

STUDY OF THERMAL AND METALLURGICAL ASPECTS OF FRICTION STIR WELDED JOINTS OF DISSIMILAR METALS

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in Partial Fulfillment of the Requirements for the
Degree of**

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in
Mechanical Engineering**

**by
SHWETANSHU GAURAV
(Roll No. 2K22/PHDME/01)**

Under the Supervision of

**DR. MOHAMMAD ZUNAID
(Supervisor)**

**PROF. RADHEY SHYAM MISHRA
(Jt. Supervisor)**



**Department of Mechanical Engineering
DELHI TECHNOLOGICAL UNIVERSITY
(Formerly Delhi College of Engineering)
Shahbad Daulatpur, Bawana Road, Delhi-110042, India.**

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SHWETANSHU GAURAV



DELHI TECHNOLOGICAL UNIVERSITY

(Formerly Delhi College of Engineering)

Shahbad Daulatpur, Main Bawana Road, Delhi-42

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(2K22/PHDME/01)

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Certified that **Shwetanshu Gaurav** (2K22/PHDME/01) has carried out their research work presented in this thesis entitled “**Study of Thermal and Metallurgical Aspects of Friction Stir Welded Joints of Dissimilar Metals**” for the award of **Doctor of Philosophy** from the Department of Mechanical Engineering, Delhi Technological University, Delhi, under our supervision. The thesis embodies the results of the original work, and the studies were carried out by the student himself. The contents of the thesis do not form the basis for the award of any other degree to the candidate or anyone else from this or any other University/Institution.

Dr. Mohammad Zunaid
(Associate Professor)
Department of Mechanical Engineering
Delhi Technological University,
Delhi-110042

Dr. Radhey Shyam Mishra
(Professor)
Department of Mechanical Engineering
Delhi Technological University,
Delhi-110042

Date:



DELHI TECHNOLOGICAL UNIVERSITY

(Formerly Delhi College of Engineering)

Shahbad Daulatpur, Main Bawana Road, Delhi-42

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Supervisor(s)

Dr. Mohammad Zunaid (Associate Professor)

Dr. Radhey Shyam Mishra (Professor)

Department: Mechanical Engineering

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ABSTRACT

Multi-material Dissimilar friction stir welding (FSW) of magnesium AZ91D and aluminium AA6061-T6 poses significant challenges due to metallurgical incompatibility. Despite their potential in lightweight applications, comprehensive optimization of process parameters in SiC-reinforced AZ91D and AA6061-T6 FSW joints remains limited. The present investigation comprehensively examines the combined effects of process parameters and nanoparticle reinforcement on the mechanical and metallurgical behavior of dissimilar AZ91D/AA6061-T6 joints. SiC particles were introduced into pre-configured grooves, and experiments were carried out by varying volume fraction (Vf) (5%,10%,15%) of SiC, tool rotational speed (TRS) (600,700,800 rpm), and traverse speed (TS) (20,30,40 mm/min).

This study investigates the influence of Tool Rotational Speed (TRS), Traverse Speed (TS), and SiC Volume fraction (Vf) on the mechanical and metallurgical properties of these joints using Central Composite Design (CCD) of Response Surface Methodology (RSM) to mathematically model FSW input parameters with key mechanical responses such as ultimate tensile strength (UTS), strain, and microhardness. The analysis of variance (ANOVA) approach identified critical parameters. It validated the model's prediction with a 95% confidence interval (CI), yielding R^2 values of 0.9973 for Ultimate Tensile Strength (UTS), 0.9319 for % strain, and 0.9951 for microhardness, indicating excellent predictive capability. Optimization revealed that the optimal microhardness, strain, and UTS in the stir zone (SZ) were 88.44 HV0.1, 6.45%, and 114.56 MPa, respectively.

Microstructural analysis revealed that the SiC nanoparticles significantly refined the grains in the stir zone (SZ). This refinement was primarily due to the pinning effect of nano-sized SiC particles, which restricted grain growth and facilitated dynamic recrystallization (DRX) during FSW, ultimately leading to a substantial reduction in grain size. Increasing SiC Vf from 5% to 15% enhanced friction stir-welded (FSWed) joints and exhibited improved mechanical characteristics. Among systematically designed experiments, the condition with 700 rpm, TRS 30 mm/min, and Volume fraction 15% achieved the highest UTS (114.98 MPa) and strain (6.67%) along with microhardness of 88.9 HV0.1, closely matching the model's predicted optimum and

confirming its accuracy and robustness across the design space. In contrast, the lowest UTS of 71.46 MPa, 32.63% joint efficiency, and 67.87 HV0.1 microhardness was recorded at 600 rpm, 40 mm/min, and 5% Vf SiC, mainly due to inadequate mixing and particle clustering, confirming the strengthening role of SiC. These findings illustrate the potential of SiC nanoparticle reinforcement in enhancing the mechanical properties of FSWed joints and demonstrate that the weld quality in AZ91D and AA6061-T6 joints strongly depends on parameter selection to ensure uniform particle distribution and defect-free welds.

