(2) The alloy with 0.5% Bi cast with a high cooling rate maintained the mechanical properties of base alloy.

(3) Bi was in very fine and spherical shaped particles on the fracture surface, however, Sn appeared on the surface of the Si phase and deteriorated the mechanical properties.

(4) The turning forces decreased with the addition of Bi and Sn.
(5) The cutting temperature was determined by Bi or Sn concentration, their distribution, and the cutting speed.

(6) Chips became shorter with the addition of Bi and Sn. All chips were discontinuous, except for the 319.2 base alloy using a WC tool, which was of the continuous type.

(7) Chip segments slid over the neighboring segments easily because Bi and/or Sn appeared within the shear bands, causing shear localization.

(8) Discontinuous chip formation is a shear fracture process.
Figures

(a)

(b)
Fig. 5.1 SEM-BEI images of Al 319.2 modified by Bi solidified at the cooling rate of 9.4 °C/s (a) with 0.5 % Bi, (b) with 1.0 % Bi, (c) with 1.0 % Bi at high magnification.
Fig. 5.2 SEM-BEI images of Al 319.2 modified by Sn solidified at the cooling rate of 9.4 °C/s (a) with 0.5 % Sn, (b) with 1.0 % Sn, (c) with 0.5 % Bi+0.25 % Sn.
Fig. 5.3 EDS spectrum of Bi or Bi+Sn particles
(a) Bi particles shown in Fig. 5.1 (a), and (b) Bi +Sn particles shown in Fig. 5.2 (c)
Fig. 5.4 Tensile stress-strain curves of Al 319.2 modified by Bi cast at the cooling rate of 9.4 \(^\circ\)C/s.
Fig. 5.5 SEM images of the tensile fracture surfaces of Al 319.2 modified by Bi cast at 9.4 °C/ 
(a) with 0.0 % Bi, (b) with 0.5 % Bi, (c) with 1.0 % Bi.
Fig. 5.6 Cross-sections near tensile fracture surface of Al 319.2 modified by Bi cast at 9.4 °C/s (a) with 0.0 % Bi, (b) with 1.0 % Bi.
Fig. 5.7 Tensile stress-strain curves of Al 319.2 modified by Sn or Bi+Sn cast at 9.4 °C/s.
Fig. 5.8 SEM images of the tensile fracture surface of Al 319.2 modified by Sn cast at 9.4 °C/s (a) with 0.5 % Sn, (b) with 1.0 % Sn.

Fig. 5.9 EDS spectrum of Sn particle shown in Fig. 5.8 (b)
Fig. 5.10 Cross-section of tensile fracture surface of Al 319.2 with 1.0 % Sn cast at 9.4 °C/s
Fig. 5.11 SEM-BEI images of Al 319.2 modified by Bi cast at the cooling rate of 26°C/s
(a) with 0.5 % Bi, (b) with 1.0 % Bi.
Fig. 5.12 SEM-BEI images of Al 319.2 modified by Sn cast at the cooling rate of 26°C/s (a) with 0.5 % Sn, (b) with 1.0 % Sn.

Fig. 5.13 Tensile stress-strain curves of Al 319.2 modified by Bi at the cooling rate of 26 °C/s
Fig. 5.14 SEM images of the tensile fracture surface of alloys cast at the cooling rate of 26 °C/s
(a) 0 % Bi, (b) with 0.5 % Bi, and (c) with 1.0 % Bi.
Fig. 5.15 Cross-sections of near tensile fracture surfaces of alloys cast at the cooling rate of 26 °C/s  
(a) 319, (b) with 0.5 % Bi, (c) with 1.0 % Bi.
Fig. 5.16 Tensile stress-strain curves of Al 319.2 modified by Sn cast at the cooling rate of 26 °C/s.
Fig. 5.17 Tensile fracture surfaces of Al 319.2 modified by Sn
(a) with 0.5 % Sn, (b) with 1.0 % Sn.
Fig. 5.18 Cross-sections near tensile fracture surfaces of alloys cast at the cooling rate of 26 °C/s (a) with 0.5 % Sn, and (b) with 1.0 % Sn.

Fig. 5.19 Typical cutting and thrust forces - time curves for 319.2 alloy cast at the cooling rate of 9.4°C
Dry turning at a cutting speed of 0.9 m/s and feed rate of 0.25mm/rev.
Fig. 5.20 Cutting and thrust forces varied with the amount of Bi addition cast at 9.4 °C/s
(a) Cutting force, (b) Thrust force.

(a) Cutting force
(b) Thrust force
Fig. 5.21 Cutting and thrust forces varied with the amount of Sn addition cast at 9.4 °C/s (a) Cutting force, (b) Thrust force.

Fig. 5.22 Turning temperature - time curves at 0.42 m/s and 0.9 m/s for alloys cast at 9.4 °C/s (denoted by numbers as: 1: with 0.5 % Bi at 0.9 m/s, 2: with 0.5 % Bi at 0.42 m/s, 3: with 1.0 % Bi 0.9 m/s)
Fig. 5.23 Turning temperature - time curves for alloys modified by Sn at the cooling rate of 9.4 °C/s
(denoted by numbers as follows: 1: 319 Al at 0.42 m/s, 2: with 0.5 % Bi+0.25% Sn at 0.9 m/s, 3: with 0.5 % Bi+0.25% Sn at 0.42 m/s).

Fig. 5.24 Turning temperature variation with the concentration of Bi for alloys cast at 9.4 °C/s.
Fig. 5.25 Turning temperature variation with the concentration of Sn for alloys cast at 9.4 °C/s

![Graph showing temperature variation with Sn concentration and cooling rate.](image)

(a) 0Sn, (b) 0.5Sn, (c) 1.0Sn, (d) 0.5Bi-0.25Sn

Fig. 5.26 The finished surface of the Al 319.2 modified by Bi cast at the cooling rate of 9.4 °C/s,
(a) Al 319.2, (b) Al 319.2 with 1.0 % Bi.
Fig. 5.27 The position of chip-root cross section is indicated in the center of chip
Fig. 5.28 The cross-sectional micrographs of chips at 0.42 m/s for alloys with 0.5 % Bi cast at 9.4 °C/s
(a) optical micrograph, and (b) SEM –BEI image shows elongated Bi stringers.
Fig. 5.29 The cross-sectional micrographs of chip root of Al 319.2 with 1.0 % Bi cut at 0.9 m/s
(a) Optical micrograph, (b) The EDS maps of the insert in (a).
Fig. 5.30 The average length of chips segments of the alloys cooling at 9.4 °C/s (a) with Bi addition (b) with Sn addition.
Fig. 5.31 Chip side view for Al 319.2 with 0.5 % Bi cooling at 9.4 °C/s and cut at 0.42 m/s

Fig. 5.32 Free surface view of chips of Al 319.2 with 0.5 % Bi cast at 9.4 °C/s and cut at 0.42 m/s
**Fig. 5.33** Chips side view for alloys cooling at the rate of 9.4 °C/s and cut at 0.42 m/s (a) Al 319.2, (b) with 0.5 % Sn. (c) with 1.0 % Sn.

**Fig 5.34** Free surface of chip for Figs. 5.33 (b)
Fig. 5.35 The cross-sectional optical micrographs of chips for alloys with 1.0 % Sn cut at 0.9 m/s (a) Optical micrograph, and (b) SEM–BEI image shown elongated Sn stringers.
Fig. 5.36 Chips morphology of alloys at the cooling rate of 9.4 °C/s using diamond tool, (a) 319.2 at 0.42 m/s, (b) with 1.0 % Bi at 0.42 m/s, (c) 319.2 at 0.9 m/s, (d) with 1.0 % Bi at 0.9 m/s.

Fig. 5.37 Chips morphology of alloys cast at the cooling rate of 9.4 °C/s using diamond tool (a) with 1.0 % Sn at 0.42 m/s, (b) with 1.0 % Sn at 0.9 m/s.
**Fig. 5.38** Chips morphology of alloys cast at the cooling rate of 9.4 °C/s using diamond tool.
**Fig. 5.39** Chips morphology using WC cutting tool at the cooling rate of 9.4 °C/s
(a) Al 319.2 at 0.42 m/s, (b) Al 319.2 at 0.9 m/s, (c) with 1.0 % Bi at 0.9 m/s, (d) with 1.0 % Sn at 0.9 m/s.

**Fig. 5.40** Cutting and thrust forces varied with the amount of Bi addition at the cooling rate of 26 °C/s
(a) Cutting force, (b) Thrust force.
Fig. 5.41 Cutting and thrust forces varied with the amount of Sn addition at the cooling rate of 26 °C/s, (a) Cutting force, (b) Thrust force.
Fig. 5.42 Turning temperature - time curves at 0.42 m/s and 0.9 m/s for alloys cast at 26 °C/s. (denoted by numbers as follows: 1: 319.2 Al at 0.42 m/s, 2: with 1.0 % Bi at 0.42 m/s, 3: with 1.0 % Bi at 0.9 m/s).

