Characterization of High-Pressure-Die-Cast Magnesium Alloy AM60

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A thesis submitted in partial fulfillment of the requirements for the degree in Master of Engineering Science
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CHARACTERIZATION OF HIGH-PRESSURE DIE-CAST MAGNESIUM ALLOY AM60

by

Hooman Baghaei Anaraki

Graduate Program in Engineering Science
Department of Mechanical & Materials Engineering

A thesis submitted in partial fulfillment of the requirements for the degree of Master of Engineering Science

The School of Graduate and Postdoctoral Studies
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Abstract

The primary goal of this study is the characterization of two new complex die castings of AM60 so that the models which are proposed by different researchers for the prediction of fracture and yielding behaviors of die-castings can be validated. Based on the location of gate system which represents the filling pattern of castings, fifteen locations on the bolster and six locations on the instrument panel are considered for microscopic and mechanical experiments. The local mechanical properties of each location are determined by tensile test, and the microstructureal features of each region are characterized by analyses of metallographic images. The results of experiments are the input variables in the proposed models. The fracture strain of samples ranges from 1.2 to 9.5%, and slight deviation of 10-15MPa is obtained for the local yield strength. The local area fraction of porosities and skin fraction are obtained from microscopic studies. The skin thickness ranges from 0.27 to 0.55 mm, and the area fraction of porosities ranges from 0.4 to 2.85% within the knit line region. By introduction of skin fraction as a ligament factor of the failure model, the fracture strain can be estimated with high accuracy. The deviation between predicted and experimental values was less than 10% for most locations. The premature failure happens to the samples containing porosities within the skin region, and actual positions of pores are considered for prediction of fracture strain. The microscopic studies of sections along the flow path also indicate that the thick defect bands are formed within the flange gates, and a higher fraction of the gas pores is detected within the region closer to the overflows, and average area fraction of porosities reaches 2.8%. To evaluate the yielding behavior of samples, the elasto-plastic transition point of the samples containing different fractions of large dendrites are determined by the analyses of strain hardening rate. The fully plastic behavior begins at 107MPa for the samples containing accumulated large dendrites within the core region, while it ranges from 121 to 132MPa for the samples with fine microstructure.

Keywords:
Magnesium die-cast, Failure model, Microstructure-properties relationship, Characterization of microstructural features
Dedication

I would like to dedicate my thesis to my family who supported me during my education. I would especially appreciate the support and encouragement which are provided by my loving wife. I would specially dedicate this work to my loving daughter and wife who gave me the strength to fulfill my career goals. I also dedicate this work to my mother who supported me with her kind words. I also dedicate the thesis to the soul of my beloved dad.
Acknowledgments

I would like to thank Meridian Technology Inc. for the help and materials that assisted me to fulfill the research. I would like to express my gratitude to Dr. Jon Weiler for the time that he spent to provide tremendous help for fulfillment of the project.

I would like to acknowledge AUTO21 Network of Centers of Excellence as second founder of the research; the project is partially funded by auto industries and government and aims to resolve the issues related to the auto industries.

The research is a part of the continuous research which has been conducted at UWO within the last decade, and I express my deepest gratitude to the former researchers who performed the characterization and modeling of the die cast magnesium. I would like to thank my co-workers who helped me to improve my knowledge and skills in testing and characterization of the materials. I would like to thank Dr. Ying Fan for her assistance in microscopic studies and academic improvement in my research.

I would like to express my deepest gratitude to Dr. Jeff Wood for the consistent support and encouragement. In the light of his academic assistance and information this work successfully progressed.
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</thead>
<tbody>
<tr>
<td>HPDC</td>
<td>High Pressure Die Casting</td>
</tr>
<tr>
<td>$\sigma_y$</td>
<td>Yield strength</td>
</tr>
<tr>
<td>$\sigma_{uts}$</td>
<td>Ultimate tensile strength</td>
</tr>
<tr>
<td>IP</td>
<td>Intensified pressure stage</td>
</tr>
<tr>
<td>$C_s$</td>
<td>Composition of solid state</td>
</tr>
<tr>
<td>$C_L$</td>
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<tr>
<td>$f_s$</td>
<td>Fraction of solid state</td>
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<tr>
<td>$f_L$</td>
<td>Fraction of liquid state</td>
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<tr>
<td>$C_0$</td>
<td>Composition of alloy</td>
</tr>
<tr>
<td>$K$</td>
<td>Equilibrium partition</td>
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<tr>
<td>$T_L$</td>
<td>Liquidus Temperature</td>
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<tr>
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<td>Solidus temperature</td>
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<tr>
<td>$T_m$</td>
<td>Melting point</td>
</tr>
<tr>
<td>$\Delta T_f$</td>
<td>Solidification Range</td>
</tr>
<tr>
<td>HTC</td>
<td>Heat transfer coefficient</td>
</tr>
<tr>
<td>ESG</td>
<td>Externally solidified grains</td>
</tr>
<tr>
<td>$d$</td>
<td>Grain size diameter</td>
</tr>
<tr>
<td>$R$</td>
<td>Cooling rate</td>
</tr>
<tr>
<td>$\sigma$</td>
<td>Flow stress/true stress</td>
</tr>
<tr>
<td>$\theta, \frac{d\sigma}{d\varepsilon}$</td>
<td>Strain hardening rate</td>
</tr>
<tr>
<td>$G$</td>
<td>Shear modulus</td>
</tr>
<tr>
<td>$E$</td>
<td>Young modulus</td>
</tr>
<tr>
<td>$Y_s$</td>
<td>Proof stress</td>
</tr>
<tr>
<td>$\rho$</td>
<td>Density of dislocation</td>
</tr>
<tr>
<td>$b$</td>
<td>Bergers vector</td>
</tr>
<tr>
<td>$\gamma$</td>
<td>Shear Strain</td>
</tr>
<tr>
<td>$L$</td>
<td>Mean free distance between dislocations</td>
</tr>
</tbody>
</table>
\begin{itemize}
  \item $\gamma$ Strain rate
  \item $f_p$ Volume fraction of dispersed particles
  \item $E_p$ Young modulus of dispersed particles
  \item $\sigma_0, K$ Friction stress, and Hall-Petch slope in Hall-Petch equation
  \item $f_t$ Fraction of field of measurement in Hall Petch equation
  \item $f$ Area fraction of porosities in failure model
  \item $\varepsilon$ True strain
  \item $n$ Strain hardening exponent
  \item $A_i$ Area of plane containing void
  \item $A_h$ Area of plane far from the void
  \item $\varepsilon_i$ Maximum critical strain
  \item $\varepsilon_h$ Homogenous strain at fracture
  \item $K_c$ Stress concentration
  \item $\delta_L$ Ligament fraction
  \item $r$ Radius of equivalent pore
  \item $t$ Thickness of casting samples
  \item $d_{w}, d_r$ Distance of pore from the edge of sample
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Chapter 1

1 Introduction

Recent growing concern for environmental issues and natural resources is the positive motivation for application of lightweight metals in the new generation of vehicles. The lower fuel consumption of lighter vehicles results in the lower amount of air pollutants. Magnesium alloys with low density, high strength to weight ratio as well as excellent castability can play an important role in the manufacturing of auto parts. Regarding low density of magnesium component, the use of magnesium components in a new generation of vehicles improves the fuel efficiency and leads to the reduction of exhaust emissions.

High pressure die casting (HPDC) is the dominant process to manufacture magnesium alloy components. In addition to the high production rate, accurate dimensions and finished surface are the superior advantages of HPDC process. The injection technique leads to the fluctuation of local mechanical properties across the castings. Therefore, the prediction and characterization of mechanical properties are of interest for both industrial and academic centers.

1.1 Research Background

In this section I will highlight the previous project, which is conducted by our research group. The challenging issue with products which are manufactured by HPDC process is the inconsistency of mechanical properties across the castings. The variation of mechanical properties across the casting is shown in Figure 1.1. The mold filling and solidification condition are the most important parameters which affect the microstructure of die-castings, and the microstructure of castings governs the local mechanical properties such as yield strength and ductility. The grain size and skin thickness are the influential parameters on the yield strength of casting, and fracture behavior of the casting is controlled by internal defects which are usually located within the core region of specimens. The main purpose of the project was to predict the local mechanical properties of the casting based on the simulation of casting processes.
The true stress-strain curve of specimens which were extracted from different locations of instrument panel beam [1].

The project began in 2001 with delivery of fifty castings named instrument panel beam. Fifty-seven locations of each casting were considered for extraction of tensile and bending specimens. The industrial partner of the project was Meridian Technologies Inc. The results of mechanical tests indicated that the fracture strain ranges from 3% to 10.5% and fracture stress is changing from 138 to 237MPa. The yield strength of the samples was between 95 to 130MPa [2].

The microstructures of samples which were extracted from different regions of the castings were investigated to determine the variation of grain size within the cross section of samples [3]. The results which were obtained from the microscopic studies were applied to determine the skin thickness of samples by choosing appropriate criteria [4]. In addition to the variation of grain size, the mechanical properties of skin and the core region of HPDC samples were examined by microindentation test. In addition to the flow curve of skin and core region, the strain hardening rates of the regions were also determined [5].
The micro indentation method was also applied to other magnesium alloys which were manufactured by gravity casting [6]. The constants of Hall-Petch equation were obtained from these investigations to predict the yield strength of the castings, and modified form of the equation was proposed to appreciate the bimodality of microstructure which is an inherent characteristic of the HPDC process [7]. To evaluate the effects of processing parameters on structural properties of the castings, the correlation between the solidification condition and structural properties was investigated. The thermal gradient and solidification conditions of different castings which manufactured by sand and permanent molds were studied, and an equation was proposed to estimate the grain size of the casting from the measured cooling rate [8].

The correlation between internal defects and fracture properties of tensile specimens extracted from instrument panel was investigated by X-ray tomography method. The size and spatial distribution of porosities as well as the maximum area fraction of porosities were determined from this investigation [1]. Moreover, the damage accumulation of defects during the consecutive loading process was studied [9]. In addition to the effect of internal porosities in decreasing of load bearing capacity, the stress concentration in the vicinity of the pore region was considered to propose the analytical model for prediction of fracture strain [10]. The distance of pores form the free surface was also considered as an important parameter, and a failure model was proposed to predict the fracture strain and stress of the castings. The ligament fraction is implemented into the failure model expresses the proximity of the equivalent pore to edge of the specimens. The skin region of the die-cast specimens is known as a pore free reign, so the least ligament fraction would be equal to the skin fraction of the specimens [11].

It’s also noted that the variation of strain hardening rate within the skin and the core region of die-cast specimens is considered by another researcher [12]. The die-cast specimens were treated as composite materials and the skin fraction and onset of fully plastic deformation is determined for the die-cast specimens of magnesium alloys with different aluminum content. It’s demonstrated that the elastic fraction of the composite was higher for alloys with a higher percentage of aluminum, and fully plastic behavior began at a higher stress.
1.2 Objective

The previous explanation implies that different models were suggested for prediction of mechanical properties of die-cast magnesium alloys. Although they were successfully applied for a certain type of complex die casting, it’s necessary to be examined for other types of complex die-castings of magnesium alloys. Therefore, two other types of complex die-castings are chosen for this study in order to validate the proposed failure model. The yielding behavior of the new complex die castings is also examined by the modified Hall-Petch equation. Moreover, the variation of yielding behaviors across the casting is studied by the analyses of strain hardening rate for the samples containing different microstructure. While the percentage of solute content can be considered relatively constant for all samples, the microstructural variations across the casting lead to different strain hardening behavior. To justify this behavior, the samples can be treated as composite materials, and the fraction of hard constituent is dependent on the diversity and distribution of grain size across the sample.

Chapter 2 includes a brief explanation about the physical and mechanical properties of magnesium alloys, and HPDC process is also introduced in this chapter. The models which are applied in this study are also explained in Chapter 2. The methods and findings of other researchers are also presented in this chapter.

The new castings which are selected for the validation of proposed models are introduced in Chapter 3. The experimental methods and procedures are designed in a way to fulfill the main purpose of this study. To characterize the local mechanical properties of the casting, tensile and hardness tests are utilized. The metallographic methods, image analysis and processing are also explained in Chapter 3, and they are utilized to characterize the microstructural features. The main purposes of microscopic studies were the determination of skin thickness, area fraction of porosities as well as characterization of grain size within the cross section of specimens. The first two are primary variables in the failure model, and the last one is an effective parameter in justification of yielding behavior.
The results of mechanical tests and morphology of fracture surfaces are presented in Chapter 4. The procedure which is applied to determine the skin thickness is explained in this chapter. The appropriate functions which are applied to the characterization of microstructural features are also introduced in Chapter 4. The relationship between microscopic studies and mechanical properties are described in Chapter 4, and the failure model is validated in this chapter. The correlation between the microstructural variation and yielding behavior are also justified by different models. The comparison between the predicted and experimental results is explained in Chapter 4.

The conclusions which are obtained from the research are summarized in Chapter 5. A brief summary about the correlation between the mechanical properties and microstructure of the specimens are presented in Chapter 5.
Chapter 2

2 Literature Review

The general properties of magnesium alloys and HPDC process are presented in this section. The formation of microstructure and plastic deformation of the die cast magnesium alloy AM60 are discussed. The various casting defects which affect the ductility of components are introduced, and the functions which can be utilized for characterization of defects are explained. The correlation between microstructure and mechanical properties of the HPDC components are explored in this chapter. The various models to predict the ductility of the HPDC magnesium alloys are studied and the effects of the skin thickness on the strength of components are described.

2.1 An Introduction to Magnesium Alloys

The global trend is to produce the new generation of vehicles which have higher efficiency and lower fuel consumption. The lightweight vehicles are fuel efficient, and produce lower greenhouse gas emission. The density of magnesium is $1.74 \text{ g/cm}^3$, or approximately one-quarter that of steel. A study by Kim [13] investigating the replacement of steel seat frames with magnesium die castings resulted in a weight reduction of 40%. The specific strength of AZ91 is remarkably higher than that of the aluminum alloy and iron. Moreover, high fluidity of molten magnesium alloys enables manufacturers to produce more complex and thin-walled castings. Therefore, magnesium alloys are good candidates for substitution of steel and aluminum in automotive parts such as instrument panel, steering wheels and seats [14].

Aluminum, as alloying element, improves the tensile strength and hardness of magnesium alloys. The formations of solid solution at 437°C and $Mg_{8_{17}}Al_{12}$ particles ($\beta$-phase) under non-equilibrium solidification condition are the main parameters in the improvement of mechanical properties. Due to the low thermal stability of $\beta$-phase particles, the rare earth (RE) metals are added to magnesium alloys to suppress the formation of $\beta$ phase and improve the creep resistance of the alloys [15]. Due to the
formation of $\text{Mg}_2\text{Si}$, the improvement of creep resistance is also achievable by addition of silicon as alloying element [16]. Other alloying elements are Zinc, Manganese and Zirconium. Zinc is usually accompanied by aluminum and improves the tensile strength of the alloy. Manganese has a strong effect in improving resistance against the corrosive environment [17]. Zirconium is added to the magnesium alloys to refine the microstructure of products. The General mechanical properties and composition of the magnesium alloys are tabulated in Table 2.1.

Table 2.1 Common magnesium alloys and their application [18]

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Composition</th>
<th>Properties and Application</th>
</tr>
</thead>
<tbody>
<tr>
<td>AM60B</td>
<td>%Al=5.5-6.5, %Mn=0.13-0.6, %Si=0.1, %Zn=0.22</td>
<td>$\sigma_y = 130\text{MPa}, \sigma_{UTS} = 240\text{MPa},$ Elongation=13% Automotive structures</td>
</tr>
<tr>
<td>AM50</td>
<td>%Al=4.4-5.4, %Mn=0.26-0.6</td>
<td>$\sigma_y = 125\text{MPa}, \sigma_{UTS} = 230\text{MPa},$ Elongation=15% Automotive structures</td>
</tr>
<tr>
<td>AZ91A</td>
<td>%Al=8.3-9.7, %Mn=0.13-0.5, %Zn=0.35-1</td>
<td>$\sigma_y = 160\text{MPa}, \sigma_{UTS} = 250\text{MPa},$ Elongation=8%</td>
</tr>
<tr>
<td>AE42</td>
<td>%Al=4, %Mn=0.1, %RE=2.5</td>
<td>$\sigma_y = 145\text{MPa}, \sigma_{UTS} = 230\text{MPa},$ Elongation=11% Applicable at high temperature, high creep resistance</td>
</tr>
<tr>
<td>AS41A</td>
<td>%Al=3.5-5, %Mn=0.2-0.5, %Si=0.5-1.5</td>
<td>$\sigma_y = 130\text{MPa}, \sigma_{UTS} = 240\text{MPa},$ Elongation=15%</td>
</tr>
</tbody>
</table>

2.2 Mg-Al Diagram Phase

The most common magnesium alloys in auto industries are based on the Mg-Al system (Figure 2.1). The equilibrium structure of the alloys consists of $\alpha$ phase. Although the eutectic transformation occurs within 13-33wt% aluminum, the non-equilibrium solidification in HPDC process leads to the formation of divorced eutectic, and $\alpha$ grains composed of layers with different solute content. The formation of eutectic phase is even observed in the magnesium alloys containing 2wt% aluminum [19].
Figure 2.1 Mg-Al equilibrium phase diagram, the red dashed line shows the composition of AM60 [6]

The segregated layers around the primary phase are shown in Figure 2.2(a), and the $\alpha$ grains are surrounded by the eutectic which is so called the eutectic grain boundary [20]. The eutectic transformation and growth depend on the diffusion mechanism; hence the solute content and cooling rate are the influential factors on the morphology of eutectic phase. The divorced eutectic will be formed once the volume fraction of one component is very small, and it appears as isolated islands in final microstructure[21]. The morphology of eutectic phase will be transformed from the fully divorced into the lamellar structure in low cooling rate and higher aluminum content [19].
Figure 2.2 Comparison between equilibrium and non-equilibrium diagram
phase(a), formation of segregated layers and eutectic boundary around the primary
phase(b) the variation of composition across the field(c) [20].

The intermetallic phases can be formed in different magnesium alloys, for example
\( \text{Mg}_{17}\text{Al}_{12} \) (\( \beta \) phase) is formed in Mg-Al binary system. If some atom sites in this ordered
phase have been occupied by wrong atoms or left vacant, then the fixed stoichiometric
combination shifts to other values [21]. The deviation from the fixed compound is
observed in the alloys that solidified in non-equilibrium condition, and the \( \beta \) phase be
replaced with another metastable compound [22]. It’s also demonstrated that the
equilibrium boundary moves from 42% aluminum towards the lower percentage of
aluminum under non-equilibrium condition. The new eutectic point is claimed to contain
around 37% [23] or 25.5% aluminum [22]. The combination effect of segregated layers
and the presence of eutectic boundary containing \( \beta \) phase with different aluminum
content leads to the discrepancies among the results which are reported for the fraction of $\beta$ phase (see Section 2.5.2).

### 2.3 An Introduction to High Pressure Die Casting (HPDC)

Figure 2.3 shows the plunger and the shot sleeve used in HPDC process. The pouring temperature is ranging from 650°C to 690°C and the die temperature ranges 200°C-240°C [18]. Figure 2.4 shows the three different stages of HPDC process, and give more clear insight about the terms which are used in Sections 2.4.3, 2.6 and 4.2.5. The first two stages consist of the slow and fast movement of the plunger, and the die cavity will be filled at the end of the second stage [24]. At the end of the second stage, there would be a slight increase in pressure (45MPa), and the velocity drops before onset of the last stage. This point is named as the impact pressure point [25]. The gate velocity usually ranges 30-50 m/s, and it reaches 100 m/s for the thin wall castings [24].

![Figure 2.3 Schematic view of the plunger and shot sleeve [24], the difference in the thermal conditions of the melt is indicated by different colors](image)

The impact point is followed by more decrease in the pressure which is called the static point (25MPa), and then the third stage is known as the intensification pressure (IP) begins. In this stage the pressure reaches much higher magnitude (80MPa) [25], and maintains constant till the ejection of the casting. The IP stage is influential in lowering the amount of porosities [26].
2.4 An Introduction to Solidification of Mg-Al Alloys

The solidification condition governs the microstructure of the magnesium alloys, and affects the grain size and fraction of non-equilibrium phases. The brief explanation about the nucleation mechanism and role of solute content in limitation of growth mechanism is provided in this section. As explained in Section 1, the die-casts can be also treated as composite materials [12]. To determine the onset of fully plastic behavior, the fraction of intermetallic phase should be estimated. The Scheil-Gulliver approach is also introduced as an analytical approach to estimate the fraction of non-equilibrium phases. Moreover, the variation of heat extraction during different stages of process is introduced to justify the bimodal microstructure of die-cast specimens.

2.4.1 Nucleation and Growth

The heterogeneous nucleation is the predominant mechanism in solidification of the castings. The difference between the energy barrier in homogeneous and heterogeneous nucleation are shown in Figure 2.5. In heterogeneous solidification, the magnitude of undercooling required to overcome the energy barrier is lower than that of homogenous
solidification. The number of nucleation sites is another influential parameter in the solidification process [27].

![Diagram](image)

\[ \Delta G^*_{het} = \Delta G^*_{hom} S(\theta) / S(\theta) \leq 1 \]

**Figure 2.5 Comparison between homogeneous and heterogeneous nucleation.** Lower undercooling is required to reach a certain nucleation rate under heterogeneous nucleation [21].

In addition to the mould walls, the intermetallics and oxides which are formed during the solidification process of the Mg alloys are other nucleation sites. The Al–Mn particles can be formed above 650°C [28], and it’s also reported that they are stable up to 700°C. Due to the early solidification process in the shot sleeve, the formation of Al–Mn compound before injection of the melt is possible and they can act as the preferred nucleation sites at high cooling rates [28,29].