Complementing the experimental investigation, a three-dimensional transient thermal simulation was performed using ANSYS, considering heat generation from both frictional sliding and plastic deformation at the tool-workpiece interface. The model was validated using thermocouple measurements placed on both the advancing and retreating sides of the joint, showing a deviation of less than 5% between the simulated and experimental temperature histories. The simulation results indicated that heat input increased with tool rotational speed and decreased with traverse speed, directly correlating with the observed material flow, grain refinement, and IMC morphology. The optimal condition (700 rpm, 30 mm/min) yielded a balanced heat input, which is sufficient for full plasticization without melting, thereby explaining the superior mechanical performance and defect-free weld morphology.

Overall, this integrated experimental-numerical-statistical framework establishes a comprehensive process-structure-property relationship for dissimilar AZ91D/AA6061-T6 FSW joints reinforced with SiC nanoparticles. The study not only validates the predictive capability of CCD-RSM for optimizing FSW parameters but also highlights the vital role of thermal input and nanoparticle dispersion in achieving defect-free, AZ91D/AA6061-T6 dissimilar joints. The outcomes provide a reliable foundation for extending nanoparticle-assisted FSW toward advanced lightweight hybrid structures in automotive and aerospace applications.

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List of Abbreviations

Abbreviation	Meaning
FSW	Friction Stir Welding
TSD	Tool shoulder Diameter
PD	Pin Diameter
PP	Pin Profile
TRS	Tool Rotational Speed
TS	Traverse Speed
TTA	Tool Tilt Angle
UTS	Ultimate Tensile Strength
PL	Pin Length
TO	Tool Offset
ST	Shoulder Type
FSWed	Friction Stir Welded
PLD	Plunge Depth
IMC	Intermetallic Compounds
Vf	Volume Fraction

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Chapter 1 : INTRODUCTION

1.1 Introduction

The growing needs of present-day society, along with the increasing global population, have placed more pressure on the world's available energy sources than ever. The increase in living standards has led to a growing reliance on personal cars and air travel, which are no longer luxuries but are now an integral part of life. Researchers and stakeholders in the industry have, therefore, strived to come up with materials that can reduce the component mass without compromising structural integrity and performance. In this regard, metals with low density and preferable mechanical properties, including magnesium and aluminium, have been utilized as a potential alternative to traditional steel. There has been significant development in high-performance, lightweight materials over the past few decades, driven by the rapidly growing transportation, defense, and energy sectors. A decrease in the weight of vehicles, aircraft, and defence platforms directly leads to increased fuel efficiency, higher functional load capacities, and reduced dangerous emissions. Consequently, an intense movement toward aluminium (Al) and magnesium (Mg) in structural applications that were formerly dominated by the denser ferrous constituents has taken place in the industrial scene [1,2]. Especially, aluminium alloys are still a subject of scholarly debate due to their desirable features, including low density, high strength-to-weight ratio, corrosion resistance, favourable thermal behaviour, and reasonable weldability. They are available in both wrought and cast forms and are traditionally divided into two main groups: non-heat-treatable and heat-treatable, thus allowing engineers to choose the material that best fits a particular task. The main series of aluminum alloy involves: The principal series of aluminum alloy entails:

- 1xxx series: This is essentially pure aluminum (more than 99 percent) with high corrosion resistance and conductivity, but low strength.
- 2xxx series (Al-Cu): High strength, primarily aerospace, although it has poor corrosion characteristics.
- 3xxx series (Al-Mn) - Moderate strength that is good in corrosion resistance and is used a lot in beverage cans and roofing.

- 5xxx series (Al-Mg) - Superior resistance to corrosion, and also weldability, and it is employed in marine and transport applications.
- 6xxx series (Al-Mg-Si) - Combination of strength, corrosion, and weldability. These are alloys that are highly versatile and find a wide range of applications in the aerospace, automotive, and structural sectors.
- 7xxx series (Al-Zn-Mg-Cu) - Highest strength, mainly used in aerospace and defense, but does tend to stress corrosion crack.