Fig. 5.43 Turning temperature - time curves at 0.42 m/s and 0.9 m/s for alloys cast at 26 °C/s. (denoted by numbers as follows: 1: 319.2 Al at 0.42 m/s, 2: with 0.5 % Sn at 0.42 m/s, 3: with 1.0 % Sn at 0.42 m/s).
Fig. 5.44 Turning temperature variation with the concentration of Bi cast at the cooling rate of 26 °C/s.

Fig. 5.45 Turning temperature variation with the concentration of Sn cast at the cooling rate of 26 °C/s.
**Fig. 5.46** The cross-sectional micrographs of chip root of the Al 319.2 with 1.0 % Bi at 0.42 m/s
(a) Optical micrograph, (b) The EDS maps of the insert in (a).
Fig. 5.47 The cross-sectional optical micrographs of chips (a) Al 319.2 alloy and (b) with 1.0 % Sn at 0.42 m/s, and (c) SEM–BEI image shown elongated Sn stringers.
Fig. 5.48 Chips length of different alloys cast at the cooling rate of 26 °C/s using diamond tool.

Fig. 5.49 Chips morphology using diamond tool of alloys cast at the cooling rate of 26 °C/s  
(a) Al 319.2 alloy at 0.42 m/s, (b) with 1.0 % Sn at 0.42 m/s, (c) Al 319.2 alloy at 0.9 m/s, and (d) 
with 1.0 % Sn at 0.9 m/s.
Fig. 5.50 Chips morphology using WC tool at the cooling rate of 26 °C/s (a) Al 319.2 alloy at 0.42 m/s, and (b) with 1.0 % Sn at 0.42 m/s.
Fig. 5.51 Cutting forces varied with turning time
(a) 319.2 Al with 1.0 % Sn, (b) 319.2 alloy.
Fig. 5.52 Chips cross section shown shear deformation.
CHAPTER 6  Mechanical Properties and Dry Turning Performances of 390 Al Modified by Bi

6.1. Introduction

In this chapter, the mechanical properties and dry turning performance of a hypereutectic Al-Si alloy 390 Al modified with Bi were evaluated using tensile tests, and by the measurement of the cutting forces, thrust forces and interfacial temperature during dry turning; the interfacial temperature being important because the melting of Bi plays an important role. The distribution and size of the Bi particles varied in the alloys solidified at different cooling rates of 9.4 °C/s and 26 °C/s. The effects of Bi modification on the mechanical properties and dry turning performance of 390 Al are discussed.

It is well known that the mechanical properties of the Al-Si alloys are strongly influenced by the microstructural differences arising from the different cooling rates during solidification. The effects of different cooling rates on the properties of the alloys (measured after casting with two different cooling rates of 9.4 °C/s and 26 °C/s) and their machining performance are discussed in Sections 6.2 and 6.3. Sec. 6.2 examines the microstructure and tensile properties of the hypereutectic Al-Si alloys with different amount of Bi addition and cast with two cooling rates of 9.4 °C/s and 26 °C/s. The discussion includes the role of the distribution and size of the Bi particles on the tensile strength, ductility and characteristics of the tensile fracture surfaces. Dry machining performance of the alloys with different amounts of Bi concentration generated under these cooling rates (9.4 °C/s and 26 °C/s) and machine at two different speeds of 0.42 m/s and 0.90 m/s are investigated in Sec. 6.3 by considering (i) cutting and thrust forces, COF
and specific cutting energy, (ii) interfacial temperature, (iii) surface roughness, (iv) cross-sectional microstructure at the chip-root, and (v) chip morphology. TEM observations were performed on the chip cross-section in order to investigate the role of Bi on the fracture of the Si particles during turning in Section 6.4. The role of Bi and the effect of the cooling rate on both tensile and machining performance of the Al-Si alloys are analyzed together and listed in Sec. 6.5.

6.2. Microstructure and Tensile Properties

6.2.1 Alloys Cast with a Cooling Rate of 9.4 °C/s

6.2.1.1 Distribution and size of the Bi particles

The SEM-BEI micrographs of the Al 390 alloy modified with 1.0 % Bi solidified at a cooling rate of 9.4 °C/s are shown in Fig. 6.1. The particles that appear bright are pure Bi. They are present in the form of spherical particles. It was observed that the Bi particles were mainly formed at the interface of the Si particles, and occasionally at the interfaces of the intermetallic particles. Bi could be rejected from the melt and to the Al/Si interface for the following reasons: (1) Bi and Al are immiscible elements based on the Bi-Al binary diagram shown in Fig. 2.31, (2) Bi is a surface-active element that decreases the surface tension of the molten alloy during solidification, (3) the Al/Si interface is an ideal location for the nucleation of Bi particles for a low nucleate energy. The distributions of Bi at the eutectic and primary Si interfaces with Al are displayed in Table 6.1. The sum of the percentage of Bi particles precipitated in the Al matrix, at the Si/Al and intermetallics/Al interfaces for each alloy was about 100%. More than half of the Bi particles precipitated at the eutectic Si/Al interface. For alloys with 1.0 % Bi and 0.5 % Bi, the percentage of the total Bi particles that precipitated at the eutectic Si/Al interface was 58 ± 7 % and 67 ± 6 %, respectively, and the percentage of Bi at the
interface of the primary Si phase was 18 ± 7 % and 8 ± 3 %, respectively. This is due to the fact that the total eutectic Si surface area was larger than that of the primary Si interface in Al matrix.

Table 6.2 lists the percentages of the primary and eutectic Si particles that had Bi particles precipitated at their interfaces. About 60 ± 15 % of the primary Si particles have Bi particles precipitated at their interfaces for the alloy with 1.0 % Bi addition. This ratio decreases to around 30 ± 11 % for the alloy with 0.5 % Bi.

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Primary Si (%)</th>
<th>Eutectic Si (%)</th>
<th>Intermetallics* (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.0 wt.% Bi</td>
<td>18±7</td>
<td>58±7</td>
<td>21±4</td>
</tr>
<tr>
<td>0.5 wt.% Bi</td>
<td>8±3</td>
<td>67±6</td>
<td>24±7</td>
</tr>
</tbody>
</table>

* Intermetallics include Al$_{15}$(Mn, Fe)$_3$Si$_2$, Al$_5$MnFeSi, Al$_2$Cu

Table 6.2 Percentages of each phase that had precipitated Bi particles at the interface when cast at 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Primary Si (%)</th>
<th>Eutectic Si (%)</th>
<th>Intermetallics (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.0 wt.% Bi</td>
<td>60±15</td>
<td>29±4</td>
<td>27±10</td>
</tr>
<tr>
<td>0.5 wt.% Bi</td>
<td>30±11</td>
<td>30±8</td>
<td>19±5</td>
</tr>
</tbody>
</table>

Figure 6.2 presents the increase in the Bi particle size and their variation with the different amounts of Bi concentration. The diameter of Bi particles was 8.7 ± 1.8 μm for 1.0 % Bi, and 6.3 ± 1.3 μm for 0.5 % Bi, cast with a cooling rate of 9.4 °C/s, indicating that the size of the Bi particles decreased along with the decrease in the amount of Bi concentration. The reasons are: (1) low Bi content in molten Al alloys, and (2) low diffusion rate of Bi in Al alloys.
6.2.1.2 Tensile properties

The tensile stress-strain curves of the Al 390 alloy with different amount of Bi concentration tested after casting with a cooling rate of 9.4 °C/s are shown in Fig. 6.3. The tensile strength decreased with an increase in the Bi addition from 0.5 % to 1.0 %.

Table 6.3 lists the tensile properties of the Al 390 alloys with different amounts of Bi additions that were solidified at a cooling rate of 9.4 °C/s. The tensile strength was found to be 247.2 ± 2.1 MPa for the base alloy (0 % Bi), 240.8 ± 2.3 MPa for the alloy with 0.5 % Bi, which is a 2.6 % decrease, and 168.0 ± 3.0 MPa for the alloy with 1.0 % Bi which is a 32.0 % decrease. The tensile strength difference was 6.4 MPa when the modification was increased in the range of 0.0-0.5 % Bi, compared to 72.8 MPa in the range of 0.5-1.0 % Bi. The change in the yield strength followed the same trend as that of the tensile strength.

Figs. 6.4 (a) and (b) show that the Bi particles appear on the tensile fracture surface of the Al 390 with both 0.5 and 1.0 % Bi, cast with a cooling rate of 9.4 °C/s. In these SEM-BEI images, the Bi particles are in an almost spherical shape and in the pure elemental state, as confirmed by the EDS spectrum as shown in Fig. 6.5.

