

IMPROVEMENT OF IMPACT RESISTANCE OF ALUMINUM AND ZINC
BASED DIE CAST PARTS BY MEANS OF TOOL STEEL INSERTS

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MURAT KAMBEROĞLU

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BASED DIE CAST PARTS BY MEANS OF TOOL STEEL INSERTS**

submitted by **MURAT KAMBEROĞLU** in partial fulfillment of the requirements
for the degree of **Master of Science in Metallurgical and Materials Engineering**
Department, Middle East Technical University by,

Prof. Dr. Canan Özgen
Dean, Graduate School of **Natural and Applied Science**

Prof. Dr. Tayfur Öztürk
Head of Department, **Metallurgical and Materials Eng. Dept.**

Prof. Dr. Ali Kalkanlı
Supervisor, **Metallurgical and Materials Eng. Dept.**

Examining Committee Members:

Prof. Dr. Ekrem Selçuk
Metallurgical and Materials Engineering Dept., METU

Prof. Dr. Ali Kalkanlı
Metallurgical and Materials Engineering Dept., METU

Prof. Dr. Şakir Bor
Metallurgical and Materials Engineering Dept., METU

Assoc. Prof. Dr. Nuri Durlu
Mechanical Engineering Dept. TOBB ETU

Prof. Dr. Rıza Gürbüz
Metallurgical and Materials Engineering Dept. METU

Date: 24.11.2011

I hereby declare that all information in this document has been obtained and presented in accordance with academic rules and ethical conduct. I also declare that, as required by these rules and conduct, I have fully cited and referenced all materials and results that are not original to this work.

Name, Last name : **Murat Kamberoğlu**

Signature :

ABSTRACT

IMPROVEMENT OF IMPACT RESISTANCE OF ALUMINUM AND ZINC BASED DIE CAST PARTS BY MEANS OF TOOL STEEL INSERTS

KAMBEROĞLU, Murat

M.Sc., Department of Metallurgical & Materials Engineering

Supervisor: Prof. Dr. Ali KALKANLI

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High Pressure Die Casting (HPDC) is low-cost technique for the mass production of complex, non-ferrous parts. Despite its benefits such as dimensional accuracy, surface quality and high production rate; some mechanical drawbacks limit use of HPDC in production of critical parts especially under dynamical loads.

This study aims to improve impact resistance and surface hardness of die cast parts by means of tool steel inserts. These inserts act as a barrier between the impactor and die casting alloy, in order to avoid surface deformation and reduce stress localization which leads crack formation. Except the impact surface, whole insert is embedded into the die casting alloy by placing them on specially machined die casting molds prior to the metal injection.

The mentioned method was evaluated by mechanical test and micro-examinations which were applied on AISI D2 tool steel inserted A518.0, A413.2 and Zamak5 alloy

samples. To see the effect of inserts on energy absorbance under single destructive loads, both monolithic (conventional) and inserted (produced by mentioned technique) samples were subjected to Charpy impact test. In order to observe its behavior under non-destructive, cyclic, low velocity impacts; a dedicated real rifle part was produced by this method and tested in the real service loads. Explicit Finite Elemental Analysis was also carried out to understand how the inserts increases the energy absorbance and protect the die cast body by simulating both destructive and non-destructive impact loads. In addition to these, micro-examinations were also conducted especially on insert-die casting alloy interface for chemical and physical interactions, defects and stability.

In regards of experimental findings, mechanical feasibility of the method was achieved. It was proved that steel inserts improve energy absorbance, stress distribution and impact-surface hardness of die cast products.

Key words: Die casting, insert, impact, Explicit FEA, metal-metal interface

ÖZ

ALÜMİNYUM VE ÇİNKO BAZLI BASINÇLI DÖKÜM PARÇALARIN DARBE DAYANIMIN ÇELİK İNSÖRTLER İLE GELİŞTİRİLMESİ

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Basınçlı döküm, karmaşık geometriye sahip demir dışı alaşımların seri üretimi için düşük maliyetli bir yöntem olup; boyutsal mükemmellik, yüzey kalitesi ve yüksek üretim hızı gibi avantajlara sahiptir. Ancak basınçlı döküm ile üretilen parçaların sahip olduğu mekanik eksiklikler; basınçlı döküm teknolojisinin, dinamik yükler altında çalışan parçaların üretiminde tercih edilmesine engel teşkil etmektedir.

Bu çalışma, çelik insörtler yardımıyla basınçlı döküm alaşımlarının darbe dayanımını ve darbe yüzeyi sertliğini arttırmayı amaçlamıştır. İnsörtlerin fonksiyonu; çarpan ile basınçlı döküm alaşımı arasında bariyer görevi görerek darbeye bağlı yüzey deformasyonlarını ortadan kaldırmak, basınçlı döküm alaşımına iletilen gerilimi yayarak çatlak oluşumunu engellemek ve/veya darbe enerjisinin daha fazla emilmesini sağlamaktır. Bu insörtler, dökümden önce kalıpta istenilen bölgelere yerleştirilmiş ve döküm sırasında darbe yüzeyi hariç tüm çevresi basınçlı döküm alaşımı tarafından kaplanmıştır. Böylelikle parça; hızlı ve düşük maliyetle, işleme

gerektirmeden basınçlı döküm ile şekillendirilmiş olup yalnızca darbe aldığı bölgelerde çelik kullanılmıştır.

Bahsedilen bu metodun verimliliği; AISI D2 takım çeliği insörtlere sahip A518.0, A413.2 ve Zamak5 alaşımlı numunelere yapılan bir takım mekanik testler ve mikro incelemeler ile denetlenmiştir. İnsörtlerin; tahribatlı darbelere karşı emilen enerji üzerindeki etkilerini görmek için Charpy darbe testi uygulanmış ve sonuçları monolitik (insörtsüz) numunelerin sonuçlarıyla karşılaştırılmıştır. Bunun yanında, tekrarlı ve tahribatsız darbelerin yer aldığı gerçek mekanizma parçalarındaki dinamik yüklere karşı dayanıklılığını test etmek amacıyla; bir piyade tüfeğinin kritik bir parçası, bu metot ile üretilerek silahta denenmiştir. Tüm bu mekanik testler; paralel olarak, dinamik sonlu elemanlara analizleri ile bilgisayar ortamında simule edilmiş ve insörtlerin mekanik etkileri incelenmiştir. Ayrıca, insört ve basınçlı döküm alaşımının oluşturduğu ara yüzey; metalurjik etkileşimleri, döküm parametrelerin etkilerini ve hataları görebilmek amacıyla mikroskobik incelemeye tabi tutulmuştur.

Çalışma sonucunda; insertlerin, basınçlı döküm ile üretilmiş parçaların darbe tokluğunu ve dayanımını ciddi miktarda arttırdığı görülmüştür. Döküm parametrelerindeki gerekli ayarlamalar ve doğru malzeme seçimi ile insörtlü basınçlı döküm teknolojisinin endüstriyel çapta kullanılabileceği tespit edilmiştir.

Anahtar sözcükler: Basınçlı döküm, insört, darbe, sonlu elemanlar analizi, metal arayüzler

To my parents and Sibel,

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CHAPTER 1

INTRODUCTION

High pressure die casting (HPDC) is low-cost process, developed especially for mass production of relatively small and complex non-ferrous metal parts. Briefly, it consists of pressurized flow of molten die casting alloy through the machined cavity in permanent steel mold. As HPDC is a molten state and high pressure process, many metallurgical factors are present and they cause some specific advantages and limitations which determine the uses of HPDC in industry. Main advantage is elimination of expensive machining processes which reduces production cost and time of parts with complex and low-tolerance geometries. However, mechanical weaknesses of dedicated non-ferrous alloys and porosity induced by gases are the main factors restricting the use of die cast parts under cyclic impulsive loads. These kinds of loads yield surface deformations, or/and cracks that lead loss of functionality or complete fracture. Thus, HPDC are rarely preferred in production of complex parts operating under dynamical loads.

In order to expand the use of HPDC; dynamic mechanical weaknesses should be avoided literally. As impact resistance of die casting alloys increases, HPDC can be assigned for production of more critical parts. Achievement of this goal brings some economic benefits since expensive shaping operations of steel have been replaced with low-cost die casting process.

It can be satisfied by designing a high strength barrier between impactor and die cast body. The barrier layer, which is called as “insert” in this study, has several functions. For parts operating under cyclic, non-destructive loads; inserts avert surface

deformations and reduce the impact stress and energy transmitted to low strength medium (die cast body). For parts subjected to single destructive loads, inserts increase the total absorbed energy by spreading the loads and avoiding localization.

Practical use of this method requires some modifications on conventional HPDC. First of all inserts should stay fixed in the die cast body strictly even after number of impacts. It is maintained by bedding into the die cast body, which means it should be placed in to the special cavities on mold of HPDC machine prior to the casting. After the injection, molten metal should cover the all over the insert except the face which is subjected to impact. Secondly, integrity of the interface between insert and die casting alloy must be maintained by appropriate adjustments in casting parameters. In order to maximize the casting quality; these modifications should be determined carefully. Thus, micro investigations (SEM, EDX and light microscopy) were held to focus on optimization of casting parameters and metallurgical interactions seen in the interface between die casting alloys and insert.

In order to prove the industrial practicability of mentioned method, many examinations and optimizations were required. This study consists of these examinations and optimizations, aimed to bring in a feasible method of production that economic benefits of die casting process can be achieved even in mechanically critical parts. Dynamic mechanical tests were carried out to evaluate alloy groups and reveal the factors influencing the effectiveness of inserts. Dynamical (explicit) Finite Element Analyses were utilized for explanation of effect of inserts observed in the mechanical test.

CHAPTER 2

THEORY AND LITERATURE SURVEY

2.1. Die Casting

Die casting is non-ferrous, liquid state metal shaping process that molten metal is forced hydraulically to fill machined cavities on the die. Rapid solidification with extreme cooling rate is maintained by elevated pressure and high thermal gradient. A plunger, which belongs to a hydraulic system, pushes the liquid metal with controlled velocity that molten metal travels the runners in the die and reaches the cavities of desired parts. Plunger continues to compress, even after the cavity is filled completely, in order to avoid shrinkage porosity. After few seconds, die opens and solidified metal is released with an exactly same shape and dimension of the cavity. Like all forming process, die casting has some benefits and limitations that determine its usage in industry [1].

2.1.1. Advantages of Die Casting Process

- High production rate: Many parts can be formed in one shot that each shot takes only few seconds. Depending on the size, complexity of parts and capacity of the die casting machine; it is able to produce hundreds of parts in an hour that none of other shaping processes can come up.
- Excellent dimensional accuracy: Dimensional tolerance is as low as 0.1 mm which is dependent on alloy and geometry. This fact eliminates requirement of extra machining operations and similar dimensional adjustments [2].

- High surface quality: Surface quality becomes extremely important for parts subjected to fatigue loading. Compared with all other processes, almost perfect smoothness (below 2.0 μm) can be maintained by die casting [3]. Die quality determines the surface smoothness and therefore dies should be periodically renewed.
- Success in low thickness: Thin walls and other detailed geometries can be perfectly casted without any following issues. Minimum thickness is directly related with characteristic viscosity of alloy, casting temperature and injection pressure. For instance, hypereutectic aluminum alloys can fill cavities with 1.0 mm wall thickness [2].
- Strength and hardness of die cast alloys are higher than their non-heat treated wrought equivalents and best in all casting processes. This distinction is based on rapid solidification and high cooling rate which lead relatively high undercooling causing elevated nucleation rate consequently. Because of this extreme undercooling, limited growth and maximum nucleation rate yield much smaller grain size. This type of microstructure limits dislocation movement effectively and cause increase in strength. As an example, comparable mechanical properties of sand cast, permanent mold cast and die cast; 390.0 and 443.0 alloys are given in Table 1. Vacuum or similar porosity reduction systems can also raise mechanical properties significantly [4].

Table 1: Basic mechanical properties of 443.0 and 390.0 Aluminum alloys casted with various processes [4]

Alloy	Process	Yield S. (MPa)	Tensile S.(MPa)	Elongation (%)
443.0	Sand Casting	55	130	8.0
	Permanent Mold Casting	62	159	10.0
	Die Casting	116	228	9.0
390.0	Sand Casting	180	180	1.0
	Permanent Mold Casting	200	200	1.0
	Die Casting	280	240	1.0

2.1.2. Disadvantages of Die Casting Process

Porosity is the major drawback of die casting which can be reduced by extra systems and adjustment of metallurgical parameters. Many reasons may lead porosity in die casting, however, air entrapment as a result of turbulent flow or insufficient vent design and gases coming out of solution during cooling and pressure relief are the main the factors yielding detrimental porosity presence [5]. As it is widely known, porosity causes notch effect in tensile loads that decreases design strength, fracture strain remarkably [6].

As a possible reason of unexpected catastrophic failure, voids (i. e. pores) decrease impact toughness too. Being a crack initiation site, die cast parts are susceptible to sudden brittle fracture under impulsive loads. Moreover, pores act like a free surface which reflects compressive stress waves to tensile. As tensile properties of all die casting alloys are weaker when compared with that under compressive loads, it rises the risk of crack propagation [7].

Porosity, especially when voids are close to surface, results as lack of resistance to wear. Metal surface collapse due these voids under abrasive loads and consequent surface defects may yield loss of functionality or moreover, crack initiation [5].

Decrease in fatigue life is another detrimental effect of porosity which is able to be reason alone to cause an unexpected failure. It is reported that when porosity percentage increase 0.1% to 0.4%, fatigue life drops by 25% [8,9].

Presence of pores hinders heat treatment capability of aluminum die cast products because of the blistering phenomenon at elevated temperatures. Distributed pores tend to agglomerate and expand at solution treatment temperatures. [10]. It is known that increasing individual mean pore size decreases the fracture strength and strain markedly due to the relatively higher stress intensity factor, K , at the front of the largest pore which tends to grow first until it reaches a free surface [11]. It is easily affirmed that same fact averts use of all die casting alloys at high temperatures.

2.1.3. Types of Die Casting

Many types of die casting process were developed;

Depending on the melting point of alloy:

- Hot chamber: Designed for low melting point materials such as zinc, tin lead alloys. Liquid metal is hold in the machine.
- Cold Chamber: Used for high melting point such as aluminum, magnesium, copper and Zn-Al alloys that an extra furnace is needed for melting and holding.

Depending on pressure:

- High Pressure Die Casting: Best for smaller and low thickness parts that injection is applied with pressure up to 70MPa. High production rate and high dimensional accuracy are the major advantage. This study is based on HPDC.
- Low Pressure Die Casting: Applied for circular rotary parts which rotationally symmetry is needed. Thus injection is performed in vertical direction. Low porosity and larger individual size are available.

In addition to these, customer needs, especially mechanical requirements, lead some evolutions. Vacuum systems, oxygen injection before shots, Acurad process are some of these improvements which are developed to reduce gas porosity in order to enhance strength, toughness and bring in heat treatment ability [12]. For much higher flexural strength, ceramic fiber or whicker reinforcement is suitable for die cast product that ceramic preforms are surrounded by squeezed die cast alloy [13,14].

2.1.4. Process Steps of High Pressure Die Casting

High pressure die casting can be considered in 3 steps which form a single cycle of production.

2.1.4.1. Preparation Step

First step includes preparations such as heating and lubrications of dies. Prior die temperature is very important factor affecting both cast quality and service life of dies due to some reasons. For instance in dies which are not preheated, a molten metal flowing through a thin section tends to solidify due the relatively higher thermal gradient than preheated dies. This fact may not be desired in some cases if the mentioned thin section is not final point where the metal is supposed to reach. This clogged flow results as incomplete cast part. Secondly non-heated gates (thin opening between cavity and runner) also decrease the temperature of metal entering the cavity unduly that raises the viscosity of molten metal that insufficient fluidity may cause some detailed sections or thin walls are missing in the die cast product. The effect of insufficient preheating on the dies is much more severe since thermal shock during the cycles is the major parameter controlling the service life of dies and circumstantial surface quality of products [15,16]. Over-heating the dies yields unfavorable results that high cooling rate, which provides the advantages of die casting process is hindered. Lubrication is another operation needed before each injection. Control of the temperature of the die and easy removal of casted parts without distortion or wear marks are maintained by lubricant which have many types depending on the viscosity and their solvents. Lubricant acts as coating between the die and metal and cuts the chemical interactions between them. Viscose ones provide better protection however surface quality is affected negatively. Low viscose lubricants, which are generally called as “oil in water”, provide better surface however these types are relatively expensive [1].

2.1.4.2. Casting Step

Phase I

Second step is the casting stage which consists of 3 phases. First phase is filling the biscuit and runners with molten metal by low speed plunger motion. The air present

in the cavity of runner and biscuit is also pushed by molten metal towards the casting cavity and vents. Higher plunger speed is unnecessary due to relatively high thickness of biscuit and runners. There is always risk of turbulence and atomization of molten metal at elevated flow velocity (Figure 1a) thus; speed of the plunger should be kept as low as a minimum velocity. The upper limit of the plunger velocity is determined by the wave velocity of the molten metal that should not be exceeded. It is expressed by;

$$v_w = \frac{\beta}{\beta - \phi} v_p \quad \text{Eq. 1}$$

β is defined as fill fraction of shot sleeve cross section behind the wave and ϕ is fill fraction of shot sleeve cross section. When plunger velocity is lower than a critical



Figure 1: Entrapped air when plunger velocity is too fast (a), and too slow (b)

value which is called as “critical slow plunger velocity” (V_{ss}), increase in porosity level was reported in previous studies. This fact is related with the entrapped air pocket in front of the plunger (Figure 1b) which is carried into the casting, when plunger velocity changes. Moreover, at the surface of the shot sleeve, where contact between die and molten metal occurs, a thin solidified film forms. This film is called as flakes which mixes into the molten metal stream may cause partial blockage at gates and thin sections that waste the casting. Extremely slow plunger gives flakes enough time to nucleate and grow which increase the mentioned probability. V_{ss} is still matter of discussion and perfect analytical derivation has not been achieved yet, despite numerous models. However, practical experiments, which are based on trial-and-error, gave some common result shown in Figure 2 [1,5,17,18].

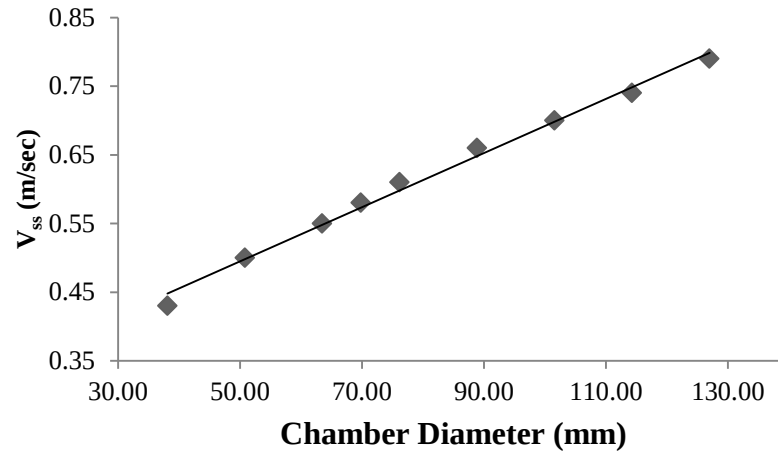


Figure 2: Experimental critical slow shot velocity of Al alloys vs. chamber diameter [1]

Phase II

Second phase is the most critical one for mold designers since it includes most of the parameters controlling the quality of the casting. Plunger speed increases up to 3-6 m/s that molten metal fills the cavity with high flow rate. Requirement of high flow rate is based on the priority that molten metal must fill all sections of the casting cavity and overflows before the solidification. If the rate of the flow is insufficient, molten metal starts to solidify before completing the flow in thin sections. Recommended flow rate at the gates which is defined as gate speed is in the range of 26-40 m/sec for aluminum alloys. Zinc alloys require gate speed about 52-60 m/sec due to relatively low superheat temperature before pouring in order to avoid dissolved gas porosity. But, gate speeds above the recommended range is detrimental for die quality since it causes the long-term serious erosion of the die steel. Excessive gate speed also brings risk of porosity as a result of turbulence within the cavity. Gate speed is calculated by a simple formula;

$$V_G = \frac{V}{tA_G} \quad \text{Eq. 2}$$

Where V_G is gate speed (m/sec), V is the volume of the cavity and overflows (m^3), t is the fill time (sec) and A_G is the gate area (m^2). PQ^2 diagrams are constructed to determine parameter range for the dedicated machine and the job. These diagrams are made of y-axis which is Pressure (P) and x-axis showing square of flow rate (Q). A general look of the diagrams shown in Figure 3 that the machine power line is expressed by the equation 3;

$$P = P_{\max} \left[1 - \left(\frac{Q}{Q_{\max}} \right)^2 \right] \quad \text{Eq. 3}$$

where the $P_{\max}$ is the maximum metal pressure when the plunger reaches the end of stroke while $Q_{\max}$ is the maximum flow rate when plunger elevates to dry shot speed. The line ascending from the origin, which is specific for mold design and alloy, is expressed by;

$$P = \frac{\rho}{2} \left(\frac{1}{C_d} \frac{Q}{A_G} \right)^2 \quad \text{Eq. 4}$$

where ρ is density and C_d is discharge resistance coefficient. The horizontal lines representing the limitations of the recommended gate speed range for alloys and vertical lines are the maximum and minimum cavity fill time which is a function of the flow rate [1, 5].