Of these, AA6061, which belongs to the 6xxx series, is of particular importance. It is a hardened alloy of magnesium and silicon as the major alloying elements, which is precipitated. AA6061 in the T6 temper condition provides an outstanding combination of tensile strength, corrosion resistance, toughness, and machinability; therefore, it is a popular material in aerospace frameworks, automotive parts, pressure vessels, and defense components. Its excellent welding capability, combined with its abundance, adds to its value in the industry. It is the 6xxx series of alloys, among them AA6061-T6, that are of particular importance in aerospace and the world of cars. Their capability to be precipitation hardened (through proper heat treatment) adds value in terms of strength, dimensional stability, and service performance. Instead, magnesium alloys are recognized as the lightest structural metals used in the modern industrial world. The primary series includes:

- Mg-Al-based alloys (AZ series, Al + Zn) - This is the most common type, which has a good strength-to-weight ratio, though it has poor creep at high temperatures.
- Mg-Mn alloys (AM series) - Mg-Mn alloy is good in corrosion resistance, but has lower strength and is used in marine and general-purpose applications.
- Mg-Zn alloys (ZK series) - Strong, aerospace and automobile usable, but less resistant to corrosion.
- Alloys containing rare-earths (e.g., WE series, AE series, ZE series) - Superior creep strength and thermal stability, finding application in aerospace and defense.
- Wrought Mg alloys (e.g., AZ31, ZK60) - More ductile and formable, commonly used in sheets, plates, and extrusions.

Its application is common due to its high castability, easy machinability, and positive vibration-damping characteristics, especially in the AZ series, such as AZ91D. These properties make magnesium alloys particularly suitable for automotive parts, such as

transmission cases, steering systems, and electronic housings, where lightweight design and vibration reduction are the most important considerations. However, magnesium alloys have some drawbacks, such as lower corrosion resistance or lower average strength compared to aluminum alloys. Therefore, despite the tendency of aluminum alloys to dominate the mechanical and corrosion performance comparison, magnesium alloys cannot be compared with them in the sphere where the main goal is to minimize the weight.

The introduction of hybrid joints, which combine aluminum and magnesium, has proven to be a good measure to embrace the capabilities of these two materials. Even though the total mass of such composite structures can be significantly decreased, their structural integrity and corrosion protection can be ensured simultaneously. For example, the addition of magnesium elements to aluminum-heavy designs can reduce component mass by up to thirty percent, resulting in significant savings of fuel and costs in the aerospace and automotive industries [3,4]. Effective joining technologies, however, are the key to the successful implementation of Al-Mg dissimilar structures.

The conventional types of fusion welding, such as gas tungsten arc welding, laser beam welding, and resistance spot welding, are particularly challenging to use for joining dissimilar aluminum (Al) and magnesium (Mg). To begin with, the melting temperature and thermal conductivity vary, leading to an uneven distribution of heat that promotes hot cracking and porosity. Secondly, high-temperature treatment decreases the wettability of Al and Mg by oxidation on the surface, resulting in the formation of inclusions. Above all, fusion welding promotes the development of brittle intermetallic compounds (IMCs) at the junction. These IMCs are tough and brittle, and as they overgrow, they may have a detrimental effect on the strength of the joints [5,6,7]. This renders the Al-Mg dissimilar joints that were made by the traditional fusion welding process to be mechanically weak and untrustworthy in the industry.

In comparison, solid-state joining processes employ temperatures that are lower than the melting points of the constituent materials, making them suitable for lightweight alloys. Among the examples is Friction Stir Welding (FSW), which was first developed by The Welding Institute (TWI) in the United Kingdom in 1991, and subsequently was an inventive technology in metal joining [8]. This is achieved through a special rotating mechanism that features a pin and a shoulder, which press the interface of two parallel

plates as it slides along the joint line. The combined action of frictional heating and severe plastic deformation leads to localized softening, allowing for the severe mixing of materials and dynamic recrystallization, which results in a defect-free bond in the solid state [5, 9, 10, 11]. As the process is not associated with melting and solidification, issues such as hot tears, solidification shrinkage, and porosity are successfully prevented. In addition, the relatively low optimum temperature suppresses the overgrowth of IMCs, which is particularly beneficial when welding dissimilar aluminum-magnesium [12,13,14].

Friction stir welding (FSW) provides a distinct advantage over traditional welding methods based on fusion. The solid-state character of the process reduces the number of defects during welding and facilitates the creation of fine and uniform microstructures, resulting in highly consistent and reliable joints. The heat production and material flow can be customized to provide good bonding even between alloys with different deformation behaviour, such as aluminum and magnesium, since critical parameters, such as tool rotational speed (TRS), traverse speed (TS), and tool tilt angle (TTA), can be tightly regulated. Additionally, FSW is an energy-saving and environmentally harmless process that requires no shielding gases or filler material. This makes it of special interest in the joining of alloys that are normally considered difficult or impossible to join through fusion techniques.