The abundance of nucleation sites and constitutional supercooling zone in front of the interface result in the high nucleation rate. The boom in the nucleation rate restricts the growth of nucleants, so the final microstructure consists of the small equiaxed grains. The characteristics of supercooling zone, built up and transportation rate of the solute elements are the influential parameters in the solidification process [21]. The growth restriction which results from this phenomenon is proportional to the equilibrium
partition of alloy (Figure 2.6). The solute content will be rejected to the liquid through the solid-liquid interface, if the equilibrium partition is smaller than the unity [30]. The equilibrium partition of AM60 is nearly 0.4 [31].

![Equilibrium solidification and distribution of solute content](image)

**Figure 2.6 Equilibrium solidification and distribution of solute content, K represent the equilibrium partition of the alloy [26]**

Non-equilibrium solidification leads to the gradient of composition which is known as coring in microstructure. The segregated layers are containing higher aluminum content, and it’s indicated on mapping images of die-cast samples of AM60 [5]. These layers are more resistant against etching solutions, and after proper image processing, they will be revealed as a dark area (see microscopic images in Chapter 4 and Appendices). The width of the region containing a higher percentage of aluminum is reported to be 2-3 μm [17]. The composition of primary grains was almost equal to the equilibrium partition ratio, and a remarkable increase of up to 10% aluminum is reported for the segregated layers [23].

From the previous explanation about highly segregated layers, and composition of primary α, it can be assumed that the diffusion of solute content in the solid state is negligible. Therefore, the fraction of non-equilibrium phases can be estimated by Scheil-Gulliver relation [32], and it is explained in Section 2.4.2. (see Section 4.1.7)
2.4.2 The Fraction of Non-Equilibrium Phase

In practice, the solidification process is non-equilibrium, and the fraction of phases cannot be derived from the equilibrium phase diagram. The composition of solid and liquid state (\(C_s\) and \(C_L\)) can be defined. The material balance is \(C_s f_s + C_L f_L = C_0\) where \(f_s\) and \(f_L\) are the weight fractions of solid and liquid. As a consequence, \(C_s\) and \(C_L\) can be determined [33]:

\[
C_s = \frac{KC_0}{KF_s + (1 - F_s)} \quad \& \quad C_L = \frac{C_0}{K(1 - F_L) + F_L}
\]  

If the solidus and liquidus are considered as straight lines, the solid fraction at every temperature can be determined [34]:

\[
f_s = \frac{(T_L - T)}{(T_m - T)(1 - K)}
\]  

Where, \(T_L\) and \(T_m\) are the liquidus temperature and melting point of pure solvent, respectively. It’s assumed that there isn’t any diffusion in the solid state, so the composition of core grain would be \(KC_0\) and does not change during the solidification process (Figure 2.7). The composition of solid and liquid at the interface can be calculated using the following mass balance equation [33]:

\[
(C_L - C_s^*) df_s = (1 - f_s) dC_L
\]  

At \(f_s = 0\), \(C_s^* = KC_0\) or \(C_L = C_0\) then:

\[
C_s^* = KC_0(1 - f_s)^{(K-1)} \quad \& \quad C_L = C_0 f_L^{(K-1)}
\]  

The liquid and solid fraction also can be determined by the following equations [34]:

\[
f_L = \left[\frac{T_m - T_L}{T_m - T}\right]^{\frac{1}{1-K}} \quad \text{and} \quad f_s = 1 - \left[\frac{T_m - T_L}{T_m - T}\right]^{\frac{1}{1-K}}
\]
As a result the mean composition of the solid state ($\overline{\mathbf{C}}_s$) would be lower than that of the solid-liquid interface, and the liquid fraction would be higher than that of the equilibrium condition at any given temperature. It also increases the solidification range, and leads to the formation of grain boundary eutectic. Therefore the solidification range is [34]:

$$\Delta T_f = T_L - T_E$$  \hspace{1cm} (2.6)

Where, $T_E$ is the eutectic temperature.

### 2.4.3 Heat Resistance in Casting Process

The rate that heat is extracted from the system has a strong influence on the growth rate of the solid-liquid interface. The heat extraction is affected by different resistances which exist in various types of the casting processes, and the layer with high resistance characteristics primarily controls the solidification rate. The total solidification time in the casting process is proportional to the volume-area ratio of the casting, and the proportionality constant depends on the thermal conductivity of melt and mold. In the case of permanent mold, this constant is decreasing as the interfacial heat transfer increases [33]. The magnitude of the interfacial heat transfer coefficient (HTC) changes during the HPDC process. There is a good contact between the mould and melt at the initial stage of the process, as the solidification process continues, the casting contracts
toward the centerline. As a result, the interface gap starts to expand (Figure 2.8). The size of air gap depends on the thermal expansion of the mould and casting [35].

![Figure 2.8 The presence of micro gap on the mold wall at the initial stage of casting (a), as the process continues the gap start to expand (b)[24]](image)

It’s reported that the highest magnitude of HTC is obtained at impact pressure point, and then it reduces to the lower magnitude in the IP stage ranging from 90 to 112 $kWm^{-2}K^{-1}$ for AZ91 [25]. The poor contact during the casting process of thin sections lead to the sharp decrease of HTC. The decreasing trend of the HTC was longer for thick sections [25,36]. The cooling rate is estimated to be more than 1000°C/S close to the die wall for the complex die cast of AM60, and it’s estimated to be about 55-222°C/s in the central region [37,8]. A cooling rate of 50°C/s is reported for the permanent mould, and a cooling rate of 106°C/s is achievable using the centrifugal casting and copper mould [17].

### 2.5 Microstructure of Mg-Al Alloys

The formation of three different phases under non-equilibrium condition is explained in previous sections. The morphology of microstructural features and the defects associated with HPDC process are discussed in this section. The criteria which are applied for determination of skin thickness are introduced in this section (see Section 4.1.2). The expression which correlates the cooling rate and size of the grains are also presented in this section. The effect of solute content on the size of the grains, and ESGs is explained. The image processing technique which is applied for characterization of intermetallic phase (see Section 4.1.7) is introduced in this section. The criteria and functions which were applied by different researchers to characterize the porosities are explained. It’s also
mentioned that the process and location of the castings should be taken into account. (see Section 4.2.5). The deformation of mush zone and the formation of defect bands are also described in this section. The defects which are resulted from the deformation of mush zone are described in this section, and similar results are obtained in the current study (see Section 4.2.5).

2.5.1 Grain Size Distribution - Skin & Core

The variations of solidification condition, HTC, nucleation rate as well as restricted growth result in the formation of an outer layer containing fine equiaxed grains. This layer is named skin, and the central region with coarse grains named core [38,39]. The skin thickness measurement can be based on the grain size variation, eutectic percentage as well as hardness profile. By selection of an appropriate threshold, the intercepted points with the general trend of variation can be considered as skin-core boundary [4]. The skin fraction is greater in thin samples, and higher degree of bimodality is expected for the thick samples [40].

The cooling rate affects the skin thickness, and the size of equiaxed grains. The slight decreasing trend of grain size is reported up to the cooling rate of -240°C/s. The steep reduction in grain size is reported at -280°C/s [28]. The correlation between the grain size diameter and cooling rate is estimated by following equation [8].

\[ d = 59R^{-0.303} \]  \hspace{1cm} (2.7)

Where \( d \) is the grain size diameter in micron and \( R \) is the cooling rate. The lower cooling rate is expected as the thickness of the casting is increasing. The grain size diameter ranges from 13 to 15 \( \mu m \) within the thin die cast specimens of AZ91 (\( t = 2.7 \text{mm} \)), whereas it is between 21 to 54 \( \mu m \) for the thick die cast specimens of AZ91 (\( t = 6.7 \text{mm} \)) [40]. The grain size measurement of a complex die cast of AM60 indicated that the average grain size changes from 5 to 9 \( \mu m \) within the skin and it was between 11 to 16 \( \mu m \) within the core region [4].
The solute content is another influential parameter on the size and morphology of grains. The predominant morphology of the Mg-Al alloys containing more than 3% aluminum were equiaxed grains, whereas the column shaped grains are observed in dilute alloys [39]. The uniform distribution of grains was reported for the dilute magnesium alloys, and the grain size diameter decreased from 15 to 7 µm in the skin region as the aluminum content increased from 0.47 to 6%. While the lower decreasing trend in grain size only continued up to 4% aluminum in the core region, and the grain size decreased from 17 to 12 µm [41]. The electron back scattered technique revealed that the average grain size diameter ranges from 5 to 8 µm for the AM60 specimens with 5 mm thickness [42]. The microstructure of the HPDC magnesium alloys is shown in Figure 2.9.

![Microstructure of the HPDC magnesium alloys](image)

**Figure 2.9** Microstructure of: (a) the intermetallic phase revealed dark using the hydrofluoric acid as etchant for AM60 sample; (b) the variation of the grain size across the cross section of the HPDC specimens of AM50 [18]

### 2.5.2 Intermetallic Phase

The β phase appears as divorced eutectic component in the grain boundary of die cast magnesium alloys containing more than 2% aluminum [19]. The variation in the composition of β phase, and high segregation can lead to the discrepancies among the results which are reported for the fraction of β phase. For instance, the average percentage of β phase in the die-cast specimens of AZ91 is reported to range from 7% to 13% in the skin and the core region, and there is a slight decrease within the thick
The morphology of $\beta$ phase particles is more dispersed in the skin of die cast AM60 alloy, and the percentage of $\beta$ phase is estimated to be 10% in the core region [43]. The image processing technique was utilized to differentiate between the $\beta$ phase and secondary $\alpha$ phase. It’s reported that the area fraction of the $\beta$ phase was relatively constant across the section and it was almost 1.7% for die cast sample of AZ91. A minor clustering tendency and finer particles were observed in the skin region [44]. The mass fraction of $\beta$ phase was estimated by Scheil-Gulliver approach and it was about 4.2% for AM60 [32]. The decrease of interconnection between $\beta$ phase particles in the core region was attributed to the presence of the coarse grains in the core region of the specimens [45].

2.5.3 Externally Solidified Grains (ESG)

The partial solidification of the melt in the shot sleeve leads to the formation of ESGs, and a mixture of ESGs and melt will be injected into the die. The sections which are located far away from the gate system have a lower fraction of the ESGs, while a higher fraction of ESGs is found in close-to-ingate regions. It’s reported that the volume fraction of ESG is about 5% in last to fill region and it increases up to 25% within the close-to-ingate region [3]. The thermal condition of shot sleeve, melt temperature and holding time of melt in the shot sleeve affects the fraction of ESGs in the close-to-ingate regions [46]. Moreover, the alloy composition is a key parameter in the formation of ESGs, and also affects the magnitude of ESGs which are formed in the shot sleeve. The fraction of ESGs for Mg-alloys was estimated to reach up to 20% [31]. The fraction of ESGs is increasing as the percentage of aluminum content increases from 2 to 5.5%, and leads to the increase of average grain size in the core region. The abrupt increase in the fraction and size of ESGs is followed by decreasing trend in the alloys containing higher amount of aluminum. The high GRF of the latter alloys constrains the formation and growth of ESGs [41]. The ESGs’ diameter can reach 50 $\mu$m [1] or even 68 $\mu$m [42].
2.5.4 Casting Defects

The porosities are well known defects among the castings, and can be classified into two main groups. The first group is gas pores including the entrained or entrapped gas pores. Moreover, dissolved gas in the molten metal is another source of the porosities in castings. The entrained gas pores are formed in the filling process because of the turbulence, and often are formed in the knit line region where two or more melt streams join each other [47]. The second group of pores is the shrinkage pores, which are formed because of the contraction at freezing point. The volume fraction of shrinkage for the magnesium alloy is about 4.2% at melting point and nearly 5% in solid state [18]. It’s reported that the formation of shrinkage porosities can be facilitated by the formation of entrapped gas pores because of the variation in the thermal conductivity [48].

The shapes of the gas pores are nearly round and the shrinkages are almost similar to the small cracks (Figure 2.10). The aspect ratio was considered as a reasonable feature for separation of gas and shrinkage pores of AM50 samples [49]. However, the processing condition and location of the porosities should be taken into account for the separation of pores. The formation of large pores within the regions close to the gate system is mainly originated from the deformation of the mush, and they can be classified as shrinkage pores [26].

Figure 2.10 Round shaped gas pore (a) and shrinkage porosities (b) in a die cast specimen of magnesium alloys [6]
The tendency of porosities for clustering can be quantified by introduction of clustering parameter. It’s the ratio of the distance between the pores to the number of the pores [50,51].

The area fraction of porosities within die–cast specimens of AM60 reaches the value of 6.7% and 6.5% [50,51]. The area fraction of large and small gas pores were 3.1% and 0.9%. The average size of shrinkage pores was 10.5 $\mu m$ and the size of gas pores ranged from 36.9 to 95.4 $\mu m$[51]. The largest pore volume was 0.55 $mm^3$, and the volumetric porosities ranged from 0.1 to 1.3% [50] The higher amounts of porosities were detected in the region close the end of the flow path [52]. It’s demonstrated that the fracture plane contains the largest pore and highest area fraction [50].

### 2.5.5 Formation of Defect Bands

The simultaneous movement and solidification of the melt into the cavity, and early solidification of the melt in the shot sleeve can lead to the highly localized deformation in some regions and formation of defect bands. The strength of melt is increasing as the fraction of solid state is increasing during the solidification process [53,54]. After a gradual increase of strength in the initial stages of solidification, there would be an abrupt increase in strength. This is the point which introduced as the maximum packing point [55]. The formation and interlocking of the solid particles are the influential parameters in the transportation of mush during the shearing process [54]. The maximum packing point shifts toward lower fraction of solid state as the morphology of solid particles is getting larger and more dendritic [55]. The resistance against the deformation of the mushy zone containing globular solid particles is lower [56]. It’s demonstrated that the inefficiency of the feeding process and severe deformation at a solid fraction higher than the maximum packing point leads to the formation of the porosities in the slip plane along the die walls. The application of pressure on the immobile skin of the die-cast leads to the concentration of large solid particles in the core region [55]. The characteristics and immobility of the skin is also an effective parameter on the distribution of grains within the die cast specimens [46]. The liquid phase will be squeezed out towards the
skins contains higher amount of the aluminum content, and the aluminum content is higher within the bands [57] and they will be revealed as a dark area along the skin-core boundary in microscopic pictures.

2.6 Effect of Processing Parameters

The HPDC process consists of different stages (Section 2.4), the effect of processing parameters on the microstructure and solidification of die cast products is explained in this section. Although the manipulation of processing parameters is not applicable to the current research, there is an outstanding resemblance between the new results and the experiments which are conducted by other researchers (see Section 4.2.5, 4.1.4).

As the velocity of plunger affects the holding time of the melt in the shot sleeve, the lower velocity of plunger results in more heat loss of melt [25]. Although it’s reported that the IP stage has a negligible effect on HTC [25, 52], the high solid fraction in the shot sleeve hinders the proper transmission of the pressure, so it’s followed with a decrease in the value of HTC [25]. The pronounced effect of the air gaps on the decreasing trend of HTC is reported, as a low magnitude of pressure is applied during the IP stage [37]. The decreasing trend of the HTC value with increasing initial temperature of the die is reported by different researchers [36, 37]. Although it’s reported that there isn’t a clear relationship between the gate velocity and the volumetric porosities [52], the average fraction of porosities is doubled as the gate velocity increased from 33 to 55 m/s. It’s attributed to the higher turbulence of flow [49].

The IP stage assists the feeding process and provides sufficient liquid to compensate the contraction. The compression of air bubbles occurs in this stage, so it has a strong influence on the amount of porosities [26, 49]. The fraction of large gas pores significantly decreases as the IP stage is applied for die cast specimens of AM50. Although the percentage of shrinkage pores remained constant, the total percentage of the porosities was 1.5 times greater in the samples without IP stage [52]. The application of IP stage on the mush containing high solid fraction resulted in the large pores and crack-like features. The main reason is the lack of enough melt for the assistance of deformation process, and
compensation of the contraction [38]. As the melt temperature is increased the percentage of the porosities is increased. It is attributed to the more dissolved gas in the melt [49].

Moreover, the die temperature and initial solid fraction affect the distribution of the ESGs. The ESGs were densely located in the center of the specimens as the initial solid fraction is lower than 7%, and the die temperature is below 200°C. The spread pattern is obtained as the initial solid fraction is increased up to 22%, and the die temperature is increased up to 350°C. The formation of the firm skin will be hindered as the die temperature is increased. The possibility of flow arrest is increasing, as the initial solid fraction is more than the coherency point [46].

2.7 Microstructure-Yield Strength of Mg-Alloys

The relationship between microstructure and yield strength are described in this section. The variation of the strain hardening is explained with the assistance of $\sigma\theta$ plots. The application of the $\sigma\theta$ plots for determination of skin fraction of different magnesium alloys is described. Regarding the diversity of grain size within different location of a complex die-cast of AM60, the same approach will be used to correlate variation of strain hardening rate and microstructure of the complex die-cast (see Section 4.1.8). The modified form of the Hall-Petch equation is introduced to correlate the local distribution of the grain size to the yield strength of the samples (see Section 4.1.4). A brief explanation about the hardening mechanism and hardness of magnesium alloys is provided.

2.7.1 Variation of Strain-Hardening Rate & $\sigma\theta$ Plots

The stress-strain curve of the materials under tensile test can be divided into different stages. Two stages are usually can be identified for polycrystals. The variation of strain-hardening rate ($\theta$) during these two stages is explained in this section. The pile-up of dislocations behind obstacles takes place in the second stage, and it's an athermal mechanism. The relationship between the flow stress ($\sigma$) and the density of dislocations ($\rho$) is [58]:

$$\sigma = C\sqrt{\rho}$$
\[ \sigma \propto \alpha G b \rho^{1/2} \]  

(2.8)

where \( G \) is the shear modulus. The pile-up rate of dislocations is related to the average free distance of dislocation. It’s established that strain-hardening rate is constant during the second stage and is proportional to the shear modulus [59].

\[ \theta = \beta G \]  

(2.9)

The proportionality constant is \( \beta \). The variation of strain-hardening rate is thoroughly investigated by Kock and Mecking. They asserted that the strain-hardening rate is athermal at low strains (stage II), and the initial strain-hardening rate \( (\theta_0) \) is always a fraction of shear modulus. The deviation from the initial strain-hardening rate in third stage can be quantified by adding recovery terms to the pile up rate of dislocations [60].

\[ \frac{d\rho}{d\gamma} = \frac{1}{Lb} - R \Rightarrow \theta = \theta_0 - \theta_r(T, \dot{\gamma}) \]  

(2.10)

The second term \( (\theta_r) \) is representing the length of dislocations which become ineffective during the recovery process and rearrangement of dislocations.

To characterize the strain hardening behavior of polycrystals, the effect of orientation and average size of the grains should be taken into account. The effect of grain orientation is quantified by Taylor factor. The mean Taylor factor is 3 for FCC (Face Centered Cubic) materials and nearly 4.5 for magnesium [58, 61]. The grain boundaries act as obstacles, and the pile-up of dislocations takes place at grain boundaries. The effect of grain size on the pile-up rate can be considered as a similar term of average free distance.

The variation of \( \alpha \theta \) with respect to \( \sigma \) is investigated by Kocks and Mecking, and the generated curves are called \( \alpha \theta \) plots. The \( \alpha \theta \) curves moved upward while the average grain size decreased by a factor of 3[58]. The steep slope in the third stage is observed for the silver sample while the average grain size is increased [62]. The decreasing trend of grain size and its effect on \( \alpha \theta \) plots of pure magnesium is also studied by Caceres et al. [61]. The \( \alpha \theta \) were also formed for the die-cast magnesium containing different amount
of aluminum (Figure 2.11). The high solute content resulted in higher initial slope, and also the concentrated alloys displayed higher upward shift [12]. The excessive initial slope which is generated from insertion of hard particles into the soft matrix can be quantified by following equation [63]:

$$\frac{d\sigma}{d\epsilon}_{p} = f_{p}E_{p}$$  \hspace{1cm} (2.11)

Where, $f_{p}$ and $E_{p}$ are the volume fraction and Young’s modulus of the dispersed particles, respectively. It’s demonstrated that the difference in an upward shift of magnesium alloys is related to the different fraction of elastic material (Skin), and they can be treated as a composite material to determine the fraction of skin and core [12]. The strain hardening-rate of the core region is considered to 1.4 \text{ GPa} [5] and it was similar to diluted alloys.

![Figure 2.11 \(\tau\theta\) plots for different magnesium alloys containing different amount of aluminum (A), volume fraction of elastic materials versus the stress (B) [12]](image)

### 2.7.2 Yield Strength of Mg Alloys & Hall- Petch Equation

The strength mechanism of the die-cast specimens originated from solid solution, dispersion and grain boundary hardening mechanisms. The difference between the solid
solution strength of skin and core is estimated to be 12MPa, and the difference is as small as 8MPa as the dispersion strengthening mechanism is considered [32]. Due to the interconnection between the \( \beta \) particles [45], the effect of dispersed particles can be considered equal to 17MPa. The main contribution in strengthening mechanism is generated from the grain boundaries [32].

The high flow stress of the skin region is attributed to the fine grain and higher percentage of the intermetallic phase in the skin region. The strain-hardening coefficient of 0.23 [5] and 0.22 [64] is obtained for the skin region of the die cast specimens of AM60, whereas it’s 0.185 [5] and 0.206 [64] within the core region of the die cast sample of AM60 [5]. The yield strength of the core region is estimated 120MPa, and it’s 130 MPa in the skin region [64].