Primary Si particles were often fractured. To further observe the broken primary Si particles, cross-sections of the tensile samples taken along the direction of the applied load of the tested samples were observed (Fig. 6.6). The fractured primary Si particles in the Al 390 alloy with 1.0 % Bi concentration were marked in Fig. 6.6 (a). The propensity for fracture of the primary Si was greater than that of the Fe-bearing intermetallic phase [Al15(Fe,Mn)3Si2]. The primary Si particles around the Fe-bearing phase were fractured, but the Fe-bearing phase itself was not. There were fewer fractured primary Si particles
in the Al 390 alloy with 0.5 % Bi as shown in Fig. 6.6 (b), and even fewer in the alloy without Bi addition as shown in Fig. 6.6 (c). Bi particles were observed to be in the vicinity of the intermetallics and the Si phases as shown in the EDS spectrum in Fig. 6.5 and the SEM image as shown in Fig. 6.6 (d); the fracture of the primary Si particles is related to Bi.

**Table 6.3** Tensile properties of Al 390 with different amounts of Bi cast at 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Yield strength</th>
<th>Tensile strength</th>
<th>Elongation</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Yield strength (MPa)</td>
<td>Change (%)</td>
<td>Tensile strength (MPa)</td>
</tr>
<tr>
<td>0 wt.%Bi</td>
<td>180.0±2.9</td>
<td>-</td>
<td>247.2±2.1</td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>176.0±3.8</td>
<td>-2.2</td>
<td>240.8±2.3</td>
</tr>
<tr>
<td>1.0 wt.%Bi</td>
<td>125.0±4.0</td>
<td>-30.6</td>
<td>168.0±3.0</td>
</tr>
</tbody>
</table>

6.2.1.3 Compressive properties

The compressive stress-strain curves of the Al 390 alloys with different Bi concentrations and cast with a cooling rate of 9.4 °C/s, compressed at 240 °C, are shown in Fig. 6.7. The base alloy and the alloy with 0.5 % Bi have almost the same yield strength, however, the yield strength of alloy with 1.0 % Bi decreased significantly. The cross-sectional microstructure after the compressive test at 240 °C shows that the cracks in the primary Si particles are parallel to the load direction (Fig. 6.8).

**6.2.2 Alloys Cast with a Cooling Rate of 26 °C/s**

6.2.2.1 Distribution and size of Bi particles

The Bi particles that were present in the form of spherical particles in the Al 390 with 1.0 % Bi when cast with a cooling rate of 26 °C/s, formed at the interface of the Si particles and occasionally at the interfaces of the intermetallic particles as shown in Fig. 6.9. The distribution of Bi at eutectic and primary Si interfaces with Al are listed in Table 6.4.
and the sum of the Bi fractions precipitated at the Si/Al and intermetallics/Al interface for each alloy was almost 100%. The percentage of total Bi precipitated at the eutectic Si/Al interface was 52 ± 3% and 57 ± 6% for the alloys with 1.0% Bi and 0.5% Bi, respectively, and the percentage of Bi at the interface of the primary Si phase is 18 ± 5% and 13 ± 6%, respectively.

Table 6.5 lists the percentages of the primary and eutectic Si particles that have Bi precipitates at their interfaces. About 66 ± 19% of the primary Si particles had Bi particles precipitated at their interfaces for the alloy with 1.0% Bi addition. The ratio decreased to around 31 ± 10% for the alloy with 0.5% Bi at the cooling rate of 26 °C/s.

Table 6.4 Percentages of the Bi particles precipitated at the Si/Al and intermetallics/Al interfaces when cast at 26.0 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Primary Si (%)</th>
<th>Eutectic Si (%)</th>
<th>Intermetallics (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.0 wt.%Bi</td>
<td>18±5</td>
<td>52±3</td>
<td>28±3</td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>13±6</td>
<td>57±6</td>
<td>28±8</td>
</tr>
</tbody>
</table>

Table 6.5 Percentages of each phase that had precipitated Bi particles at the interface when cast at 26 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Primary Si (%)</th>
<th>Eutectic Si (%)</th>
<th>Intermetallics (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.0 wt.%Bi</td>
<td>66±19</td>
<td>49±21</td>
<td>41±20</td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>31±10</td>
<td>31±6</td>
<td>32±7</td>
</tr>
</tbody>
</table>

The average diameter of the Bi particles were about 2.9 ± 0.4 μm in Al 390 with 1.0% Bi and 2.0 ± 0.3 μm in Al 390 with 0.5% Bi when cast with a cooling rate of 26 °C/s, much smaller than those of the alloys that were cast with a cooling rate of 9.4 °C/s, which were 8.6 ± 0.9 μm and 6.3 ± 0.8μm (Fig. 6.2).

Comparing the distribution of Bi in the Al 390 alloys with 0.5 and 1.0% Bi cast with cooling rates of 9.4 and 26 °C/s, most of the Bi precipitated around eutectic Si particles. Around 30% of the primary Si had Bi precipitates at their interfaces in the alloy with 0.5% Bi addition, however, more than 60% of the primary Si had Bi
precipitates at their interfaces in the alloy with 1.0 % Bi addition. Neither the distribution of total Bi at the interfaces of each phases (Tables 6.1 and 6.4) nor the percentage of each phase with Bi precipitation (Tables 6.2 and 6.5) was influenced by the cooling rate; they were mainly determined by Bi concentration.

6.2.2.2 Tensile properties

The stress-strain curves of the alloys with different Bi concentration cast with a cooling rate of 26 °C/s during solidification are shown in Fig. 6.10, and the tensile properties of the the Al 390 alloys with different amounts of Bi addition cast with a cooling rate of 26 °C/s are listed in Table 6.6.

The tensile strength of the base alloy that was cast with a cooling rate of 26 °C/s was 275.2 MPa with 0 % Bi, 266.3 MPa for the alloy with 0.5 % Bi, and 235.6 MPa for the alloy with 1.0 % Bi. The tensile strength difference was 9.3 MPa in the range of 0.0-0.5 % Bi, compared to 30.7 MPa in the range of 0.5-1.0 % Bi. The change in properties for the alloys cast at a cooling rate of 9.4°C/s were similar, and the decrease in yield strength was insignificant between the base alloy and the alloy with 0.5 % Bi.

Table 6.6 Tensile properties of Al 390 alloys with different amounts of Bi cast at 26 °C/s

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Yield strength</th>
<th>Tensile strength</th>
<th>Elongation</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Yield strength (MPa)</td>
<td>Change (%)</td>
<td>Tensile strength (MPa)</td>
</tr>
<tr>
<td>0 wt.%Bi</td>
<td>182.0±2.2</td>
<td>0.0</td>
<td>275.6±1.9</td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>179.0±3.5</td>
<td>-1.6</td>
<td>266.3±2.0</td>
</tr>
<tr>
<td>1.0 wt.%Bi</td>
<td>136.0±3.3</td>
<td>-25.3</td>
<td>235.6±2.3</td>
</tr>
</tbody>
</table>

A comparison of Fig. 6.3 with Fig. 6.10 of the tensile stress - strain curves reveals that the tensile properties in the Al 390 alloy cast with cooling rates of 9.4 °C/s and 26 °C/s are comparable, except that the tensile stresses of the alloys solidified at the
cooling rate of 26 °C/s exhibited a higher and narrower distribution with different Bi addition levels. Tables 6.3 and 6.6 also show that the higher cooling rate increased the tensile strength and hardness, and alleviated the negative effects of the Bi addition.

**Figures 6.11 (a) and (b)** show the Bi particles on the tensile fracture surface of Al 390 alloys cast with a cooling rate of 26 °C/s. The Bi particles are still in an almost spherical shape and in the pure elemental state. The size of Bi particles in the fracture surface is related to the cooling rate and the amount of Bi addition, which is consistent with the results shown in **Fig. 6.2**.

A comparison of **Fig. 6.11** with **Fig. 6.4** also reveals that there were fewer fractured primary Si particles in the Al 390 alloy cast with a cooling rate of 26 °C/s than those cast with a cooling rate of 9.4 °C/s, and there is no difference in the case of eutectic Si particles. Primary Si particles are mostly fractured. To further observe the broken primary Si phase, cross-sections of the tensile samples taken along the direction of the applied load of the tested samples were observed. The cross-sections of tensile samples with 1.0, 0.5, and 0 % Bi cast at a cooling rate of 26 °C/s are shown in **Fig. 6.12**. There are cracks in the primary Si particles in the alloy with 1.0 % Bi (**Fig. 6.12 (a)**). There are fewer fractured primary Si particles in the alloy with 0.5 % Bi and Al 390 alloy (**Fig. 6.12 (b), (c)**).

6.2.2.3 Compressive properties

The compressive stress-strain curves of the Al 390 alloys with different Bi concentrations, cast with a cooling rate of 26°C/s and compressed at 240 °C are shown in **Fig. 6.13**. It is evident that the base alloy and alloy with 0.5 % Bi have almost the same yield strength, however, the compressive strength of alloy with 1.0 % Bi is lower.
The cross-sectional microstructure after a compression test at 240 °C for alloy with 1.0 % Bi shows that there are less cracks in the primary Si particles, which are parallel to the loading direction (Fig. 6.14 (a)). Some cracks are associated with the Bi particles as indicated in Fig. 6.14 (a), which are confirmed by the SEM-BEI image (Fig. 6.14 (b)) and EDS spectrum (Figs. 6.14 (c)).