The new inventions in this area have introduced the concept of nanoparticle reinforcement in the stir zone, which improves the strength of the welded joints. Various techniques have been introduced to add reinforcements such as silicon carbide (SiC), aluminum oxide (Al_2O_3), carbon nanotubes (CNTs), and many more to the welded region. These techniques include surface grooves, pre-placed powders, or coatings. These nano-additives assist in strengthening the mechanism of welded joints in many ways, including Zener pinning-induced grain refinement, Orowan looping, dislocation inhibition, and effective load transfer. Among all these, SiC nanoparticles are considered the preferable option as they are hard, thermally stable, and can be utilized with aluminum and magnesium matrices. The tensile strength, hardness, and wear resistance have been improved by the addition of intermetallic compounds in the stir zone, and the brittle nature of the compounds has been alleviated. Hence, nanoparticles-

assisted FSW is widely recognized as a promising method for producing high-performance and advanced dissimilar joints.

Despite the reinforcement strategies that have improved mechanical and metallurgical performance, the thermal and flow behavior of FSW should be properly comprehended. Experimental thermocouple data are helpful but must be limited in quantity and scope. Computational modeling methods, such as the finite element method (FEM) and computational fluid dynamics (CFD), have been widely employed to model FSW. The models predict trends of heat production, temperature distribution, and material flow in the weld region that cannot be easily identified during experiments. The research on CFD methods, in particular, demonstrates asymmetric heat transfer between the moving and receding sides, as well as vortex-like movement in the stir zone. Experimental studies can be supplemented with these simulations to establish thermal cycles with various process parameters.

In general, welding of aluminium and magnesium alloys reinforced with SiC nanoparticles and friction-stirred will directly respond to the industrial requirement of lightweight, strong, and economical structures. This technique not only fills the gaps of conventional fusion welding but also enhances the thermal and metallurgical characteristics of dissimilar joints, thereby the optimization of structures in the most vital applications. These trends form the basis of the present research, which includes a systematic study of thermal, metallurgical, and mechanical properties of AZ91D-AA6061-T6 dissimilar FSW joints with the aid of experimental validation and numerical simulation with ANSYS 22 R1.

1.2 Friction Stir Welding

Friction Stir Welding (FSW) is a solid-state welding method that was initially invented at The Welding Institute (TWI), UK, in 1991. The method has achieved widespread acceptance in key industries, including aerospace, automotive, shipbuilding, railway, and defense production, since its inception. The increased focus on lightweight and fuel-efficient structures has led to a rise in their use, particularly with alloys such as aluminum and magnesium, which are challenging to weld with traditional fusion methods. The process does not require heating and solidification of the material, unlike conventional welding methods. FSW works at a relatively low temperature of the base

metals. The principle of the working of FSW is fairly simple. This kind of non-consumable tool is specially designed with a pin and a shoulder, used in rotating and setting in the joint line of two plates in contact. The frictional heating between the shoulder and the workpiece, combined with the extreme plastic deformation of the material around the rotating pin, hardens the material surrounding the rotating pin without melting it off. The softened material is stirred, carried, and condensed behind the tool as it passes through the joint to form a solid bond that is defect-free in its solid state.

FSW has been used on a large scale to join aluminium alloys, magnesium alloys, copper, titanium, steels, and dissimilar combinations. It is particularly suitable for lightweight alloys such as AA6061-T6 (Al-Mg-Si series) and AZ91D (Mg-Al-Zn series), which are challenging to weld using fusion processes due to the formation of brittle intermetallic compounds (IMCs) or other defects. FWS has become a good and dependable method of joining due to the growing need for weight reduction in aerospace, marine, and automotive structures. Figure 1.1 shows a schematic of the FSW process and nomenclature