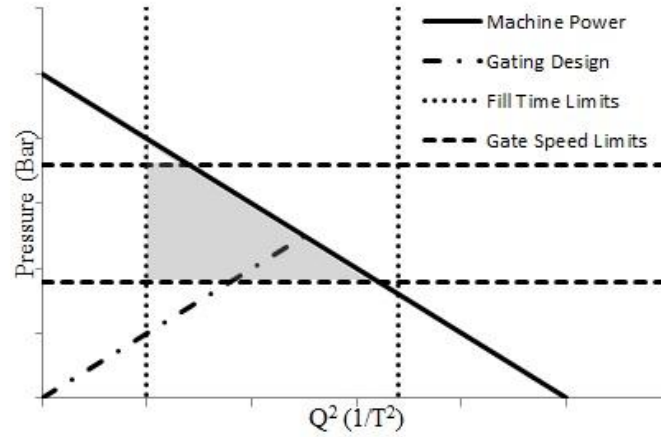


Figure 3: Schematic PQ^2 diagram. Grey region represents recommended working area.

Phase III

Third phase starts after the whole cavity is filled with molten metal and high compression is applied to biscuit to eliminate shrinkage during the solidification. It is a static phase that plunger velocity is almost zero; however the machine power rises to maximum to feed solidifying metal. Magnitude of pressure gets important at the sections where weight/surface area ratio is high due to delayed solidification. Since these sections are the most susceptible to shrinkage voids, they are attended during pressure calculations. Density change during solidification and thermal expansion coefficient of both molten and solid metal are other factors forming shrinkage behavior of the alloy which decide pressure required in third phase. Regarding these factors, pressure applied at the third phase is generally between 200-1200 bars. High pressure is also critical for controlling gas porosity that dissolved gases come out of solution during solidification and tends to agglomerate due to thermal softening at elevated temperatures when compression is not applied [1,5]. Pressure, plunger velocity and position in each phase can be seen in Figure 4 which is schematic output diagram of an individual cycle.

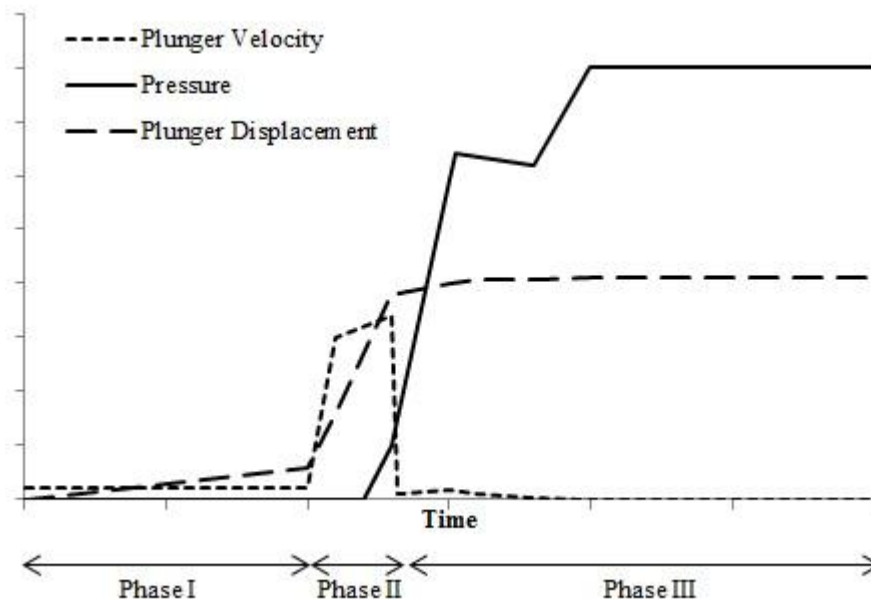


Figure 4: Schematic graph showing variations of pressure, plunger velocity and pressure displacement with respect to time and phases

2.1.4.3. Removal Step

After the solidification and specified time to cool down, dies of the die casting machine opens for removal of the casting. Extra apparatus integrated into the molds steps in to push the casting out of the dies. Detached parts fall in to the basket or belt conveyor or taken by a programmed robot.

Common defects, related to this step, are blisters appear at the surface of the casting and die soldering problem. Main reason of blisters is that entrapped gases near the surfaces, which are compressed with pressure about hundreds of bars during third phase, reach hydrostatic equilibrium. When the compression is released, hydrostatic equilibrium disappears and gases start tends to expand. Since the metal removed from the dies is still hot (120 - 250°C) and behaves plastically even under very low stresses, local visible permanent deformations may formed on surface. This can be eliminated by retarding the removal step and increasing third phase for required temperature drop. [1]

Die soldering is very common case that reduce production rate and cause deformations on castings. Hot die casting alloy sticks to die steel due to absence of protective layer between them that casting cannot be removed easily. This adhesion is bases on chemical interaction between die steel and die casting alloy that aluminum (aluminum in the zinc alloys too, even very low concentrations) tends to diffuse into the steel and form intermetallics. This welding-like behavior results as requirement of extreme effort for removal which wastes time and labor force. Furthermore, die cast part and die surface can be severely damaged during these efforts. Die soldering can be avoided by decreasing super-heat temperature of molten metal and pre-heating temperature of dies. Lubrication should be revised and gate speed can be reduced that molten metal hits die surface more gently. In order to slow down diffusion rate, iron content in die cast alloy can be increased to above 1% by compromising ductility [1,19,20].

2.2. Impulsive Loading on Bilayered Beams

Many studies have been made of dynamic behaviors of layered materials, including laminate composites and multi-layered metal plates. These previous studies generally compared energy absorbance and deflection angles of monolithic (single body) beams and layered structures subjected to hypervelocity particles or low velocity impacts. In this thesis, rate of impulsive loads can be classified as low velocity since the parts casted by the mentioned process are planned to be used in mechanisms belong to defense or automotive systems that continuous and stable operation is required. However, high velocity impact studies were also taken as reference due to some common points such as momentum and energy transfer across the boundaries.

2.2.1. Impact-Induced Stress Waves

All solid materials yield flexural wave propagation when they are subjected to impulsive load. Velocity and amplitude of these waves are dependent on many parameters such as material, impact velocity, geometry of body, etc. Since the energy of impactor is transferred as stress which is distributed by the mentioned waves, failure occurs when the localized stress exceeds a critical level. Main requirement to avoid a failure or to achieve maximum energy absorbance, localization should be avoided by scattering the waves in very short time. [21].

Impact initiates three kinds of elastic waves which are longitudinal waves, shear waves and Rayleigh waves. Longitudinal waves can also be called as p-waves which based on compression and travel with the speed of C_L . Velocity of longitudinal wave is calculated with formula;

$$C_L = \sqrt{\frac{M}{\rho}} \quad \text{Eq. 5}$$

where M is p-wave modulus which is determined by E and ν ;

$$M = \frac{E(1 - \nu)}{(1 + \nu)(1 - 2\nu)} \quad \text{Eq. 6}$$

Longitudinal wave is fastest wave that its vibrations are parallel to the travel direction of wave energy. Since it has half-sphere shape, decay rate is minus square root of wave radius [22,23]. Shear wave (S-wave) is slower wave that particle vibration axis is perpendicular to the travel direction. Its velocity is found by formula;

$$C_S = \sqrt{\frac{G}{\rho}} \quad \text{Eq. 7}$$

where G is shear modulus of material. Shear wave decays with rate about r^{-1} that it makes shear wave travels much longer than longitudinal wave despite relatively lower velocity [22]. Rayleigh wave is also called as surface wave which propagates at shallow depths. It is slowest but most durable wave with decay rate $r^{0.5}$ and velocity, C_R ;

$$C_R = \frac{0.862 + 1.12\nu}{1 + \nu} C_S \quad \text{Eq. 8}$$

If wavelength of the longitudinal wave initiated at the impact point, is smaller than the thickness of the plate, Rayleigh wave mechanism become dominant along whole body. [23].

2.2.2. Behavior of Stress Waves in Layered Structures

Importance of the behavior waves is due to the fact that impulsive stress load is transferred with these waves in solids. The idea is that these waves should be reduced or even avoided to pass in to the susceptible component which, therefore, can operate without failure. Many experimental and numerical studies have been made in last century in order to improve the impact performance of solids that researchers have revealed the load spreading effect of layered structures. Layered components have a specific ability to reduce the stress passing thorough body with some physical contributions of presence of interfaces [24].

When impulsive load is applied on a surface of a layered solid by an impactor, an initial stress waves, which longitudinal one is called “head wave”, is propagated. Amplitude of this incident wave is pressure or stress which can be formulated as;

$$\sigma_0 = \frac{\rho_i c_i \rho_A c_A v_0}{\rho_i c_i + \rho_A c_A} \quad \text{Eq. 9}$$

where, ρ_i, c_i and ρ_A, c_A are the density and longitudinal wave velocity of the impactor and target plate respectively and v_0 is the impact velocity. It should be underlined that the amplitude of the first incident wave does not depend on the mass of the impactor or mechanical and geometrical properties of the target layer. In order to simplify the equations, in the science of dynamic mechanics and acoustics the product of ρ (density) and c (wave velocity) has been called as mechanical impedance which is a specific medium property [25].

When an impact induced longitudinal waves hit the interface between A and B (Figure 5) some amount of energy carried by the wave is reflected back through impact surface while the rest passes to layer B and interface as surface waves. This phenomenon is based on the physical behavior of waves encountering change of impedance. Although initial wave is longitudinal, additional transverse waves appear in each medium as both reflected and refracted [22].

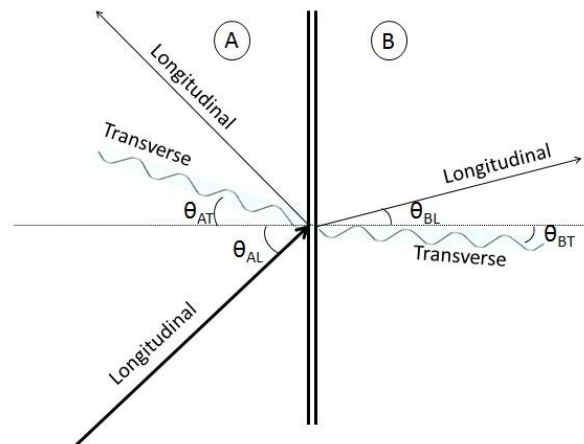


Figure 5: Reflection and refraction of a longitudinal wave at the solid-solid interface.

$$\frac{\sin\theta_{AL}}{c_{AL}} = \frac{\sin\theta_{AT}}{c_{AT}} = \frac{\sin\theta_{BL}}{c_{BL}} = \frac{\sin\theta_{BT}}{c_{BT}} \quad \text{Eq. 10}$$

When the interfacial waves are neglected, it can be assumed that;

$$U_I + U_R = U_{Ref} \quad \text{Eq. 11}$$

Where, U_I , U_R and U_{Ref} are particle velocity in incident, reflected and refracted waves respectively. A relation between stress and particle velocity is derived from equation of conservation of energy, $Fdt=d(mU_p)$, that;

$$\sigma = \rho c U_p \quad \text{Eq. 12}$$

from the equation 10, when incident wave is normal to interface, the amplitude ratio of reflected and transmitted (refracted) longitudinal waves can be derived as;

$$\frac{\sigma_{AL(Reflected)}}{\sigma_{AL(Incident)}} = \frac{\rho_B c_B - \rho_A c_A}{\rho_A c_A + \rho_B c_B} \quad \text{Eq. 13}$$

And

$$\frac{\sigma_{BL(transmitted)}}{\sigma_{AL(Incident)}} = \frac{2\rho_B c_B}{\rho_A c_A + \rho_B c_B} \quad \text{Eq. 14}$$

It should be noted that this formulation is valid for both condition where impedance of A is higher or lower than B. However it is easily noticed that amplitude ratio of reflected wave (Eq. 13) is found lower than zero when impedance of second medium (B) is lower than A. This fact points that reflected wave has opposite sign, which means compressive incident wave turn into tensile wave. At higher impact velocity conditions (supersonic) this shift yield spall type failure due to the remarkable inequality between tensile and compressive strength especially at high strain rates and brittle materials. The condition which, impedance of B is lower than A, has another circumstance obtained from equation 14 that transmitted stress wave

amplitude is lower than the incident wave. In other words, stress is passed to low impedance medium with decreasing amplitude [22,26]. This fact is one of the key points that underlie the feasibility of the study which will be mentioned in following sections.

2.2.3. Load Spreading

Spreading the impulsive load along axis perpendicular to impact direction delay or even avoids localization of stress at the impact point. In a monolithic medium, momentum is transferred by longitudinal wave which travels in a perfect half sphere shape. However, when impulsive load is applied on layered body, horizontal and vertical components of momentum transfer are separated and move independently due to the impedance difference between layers as it is seen in Figure 6. Momentum transfer along horizontal direction (H) stays in first layer and load carried by this momentum transfer does not transmit to second medium. This fact decrease the total momentum transferred to second medium. Lateral distribution of momentum and reduced vertical (V) transfer leads load spreading concept which enhance energy absorption remarkably especially in cases where second medium has lower toughness [27].

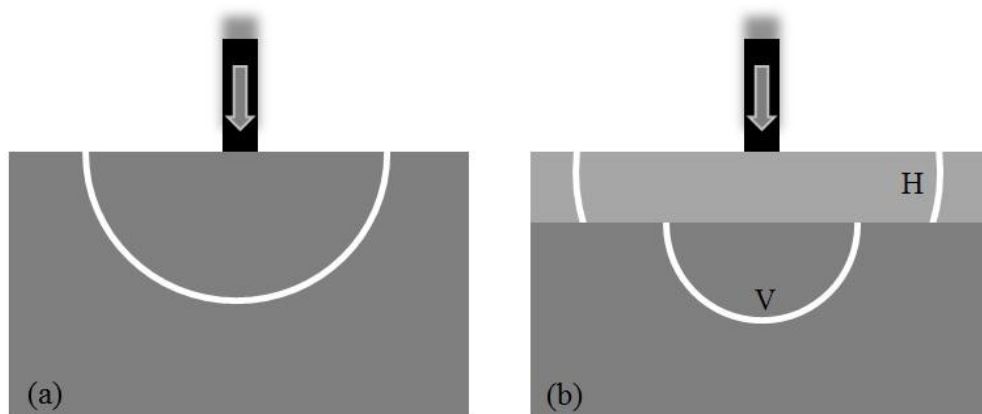


Figure 6: Momentum transfer behavior of monolithic (a) and layered body (b).

The point is to increase momentum transfer the ratio H/V in order to absorb maximum amount of energy. Efficient load spreading is maintained by first layer

which has high wave speed, impedance and strength. However high strength and impedance may also increase the load transmitted to second medium which makes them not consistent parameters. Thus in ref [27] a quantitative measurement of load spreading with the help of a term called “Normalized Dissipative Energy”, (NDE). It is degree of localization of dissipative energy in second layer and achievement of load spreading can be maintained from distribution of NDE.

$$\text{NDE} = \frac{D}{T} = \frac{(\frac{\rho_0}{\rho}) \int \sigma_{ij} d\epsilon_{ij}^p}{E_i + E_k} \quad \text{Eq. 15}$$

$d\epsilon^p$ is plastic increment, E_i and E_k are internal energy and kinetic energy respectively. For effective load spreading, in other words larger H/V ratio, radial NDE distribution should have largest extent and lowest gradient. In addition to these, wave speed of first layer should be as high as possible [27]. In this study, NDE distribution can be measured in finite elemental analyses in order to compare load spreading efficiency of insert-die cast alloy combinations.

2.2.4. Transmission of Spread Waves

The reflected tensile longitudinal wave continues its propagation in opposite direction of the load until it reaches surface where impactor and layer A contact (Figure 7). If the impactor has not been eroded (destroyed) or retreated right after the impact, same transmission and reflection mechanism seen in the A-B interface occurs at the A-impactor interface. The impedance of the impactor becomes important at this point since the transmitted and reflected wave amplitude ratios maintained by equations 13 and 14 in this interface too. The transmitted over incident (previously reflected from A-B interface) stress at the impact surface is important since, usually it is wanted to maximize amount of energy transmitted back to impactor and minimize the stress stayed and reflected through the layered target. The case that contact between layer A and impactor due to fragmentation or etc., impact surface of layer A becomes free surface that complete reflection with

opposite sign (tensile to compressive) is observed according to equations 13 and 14 [25,26].

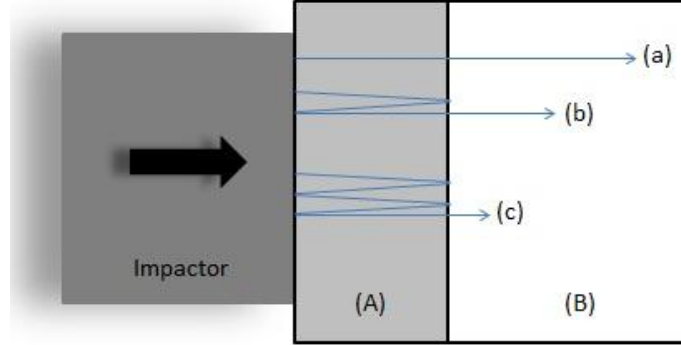


Figure 7: Longitudinal wave behavior after impact.

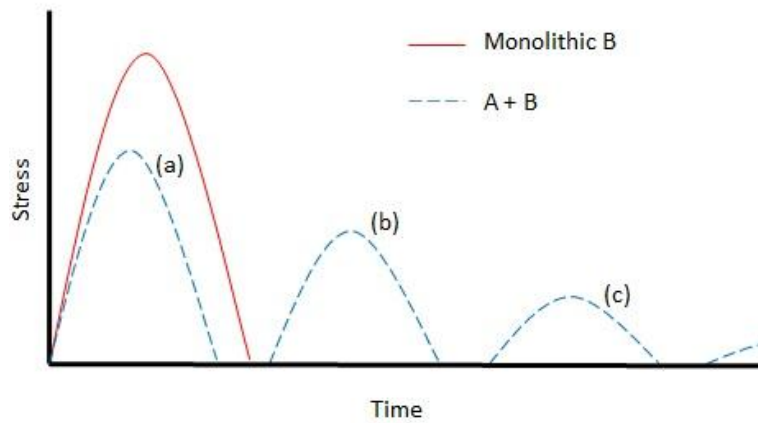


Figure 8: Schematic stress measurement at a point in medium B. Straight red line represent monolithic B body and Blue dashed line belongs to layered body shown in Figure 7. Note that, it is valid when impedances of B is lower than A.

Reflected longitudinal wave from the impact surface continues interface interactions like a bouncing ball between two walls but with continuously decreasing its amplitude as shown in Figure 7 and Figure 8. As a result of this wave motion, contrary to single but high volume stress wave propagation observed in monolithic bodies, several decaying waves with smaller amplitudes transmit into the layer B [25].

Transmitted and reflected Energy, which is carried by the stress waves, was determined according the equation 16. Impedance mismatch is the main factor

determines the energy of reflected and transmitted waves. [22] As total energy was conserved, increased impedance mismatch leads more energy was reflected from interface which is preferred in this study.

$$E_{BL(\text{transmitted})} = 1 - \left[\frac{\rho_B c_B - \rho_A c_A}{\rho_A c_A + \rho_B c_B} \right]^{1/2} \quad \text{Eq. 16}$$

2.2.5. Impact-Induced Plasticity

Response of metallic materials to impulsive load is determined by material and energy of impact. This response may be completely elastic that medium obeys Hooke's law and only mentioned elastic waves propagate. Second possible response is initiation of plastic waves where stress is beyond plastic limit. Finally at much higher amplitudes, behavior of material approaches to compressible fluid instead of elasto-plasticity that hydrodynamic flow dominates effect of deviatoric stress component. This phenomenon is called as "shock" [26].

In material science, plasticity in metals has been described as stress induced irreversible flow of material based on slip mechanism. Due to facts of metallic bonding, metal atoms tend to displace on slip plane and bond with new neighbor atoms without change in crystal structure. From another point of view, displacement of atoms can also be explained by movement of crystal defects called "dislocations". Properties of dislocation movements, such as possible directions (slip systems), unit distance (Burger's vector) and required energy are determined by crystal structure, metallic bonding and lattice parameter of metal. These properties directly influence temperature and deformation rate dependence of plasticity. At relatively higher strain rates, some materials (especially BCC because of high energy and lack of closed-packed planes and some of HCP due to low slip systems) perform remarkable increase in yield strength and drop in ductility. This elevation is related with the restriction or limitation of slip mechanism due to the time dependence of dislocation movement. In addition to this, at high strain rates (and especially for alloys having

high stacking fault energy), twinning may occur instead of shear that plastic flow is retarded. Because of reduced plasticity, plastic work is limited that rest of impact energy stayed non-absorbed which leads brittle failure [28,29,30].

For parts working under cyclic, non-destructive loads, defects such as inelastic deformation are not desired. Localization of stress mentioned previously, is the main factor plastic yielding at impact surface. Another important point affecting local plastic deformation especially at higher impact rates is the thermal localization of heat. Contrary to hardening phenomena, drop in yield strength may be observed due to the fact that impact energy is converted to heat. Instead of an isothermal deformation, local thermal softening leads easier plastic deformation which increases the temperature further. This cascade cycle cause macro scale ductile failure at impact point [31]. Since there is not any accurate mathematical relation between material properties and dynamic behaviors, constitutive models (i.g. Johnson-Cook, Cowper-Symonds), which are based on constants obtained from experimental data, are used for prediction of dynamic yielding.