A comparison of Fig. 6.7 with Fig. 6.13 reveals that the alloys cast with the cooling rate of 26 °C/s have higher compressive strength compared to the alloys cast with the cooling rate of 9.4 °C/s. Si particles were also more refined when cast with a cooling rate of 26 °C/s, and less Si particles were broken during compression at 240 °C.

6.3. Dry Machining Performance at 0.42 m/s and 0.9 m/s

6.3.1 Al 390 alloys cast with a cooling rate of 9.4 °C/s

6.3.1.1 Cutting and thrust forces

Al 390 alloy samples with different Bi additions and solidification rates were subjected to dry orthogonal cutting tests, using the parameters listed in Section 3.5.

6.3.1.1.1 Turning with a speed of 0.42 m/s

The average cutting force measured during turning at 0.42 m/s was 490.6 N for the Al 390 base alloy, and 429.5 N for the alloy with 0.5 % Bi, both cast with a cooling rate of 9.4 °C/s, a decrease of 12.5 % (Table 6.7, Fig. 6.15 (a)). The average cutting force was 425.5 N for the alloy with 1.0 % Bi, a decrease of 13.3 %. The decrease in the cutting force was less significant with further increase in the Bi amount from 0.5 % to 1.0 %.

The average thrust force generated during dry turning at 0.42 m/s did not follow the same trend as that of the cutting force, and the thrust force decreased almost linearly with the increase in Bi addition (Table 6.7, Fig. 6.15 (b)). The average thrust force for
the Al 390 base alloy was 164.1 N, 135.7 N for the alloy with 0.5 % Bi, and 126.3 N for the alloy with 1.0 % Bi, all cast with a cooling rate of 9.4 °C/s, a decrease of 17.3% and 23.0% respectively.

6.3.1.1.2 Turning at a speed of 0.9 m/s

The average cutting force at 0.9 m/s decreased from 478.0 N for the Al 390 base alloy to 410.2 N for the alloy with 0.5 % Bi, a decrease of 14.2 % (Table 6.7, Fig. 6.15 (a)). The average cutting force was 403.8 N for the alloy with 1.0 % Bi. The decrease in the cutting force was less significant with a further increase in the Bi amount, from 0.5 % to 1.0 %, both cast with a cooling rate of 9.4 °C/s.

The average thrust force generated during cutting does not follow the same trend as the cutting force, and the thrust force decreased almost linearly with the increase in Bi addition (Table 6.7 and Fig. 6.15(b)). The average thrust force for the Al-Si base alloy is 153.6 N for the base alloy cast with a cooling rate of 9.4 °C/s, 131.7 N for the alloy with 0.5 % Bi, and 120.6 N for the alloy with 1.0 % Bi.

Table 6.7 Average cutting and thrust forces of Al 390 with different amounts of Bi cast at 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Speed (m/s)</th>
<th>Cutting force</th>
<th>Thrust force</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>Cutting force (N)</td>
<td>Change (%)</td>
</tr>
<tr>
<td>0 wt.%Bi</td>
<td>0.42</td>
<td>490.6±24.5</td>
<td>164.1±7.8</td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>0.9</td>
<td>429.5±21.1</td>
<td>-12.5</td>
</tr>
<tr>
<td>1.0 wt.%Bi</td>
<td></td>
<td>425.5±21.0</td>
<td>-13.3</td>
</tr>
<tr>
<td>0 wt.%Bi</td>
<td></td>
<td>478.0±23.2</td>
<td>153.6±7.6</td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td></td>
<td>410.2±20.0</td>
<td>-14.2</td>
</tr>
<tr>
<td>1.0 wt.%Bi</td>
<td></td>
<td>403.8±19.3</td>
<td>-15.5</td>
</tr>
</tbody>
</table>

The calculated COFs (Equation 2.9) of the Al 390 alloy with different amounts of Bi additions solidified at the rate of 9.4 °C/s (Table 6.8) indicated that the COF is 0.33
for the base alloy, 0.29 for the alloy with 1.0 % Bi when turning at a speed of 0.42 m/s, a decrease of 12.1 %. However, when turning at the speed of 0.9 m/s, the change in the COF was insignificant, the COFs of the alloys with 0 and 0.5 %Bi were same, while the COF of alloy with 1.0 % Bi decreased by 6.3 %.

The specific cutting energy was calculated (Table 6.8). The specific cutting energy was 872.9 MJm$^{-3}$ for the Al 390 base alloy, 741.1 MJm$^{-3}$ for the Al 390 alloy with 1.0 % Bi at the turning speed of 0.42 m/s, a decrease of 15.1 %. The specific cutting energy decreased from 842.1 MJm$^{-3}$ for the Al 390 base alloy to 699.2 MJm$^{-3}$ for the Al 390 alloy with 1.0 % Bi at the turning speed of 0.9 m/s, a decrease of 17.0 %, the drop percentage is similar.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>COF</th>
<th>Change (%)</th>
<th>Specific cutting energy (MJm$^{-3}$)</th>
<th>Change (%)</th>
<th>Cutting Speed (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0 wt.%Bi</td>
<td>0.33</td>
<td>-</td>
<td>872.9</td>
<td>-</td>
<td>0.42</td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>0.32</td>
<td>-3.0</td>
<td>748.3</td>
<td>-14.3</td>
<td></td>
</tr>
<tr>
<td>1.0 wt.%Bi</td>
<td>0.29</td>
<td>-12.1</td>
<td>741.1</td>
<td>-15.1</td>
<td></td>
</tr>
<tr>
<td>0 wt.%Bi</td>
<td>0.32</td>
<td>-</td>
<td>842.1</td>
<td>-</td>
<td>0.9</td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>0.32</td>
<td>0.0</td>
<td>722.5</td>
<td>-14.2</td>
<td></td>
</tr>
<tr>
<td>1.0 wt.%Bi</td>
<td>0.30</td>
<td>-6.3</td>
<td>699.2</td>
<td>-17.0</td>
<td></td>
</tr>
</tbody>
</table>

6.3.1.2 Interfacial temperature

Temperature at the interface between the chips and the rake face of the WC cutting tool is shown in Fig. 6.16. The temperature-turning time curves (Fig. 6.16) of all the alloys indicate that the temperature increased quickly at the beginning, and then reached an equilibrium value. The temperature at the interface decreased with an increase in the amount of Bi in the modified alloys.
Table 6.9 Interfacial temperature between chips and the rake face cast with a cooling rate of 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>0.42 (m/s)</th>
<th>0.9 (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Temperature (°C)</td>
<td>Change (%)</td>
</tr>
<tr>
<td>0 wt.%Bi</td>
<td>222.3±11.2</td>
<td>239.5±11.9</td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>200.3±9.8</td>
<td>-9.9</td>
</tr>
<tr>
<td>1.0 wt.%Bi</td>
<td>155.5±7.8</td>
<td>-30.0</td>
</tr>
</tbody>
</table>

Temperature (°C) at the interface between the chips and the rake face of the WC cutting tool cast with the cooling rate of 9.4 °C/s are listed in Table 6.9 and Fig. 6.17. The temperature was 222.3 °C in the Al 390 base alloy, 200.3°C in the alloy with 0.5 % Bi, 155.5 °C in the alloy with 1.0 % Bi, all turned with a speed of 0.42 m/s. The interface temperature decreased from 239.5 °C for the base alloy to 210.7 °C for the alloy with 0.5 % Bi, and to 161.2 °C for the alloy with 1.0 % Bi at the turning speed of 0.9 m/s.

6.3.1.3 Surface roughness

The finished surfaces of the Al 390 base alloy and the alloy with 1.0 % Bi cast with a cooling rate of 9.4 °C/s observed by SEM are shown in Fig. 6.18, and the average surface roughness of the Al 390 with different amounts of Bi addition are listed in Table 6.10. The average surface roughness (Ra) of the alloys are in the range of 0.859 µm to 0.934 µm at the turning speed of 0.42 m/s, and in the range of 0.821 µm to 0.949 µm at the turning speed of 0.9 m/s. The variation is insignificant.

Table 6.10 Average surface roughness of Al 390 alloys with different amounts of Bi cast at 9.4 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Turning speed of 0.42 (m/s)</th>
<th>Turning speed of 0.9 (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Surface roughness (µm)</td>
<td>Change (%)</td>
</tr>
<tr>
<td>0 wt.%Bi</td>
<td>0.911±0.264</td>
<td>-9.9</td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>0.934±0.128</td>
<td>2.5</td>
</tr>
<tr>
<td>1.0 wt.%Bi</td>
<td>0.859±0.220</td>
<td>-5.7</td>
</tr>
</tbody>
</table>
6.3.1.4 Cross-sectional microstructures of the chip-root

The fractured surface between the segments of the chip formed from turning at 0.9 m/s and 0.25mm/rev show that there are sliding lines on the segment (Fig. 6.19), indicating that sliding and severe plastic deformation occurred during chip formation. Microstructure of chip-root section can show evidence of the shear deformation during dry turning. The position of the chip-root cross section is indicated along the red line in the center of chip in Fig. 6.19. The chip-root itself was taken by interrupting the cutting process.