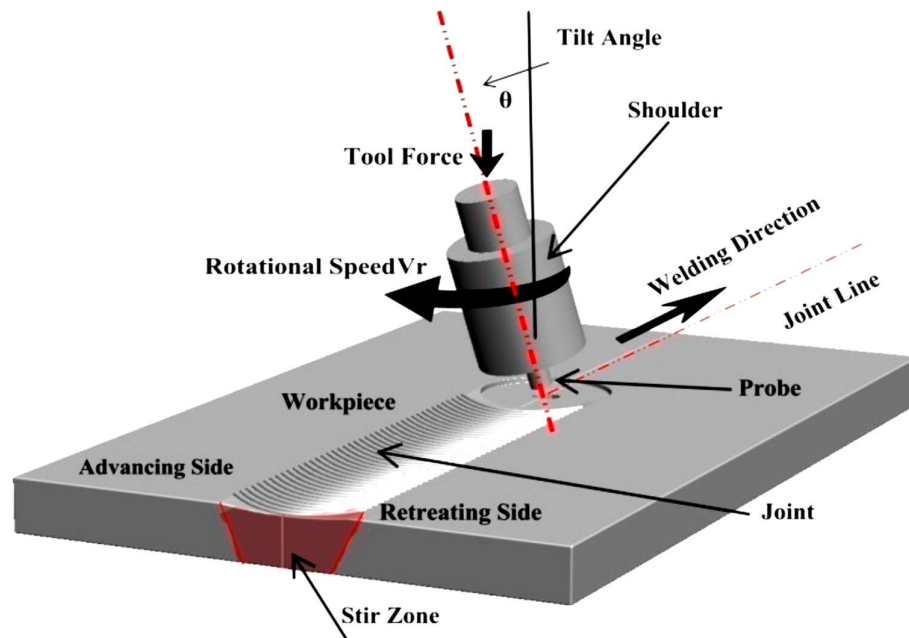


Figure 1.1: Schematic of the FSW Process and nomenclature [15]

1.2.1 Nomenclature of FSW

To understand the process, it is essential to define the major components and terms:

- **Tool Shoulder:** The part of the tool that remains in touch with the workpiece's top surface is called the shoulder. Almost 80%-92% of the heat generated during friction stir welding comes from it, making it the primary heat source. The plastic deformation of the material beneath the shoulder and the friction at the tool-workpiece interface are the main sources of this heat.
- **Pin (or Probe):** The pin is the projecting part of the tool that enters the workpiece, allowing the material to flow across the thickness of the joint. Its contribution to heat generation is relatively small, usually between 10% to 20%. This heat is mainly generated by vigorous stirring and plastic deformation of the pin surface, with a small proportion originating from frictional contact.
- **Advancing Side (AS):** This is the part of the workpiece where the rotating tool's direction matches the welding direction.
- **Retreating Side (RS):** Located opposite to the advancing side, here the tool rotation moves in the opposite direction to the welding direction. Due to this relative velocity difference, the welding process is inherently asymmetric, with more heat being generated and plastic flow on the AS compared to the RS.
- **Leading Edge (LE) & Trailing Edge (TE):** The tool portion that first encounters the material during welding is termed the leading edge, whereas the part that follows is known as the trailing edge.
- **Plunge Depth:** This is a parameter that is used to determine the vertical distance between the shoulder of the tool and the workpiece top surface once fully penetrated. An increased depth of plunge increases the contact area, thus increasing heat generation in the workpiece. Normal plunge depths range from 0.05mm to 0.5mm.
- **Tool Tilt Angle:** This is the tool's angle that is tilted slightly retrograde to the direction of the welding to provide a greater compressive force to the joint. This change reduces the chances of defects and flash. The tilt angle is usually between 0° to 3°.

- **Axial Force (Plunge Force):** This downward force ensures that there is proper contact of the shoulder and workpiece and also helps in consolidating the weld nugget.

1.2.2 Stages of FSW

FSW is performed in four stages (Figure 1.2), each of which is crucial for heat generation, material flow, and joint formation.

(1) Plunging:

This is initiated by the rotating tool being forced into the joint line of the metal plate, similar to drilling. The diving process will continue until the shoulder of the tool comes into contact with the surface of the material. By this stage, friction at the tool-workpiece interface is the primary source of heat generation, which is supported by plastic deformation of the material. Heating due to friction is the prevailing parameter, with almost 70 to 80 percent indicating the total heat input [7]. The material softens and begins to flow upward, then under the tool shoulder, and is compacted.

(2) Dwelling:

Once it has plunged, the tool can be left to rotate in the same position without progressing. This phase is referred to as dwelling, and it helps increase the temperature of the area sufficiently to soften the material, making the stirring process more efficient. It is during this period that the plastic deformation beneath the tool shoulder increases. Dwelling is not a mandatory process, and its time is usually varied in relation to the preferred heat input and materials flow conditions.

(3) Welding:

This step involves rotating the tool over the joint line. The molten material is pushed from the front of the tool to the back, thus creating a solidified weld.

(4) Retracting Stage

This is performed at the conclusion of the weld, whereby the tool is pulled back, leaving a small keyhole. The success of the next stage relies on the effectiveness of the previous stages.