When the parts are used for purposes that they are sacrificial for mechanical protection of another subject (e.g. security locks, impact absorbers, etc.), the energy absorbed prior to the failure become important. As these parts are expected to operate one-time, maximum plastic deformation was desired in order to convert the impact energy into the plastic work. Thus, for sacrificial parts, which operate under destructive loads, should perform maximum plastic deformation before the failure. Since the local maximum inelastic deformation was limited for die cast parts, this can be achieved by reducing the localization of the stress and distribute the inelastic deformation as wide as possible.

2.3. Interactions Between Steel and non-Ferrous Alloys

Contact of two chemically different metals excited many metallurgist's and researcher's attention due to some special interactions. In this study, interactions between steel and non-ferrous (aluminum and zinc based) die cast alloys are taken into consideration. At high temperature, diffusion of atoms of die cast alloy into the steel or the converse of it, lead a local mixture of atoms regardless of low solubility at room temperature. Phase diagrams come into prominence at this point that mixture may form a meta-stable solid solution or lead super saturation that yields an extra local compound layer. As chemical behavior these alloys influences the mechanical stability of the interface between the insert and die cast alloy, which is not only critical for the load spreading and transmission, but also vital to many points such as interface integrity and toughness of the whole component.

2.3.1. Steel – Al alloy interface

Steels and aluminum alloys have been most widely used structural materials in last century. Both alloys have their own beneficial characteristics for appropriate usages. However when these two distinct alloy groups are planned to be used together in order to combine their advantages, some limitations appear in both production and service. One of the most important limitations is the temperature due to the requirement that type and thickness of intermetallic layer should be under control.

Intermetallics are either stoichiometric or non-stoichiometric phases formed by two or more metals that they have their own crystal system or lattice parameter and shown like a chemical compound (ex: FeAl_3). Compositional homogeneity is needed for presence in large scale but and when metals are divided with a border such as an interface, diffusion into each other medium is required for growth. As an evident fact, diffusion rate has exponential relation with temperature that small increment in the temperature yields remarkable increase in diffusion coefficient. This effect is important since shaping and forming of metals are performed at relatively high temperatures or even in liquid state. It is reported that, formation of almost all

intermetallic layers between steel and aluminum alloy substrates are diffusion controlled. [32,33]. Thus temperature and time are the major parameters which determine if and which intermetallic(s) are to form. However there is not a rule that every presence of diffusion yields formation of intermetallic compounds. Nucleation of intermetallic layer requires super saturated solution that thermodynamic instability becomes the driving force. Therefore there is an incubation time that diffusing atoms reach super saturated solution by diffusion. Formation of these intermetallic layers generally needs immersion of steel into the liquid aluminum alloy or even extra annealing process for diffusion. As another requirement for nucleation, kinetic activation energies of each intermetallics should also be satisfied. There more factor affecting nucleation, beside temperature and thermodynamic stability that some inhibitors and activators which is to be considered in this section. [33,34,35]

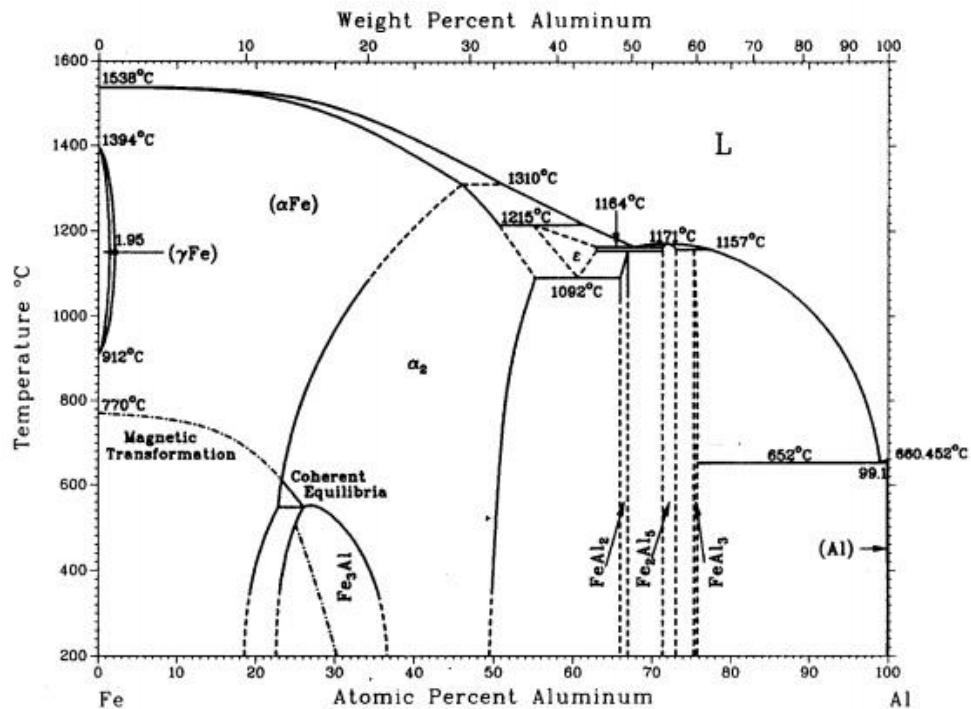


Figure 9: Phase Diagram of iron and aluminum. [36]

As it is indicated in the phase diagram (Figure 9), five intermetallic compounds are possible which are Fe₃Al, FeAl, FeAl₂, Fe₂Al₅, FeAl₃. These intermetallics exist when almost pure aluminum alloys are used. When the immersion is applied at temperature about the melting point of aluminum alloy, Fe₂Al₅ layer exist at the

interface. As the temperature increases, iron rich alloy, Fe_3Al become preferential [32]. FeAl and FeAl_2 has not been observed after immersion without and subsequent heat treatment [38].

Presence of silicon has massive effect on both diffusion rate of atoms and intermetallics compositions [32-40]. Silicon atoms inhibit the diffusion at the interface and reduce the thickness of intermetallic layers [32,34,39]. Especially in Al-Si alloys, Si atoms take the lattice positions of Al atoms in Fe_3Al (FCC). This is proved by the decrease of lattice parameter of Fe_3Al by increasing Si content. Because of this reason, it is generally preferred to show as $\text{Fe}_3(\text{Al}_x\text{Si}_{1-x})$ [35]. Fe_2Al_5 layer, which is most commonly and rapidly formed intermetallic at interface between pure Al alloy and iron, has also affected by presence of silicon especially at high concentrations (>6%). Fe_2Al_5 compound transform to ternary $\eta\text{-Al}_5\text{Fe}_2(\text{Si})$ and/or $\beta\text{-Al}_5\text{FeSi}$ with respect to silicon content. $\beta\text{-Al}_5\text{FeSi}$ layer, which is seen at the interfaces between steel and Al-Si alloys, nucleates about 555°C and grows till temperature drops below 510°C [33,41,42,43]

Other alloying elements are also effective on nature and thickness on intermetallic layers. Chromium content in steel inhibits diffusion of aluminum atoms by decreasing its activity coefficient in iron while other common alloying elements such as Mn, Ni, Co have increasing effect [40]. Strontium, which is used as modifier in Al-Si cast alloys, has inhibitor effect on intermetallic formation since it decreases the chemical potential of iron atoms in the diffusion zone. Titanium, which is commonly used grain refiner in Al-Si alloys, limits the diffusion of aluminum atoms and reduces the reaction rate but it is not effective as much as silicon or strontium. Presence of pre-dissolved iron in the aluminum alloys has also decrease the rate of intermetallic formation by decreasing chemical potential gradient at the interface which is driving force of iron diffusion from steel. As an interesting note, because of being a diffusion retardant; strontium, titanium and iron are used as a solution for common die soldering problem in die casting instead of the major reasons of use such as modifier, refiner or hardener [33,43].

Effect(s) of intermetallics formed at the Fe-Al interface is another point which should be considered. Results of related studies show that minimizing or even eliminating the intermetallic layer, improve the mechanical properties under both single and cyclic loads. Reasons of this fact are:

- Intermetallics are extremely brittle that cracks can easily propagate through the steel and/or aluminum alloy easily which result as catastrophic failure [44,45].
- Local density change during growth and mismatches of thermal properties of intermetallic layer and substrates (Steel and Al alloy) such as thermal expansion coefficient, conductivity and specific heat may lead thermo-mechanical residual stresses along the interface that component become susceptible for crack initiation even under weak impact loads [34,45].
- Difference in diffusion rate between iron and aluminum may lead porosities (similar to Kirkendall effect) since the position of the interface stays unchanged [46-48]
- Finger-like morphology of intermetallic layer (especially in Fe_2Al_5 derivatives) is detrimental since fingers grow along at the grain boundaries of steel and aluminum that increases the potential of intergranular crack propagation [37,48].

Despite the possible effects mentioned above, intermetallic phase might be supposed to improve the wetting behavior of the aluminum on steel insert and bind them which may avoid delamination and gaps along the interface and enhance mechanical properties.

2.3.2. Steel – Zn Alloy Interface

Similar to the case in Fe-Al interface, iron has a tendency to form intermetallics with zinc according to binary phase diagram shown is Figure 10. Four main intermetallic phases, which are Γ ($\text{Fe}_3\text{Zn}_{10}$), Γ_1 ($\text{Fe}_{11}\text{Zn}_{40}$), δ (FeZn_{10}) and ζ (FeZn_{13}), form at the interface between pure zinc and steel [49]. The common property of Fe-Zn

intermetallics are the extreme brittleness and mechanical instability even more deleterious than Fe_2Al_5 . Therefore, aluminum is added (0.2-0.4%) into the Zn bath during galvanizing process in order to form a thin Fe_2Al_5 inhibition layer at the interface [50,51]. If further diffusion is allowed, zinc atoms tend to diffuse into the Fe_2Al_5 layer and form $\text{Fe}_2\text{Al}_5\text{Zn}_x$ [52]. Presence of Zn (even lower than 2%) raises residual stress along the layer which causes perpendicular cracks at the layer. [53,54]

Recent Zn-based die cast alloys contain aluminum above 3.5%. This means Fe_2Al_5 layer formed preferentially before Fe-Zn intermetallics [51,55,56]. Furthermore required diffusion for nucleation and growth is much higher than the allowed amount during solidification in die casting process. In other words, Solidification in HPDC is extremely fast that diffusion is not enough for super saturation-induced intermetallic nucleation and growth [57]. However a very thin Fe_2Al_5 intermetallic might have been expected due to liquid alloy-steel contact in phase I.

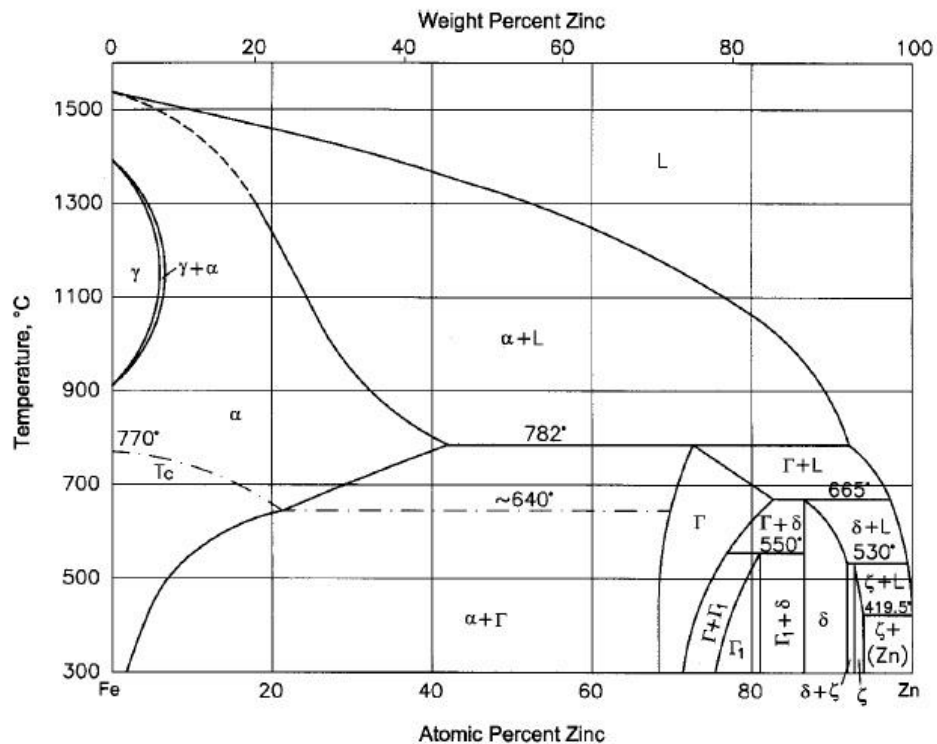


Figure 10: Phase Diagram of iron and zinc. [49]

CHAPTER 3

EXPERIMENTAL PROCEDURE

3.1. Die Casting

As a cost efficient and fast process, Die casting is the most feasible method to produce complex shaped mechanism components. This study has focused on improvement of dynamic properties of die cast parts with the help of steel inserts at critical impact points. For an accurate evaluation of this improvement; die casting alloys, inserts and process parameters were planned carefully.

3.1.1. Alloys

Three different die casting alloys were used in experiments in order to see the influence types alloys and effects of their physical properties such as impedance, density etc. Two of these alloys are aluminum based alloys, A518.0 (Al-Mg) and A413.2 (Al-Si) and the other one is zinc based Zamak5. These alloys were chosen as they are commonly used die casting materials. In addition to this, comparison between these two aluminum alloys has showed the effects of silicon and magnesium as main alloying elements. Zamak5 was chosen in order to compare the chemical and mechanical behavior of zinc and aluminum based alloys. Their chemical compositions, physical and mechanical properties are shown in Table 2, Table 3 and Table 4 respectively.

Table 2: Chemical composition of die casting alloys. (Spectral analysis in weight %)

Alloy	Al	Zn	Si	Mg	Fe	Cu	Ti
A518.0	Bal.	0.08	0.21	8.19	0.70	0,15	0,03
A413.2	Bal.	0.11	12.9	0.12	0.66	0.09	0.22
Zamak5	4.20	Bal.	0.13	0.07	0.07	1.16	-

Table 3: Physical properties of die casting alloys [58-60]

	A518.0	A413.2	Zamak5
Density (g/cm ³)	2.57	2.64	6.70
Thermal Conductivity (W/m°C)	96.2	121.0	109.0
CTE (10 ⁻⁶ mm/mm°C)			
20-100 °C	24.1	20.4	27.0
20 - 300°C	26.5	22.4	27.4

Table 4: Mechanical properties of die casting alloys

	A518.0	A413.2	Zamak5
Yield Strength (MPa)	170	246	-
Tensile Strength (MPa)	322	271	328
Compressive Strength (MPa)*	-	380	600
Shear Strength (MPa)*	205	170	260
Hardness (HV)	90	98	112
Elongation (%)	7.0	4.5	8.6
Unnotched Impact Strength (J)	16	12	24
Young's Modulus (GPa)	70.2	72.9	130
Poisson's ratio *	0.33	0.33	0.28

* taken from reference [58]

Both aluminum alloys were melted in injection furnace and heated up to $750 \pm 5^\circ\text{C}$. 50-60°C loss was estimated during transportation to the die casting machine which means that casting temperature was about $690 \pm 10^\circ\text{C}$. Zamak5 was melted and heated in same furnace up to $480^\circ \pm 5^\circ\text{C}$ and casting temperature was 420 ± 10 . While molten metal was waiting in furnace, argon degassing operation has been held for

10-15 minutes. Temperature drop during the argon passage was taken into account and alloy was heated up to required temperature.

3.1.2. Inserts

Steel inserts were made of AISI D2 (DIN 1.2379) cold work tool steel whose composition is shown in Table 5. AISI D2 is wear-resistant steel with moderate impact toughness. D2 was chosen as insert alloy since:

- Moderate toughness allowed us to see influence of insert design, interface and other casting-related factors.
- High hardness (59 ± 1 HRC), which enhance wear resistance, can be easily gained by heat treatment without any extra surface treatments.
- Availability of D2 is better than other premium steel alloys as it is widely used.

Heat treatment processes are also given in Table 6 and mechanical properties at final condition are featured on Table 7. Heat treatment steps were determined for the highest toughness of D2. Final thickness of steel inserts was 1.5 ± 0.05 mm. Geometries of inserts used in impact test sample and hammer are shown in Figure 11 and Figure 12 respectively.

Table 5: Average Spectrometer Results of AISI D2 tool steel

C	Cr	Si	V	Mo	Mn	P	S
1.52	11.40	0.29	0.89	0,78	0,46	0,02	0,03

Table 6: Heat Treatment of AISI D2 tool steel.

Steps	Notes
Pre-Heating	Heat up to 600°C with rate 200°C/h, hold 20 min
Pre Heating	Heat up to 800°C with rate 200°C/h, hold 15 min
Austenization	Heat up to 1030°C, hold 33 min
Quenching	Forced Air quenching with 1.0±0.2 Bar compressed air
Tempering	Heat up to 265°C in 1 hour, hold 110 min, cool in air
Tempering	Heat up to 275°C in 1 hour, hold 110 min, cool in air
Tempering	Heat up to 285°C in 1 hour, hold 110 min, cool in air
Tempering	Heat up to 290°C in 1 hour, hold 110 min, cool in air

Steel inserts was kept away from air contact at elevated temperature due to corrosion tendency and decarburization risk. Controlled treatment atmosphere and airtight enclosure were applied.

Table 7: Mechanical and Physical Properties of AISI D2 treated according to table 6.

Mechanical Properties		Physical Properties	
Hardness (HRC)	59	Density (g/cm ³)	7.70
Compressive Strength (MPa)	2170 - 2450	Thermal Conductivity (W/m°C)	20.0
Young's Modulus (GPa)	215.0	CTE (10 ⁻⁶ mm/mm°C)	
Poisson's Ratio	0.29	20-200°C	11.2
Max. Elongation (%)	-	20-500°C	12.0
Unnotched Impact Strength (j)	50-53	20-800°C	12.4

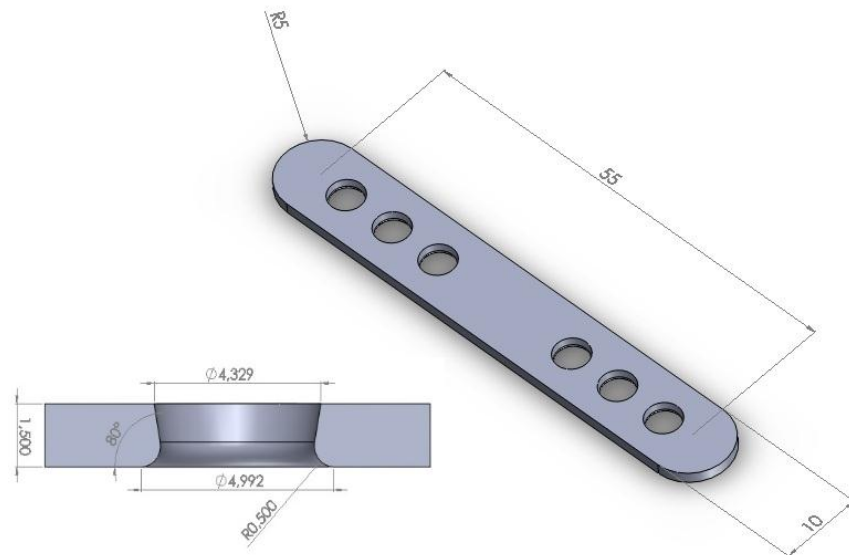


Figure 11: Insert of impact test sample

Surfaces of all treated steel inserts, except an experimental group which is to be mentioned soon, were polished with 1 μ m diamond paste in order to reduce any oxide layer, surface flaws or roughness.

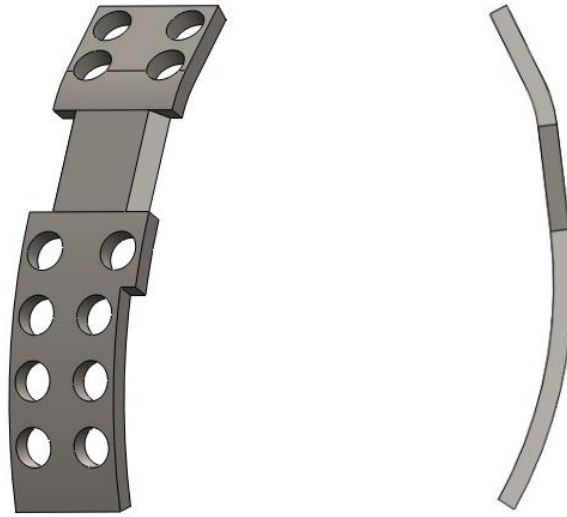


Figure 12: Insert of hammer of H&K 33 rifle. Dimensions are confidential

3.1.3. Casting

Casting operation was performed via HIDROTEKNIK MEP200 HPDC machine (see Figure 13). Machine settings are tuned regarding corresponding PQ^2 diagram (Figure 14) and alloys. Casting data such as pressure, plunger velocity, plunger displacement and filling time were graphed via EMTEC DC200 device after each injection. All result graphs showed exact similarity which points the consistency of the casting control. One of these graphs is shown in Figure 15. Using same casting data, a casting simulation was held with NOVACAST software in order to evaluate parameters and mold design.



Figure 13: HPDC machine used in experiments.