Figures 6.20 (a), (b) and (c) are the cross-sectional optical micrographs of the chips formed from the Al 319 base alloy and the alloys with 0.5, 1.0 % Bi during dry turning at a speed of 0.42 m/s. The segments were shorter with Bi addition (Table 6.11), and chip cracks appear more frequently. This trend was the same as that seen at 0.9 m/s (Figs. 6.21 (a), and (b)).

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Turning speed of 0.42 (m/s)</th>
<th>Turning speed of 0.9 (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Size (µm)</td>
<td>Change (%)</td>
</tr>
<tr>
<td>0 wt.% Bi</td>
<td>510±22</td>
<td></td>
</tr>
<tr>
<td>0.5 wt.% Bi</td>
<td>350±18</td>
<td></td>
</tr>
<tr>
<td>1.0 wt.% Bi</td>
<td>290±14</td>
<td></td>
</tr>
</tbody>
</table>

Figure 6.22 is the cross-sectional optical micrographs of the chips at high magnification, and alloys cast with a cooling rate of 9.4 °C/s. Fig. 6.22 (a) shows the cross-sectional optical micrographs of the material ahead of the chip root for the Al 390 alloy with 1.0 % Bi. The alloy generated discontinuous-type chips, with short chip segments mostly loosely attached to each other. Fig. 6.22 (b) shows the SEM micrograph
of an area between two segments of the chip, separated by a shear crack (taken from the location labeled as “A” in Fig. 6.22 (a)). The crack shows the presence of Bi particles along the shear crack between the two segments. Fig. 6.22 (c) is the cross-sectional optical micrograph of the Al 390 base alloy. The cross-section of the chip also exhibits cracks, but their number was less compared to that of 1.0 % Bi alloy and the chip had a less segmented appearance. The elongated Bi stringers were located in the crack near the tool tip based on the EDS maps (Fig. 6.23); gross crack formation was related to Bi stringers.

Morphology of the eutectic and primary Si particles in the chips of the Al 390 with 1.0 % Bi machined dry at a turning speed of 0.42 m/s and feed rate of 0.25 mm/rev indicated that the size of the eutectic Si phases after the turning tests was 2 µm, and the primary Si was broken into particles of 10.2 µm. The histograms of the eutectic and primary Si particles size are shown in Fig. 6.24.

Stringers of Bi were present in the cross-sectional microstructure observed by SEM-BSI (appearing bright) in the Al 390 alloy with 1.0 % Bi at a turning speed of 0.9 m/s and 0.25 mm/rev (Fig. 6.25). It was confirmed by an EDS spectrum to be pure Bi. The Bi underwent strong plastic deformation and melted.

Second phase particles or the impurities were responsible for the inhomogeneous strain distribution and may have acted as stress concentrators within the chip. The variations in the average cutting and thrust forces mirrored the role of the Bi addition: Bi promoted a discontinuous chip formation and decreased the friction between the tool and the chips.
6.3.1.5 Chip morphology

Chip length and morphology of the Al 390 base alloy and with 0.5 % and 1.0 % Bi cast with a cooling rate of 9.4 °C/s using a diamond cutting tool are shown in Fig. 6.26. The morphology of the chip formed from the alloys cast with a cooling rate of 9.4 °C/s is shown in Figs. 6.26 (a), (b), and (c) at a turning speed of 0.9 m/s and 0.25 mm/rev. Chip length decreased with increasing Bi concentration. It decreased from 3 mm with 0 % Bi to 0.9 mm with 1.0 % Bi for alloys cast with a cooling rate of 9.4 °C/s.

6.3.2 Al 390 Alloys Cast with a Cooling Rate of 26 °C/s

6.3.2.1 Cutting and thrust forces

Two speeds were used to test the Al 390 alloys with different amounts of Bi concentration cast with a cooling rate of 26 °C/s.

6.3.2.1.1 Turning at a speed of 0.42 m/s

The average cutting force was 469.9 N for the Al 390 base alloy, 395.3 N for the alloy with 0.5 % Bi, a decrease of 15.9 % (Table 6.12 and Fig. 6.27(a)). The average cutting force was 384.8 N for the alloy with 1.0 % Bi. The decrease in cutting force was less significant with a further increase in the Bi amount from 0.5 % to 1.0 %.

The average thrust force generated during cutting did not follow the same trend as the cutting force, and the thrust force decreased almost linearly with an increase in Bi addition (Fig. 6.27 (b)). More specifically, the average thrust force decreased from 122.8 N for the Al 390 base alloy to 103.4 N for the alloy with 0.5 % Bi, and to 80.6 N for the alloy with 1.0 % Bi (a 15.8 % decrease for the alloy with 0.5 % Bi, and a 34.4 % decrease for the alloy with 1.0 % Bi, compared to the base Al-Si alloy).
### Table 6.12 Average cutting and thrust forces of Al 390 with different amounts of Bi cast at 26 °C/s

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Cutting force</th>
<th>Thrust force</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Cutting force (N)</td>
<td>Change (%)</td>
</tr>
<tr>
<td>0 wt.%Bi</td>
<td>469.9±20.8</td>
<td>-</td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>395.3±18.2</td>
<td>-15.9</td>
</tr>
<tr>
<td>1.0 wt.%Bi</td>
<td>384.8±18.1</td>
<td>-18.1</td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>396.2±18.3</td>
<td>-16.3</td>
</tr>
<tr>
<td>1.0 wt.%Bi</td>
<td>393.5±18.2</td>
<td>-16.9</td>
</tr>
</tbody>
</table>

6.3.2.1.2 Turning at a speed of 0.9 m/s

The average cutting force was 473.4 N for the Al 390 base alloy, and 396.2 N for the alloy with 0.5 % Bi both cast with a cooling rate of 26 °C/s, a decrease of 16 % (Table 6.12 and Fig. 6.27 (a)). The average cutting force was 393.5 N for the alloy with 1.0 % Bi cast with a cooling rate of 26 °C/s, which is a 16.9 % decrease from that of the base alloy. The decrease in cutting force was less significant with a further increase in the Bi amount from 0.5 % to 1.0 %.

The thrust force decreased almost linearly with the increase of Bi addition (Table 6.12 and Fig. 6.27 (b)). When cast with a cooling rate of 26 °C/s, the average thrust force was 135.1 N for the base alloy, 115.9N for the alloy with 0.5 % Bi, and 84.6 N for the alloy with 1.0 % Bi, a decrease of 37.4 % totally.

Calculated COFs of the Al 390 with different amounts of Bi addition cast with a cooling rate of 26 °C/s (Table 6.13) indicated that the COF decreased from 0.26 for the base alloy to 0.21 for the alloy with 1.0 % Bi at a turning speed of 0.42 m/s, and from 0.29 to 0.21 at a turning speed of 0.9 m/s.
The specific cutting energy was also calculated (Table 6.13). The specific cutting energy was 790.4 MJm⁻³ for the Al 390 base alloy, and 620.6 MJm⁻³ for the Al 390 alloy with 1.0 % Bi at the turning speed of 0.42 m/s, and 811.3 MJm⁻³ for the Al 390 base alloy and 637.4 MJm⁻³ for the Al 390 alloy with 1.0 % Bi at the turning speed of 0.9 m/s.

The frictional force between the tool rake face and the chips can be determined by the thrust force, according to the Merchant circle diagram (a composite cutting force circle). It can be inferred that the frictional force decreased with the increase in Bi addition for both alloys cast with a cooling rate of 9.4 and 26 °C/s.

**Table 6.13** COF and specific cutting energy of Al 390 with different amounts of Bi cast at 26 °C/s

<table>
<thead>
<tr>
<th>Alloy</th>
<th>COF</th>
<th>Change (%)</th>
<th>Specific cutting energy (MJm⁻³)</th>
<th>Change (%)</th>
<th>Speed (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0 wt.%Bi</td>
<td>0.26</td>
<td></td>
<td>790.4</td>
<td></td>
<td>0.42</td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>0.26</td>
<td>0.0</td>
<td>664.9</td>
<td>-15.9</td>
<td></td>
</tr>
<tr>
<td>1.0 wt.%Bi</td>
<td>0.21</td>
<td>-19.2</td>
<td>620.6</td>
<td>-21.5</td>
<td></td>
</tr>
<tr>
<td>0 wt.%Bi</td>
<td>0.29</td>
<td>-19.2</td>
<td>811.3</td>
<td></td>
<td>0.9</td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>0.29</td>
<td>0.0</td>
<td>682.8</td>
<td>-15.8</td>
<td></td>
</tr>
<tr>
<td>1.0 wt.%Bi</td>
<td>0.21</td>
<td>-27.6</td>
<td>637.4</td>
<td>-21.4</td>
<td></td>
</tr>
</tbody>
</table>

6.3.2.2 Interfacial temperature

Temperature at the interface between the chips and the rake face of the WC cutting tool is shown in Table 6.14 and Fig. 6.17. The temperature at the interface decreased with increasing amount of Bi addition.