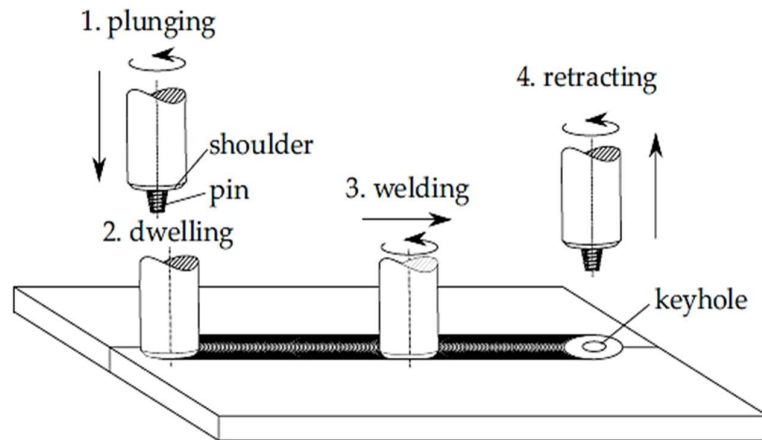


Figure 1.2: Stages of FSW [16]

1.2.3 FSW Tool Design

Tool design plays a crucial role in determining heat generation, material flow, and the quality of the welds in FSW. The tool could not be consumed, so its performance is related to the tool material, the shoulder geometry, and the pin profile. The design of tools must be well-optimized to ensure the absence of defects in welds, as well as to refine the microstructure of the joint and its mechanical characteristics. A detailed description of the tool material, shoulder geometry, and pin geometry is given in the following subsections.

1.2.3.1 Tool Material

The material used for producing the FSW tool is of utmost importance because it directly relates to the durability, stability, and performance of the tool in the welding process. In contrast to consumable electrodes used in fusion welding, the FSW tool is non-consumable and must sustain successive repeated cycles of extreme mechanical and thermal loading without subjecting itself to quick wear or distortion. The material selection for the tool is influenced by various factors, including the nature of the base metal to be joined, the presence of reinforcements or abrasives, and plate thickness.

➤ Tool Material Requirements

The material of the FSW tool should meet the following requirements:

- **High hardness:** The tool must maintain its hardness and strength at elevated temperatures generated during welding. In the case of aluminum and magnesium alloys, welding temperatures can reach 450°C or 550°C. In contrast, for high-melting alloys such as steels or titanium, the welding temperature can be even higher, exceeding 1000°C.
- **High wear resistance:** The tool is abrasively worn by oxide layers (especially in aluminium and magnesium) and, in reinforced systems, by hard ceramic particles such as SiC or Al₂O₃.
- **Good fracture toughness:** The tool should not fracture due to torsional and axial stress.
- **Thermal fatigue resistance:** The tool material must be able to withstand repeated heating and cooling under thermal cycling because, during welding, the material undergoes repeated heating and cooling.
- **Economical and machinable tool material:** The tool material must be cost-effective and easy to machine into the required geometry as per industrial needs.

1.2.3.2 Common Tool Materials for FSW

The material used in the tool selection is primarily determined by the melting point of the workpiece's alloys.

➤ **For low-melting alloys (Aluminium and Magnesium)**

- **Tool Steels (e.g., H13, AISI 4140):** Widely used because they provide strength, hardness, and toughness up to 600°C. They are economical and easily machinable, making them suitable for Al-Mg welding.
- **High-Speed Steels (HSS, e.g., M2, M42):** Offer improved hot hardness and wear resistance compared to standard tool steels, extending tool life in high-speed or longer welding runs.
- **Tungsten Carbide (WC-Co composites):** Provides higher wear resistance than steels, particularly effective when abrasive particles are in the stir zone.

➤ **For intermediate and high-melting alloys (Steels, Titanium, Nickel-based alloys)**

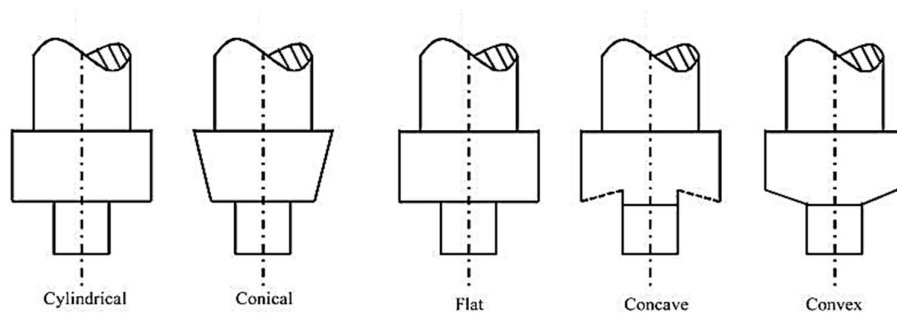
- Tungsten-Rhenium Alloys: Withstand temperatures up to 1200°C, providing excellent strength at high heat input.
- Cermets (Ceramic-Metal Composites): Provide toughness and hardness at elevated temperatures.
- Polycrystalline Cubic Boron Nitride (PCBN): Currently one of the best materials for FSW of steels and titanium alloys. PCBN exhibits extremely high hardness and excellent wear resistance, but is very costly, which limits its industrial use.