Although the effect of the grain size on yield strength can be evaluated by Hall-Petch equation, the experimental yield strength was lower for the samples with high diversity in grain size. It’s demonstrated the samples can be treated as composite, and the fraction of grains can be considered as modifying parameter in Hall-Petch equation [65,7]. The fraction of each field in the modified form of Hall-Petch equation is represented by \( f_i \).

\[
\sigma_y = \sum f_i (\sigma_0 + Kd^{-\frac{1}{2}}) \tag{2.12}
\]

It’s demonstrated that \( \sigma \) is constant for most magnesium alloys and independent of aluminum concentration, whereas the Hall-Petch slope \( K \) is increasing from 220 to 300 MPa.m\(^{0.5}\) as the aluminum content is increasing from 0 to 11.7% [32]. The mean yield strength was reported to be 106 MPa for die cast specimens of AM60, and a higher deviation was observed in the samples containing a higher percentage of the shrinkage porosities [7]. The maximum yield strength of AM60 specimens was 126MPa and it reduced to 115Mpa as the area fraction of the porosities increased to more than 10% [66]. The Hall-Petch parameters are determined by different researchers and results are presented in Table 2.2.
The discrepancies among the results of hardness tests of die-cast specimens of AZ91 and AM60 are reported. The localized distribution of grains and β phase were the main reasons for inconsistent results. The abrupt decrease is attributed to the presence of micro porosities [68,1]. The wider core region and low hardness are reported for thick samples [68]. The micro indentation technique is utilized to determine the skin thickness. It’s reported that the hardness of large dendrites was half of the average value of hardness [4].

<table>
<thead>
<tr>
<th>Material</th>
<th>( K(Mpa,\mu m^{1/2}) )</th>
<th>( \sigma_0(MPa) )</th>
</tr>
</thead>
<tbody>
<tr>
<td>AM60-complex die cast[67]</td>
<td>297</td>
<td>24</td>
</tr>
<tr>
<td>AZ91- die cast [66]</td>
<td>252.3</td>
<td>58.2</td>
</tr>
<tr>
<td>AZ91- Gravity Cast[66]</td>
<td>167.2</td>
<td>74.2</td>
</tr>
<tr>
<td>AM60 - Gravity Cast[6]</td>
<td>152.9</td>
<td>62.1</td>
</tr>
<tr>
<td>AZ91- Gravity Cast[6]</td>
<td>142.2</td>
<td>73</td>
</tr>
<tr>
<td>AM60[32]</td>
<td>280</td>
<td>12</td>
</tr>
</tbody>
</table>

The reduction of yield strength was reported for die-cast of AM60 without skin region [3]. As the aluminum content is increasing in the Mg-Al alloys, the skin fraction is increased, and the elasto-plastic transition point is increased as well [12]. A reduction of yield strength from 160MPa to 120MPa is reported for die-cast of AZ91, as the thickness increased from 2mm to 10mm [68]. The sensitivity of yield strength to the sample thickness is reduced for the samples with thickness higher than 3.9mm. The higher flow
stress in thin samples was attributed to the high solidification rate and finer microstructure [69].

2.8 Internal Defects-Fracture Strength of Mg-Alloys

The presence of the internal porosities puts some constraints on extensive usage of the HPDC products [70]. Introduction of void leads to the formation of non-uniform stress distribution. In addition to the reduction of load bearing capacity, micro voids act as the stress riser regions in the homogeneous structure. The effect of porosities on the fracture behaviour is described, and the failure models for the prediction of fracture properties are also introduced in this section.

Although it’s demonstrated that there isn’t a clear relationship between the volumetric porosities and fracture properties of die casts [71,2], it’s claimed that there is an empirical equation for the sample containing large pores, and the fracture strain decreased from 16 to 3% as the volume fraction of the porosities increased from 0.8 to 3.5% [40]. The parabolic trend in reduction of fracture strain against the area fraction of the porosities ( \( f \) ) was reported by Surappa et al [71]. The fracture plane is reported as the plane containing the highest area fraction of porosities and the largest pore [50]. The elongation of die-cast AE44 was nearly 12 % at \( f =0.01 \) and it decreased to 4.5% as the \( f \) increased up to 3.5% [72]. The die-cast specimens of AM50 and AM60 were modeled by finite element methods. And the elongation of models containing pores with a large diameter of 40 \( \mu \text{m} \) was almost one-third of the samples containing small pores with the same volume fraction of porosities. The large pores are introduced as the preferred sites for crack initiation [73,74]. Although the void opening was negligible in the elastic region, it increases linearly during plastic deformation [74].

2.8.1 Failure Models

The visual schematic for the presence of a spherical pore is shown in Figure 2.12. The flow behavior of most metals can be quantified by the power law equation \( \sigma = K\varepsilon^n \). Where, \( \sigma \) and \( \varepsilon \) are the true stress and strain, \( K \) is a constant and \( n \) is the strain hardening exponent. By considering the force balance between the homogeneous plane
\( A_h \), and the plane containing pores, \( A_i \), the following relationship between \( \varepsilon_{h} \) as a function of \( \varepsilon \) can be obtained [75]:

\[
(\varepsilon_i)^n (1 - f) \exp(-\varepsilon_i) = (\varepsilon_h)^n \exp(-\varepsilon_h)
\]  
(2.13)

The equation is known as the **constitutive model**, and can be solved for given values of \( f \) and \( n \), to predict the fracture strain of the sample. To solve this equation, it’s necessary to define the upper bound of the critical strain in the plot. The tensile tests should be conducted on a large number of tensile specimens to determine the highest fracture strain.

![Figure 2.12 Material containing spherical void [80]](image)

This approach has been taken by different researchers to predict the elongation of the samples containing porosities [50, 76, 77, 78]. The fracture strain of as-cast samples of aluminum samples was predicted with reasonable accuracy, the higher deviation is observed for the samples containing higher area fraction of porosities [76]. The prediction of fracture strain for die-cast of AM60 indicated that the dependency to \( f \) is also pronounced as the strain hardening and grain size are increasing [78]. It’s also reported that the definition of instability point is not supported by constitutive model. The area fraction of eutectic phase is subtracted from the load bearing area for die-cast
aluminum, and location of the pore and $n$ implemented into the model for prediction. A deviation of less than 2% is obtained [79].

In addition to the some discrepancies among the results, the model was more accurate for the centered pores. The variation of stress concentration due to the proximity of the pore to the edge of the sample is considered by Weiler et al [10,11]. The localized stress around the spherical void is also taken into account, and stress concentration is implemented into the new model, which is called Modified Model. The relationship between the stress concentration around a spherical void and the area fraction of void in the infinite body is determined [80]. The effect of plastic deformation on the stress concentration factor of a plate containing a hole was investigated by Stowell. He mentioned that as the amount of plastic deformation is increasing, the elastic stress concentration factor decreases and the strain concentration factor is increasing significantly [81]. The plastic stress and strain concentration can be characterized by secant modulus at the edge of the hole and secant modulus at the region far from the void [81,82]. It’s demonstrated that the instability point occurred at the cross section containing imperfections. Due to the stress concentration around the void, the strain only reaches a fraction of the ideal magnitude at the net cross section [10]:

$$n^n (1 - f) \exp(-n) = (K_c \cdot \varepsilon_h)^n \exp(K_c \cdot \varepsilon_h)$$  \hspace{1cm} (2.14)

$$K_c = 1 + 2.51 \cdot e^{1.65f} = \frac{n^{n-1}}{(\varepsilon_h)^{n-1}}$$ \hspace{1cm} (2.15)

The effect of an eccentric imperfection within the die-cast of AM60 was investigated and the equivalent pore was considered as the representative of the area fraction of porosities. The distance of the void from the edge of the sample is characterized by the implementation of the ligament fraction ($\delta_L$) into the failure model [11].

$$\delta_L = \frac{d_i - r}{t}$$  \hspace{1cm} (2.16)
Where, \( d_t \) is the distance of the centre of the equivalent pore to the edge of the sample, \( r \) is the radius of the equivalent pore and \( t \) is the thickness of the specimen. The centered pore has the highest value of the ligament fraction. The smallest ligament fraction in die-cast samples occurs once the equivalent pore is located in the vicinity of the skin. The modeling of die-cast AM60 is conducted with the assistance of the finite element method, and the following expression was introduced as the **Failure Model** for prediction of fracture behavior [11].

\[
\varepsilon_f = 0.016 f^{-0.3} \left[ 0.18 \ln(\delta_L) + 1.17 \right] \tag{2.17}
\]

Respectively the fracture stress can be predicted by the following expression [11]:

\[
\sigma_f = 146 f^{-0.1} \left[ 0.06 \ln(\delta_L) + 1.05 \right] \tag{2.18}
\]

The cross section of tensile coupons is rectangular, so the proximity to the edge of the samples can be defined for both sides. Therefore the model can be rewritten in the general form and the ligament fraction can be replaced with the ligament area fraction (\( \delta_A \)) [11].

\[
\delta_A = \frac{(d_t-r)(d_w-r)}{A_0} \tag{2.19}
\]

Where, \( d_t \) is the distance between the centre of the equivalent pore and the edge of the sample, \( d_w \) is the distance from the width of the sample and \( A_0 \) is the initial cross section area. The following expressions describe the general form of the failure model, and fracture strain and stress can be determined using the following equations:

\[
\varepsilon_f = 0.016 f^{-0.3} \left[ 0.17 \ln(\delta_A) + 1.26 \right] \tag{2.20}
\]

\[
\sigma_f = 146 f^{-0.1} \left[ 0.04 \ln(\delta_A) + 1.07 \right] \tag{2.21}
\]

The failure model is applied for prediction of ductility for both die cast of AM60 and aluminum A356, the deviation from the experimental values were less than 7% [11,79].
Chapter 3

3 Materials & Experimental Works:

The experimental methods and appropriate procedures which are taken to examine the mechanical properties are introduced in this chapter. Also the brief explanation is provided to express the difference between the gate systems of the castings provided by the Meridian Technologies Inc. Due to the confidentiality of engineering drawings and detailed simulation, other references are to be used to provide a necessary visual picture. The two mainstreams of the current investigation are the characterization and prediction of the mechanical properties of the HPDC magnesium alloys. The experimental procedures are designed in a way to provide the general view of the mechanical and microstructural properties, and aims to provide input information for the recent failure model.

Based on the design and location of the gate system of the castings, different regions are selected for the current investigations. The different regions are named and characterized. The naming procedure of different regions is mainly borrowed from the reference [1]. The Archimedes’ method is used to determine the volumetric porosity and provide a general view about the spatial distribution of the porosities across the casting. To characterize the local mechanical properties of the casting tensile test is performed on the specimens cut from different locations. The metallographic studies are performed on the samples cut from the gauge length of the tensile specimens in order to explore the relationship between the local microstructure and mechanical properties. To examine the fraction and distribution of the porosities before tensile test, the metallographic specimens are extracted from the locations along the flow path. The hardness test is performed to assess the mechanical properties of the skin and core region. The microscopic studies are also performed on the fracture surfaces of the tensile specimens.
3.1 Materials and Characterization of Locations

The mechanical and microscopic experiments are conducted on two different AM60B castings provided by Meridian Technologies Inc; located in Strathroy, Ontario. The weight percentages of the alloying elements in AM60B are shown in Table 3.1.

<table>
<thead>
<tr>
<th></th>
<th>Al</th>
<th>Mn</th>
<th>Si</th>
<th>Zn</th>
<th>Fe</th>
<th>Cu</th>
<th>Ni</th>
<th>Mg</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>5.5-6.5</td>
<td>0.25-0.6</td>
<td>&lt;0.1</td>
<td>&lt;0.22</td>
<td>&lt;0.05</td>
<td>&lt;0.01</td>
<td>&lt;0.002</td>
<td>Balance</td>
</tr>
</tbody>
</table>

Table 3.1 Composition of AM60B [6]

A bin including 7 identical castings named bolster and a new instrument panel beam are the materials studied in this research. (Figure 3.1 and Figure 3.2)

Figure 3.1 The magnesium casting – bolster used in this study

Figure 3.2 The magnesium casting – new instrument panel beam
The engineering drawing provided by Meridian assists us to anticipate the filling pattern of the casting. In addition to the solidification condition and fill pattern which are important parameters to select the location of tensile specimen, the position of the overflows are considered in the selection of the locations.

Generally gate system for HPDC process is designed in a way that the melt stream flows via the minimum length across the die cavity, and the stream starts from the ingate and the overflows are located at the end of the flow path [70]. It’s noticed that there is a difference in the design of the gate systems between the bolster and the new instrument panel beam. The in-gate system of the instrument panel beam was similar to the previous casting investigated by our group [1]. The fan runners are designed as the gate system for the bolster, and the wider fill pattern is achievable as the melt progress into the mould cavity. The control over the flow melt can be conducted more efficiently as the conventional runners are designed for the die cast. The higher velocity is achievable by conventional runners [70]. The conventional and fan runners are shown in Figure 3.3.

**Figure 3.3 Diagrams showing runner systems:** (a) the overflows, vents as well as fan type runners in the bolster, (b) the green dashed ellipse showing the conventional runners of the new instrument panel.

The knit line regions are the junctions of two or more molten flows. According to the studies conducted by other researchers the turbulence of molten flow leads to higher
amount of porosity and oxide entrapment in this region [47]. Therefore, there would be a higher possibility that these locations are prone to the unexpected failure during the service life. The overflows are other sections of the gate systems connected to the vents, and collect oxidized front of molten flow, and it can also act as heat storage near thin parts of the casting [70].

To evaluate the local mechanical and microstructural properties of the bolster, fifteen locations are selected. The different knit line regions can be identified in the bolster with respect to the gate system, and classified into two different groups. The first group is the knit line region with three overflows nearby and the tensile specimens cut from this knit line are named locations 8 and 7. The other group has only two overflows in a distance of 10 cm, and the tensile specimens cut from this region are named locations 4, 5, 10 and 11. Figure 3.4 shows the location of the tensile specimens cut from the knit line locations, one of which is closer to the overflow and the other one is closer to the gate system in a distance of nearly 10 cm. Locations 10, 7 and 4 are close to the overflow position, and locations 8, 5 and 11 are close to the gate system.

![Figure 3.4 Locations of tensile specimens which are extracted from the knit line regions](image)

The regions that are away from the gate system are named last-to-fill regions. Two groups of the latter regions are recognized in the bolster. The distance from the first
group of the last-to-fill region to the closest in-gate is 40 cm. The tensile specimens 3 and 14 are cut from this region and they are the specimens with the lowest thickness of 2mm among the other specimens (Figure 3.5).

![Image](image-url)

**Figure 3.5 The locations of different regions which are considered for extraction of tensile and metallographic specimens**

The second group of the last-to-fill regions are located at the two ends of the casting, named locations 1 and 15 (Figure 3.6). The distance from this region to the closest in-gate system is almost 26 cm. The thickness of this group is about 2.2mm which is higher than the first group.

![Image](image-url)

**Figure 3.6 The location of the last-to-fill region (1 and 15) at two ends of the bolster**
Two tensile specimens are cut from the regions close to the in-gate system at a distance of 2cm from the flange gate and they are locations 6 and 9 (Figure 3.7).

Figure 3.7 The regions which are located in the vicinity of the flange gate and are the first regions that receive the melt

Three samples are extracted from the leg-shape section of the casting, namely locations 2, 12 and 13 (Figure 3.5). In addition to the regions which are selected for tensile test, a numerous sections along the flow path are prepared for microscopic investigations (Figure 3.5-JK and Jll regions).

Figure 3.8 shows the sections extracted from the JK and Jll regions. The white dashed ellipse shows the locations of the flange gates in JK and Jll sections, and they are identified as weak points according to the tests conducted by Meridian. Therefore, they are subjected to further investigation. Based on the location of flange gates, two different groups can be realized. White dashed rectangles in Figure 3.8 show the locations of the flange gate. JK samples are cut from the knit line with three overflows nearby (middle knit line), and Jll sections are extracted from the knit line with two overflows nearby (side knit line). For instance the samples cut from the side knit lines (Jll) covers the tensile specimens which are extracted from Locations 10 and 11. The JK and Jll samples are including the ribs and screw plug shapes. The in-plane sections between the JK and Jll are also prepared for microscopic studies. The main purpose of the microscopic studies from flange gate to the overflow is to characterize the porosities along the flow path, and the microscopic analyses of these sections provide better insight about the porosities before tensile test.
Figure 3.8 JK and Jll sections start from the flange gates and end up to the vicinity of the overflow positions.

The locations of new instrument panel beam are assigned, tested and simulated by Meridian Inc. Figure 3.9 shows the different sections of the new instrument panel used to prepare the metallographic samples. The fluid flow simulation of casting process shows that the locations 2, 3 and 4 are the last-to-fill regions, with thicknesses of 2.2, 2.7 and 2.8mm, respectively. Locations 1 and 5 with a thickness of 4.7 and 4.4 mm are classified as the last-to-solidify region. Location 6 is the thinnest amongst all locations of the new instrument panel beam, and special characteristic is not assigned to this location. These locations are considered for the microstructure investigations.
Figure 3.9 Six different locations of the new instrument panel beam which are selected for microscopic studies

Table 3.2 shows the codes which are assigned for different specimens. For instance specimen Cs3L5 is the sample cut from the fifth location of the third casting named bolster.

Table 3.2 Coding system of the samples

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Casting Name – Number</th>
<th>Location</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cs3L5</td>
<td>Bolster- Third Casting</td>
<td>5</td>
</tr>
<tr>
<td>CsNL2</td>
<td>New Instrument Panel</td>
<td>2</td>
</tr>
<tr>
<td>JK3</td>
<td>Bolster – Flange Gate- Middle Knit Line</td>
<td>3</td>
</tr>
<tr>
<td>Jll2</td>
<td>Bolster – Flange Gate – Side Knit Line</td>
<td>2</td>
</tr>
</tbody>
</table>
3.2 Density Measurement

Although the relationship between the volumetric porosity and fracture strain [2,71], it’s claimed that the decreasing trend of fracture strain with volumetric porosity would be tangible if the sample contains large pores[40]. Therefore, the volumetric porosities of the tensile specimens of bolster are determined before tensile test. The Archimedes’ method is applied to determine the density and the volumetric porosity. The schematic images of apparatus used in this experiment are shown in Figure 3.10.

![Schematic view of apparatus](image)

**Figure 3.10** The schematic view of balance and the methods used to hold the specimen in the liquid (ASTM B311-08)

The nominal density of AM60 is 1.8 g/cm³. The procedure is based on ASTM B311-08 and the following equation is used to calculate the density of tensile specimens:

\[
D = \frac{(A \times E)}{(A - B + C)}
\]  

(3.1)

D= density of tensile specimen, g/cm³,  
A= mass of test specimen in air, g,  
B= apparent mass of test specimen and specimen support in water, g,  
C= mass of specimen support immersed in water, g,  
E= density of water in g/cm³.

3.3 Tensile Test

The tensile tests are performed to determine the local ductility of samples which are extracted from different regions of the casting. The tensile specimens were cut according
to the specific dimensions mentioned in ASTM standard B557-M as shown in Figure 3.11.

![Tensile Specimen Dimensions](image)

**Figure 3.11 Dimensions of the tensile specimen in mm, and \( t \) is the sample thickness**

The tensile test was performed using an Instron-8804 machine and a load cell of 5kN. The constant crosshead speed of 1mm/min is applied and tests are performed at room temperature. The initial dimension of each specimen was measured by caliper to find the average cross section area. The strain was measured continuously using an extensometer with a gauge length of 25 mm.

### 3.4 Vickers Hardness Test

The hardness test is utilized for die-cast specimens of Mg alloys to determine the skin thickness[4], and explore the relationship between yield strength and hardness of specimens with different thickness[32]. As the thicknesses of the samples which are extracted from the bolster are relatively identical, the hardness test is applied to compare the hardness of the skin and the core region of samples. The Vickers hardness tests are performed to compare the hardness of the skin and the core region of samples cut from different regions of bolster. The samples are polished and mounted for hardness test. A square based diamond is used as an indenter. The Vickers hardness numbers are determined by dividing the applied load to the surface area of indentation.

\[
VHN = \frac{1.854P}{L^2} \tag{3.2}
\]

Where, \( P \) is the applied load in \( kg \), and \( L \) is the average length of diagonal in \( mm \). The image analysis software is applied to measure the area of indentation mark and length of
the diagonal. Also the distance from the indentation marks to the edge of the samples and other indentation marks were more than $2.5d$ to avoid the effect of other test marks, as recommended in the ASTM E348-11.

3.5 Metallography

The main purposes to perform metallographic experiments are listed below.

- Determination of the area fraction of the porosities which are located beneath the fracture surface of polished specimens

- Determination of skin fraction in different locations of the castings, via measuring grain size procedure and consideration of other criteria

- Providing the polished specimens for Vickers test.

The metallographic procedure is followed according to the manual provided by ASM-Metallography and Microstructure of the Magnesium alloys. The samples are sectioned using low speed diamond saw (Figure 3.12).

![Low speed diamond saw](image)

**Figure 3.12** Low speed diamond saw which is utilized to prepare in-plane sections parallel to the gauge length and vertical sections are perpendicular to the tensile load

To determine the area fraction of porosities the fracture surfaces of samples are removed and the plane below the fracture surface is considered for analyses. The in-plane sections are also prepared to determine the spatial distribution of the porosities. Our results indicate that this is an appropriate way for the prediction of local ductility (see Section 4.2.4). The 320-1200 grit SIC papers are used in grinding steps, and then the polishing
procedure is performed in two steps. The suspension mixture of water and 1 \( \mu m \) alumina is used in the first step, and 0.05 \( \mu m \) used in the second step. Due to the high reactivity of the HPDC magnesium alloys, the polishing procedure can be a challenging step in preparation of the samples. Therefore, the polishing time should be shortened, or the polishing procedure should be accompanied frequently with rinsing by alcohol and water to prevent the formation of the oxide film on the samples. Also a mixture of alcohol and 25% glycerol is used instead of water for preparation of polishing liquid. As-Polished samples are used for the characterization of the porosities. To detect the microstructureal features such as the grain size and intermetallic phase, the samples are etched with different etchants. The etchants used in this study are the 5% citric acid to reveal the grain boundaries and grain size measurement, and the 10% hydrofluoric acid is used to reveal the \( \beta \) phase.