The interfacial temperature was 217.4 °C for the Al 390 base alloy, which decreased to 198.5 °C for the alloy with 0.5 % Bi, and to 157.5 °C for the alloy with 1.0 % Bi at the turning speed of 0.42 m/s. At a higher dry turning speed of 0.9 m/s, it decreased
from 230.1 °C for the Al 390 base alloy to 208.2 °C for the alloy with 0.5 % Bi, and to 163.4 °C for the alloy with 1.0 % Bi at the turning speed of 0.9 m/s.

Comparing the values in Table 6.9 and Table 6.14, when the cooling rate was increased, the interfacial temperature decreased when using the same cutting speed. This corresponds to the change in the cutting forces, thrust forces and friction.

**Table 6.14** Interfacial temperature between the chip and the rake face of alloys cast at 26 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Turning speed of 0.42 (m/s)</th>
<th>Turning speed of 0.9 (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Temperature (°C)</td>
<td>Change (%)</td>
</tr>
<tr>
<td>0 wt.%Bi</td>
<td>217.4±9.1</td>
<td></td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>198.5±8.9</td>
<td>-8.7</td>
</tr>
<tr>
<td>1.0 wt.%Bi</td>
<td>157.5±8.0</td>
<td>-27.6</td>
</tr>
</tbody>
</table>

6.3.2.3 Surface roughness

The finished surfaces of the Al 390 base alloy and that with 1.0 % Bi cast with a cooling rate of 26 °C/s observed by SEM are shown in Fig. 6.28. The average surface roughness of the Al-16Si alloys with different amounts of Bi addition cast with a cooling rate of 26 °C/s are also listed in Table 6.15. The average surface roughness was in the range of 0.854 µm to 0.925 µm at the turning speed of 0.42 m/s, and in the range of 0.884 µm to 0.899 µm at the turning speed of 0.9 m/s. The difference is insignificant.

It was also observed that the Bi particles with the size of 1.5 µm appeared on the finished surface of the Al 390 alloy with 1.0 % Bi and cast with a cooling rate of 26 °C/s, which indicated that Bi could have played the role of a lubricant on the tool rake face. (Fig. 6.29).
Table 6.15 Average surface roughness of Al 390 modified by Bi and cast at a cooling rate of 26 °C/s

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Turning speed of 0.42 (m/s)</th>
<th>Turning speed of 0.9 (m/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Surface roughness (µm)</td>
<td>Change (%)</td>
</tr>
<tr>
<td>0 wt.%Bi</td>
<td>0.925±0.109</td>
<td></td>
</tr>
<tr>
<td>0.5 wt.%Bi</td>
<td>0.866±0.130</td>
<td>-6.4</td>
</tr>
<tr>
<td>1.0 wt.%Bi</td>
<td>0.854±0.148</td>
<td>-7.7</td>
</tr>
</tbody>
</table>

6.3.2.4 Cross-sectional microstructures of the chip-root

Figures 6.30 and 6.31 are cross-sectional optical micrographs at low magnification. Micrographs of the material ahead of the chip root were taken by interrupting the cutting process at 0.42 m/s and 0.9 m/s. A diamond cutting tool was used. The alloy generated discontinuous-type chips, with short chip segments, most of them loosely attached to each other.

Figure 6.30 shows the cross-sectional optical micrographs of alloys with different concentrations of Bi turned at the cutting speed of 0.42 m/s. The chip segment is short for Al 390 with Bi addition (Figs. 6.30 (a), and (b)), and chip cracks appeared more frequently in the alloys with Bi addition than in the base alloy (Fig. 6.30 (c)). Figs. 6.31(a), (b) shows the cross-sectional micrographs of the alloy with 0.5 and 1.0 % Bi turned at a speed of 0.9 m/s. The trend of the crack in chips is as the same as those turned at 0.42 m/s.

Figures 6.32 and 6.33 are the cross-sectional optical micrographs at high magnification. Figure 6.32 (a) shows the cross-sectional micrographs of the base alloy turned at 0.42 m/s. Figs. 6.32 (b) and (c) are the cross-sectional optical micrographs of the alloys with 0.5 % Bi and 1.0 % Bi turned at 0.42 m/s. The cross-section of the chip
exhibits cracks too, but their number was less compared to that of 1.0 % Bi alloy and the chip had a less segmented appearance.

**Figures 6.33 (a), (b) and (c)** show the cross-sectional micrographs of alloy with 0, 0.5 and 1.0 % Bi turned at a speed of 0.9 m/s. The trend of crack in chips was the same as those turned at 0.42 m/s.

6.3.2.5 Chip morphology and surface morphology

The morphology of the Al 390 base alloy and with 0.5 % and 1.0 % Bi cast with a cooling rate of 26 °C/s is shown in **Figs. 6.34 (a) and (b)**; the chip length is 4 mm for the Al 390 without Bi, 0.4 mm for the Al 390 with 1.0 % Bi, cast with a cooling rate of 26 °C/s.

6.4. TEM Observation of the Chip

Bi particles were also observed by TEM observation of the chips that were subjected to large plastic deformation. The TEM samples were taken from near the root of the chips (**Fig. 6.35 (a)**). There are Bi particles (of about 0.6 μm) associated with the Si particles in the alloys with 1.0 % Bi cast with a cooling rate of 26.0 °C/s (**Fig. 6.35 (b)**, and 9.4 °C/s (**Fig. 6.35 (c)**). The dark particles were confirmed by SAED pattern and EDS spectrum to be pure Bi (**Fig. 6.35 (d), and (e)**). There were also cracks and deformation in the Si particles initiated from the Bi particles (**Fig. 6.35 (b), and (c)**), Si deformed and cracked initiated near Bi particles. However, the Si particles in the alloy with 1.0 % Bi that was cast with a cooling rate of 26.0 °C/s was completely fractured, with Bi particles whose size measured up to 2.4 μm (**Fig. 6.36**). There was no interface failure through debonding at the interface of Al/Si in all those observed, which indicated
that Bi does not deteriorate the interfacial bonding between the Si particles and the Al matrix.

6.5. Discussion

6.5.1 Bi particle formation and refinement

According to the equilibrium binary phase diagrams of Si-Bi (Fig. 6.37) and Bi-Al [54, 91], Bi has a low melting point (544 K), and is mutually immiscible in solid Al and Si [15]. The Bi-Al equilibrium system exhibits a monotectic transformation (Al-Bi monotectic composition is 4.11 % Bi) [54] to solid Al and liquid Bi at 930 K with a domain of liquid Bi immiscibility, and a eutectic reaction which extends from 0-100% Bi at 543 K. The Bi-Si phase diagram shows that there is a monotectic transformation at 1400 °C; and a eutectic reaction at 271 °C. In the Al-Si binary phase diagram, there is a eutectic reaction at 577 °C.

In this work, the alloy system is very complex (Table 3.1); the solidification has the followings stages in the equilibrium state: (1) the concentration of Bi is low at a level of 1.0 %, less than the monotectic point (4.11 % Bi) shown in binary phase diagram, so all the alloys are hypomonotectic Al–Bi alloys at 930 K, and the molten Al alloy is a homogenous single liquid phase (Fig. 6.38 (a)). (2) The calculated liquidus temperature of the Al-Si alloys with 1.0 % Bi used are 910.4 K (Table 3.1), based on the element-equivalency concept method [92], i.e. primary Si particles start to precipitate from the molten Al alloy at 910.4 K during solidification in the equilibrium state. The alloy system is a solid Si phase (and some intermetallic phases) with a liquid phase until near the Al-Si eutectic temperature, in which Bi accumulated to a high concentration at the solidification frontier of the precipitated primary Si particles (Fig. 6.38 (b)), and a high
concentration Bi area formed because the solidification process passed the monotectic transformation temperature of Al-Bi (930 K) and Si-Bi systems (1400 °C). (3) The pure liquid Bi droplet started to form at the temperature of the main Al-Si eutectic reaction (~577 °C) with most of the eutectic Al and Si in the solid states (Fig. 6.38 (c)). (4) There is still some amount of liquid Al-Si-Bi phases till the eutectic temperature of Al-Si-Bi is reached. Pure solid Bi started to form mainly at the Al-Si-Bi eutectic temperature (around 543 K). The final structure is demonstrated in (Fig. 6.38 (d)). The solidification process at the cooling rate of 9.4 °C/s took place as described above in the equilibrium state.