1.2.3.3 Tool Shoulder Geometry

The FSW tool with the shoulder is the primary component that generates frictional heat and confines the plasticized material under the tool. Its geometry dictates its effectiveness in producing heat, controlling the material, and the quality of the surface finish that can be obtained. There are several types of shoulder shapes (Figure 1.3) employed, each with its benefits and drawbacks.

- A **cylindrical shoulder** is one of the simplest designs, where it is composed of a flat circle that is in direct contact with the workpiece. It creates an even distribution of heat and is comparatively simple to produce. Sometimes, it lacks the ability to provide sufficient confinement of the plasticized material. This may lead to an increased probability of flash formation and surface defects, especially when dealing with softer alloys.
- The **conical shoulder**, in its turn, is inclined downwards, and this increases forging pressure and causes the softened material to be forced into the weld zone. This design is likely to increase the thickening of the plates, although in some cases, it may result in a thinning of the weld crown if the parameters are not optimized.
- Another simple setup that has a flat surface that touches the plate is a **flat shoulder**. It offers a consistent heat supply and reliable material circulation. Nevertheless, it cannot be used in demanding applications because it is susceptible to excessive flash due to the absence of confinement.

- The underside of the **concave shoulder** is shallowly shaped in a dish-shaped form, forming a pocket to restrict the material. This is a major characteristic that enhances the containment of plasticized metal, minimizing flash propensity and increasing surface smoothness. Downward forging action is also facilitated by the concave shape, ensuring that nuggets are uniform in composition. For this reason, it is one of the most widely used shoulder geometries in industrial FSW applications.
- Conversely, the profile of the **convex shoulder** is bulged outwards. The geometry enhances pressure during forging, thereby improving material consolidation. However, it can also lead to thinning of the weld surface. The convex shoulders are less effective in controlling flash compared to the concave ones and are therefore used less in reinforced or dissimilar welds. However, they may be beneficial in cases where additional action, such as forging, is required.



SHOULDER GEOMETRY

Figure 1.3: FSW tool shoulder geometry [17]

1.2.3.4 Tool Pin Geometry

The pin or probe, which emerges from the lower portion of the shoulder, accesses the joint line and is crucial for stirring and mixing the material, as well as for transporting it to the weld area. The design of the wire has a direct impact on the nuggets zone size, refinements, and removal of defects in the form of voids or tunnels. The pin geometry (Figure 1.4) may be chosen according to the plate's thickness, the nature of the materials being welded, and the reinforcements.

- A **cylindrical pin** is a straight-profile pin with a fixed diameter. It generates comparatively constant and predictable flow material and is therefore suitable for simple joints and thin plates. Nevertheless, due to the comparatively weak turbulence that it creates in the plasticised material, cylindrical pins tend to create tunnel defects in thicker plates or where increased mixing is necessary.
- The **triangular pin** has a polygonal cross-section that will introduce a pulse stirring action. This geometry disrupts material flow more aggressively, enhancing turbulence and promoting better mixing of the alloys. It works especially well with dissimilar joints, where the interfacial oxides need to be broken up by vigorous stirring to promote metallurgical bonding. The disadvantage is that the sharp edges cause more stress to the tool, which decreases the tool life, particularly in the presence of abrasive particles.
- A **conical pin** becomes smaller at the tip in a uniform manner. This design will enable the downward movement of material and ensure easy consolidation of material at the base of the weld. The conical pin helps to avoid the development of voids due to its low likelihood of not penetrating fully. Although it does not offer the level of turbulence found in polygonal designs, it strikes a good balance between flow control and weld quality.
- The **tapered pin** looks like the conical type, but in most cases has facets or progressive alteration of the diameter instead of having a smooth cone. This geometry will increase the vertical velocity of the plasticized material, which decreases the likelihood of flaws, such as voids or non-bonding areas. It is also easier to insert and withdraw tools. It is due to these reasons that tapered pins are commonly employed in similar and dissimilar alloy welding, and they are particularly well-suited where nanoparticles, such as SiC, are incorporated, as they facilitate the attainment of smoother reinforcement dispersion in the stir zone.