Molds, which were machined and treated by RUTAŞ® Die Casting Co. Ltd., had been designed in the way that inserts could be placed easily and hang on during high pressure metal injection without any deformation or displacement (see Figure 16). This requirement was satisfied by extra deeper cavities for inserts that they could be seated and fixed tightly. However over tightening would have extended the time of removal step and caused material loss which would be waste of money, time and labor. Thus, inserts cavities were almost near net machined by EDM (Electrical Discharge Machining)

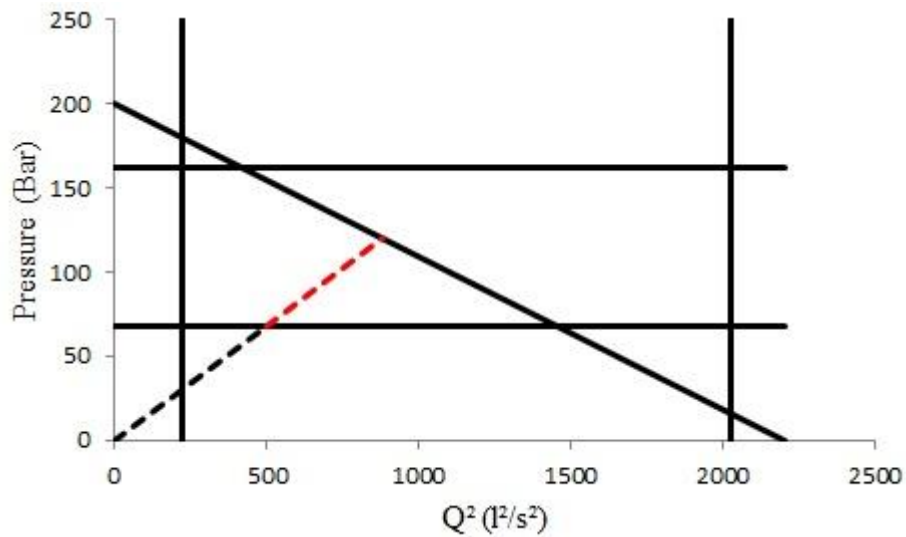


Figure 14: PQ^2 diagram drawn for the dedicated mold and HPDC machine. Red dotted line represents optimal operating parameter set for phase-II.

The casting process has 7 steps:

- 1) Lubrication of pre-heated mold (5-10sec)
- 2) Seating inserts tightly into their cavities (5-10 sec)
- 3) Locking the clamps (4 sec)
- 4) Pouring the super-heated molten metal into the chamber (2-5 sec)
- 5) Metal Injection (6-10sec)
- 6) Relief of clamps and removal of the part (6 sec)
- 7) Grinding and trimming (15-60 sec)

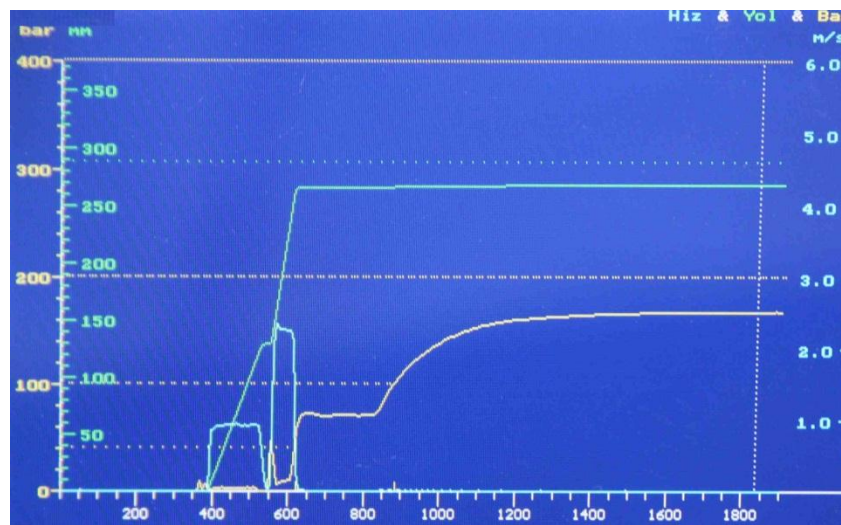


Figure 15: Graph of one of experiments captured from casting monitor where pressure, plunger velocity and plunger positions were shown. X-axis is time in milliseconds. For detailed information see Figure 4.

During the casting of samples, same two lubricants, which are water based (for thin sections) and graphite based (for runners and biscuit), were used together. The pre-heating temperature of the mold was kept at same temperature as much as possible which is about $120 \pm 10^\circ\text{C}$.



Figure 16: AISI 2344 Die casting molds used in experiments. Nitriding operation was held on whole casting surface to obtain hardened protection layer.

3.1.4. Casted Samples

After the removal step, casted samples were left to cool to room temperature in air and carried to grinding operation to reduce burrs on surface. After a macro-examination both impact test samples (Figure 17a) and hummers (Figure 17b) are grouped and stored.

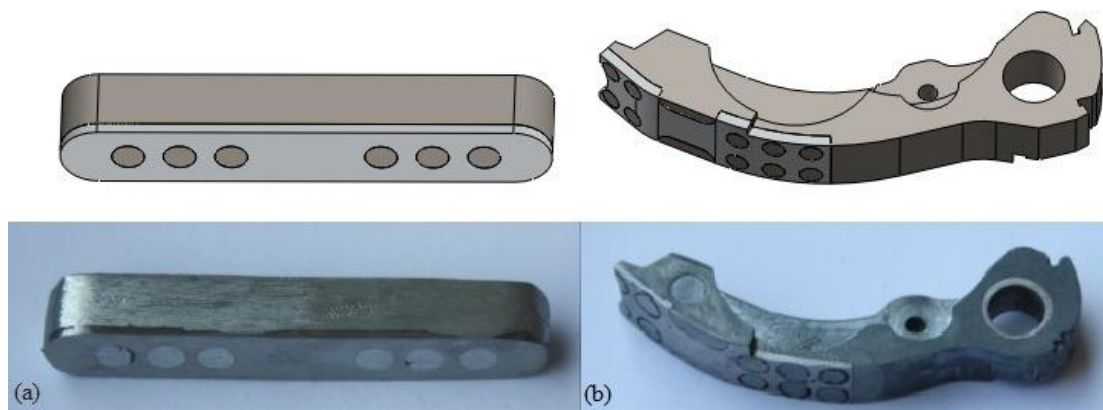


Figure 17: Casted samples; impact test sample (a), hummer of H&K 33E (b)

3.2. Weapon Test

In order to evaluate the functionality and mechanical durability of the steel inserted die cast products under non-destructive dynamic service loads, both inserted and non-inserted die cast hummers were tested in rifle.

After grinding and dimension control steps, as-casted hummers were brought to small arms weapon factory of Mechanical & Chemical Industrial Corporation (Makina ve Kimya Endüstrisi Kurumu, MKEK) which was established in Kırıkkale, Turkey. The hummers were assembled into a perfect H&K 33E rifle, under the supervision of quality control experts of the factory. Perfectly assembled weapons were fired about 180 times in full auto mode (750 rounds / minute). Calculations gave that former steel hummers had hit with velocity of about 3.1m/s and impact energy was estimated as 1.92 joules. Because of the density difference, impact velocity of die cast hummers were varied, as given in Table 8.

Table 8: Mass and respective impact velocity of hummers made of former steel and inserted die casting alloys.

Alloy	Mass (g)	Impact Velocity (m/s)
Former Machined Steel	36.7	3.87
D2 insert – 413.2	14.4	5.14
D2 insert – Zamak5	32.2	4,21

Firing quality and rapidity was checked in shooting range by experienced technicians and operated hummers were disassembled for macro and micro inspections for defects such as plastic deformations, macro and/or micro crack propagations etc.

3.3. Charpy Impact Test

Charpy impact test was carried out to see the effect of inserts under impulsive destructive loads. In addition to comparison of monolithic and inserted samples,

extra experimental grouping was done according to interface- related factors which were predicted to have effects on energy absorbance and impact behavior of components (Table 9). In order to maintain statistical accuracy every group had three perfect (perfect dimension and shape, no surface defects, no visible delamination) samples. Tests were held on same day (24°C) and by same equipment (Figure 19) which was Tinius Olsen 358 joule testing machine. ASTM E-23 [61] was taken as reference in design of samples. Curvature with 5mm radius was present at the far edges of sample in order to make removal step easier. Since machining operations had risk to effect test results, this curvature was not able to be cut, it was not needed anyhow as all samples had same dimensions and shape. Insert design was developed for this study regarding maximum efficiency.

Table 9: Charpy impact test sample groups.

Group	Insert	Code	Alloys		
			A518.0	A413.2	Zamak5
Normal Group	Yes	CG-I	3	3	3
	No	CG	3	3	3
Sand Casted	Yes	SC-I		3	
	No	SC		3	
Surface-oxidized inserts	Yes	OIG	3		
Immersed inserts					
	5 sec.	Yes	LIG5	3	
	10 sec.	Yes	LIG10	3	

Normal Group: Normal group was used to see the effect of inserts and influence of alloys. It was also taken as reference in observing the effects of factors below.

Sand Casted: Casting into the sand mold (silica sand with clay bonder) has very low cooling rate whose effect on interface chemistry (intermetallic) was considered. The efficiency of insert was compared between sand cast and die cast samples.

Surface-oxidized inserts: Oxide layer on surface, which was maintained by flame heating, might influence diffusion and wetting behavior of the interface whose mechanical effects were examined.

Immersed inserts: Inserts of these groups have been immersed into the molten die cast alloy for 5 and 10 seconds, shortly before casting in order to force to nucleate intermetallic layer at the insert surface before the following die casting operation. This was important to detect the mechanical effects of intermetallic layer. Instead of immersion, heat treatment of inserted casted parts might form intermetallic layer too, however, at high temperatures agglomeration and expansion of pores (see section 0) in the die cast body may also decrease the toughness dramatically that effect of intermetallic layer could not have determined clearly as single parameter.

In addition to these groups, inserts were also subjected to impact test alone. Instead of direct comparison, result was used in discussion of mechanical contribution of the inserts.

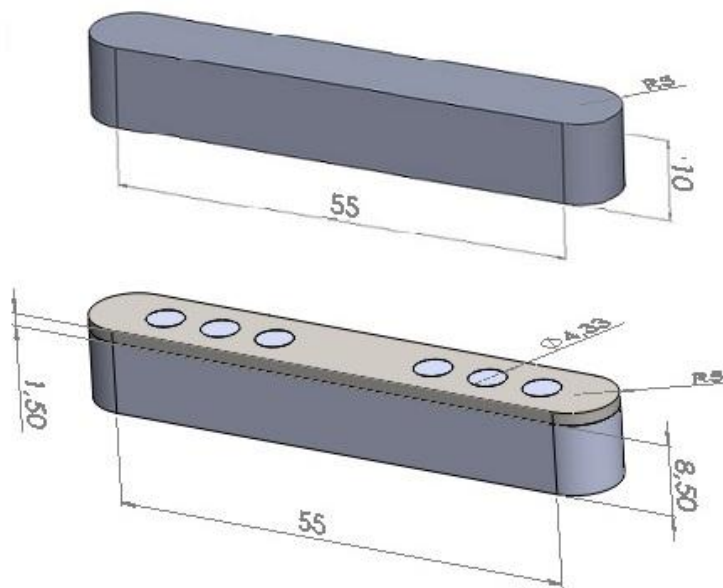


Figure 18: Charpy Impact Test Samples



Figure 19: Charpy impact test equipment used in experiments

3.4. Three Point Flexural Test

Observation of effect of inserts under static loads may help to reveal dynamic influence of them. Thus, inserted and monolithic A413.2 die cast samples were subjected to three point bending test. Test was held via Shimadzu AG-X universal testing machine whose span length was fixed to 40 mm (same as charpy test). Test samples were exactly same as used in Charpy impact test. Load was applied with stroke velocity of 1 mm/min and test ended when aluminum failed.

3.5. Interfacial Examinations and Characterization

Field-emission Scanning Electron Microscopy (FEI Nova NanoSEM 430) was main tool for micro-examination of the interface. Surface topology and defects were visualized by secondary electron mode and elemental contrast was maintained by backscattered electron mode. Energy Dispersive X-ray Spectrometry (EDS) was employed for quantitative elemental analysis especially to observe diffusion near the interface (line scan) and characterization of phases.

Large defects such as porosity, delamination, deformations, cracks and oxides were also inspected by metallographic light microscopy. After the weapon test, impact points of operated hummers were also examined by light microscope for impact-induced defects.

3.6. Finite Element Analyses

In this study, effects of steel inserts were investigated by comparative dynamic finite element analyses of both parts produced by conventional (monolithic, i.e. single die cast body) and developed (inserted, i.e. steel inserts at critical points) die casting processes. This comparison showed the mechanical performance of the developed process, which was critical for feasibility of it.

Some physical measurements and analyses are not able to be made in real world, or some conditions may need to be evaluated without testing in physical experiments due to cost and time limitations. In these cases, a mathematical model, which is simulated with the help of numerical solutions, should be developed as close as possible to the real physical problems. As a methodology Finite element method (FEM) is the procedure which solid components and loads are divided into smaller simple elements and nodes and simulated with some initial and boundary conditions. In addition to basic static analyses, dynamic engineering problems may also be simulated via dynamic explicit solvers (E-FEM) which are comprehensive FEM analyses including energy and momentum calculations.

Real mechanism motions and loads of both Charpy impact test and weapon mechanism test were simulated in analyses. Charpy impact tests of both inserted and monolithic (non-inserted) die cast alloys were modeled in analyses that influence of the presence of insert on energy absorbance and stress behavior was examined. Dynamical stress propagation behaviors of inserted and monolithic hummers were also compared under non-destructive, low energy impacts.

ANSYS 12.1/Workbench interface was used for Pre-processing of simulations where exact CAD files of parts, material models, meshing parameters, initial conditions, boundary conditions and other analysis settings were entered as analysis input. ANSYS –AUTODYN was preferred as solver due to satisfactory performance and accuracy of it. Post-processing (assessment of result file) was performed on same ANSYS 12.1/Workbench interface.

3.6.1. Material Models

All necessary properties of D2 tool steel, Al 518.0, Al 413.2 and Zamak5 were added into engineering data of ANSYS 12.1. Mechanical behaviors such as elasticity and plasticity were defined regarding strain rate dependency and dynamic conditions. Plastic behavior of D2 tool steel and die casting alloys were defined as Johnson-Cook constitutive model whose formula which is shown in equation 17,

$$\sigma_Y = (A + B\bar{\epsilon}_p^n)(1 + C\ln\dot{\bar{\epsilon}}_p^*)(1 - T^{*m})$$

Eq. 17

$$\dot{\bar{\epsilon}}_p^* = \frac{\dot{\bar{\epsilon}}_p}{\dot{\bar{\epsilon}}_0} \quad T^* = \frac{T - T_A}{T_M - T_A}$$

where σ_Y is yield strength and $\bar{\epsilon}_p$ is equivalent plastic strain. $\dot{\bar{\epsilon}}_p^*$ is normalized equivalent plastic strain rate while $\dot{\bar{\epsilon}}_0$ is reference which is generally $1s^{-1}$. T^* is homologous temperature where T_M and T_A are melting temperature and ambient temperature respectively [30]. A, B, C, n and m are specific material parameters which had been estimated experimentally [58,62] and given in Table 10.

Table 10: Johnson-Cook constitutive law coefficients of materials [58,62]

	A (MPa)	B (MPa)	C	n	m	$T_M(^{\circ}C)$
AISI D2	1776	904	0.012	0.312	3.38	1460
A518.0	193	307	0.005	0.180	0.76	535
A413.2	246	275	0.006	0.217	1.12	644
Zamak5	290	462	0.010	0.451	4.22	393

Failure criteria were determined regarding dynamic behaviors of alloys. Johnson Cook failure model was defined for the materials that dynamic fracture strain which was influenced by local stress, strain rate and local temperature as it is shown in equation 18.

$$\bar{\epsilon}^f = [D_1 + D_2 \exp(D_3 \tilde{\sigma})] [1 + D_4 \ln|\dot{\epsilon}^*|] [1 + D_5 T^*] \quad \text{Eq. 18}$$

where ϵ^f is dynamic fracture strain and $\tilde{\sigma}$ is stress triaxiality. Material constants, D_1 , D_2 , D_3 , D_4 and D_5 are specific for each material and its background. Constants of hardened D2 (59 HRC) and die casting alloys are given in Table 11.

Table 11: Johnson Cook Failure model material constants of AISI D2 tool steel.

	D_1	D_2	D_3	D_4	D_5	$T_M(^{\circ}\text{C})$
AISI D2	-0.6	1.7	-0.3	0.005	0.09	1460
A518.0	0	0.612	-2.980	0.005	N/A	535
A413.2	0.02	0.446	-2.241	0.004	0.16	644
Zamak5	0.01	0.655	-1.790	0.006	N/A	393

Failure of the insert occurs when;

$$D = \sum \frac{\Delta \epsilon_p}{\bar{\epsilon}^f} > 1 \quad \text{Eq. 19}$$

Equivalent plastic strain, $\bar{\epsilon}_p$, is found by integration of equivalent plastic strain rate as it is seen in equation 20.

$$\bar{\epsilon}_p = \int \dot{\bar{\epsilon}}_p dt = \int \sqrt{\frac{2}{3} \dot{\epsilon}_{ij}^{\text{dev}} \dot{\epsilon}_{ij}^{\text{dev}}} dt \quad \text{Eq. 20}$$

3.6.2. Analysis Settings

In Charpy impact test simulation, pendulum speed was defined as initial condition which is 5.124 m/s. Density of pendulum was adjusted to maintain 27.269 kg mass (real weigh of the pendulum) which gave 358.0 joule impact energy. Geometry and position of anvils and pendulum have been designed (Figure 20) according to both real test equipment and ASTM E-23 [61] standard. Position of anvil was fixed in all directions as boundary condition and displacement of pendulum was allowed in only impact direction. Pendulum and anvils were completely rigid that they were subjected to neither stress nor deformation. Meshing was controlled by max. element sizing about 1.0×10^{-4} m with tetrahedron method which is very efficient and accurate in dynamic analyses.

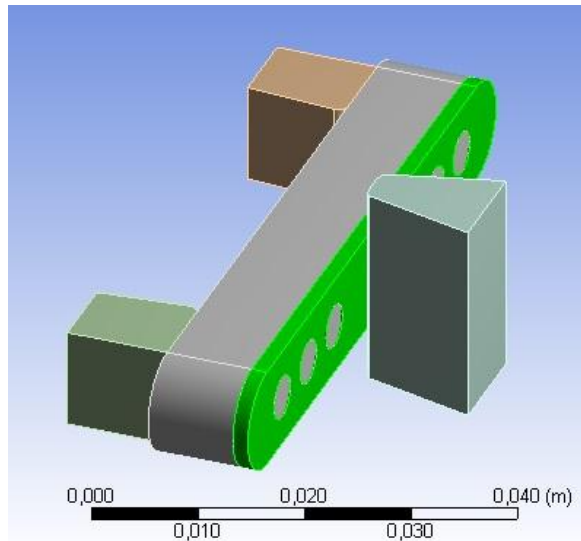


Figure 20: Geometry of analysis model captured from ANSYS 12.1/Explicit Workbench GUI

In the analyses of weapon test (Figure 21), impact mechanism of hummer was simulated exactly that the hummer (both monolithic and inserted) was subjected to an angular velocity, 68.86 rad/s (3.1 m/s at impact point) as initial condition. A target was pin of the rifle (5mm radius cylindrical alloyed steel) which is responsible for initiation of explosion of gunpowder in the shell. As boundary conditions, pin was fixed and hummer motion was limited to direction of angular velocity. Tetrahedron

meshing method was held with extra contact meshing protocol for accurate contact analysis.

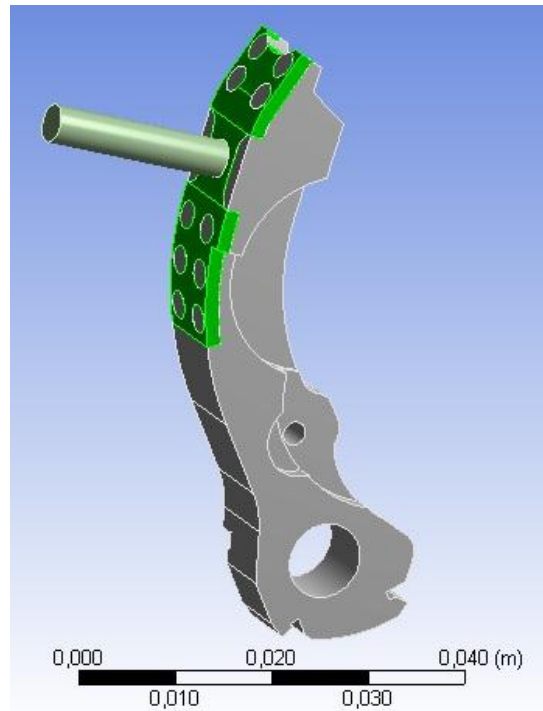


Figure 21: Geometry of analysis model captured from ANSYS 12.1/Explicit Workbench GUI

CHAPTER 4

RESULTS & DISCUSSION

4.1. Interfacial Examinations

4.1.1. AISI D2 insert – A518.0 die casting alloy

SEM image of die cast D2 – A518.0 interface is shown in Figure 22. Both image and EDX line scan results affirmed that neither intermetallics nor similar interaction phases were present along the interface. Thickness of diffusion zone was measured about 9 μ m which was above the value measured in die cast D2 - A413.2 interface (see section 4.1.2.).