In this work, the alloy system cooling at 26 °C/s was a non-equilibrium solidification process. The temperature of the primary Si formation was 890.5 K based on the cooling curve of the Al-16Si alloy with 1.0 % Bi, in which the Bi accumulated to high concentrations at the solidification frontier of the precipitated primary Si particles (Fig. 6.38 (b)). The liquid Bi was rejected and pushed by the solidifying Si particles during solidification until the Al-Si eutectic temperature (827 K), as shown in Fig. 6.38 (c), or easily engulfed by Si particles (Fig. 6.35(b)). These accumulated Bi droplets could have adversely impact on and cause microdefects to the primary Si particles during their growth stage. Bi particles form and reside in the boundaries of the Si particles at the end of solidification due to the Bi droplet accumulation (Fig. 6.38(d)). The degree of accumulation is a function of the concentration of Bi. There were more Bi droplets associated with the primary Si particles in the alloy with 1.0 % Bi compared with 0.5 % Bi concentration. However, Bi was in the spherical shape because it did not wet the grain boundaries of the Al matrix and the primary Si phase, which resulted in no effect on the bonding of the Al and Si particles.
In contrast to solid-state transformations, the segregated droplets during decomposition in immiscible liquids are not fixed in space but move freely owing to various forces [93] (gravity, natural convection, forced convection) and also move relative to each other (Stokes motion or Marangoni motion). Thus the droplets can come so close to each other that they combine and form a bigger droplet. The observation of the relatively large size of the Bi particles in the alloy with 1.0 % Bi compared with 0.5 % Bi concentration could be attributed to the coagulation process.

At a high cooling rate, the time used for the combination of the segregated droplets is very short, and the final Bi particles size is refined. Traditionally, all kinds of phases undergo two major events: nucleation and the following growth, and the creation of a nucleus implies the formation of an interface at the boundaries of a new phase. The free energy needed to form a spherical cluster of radius \( r \) is:

\[
\Delta G = -\frac{4}{3} \pi r^3 G_v + 4\pi r^2 \sigma
\]  
\( (6.1) \)

\( G_v \) : liberates. Where the first term shows the energy gain of creating a new volume and the second term shows the energy loss due to surface tension \( \sigma \) of the new interface.

When \( \frac{d\Delta G}{dr} = 0 \), the critical radius reaches:

\[
y* = -\frac{2\sigma}{G_v}
\]  
\( (6.2) \)

at that point growth of the cluster is no longer limited by nucleation.

The greater the supercooling, the smaller the critical radius and the less the energy needed to form it. The free energy can be calculated as:
\[ \Delta G_v = \frac{\Delta H_v}{T_m} \Delta T \]  

(6.3)

So the greater the supercooling, the smaller the critical radius and the less energy needed to form it. At the higher cooling rate of 26 °C/s with a greater supercooling, all the Si and Bi phases are refined. When the eutectic Al-Si zones are refined, Bi is separated into smaller volumes, these also promote the refinement of the Bi phases under formation.

6.5.2 Fracture of the Si particles

Bi particles are engulfed in the Si crystals because of the Bi accumulation in the solidification frontier of the primary Si phase (shown in Fig. 6.35(b)). This observation is similar to other studies on the Al-Si alloys that excess aluminum is engulfed by the primary Si particles. Some of the primary Si crystals contain interior regions or islands of Al alloys that are completely enclosed in the corresponding Si crystals \([94, 95]\) when cast in a permanent mold for an unmodified Al-12 % Si-1 % Ni base alloy. The primary silicon plates formed under a low cooling rate commonly show stratified internal deposits of aluminum parallel to the plate surfaces \([96]\). The stratified internal deposits of aluminum are actually the regions of excess aluminum engulfed by the growing Si crystal.

From the results presented above, it can be concluded that cracks initiate in Si particles in contact with Bi present in particles whose size is in the micrometer range as shown in Fig. 6.35 b), c). The average size of the Bi particles in the micrometer range is a strong function of the cooling rate, also of the distribution of the Bi particles around the primary Si particles, determined by the Bi concentration. So alloys with 0.5 % Bi cast with a cooling rate of 26.0 °C/s provided better tensile strength and elongation resulting from the refined Bi particles (at the higher cooling rate) and less Bi distributed around the
primary Si particles. Refined primary Si particles cast at the high cooling rate also improved the mechanical properties of the Al-Si alloys.

Larger Si particles have a higher probability of containing a flaw of critical size than smaller ones. They usually fracture at considerably lower tensile strengths, reported [97, 98] as low as 0.2 GPa. The fracture initiates once the applied stress on silicon exceeds the silicon strength. The statistical presence of microdefects (cavity or scratch) in Si particles is generally considered to be the source of such results [98-101]. The fracture of Si particles usually occurs along the <111> cleavage plane [102]. The fracture toughness of Si (K1c= 0.83-0.95 MPa.m1/2) is low, leaving it sensitive to defects or cracks. Bi particles cause microdefects in the boundary of Si particles during solidification of the Al-Si alloys, and there are cracks in the Si particles that initiate from the position related to Bi particles.

6.5.3 The Role of Bi on the Machining Performance

When cast with the higher cooling rate of 26.0 °C/s, the primary and eutectic Si particles, the aluminum matrix and the Bi particles are all refined. The mechanical properties of Al-Si alloys are maintained for the alloy with 0.5 % Bi and the machining performance are improved.

For Al-16Si with 0.5 % Bi alloys compared with the base alloy cast with a cooling rate of 26 °C/s and turned at 0.42 m/s, the cutting and thrust forces decreased by 15.9 % and 15.8 % respectively, the specific cutting energy decreased by 15.9 %, the average temperature at the interface between the chips and the rake face decreased from 217.4 °C to 198.5 °C, and the surface roughness decreased from 0.925 μm to 0.866 μm. The average length of the chip segments decreased. Chips were of the discontinuous type. Bi
decreased the cutting force and the tool rake temperature, promoted a discontinuous chip formation and decreased the friction between the tool and the chips.

In this work, the microstructures of the Al-16Si alloy consist of large primary Si, eutectic Si and Al matrix, and intermetallics. During the cutting process, the primary Si and eutectic Si are broken to small fragments, and the Al adhered on the tool surface.

This situation changed with the addition of Bi. (1) The Bi particles melted during the orthogonal cutting as discussed in Sec. 5.3.4. The tested chip bulk temperature was lower than the melting point of Bi, however, but the temperature in the shear bands were high enough to cause the melting of Bi. (2) Melted Bi lubricated the tool rake face, reduced Al adhesion, and promoted chip segment formation. (3) Large primary Si particles deteriorated the alloy machining performance. They easily fractured when the Bi was associated with the primary Si, and decreased the cutting forces and energy. (4) Refined primary Si particles cooled at a high cooling rate also decreased the cutting forces and energy.

The optimum alloy was found to be Al-16Si with 0.5 % Bi cast with a high cooling rate. Bi in the alloy decreased the tool rake face temperature and cutting forces.

**6.6. Conclusions**

This chapter analyzed the microstructures and the cross sections of the tested samples to determine the effect of Bi modification in the alloys cast with two different solidification rates on the room-temperature tensile properties and dry machining performance of hypereutectic Al-16% Si alloys. The main conclusions arising from this work are as follows:
(1) The size of the Bi particles increased when the cooling rate was decreased and the amount of Bi increased. However, most of the Bi particles precipitated at the interface of eutectic Si, regardless of the cooling rate, and the percentage of each phase associated with precipitated Bi particles was also mainly decided by the Bi concentration.

(2) About 60% of primary Si particles had Bi particles precipitated at their interface in the alloy with 1.0% Bi. Only 30% of the particles had Bi precipitated at their interface in the 0.5% Bi alloy.

(3) The higher cooling rate enhanced the mechanical properties by refining the Si and Bi particles. However, the mechanical properties of the Al-Si alloys with 1% Bi decreased even upon casting with a high cooling rate, as the Bi particles precipitated on the primary Si particles and induced Si particle fracture.

(4) Interfacial temperatures and cutting and thrust forces decreased with the increase in the amount of Bi addition. Bi promoted a discontinuous chip formation by melting during dry turning because of the large shear localization experienced in the shear band.

(5) Al-16% Si alloys with 0.5% Bi subjected to a high solidification rate maintained the mechanical properties of the base alloy, while improving the dry machining performance.
FIGURES

Fig. 6.1 SEM-BEI images of Al 390 with 1.0 % Bi cast at the cooling rate of 9.4 °C/s.

Fig. 6.2 The size of Bi particles varied with the amount of Bi addition and cooling rates.

Fig. 6.3 Typical tensile engineering stress-strain curves of Al 390 with different amounts of Bi cast at 9.4 °C/s.
Fig. 6.4 SEM-BEI images of the tensile fracture surface of the alloys cast at 9.4 °C/s. (a) 1.0 % Bi, (b) 0.5 % Bi.

Fig. 6.5 EDS spectrum shows pure Bi particles in Fig. 6.4.
Fig. 6.6 Cross sections of alloys cast at 9.4 °C/s after tensile tests. (a) with 1.0 % Bi, (b) with 0.5 % Bi, (c) 0 % Bi, (d) SEM image of (a)
Fig. 6.7 Compressive stress-strain curves of alloys cast at the 9.4°C/s and compressed at 240°C.

Fig. 6.8 Cross-sectional microstructure for Al 390 with 1.0% Bi cast at 9.4°C/s after compressive test.
Fig. 6.9 SEM-BEI images of Al 390 with 1.0 % Bi cast at the cooling rate of 26 °C/s.