### 3.6 Optical Microscopic Studies

Two different optical microscopes are utilized in this study. The stereo microscope is used to determine the morphology of fracture surfaces after tensile test, and also the characterization of observed features is performed using the image analysis software-Clemex.

The study of microstructure required higher magnification and resolution. A powerful optical microscope is used to provide microscopic images for further analyses which include the grain size measurement and determination of the area fraction of porosities.

### 3.7 Characterization of Microstructural Features by Image Analysis (IA) and Image Processing

The detection of microstructural features is based on the binary operation which is the main function of IA Software named Clemex. The gray levels of images are started from the dark to the light gray level. The selection of an appropriate gray threshold is important in the characterization of microstructural features. The gray boundary threshold can be selected automatically or manually. The Flicker method is the manual
selection of binary level. It's based on the comparison between the microscopic and binarized image. The image processing can be applied to narrow gray-level range. Further explanations about these two methods are provided in Appendix-D-14.

The auto-binary operation is utilized for the characterization of porosities. To produce the quantitative pictures of the metallographic specimens the panorama pictures are made by stitching numerous consecutive pictures of one specimen. The software which is utilized to perform this procedure is Irfan View. The area fractions of the porosities are determined from the montage images. The differentiation between gas and shrinkage pores is carried out using the Roughness function of IA software (see Section 4.3.5).

The flicker method is considered for the grain size measurement. The grain size measurement can be performed by two different methods, and there would be a slight difference between the results obtained from each method. The first method is the field view measurement which is based on the intercept method (Figure 3.13-A). The second method is the object measurement, and the computation of grain size is based on the number of the grains per unit area (Figure 3.13-B). The diversity of the grain size across the field of view can be determined using the second method. To ensure the grain size measurement is performed accurately, both methods are applied for each field of the view. According to the ASTM E1382 the other features of IA software such as thinning and erosion can be used during the measurement of the grain size. The image processing features of Irfan View is also utilized to characterize the intermetallic phase of the specimens. Both components of eutectic phase are darkened by the etchant of hydrofluoric acid, so the separation of two components is performed by the gamma correction of the images (see Section 4.2.2 and [14]).

In spite of utilizing different techniques to characterize the microstructural features of the samples, the measurement process can be affected by different parameters. According to the ASTM E-112 there would be a deviation of 0.5 to 1 \( \mu m \) in the results of the grain size measurements using different techniques. The differentiation between the primary grains and segregated layers is conducted by the binary method and image processing features. These operations lead to the deviation of the results.
Figure 3.13 Comparison between two methods which are applied for grain size measurement (a) (b)
Figure 3.14 Shows the difference between two microscopic images of the sample CsNL2 which is subjected to the image processing by thinning and erode features.

![Image of Figure 3.14](image)

**Figure 3.14** Comparison of two microscopic images with gas pores (large black spots) before (a) and after (b) erode and thinning operation

To evaluate the deviation in the grain size measurement, the size of 9 individual grains is measured by IA Software and intercept methods. The length of the intercept lines are intentionally considered longer than the grain diameters, and two ends of them protrude into the dark area around the grains. In that way, the segregated layers are now taken into account to determine the actual grain size. The results show that the magnitude of the underestimation is about 1-1.5 $\mu m$ (The procedure is shown in Appendix D.1). Despite the existence of the deviations, the originality of experimental values is preserved, and they are not modified.

It’s also noted, there is a possibility of inaccurate results while a high number of the grains in each field of measurement are considered during the grain size measurement. As the large ESGs are eliminated from the fields which are used for determination of skin thickness, the grain diversity is low and numbers of counted grains do not have a significant effect on grain size measurement. To analyze the effect of grain numbers on the semi-automatic measurement of grain size, three fields of view from three different specimens is studied. The number of the grains in each measurement is decreased to characterize the variation of the grain size measurements. The results indicate that the
number of the grains in each field of measurement does not affect the measurement significantly (The related pictures are presented in Appendix D2-3).

However, a higher degree of the diversity is obtained from the samples containing large ESGs. Therefore the number of grains which are counted within each field of view affects the results which are obtained from the measurement of grain size. To appreciate the local increase of the average grain size within each field of view, the cross section of samples is divided into the smaller squared grids and average grain size is determined for each field and the fraction of each field \( f_i \) is applied to the prediction of yield strength by the modified form of the Hall-petch equation. Further explanation about the effect of local accumulation of large ESGs are presented in Appendix E.4
Chapter 4

4 Results & Discussions

In the previous chapter, different experimental methods were introduced. The results of tensile tests and microscopic investigations are presented in this chapter. The fluctuation of local mechanical properties of the die-castings of AM60 results from the microstructural variations across the castings. The results which are obtained from the microstructural investigation are the primary variables in the proposed models, and the model prediction is compared with the actual results which are obtained from the tensile tests.

In the first section the results of local yield strength are presented, and the experimental results are compared with the values which are predicted by the modified form of the Hall-Petch equation (2.12). Moreover, the skin thicknesses of the samples are determined, and the effect of cooling rate on the grain size is evaluated by the proposed relationship (2.7). The decreasing trend of yield strength with the area fraction of ESGs is presented, and the effect of grain distribution on the yield strength is described.

The variations of yielding behavior among the samples are justified by \( \sigma \theta \) plot in the second section. The onset of fully plastic behavior is determined and the effect of grain distribution on the strain hardening behavior of samples is explained. The variation of the intermetallic phase and hardness through the cross section of the samples are presented in this section, and the fraction of intermetallic phase is determined by Scheil-Gulliver equation.

The results of local fracture behavior and the correlation with volumetric porosities of different locations are presented in last section. The characterizations of porosities within the samples after tensile tests, and along the flow path are presented in this section. The failure model is validated using the data which is obtained from the determination of skin thickness and the area fraction of the porosities.
4.1 Effect of Microstructure on Yielding Behavior

The variation of solidification condition during the HPDC process leads to the bimodality of microstructure within the die cast specimens. The microstructure of die cast sample is composed of skin and core region, and the diversity of grains is shown in Figure 4.1 (see AppendixE-FigureE.5). As a result, different yielding behaviors can be expected in different regions. The local yield strength is determined by performing tensile tests. As explained, the contribution of grain size to the strengthening mechanism can be quantified by the modified form of Hall-Petch equation (2.12). The grain size measurement is carried out in different locations of the casting, and the skin thicknesses of the samples are determined.

![Image showing fine equiaxed grains within the skin region, and presence of large dendrites within the core region](image)

**Figure 4.1** Fine equiaxed grains within the skin region, and presence of large dendrites within the core region

4.1.1 Local Yield Strength

The yield strength of the samples is governed by the grain size of the samples. The results of tensile tests revealed that the yield strength of the samples fluctuates among the samples which are extracted from different locations (Figure 4.2 and Figure 4.3). The variation of the average grain size and distribution of grains results in the variation of the yield strength across the casting. While the yield strength ranges from 105 to 115 Mpa for
the Knit-line and last-to fill samples, a decrease of 10MPa is obtained from the close-to-ingate samples. The latter samples contain a high fraction of ESGs. Depending on the distribution pattern of ESGs, the decreasing on the yield strength is different. The samples containing accumulated ESGs within the core region has lower yield strength than the samples with dispersed pattern. Moreover, samples containing ESGs within the skin region has the lowest yield strength among the samples. The lowest yield point of 93 MPa is obtained for sample Cs1L6. The decreasing effect of the local ESGs on the yield strength is presented as $f$, in the modified form of the Hall-Petch equation (see Appendix E- Figure E.6-10 and Section 4.1.4).

![Graph showing yield strength variation in different locations.](attachment://image.png)

**Figure 4.2** variation of yield strength in different locations of the bolster, the error bars representing the standard deviation of seven tensile tests for each location

The overall yield strength of samples which are extracted from instrument panel was higher than the bolster, and it was $123 \pm 5\text{MPa}$ (Figure 4.3).
Figure 4.3 variation of yield strength in different locations of the new instrument panel, the standard deviation for each location is shown by the error bars.

4.1.2 Grain Size Distribution- Skin Thickness

The skin region is composed of finer grains, and a higher fraction of the skin region leads to the decrease in average grain size of the samples. Therefore, the fractions of skin in die-cast specimens affect the yield strength of samples. The grain size distribution, location of the shear bands as well as the presence of the ESGs are the criteria which are considered for determination of skin thickness [4]. The results which are obtained from the measurements of grain size and skin thickness will be used for the validation of the failure model (Section 4.3.6) and modified form of the Hall-Petch equation (2.12).

In this study, the skin thickness measurement is based on the analyzing of the grain size distribution and also it’s also assessed by other criteria. For instance, the analysis of grain size variation shows that the skin thickness of the sample Cs4L3 is 0.36mm, and this value matches well with the location of shear band in the sample (Figure 4.10, and Appendix D-Figure D.10). It’s also noted that shear ban does not follow the contour of the sample, and skin thickness varies from 0.36 to 0.2mm in the one side of the sample. Such deviation accounts for a deviation of 3Mpa in the prediction of yield strength by the modified form of Hall-Petch equation.
As the fraction of skin plays represents as the fraction of fine grains in the modified form of the Hall-Petch equation, the existence of large ESGs reduce the related fraction of fine grains ($f_i$) in the modified form of the Hall-Petch equation. For instance, large ESGs are located within the region close to the free surface of the samples Cs1L6 and Cs4L6 (Appendix E-Figure E.8-10). Therefore, the location of ESGs can be considered as the location of skin-core boundary, and the skin thickness would be 0.1 mm.

To determine the skin thickness of the samples by analyses of grain size measurement, the average grain size is considered as threshold value to define the skin-core boundary of the samples. The cross section of the metallographic specimens is investigated by optical microscope and the consecutive fields of view are prepared for the analyzing process. To avoid any biased measurements three different rows within the cross section of samples are chosen randomly. Each data point in Figure 4.4, 7, 9, 11, 12 is the average gain size obtained from three fields of view and the error bars define the deviation from the average value in each field of measurement. The horizontal axes in the figures represent the thickness of the metallographic specimen, and the position of each measurement is defined in the figures. The red dashed line defines the value of the average grain size, and column bars are the average number of grains in each field of view. The blue lines show the location of the skin-core interface. To provide a clear view about the variation of skin fraction, the results of three experiments on bolster and two of the new instrument panel are presented.

Figure 4.4 shows the grain size distribution across the specimen Cs7L11. The gradual increase of the grain size in the center of the casting is accompanied by a reduction in the number of the primary $\alpha$. It indicates that the coarse grains are located in the center of the casting.
Figure 4.4 Variation of grain size within sample Cs7L11, Column bars show the average number of grains and data points are the average grain size.

Figure 4.5 The ESGs are located close to the center line of sample Cs7L11, the Vickers indentations are visible as black spot.
The comparison between the measured skin thickness in Figure 4.4 and the location of shear bands (Appendix D – Figure D.4) indicates that skin thickness can be defined by both criteria. The location of shear bands matches well with the analyses of the grain size distribution. The images of etched sample (Figure 4.5) show that the ESGs are located closer to the centerline of the cross section, but not in the skin-core boundaries. Therefore, the onset of ESGs is not considered as an appropriate criterion for the skin measurement, and grain size measurement and location of shear bands are selected as preferred criteria. It’s also noted that the skin fraction is not uniform in the sample. The skin thickness in one side ranges from 0.2 to 0.3 mm and it is almost 0.5 mm in the other side. The same pattern is also observed in the samples cut from the locations 4, 5, 7, 8, 10 and 11 of the bolsters. Depending on the processing parameters and geometry of casting, the cooling rate is changing within different location of the casting, and affects the microstructure of the specimens. The casting geometry affects the local cooling rate. The heat flow rate is higher across the convex-shaped side than that of the concave-shaped side [33] (Figure 4.6).

**Figure 4.6 Two different heat flow regimes can lead to various skin thicknesses on both sides**

Figure 4.7 and Figure 4.9 show the grain size distribution of the other samples cut from different locations of the bolster. The grain size distributions of the specimens cut from the New Instrument Panel are shown in Figure 4.11 and Figure 4.12.
Figure 4.7 Variation of grain size within sample Cs4L8, Column bars show the average number of grains and data points are the average grain size.

Figure 4.8 Cross section of etched sample Cs4L8, the skin thicknesses are 0.31 and 0.51mm, white arrows help us to find the transition line.

More microscopic pictures of Cs4L8 are shown in Appendix D.6.
Figure 4.9 Variation of grain size within sample Cs4L3. Column bars show the average number of grains and data points are the average grain size.

Figure 4.10 Cross section of sample Cs4L3 with different skin thickness.

The microscopic pictures of sample Cs4L3 and the variation of grain size in other locations are shown in Appendix D.7.
Figure 4.11 Variation of grain size within sample CsNL2, Column bars show the average number of grains and data points are the average grain size.

Figure 4.12 Variation of grain size within sample CsNL4, Column bars show the average number of grains and data points are the average grain size.
The microscopic pictures of sample CsNL2-4 and the variation of grain size within other locations are shown in Appendix D (Figure D.8-10). Unlike the sample Cs7L11, the ESGs are located close to the skin-core boundary of samples Cs4L8 and Cs4L3 (Figure 4.8 and Figure 4.10). In addition to the difference in heat extraction rate, the skin thickness does not follow the contour of geometry (Figure 4.10). As a result, the non-uniform skins are formed in these samples. The variation of skin thickness is labeled in Figure 4.10. It’s also noted that some smaller ESGs are located within the skin region of the sample Cs4L3, and local increase of average grain size can be appreciated using the modified form of the Hall-Petch equation.

The high fraction of ESGs within the core regions of the samples which are extracted from the close-to-ingate region leads to the local increase of average grain size within the core (Figure 4.13, Appendix.D FigureD.29). The average grain size of the in-plane sections cut from the close-to-in-gate regions (locations 6 and 9) is about 16 μm.

![Microscopic pictures of sample CsNL2-4](image1.png)

**Figure 4.13 In-plane sections of the sample Cs1L9 with two different magnifications**

The microstructural characteristics of some samples are shown in Table 4.1. Two skin thicknesses are introduced as bottom and top in the following table. The morphology of grains within the skin thickness of the samples which are extracted from the close-to-ingate region is branched shape.
Table 4.1 Microstructural characteristics and yield strength of the samples

<table>
<thead>
<tr>
<th>Sample</th>
<th>Location</th>
<th>Average Grain Size (μm)</th>
<th>Skin Thickness (mm)</th>
<th>ESGs (%)</th>
<th>(\sigma_y) (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cs4L8</td>
<td>Center-Knit Line</td>
<td>Skin: 4.8 Core: 5.4</td>
<td>Bottom: 0.51 Top: 0.3</td>
<td>1.1</td>
<td>115.3</td>
</tr>
<tr>
<td>Cs7L11</td>
<td>Side-Knit Line</td>
<td>Skin: 5.6 Core: 6.1</td>
<td>Bottom: 0.5 Top: 0.27</td>
<td>0.6</td>
<td>108.6</td>
</tr>
<tr>
<td>Cs4L3</td>
<td>Last-to-Fill</td>
<td>Skin: 5.5 Core: 6.3</td>
<td>Bottom: 0.33 Top: 0.47</td>
<td>1.5</td>
<td>108.2</td>
</tr>
<tr>
<td>Cs6L13</td>
<td>Various</td>
<td>Skin: 5.8 Core: 6.9</td>
<td>Bottom: 0.31 Top: 0.45</td>
<td>2.4</td>
<td>104.8</td>
</tr>
<tr>
<td>Cs6L8</td>
<td>Center-Knit Line</td>
<td>Skin: 5.3 Core: 6.2</td>
<td>Bottom: 0.5 Top: 0.32</td>
<td>1.3</td>
<td>108.0</td>
</tr>
<tr>
<td>Cs1L9</td>
<td>Close-to-ingate</td>
<td>Skin: 6.8 Core: 16</td>
<td>Bottom: 0.4 Top: 0.31</td>
<td>4</td>
<td>96.0</td>
</tr>
<tr>
<td>Cs4L9</td>
<td>Close-to-ingate</td>
<td>Skin: 6.1 Core: 13</td>
<td>Bottom: 0.37 Top: 0.25</td>
<td>5.5</td>
<td>106.4</td>
</tr>
<tr>
<td>Cs1L6</td>
<td>Close-to-ingate</td>
<td>Skin: 11.5 Core: 12</td>
<td>Bottom: 0.16 Top: 0.25</td>
<td>2.6</td>
<td>93.0</td>
</tr>
<tr>
<td>Cs4L6</td>
<td>Close-to-ingate</td>
<td>Skin: 9.8 Core: 10.7</td>
<td>Bottom: 0.28 Top: 0.18</td>
<td>1.9</td>
<td>98.3</td>
</tr>
<tr>
<td>CsNL4</td>
<td>Last-to-Fill</td>
<td>Skin: 5.7 Core: 6.4</td>
<td>Bottom: 0.57 Top: 0.54</td>
<td>-</td>
<td>127.9</td>
</tr>
<tr>
<td>CsNL6</td>
<td>Various</td>
<td>Skin: 6.5 Core: 7.2</td>
<td>Bottom: 0.4 Top: 0.39</td>
<td>0.04</td>
<td>122.6</td>
</tr>
</tbody>
</table>
4.1.3 Effect of Skin Fraction on Yield Strength

It’s also demonstrated that an increase of skin fraction from 20% to 40% leads to an increase of 13% in the yield strength of die cast AM60 samples [3,69]. The study of grain size distribution reveals that the skin thickness across the bolster ranges from 0.2 to 0.5mm. A significant difference between the skin fractions of samples which are extracted from the knit-line location of the bolster is not observed. In spite of an apparent decrease in the skin thickness of close-to-ingate samples, the lower yield strength of these samples mainly results from the presence of ESGs within the core region. Due to the presence of ESGs and branched shape grains within the skin region of samples, the average grain size of the samples is increased up to $7.4 \mu m$. Therefore, the effect of skin fraction cannot be considered as sole parameter for the prediction of yield strength. The fraction of skin is represented in the modified form of the Hall-Petch equation.

To provide a clear insight about the previous explanation, the increase of skin fraction in the samples which are extracted from the new instrument panel is studied. The comparison between Figure 4.11 and Figure 4.12 indicates that the thickness of the skin is increased in the thicker samples. The variation of the yield strength versus the skin fraction of the samples cut from the new instrument panel is shown in Figure 4.14. In spite of the increase in skin thickness of the samples, the skin fraction and yield strength do not change considerably. Sample CsNL6 is the thinnest sample with a thickness about 2 mm, the skin fraction of this sample is higher than locations 2, 3 and 4. However, the average grain size of this sample is 6.5 $\mu m$ in the skin region, and it is slightly larger than that of CsNL2 and CsNL4 of about 6 $\mu m$. The average grain size within the core region of CsNL6 is 7.2 $\mu m$. The average yield strength of CsNL6 is slightly lower than that of the other samples.

The comparison between the microscopic images of samples CsNL2 and CsNL4 reveals that the microstructure of the first sample is mainly composed of the equiaxed grains, whereas it is branched-shape in the second sample (Figure 4.15). It’s more likely that this section of the casting experiences different solidification conditions (Figure 3.9), and
simulated results indicate that the cooling rate of this section ranges from 90°C/S to 120°C/S which is lower than that of the section 2, 3 and 4.

![Figure 4.14](image1.png)

**Figure 4.14** Variation of yield strength skin thickness of samples extracted from new instrument panel

![Figure 4.15](image2.png)

**Figure 4.15** Comparison between the morphology of the grains within samples CsNL2 and CsNL4

The grain size measurement is also compared with the results obtained from the simulation of the process for both bolster and new instrument panel. The simulation
results are presented in Appendix D.15. The simulated value of cooling rate represents the predicted value of cooling rate within the core region of specimens. This value can be used to predict the average size of the grains within the core region of the samples, and equation (2.7) is applied for the prediction. According to the simulation of the HPDC process, the cooling rate ranges from 200°C/S to 250°C/S within the core region of the tensile specimens which are extracted from the knit line region of bolster (Figure 4.16). Locations 4, 5, 7, 8, 10 and 11 are the tensile specimens of this region. However, the different cooling rates between the two sides of the locations are not clearly identified by the simulation of processes.

![Solidification Rate](image.png)

**Figure 4.16 Variation of solidification rate from the flange gate to the overflow within the JK section of bolster**

The simulated results also indicate that there is an increase in solidification rate as the thickness of the samples is decreasing. The cooling rate is decreased from 180°C/S to 90°C/S as the thickness of the sections increases from 2.2mm to 4.7mm in the new instrument panel. The cooling rate of locations 2, 3 and 4 of the new instrument panel
ranges from 150°C/S to 180°C/S, while it’s about 90°C/S to 120°C/S for locations 1 and 5 of the new instrument panel. Location 6 of the new instrument panel is an exception, the thickness of this sample is about 2mm and the predicted cooling rate ranges from 90°C/S to 107°C/S. Figure 4.17 shows the predicted and experimental grain sizes. Depending on the cooling rate the predicted grain size ranges from 11.5 μm to 14.5 μm. The size of the grains in the core region of the bolster is almost half of the predicted values, and the difference between the predicted and experimental values of thick samples is less than 4 μm.

![Bar chart showing predicted and experimental grain sizes](image)

**Figure 4.17 Comparison between the predicted and experimental values of the grain size**

In spite of the difference between predicted and experimental values, the decreasing trend of the grain size versus the increase of cooling rate is also captured with experimental results. It’s also noted that the morphology of grains is especially different within the core region of thick samples. The branched grains are formed in the core region (Appendix D-Figure D.37). There is a possibility that the textures on the surface of the bolster improves the adhesion of casting to the mould, and deteriorate effect of the air gap on the heat transfer coefficient is reduced [24, 37].
4.1.4 Prediction of Yield Strength

The contribution of grain size to hardening mechanism can be characterized by Hall-Petch equation. The high diversity of the grain sizes across the cross section of die-cast specimens cannot be evaluated properly by the general form of the Hall-Petch equation. To deal with the high diversity, the average grain size can be determined for each field of view, and then the fraction of each field ($f_i$) can be attributed to the general formula. Therefore the modified Hall-Petch relationship (equation (2.12)) is considered to determine the theoretical yield point of the specimens [7]. To predict the yield strength of samples the friction stress $\sigma_0$ is considered to be equal to 12 MPa and the slope $K$ is equal to $280 \text{ MPa}\sqrt{\mu m}$ [32]. The comparison between the predicted and experimental values of tensile strength of the samples is shown in Figure 4.18.