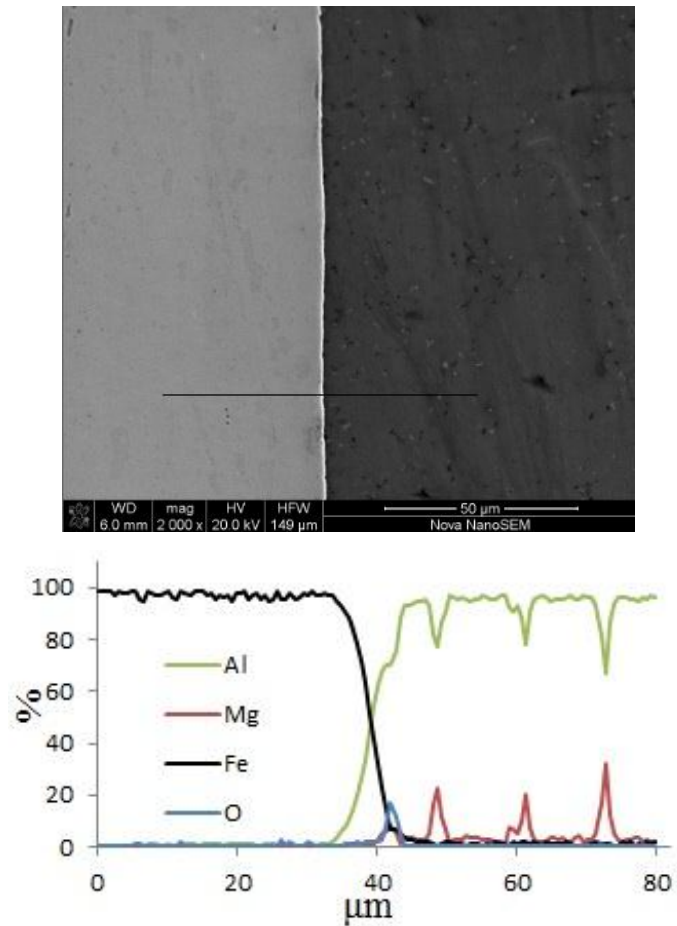


Figure 22: SEM image (S.E.) of the interface between die cast AISI D2 - A518.0 (up) and EDX line scan analysis (left) run on the path shown in the image.

Complete wetting was observed along the interface that there did not exist any interfacial porosity or local delamination. EDX line scan showed that in addition to the elevated oxygen concentration at the interface, interfacial magnesium content was high (9%) relative to the matrix of A518.0 which can be explained by presence of common (Al-Mg-O) oxide film which was formed on surface of solid A518.0 die cast alloy. This protective oxide layer forms the basis of improved corrosion resistance of A518.0 alloy.

4.1.2. AISI D2 insert – A413.2

4.1.2.1. Die cast samples

Die cast D2-A413.2 interface was examined. Local delamination was present along the interface which was probably based on thermal contraction during cooling. The fact which supports this reason is the EDX line scan (Figure 23a) that diffusion of foreign elements across the interface occurred which must have happened before the contact was lost. When the EDX line scan of delaminated interface was compared with non-delaminated one (Figure 23b), It was seen that thickness of diffusion zone of delaminated interface was smaller which shows contact was lost before the diffusion was not completed yet. These facts prove delamination was induced by thermal contraction (i.e. shrinkage). At the regions close to the gates, delamination was overcome by raising the phase-III pressure that higher compression helped to feed the shrinkage probably. However, the effect of elevated phase-III pressure was found to be limited at far regions from the gate of the die.

When the full contact D2-A413.2 interface is focused, EDX line scan results indicated that interfacial diffusion zone was relatively thin which can be related with the effect of silicon content (11.5%) of A413.2 (Figure 23b). The absence of plateau in the EDX line scan analysis pointed that there was no intermetallic formation along the interface. This result supported the insufficiency time spent above the critical temperature for required diffusion and kinetic energy barrier because of the extreme cooling rate.

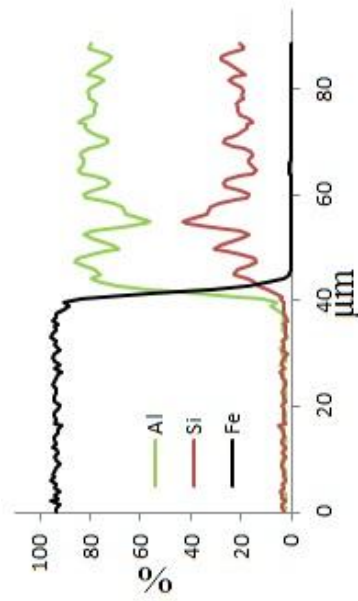
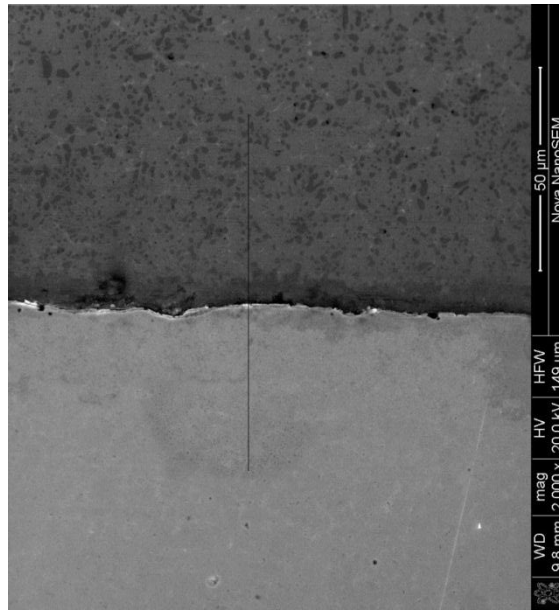
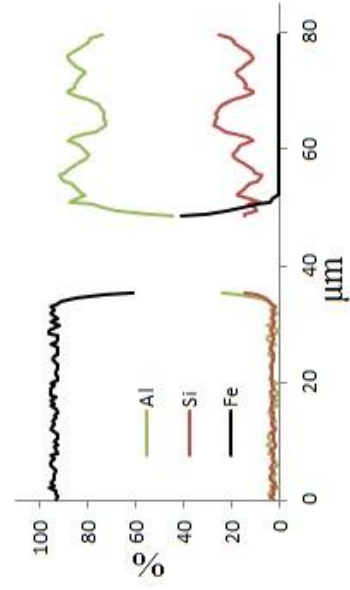
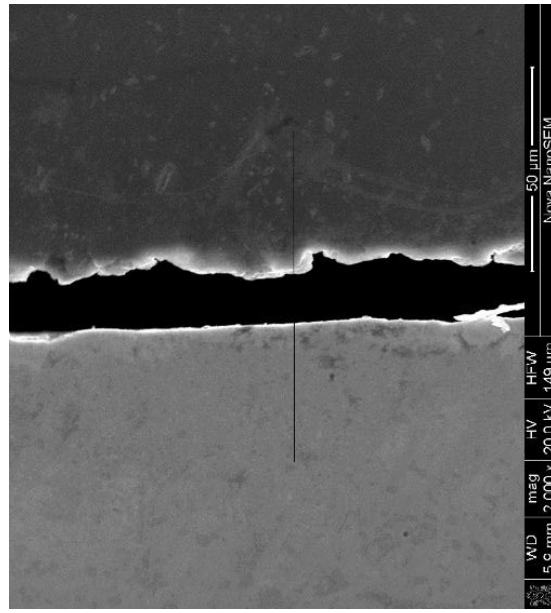


Figure 23: SEM image (S.E.) and EDX line scan analyses of the interfaces between die cast AISI D2 - A413.2 which were belong to delaminated (a) and contacted (b) couples.

Casting temperature was revealed as important in interface morphology. As the total shrinkage is directly related with it, delamination decreased by decreased casting temperature (Figure 24). Thus, casting temperature should be as low as possible.

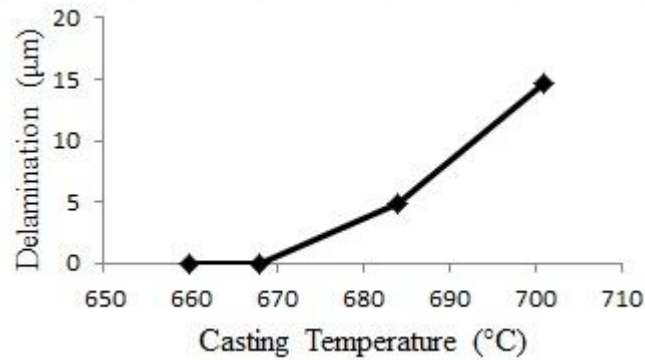


Figure 24: As-cast delamination vs. Casting temperature. Data are taken from the same position of samples casted with same phase-III pressure

Accidentally that one of the shots was poured at 613 °C, which is below the lower limit (~650°C). SEM examination of interface of these samples showed that flakes which was solidified prematurely during the phase-I and phase-II (see section 2.1.4.2.) appeared along the interface as in Figure 25. Solidified flakes, whose elemental composition was 19.5 % Si and 80.5% Al (EDX analysis), were predicted to be carried and stacked to the insert surface by the flow during phase-II. Mechanical influence the flake layer is not able to be guessed; nevertheless it can be qualified as detrimental due to the irregular morphology.

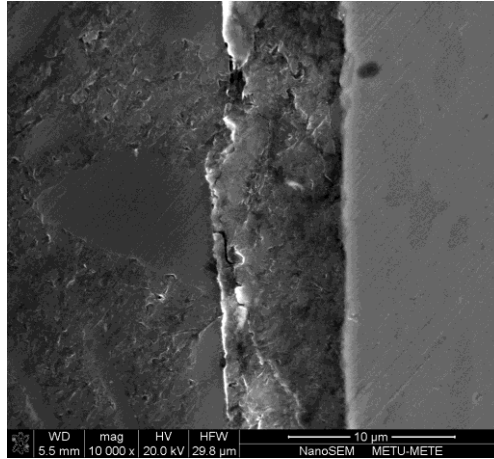


Figure 25: Flake layer observed at the interface of AISI D2 - A413.2 die cast sample whose casting temperature was below the minimum limit. Overall thickness was 18 μ m.

4.1.2.2. Sand Casted Samples

Sand casted D2-A413.0 samples were also examined. Because of low thermal gradient, cooling rate was expected to be quite low with respect to die casting. Thermal analysis results showed that sand casted A413.2, which was superheated up to 690°C, reached 440°C in 17.2 min (1034 sec) with average cooling rate of 0.24 C/sec (as reference, cooling rate of die cast A413.2 was 80-130C/sec). It was expected that slow cooling rate would have allowed further diffusion in sand casted D2-A413.2 samples that a thick intermetallic layer would have appear due to the extreme super-saturation. However, instead of a homogeneous and complete intermetallic layer, locally grown β -Al₅FeSi intermetallic plates appeared as in Figure 26.

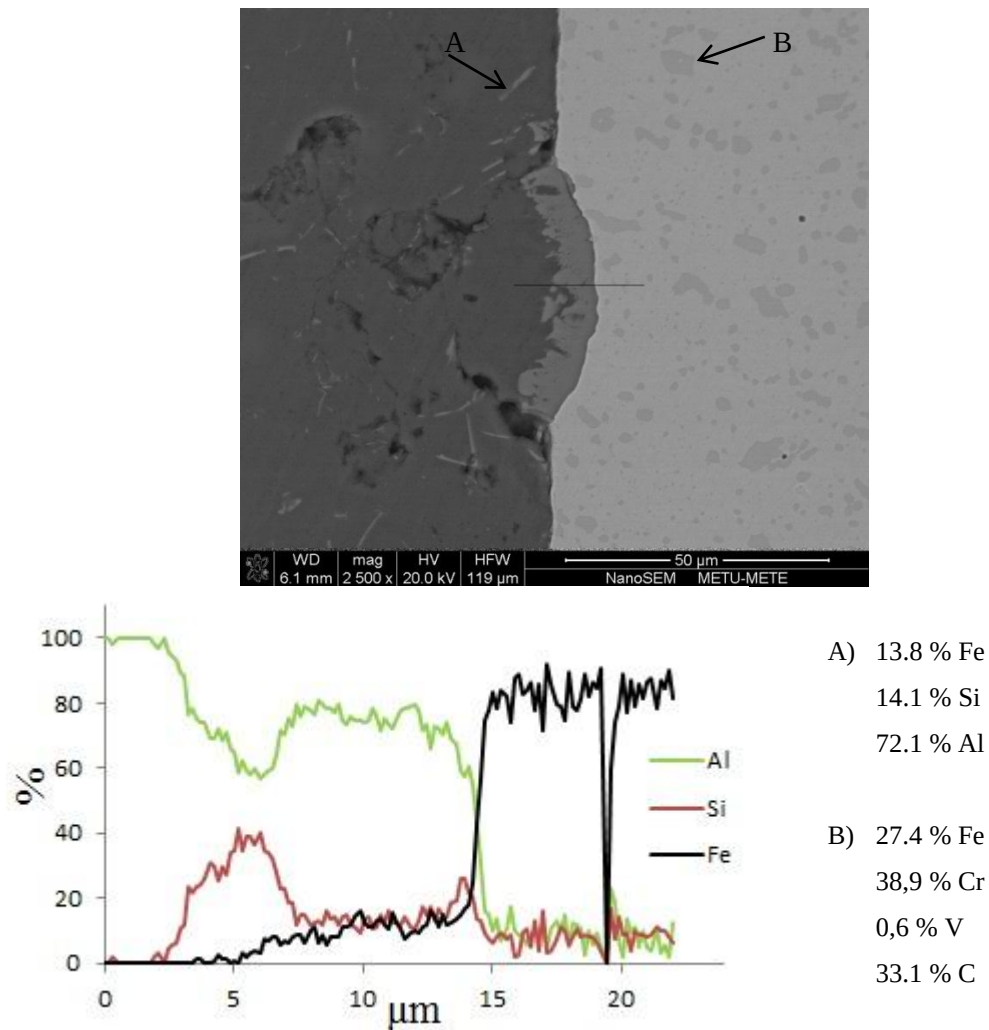


Figure 26: SEM image (Back Scattered) of sand cast AISI D2 - A413.2 interface. EDX line scan analysis proved the presence of β -Al₅FeSi intermetallic layer which was given in below the figure. Other local EDX analysis results (atomic %) of A and B was given below. Analyses showed that A was intermetallic layer and B was (Cr,Fe)₇C₃ carbide.

The cracks and porosity at the A413.2 side were evident in Figure 26. These defects were observed to become frequent at regions close to the concave intermetallic plates. As their presence is critical due to the mechanical consequences, it was important to understand the reason which was associated with combined effects of thermal stresses and porosity. Soon after the casting, β -Al₅FeSi intermetallic plate nucleated and grew at the interface. Because of the remarkable density difference between the A413.2 (2.64 g/cm³) and β -Al₅FeSi (3.60 g/cm³) induced a tensile stress which continuously increased during the growth of β -Al₅FeSi. In the meantime, the rest of the A413.2 started to solidify and continued to cool with thermal contraction. This

shrinkage acted as source of another tensile stress which was perpendicular to the interface. This stress was not able to be relieved by delamination (as it was seen in die cast samples, section 4.1.2.1.) due to the high strength of $\beta\text{-Al}_5\text{FeSi}$ which conserved the contact at the interface. Because of stress concentration sites such as pores close to the interface, combination of these two tensile stress sources was sufficient to propagate cracks. This mechanism was also described in Figure 27 with comparison between die cast and sand cast.

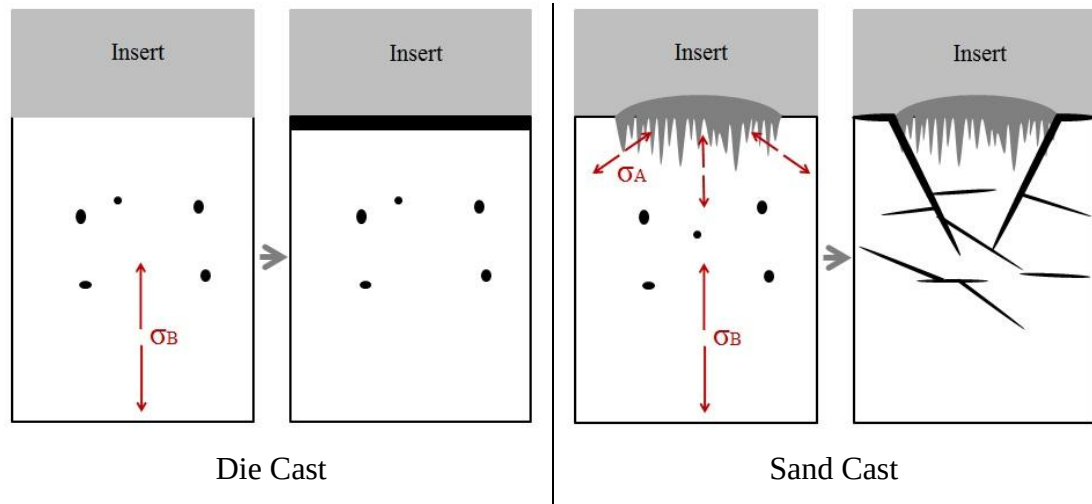


Figure 27: Comparison of inserted samples produced by die casting and sand casting. Crack propagation was occurred in sand cast samples due to the stresses induced by intermetallic layer. σ_A is the stress applied by the density change during the growth of the intermetallic layer $\beta\text{-Al}_5\text{FeSi}$. σ_B is the stress occurred during the shrinkage of the die casting alloy.

As in Figure 26, it was interesting that $\beta\text{-Al}_5\text{FeSi}$ did not only form along the interface. At regions close to the interface, needle-like $\beta\text{-Al}_5\text{FeSi}$ intermetallic was present in the matrix of A413.2 which was labeled as “A” in Figure 26. This result supports the cooling rate dependence of intermetallic formation. Despite the same Fe concentration (0.6%) of in both die cast and sand casted A413.2 alloy before the casting, the fact, dispersed intermetallic β -platelets existed in sand casted but not in the die cast samples, was directly related with three reasons.

- 1) As the cooling rate was relatively low in sand casting, sufficient thermally-induced diffusion occurred for super-saturation which is the driving force of intermetallic formation.

- 2) In equilibrium condition which was satisfied by sand casting, according to the Al-Fe-Si ternary phase diagram (Figure 28), diffused iron concentration was above the critical limit where $L \rightarrow (Al) + Al_5FeSi$ reaction took place. However in non-equilibrium condition where extreme cooling rate existed as in the case of die casting, the limit shifted to higher Fe concentration. This change in phase diagram meant solubility of iron was increased that super-saturation might have not been reached to form $\beta-Al_5FeSi$.
- 3) Even if reaction had took place in HPDC, $\beta-Al_5FeSi$ would have had very small size due to the low dendrite arm spacing (supposing primary Al dendrites formed due to the shift of eutectic composition to higher Si concentration due to high undercooling in HPDC).

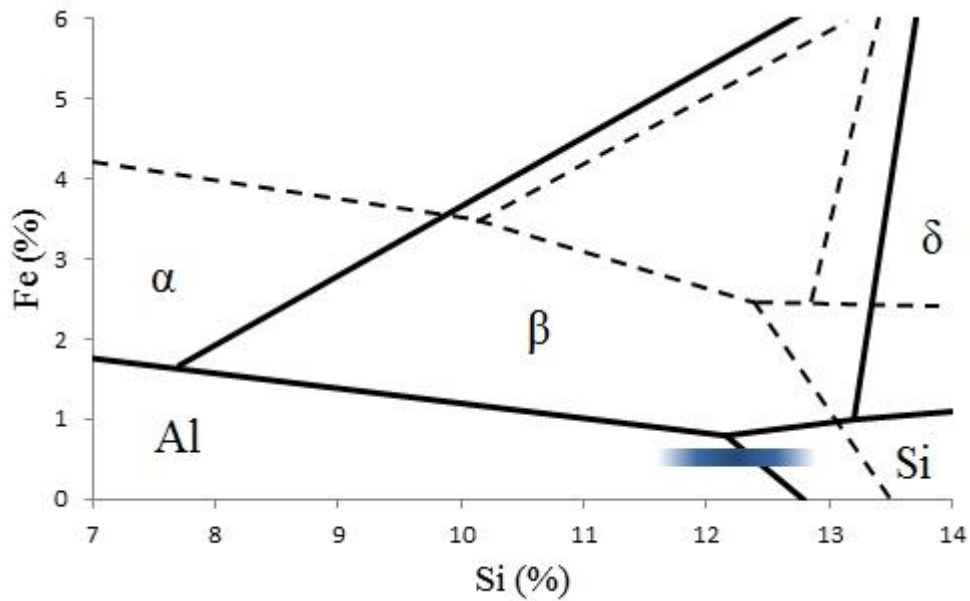


Figure 28: Al corner of Al-Fe-Si ternary phase diagram. Straight line is the equilibrium phase boundaries, while dashed lines symbolizes non-equilibrium condition with cooling rate about 100°C/s. Lines of non-equilibrium condition was drawn according to references [43,63]. Blue region represents the interval of composition of A413.2 alloy.

Mechanical effects of dispersed $\beta-Al_5FeSi$ platelets were considered as harmful due to two common consequence of intermetallic formation. Because of the needle-like

morphology, β -Al₅FeSi platelets contribute crack propagation, besides, intermetallics lead significant increase in micro-porosity [43]. (see section 2.3.1.)