Fig. 6.10 The tensile stress - strain curves of Al 390 modified by Bi cast at 26.0 °C/s.
Fig. 6.11 SEM-BEI images of the tensile fracture surface of alloys cast at the cooling rate of 26 °C/s
(a) with 1.0 % Bi, and (b) with 0.5 % Bi.
Fig. 6.12 Cross-sections of tensile samples for Al 390 modified by Bi cast at 26 °C/s
(a) 1.0 % Bi, (b) 0.5 % Bi and (c) 0 % Bi.

(c)

Fig. 6.13 Compressive stress-stain curves at 240 °C of Al 390 modified by Bi cast at 26 °C/s
Fig. 6.14 The compressive properties of Al 390 with 1.0 % Bi cast at 26 °C/s (a) cross-sectional microstructure, (b) SEM-BEI image, and (c) SEM-EDS spectrum of Bi particle
Fig. 6.15 Cutting and thrust forces of Al 390 with different amounts of Bi cast 9.4 °C/s (a) Cutting force, (b) Thrust force.

Fig. 6.16 Temperature-time variation of Al 390 modified by Bi cast at 9.4 °C/s turning at 0.42 m/s
Fig. 6.17 Temperature variation for alloys modified by Bi during dry turning (a) at 0.42 m/s, and (b) at 0.9 m/s.

Fig. 6.18 SEM images of finished surfaces of alloys cast at 9.4 °C/s (a) Al 390, and (b) Al 390 with 1.0 % Bi.
**Fig. 6.19** Fracture face between segments of chip of Al 390 with 1.0 % Bi turning at 0.9 m/s, 0.25mm/rev
Fig. 6.20 Cross-sectional micrographs of chips for alloys cast at 9.4 °C/s and turning at 0.42 m/s
(a) 0.0% Bi, (b) 0.5 % Bi, and (c) 1.0 % Bi.
Fig. 6.21 The cross-sectional micrographs of chips for alloys cast at 9.4 °C/s and turning at 0.9 m/s
(a) 1.0% Bi, (b) 0.5% Bi.
Fig. 6.22 Cross-sectional micrographs of chip root at turning speed of 0.42 m/s (a) Al 390 with 1.0 % Bi, (b) Bi in the crack from insert in (a), (c) Al 390 alloy.
Fig. 6.23 EDS maps of Al 390 with 1.0 % Bi cast at a cooling rate of 9.4 °C/s.
Fig. 6.24 Histograms of Si particles in chips of Al 390 with 1.0 % Bi cast at a cooling rate of 9.4 °C/s
(a) primary and (b) eutectic Si turning at the speed of 0.42 m/s
Fig. 6.25 Elongated pure Bi stringers in chips of Al 390 with 1.0 % Bi cast at 9.4 °C/s cutting at 0.9 m/s.
Fig. 6.26 Chips morphology using diamond tool for alloys cast at the cooling rate of 9.4 °C/s
(a) 0.0 %Bi, (b) 0.5 %Bi and (c) 1.0 %Bi
Fig. 6.27 Cutting and thrust forces varied with the amount of Bi addition at the cooling rate of 26 °C/s
(a) Cutting force, (b) Thrust force.

Fig. 6.28 SEM images of finished surface of alloys cast at a cooling rate of 26 °C/s
(a) Al 390, and (b) Al 390 with 1.0 % Bi.
Fig. 6.29 The finished surface of Al 390 observed by SEM shown Bi particle.

(a) SEM image with EDS, (b) Insert in (a)
Fig. 6.30 The cross-sectional micrographs of chips for alloys cast at 26 °C/s and turning at 0.42 m/s
(a) 0.0 % Bi, (b) 0.5 % Bi, (c) 1.0 % Bi.
Fig. 6.31 The cross-sectional micrographs of chips for alloys cast at 26 °C/s and turning at 0.9 m/s
(a) 0.5 Bi, (b) 1.0 % Bi.
Fig. 6.32 Cross-sectional micrographs of chips in high magnification for alloys cast at 26
°C/s and turning at 0.42 m/
(a) 0.0 % Bi, (b) 0.5 % Bi, (c) 1.0 % Bi.
Fig. 6.33 Cross-sectional micrographs of chips in high magnification for alloys cast at 26
°C/s and turning at 0.9 m/s
(a) 0.0% Bi, (b) 0.5% B, (c) 1.0 Bi.
Fig. 6.34 Chips morphology of alloys cast at the cooling rate of 26 °C/s (a) with 0.0 % Bi, and (b) 1.0 % Bi.
Fig. 6.35 TEM observation of Bi in chips
(a) Position of TEM sample in chips, (b) TEM micrograph of alloys with 1.0 % Bi cast at the cooling rate of 26.0 °C/s. (c)TEM micrograph of alloy at the cooling rate of 9.4 °C/s, (d) indexed SAED pattern of Bi in (b), (e) EDS spectrum of Bi particles in (b).
Fig. 6.36 TEM micrographs of chip from alloy with 1.0 % Bi cast at the cooling rate of 26.0 °C/s
Illustrating fractured Si with Bi

Fig. 6.37 Si-Bi equilibrium binary phase diagram
(a) > 910.4K at equilibrium state or > 893.5k at 26°C/s

(b) 910.4K–850K at equilibrium state or 893.5–820K at 26°C/s

(c) 850K at equilibrium state or 820K at 26°C/s

(d)~543K

**Fig. 6.38.** Schematics diagrams show Bi solidified on Al/Si interface
(Solid black dot is solid Bi, Open black dot is liquid Bi)
(a) Molten Al alloy is homogenous one liquid phase (b) Primary Si particles precipitated from molten Al, Bi began to separate into globules of bismuth dispersed in a liquid matrix of aluminum alloy, (c) Segregated liquid Bi droplets separated by Si Al phases. (d) Solid Bi particles form and segregated in the boundaries of the Si particles
CHAPTER 7 Conclusions & Recommendations for Future Work

7.1. Conclusions

This work systematically studied the effect of two low-melting-point elements, Bi and Sn, on the microstructure, mechanical properties and dry machining performance of Al-Si alloys. The key conclusions drawn from the observations of this work are as follows:

(1) For the hypoeutectic Al-Si alloy, the eutectic silicon morphology is coarsened with the increase of Bi under the lower cooling rate; the morphology is changed marginally at the higher cooling rate. Sn has no impact on the silicon morphology.

(2) For the hypereutectic Al-Si alloy, Bi has little effect on the size of the primary Si particles. Eutectic Si particles are increasingly refined as the amount of Bi is increased.

(3) Most Bi particles precipitate at the interface of the Si and Al phases, and the percentage of the different phases associated with the precipitated Bi particles is mainly decided by the Bi concentration.

(4) Turning forces decrease with the addition of Bi and Sn.

(5) The cutting temperature decreases with the addition of Bi and Sn.

(6) The addition of Bi and Sn forms shorter chips, and Bi and Sn appear within the shear bands.
(7) Bi and Sn melt during dry turning because of the large shear localization in the shear band, and promote discontinuous chip formation, thereby improving the dry machining performance of the alloys modified by them.

(8) The optimum concentration of Bi or Sn has been found to be 0.5%.

In summary, the practical recommendations that can be made to improve the dry machining performance of Al-Si alloys modified with Bi and Sn are: (a) Bi and Sn can both be used to improve the dry machining performance of Al-Si alloys, (b) mechanical properties can be maintained with Bi modification, but is compromised by Sn modification.

7.2. Recommendations for Future Work

The dry machining performance of alloys modified by Bi or Sn was improved based on the decrease of cutting force and thrust force, the interfacial temperature of the chip and tool rake face, and the chip morphology. Both the amplitude and period of the cutting and thrust forces decreased (Fig. 5.51) for alloys modified by Bi. The machining processing could be more stable and BUE could be delayed. Tool life should be improved and the economical benefits will be significant for automotive industry. However, future work that may be done to carry this work forward would include:

(1) Die cast aluminum alloys have been in automotive applications such as the transmission box, engine and piston. As die casting processes have the advantage of high cooling rates and do not require Sr modification. The interaction between Bi and Sr is avoided, and a high cooling rate will refine
the microstructure and maintain the mechanical properties of the base alloys, while improving the dry machining performance of die cast parts with the addition of Bi or Sn.

(2) The effect of the high cooling rate and the size and distribution of Bi and Sn on the dry machining performance of die cast Al-Si alloys need further investigation. The dry machining performance in dry drilling and tapping need special attention, because auto parts of these alloys, such as the engine and piston, require drilling and tapping processes in addition to turning after casting.

(3) The study of high cooling rates of casting such auto parts is important in order to apply this knowledge in industrial applications. The undercooling degree of the primary and eutectic reactions need further systematic research in these complex alloys system. Bi particle movement and formation during high rate solidification need further investigation to better control their size and distribution, along with a deeper understanding of the interaction of Bi with Si particles.

(4) Research on different tools and tool materials for the dry machining performance of die cast parts with Bi and Sn modification is another aspect of the machinability of materials that will greatly benefit the automotive industry.

(5) Evaluation the weight and rank of the amount of Bi addition and the cooling rate.
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