![Graph showing comparison between predicted and experimental yield strengths](image)

**Figure 4.18** Experimental yield strength versus predicted yield strength
The predicted values match well with experimental data which are obtained from the tensile experiments of the new instrument panel. The maximum deviation between two values was less than 5%. However, an overestimation of more than 10% is obtained from the samples which are extracted from the bolster. A higher fraction of porosities within the samples can lead to the lower value of yield strength [7,66]. The microscopic images of the samples revealed that the defect bands containing higher fraction of porosities, and they are formed within the skin-core boundaries of the samples (see Section 4.3.5). The deviation between the predicted and experimental values can be related to the deviation in the measurement of grain size (see Section 3.7). The main purpose of this study is the validation of the modified form of the Hall-Petch equation, so the friction stress and slope of the equation is recalculated by plotting the average grain size versus the experimental values of yield strength (Figure 4.19). Then, the general form of the Hall-Petch relationship is:

\[
\sigma_y = 8 + 273d^{-0.5}
\]  

(4.1)

![Diagram](image_url)  

Figure 4.19 Experimental yield strength versus grain size diameter (\(\mu m\))
To compare the difference between the predictions which are made by the general and modified form of the Hall-Petch equations, the new expression (4.1) is applied for predicting of the yield strength (Figure 4.20).

![Graph showing predicted yield strength versus experimental yield strength](image)

**Figure 4.20 Experimental yield strength versus predicted yield points which are obtained from two different approaches**

The comparison between two predicted values indicates that the difference between the predicted is low for the samples containing low fraction of ESGs. However, a difference of more than 10% is obtained from the samples containing higher fraction of ESGs in the core region. The local grain size of the samples increases within the field containing higher fraction of ESGs (see Appendix E-Figure E.6-10). For instance, the predicted yield strength of sample Cs1L9 is 97.5 Mpa, and the predicted value which is obtained
from the general form of the Hall-Petch is 91.2MPa. As the diversity of grain size is increasing within the samples, the modified form of the Hall-Petch equation would be the preferential model for the prediction of the yield strength. The cumulative effect of the regions with larger/smaller average grain size is taken into account in the modified form of the Hall-Petch Equation, while their effect is lessened in the general form of the Hall-Petch equation.

The previous explanations indicate that the presence of ESGs within the samples, leads to the reduction of the yield strength. Although the higher fractions of ESGs are detected within the samples which are extracted from the close-to-ingate region, the distribution patterns of ESGs are different among different samples (Figure 4.21).

Figure 4.21 ESGs are packed in the center line of the Cs1L9 and they are more dispersed in the sample Cs4L9.
As a result, the increasing trend of average grain size would be different within different field of measurement. For instance, the presence of ESGs within the skin region of the samples (see Appendix E-Figure E.9-10) eliminates effect of hard skin on the strengthening of the samples. The yield points of these samples were lower than the samples containing the same fraction of ESGs within the core region. Therefore, the yield strength can be correlated to both fraction and the distribution pattern of ESGs. The flow pattern and solidification conditions are the influential parameters affecting the position and transportation of the ESGs. The distribution pattern of ESGs depends on the strength of the skin and solidification process in the shot sleeve. A more distributed pattern is resulted from weak skin and a higher fraction of ESGs which are formed in the shot sleeve [46]. The procedure which is taken to determine the fraction of ESGs is shown in Appendix E.3. The general trend in Figure 4.22 shows that the yield strength is decreasing while the fraction of large ESGs (> 25 μm) is increasing. However, the decreasing trend is not applicable for all samples.

![Experimental yield strength versus the percentage of the large grains](image)

*Figure 4.22 Experimental yield strength versus the percentage of the large grains*
The diversity in grain size distribution is decreasing under the condition that the ESGs are more dispersed. Therefore, the local increase in the average grain size is not observed for the dispersed distribution within the core region. In spite of the high percentage of ESGs, higher yield strength is obtained in sample Cs4L9 compared to the samples that the ESGs mainly accumulated in the center. The accumulation of ESGs within the core region leads to the increase of average grain size within the core region (16 \( \mu m \)). Moreover, the presence of ESGs within skin region of the samples leads to the local increase of grain size. As a result, the yield strength of theses samples is decreasing, and the yield behavior of theses samples is similar to the samples containing higher fraction of ESGs within the core region. Further information about the cumulative effect of large grains on the average grain size is presented in Appendix E.

4.2 Effect of Microstructure on Strain Hardening

The diversity and distribution pattern of the grain sizes affect the yield strength and strengthening of the samples. It’s also demonstrated that the grain size is the most influential parameter, and fraction of intermetallic phase is another effective parameter in the strengthening mechanism of the Mg-Al alloys [32]. The variation of hardness and intermetallic phase through the cross section of samples is determined to evaluate these effects. Moreover, the maximum fraction of intermetallic phase is obtained using Scheil-Gulliver equation. The analytical results are used to determine the onset of fully plastic deformation of each sample. The variation of strain hardening among the samples containing different diversity of the grains is explored with the assistance of \( \sigma \theta \) plots.

The main purpose of these analyses was to highlight the effect of grain size distribution on strengthening of the die-cast samples of AM60. This effect is presented as a fraction term in the modified form of Hall-Petch equation. Moreover, the onset of fully plastic deformation for each sample is determined with the assistance of these plots.

4.2.1 Hardness Variation & \( \beta \) Phase Distribution

The Vickers hardness was performed on the various samples including the flange gate, knit line as well as close to-ingate samples. Although the lower average hardness within the core region is common among the samples, high discrepancies among the results were
evident. The higher hardness within the skin region is mainly resulted from the fine grains, a higher fraction of intermetallic phase as well as the smaller interconnection of the intermetallic particles [68]. The related figures and results of harness tests are presented in Appendix E.

The variations of $\beta$ phase across the section of three samples are also studied. Both components of the eutectic phase usually appear as dark areas after the etching process. Therefore, an appropriate image processing should be applied to differentiate between these two microstructural features [44], and it’s possible by changing of contrast, brightness as well gamma correction of the images (Figure 4.23). The average area fractions of the $\beta$ phase across the section of samples are determined for each field of view. The data points in Figure 4.24 indicate that there is a slight decrease in the area fraction of the $\beta$ phase within the core region of all samples.

Figure 4.23 Image processing for determination of $\beta$ phase, as pointed by the arrow
The distance between dispersed particles affects the contribution of them to the strengthening process, so the mean distance between $\beta$ particles is determined by intercept method (Figure 4.25). The intercept lines are shown in Appendix D (Figure D.38). The spacing distance between particles is higher within the core region of the samples which are extracted from close-to-ingate regions. As the divorced eutectic is located in the boundary of grains, the increase in distance between the $\beta$ phase particles in the core region of the samples is predictable. The average spacing of $\beta$ particles in Cs1L9 is 1.3 times larger than that of sample Cs6L3, and it’s almost equal to the average grain size within the core region of sample Cs1L9. In addition to the effect of dispersed particles on strengthening mechanism, grain size is the most influential parameter in the strengthening mechanism of the die cast specimens. The yield strength of sample Cs1L9 is about 10% lower than that of the sample Cs6L3.
Although the image processing technique can improve the separation of the eutectic components, the differentiation between the $\beta$ phase and eutectic phase cannot be performed properly. The average area fraction of $\beta$ phase within the skin region ranges from 1.5% to 4.5%, and the average area fraction is about 3.3% within the samples. Due to the localized distribution of the $\beta$ phase, scattered results are obtained from different regions. The average area fraction which is reported for die-casts of AZ91 [44] was lower than the current experimental values.

To determine the onset of fully plastic deformation of each sample, it’s necessary to estimate the maximum fraction of the intermetallic phase within the sample. The intermetallic particles are the last fraction of elastic which remains within the samples before beginning of fully plastic deformation. The Scheil-Gulliver approach is also applied to estimate the maximum percentage of the non-equilibrium eutectic phase within the magnesium alloys containing different fractions of aluminum. Figure 4.26 shows the non-equilibrium diagram phase of different magnesium alloys.
Figure 4.26 Non-equilibrium solidification of different magnesium alloys, as the aluminum content is increasing, the higher fraction of liquid reaches the eutectic composition.

The liquidus ($T_L$) and solidus ($T_S$) lines are considered as the straight lines and the equivalent equations are shown in Figure 4.26. The fraction of liquid state at each temperature can be determined by the equation (2.5), and the weight percentage of solute content in solid state can be computed from equation (2.4). As the percentage of aluminum is increasing, the fraction of the non-equilibrium eutectic is increasing. Table 4.2 indicates that the maximum weight fraction of the eutectic is less than 6% and 12% for AM60 and AZ91, respectively. The volume fraction of the $\beta$ phase was reported to be nearly 3.6% for AM60 [32]. The second column of the table ($C_{S/equil}^{Scheil}$) defines the intercept point of non-equilibrium solidus line and eutectic line in the diagram. And the solidification range is also determined using the equation (2.6).
Table 4.2 Weight fraction of non-equilibrium phases of Mg-Al alloys

<table>
<thead>
<tr>
<th>Alloy</th>
<th>$C_{S/437}^{Scheil}$</th>
<th>Eutectic $%$</th>
<th>$Mg_{17}Al_{12}$ $%_{Scheil}$</th>
<th>Secondary $\alpha$ $%_{Scheil}$</th>
<th>Solidification Range</th>
</tr>
</thead>
<tbody>
<tr>
<td>AM20</td>
<td>1.7</td>
<td>1</td>
<td>0.69</td>
<td>0.31</td>
<td>200</td>
</tr>
<tr>
<td>AM60</td>
<td>4.3</td>
<td>6</td>
<td>4.1</td>
<td>1.9</td>
<td>174</td>
</tr>
<tr>
<td>AZ91</td>
<td>5.8</td>
<td>12</td>
<td>8.3</td>
<td>3.7</td>
<td>155</td>
</tr>
</tbody>
</table>

4.2.2 Effect of Grain Size on Strain Hardening- $\sigma$ Plots

The inhomogeneous microstructure of the HPDC specimens results in the fluctuation of yield strengths in the skin and core regions of the die cast specimens. It’s established that the skin is harder than the core. The primary reasons for higher hardness within skin region are smaller grains and closer $\beta$ phase particles. During the early stage of tensile test, the skin is supposed to remain elastic, while the plastic deformation has already begun in the soft core. Figure 4.27 shows that at lower applied stress both skin and core are elastic and the stress-strain curve is relatively linear. Once the applied stress reaches the yield point of the core region, a tensile stress should be applied from the skin to maintain the uniform deformation of the geometry. Therefore the state of the stress will be changed to the tri-axial condition. As a result the yield strength and onset of fully plastic behavior among the samples containing different fraction of soft region is different. The extent of microplasticity regime can be evaluated by analyzing of strain hardening behavior. The analyzing procedure is applied by Caceres [12] to determine the fraction of the skin in magnesium alloys with different aluminum content. The skin fraction of the samples containing a higher percentage of aluminum is higher than the diluted alloys; therefore the plastic deformation is constrained by the elastic portion of the samples [12]. In the current case, the skin fraction is relatively constant for the samples which are extracted from the knit-line and last-to-fill regions of the bolster. Due to the presence of ESGs, the skin thickness of the samples which are extracted from the
location 6 is lower than the other samples. Moreover, core region of the samples are different from each other. It ranges from the relatively fine grains (Knit line samples) to the samples containing higher fraction of ESGs with different distribution pattern.

Analyzing of the strain-hardening rate is performed for the specimens of AM60 cut from different locations of the bolster. To provide clear insight about the microstructure of the samples, the grain size profiles of the samples are shown in Figure 4.28 and Table 4.1. A high number of ESGs are located within the samples cut from locations 6 and 9 with different distribution pattern (Figure 4.21 and Appendix E.4).

The true strain of 0.005% is chosen to define the proof yield strength in the elastic region (\(Y_s\)), in that way our analyses cover both elastic and plastic region of the stress-strain curve. The initial proof stress at \(\varepsilon_{True}^{Plastic} = 0.005\%\) varies among the samples, as presented in the first row of Table 4.3. The proof stress is lower for samples containing higher

**Figure 4.27 True stress-Strain curves of different samples containing different diversity and distribution pattern of grains**
fraction of accumulated ESGs. The true plastic strain is calculated by the following expression:

\[
\varepsilon_{\text{True Plastic}} = \varepsilon_{\text{Total}} - \frac{\sigma_{\text{True}}}{E}
\]  

(4.2)

where, \( E \) is the Young’s modulus. To analyze the variation of the strain-hardening rate among different samples, the \( \sigma \theta \) plots are formed for different samples, and the following equation is used to sketch the plots (Figure 4.29) [12].

\[
(\sigma - Y_s) \frac{d\sigma}{d\varepsilon} = (\sigma - Y_s)\theta = (\sigma - Y_s) \frac{E}{32}
\]

(4.3)

Where, \( \sigma \) and \( \varepsilon \) are the true stress and true strain, respectively.

Figure 4.28 Grain size profile of different samples, samples Cs1L6, Cs1L9 has a higher fraction of ESGs
Figure 4.29 The $\sigma\theta$ plot for the samples with different grain size distribution cut from different locations of the bolster

The Young’s modulus is $44 \text{ GPa}$ for the pure magnesium and the strain-hardening rate at low strains would be $\frac{44}{32} = 1.4 \text{GPa}$ [12] which is shown by the dashed line in Figure 4.29.

The mean strain-hardening rate ($\frac{d\sigma}{d\epsilon}$) is determined by averaging the slope of 10 data points. The initial slope was nearly 22 for all samples within the initial stages of tensile test, and then it dropped to nearly 17 for all the samples with the exception of sample Cs6L8 (Table 4.3). The initial slope of 16.3 is also reported for the magnesium alloys with 0.47% aluminum and it resembles the plastic behavior of the core region [12]. Then the slopes of the $\sigma\theta$ curves are decreasing and all the lines reach the strain-hardening rate of the polycrystal magnesium with the purity of 99.7% (Figure 4.29-dashed line). The higher upward increase is obtained from the samples containing higher fraction of
fine grains within the core region (knit-line samples Cs6L8 and Cs4L3), and the sample containing dispersed pattern of ESGs within the cross section (Cs4L9) also displayed the same behavior. The yield strengths of these samples were higher than the samples containing accumulated ESGs within the core region (Table 4.3). It is also noted that the hump shape of the curve is more extended for the samples with higher yield strength, and the samples containing high fraction of accumulated ESGs reach the 1.4GPa line sooner than the samples with finer structure.

It’s assumed that the volume fraction of intermetallic phase is the same for all samples, and which is equal to 0.036 obtained from the Scheil approach (Table 4.2 and [32]). The contribution of the intermetallic phase in the increase of the initial slope can be estimated by equation (2.11), and the Young’s modulus of the $\beta$ phase is 58 GPa. The contribution of the $\beta$ phase in initial slope is about 2.1. The latter decrease in slope is mainly related to the high diversity in grain size, and this effect is pronounced in samples containing high fraction of accumulated ESGs within the core region. The higher strain hardening rate is well extended for the samples with lower fluctuation in local average grain size.

Table 4.3 The information obtained from Figure 4.29& Figure 4.30

<table>
<thead>
<tr>
<th>Samples</th>
<th>Cs1L6</th>
<th>Cs1L9</th>
<th>Cs4L3</th>
<th>Cs6L8</th>
<th>Cs6L13</th>
<th>Cs4L9</th>
</tr>
</thead>
<tbody>
<tr>
<td>$Y_s$ (at $\varepsilon = 0.0005$)</td>
<td>62.5</td>
<td>62.5</td>
<td>75.9</td>
<td>66</td>
<td>69.5</td>
<td>67.2</td>
</tr>
<tr>
<td>Slope (25 data points)</td>
<td>17.1</td>
<td>17</td>
<td>16.4</td>
<td>20.2</td>
<td>16.4</td>
<td>16.2</td>
</tr>
<tr>
<td>Stress at Lowest Elastic Fraction (MPa)</td>
<td>45</td>
<td>44.6</td>
<td>55.7</td>
<td>62.5</td>
<td>51.5</td>
<td>54.6</td>
</tr>
<tr>
<td>End Point of Elastic-Plastic Region (MPa)</td>
<td>107.5</td>
<td>107.1</td>
<td>131.6</td>
<td>128.5</td>
<td>121</td>
<td>122.1</td>
</tr>
<tr>
<td>$Y_s$ (Yield Strength at $\varepsilon = 0.002$)</td>
<td>93</td>
<td>96</td>
<td>111</td>
<td>108</td>
<td>105</td>
<td>106</td>
</tr>
</tbody>
</table>
The die-cast specimens are composed of two different regions, so they can be treated as the composite material. The total volume fraction of the material which remains elastic \( f \) can be calculated by the application of the load sharing equation in composite material [12].

\[
\sigma \theta = \epsilon E f + \sigma \theta_{\text{core}} (1-f) \tag{4.4}
\]

Where, \( \theta \) and \( \theta_{\text{core}} \) are the strain-hardening rates of the composite (HPDC specimen) and the core region, respectively. To solve the equation (4.4), the magnitude of the variables should be defined. The magnitude of \( \theta \) is already determined. According to the microhardness experiments which are conducted by Weiler et al [5], the strain-hardening rate of the core region \( (\theta_{\text{core}}) \) is nearly 1.4GP for die-cast specimens of AM60. By substitution of the known values, the total volume fraction of the elastic material \( f \) can be determined by the following equation.

\[
f = \frac{\theta - \theta_{\text{core}}}{E - \theta_{\text{core}}} = \frac{\theta - 1.4}{42.6} \tag{4.5}
\]

The volume fraction \( (f) \) includes the elastic material in both skin and core. The total volume fractions of the elastic material are calculated for the samples and the results are shown in Figure 4.30. The beginning points for different knit line samples are almost the same, and a slight deviation is obtained for the samples which are extracted from the close-to-ingate regions. The sharp initial decrease is obtained from the samples containing accumulated ESGs within the core region, while it’s postponed for the sample containing dispersed ESGs. The initial decrease can be related to the early onset of plastic deformation of the core region.
Figure 4.30 Total volume fraction of the elastic material within different specimens

A higher strain-hardening rate is obtained from the samples with fine microstructure and the plot is extended up to the higher values of stress. In spite of the similar behavior between the samples with fine microstructure and the sample with dispersed ESGs up to the yield point (0.2% offset), the decreasing trend is accelerating in the third stage for the sample Cs4L9. It implies that the recovery stage is faster for this sample, which is mainly related to the global increase of average grain size. The only remaining elastic fraction at this point is the intermetallic fraction, and by subtraction of the stress associated with this value (third row-Table 4.3) from the true stress the onset of fully plastic behavior can be defined for the samples (fourth row-Table 4.3). The onset of fully plastic deformation is labeled with (X) in Figure 4.12. Although the maximum difference between the yield point (0.2% offset) of the samples is below 18 MPa, the difference between the end point of the elastic-plastic region among the samples ranges from 14 to 24MPa (row 4 and 5 in Table 4.3).
4.3 Effect of Microstructure on Fracture Behavior

The results of tensile tests on the bolster and new instrument panel are presented in this section. The characteristics of the porosities are studied in this section, and appropriate functions are utilized to characterize the distribution and types of porosities. To evaluate the fracture behavior of die-cast samples of AM60, the recent failure models [10,11] are used to predict the fracture stress and strain of the samples which are extracted from different locations of the bolster and new instrument panel. The fraction and distribution of the porosities are determined from the plane below the fracture surfaces of tensile specimens as well as the information which is obtained from the analyses of the location before tensile test.

4.3.1 Local Fracture Behavior & Fracture Surface Morphology

The local ductility of different locations of the casting is determined by the performance of the tensile test on 105 tensile coupons of the bolster. The tensile tests were performed on different location of the new instrument panel by Meridian Inc. Figure 4.31 shows the true stress-strain curves obtained from the tensile test of seven specimens. They are representatives of 105 tensile test experiments and show the typical mechanical results of each location.

Although there are significant difference in fracture strain and stress, the flow curves of the samples are almost similar. The following expression represents the average power law equation derived from the plastic section of the true stress-strain curves of the samples.

\[ \sigma = 382\varepsilon^{0.235} \]  \hspace{1cm} (4.6)

The power law equation and the average fracture stress and strain of each location is shown in Figure 4.32. The fracture behavior of each location follows the trend which is obtained from the power law equation.
Figure 4.31 True stress-strain curve of the samples extracted from the bolster

Figure 4.32 Average fracture behavior of each location is compared with the power law equation
The mechanical properties of each location are shown in Figure 4.33. The true fracture strain ranges from 1.2 to 9.5% and the true fracture stress ranges from 143 to 228MPa. The lowest fracture strain belongs to one of the knit line region (Location 8) and the specimens cut from one of the last-to-fill regions (Locations 3 and 14). The highest fracture strain is observed in the specimens cut from the close-to-in-gate regions. The reduction of almost seven percent in the yield strength is also observed in this location. The significant variation of the fracture strain is obtained in the samples cut from the locations 7 and 11. The fracture strain ranges from 3.6 to 7.8% in location 11. Although the average strain hardening exponent is 0.235 for most of the samples, the apparent lower value is obtained for the samples experienced unexpected failure, as the damage accumulation accelerates the failure mechanism. The detailed information about the variation of mechanical properties is presented in Appendix A (Table A.1).