4.1.2.3. Die Cast A413.2 with Immersed D2 inserts

Samples, whose inserts were immersed into the molten A413.2, were examined to observe the microstructural effects of completely grown intermetallic layer along the interface. Samples have been hold in molten A413.2 (700°C) for 5, 10 and 20 seconds whose SEM images were shown in Figure 29 respectively. Immersion led rapid formation of intermetallic layer which was characterized as β -Al₅FeSi (by EDX analysis shown in Figure 29). Because the diffusion coefficient and iron solubility in aluminum was extremely high relative to the solid A413.2, very short immersion time was enough for growth of β -Al₅FeSi. It was observed that when the molten state diffusion was allowed, local β -Al₅FeSi plates tended to grow along the direction parallel to the interface and after complete cover, it grew through A413.2 matrix.

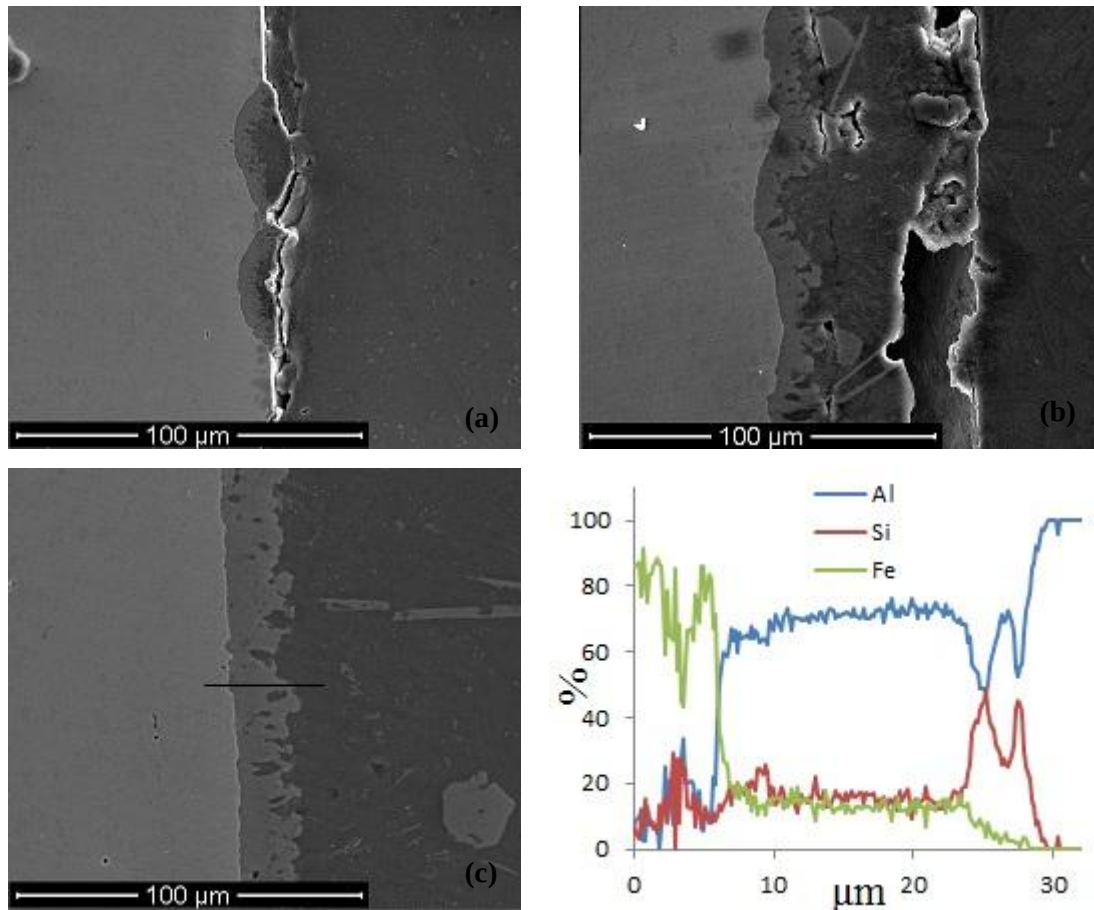


Figure 29: SEM images (S.E.) of the interface between die cast AISI D2 - A413.2 samples whose inserts were immersed into molten A413.2 alloy for 5 seconds (a), 10 seconds (b) and 20 seconds (c). EDX line scan analyses run on path on the sample (c).

The β - Al_5FeSi -induced subsequent crack formation was observed in samples whose inserts had been immersed. Similar to the case of sand casted samples, intermetallic layer led direct and indirect mechanisms which induced tensile stress in direction perpendicular to the interface. This stress causes crack propagation parallel to interface.

Another consequence of enhanced diffusion which was already mentioned at sand casted samples was also seen in immersed D2- A413.2 die cast samples. Dispersed β - Al_5FeSi intermetallic platelets appeared in A413.2 matrix, especially at regions close to the interface. Platelet of both samples showed similarity in dimension and morphology, but there existed some macro-scale β - Al_5FeSi particles with non-uniform morphology (Figure 29). Dimensions of some of these particles transcended

the thickness of the interfacial β -Al₅FeSi layer that it was unexpected fact. During the immersion, induction furnace was paused that dissolved iron was not able to be mixed by magnetic induction. Reason was explained in ref [63] that while most of the β -Al₅FeSi nucleates after eutectic growth, some primary (formed before the eutectic reaction) β -Al₅FeSi particles existed due to the non-homogeneous iron dissolution.

Mechanical effects of intermetallic layer were discussed in section 4.2.2. where die cast impact test samples whose inserts were immersed into molten A413.2, prior to the casting.

4.1.3. AISI D2 insert – ZAMAK 5

Interface between die cast zinc-based Zamak5 and D2 insert was affected by porosity. Zamak5 is generally casted within very narrow superheat temperature (~max. 450°C) limits since it is very prone to dissolved gas porosity and shrinkage at high temperatures. This limitation led some problems in this study that since inserts were not able to be pre-heated, extreme thermal gradient existed between insert and molten Zamak5 which led faster and earlier solidification relative to the other regions of the cavity. The entrapped air carried by the flow and dissolved gases were not able to leave the cavity due to mentioned earlier solidification. As a consequence, gas porosity band was seen near the interface (Figure 30)

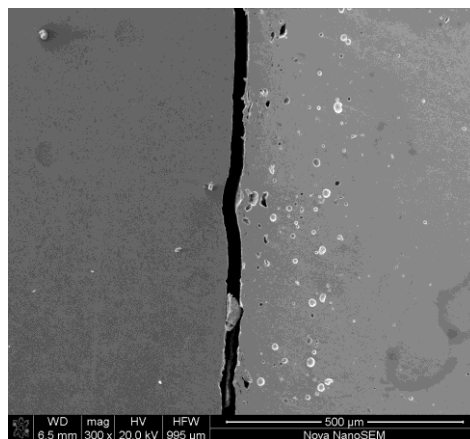


Figure 30: Porosity band and delamination seen in the interface of die cast AISI D2-Zamak5 samples.

Gas porosity band was also found to be related to distance from the gate of the cavity which showed the effect of feeding efficiency. As it was seen in Figure 31, porosity density increased by distance travelled from the gate. It was not unexpected that since the high-pressure feeding was maintained from the gates where the solidification and temperature drop was delayed relative to the far regions, gases (air + dissolved gases) was able to be forced to leave by compression.

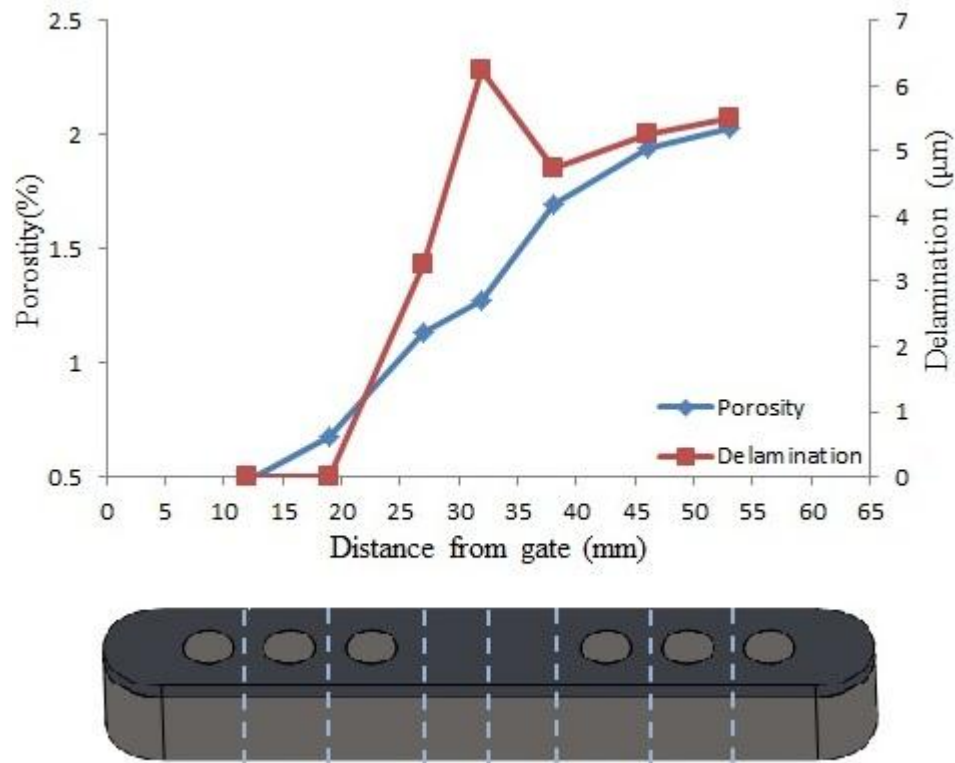


Figure 31: Porosity band and delamination distribution along flow direction. These data were taken from sample casted without any precautions or modifications.

Porosity band would have been overcome by increasing superheat temperature of molten alloy (i.e. pouring temperature) however it led another problem which was the reason of strict superheating limits in practices of die casting of zinc alloys. When the alloys was poured at 510°C, Due to the extreme shrinkage and dissolved gases, centerline porosity (Figure 32) occurred in where the latest solidification was predicted.

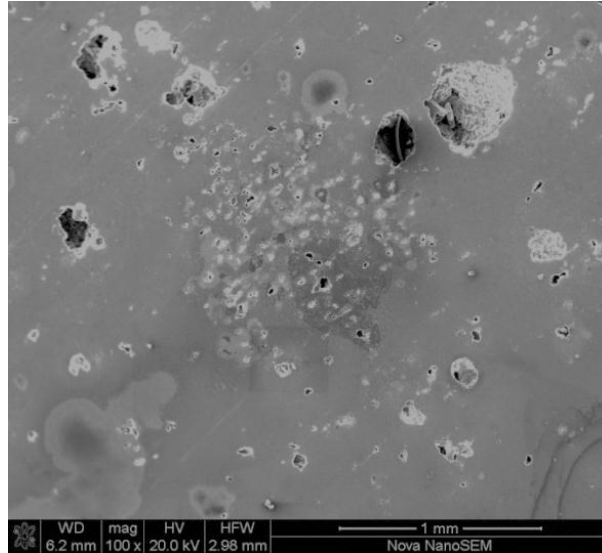


Figure 32: Centerline porosity which was induced by shrinkage of the Zamak5. Over-superheated samples has tendency to have such shrinkage-related defects.

Porosity band was avoided effectively by increasing gate speed and phase-III pressure (above 160 bars). Accelerated flow (by increasing phase-II velocity above 2.3m/s) provided molten metal to fill the whole cavity faster that porosity was able to leave the alloy prior to the solidification. Elevated phase-III pressure forced porosity toward the vents by extreme ductility (due to thermal softening) at temperature just below solidus. Shrinkage-induced interfacial delamination was also overcome by these adjustments. Morphologically perfect interface was maintained as in Figure 33.

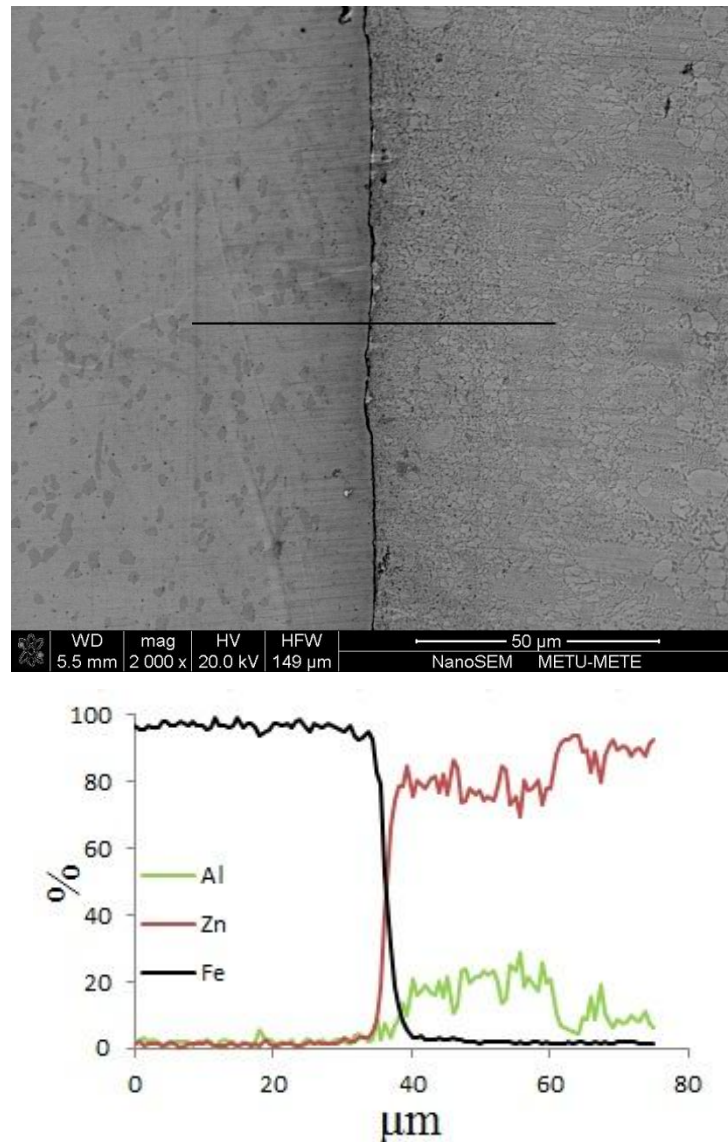


Figure 33: SEM image (S.E.) of the interface between die cast AISI D2 - Zamak5 (up) and EDX line scan analysis (left) run on the path shown in the image.

As it is seen in Figure 33, there existed no intermetallic phase along the interface of D2 - Zamak5 die cast couple. This result was expected because of the similar results found in other couples that the insufficiency of diffusion and rapid heat loss avoided both nucleation and growth of intermetallic compounds. Besides, aluminum content of Zamak5 alloy was quite high that Al-Fe interaction dominates all other reactions. It was not expected to see Al-Fe intermetallics at Zn-based samples while no intermetallics formed at Al-based samples.

4.2. Mechanical Tests

4.2.1. Three Point Flexural Test

3 point bending test results were given in Figure 34. Test stopped immediately after A413.2 has failed. It was not surprising that under quasi-static loading, inserts had very limited amount of strengthening effect. Static loads were not able to be reduced by inserts due to elevating strain. Increasing elastic deformation raised tensile stress on the faces where the cracks grew and elasto-plastic failure occurred. It should be noted that the inserts and A413.2 did not fracture simultaneously (Figure 35). Inserts persisted in going on plastic deformation even after twice of stroke movement which aluminum section has failed.

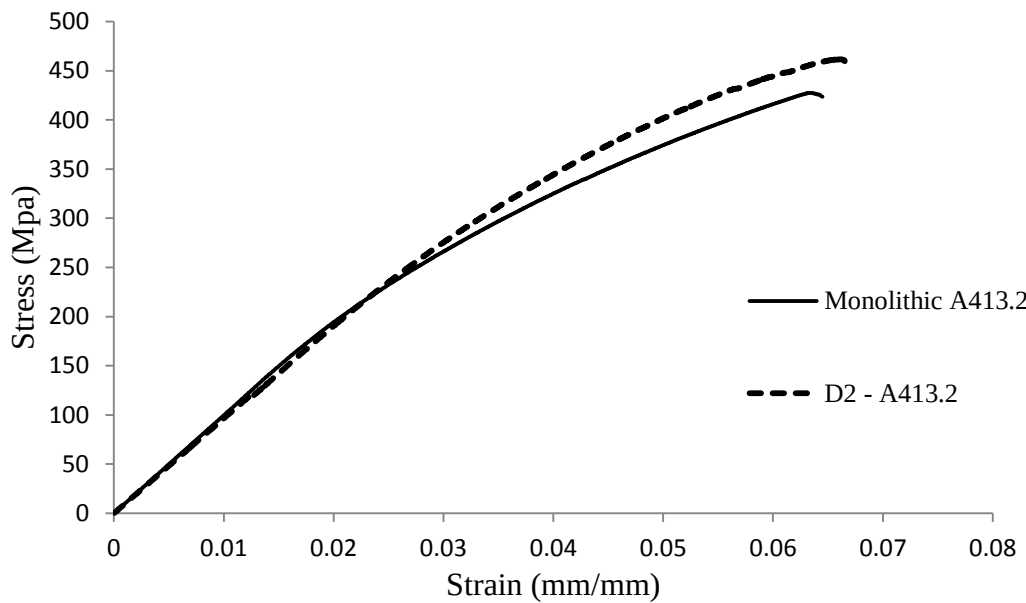


Figure 34: Three point bending test results of both monolithic (A413.2) and inserted (D2-A413.2) samples. Tests ended when aluminum failed.

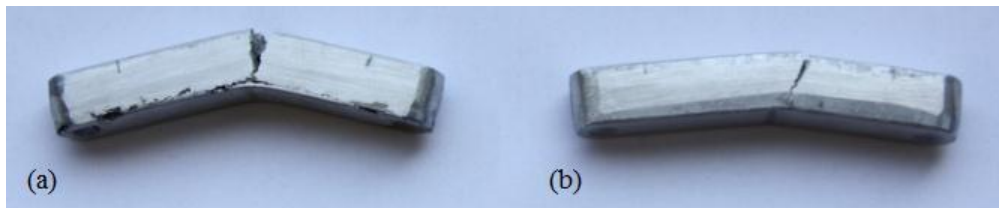


Figure 35: Bending test samples; inserted D2-A413.2(a) and monolithic 413.2 (b)

4.2.2. Charpy Impact Test

Impact test results are shown in Table 12. Standard deviations of each experimental group were low enough to make accurate measurement and comparison. In normal group, inserts had positive effect on energy absorbance of samples as it was expected and this increase was above the energy absorbance of inserts. This fact supported the benefits of impedance mismatch in layered parts.

Table 12: Charpy impact test results in joules (j)

Alloy	Group Code Name	Test1	Test2	Test3	Mean
A518.0	CG-I	24	25	25	24.67 ± 0.58
	CG	15	14	15	14.33 ± 0.58
	OIG	16	17	17	16.67 ± 0.58
A413.2	CG-I	19	18	18	18.33 ± 0.58
	CG	11	11	11	11.00 ± 0.00
	LIG5	8	8	9	8.33 ± 0.58
	LIG10	6	7	7	6.67 ± 0.58
	SC-I	9	9	8	8.67 ± 0.58
	SC	6	6	7	6.33 ± 0.58
Zamak5	CG-I	30	30	30	30.00 ± 0.00
	CG	19	21	21	20.33 ± 1.15
Insert (alone)	I	5	5	5	5.00 ± 0.00

Using inserts yielded increase in energy absorbance which was measured as 72.1 % (10.33 j), 66.7% (7.33j) and 47.6% (9.67) in 518.0, 413.2 and Zamak5 die cast normal group samples respectively. There was correlation ($r=0.92$) between percentage increases and impedance ratio (between D2 insert/ die casting alloy, see Table 13) whose mechanism was not able to be determined only with these data unfortunately, since Charpy impact test gave only energy absorbance of total sample. For this purpose, Finite Element Analysis results were considered (see section 4.3.1.).

Inserts, whose surface was oxidized by flame prior to casting, had very limited enhancement in energy absorbance of A518.0 alloy. It was probably based on the facilitation of crack growth which was induced by surface corrosion. (see Figure 36) Despite the high chromium content in D2 tool steel, oxide layer was iron based, because most of the chromium content tended to precipitate as $(Cr,Fe)_7C_3$ (chromium carbide which gives required hardness) instead of forming an protective oxide layer like a stainless steel.

Table 13: Calculated transmission and reflection ratio of stress and energy at interfaces according to equations 13 and 14.

Interface	Impedance Ratio	Transmission Ratio		Reflection Ratio	
		Stress Amp.	Energy	Stress Amp.	Energy
D2 – A518.0	2.877	0.516	0.766	-0.484	0.234
D2 - A413.2	2.787	0.528	0.776	-0.472	0.223
D2 - Zamak5	1.723	0.735	0.070	-0.265	0.930

Negative value represents change in stress sign, compressive to tensile and vice versa.

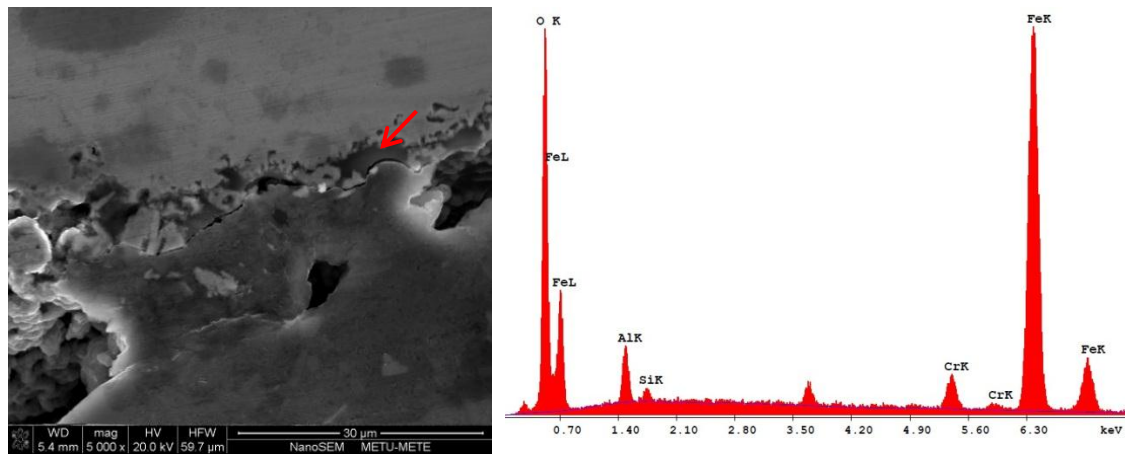


Figure 36: SEM image (S.E.) and EDX analysis of the oxide layer on flame-heated inserts.

Impact test results of samples (LIG5 AND LIG10), whose inserts had immersed into molten die casting alloy for 5 and 10 seconds respectively, showed the effect of detrimental intermetallic layer at the interface. Test results of these samples were even below the test results of monolithic A413.2. It means immersed inserts did not enhance toughness, moreover they damage it dramatically. The reason was associated with the mechanisms mentioned in section 4.1.2.2. Intermetallic layer

indirectly induced residual stresses and cracks which contributed to the drop of impact toughness. As the immersion time increased, intermetallic layer grew and its effects become more severe. Decrease in energy absorbance was measured about 24.3% (2.67 j) and 39.4% (4.33 j) in samples immersed for 5 seconds (LIG5) and 10 seconds (LIG10) respectively.

Alone D2 inserts, which were treated and shaped exactly same to ones used in normal group, were subjected to the impact test to find the monolithic toughness of them (I). All three samples gave the same energy absorbance value, 5 joules. This data was used as reference in comparisons.

4.2.3. Weapon Mechanism Test

Three die cast test groups, which are monolithic Zamak5 (HZ) and D2 inserted Zamak5 (HZ-I), were tested in dedicated H&K 33E rifle with real ballistic cartridge.

One of the monolithic Zamak5 hummer failed after the first impact on the pin. As it was seen in Figure 37a failure was described as fracture and it was associated with sudden crack propagation induced by dynamic bending stress which was concentrated by porosity. It was proved by explicit finite element analysis shown in Figure 37b. Since the hummer of the H&K 33E rifle should perfectly operate even after 3000 shots [64], monolithic Zamak5 sample gave very unsatisfactory result, as it was expected.

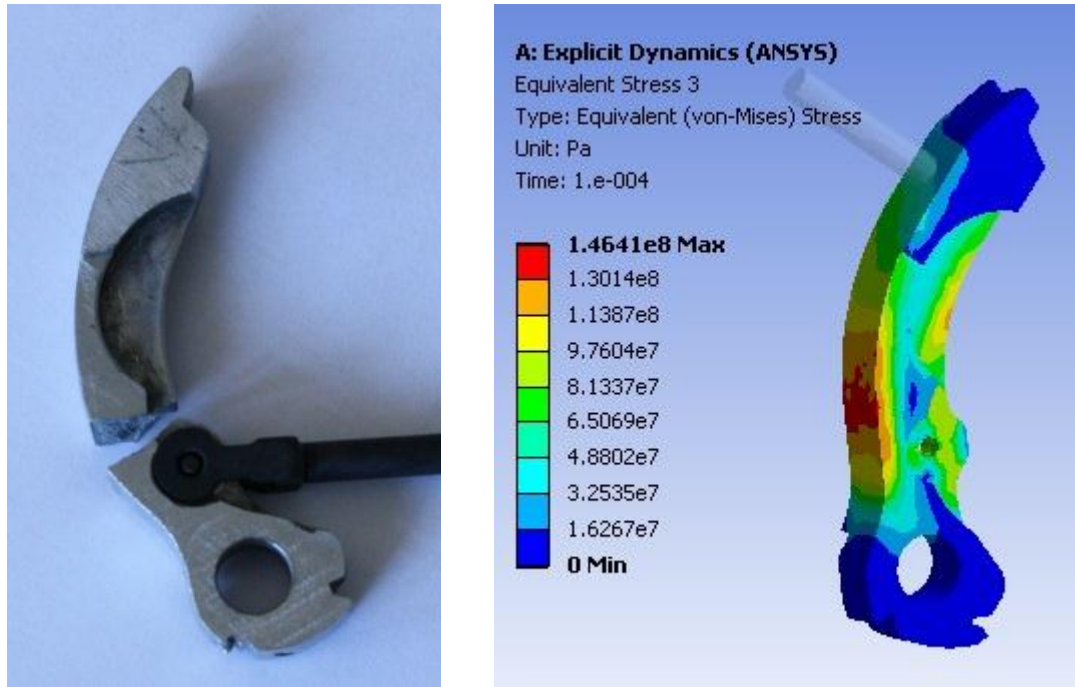


Figure 37: Fracture of monolithic Zamak5 hammer tested in rifle (a) and equivalent stress distribution after the impact (b).

180 cartridge perfectly shut by low porosity monolithic Zamak5 samples. However, after the disassembly, severe plastic deformation was observed on the impact point which was not acceptable according to specifications [64]. A crater was formed due to the impact with 0.4 mm depth and 5mm diameter (pin diameter) which was shown in Figure 38. Neither fracture nor initiated cracks were detected in whole body which pointed that impact toughness of the die cast Zamak5 alloy was satisfactory but due to the low hardness (108 HB) and, impact surface was inelastically deformed by the pin whose hardness was read as 58 HRC (~620 HB). Thermal softening was also possible factor enhancing plastic deformation because of the heat generated by dynamic loading.



Figure 38: Operated monolithic (non-inserted) Zamak5 hammer. Despite perfect shots, inelastic deformation was occurred at the impact point.

D2 steel inserted Zamak5 die cast samples performed 180 perfect shots without any fracture, premature cracks or deformation (Figure 39). Because of the extreme surface hardness (58-59 HRC) of AISI D2 tool steel insert, there was not even sign of impact at the contact point. Macro delamination, which was expectable, was not detected at the interface between insert and die cast part.



Figure 39: Die cast AISI D2-Zamak 5 inserted hammer. Neither deformation nor cracks were visible after hundreds of cycles.

Micro inspections were also carried out on operated inserted samples that the interface was focused for possible deformations and defects. In the light microscopy images of the cross section exactly below the impact point (cross section where the impact axis lies on), there was no local delamination, deformation or premature cracks as it is shown in Figure 40.

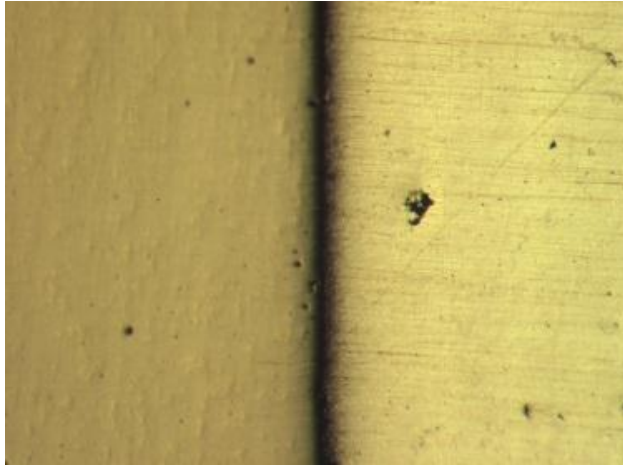


Figure 40: Light microscopy image of the interface just below the impact point of the inserted D2-Zamak5 sample (50x, No etching)

In other cross sections there was not any defects too, except a crack initiated at one of the corners of the inserts which was reamed insufficiently. This sharp corner, shown in Figure 41, probably acted as stress concentrator which is most common reason of cracks lead catastrophic failures. This result showed the severity of the sharp corners which should be strictly avoided by reaming or lowering the angle of corners during design and machining of inserts.

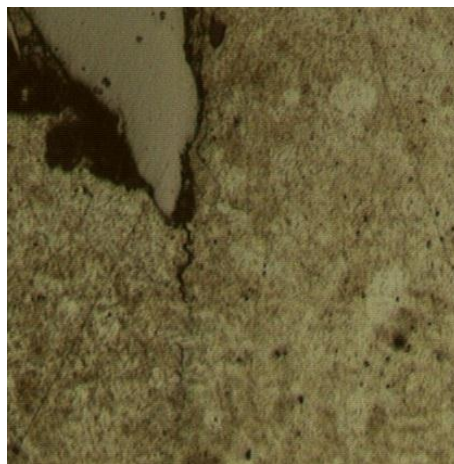


Figure 41: Light microscopy image of the crack initiated from the sharp corner of insufficiently reamed insert of operated D2-Zamak5 sample (200x, no etching)

Weapon test results showed the efficiency of inserted die cast alloys under non-destructive, cyclic, impulsive loads. In order to reveal the mechanism of the efficiency, finite element analyses were focused.

4.3. Finite Element Analyses

4.3.1. Charpy Impact Test Analyses

Charpy impact test analysis results were given in Table 14. These values represented the kinetic energy loss of pendulum after the fracture which meant the energy absorbed by the sample.

Table 14: Results of Charpy impact analyses of inserted and monolithic samples

	Absorbed Energy (j)		% Increase
	<u>Monolithic</u>	<u>Inserted</u>	
A518.0	13.91	20.15	44.9
A413.2	13.48	18.88	40.1
Zamak5	22.71	28.72	26.5

The results of the analyses supported the efficiency of the inserts on improvement of toughness which was already proved by real test results. This improvement was provided by increase in plastic work (energy absorbed during the plastic deformation) done by the die casting alloy showed in Table 15.

Table 15: Energy absorbed by only plastic deformation calculated in Charpy impact analyses.

	Plastic Work (j)		% Increase
	<u>Monolithic</u>	<u>Inserted</u>	
A518.0	8.69	13.19	51.8
A413.2	7.59	10.38	36.8
Zamak5	11.92	15.44	29.5

As all analyses were performed under same boundary and initial conditions, the reason of increase in the plastic work was associated with the brittle to ductile transition of die casting alloy even under same impact velocity and temperature. Such transition was able to be maintained only by reducing the localization of the energy and stress. As it was seen in Figure 42, just before the crack initiated ($t=2.4 \times 10^{-4}$ s), magnitude of the equivalent stress (Von Mises) measured in monolithic A413.2 (a) was almost equal to the inserted A413.2 (b). However, in inserted A413.2 (b) it was visible that stress was distributed widely and did not localized as much as seen in monolithic A413.2 (a). Reason of this difference might be associated with the fact that, in inserted A413.2, load was not applied by pendulum but by inserts whose contact surface was much larger. As it was mentioned before, dynamic load sources such as impacts lead localization of stress since they do not allow stress to distribute as perfect as in static cases due to limited time and on-going deformation. In monolithic A413.2 (a), pendulum directly hit on the surface of A413.2 alloy that load was applied by pendulum which had very small contact surface. As a result, stress was localized at the region falls below the impact surface.

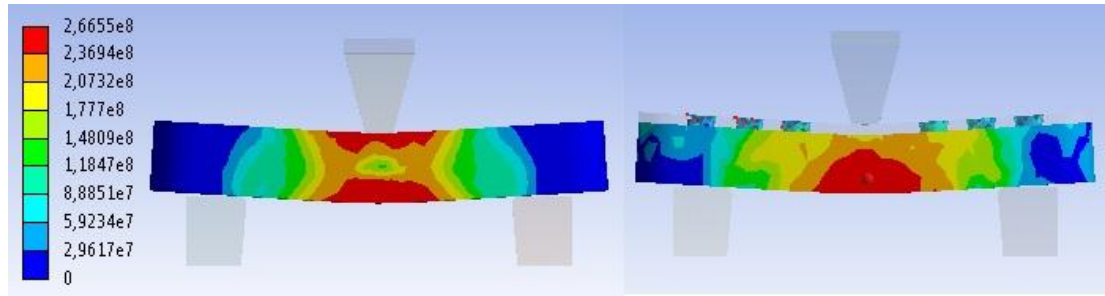


Figure 42: Comparison of distribution of Equivalent Von-Mises stress (in Pa) on monolithic (a) and inserted (b) samples just before the crack initiated ($t=2.4 \times 10^{-4}$).

In the case of inserted A413.2 (b), pendulum hit the insert and the energy of the impact, carried by stress waves, was transmitted to die cast body. However this time, some amount of the energy was reflected back by the insert-die cast interface and carried by horizontal (perpendicular to impact direction) stress waves traveled in the insert. When these spread stress waves reached the interface again, some of the

energy was transmitted to die cast body. However this subsequent transmission was performed at regions far from the impact point which provided larger load transmission zone and reduction of localization.

Figure 43 showed that the equivalent plastic strain distribution was affected by this fact. Inserted A413.2 sample (b) evidently faced higher plastic deformation prior to the fracture. As load transmission zone was extended, localization of stress was partially eliminated and stress was distributed widely. As a result, plastic deformation occurred at larger volume and more kinetic energy converted into plastic work relative to that of monolithic A413.2 sample (a). Plastic strain rate is directly related with the shear wave velocity which is quite slower than longitudinal waves. Thus localization is very important for plastic behavior of the beams under dynamic flexural loads. Overcoming localization allows more plastic strain occurred prior to the crack propagation by incident stresses.

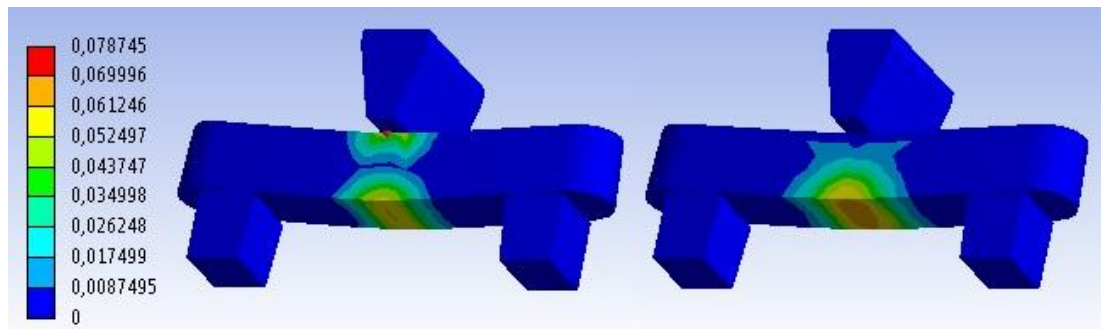


Figure 43: Comparison of equivalent plastic strain on monolithic (a) and inserted (b) samples just before the crack initiated ($t=2.4 \times 10^{-4}$).

Efficiency of inserts was influenced by the impedance mismatch between the insert and the die casting alloy. The higher the difference, the more stress and energy reflected from the interface which was desired result. This fact was strongly supported by results of both real impact tests and analyses (Figure 44). It should be underlined that mechanical impedance of inserts must be higher than impedance of die casting alloy for the purpose.

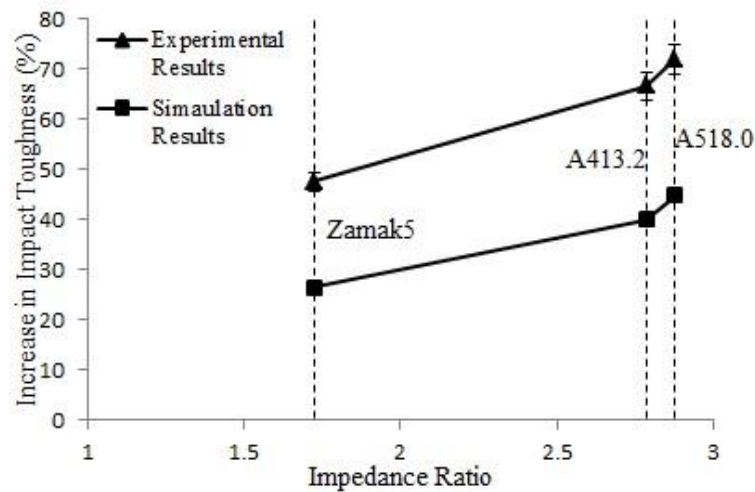


Figure 44: Percent improvement of energy absorbance vs. impedance ratio (between insert and die casting alloys) graph according to real Charpy test and Explicit FE analysis results.

Both real and simulated (FEA) Charpy impact test results proved that inserts had positive effect on energy absorbance of die cast samples under destructive loads. However there are remarkable difference between quantitative results of analyses and real test as seen in Figure 44. Real test results gave percent increase in energy absorbance which was almost twice of that of finite element analysis results. This difference was considered to be related with metallurgical effects of inserts such as cooling rate and approximation errors in variable geometries. These factors were not able to be taken into account in numerical calculations of FE analyses due to being specific for each case.

The load spreading ratio, Horizontal (radial) / Vertical (axial), indicates spreading capability of the inserts which should be maximized as much as possible. As it was mentioned in section 2.2.3. , this ratio was influenced by impedance mismatch and strength of the materials but the most accurate factor was determined as NDE (Normalized Dissipative Energy) distribution along the radial axis. Formulation of NDE was given in equation 15. It was mentioned that best load spreading capability (i. e. high spreading ratio) is belong to parts with lowest NDE gradient. As it was seen in Figure 45, radial NDE distribution gradient of D2-A518.0 was lowest and D2-Zamak5 had the highest. This result showed correlation with the percent increase

in energy absorbance provided by both experimental Charpy tests and analyses. High NDE gradient meant extreme localization of energy at the impact point which pointed that stress was not able to be spread along horizontal direction. Thus it is desired to have a NDE distribution gradient as low as possible for minimum energy (also momentum) transmitted to die casting alloy.

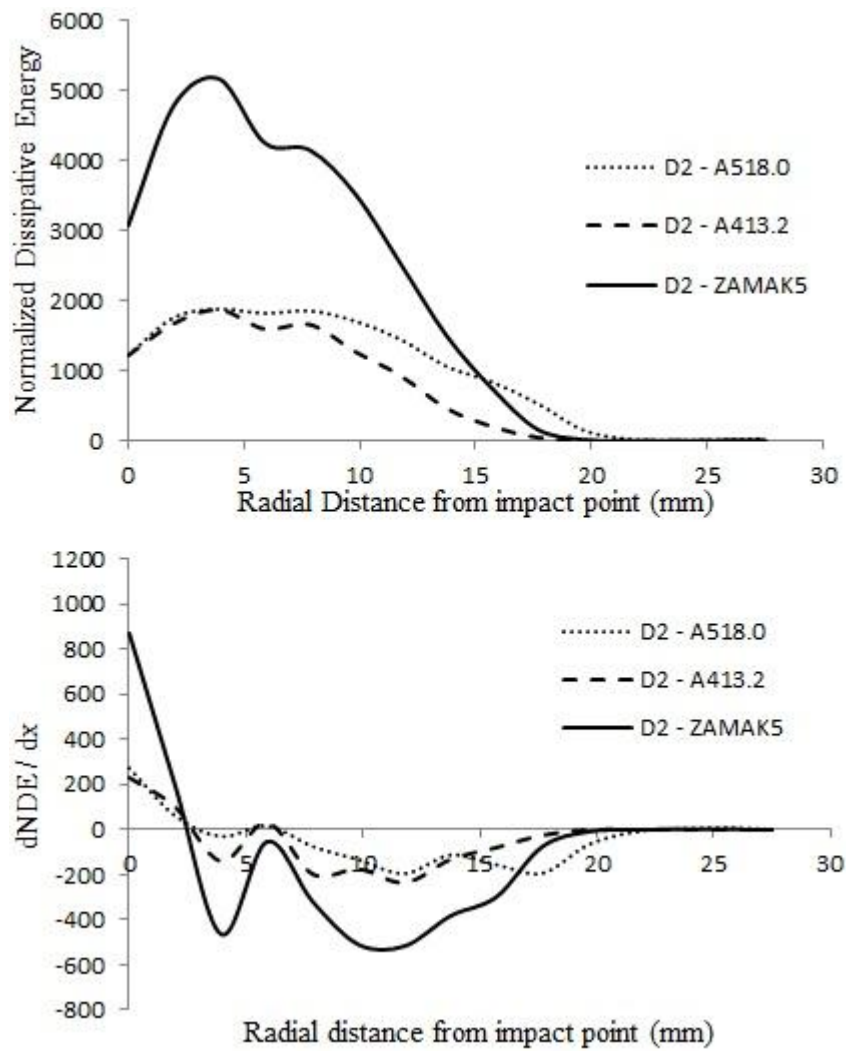


Figure 45: Radial NDE distribution of inserted samples (a) and slope of distribution (b)

4.3.2. Analysis of Operating Hummer of H&K33

Service conditions of hummer of H&K33 infantry rifle was simulated via ANSYS 12.1 FEA software in order to see the effect of inserts under non-destructive loads. Analysis results showed that the plastic deformation at the impact point was totally avoided by the insert. Even the incident stress at the impact point of inserted samples was higher than that of monolithic ones; no plastic deformation was occurred due to high strength impact surface.