![Graph showing variation of mechanical properties across locations](image)

Figure 4.33 Variation of local mechanical properties across the bolster (The detailed information is shown in Appendix A – Table A.1)
Figure 4.34 shows the results obtained from the tensile tests which are conducted on different samples of the new instrument panel. The tensile tests are performed by Meridian Inc.

![Figure 4.34 Local mechanical properties of new instrument panel](image)

4.3.2 Characteristics of Fracture Surfaces

The fracture surfaces of the tensile specimens are investigated using the low magnification stereo microscope and SEM. The fracture plane reaches the instability point sooner than other neighboring planes, and also the largest imperfections are usually located in this plane. The internal necking usually occurs locally within these planes. The fibrous and dimpled features which are the characteristics of this mechanism are shown on the SEM images of the fracture surfaces (see Appendix B-Figure B.56-63). The stereo-images of the fracture surfaces are presented in Appendix B (Figure B.1-58). The features on the fracture surfaces are characterized by the IA software.

*Fracture surfaces with distributed porosities and inclusions:* The porosities and inclusions are observed on the fracture surfaces of samples cut from locations 8, 3, 7 and 14 (Figure 4.35). They are extracted from the knit-line and last-to-fill regions. The area fraction of the impurities on the fracture surfaces ranges from 0.3 to 1.6%. The fracture
strains of samples with large inclusion and close to the edge of the specimens are lower than 5%. Although the inclusions can be considered as internal discontinuities, they are not the only cause of the premature fracture. The red dashed ellipse in Figure 4.35 shows that the porosities are located close to the edge of the sample. The chemical analysis of inclusions on the fracture surface of the sample is conducted by energy dispersive X-ray analysis (EDX), and it’s revealed that the inclusions are mainly composed of calcium (The result of EDX is shown in Appendix B.2).

![Figure 4.35](image)

**Figure 4.35** Large inclusions on the fracture surfaces of the specimens with two different magnifications

**Tongue and groove shaped fracture surfaces:** The higher densities of the porosities, large pores as well as long cracks are observed on the tongue and groove shaped fracture surfaces. This group consists of the fracture surfaces of some samples cut from the knit line region (Locations 4, 5, 10 and 11). The fracture strains of these samples are low and range from 3 to 5.5%. As shown in Figure 4.36, the SEM images of the fracture surfaces show that the long propagated crack passes through the large pore.
Figure 4.36 Tongue and groove shaped fracture surfaces, the red dashed circle indicates the location of the large pore

*The fracture surfaces with high density of shrinkage cavities:* A high fraction of the shrinkage porosities is observed on the fracture surface of the samples cut from locations 3 and 14. The SEM images of the fracture surface of the sample Cs3L3 show that the location of the shrinkage pores can be easily identified by the presence of intact dendrites within the pores [72](Figure 4.37 samples Cs3L3 (a) and (b). The large shrinkage cavity is also observed on the fracture surface of the same sample (Figure 4.37 samples Cs3L3 c and d). The microscopic pictures of the plane below the fracture surface reveal that a higher density of pores is located within the defect bands of the samples, and they act as the preferential sites for crack initiation. Figure 4.37(Cs2L3a,b) is a clear evidence that shows the propagation of a crack from the defect band towards the edge of the sample. As the location of defect band matches with the location of skin-core interface,
application of skin fraction for determination of the lower bound of fracture surface is reasonable. The fracture strain of the samples which are extracted from locations 3 and 14, ranges from 2 to 5.5% (The microscopic pictures of the shrinkage pores are shown in Appendix A-Figures A.11-13 and Appendix B.2).

Figure 4.37 Microscopic images showing intact dendrites (Cs3L3-a-b), large shrinkage pore (Cs3L3-c-d), propagation of cracks from the defect band of sample Cs2L3

4.3.3 Volumetric Porosity - Mechanical Properties

Although it’s demonstrated that the clear relationship between volumetric porosities and mechanical properties [2,71], the decreasing trend of fracture strain against volumetric
Porosities is reported in [40]. Therefore, the variation of volumetric porosities among the samples is studied and the results are shown in Figure 4.38. The density of the HPDC magnesium alloys is lower than the nominal value due to the formation of the porosities in the process. As a result, the average density of the bolster is $1.766 \pm 0.002 \text{ g/cm}^3$. Depending on the flow pattern and solidification condition the fraction and location of the porosities vary across the casting.

![Figure 4.38 Variation of volumetric porosity and ductility across the bolster, column bars are the local volumetric porosity of the bolster](image)

The results of volumetric porosity highlight the efficiency of overflow position in lowering the fraction of porosities. The samples with three overflows have the lowest amount of volumetric porosity. However, it cannot be interpreted into better ductility. To provide a clear view about the variation of volumetric porosities, the samples are divided into five different groups.
The first group of tensile specimens cut from the knit line location has the lowest level of volumetric porosity (Locations 7 and 8). Although the ductility of most samples cut from location 7 is higher than 6%, the least ductility is obtained from the samples of location 8. The first group is subjected to further investigations, and the microscopic images revealed that an array of porosities is located beneath the free-surface of the specimens (Figure 4.39-A). The subsurface pores are the predominate cause of the premature failure in these samples.

Figure 4.39 Subsurface pores of sample Cs1L8 with two different magnifications (red dashed rectangle), showing the propagated cracks through the sub-surface pores in SEM image of the same sample(C)

The second and third groups of the samples are extracted from the knit line region with two overflows nearby (Locations 4, 5, 10 and 11). A significant fluctuation of the mechanical properties is obtained from the samples which are extracted from the knit line region with two overflows nearby. The dramatic decrease in density of the samples which are extracted from Locations 10 and 11 is followed by an abrupt reduction in fracture strain (Appendix A- Figure A.2). It’s noted that the large pores are also found in these locations (Figure 4.42)[40]. The second highest volumetric porosity belongs to the fourth group of the samples (Locations 3 and 14). This group is cut from the last-to-fill region, consisting of the thinnest samples.
The fifth group of the specimens is also cut from the last-to-fill region (Locations 1 and 15). Although they have higher volumetric porosity, the mechanical properties of the samples are almost consistent, and the deviation of the fracture strain is less than 0.013mm/mm. The results obtained from microscopic studies indicate that the fifth group mainly contains the dispersed gas pores, on the other hand the fractography and microscopic images show that more shrinkage porosities exist on the fracture surface of Locations 3 and 14 (Figure 4.37 and Appendix B.2). The fluctuation of local mechanical properties is explained, and the information obtained from the study of the fracture surfaces shows that the main defects causing failure are porosities and inclusions. Although the decreasing trend of fracture strain with volumetric porosities is obtained for some samples containing large pores, the clear correlation between the volumetric porosities and fracture strain did not obtain for most samples, and other characteristics of defects such as the distribution and location can contribute to the failure mechanism of the samples.

4.3.4 Characterization of the Porosities within Tensile Specimens

The fracture mechanism is composed of the growth, coalescence and nucleation of the pores which they can be classified as pre-existed and nucleated pores. The failure model is mainly based on the investigations which are conducted on the specimens before the performance of the tensile test. To determine the porosity level which is the input variable for the failure model, the plane adjacent to the fracture surface is analyzed. The location of the equivalent pore is considered in the vicinity of the skin-core boundary. To evaluate our procedures, it’s necessary to study the distribution and variation of the area fraction of porosities within the tensile specimens. Therefore, the distributions of porosities within both vertical and in-plane sections of the samples are studied.

The different flow patterns and solidification conditions lead to different distributions of the porosities. Figure 4.40 shows two in-plane sections which are extracted from the gauge length of the tensile specimens. To determine the average distance from the neighboring pore, the parallel lines are sketched in two different directions of X and Y, and then the intercept lines are measured along two different directions. The procedure
and images of other in-plane sections are shown in Appendix A (Figure A.3-8). The width of the samples is 6 mm and four measurements are carried out in each direction.

![Diagram showing mean intercept length in X and Y directions for samples Cs5L1 and Cs5L11.](image)

**Figure 4.40** In-plane sections of samples Cs5L1 and Cs5L11, mean intercept length are measured in X and Y directions, in micron

The non-uniform distribution of the porosities along the Y-direction in sample Cs5L11 is evident. The red dashed ellipse and rectangle in Figure 4.40 account for 70% of the area fraction of porosities in each in-plane section. The area fraction of the porosities is 1.3
and 3.5\% for samples Cs5L1 and Cs5L11, and the fracture strain of the specimen Cs5L11 is half of sample Cs5L1. The average distance between the porosities within the red dashed rectangle is almost one third of the mean distance in the red dashed ellipse. The mean distance is less than 100 \( \mu m \) for the sample Cs5L11 along the Y-direction. The comparison between the number and size of the pores indicates that the size of pores within the sample Cs5L11 is two times larger than that of the Cs5L1. Although there is no significant difference between the numbers of the pores within both samples, more than 80\% of the porosities are accumulated in one side of sample Cs5L11.

In spite of the fact that the clear insight about the distribution of the pores cannot be derived from the measurement of the average distance between the pores in a single plane, the basic definition of the randomness assists us to perceive the various distributions of the porosities. Assuming each field of measurement is the top plane of a cubic, and the pores within the field of view are the only pores in the volume, then the intercept lines would be the average distance to the nearest pore \( d \). Then the equivalent volume in the field of view is 4.8 mm\(^3\). The number of pores \((N)\) in the field of view can be considered as same as the number of pores within the equivalent volume. Then the randomness parameter (RP) for the field of view can be determined by the following expression [50]:

\[
RP = 1.23d(\pi \frac{N}{4.8})^{\frac{1}{3}}
\]

\( (4.7) \)

The randomness parameter is about 0.48 in the red dashed rectangle, and it’s about 0.94 for most fields of sample Cs5L1. The RP value is 1 for random distribution, and is smaller than unity as the tendency in the formation of clustered pores is increasing [50]. Moreover, the microscopic images of the vertical and longitudinal sections show that the relatively uniform pore distribution is more common in the samples cut from Locations 1 and 15, and more consistent mechanical results are obtained from these samples. On the other hand, the non-uniform distribution of pores in Y-direction is observed within the samples cut from the knit line region, and significant scatter among the tensile results are obtained from the same region. Although it can be assumed that the steep reduction in the
average distance between the pores is mainly originated from non-uniform distribution of the porosities, the coalescence and evolution of the porosities during the tensile test should be taken into account. In other word, the possibility of the existence of pores close to the skin region is higher for the samples which are extracted from the knit line region. The growth and coalescence of these pores along the direction perpendicular to the load axis leads to the lower fracture strain.

Similarly, four in-plane sections of the samples cut from locations 1, 10 and 11 are analyzed, and the results are shown in Figure 4.41 (The microscopic pictures of the in-plane sections are presented in Appendix A). The characterizations of porosities are performed by determination of the roundness and the area fraction of the porosities with respect to the distance from the fracture surface. The circularity of porosities can be characterized by two different parameters. The roundness parameter can be applied to 2D images and it is similar to the sphericity coefficient for 3D images. The following equation is the basic definition of the roundness factor in IA software. As the pores are not the perfect circle, the feret (L) is replaced with diameters in most microscopic characterization. The non-uniform objects can be restricted by different parallel planes, and the average distance between these planes is the feret diameter.

\[
RoundnessFactor = \frac{4 \times Area}{\pi \times L \times L}
\]  

(4.8)

where L is the longest feret diameter, and the roundness is equal to 1 for a perfect circle object. There is an increase in area fraction of the porosities in the vicinity of the fracture surfaces, while the roundness of the porosities is decreased in all four samples. The growth and coalescence of the porosities are the main reason for the lower degree of roundness among the porosities close to the fracture surface. Despite the variation of the roundness in all samples is relatively similar, the significant increase in area fraction of porosity is observed in the sample Cs5L11.
Figure 4.41 Variation of the area fraction of the porosity (column bars), and roundness (Lines) from the fracture surface of the samples

It’s important to note that, the area fractions of the porosities for the other samples are relatively constant in a distance of 3 mm from the fracture surface. If the steep variation in the area fraction of porosities of sample Cs5L11 is implemented in the failure model, the resulted deviation would be less than 10%.

Figure 4.42 shows the variation in size of the largest pores as the distance from the fracture surface is increasing. Both samples Cs4L10 and Cs5L11 contain larger pores, and the size of the pores in the samples cut from the knit line region is almost twice of the pores in samples cut from the location1. The largest pore has been detected in Cs4L10 and its radius is about 125 μm. The fracture strain of the samples Cs5L1 and Cs7L1 are 7.7% and 6% respectively, while the fracture strains of both knit line samples are almost the same and it’s about 4%.
According to the obtained results (Figure 4.41, microscopic images of in-plane sections), the area fraction of the porosities which is determined on the plane below the fracture surface can reasonably apply to the prediction of fracture behavior. The non-uniform distribution of the porosities in Y-direction implies that the proximity of the equivalent pore to the edge of samples should be taken into account. The high tendency of clustering is also observed for these types of samples. The presence of the large pores and a higher fraction of the porosities are the main reasons for low ductility and high scatter in fracture strains of the samples which are extracted from the knit line region. Moreover, the slight variations of porosity level from the fracture surface indicate that the fracture strain can be reasonably predicted by the failure model.

4.3.5 Characterization of the Porosities along the Flow Path

To evaluate the formation and distribution of porosities across the casting, the microstructure of the casting along the path of melt is studied. A number of metallographic specimens are extracted from the flange gate of knit line region to the overflows of the bolster (Figure 3.8-JK and Jll locations).
Figure 4.43 Defect bands are formed parallel to the wall of the flange gates, the red arrows show the flow path

Figure 4.43 shows the sections of the flange gate which is extracted from the JK section. The thick defect bands stretch from the flow entrance to the thinner sections of the casting, and they contain a high density of porosities. The large pores and crack-shaped pores are also observed within the flange gates. The porosities which are detected in the flange gate samples cut from the JK location are larger than that of the specimens cut from the Jll location (Appendix C-FigureC.5-10). The area fraction of porosity of the samples JK2 and JK3 are 3.6 and 4.3%. According to the mechanical tests which are conducted in Meridian Inc, these regions are weak points of the bolster. The flange gates receive the last portion of the melt, and it contains a high fraction of ESGs. The coherency and maximum packing point occurs at a lower solid fraction for mush zone containing dendritic solid particles. It’s also demonstrated that the crack-like features,
and large pores within the entrance region of melt is attributed to the deformation of mush containing high fraction of solid particles, and these pores can be categorized as shrinkage pores [26,38]. In addition to the deformation of mush with a high fraction of solid state, lack of sufficient melt leads to the formation of the large pores and thick defect bands. The high fraction of solid state in mush can lead to the interruption of flow and unsuccessful ejection of gas from the die cavity (Appendix C-Figure C.14).

The in-plane sections of flange gate which are extracted from the JK region indicate that the pores are widely distributed across this region (Figure 4.44). Due to the localized distribution of solid particles, the curved and parallel shear bands are also formed along the curved flow patterns. The related microscopic images are presented in Appendix C (Figure C1-4).

![Image of microscopic image](image_url)

**Figure 4.44** High fraction of pores within the in-plane section of flange gates cut from JK section, the area fraction of the porosities is 3.1%

The microscopic results indicate that the high fractions of the porosities are formed within the thick sections which are located close to the flange gate of JK region. However, large pores and thick defect bands are not observed within the upper section of JK location. The upper sections of JK mainly contain shrinkage pores, and they are located within the defect bands (The pictures are presented in Appendix C-Figure C11-13.
Although narrower defect bands are observed within the flange gates of JIl region, higher fractions of porosities are noted within the upper section of this region. These sections are closer to the overflows. Figure 4.45 shows the JIl section and the flow path of the molten metal from the flange gate (red circle – A) to the overflow. A high percentage of the porosities are observed within the sections B and C. Both sections are the place where the tensile specimens are extracted, and they’re already named as locations 10 and 11. There is a similarity between the distribution of pores within the section C and tongue and groove shaped fracture surfaces (Figure 4.36). The related figures are shown in Appendix C (Figure C15-16).

Figure 4.45 Distribution of porosities within the upper sections of JIl2 section, the magnified images of dashed rectangles are shown beside the montage microscopic images

The image processing and analyses techniques are used to determine the area fraction of the shrinkage and gas pores separately. From the microscopic images of flange gate, it
can be realized that the separation of pores by consideration of aspect ratio function [49] leads to the misinterpretation of the results. Therefore, the separation of pores is conducted by roughness factor. The roughness factor indicates the jaggedness of the microstructural features and is determined by the following expression in image analysis software.

$$\text{Roughness} = \frac{\text{ConvexPerimeter}}{\text{Perimeter}}$$ \hfill (4.9)

Where, the convex perimeter is the line attached the ferret tangent points.

$$\text{ConvexPerimeter} = \sum \text{Ferets}(2\tan) \left[ \frac{\pi}{2(\text{NumberOfFerets})} \right]$$ \hfill (4.10)

To provide a clear insight about the feature which are used to differentiate gas and shrinkage pores, the microscopic images and the values which are obtained by image analyzing are presented in Appendix A(Figure A.9-13). Figure 4.46 shows the procedure which is applied to differentiate between the gas and shrinkage pores. The pores with feret diameter larger than 20 $\mu m$ are separated from the field of view, and colored with green paint. Then, the pores with roughness larger than 0.9 is considered as the shrinkage pores (red features) and the rest are considered as small gas pores (blue features). The area fraction of each type of the porosities is determined within each field of measurement and the results are shown in Figure 4.47. The comparison between the fraction of large gas pores in section (B) and (C) shows that the percentage of the large gas pores in section (C) is almost two times larger than that of section (B). The average area fraction of the porosities is about 2.1% for section B and about 2.6% for section(C). As each tensile specimen is composed of three fields of measurement, the highest fraction of the porosity among the three fields of measurements is estimated to be 2.8%. The microscopic figures of section B and C indicate that the relatively thick defect bands are formed in the section B, while they are getting thinner within the section C, and the average area fraction of the shrinkage pores reduced from 1 to 0.5% within the section C.
Figure 4.46 Boolean operation is used for separation of gas and shrinkage pores in each field of measurement.

Figure 4.47 Variation of the area fraction of porosities in section (B) and (C) of JII2 section.
An increase in fraction of large gas pores and volume fraction of porosities is reported, once the pressure during the IP stage is reduced [26,49,52]. From the microscopic studies and results which are obtained from the characterization of the porosities, it can be suggested that the high solid fraction has deteriorated effect on the function of the IP stage.

4.3.6 Prediction of Fracture Behavior

To evaluate the fracture behavior of die-cast samples of AM60, the failure model which is proposed by J.Weiler [10,11] is used to predict the fracture stress and strain of the samples which are extracted from different locations of the bolster and new instrument panel. The fraction and distribution of the porosities are determined from fracture surface, plane below the fracture surfaces of tensile specimens as well as the information which is obtained from the analyses of the location before tensile test.

4.3.7 Prediction of Fracture Strain by Analysis of Fracture Surfaces

By considering the porosities and inclusions as initiation sites for fracture, the proximity of them to the free surface can be examined by the failure model [10,11]. The inclusions are treated as large pores. The location of the largest impurity is regarded as the location of the equivalent pore, and the ligament area fraction is determined based on the position of the inclusions (equation(2.19)). Figure 4.48 shows the coordinates of the large inclusions on the fracture surface. The area fractions of the inclusions and porosities as well as the position of the largest impurities are tabulated in Table 4.4.

Figure 4.48 Location of the largest inclusion within two fracture surfaces
Table 4.4 Predicted fracture strain using two different assumptions

<table>
<thead>
<tr>
<th>Sample</th>
<th>$f$ %</th>
<th>$d_w$ mm</th>
<th>$d_r$ mm</th>
<th>$r_{eq}$ mm</th>
<th>$\varepsilon^{Predicted\ CenteredPore}$</th>
<th>$\varepsilon^{Predicted\ FailureModel}$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cs5L8</td>
<td>1.9</td>
<td>1.5</td>
<td>1</td>
<td>0.225</td>
<td>5.2</td>
<td>4.2</td>
</tr>
<tr>
<td>Cs1L7</td>
<td>0.3</td>
<td>1.5</td>
<td>1.1</td>
<td>0.11</td>
<td>9.1</td>
<td>7.7</td>
</tr>
<tr>
<td>Cs1L8</td>
<td>0.6</td>
<td>1.2</td>
<td>0.6</td>
<td>0.1</td>
<td>7.4</td>
<td>1.9</td>
</tr>
</tbody>
</table>

To evaluate the effect of the location of an inclusion, the fracture strain is predicted with two different assumptions. The first prediction is made by assuming the centered pore condition, and the actual position of the largest impurity assumed as the location of equivalent pore in the second approach. The fracture strain is predicted using the generalized failure model (equation (2.20)). The comparison between the predicted values shows that there is a reduction of 20% in elongation, as the area ligament fraction of the large inclusion is considered for prediction of fracture strain. The predicted values of fracture strains are relatively accurate for the first two samples, the relative deviations from the experimental values are 4 and 12 %, respectively. The main sources of deviation can be related to the measured area fraction of the porosities on the fracture surface. The porosities appear like black spots on the fracture surface, once a large number of the porosities are gathered in a small region their appearance would be dirty or sooty in low magnification. As a result, inaccuracy in the measurement of the area fraction of the porosities leads to the deviation in predicting the value of elongation. However, the deviation from the predicted value is higher in the third sample which contains clustered pores below the free surface (Figure 4.39). The effects of subsurface pores on the prediction of fracture behavior of the samples are explained in the next section.