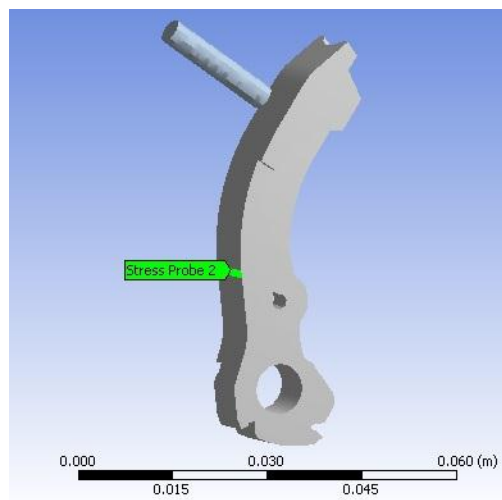


Figure 46: Position of stress probe used in E-FEA simulations.

The critical region, where monolithic Zamak5 sample was fractured (see Figure 37), was also focused. A probe was attained (Figure 46) to this region that measure maximum stress by time. After the analysis of both monolithic and inserted samples, collected data were compared. As it was seen in Figure 47 It was found that maximum stress read in inserted samples was almost quarter of stress measured in monolithic samples.

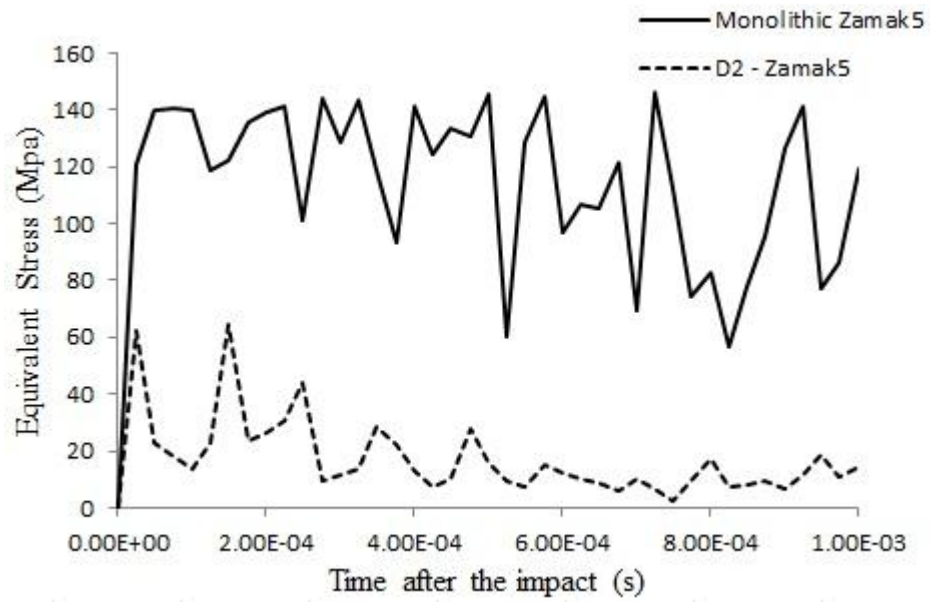


Figure 47: Equivalent (Von-Mises) Stress calculated in critical region of monolithic Zamak5 (straight line) and D2 – Zamak5 (dotted line)

The maximum stress measured by the probe of monolithic sample was 146.41 MPa while it was 64.11 MPa in inserted one. The effect of inserts on reducing the stress, transmitted to the die cast medium, was strongly supported by the explicit finite element analyses. Energy and load was reflected from the insert – die casting alloy interface due to the impedance mismatch. Some amount of reflected load was spread in the insert perpendicular to the impact direction while some amount of it was transmitted back to the impactor by elastic strain energy.

CHAPTER 5

CONCLUSIONS

In this study, it was focused on the improving impact resistance of the die cast parts by employing steel inserts, in order to widen economic benefits provided by die casting process. Results of the study pointed that the goal was achieved and the improvement was verified as satisfactory for industrial applications. While this thesis was being written, this study was put into practice in light weapon factory of Mechanical & Chemical Industry Co. with some classified modifications.

Other important conclusions derived from the tests and analyses were given above:

- 1- *Effectiveness of insert increases with higher impedance mismatch.* Since the mechanical impedance difference between insert and die casting alloy is the main factor controlling energy and stress transmission, the mismatch should be maximized as much as possible to reduce energy and load transmitted through the die casting alloy. During the material design, it should be kept in mind that impedance of the inserts must be higher than that of die casting alloys.
- 2- *Inserts are effective under dynamic loads.* Inserts which are perfectly designed as impact barrier may not improve static durability of the part. It is due to the fact that inserts were used for purposes (1) increasing impact-surface hardness to avoid deformations and wear, (2) reducing the amplitude of incident waves transmitted to low strength region(die casting alloy), (3)

Dividing a single incident stress wave into several weak waves, (4) Spreading the loads along perpendicular direction of impact. These mechanisms are not effective on strengthening whole part against static flexural or tensile loads. But if it is insisted, the thickness should be increased.

- 3- *Intermetallic layers are detrimental.* Intermetallic layer between steel insert and non-ferrous alloy, induce tensile stress by local density change during its growth and shrinkage during solidification of die casting alloy. This stress may remain under elastic limit as residual stress which amplifies the loads and accelerates the failure. In worse case, it may exceed yield strength around the stress intensity sites such as pores which leads crack initiation even without any service load.
- 4- *In High Pressure Die Casting, intermetallic layer was inhibited by high thermal gradient.* Fortunately intermetallic layers do not occur in inserted die cast parts due to three reasons; (1) High cooling rate causes sudden drop of temperature especially at the interface that, diffusion of atoms in foreign side was limited, (2) High thermal gradient leads non-equilibrium cooling which reduces the driving force of super-saturation of diffused atoms, (3) sudden loss of kinetic energy (temperature), makes it difficult to overcome the activation energy of the reactions yielding intermetallic layer.
- 5- *Alloys should be chosen regarding geometry and purposes.* Casting practices showed that at regions where narrow insert holes or thin casting walls present, high fluidity is required that Si-rich aluminum alloys or Zn-based alloys were suitable. Due to the low fatigue strength but high impact resistance, Zn-based alloys should be preferred at parts subjected to high magnitude, non-cyclic loads. For low-magnitude, cyclic loads, Al-based alloys are much more convenient.
- 6- *Casting Parameters should be adjusted according to geometry and alloys.* For Al-based parts, which have thin sections and complex geometries, pouring temperature should be kept slightly high (for A413.2 not above 670°C) and phase-II velocity should be raised (above 2.2m/s for 21mm sleeve diameter) so that molten metal become able fill whole cavity prior to solidification. But for thick parts, since high pouring temperature magnifies

effects of shrinkage (cause porosity and delamination) and porosity; it should be kept as low as possible (for a413.2, not below 630°C). Furthermore, phase-III time and pressure should be increased for parts having thick wall thickness. Superheat temperature (pouring temperature) must be kept as low as possible (for Zamak5, below 420°C) for Zn-based alloys in whatever case due to some metallurgical issues, thus gate velocity must be raised to accelerate the flow. Since viscosity of Zn-based alloys are low enough even at temperature close to the solidus line, high pouring temperature is not required for thin sections.

REFERENCES

- 1) Andersen B., Die Casting Engineering: A Hydraulic, Thermal and Mechanical Process, Mercel Deker, New York, 2005, p. 21 - 286
- 2) Sully L.J.D., Die Casting, ASM Handbook, Vol. 15: Casting, ASM International, Cleveland, 1998, p.611 - 631
- 3) NADCA, Product Specification Standards for Die Casting, North American Die Casting Association, Wheeling , 2003, p. 6.8
- 4) Davis J.R., Concise Metals Engineering Data Book, ASM International, Cleveland, 2009, p. 113 - 114
- 5) Bar-Meir G. Fundamentals of Die Casting Design, Free Software Foundation, Boston, 2002, p. 33 - 116
- 6) Weiler J.P., Wood J.T., Klassen R.J., Maire E., Berkmortel R., Wang G., Relationship between internal porosity and fracture strength of die cast AM60B alloy, Material Science Engineering A, 2005, 395: p. 315 - 322
- 7) Glass C.M., Greenhunt V.A., Kingman P.W., Porosity and Spall Fracture of Shock Loaded Materials, Defense Technical Information Center, 1976, DTC Online
- 8) Linder J., Arvidsson A., Kron J., The Influence of Porosity on The Fatigue Strength of High-Pressure Die Cast Aluminum, Fatigue Fract. Engng. Mater. Struct., 2006, 29: p.357 - 363
- 9) Mayer H., Papakyriacou M., Zettl B., Stanzl-Tshegg S.E., Influence of Porosity on The Fatigue Limit of Die Cast Magnesium and Aluminum Alloys, 2003, 25: p.245 - 256
- 10) Lumley R.N., Polmear I.J., Groot H., Ferrier J., Thermal Characteristics of Heat Treated Aluminum High-Pressure Die-Casting, Scripta Materialia, 2008, 58: p.1006 - 1009

- 11) Weiler J.P., Wood J.T., Klassen R.J., Maire E., Berkmortel R., Wang G., Porosity Damage Evolution and Fracture in Die-Cast Magnesium Alloy AM60B, 2007, 15-17: p. 455 – 460
- 12) Binney M., Dargusch M., Nave M., McKinney P., Porosity Reduction in High Pressure Die Casting Through The Use of Squeeze Pins, 22nd International Die Casting Congress, Indianapolis, 2003, T03:065
- 13) Türkyılmaz G., Processing and Assessment of Aluminum Ceramic Fiber Reinforced Metal Matrix Composite Parts for Automotive and Defense Applications, M.Sc. Thesis, Graduate School of Natural and Applied Science of Middle East Technical University, 2009, p. 109 – 111
- 14) Scott V.D., O'Rourke J.M., Bushby R.S., Manufacture and Mechanical Properties of Saffil-Reinforced Aluminum Alloys, Composites Part B, 1999, 30: p. 1 – 7
- 15) Kosec B., Kosec G., Sokovic M., Case of Temperature Field and Failure Analysis of Die-Casting Die, Journal of Achievements in Material Science and Manufacturing Engineering, 2007, 20: p. 471 – 474
- 16) Persson A., Hogmark S., Bergström J., Simulation and Evaluation of Thermal Fatigue Cracking of Hot Work Tool Steels, International Journal of Fatigue, 2004, 26: p. 1095-1107
- 17) Haijing M., A Numerical Study of Externally Solidified Products in The Cold Chamber Die Casting Process, Ph.D. Thesis, Graduate School of The Ohio State University, 2004, p. 140 – 144
- 18) Faura F., Lopez J., Hernandez J., On The Optimum Plunger Acceleration Law in The Slow Shot Phase of Pressure Die Casting Machines, International Journal of Machine Tools and Manufacture, 2001, 41: p. 173 - 191
- 19) Han Q., Viswanathan S., Analysis of Mechanism of Die Soldering in Aluminum Die Casting, Metallurgical and Materials Transactions A, 2003, 34: p. 139 – 146
- 20) Zhu H., Guo J., Jia, J., Experimental Study and Theoretical Analysis on Die Soldering, Journal of Material Processing Technology, 2002, 123: p. 229 – 235
- 21) Aksoy H.G., Şenocak E., Wave Propagation in Functionally Graded and Layered Materials, Finite Elements in Analysis and Design, 2009, 45: p. 876 – 891

- 22) Meyers M.A., Dynamic Behavior of Materials, John Wiley & Sons, New Jersey, 1994, p. 23 – 97
- 23) Özşahin E., Alüminyum Levhaların Yüksek Hızlı Çarpma Yükleri Altındaki Davranışları, Ph.D. Thesis, Graduate School of Science Engineering and Technology of Istanbul Technical University, 2008, p. 11 – 42
- 24) Dey S., Børvik T., Teng X., Wierbicki T., Hopperstad O.S., On the Ballistic Resistance of Double-Layered Steel Plates: An Experimental and Numerical Investigation, International Journal of Solids and Structures, 2007, 44: p. 6701 – 6723
- 25) Chen X., Chandra N., The Effect of Heterogeneity on Plane Wave Propagation Through Layered Composites, Composite Science and Technology, 2004, 64: p. 1477 – 1493
- 26) Wasley R.J., Stress Wave Propagation in Solids: An Introduction, Marcel Dekker, New York, 1973, p. 98 – 125
- 27) Gupta Y.M., Ding J.L., Impact Load Spreading in Layered Materials and Structures Concept and Quantitative Measure, International Journal of Impact Engineering, 2002, 27: p. 277 – 291
- 28) Callister W.D., Material Science and Engineering: An Introduction, 7th Ed., John Wiley & Sons, New York, 2007, p. 225 – 227
- 29) Dieter G.E., Mechanical Metallurgy, McGraw-Hill, Singapore, 1988, p. 496 – 499
- 30) Johnson G.R., Cook W.H., A Constitutive Model and Data for Metals Subjected to Large Strains, High Strain Rate and High Temperature, 7th International Symposium on Ballistics, 1983, The Hague, 1983, p. 541 – 547
- 31) Tresca M.H., On Further Applications of The Flow of Solids, Proc. Inst. Mech. Eng., 1878, 30: p. 301 – 345
- 32) Kobayashi S., Yakou T., Control of Intermetallic Compound Layers at Interface Between Steel and Aluminum by Diffusion-Treatment, Material Science and Engineering A, 2002, 338: p. 44 – 53
- 33) Nazari K.A., Shabestari S.G., Effect of Micro Alloying Elements on The Interfacial Reactions Between Molten Aluminum Alloy and Tool Steel. Journal of Alloys and Compounds, 2009, 478: p. 523 – 530

- 34) Rathod M.J., Kutsuna M., Joining of Aluminum Alloy 5052 and Low-Carbon Steel by Laser Roll Welding, *Welding Journal*, 2004, 83: p.16 – 26
- 35) Awan G.H., Hasan F., The Morphology of Coating/Substrate Interface in Hot-Dip-Aluminized Steels, *Material Science and Engineering A*, 2008, 472: p. 157 – 165
- 36) Kattner U.R., Burton B.P., Al-Fe (Aluminum – Iron), *ASM Handbook: Alloy Phase Diagrams*, Vol. 3, ASM International, 1992 p. 294
- 37) Cheng W.J., Wang C.J., Observation of High-Temperature Phase Transformation in the Si-Modified Aluminide Coating on Mild Steel Using EBSD, *Material Characterization*, 2010, 61: p.467 – 473
- 38) Chen Z.W., Fraser D.T., Jahedi M.Z., Structures of Intermetallic Phases Formed During Immersion of H13 Tool Steel in An Al-11Si-3Cu Die Casting Alloy Melt, *Material Science and Engineering A*, 1999, 260: p. 188 – 196
- 39) Mekhrabov A.O., Akdeniz M.V., Effect of Ternary Alloying Elements Addition on Atomic Ordering Characteristics of Fe-Al Intermetallics, *Acta Materialia*, 1998, 47: p. 2067 - 2075
- 40) Akdeniz M.V., Mekhrabov A.O., The Effect of Substitutional Impurities on The Evolution of Fe-Al Diffusion Layer, *Acta Materialia*, 1998, 46: p. 1185 – 1192
- 41) Dezellus O., Digonnet B., Sacerdote-Peronnet M., Bosselet F., Rouby D., Viala J.C., Mechanical Testing of Steel/Aluminum-Silicon Interfaces by Pushout, *International Journal of Adhesion and Adhesives*, 2007, 27: p. 417 – 421
- 42) Wang J., Lee P.D., Hamilton R.W., Li M., Allison J., The Kinetics of Fe-Rich Intermetallic Formation in Aluminum Alloys: In Situ Observation, *Scripta Materialia*, 2009, 60: p. 516 - 519
- 43) Taylor J.A., The Effect of Iron in Al-Si Casting Alloys, 35th Australian Foundry Institute National Conference, 2004, Adelaide, p: 148 – 157
- 44) Molinari A., Pellizzari M., Straffelini G., Pirovano M., Corrosion Behavior of A Surface-Treated AISI H11 Hot Work Tool Steel in Molten Aluminum Alloy, *Surface and Coating Technology*, 2000, 126: p. 31 – 38
- 45) Zhang H., Liui J., Microstructure Characteristics and Mechanical Property of Aluminum Alloy/Stainless Steel Lap Joints Fabricated by MIG Welding-

Brazing Process, Material Science and Engineering A, 2011, 528: p. 6179 – 6185

- 46) Springer H., Kostka A., Dos'Santoz J., Raabe D., Influence of Intermetallic Phases and Kirkendall-Porosity on The Mechanical Properties of Joints Between Steel and Aluminum Alloys, Material Science and Engineering A, 2011, 528: p. 4630 – 4642
- 47) Rohr M.V., Development of Coatings for Steels of Heat Exchangers and Improving Their Corrosion Resistance in Environmental Simulating Smokes of Burning Coal, Ph.D. Thesis, L'Institut National Polytechnique De Toulouse, 2005, p: 80 – 81
- 48) Cheng W.J., Wang C.J., Study of Microstructure and Phase Evolution of Hot-Dipped Aluminide Mild Steel, Applied Surface Science, 2011, 257: p. 4663 – 4668
- 49) Raghavan V., Fe-Zn (Iron-Zinc), Journal of Phase Equilibria, 2003, 24(6): p. 544 – 545
- 50) Culcasi J.D., Sere P.R., Elsner C.I., Di Sarli A.R., Control of The Growth of Zinc-Iron Phases in The Hot-Dip Galvanizing Process, Surface and Coating Technology, 1999, 122: p. 21 – 23
- 51) Feliu S., Barranco V., XPS Study of The Surface Chemistry of Conventional Hot-Dip Galvanised Pure Zn, Galvanneal and Zn-Al Alloy Coatings on Steel, Acta Materialia, 2003, 51: p. 5413 – 5424
- 52) Marder A.R., The Metallurgy of Zinc-Coated Steels, Progress in Material Science, 2000, 45: p. 191 – 271
- 53) Reumont G., Vogt J.B., Iost A., Foct J., The Effect of An Fe-Zn Intermetallic-Containing Coating on The Stress Corrosion Cracking Behavior of A Hot-Dip Galvanized Steel, Surface and Coatings Technology, 2001, 139: p. 265 – 271
- 54) Semoroz A., Durandet T., Rappaz M., EBSD Characterization of Dendrite Growth Directions, Texture and Misorientations in Hot-Dipped Al-Zn-Si Coatings, Acta Materialia, 2001, 49: p. 529 – 541
- 55) Elvins J., Spittle J.A., Worsley D.A., Microstructural Changes in Zinc Aluminium Alloy Galvanising as a Function of Processing Parameters and Their Influence on Corrosion, Corrosion Science, 2005, 49: p. 2740 – 2759

- 56) Manna M., Effect of Fluxing Chemical: An Option for Zn-5wt.%Al Alloy Coating on Wire Surface by Single Hot Dip Process, *Surface & Coatings Technology*, 2011, 205: p. 3716 – 3721
- 57) Adachi Y., Arai M., Transformation of Fe-Al Phase to Fe-Zn Phase on Pure Iron During Galvanizing, *Material Science and Engineering A*, 1998, 254: p. 305 – 310
- 58) ASM Alloy Center:, <http://products.asminternational.org/matdb/index.jsp>, American Society for Metals, last visited on August 2011
- 59) ASTM, Standard Specification for Zinc and Zinc-Aluminum (ZA) Alloy Foundry and Die Castings, American Society for Testing and Materials, 2011, B86-10a: p. 1 – 8
- 60) ASTM, Standard Specification for Aluminum-Alloy Die Castings, American Society for Testing and Materials, 2010, B85/85M: p. 1 – 9
- 61) ASTM, Standard Test Methods for Notched Bar Impact Testing of Metallic Materials, American Society for Testing and Materials, 2007, E23 REV. A: p. 1 – 28
- 62) Becze C.E., Elbestawi M.A., A Chip Formation Based Analytic Force Model for Oblique Cutting, *International Journal of Machine Tools & Manufacture*, 2002, 42: p. 529 – 538
- 63) Khalifa W., Samuel F.H., Gruzleski J.E., Iron Intermetallic Phases in The Al Corner of The Al-Si-Fe System, *Metallurgical and Materials Transactions A*, 2003, 34: p. 807 – 826
- 64) Turkish Armed Forces, Specification of H&K 33 Rifle: A5 103 (H&K 33 E), Ministry of National Defense of The Republic of Turkey, 2001, p. 12

APPENDIX A

CHARPY IMPACT TEST SAMPLE PHOTOS

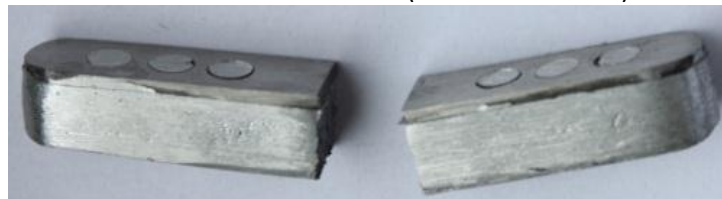
AISI D2 – A413.2 die cast



AISI D2 – Zamak5 die cast



AISI D2 – A413.2 die cast (immersed inserts)



AISI D2 – 413.2 Sand Cast



Figure 48: Broken Charpy impact test samples