4.3.8 Prediction of Fracture Strain by Analyzing of Plane below Fracture Surface, and Specimens before the Tensile Test

The failure model is applied to predict the fracture strain of samples which are extracted from the bolster. As explained in Section 2.8.1, two variables should be defined for the prediction. They are skin fraction and the area fraction of porosities. The skin fractions of
the samples are determined in Section 4.1.2, the least skin thickness is about 0.27 mm for
the samples with thickness of 2.2mm, and the skin fraction would be 0.12. To determine
the area fraction of the porosities, the fracture surfaces of the tensile specimens are
removed, and the plane below the fracture surfaces are considered. There are two issues
which are involved in this approach. The first one is that the fracture surfaces are the
planes containing the highest fraction of the porosities within the tensile specimens, and it
can be expected that the plane below the fracture surface has a lower fraction of
porosities. It’s established in Section 4.3.4 that the distribution of the porosities along the
load axis is relatively uniform, and the higher fluctuation of the porosity level results in
slight deviation in predicted results (Figure 4.41). The other issue is the evolution and
growth of pores can lead to the overestimation of the area fraction of porosities. To deal
with the latter issue the auto binary function with the minimum gray level threshold is
applied, and the microscopic figures which are obtained from the planes are presented in
Appendix D.16. Also the determination of the area fraction of porosities is conducted by
Meridian Inc. by simulation of the process. The comparison between the simulated and
experimental results is shown in Figure 4.49.

![Comparison between the simulated and experimental values of the area fraction of porosities in the bolster](image-url)
Obvious differences between the simulated and experimental values for most locations of bolsters are observed. The measured average area fractions of the porosities are higher than that of the simulated results, with the exception of locations 1 and 15, which are almost half of the simulated result. The relatively similar results are obtained from the locations 3 and 14, which contain a higher fraction of shrinkage pores. According to the microscopic pictures of the plane below the fracture surfaces of knit line locations (Appendix D.16 and Figure 4.36,39,45), the high fraction of porosities is observed within the knit line region (Locations 4, 5, 10, 11 and Locations 7, 8). The increase in area fraction of the porosities within the knit line region is not detected by simulation, and mainly shrinkage pores are detectable by simulating of the process. The comparison between the results which are shown in Figure 4.47 and predicted results for knit line locations shows that almost the same fraction of shrinkage porosities are determined from both techniques.

The results which are obtained from the simulation and experiments of the new instrument panel are tabulated in Table 4.5. It can be seen that experimental values are higher than the simulation, with the exception of location 5. The highest differences between the results are obtained from the locations 2 and 3.

Table 4.5 Comparison between the simulated and experimental values of area fraction of porosities of new instrument panel

<table>
<thead>
<tr>
<th>Location</th>
<th>1</th>
<th>2</th>
<th>3</th>
<th>4</th>
<th>5</th>
<th>6</th>
</tr>
</thead>
<tbody>
<tr>
<td>Simulated Results</td>
<td>0.56</td>
<td>0.62</td>
<td>0.69</td>
<td>0.41</td>
<td>3.27</td>
<td>0.35</td>
</tr>
<tr>
<td>Experimental Results</td>
<td>1±0.3</td>
<td>0.55±0.21</td>
<td>1.2±0.18</td>
<td>0.9±0.15</td>
<td>2.1±0.9</td>
<td>0.5±0.15</td>
</tr>
</tbody>
</table>

The comparisons between the experimental and predicted values of fracture strain are shown in Figure 4.50. Two different assumptions are considered for evaluation of the failure model. Equations (2.16) and (2.17) are used to predict the fracture strain of
samples with different fractions of porosities. Also the dashed lines in Figure 4.50 represent the prediction which is made by application of constitutive and modified models (equations (2.13) and (2.14). The prediction which is made by these two models is mainly similar to the centered pore assumption of failure model, and the fracture strain of some samples is over predicted. By introduction of skin fraction as the ligament factor in our prediction, the predicted value would be almost 21% lower than the value predicted by centered pore assumption. As explained, underestimation in the prediction of fracture strain is mainly originated from the plane containing the grown pores which turn to the crack-shaped features. The microscopic figures of two labeled samples indicate that pores grow and turn to the crack-shaped features within the plane below the fracture surface of samples Cs5L5 and Cs4L9. The deviation between the experimental and predicted values for the samples Cs3L4 and Cs3L7 are about 10%. Previous explanation implies that the introduction of skin fraction enables us to predict the fracture strain of most samples with relatively high accuracy.

![Figure 4.50](image-url)

**Figure 4.50** Evaluation of failure model by experimental results obtained from the planes below the fracture surfaces of a bolster
It’s noted that some specimens cut from locations 3, 8 and 14 have some porosities in the vicinity of the free surface and the elongation of these samples are below 3%. Figure 4.51 shows that three large pores are located in the vicinity of both sides of sample Cs3L8 as a result fracture strain of this sample is about 1%. The SEM images of the samples which are extracted from locations 3 and 14 reveal that the regions containing intact dendrites are located close to the free surface of the samples (Figure 4.51 and Appendix B-FigureB.58-60). The predominant mechanism of the fracture is dictated by the high magnitude of local stress around the pores below the free surface. Therefore the microscopic images are used for the determination of the area fraction of porosities and ligament area fraction (equation(2.19))

Figure 4.51 SEM images showing the position of sub-surface pores within samples: (a) Cs3L3; (b) Cs3L14 and (c) Cs3L8 with three sub-surface pores
The ligament fraction of these samples is nearly 0.004, and the assumption of sub-surface pores is applied to predict the fracture strain of these samples. The comparison between experimental and predicted values of fracture strains by application of sub-surface pore assumption are shown in Figure 4.52.

![Fracture Strain vs Area Fraction of Porosities](image)

**Figure 4.52** Comparison between the predicted and experimental values of fracture strains considering sub-surface pore assumption

Similarly, the prediction of fracture strain for the instrument panel is conducted by determination of the area fraction of porosities within the locations which are assigned by Meridian Inc. The third row of Table 4.5 shows the skin fraction of the specimens, and the prediction of fracture strain is based on the maximum area fraction of the porosities which are obtained from the microscopic studies and simulation. The skin fraction assumption is applied for prediction of fracture strain and the results are shown in Figure 4.53. As the arbitrary planes within the specimens are not necessarily the plane with the highest area fraction of porosities, the overestimation of the fracture strain can be expected for some samples. The overestimation of fracture strain is more pronounced as
the simulated results are considered for prediction. It’s also noted that the extraction of tensile specimens from these locations can lead to the new position of the pores within the tensile specimens. For instance, there is a large pore within sample CsNL5 (associated figure is presented in Appendix D), and the relocation of this pore in the vicinity of free surface is possible if the tensile specimens are extracted from this location.

![Figure 4.53 Predicted fracture strain for new instrument panel using the information obtained from experiments and simulation](image_url)

To set an appropriate lower bound for the bolster the highest average area fraction of the porosities which is obtained from section C (Figure 4.45) is considered as a reliable criterion. The resemblance between the fracture pattern of the least ductile specimens (Appendix D) and distribution of the porosities within this section, as well as high degree of scattering in mechanical properties of the tensile specimens from this location are the main reasons for this selection. The highest average area fraction of the porosities is about 2.8% in section C, the lowest skin thickness among the samples cut from that region is about 0.27 mm, and the average thickness of the samples is 2.2 mm. As a result, the predicted fracture strain is about 3.7% by assuming skin fraction condition.
Figure 4.54 shows the comparison between the predicted and experimental values. The green dashed line defines the lower bound of the fracture strain, and the blue line represents the lower bound of the fracture stress.

Figure 4.54 Variation of the mechanical properties across the bolster and the lower bounds of the fracture stress and strain

The predicted lower bound matches properly with most of the experimental values. The exceptions are locations 3, 8 and some samples cut from location 14. The low ductility of these locations was discussed in previous sections, matches with the experimental values which are obtained from different locations.
Chapter 5

5 Summary

The characterization of two complex die casts of AM60 and validation of the models which are proposed for the prediction of yield strength and fracture behavior were the main purposes of this study. The effects of grain size diversity on the yielding behavior of the die-cast of AM60 are also studied using the new approach. The experimental data which is obtained from the mechanical and microscopic studies are applied for validation of the models. To achieve these goals, metallography, image analyses and image processing techniques were utilized to characterize the microstructural features within different location of the castings.

As-Cast Microstructure:

In the first step, characterization of grains within the cross section of samples was carried out with the assistance of optical microscopy and Image Analysis software. The analyses of grain distribution revealed that the skin fraction is not uniform within the cross section of samples which are removed from the castings. A higher degree of non-uniformity in the skin thickness is obtained from the knit line specimens of the bolster. The different heat flow rate across both sides of the samples can lead to different solidification conditions. The smallest skin thickness was found to be 0.27mm among these samples. The average grain size ranges from 5.5 to 7 μm within the skin region, and it ranges from 7.5 to 9 μm within the core region. The large and branched grains within the core region of samples with higher thickness are detected. Both experimental and predicted sizes of the grains confirm that the average grain size is increasing while the cooling rate is decreasing. However, the predicted value which is obtained by simulating the process was higher than the experimental results. The distributions of large dendrites were different across the bolster and two different types of distribution patterns are distinguished. The first one was the accumulated ESGs within the core region of samples, and local average grain size is increased up to 16 μm. The dispersed pattern is
distinguished as second type, and didn’t have a significant effect on the local average grain size.

As-polished specimens were observed using optical microscopy in order to characterize the level of porosity within the specimens. The microscopic images of in-plane sections of the tensile specimens revealed that the variation of porosity level up to the regions close to the fracture surface is relatively uniform along the load axis. Although the higher deviation of the porosity level is obtained for some knit-line samples, such deviation has no significant effect on the prediction of fracture behavior. The area fraction of porosity within the knit line regions ranges from 0.4 to 2.8%, and results in a high fluctuation of the fracture strain. The comparison between the experimental and simulation results indicate that the proper estimation of the pores’ fraction cannot be made by simulating the process. This is because simulation software is not able to predict gas entrapment. Moreover, non-uniform distribution of pores and higher tendency of clustering is established in these regions. The characterization of porosity before tensile testing revealed that, there is an abrupt increase in fraction of the large gas pores within the sections closer to the overflow position. The average area fraction of the large pores is almost two times greater than that of the sections which are far away from the overflow position, and the porosity level reached values up to 1.4% in sections close to the overflows. It’s also noted that the fraction of shrinkage porosity within these samples were relatively similar to the value which is obtained from the simulation of process. This increase can be related to the inefficiency of the IP stage of die-casting process. Moreover, the presence of large pores and crack-like features within the flange gate is attributed to a higher fraction of solid particles, and can lead to the unsuccessful performance of the IP stage.

The characterization of the intermetallic phase is performed with the assistance of image processing software. The area fraction of the intermetallic phase ranges from 1.5% to 4.5% within the skin region, while it’s between 1.2% and 3% within the core region. Due to the accumulation of ESGs within the core region of samples which are extracted from the close-to-ingate region, the distance between $\beta$ particles is increased by a factor of 1.3.
Yield Strength:

The yielding strengths of the samples are predicted using the modified form of the Hall-Petch equation. The predicted yield strength of the samples which are extracted from the new instrument panel well matched with experimental values. The comparison between the experimental and predicted yield strength of the bolster indicate that the difference between two values is about 10%, and it can be related to the higher fraction of shrinkage pores which are located within the shear bands of the samples. The parameters of Hall-Petch equation is obtained by plotting the yield points versus the average grain size of the samples. The values which are predicted by general and modified form of the Hall-Petch equation are compared with each others. As the diversity of grain increases within the samples, the prediction of yield point by the general form of the model leads to a deviation of more than 10%. The variation of the local grain size within the sample is taken into account as the modified form of the model is utilized. As a result, the deviation between the experimental and predicted values were less than 6%. Moreover, the different distribution patterns of ESGs affect the yield strength of the samples. The effect of grain size diversity represent as the fraction in modified form of the Hall Petch equation (2.12).

The decreasing trend of yield point with the fraction of ESGs indicates that as the percentage of the ESGs increase from 0.2 to 4%, the yield strength of samples reduces from 113 to 96 MPa. The effect of grain size diversity is also examined by analyzing of strain-hardening rate of the samples with different grain size distribution. The new approach assists us to appreciate the effect of local average grain size by determination of tangible values including the fraction of elastic material and onset of fully plastic behaviors. The variations of strain-hardening rate of the samples with dispersed pattern of ESGs were relatively similar to the samples with fine microstructure with the exception of the accelerated recovery stage at a stress higher than the yield point (0.2% offset). The onset of the recovery stage occurs at much lower stress for the samples with accumulated ESGs within the core region. The accumulation of the ESGs in the core region of specimens leads to the decrease in the elastic fraction of the samples by a factor of 1.7.
Moreover, the elasto-plastic transition point of these samples is lower than that of the samples with fine grain structures. The fully plastic behavior of the samples with fine structure and dispersed pattern of ESGs occurs at higher stress values by a factor of 1.3.

**Fracture Strength:**

The fracture behaviors of samples are predicted by failure model. The area fraction of porosity is determined from the plane below the fracture surface of samples, and the skin fractions of samples are introduced as the ligament factor for prediction of fracture behavior. The comparison between the predicted and experimental results indicates that the failure model is able to predict the fracture strain of samples with high accuracy. The deviation between prediction and experiment is less than 10% for most samples which are extracted from the bolster. The microscopic studies and SEM images revealed that some samples contain sub surface pores. The prediction of fracture behavior is made by utilizing the ligament fraction which is computed from the actual position of pores. The failure model is applied to predict the fracture strain of these samples, and the predicted results well matches with the experimental values. Moreover, the lower bound for fracture properties can be defined by application of the data which is obtained from the microscopic studies of the knit line region before tensile test.

According to the results which are obtained from the mechanical and microscopic experiments, it’s established that the failure model is applicable for the die cast products, and it’s able to predict the fracture properties of the casting with high accuracy. The practical way for determination of area fraction porosities is introduced. The results obtained from the microscopic studies of sensitive regions such as knit line can be applied as input variables of failure model. As a result the lower bound of ductility can be defined properly by failure model.
References or Bibliography


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Appendices

Appendix A: Variation of Local Mechanical Properties & Fraction of Porosity

This section includes the detailed results of the uniaxial test, density measurements. The microscopic pictures of inplane sections, and the procedure which is taken to characterize the microstructural features are presented in this section.

1. The typical stress- strain curve of die-cast sample, the 0.2% off set is sketched to determine the yield point(). The variation of the mechanical properties are shown in , the weak points are highlighted by pink color, and the highest fluctuation is highlighted by blue color.

![Stress-strain curve](image)

Figure A. 1 The engineering stress-strain curve of the sample cut from the bolster
Table A.1: The variation of mechanical properties across the bolster

<table>
<thead>
<tr>
<th>Location</th>
<th>σy (Mpa)</th>
<th>σuts (Mpa)</th>
<th>εf (%)</th>
<th>K</th>
<th>n</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>107.5±2.9</td>
<td>192±17.7</td>
<td>5.9±1.3</td>
<td>370</td>
<td>0.232</td>
</tr>
<tr>
<td>2</td>
<td>110.9±4.9</td>
<td>185.6±5.1</td>
<td>6±0.6</td>
<td>347</td>
<td>0.217</td>
</tr>
<tr>
<td>3</td>
<td>108.5±5.8</td>
<td>148.7±6.4</td>
<td>2.6±0.7</td>
<td>260</td>
<td>0.153</td>
</tr>
<tr>
<td>4</td>
<td>113.2±5.7</td>
<td>198.7±12.5</td>
<td>6.1±0.7</td>
<td>387</td>
<td>0.232</td>
</tr>
<tr>
<td>5</td>
<td>111.1±5.5</td>
<td>199.4±17.2</td>
<td>6.6±1.1</td>
<td>399</td>
<td>0.243</td>
</tr>
<tr>
<td>6</td>
<td>103.6±5.6</td>
<td>191.4±14.2</td>
<td>7±1.3</td>
<td>396</td>
<td>0.261</td>
</tr>
<tr>
<td>7</td>
<td>113.2±4.7</td>
<td>204.5±23.4</td>
<td>7±1.9</td>
<td>417</td>
<td>0.252</td>
</tr>
<tr>
<td>8</td>
<td>115.3±7.6</td>
<td>157.2±12.7</td>
<td>2.4±1.2</td>
<td>280</td>
<td>0.156</td>
</tr>
<tr>
<td>9</td>
<td>109.2±6.1</td>
<td>181.3±13.7</td>
<td>8.5±1</td>
<td>424</td>
<td>0.268</td>
</tr>
<tr>
<td>10</td>
<td>111.1±8.5</td>
<td>194.7±23.6</td>
<td>6.5±1.4</td>
<td>403</td>
<td>0.248</td>
</tr>
<tr>
<td>11</td>
<td>113.4±7</td>
<td>188.8±30.6</td>
<td>5.7±2.1</td>
<td>417</td>
<td>0.252</td>
</tr>
<tr>
<td>12</td>
<td>101±3.5</td>
<td>176.2±9.7</td>
<td>5.8±0.8</td>
<td>360</td>
<td>0.243</td>
</tr>
<tr>
<td>13</td>
<td>109±4.4</td>
<td>179.5±17.6</td>
<td>5.6±1.3</td>
<td>357</td>
<td>0.226</td>
</tr>
<tr>
<td>14</td>
<td>110.7±6.1</td>
<td>158.7±17.5</td>
<td>3.4±1.8</td>
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<td>0.182</td>
</tr>
<tr>
<td>15</td>
<td>107.2±3.7</td>
<td>174.5±16.3</td>
<td>4.9±1.3</td>
<td>341</td>
<td>0.213</td>
</tr>
</tbody>
</table>
2. The local volumetric porosity of each location

![Figure A. 2 Density profile for each location of the casting](image)

3. Application of intercept line to determine the average distance from the neighbouring pore: The analyses of microscopic pictures of in-plane sections

![Figure A. 3 The intercept lines are sketched in two directions and Boolean functions is applied to determine the average distance](image)
4. The in-plane cross sections of other samples cut from the gauge of the tensile specimens:

Figure A. 4 The longitudinal cross section of the sample Cs3L3(a) and Cs5L11(b), the area fractions are 0.9% and 3.5% respectively.

Figure A. 5 The longitudinal cross section of the samples Cs7L1(a) and Cs4L6(b), the calculated area fractions are 1% and 1.2% respectively.
Figure A. 6 The longitudinal cross section of the samples Cs4L10(a) and Cs5L5(b), the calculated area fractions are 3.7% and 2.8% respectively.

Figure A. 7 The longitudinal cross section of the samples Cs1L6 (a) and Cs3L8 (b), the calculated area fractions are 0.7% and 0.5%, respectively.

5. The number of the porosities in-plane section of the different samples are shown by column bars. The number of the porosities is decreasing in the planes far away from the
fracture surface, the steep decrease is observed in sample Cs4L10. Data points are the radius of the largest pore in Figure 8, and overall slight decrease in the radius of the largest pore is obtained in planes far away from the fracture surface.

![Graph showing pore distribution](image1)

**Figure A.8** The distribution of the pores across the in-plane sections cut from different locations

6. **Characterization of the porosities:** The gas and shrinkage porosities are characterized using various criteria including the aspect ratio, roundness and sphericity.

![Images of pore structures](image2)

**Figure A.9** The sphericity coefficient of the shrinkage cavities were below 0.5
Figure A. 10 The roundness of the gas pore is considered between 0.5 to 1, the roundness of an ideal circle is equal to one, and the average roundness of the shrinkages were below 0.4, and the average aspect ratio of the shrinkages were more than 1.5.

Figure A. 11 the propagation of the crack from the defect band of the sample Cs2L3
Figure A.12 The higher aspect ratio is the characteristic of the shrinkage pores, arrow indicate the location of the shrinkage porosities

Figure A.13 Shrinkage porosities are formed close to the gas pores
Appendix B: Microscopic Images of Fracture Surfaces

To provide clear insight about the fracture behavior of the samples, the images of the fracture surfaces are also presented.

Figure B.1 Cs1L1 inclusions and pores in middle

Figure B.2 Cs2L1

Figure B.3 Cs3L1 Crack & brittle fracture - light angle 45

Figure B.4 Cs4L1

Figure B.5 Cs5L1

Figure B.6 Cs6L1 Higher density of pores in the middle

Figure B.7 Cs7L1

Figure B.8 Cs1L3 Porosities and inclusions

Figure B.9 Cs1L3 Different magnification
Figure B.10 Cs2L3

Figure B.11 Cs2L3

Figure B.12 Cs2L3 Inclusions -higher magnification

Figure B.13 Cs3L3 Crescent shape crack

Figure B.14 Cs4L3

Figure B.15 Cs5L3

Figure B.16 Cs6L3

Figure B.17 Cs1L5 Pores and micro cracks

Figure B.18 Cs4L5 Pores and crack
Figure B.19 Cs1L6 Brittle fracture surfaces

Figure B.20 Cs5L6 Pores and micro cracks

Figure B.21 Cs1L7 Inclusions on both fracture surfaces

Figure B.22 Cs3L7 Presence of impurity

Figure B.23 Cs3L7 inclusion

Figure B.24 Cs4L7 Presence of pores and inclusions

Figure B.25 Cs5L7 Presence of pore

Figure B.26 Cs6L7 Presence of pores

Figure B.27 Cs7L7 Presence of inclusion and pores
Figure B.28 Cs1L8 Impurities

Figure B.29 Cs1L8 inclusion different magnification

Figure B.30 Cs1L8 Inclusion nmagnification

Figure B.31 Cs2L8 Presence of impurity and inclusion

Figure B.32 Cs2L8 Presence of inclusion with different magnification

Figure B.33 Cs3L8 Presence of brittle fracture surface

Figure B.34 Cs3L8 Brittle fracture surface with different magnification

Figure B.35 Cs5L8 Large inclusion in both fracture surfaces

Figure B.36 Cs5L8 Inclusion with different magnification
Figure B.37 Cs6L8  
Impurities and pores

Figure B.38 Cs6L8  
Impurity with different magnification

Figure B.39 Cs2L10  
Pores and impurities

Figure B.40 Cs4L10  
Pores and cracks

Figure B.41 Cs5L10  
Pores and cracks

Figure B.42 Cs1L11  
Pores and cracks

Figure B.43 Cs2L11  
Pores and impurities

Figure B.44 Cs4L11  
Pores and impurities

Figure B.45 Cs5L11  
Pores and impurities
Figure B.46 Cs1L12
Brittle fracture surfaces in both fracture surface

Figure B.47 Cs1L12
Fracture surface with higher magnification

Figure B.48 Cs1L12

Figure B.49 Cs1L13
Brittle fracture surfaces under different light angle

Figure B.50 Cs1L14
Impurities on both fracture surfaces

Figure B.51 Cs2L14
Impurities on both fracture surfaces

Figure B.52 Cs4L14
Impurities

Figure B.53 Cracks and impurities on both fracture surfaces

Figure B.54 Pores and impurities on both fracture surfaces
The results obtained from the EDX analyses indicate that the inclusions of the fracture surfaces are mainly composed of calcium.

![EDX analysis graph]

**Figure B.55** The red-dashed circle shows the location that elemental analysis is performed on the fracture surface of the specimen cut from the knit line

The SEM images of the fracture surfaces indicate that there are high fractions of the shrinkage porosities are on the fracture surface of the samples cut from the location 3 and 14, and some pores are located within the skin region and they are close to the free surface.

![SEM image]

**Figure B.56** The big shrinkage cavity on the fracture surface of the sample Cs3L3
Figure B.57 The shrinkage pores on the fracture surface of the sample Cs3L3, the area fraction of the porosities is about 3.5%

Figure B.58 The shrinkage pores in the vicinity of free surface Cs3L14
Figure B.59 The large shrinkage cavity in the vicinity of free surface Cs3L14, the area fraction of the porosities is more than 5% and the fracture strain is 1.6%.

Figure B.60 The intact dendrites are evident in higher magnification images of figure B.59
Figure B.61 The red circle shows the location of the porosities within the defect band (a) and the intact dendrites are evident in higher magnification image (b)

Figure B62. SEM images of the fracture surface samples Cs1L4
Figure B.63 The red circle shows the location of the porosities within the core region of sample Cs6L1(a) and higher magnification field is also presented(b)
Appendix C: Images of Defect Bands and Porosities before Tensile Test

The microscopic pictures of curved sections revealed that the curved defect bands are formed along the contour of these sections.

Figure C1. The curved defect band is shown in montage microscopic image (1) and figures (2) and (3) are the magnified image of the defect bands and porosities.
Figure C2. Formation of the defect bands along the curve path

Figure C.3 Additional force toward the centre of the curved paths leads to the localized viscosity of the mush zone. The formation of porosities along the curved flow path is resulted from the local difference in solid fraction. The scales are 200 $\mu$m (a) and 50 $\mu$m (b)
Figure C.4 The presence of shear bands along the walls within the flange gates of bolster, due to the movement of the packing points, parallel shear are formed (a), the shrinkage porosities are formed along the shear path (b)
The flange gates which are extracted from the JK sections are containing the thick defect bands, and high fractions of the shrinkage pores are located within the bands.

Figure C.5 The red dashed rectangle shows the thick defect band of the flange gate extracted from the JK section

Figure C.6 Shrinkage pores within the flange gate sections cut from the JK sections. Scale bar in the pictures is 50 μm
Figure C.7 The defect bands of the flange gate extracted from the Jll location were thinner than JK locations

Figure C.8 Shrinkage pores within the flange gate sections cut from the Jll sections
Figure C.9 High fraction of shrinkage pores within the inplane section of flange gate

Figure C.10 The high percentage of shrinkage pores in the longitudinal section of the flange gate JK
Figure C.11 The distribution of the pores in thicker sections close to the flange gate (JK location)

Figure C.12 The microscopic images of the upper sections of JK location, they are containing low fraction of the shrinkage pores within the location of shear bands
Figure C.13 Three different magnified images of the defect band which is formed within the upper section of the JK location

The unsuccessful ejection of the gas in the region close to the overflow position of bolster is shown in Figure C.14. There is a possibility that the higher fraction of solid state in mush prevent the complete ejection of the entrapped gas from the die cavity.
Figure C.14 The gas pores in the section close to the overflow is highlighted by red rectangle. The entrapped air is not dissipated during the filling process. Scale bars are 1mm.

To provide more clear insight about the presence of larger gas pores in the sections close to the overflow, larger microscopic and more informative images are presented in this section. First, the images which are associated with Figure 4.46 are presented, and then the microscopic and binary images of other sections are shown. Figure C.15 shows the microscopic images of section (B) with two different magnification. The decreasing tend in fraction of shrinkage porosities is evident, Figure C.15-1 is one of the microscopic images of section(B) and Figure C.15-2 ,3 are obtained from section(D). Figure C 15-2 is the field closer to the section(B).
Figure C.15 The microscopic images of section (B) with two different magnification, defect bands are formed in section (B), The higher fraction of large gas pores and lower fraction of shrinkage are observed in section closer to the overflow (2,3)
Figure C.16 shows that the higher fraction of shrinkage porosities are formed within the Section B, while higher fraction of large pores are observed within the Section C.

**Figure C.16** The comparison between the microscopic figures which are obtained from Section B and C of Jll region
Appendix D: Variation of Grain Size across Various Sections

The microscopic images which are applied to determine the average grain size of different sections. In addition to the figures which are presented in section 4.2.3, the variations of the grain size across the cross section of other samples are presented in this section. The procedures which are taken to characterize the deviation between the different grain size measurement techniques are presented in this section.

1. The deviation between the intercept method and automatic measurement of the grain size.

![Image of microscope view](attachment:image.png)

**Figure D.1** The intercept lines are shown with different colors, the difference between the two methods are compared with each other
2. To analyze the effect of the number of grains on the grain size measurement for the field with lower grain diversity, three fields of view from three different specimens are considered, and the following procedure is taken to explore the deviation of the grain size measurement. Grain size measurement is performed using the object and field measurements. The difference between two techniques was less than 0.5 \( \mu m \)

![Figure D.2 The number of grains is reduced in each measurement](image)

![Figure D.3 Variation of the measured grain size with the number of grains in each field of measurement](image)
3. The variation of the grain size across the samples which are extracted from different regions of bolster.

Figure D.4 Variation of grain size across sample Cs4L12, red dashed line is the average grain size, blue lines are defined as skin-core boundaries

Figure D.5 Variation of grain size across sample Cs6L13, red dashed line is the average grain size, blue lines are defined as skin-core boundaries
Figure D.6 Variation of grain size across the sample Cs1L3, red dashed line is the average grain size, blue lines are defined as skin-core boundaries.

Figure D.7 Variation of grain size across the sample Cs1L7, red dashed line is the average grain size, blue lines define skin-core boundaries.
Figure D.8 Variation of grain size across the sample Cs7L1, red dashed line is the average grain size, blue lines define skin-core boundaries.

Figure D.9 Variation of grain size across the sample Cs5L5, red dashed line is the average grain size, blue lines are define skin-core boundaries.
4. The variation of the grain size across the samples which are extracted from different regions of the new instrument panel.

Figure D.10 Variation of grain size across sample CsNL3, red dashed line is the average grain size, blue lines define skin-core boundaries

Figure D.11 Variation of grain size across the sample CsNL6, red dashed line is the average grain size, blue lines define skin-core boundaries
5. Microscopic pictures of the sample Cs7L11.

Figure D.12 The comparison between the both skins located in both sides of sample
Figure D.13 The microstructure of the core region within sample Cs7L11, the variation of the grain size is low with the exception of ESGs.

Figure D.14 The variation of microstructure toward the core region is gradual.
Figure D.15 Microstructure of the core region and indentation marks within the core sample Cs7L11

Figure D.16 The fields of view which are used for grain size measurement

Figure D.17 The ESGs are distributed across the cross section of the sample, skin reveals slightly darker.
Figure D.18 The ESGs can lead to the localized increase in grain size measurement
Figure D.19 The skin-core boundary can be recognized easily

7. Microscopic pictures of the sample Cs4L3.

Figure D.20 The location of the shear band matches well with analyses of grain size variation, the texture of the surface leads to the non-uniform skin thickness
Figure D.21 The microstructure of the core region within sample Cs4L3

Figure D.22 The microstructure of the skin region of sample Cs4L3
Figure D.23 Fields of view which are used for grain size measurement Cs4L3

8. Microscopic pictures of sample CsNL2.

Figure D.24 The difference between the microstructure of skin and core region in sample CsNL2
Figure D.25 Fields of view used for grain size measurement in sample CsNL2


Figure D.26 The difference between the microstructure of skin and core region in sample CsNL4
Figure D.27 Fields of view used for grain size measurement in sample CsNL4

10. Microscopic pictures of sample CsNL3.

Figure D.28 The presence of large pores in sample CsNL3

11. Microscopic pictures of core region of the sample which extracted from the location 9 of the bolster.
Figure D.29 High fraction of ESGs within the core region of these samples lead to the increase of average grain size in this region.
The Microstructure of the different sections from the flange gate to over flow

Figure D.30 Longitudinal section of the flange gate with two different magnifications (JK sample)
Figure D.31 The microstructure of the flange gate with two different magnifications (JK sample)

Figure D.32 The gradual increase of grain size can be observed from the edge to core region of the flange gate cut from the JK location.
Figure D.33 Three microscopic images from the two parallel shear bands formed along one side of the flange gate JK, it indicates the occurrence of two high solid fraction during the injection of the melt.
Figure D.34 The microstructure profile of the specimen cut from JLL10, R1 and R7, showing the microstructure of the sample in the skin region.
To clarify the effect of sample thickness the variation of the grain size across the flange gate and microscopic images of the samples which are extracted from the locations 1 and 5 of the instrument panel are presented here. The thicknesses of the samples are
greater than 4mm and average grain size is higher than thin samples. Figure D.37 shows the field of view which is obtained from the location of the new instrument panel.

**Figure D.36 Variation of grain size across the flange gate cut from JK location**

**Figure D.37 Variation of grain size across the flange gate cut from Jll location**

Due to the penetration of the ESGs in the skin region of the flange gate, a high scatter amongst the grain size measurements is obtained. Considering the calculated deviation
between the different techniques of grain size measurement, the average grain size within the core region of the thick samples is more than 9 \( \mu m \).

![Microscopic images](image)

**Figure D.38** the microscopic images of the core region of the sample which is extracted from the location 5 of the new instrument panel

A higher number of branched grains are observed within the core region of the thick samples.

**13.** The intercept technique is applied to determine the average distance between the intermetallic phases.
Figure D.39 The image processing and Boolean function used to measure the distance between the intermetallic phases

14. To set the gray level for the binary operation the comparison between the microscopic and binary image is performed. Figure D.39 shows the difference between the application of the auto-binary operation and the method which is applied for the images containing small microscopic features.

Figure D.40 The microscopic image before applying binary operation
The comparison between two techniques indicate that higher fraction of the porosities is obtained from the second method. The area fraction of the porosities is about 0.6% from the auto-binary results and it's about 1.1% from the second technique. It should be noted that setting the gray value of binary operation manually sometimes leads to the overestimation.
Figure D.43 the area fraction of the porosities are 1.9 and 2.6 percents in (A) and (B) respectively, according to the original picture there is an overestimation in figure(B)

Therefore both techniques are considered for binary operation of the microscopic images and binary images are compared with original pictures to ensure the proper binary operation.
Figure D.44 Both techniques are applied, and the comparison is made to set the gray value for binary operation, the lower percentage of 1.8% is obtained once the auto-binary image was the choice.
15. The simulation results are presented in this section. Figure D.45 the solidification rate is almost constant across the knit line locations and ranges from $200^\circ C/S$ to $250^\circ C/S$. The thickness of samples is almost 2.2mm. Figure D.46 the solidification rate ranges from $150^\circ C/S$ to $180^\circ C/S$ across the locations 2,3,4,of the new instrument panel. The thickness of these samples are 2.2, 2.7 and 2.8 mm respectively.
Figure D.47 the solidification rate ranges from 90°C/S to 120°C/S across the locations 1 and 5 of the new instrument panel. The thickness of these samples are 4.4 and 4.7 mm respectively.

Figure D.48 the solidification rate ranges from 90°C/S to 120°C/S across the locations 6 of the new instrument panel. The thickness of these samples is 2mm.
16. The microscopic images of planes which are applied to determine the area fraction of the porosities are presented in this section. First, the planes beneath the fracture surfaces of the tensile specimens which are cut from the bolster are shown.

![Figure D.49](image)

**Figure D.49** the figures which are obtained from location 1 and 15. The cross section of these samples are $2.2 \times 6\text{mm}$, and the scale bares are $200\mu\text{m}$.

The dispersed and uniform distribution of the porosities is the common characteristic which is observed within the samples which are extracted from the locations 1 and 15. For instance the distance of the large pore within the sample Cs3L15 from the edge of the sample is $1045\mu\text{m}$. The average distance of the large pores from the edge of the samples Cs7L1 and Cs5L1 is $960\mu\text{m}$ and $890\mu\text{m}$. Therefore, it’s most likely that the centered pore assumption of failure model can perfectly predict the fracture behavior of these samples.
Figure D.50 the figures which are obtained from the knit line location with two overflow nearby. These locations are 4, 5, 10 and 11. The cross section of these samples are $2.2 \times 6\text{mm}$, and the scale bars are $200\mu\text{m}$.

The red dashed circle is the location of the large pores which are usually observed within the samples of knot line location. They are the preferred sites for crack initiation and red arrows show the propagation path of the crack [74]. A high scatter in area fraction of the porosities leads to the significant difference in fracture behavior of the samples. As the pores are turned to the crack type features, it’s more likely that determination of area fraction of the porosities leads to the overestimated values.
Figure D.51 the figures which are obtained from the knit line location with three overflow nearby. These locations are 7 and 8. The cross section of samples are nearly $2.2 \times 6\, \text{mm}$, and the small scale bares are $200\, \mu\text{m}$.
The sub-surface porosities are observed within the samples which are extracted from this location. The microscopic figure of the subsurface and shrinkage pores are shown in figure D.50.

Figure D.52 The first row shows the pores beneath the surface of the sample Cs1L8, and the jogged characteristics of shrinkage pores are shown within the sample Cs2L8 (second row).

The microscopic pictures of locations 3 and 14 revealed that they are containing higher fraction of the shrinkage pores, and higher density of pores are observed within the location of shear bands. Figure D.51 shows that defect bands are the preferred locations for crack initiation (Sample Cs2L3). The microscopic pictures of shrinkage pores with higher magnifications are shown in Figure D.52.
Figure D.53 High fraction of shrinkage pores within the locations 3 and 14. The cross section of samples are nearly $2 \times 6\text{mm}$, and the skin thickness is decreasing where the width of the sample is decreasing.
Figure D.54 The microscopic picture of shrinkage pores within the samples which are extracted from location 3 and 14

Figure D.53-54 shows the microscopic pictures of locations 6, 9, 2, 12 and 13. The red dashed rectangle shows the grown pore within the sample Cs4L9 (Figure D.53). The row of black spots within the sample Cs6L13(Figure D.54) is the location of indentation test.
Figure D.55 The microscopic picture of location 6 and 9 (2.2×6mm cross section)

Figure D.56 The microscopic picture of locations 2, 12 and 13. The cross section of samples are 2.2×6mm
Figure D.57 The microscopic picture of samples cut from different locations of new instrument panel
Figure D.58 The high magnified figures of new instrument panel
Appendix E: ESGs and Hardness Test

The microscopic pictures and graph which is associated with the hardness test and presence of ESGs are presented in this section.

The Vickers hardness was performed on the various samples. Due to the limitation which is resulted from the intended area and the minimum distance between them, the flange gate with high cross section is chosen to appreciate the variation of hardness. A load of 300 $gf$ is applied during the hardness test of the flange gate. The hardness map and variation across the section of the samples are shown in Figure E.1.

Figure E.1 Slight reduction of hardness in the core region of flange gate cut from JK section, the red dashed rectangle showing the locations that indenter penetrated through the pores

Although the lower average hardness within the core region is common among the samples, high discrepancies among the results were evident. The higher hardness within the skin region is mainly resulted from the fine grains, a higher fraction of intermetallic phase as well as the smaller interconnection of the intermetallic particles. The mean Vickers hardness is 64 within the skin of flange gate, while it is around 61 in the internal
region. Although the slight decrease of hardness in the core region is obtained, the results are not consistent. The dramatic decrease of the hardness is related to the presence of the subsurface porosities, which are collapsed because of the force applied during the Vickers hardness test. In spite of the presence of ESGs within the core regions of samples Cs1L9 and Cs4L9, the significant decrease within the core region of theses samples is not obtained. The Vickers hardness within the core region of sample Cs1L9 ranged from 63 to 67, while it was $67 \pm 6$ for sample Cs4L9. The difference in local distribution of $\beta$ phase and the variation of grain size resulted in the variation of the intended area. Therefore, it’s probable that the intended area does not reflect the average properties of the region. The localized hardness results are also reported by other researchers [68].

Penetration of the indenter into the locations close to the defect bands lead to the dramatic decrease in the hardness.

![Contact of indenter with porosities along the defect band](image)

**Figure E.1 Contact of indenter with porosities along the defect band**

2. The random location of the in-plane section of sample Cs1L9 is chosen to determine the hardness across the section, and different load cell is applied to perform the hardness test.
Figure E.2 The hardness of the in-plane section which represent the core region of the sample Cs1L9, the significant reduction in the core region is not observed.

Figure E3 shows the average hardness within the different regions of the samples cut from knit line regions of the bolster (Sample Cs3L8 and Cs6L8). A load cell of 0.1 kgf/mm² is applied to measure the hardness within the skin and core region of samples. Although considerable difference between the hardness of core and one of the skin region is not obtained, the hardness was slightly higher within the other one. It ranges from 67 to 72.

Figure E.3 Slight reduction of the hardness in the core region of samples Cs3L8 and Cs6L8.
3. The fractions of the ESGs within the cross section of the tensile specimens are determined. The image processing is performed by erode function of image analyses software, then Boolean function is applied for separation of large grains.

Figure E.3 The ESGs are differentiated from the small grain in inplane section of the sample which is extracted from the close to the ingate region

Figure E.4 the fraction of the ESGs are determined within in-plane section, the fraction of the ESGs within the plane below the fracture surfaces is also determined
4. The core region of close-to-ingate samples are composed of ESGs and fine grains. Figure E.5 shows in-plane section of the sample Cs1L9 and grains with different grain diameter are colored to show the distribution of grains across the core region of sample.

<table>
<thead>
<tr>
<th>Color</th>
<th>Grain Size(GS)µm</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>GS&gt;35</td>
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<tr>
<td></td>
<td>25&lt;GS&lt;35</td>
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<td>15&lt;GS&lt;25</td>
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<tr>
<td></td>
<td>GS&lt;15</td>
</tr>
</tbody>
</table>

Figure E.5 The grain size diversity is shown with different color within the core region of samples Cs1L9

The effect of grain size diversities on the measurement of the average grain size is represented as $f_i$ in modified form of Hall-Petch equation (2.12), and it’s the fraction of each field of measurement. To appreciate the increasing trend of grain size within the field containing higher fraction of ESGs, the field of measurement for the in-plane sections of sample Cs1L9 and Cs4L9 is getting smaller in each step which is shown in Figure E.6.
Figure E.6 Red-colored grains represent each field of measurement, the number of counted grains decreased in each step from (A) to (C)

The variation of average grain size against the decreasing number of counted grains is shown in Figure E7.
Figure E.7 Increasing trend of grain size as the number of grains is decreasing within the field of measurement

The average grain size for the largest field of measurement was about 10.5 μm for both samples, and increased up to nearly 13 μm while the number of grains is decreased within the field for sample Cs1L9. However, the decreasing trend is followed with lower slope for sample Cs4L9 and the deviation from the average grain size of large field was about 2 microns. The comparison between two microscopic pictures indicated that the large ESGs are surrounded by small grains while the large fraction of ESGs can be counted within each field of measurement for sample Cs1L9. The cumulative effect of grains larger than 11 μm is more than 30% for the small field of measurement of sample Cs1L9, while it’s about 2% for smaller field of sample Cs4L9.

To appreciate the effect of accumulated ESGs on the average grain size of each field, the thickness of the samples (Cs1L9 and Cs4L9) is divided into 13x13 squared-grids (Figure E8), the higher numbers of ESGs are located within the centered grids of sample Cs1L9. Although the high area fraction of ESGs is determined for sample Cs4L9, most grids have a few numbers of ESGs. Therefore it can be expected that the average grain size within each field of measurement is closer to the average grain size which is calculated for the core region, while an increase of 13-15 μm in average grain size of the field containing more than 30% ESGs is expected.
Figure E.8. Each squared-grid represents one field of measurement
Moreover the presence of large ESGs within the skin region of sample leads to reduction of yield strength. The local average grain size within the skin region of the sample also increases up to 13-15 \( \mu \text{m} \). Figure E9 and E10 shows the cross section of samples which are extracted from close-to ingate- region (Location6), and dispersed ESGs are located within the skin regions of the samples. As a result, the yield strength of the sample Cs1L6 is decreased up to 93 MPa, and it’s 98MPa for sample Cs4L6. Although the lower fractions of ESGs are located within the core region of samples, the yield strength and \( \sigma \theta \) plots (Figure 4.29 and Figure 4.30) of these samples resembles with the samples containing higher fraction of ESGs within the core region.

![Figure E.9 The disperse ESGs are located within the skin region of sample Cs1L6](image-url)
Figure E.10 The disperse ESGs are located within the skin region of sample Cs4L6
Distribution pattern of ESGs are shown in the following figures.

Figure E.11 The presence of ESGs within the skin region
Figure E.12 Lower fraction of ESGs within last-to-fill region
Figure E.13 Lower fraction of ESGs within Knit line regions
Figure E.14 ESGs within Location 1
Figure E.14 ESGs within various locations
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