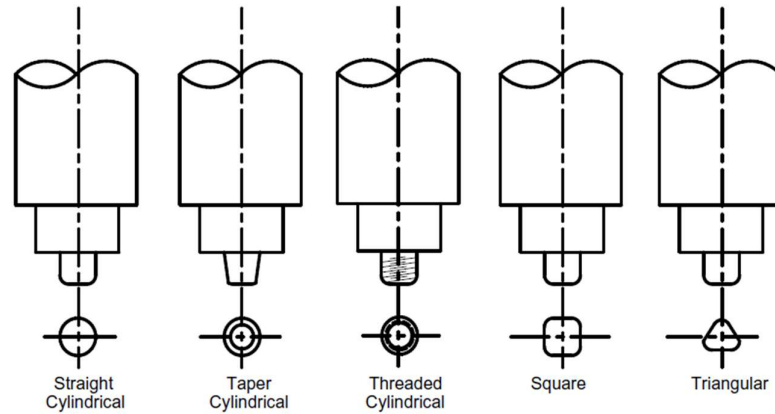


Figure 1.4: FSW tool pin geometry [18]

The shoulder and pin geometries of the tool play a significant role in determining heat generation, material flow, and weld integrity in the friction stir welding of aluminum and magnesium alloys. Concave or scrolled shoulders may be used, which increases the material's confinement to fit it more securely. However, the cylindrical shoulders are very popular since they achieve a consistent distribution of heat and maintain constant process conditions, especially when the tilt angle and plunge depth are optimized properly. Equally, although pin shapes like threaded or polygonal will elevate turbulence and mixing, the pins that create perfect welds are cylindrical, provided that the process parameters are well managed. The cylindrical shape of aluminium gives a consistent heat input, and the surface is clean and smooth; on the other hand, magnesium offers sufficient plasticization without excessive thinning. A cylindrical shoulder and pin are therefore justifiable and valid as a viable and lucrative option in welding dissimilar Al-Mg systems.

1.2.4 Process Parameters

Process parameters have a significant influence on the performance of friction-stir-welded joints. In comparison to the fusion welding method, which focuses on regulating the melting and solidification process, FSW relies on the thermal-mechanical environment created by the rotating tool. These conditions govern material plasticization, mixing, and development of microstructure; thus, they dictate the strength and integrity of the weld. The most significant ones are the tool rotational

speed, the traverse speed, the tool tilt angle, the plunge depth, and the tool geometry, which interrelate to control heat input, material flow, and joint consolidation.

1.2.4.1 Tool Rotational Speed

The most critical parameter determining the amount of heat generated and the degree of stirring of the material is the rotational speed. The higher the rotational speed, the greater the magnitude of frictional heating and plasticity provided to the workpiece, thereby increasing flow and mixing in the stir zone. However, very high speeds can cause overheating, grain coarsening, and the formation of fairly thick intermetallic compounds (IMCs) in dissimilar joints, especially in Al-Mg systems, which in turn can reduce the integrity of the joint. On the other hand, low speeds can be too slow to produce enough heat, resulting in insufficient mixing and ultimately leading to defects, voids, or incomplete bonding in the tunnel. In this respect, selecting an optimal range of speeds that strikes a balance between the amount of heat generated and the quality of the joints is a critical design choice.

1.2.4.2 Traverse Speed

The traverse (or welding) speed controls the amount of dwell time the tool is at a given locus and thus, the amount of heat per unit length. Decreased traverse speed increases the time the tool spends inside the material, thereby increasing the amount of heat deposited into the material and enlarging the size of the heat-affected zone (HAZ), which may reduce hardness in aluminum alloys. Faster traverse rates can truncate thermal cycles and inhibit the formation of deleterious microstructures; however, they can also cause the material flow to be hindered, resulting in incomplete bonding or the formation of voids. In this way, the interaction between rotational and traverse speeds is the most important: the former determines the heating of the joints, and the latter regulates the spread of this heat along the joint line.

1.2.4.3 Tool Tilt Angle

The tilt angle of the tool is what helps in the effective forging of plasticized material behind the tool. Even with a slight positive tilt, the trailing edge of the shoulder presses down on the softened material, consolidating the joint and eliminating the voids.

Excessive tilt will cause poor consolidation, and excess tilt will produce flash and decrease the thickness of the joint. The choice of the right tilt angle enhances joint soundness, surface finish, and mechanical performance.

1.2.4.4 Plunge Depth

The depth of the plunge is determined by the depth to which the shoulder enters the work surface. A sufficient plunge depth ensures optimal frictional contact, sufficient heat generation, and stirring. Common defects attributed to inadequate plunge include the development of tunnels and partial fusion, while excessive penetration may lead to excessive flash, plate thinning, and tool wear. In dissimilar systems, especially that of aluminum-magnesium welds, plunge depth is supposed to be varied to reduce the excess of interfacial reaction and total mixing.

1.2.4.5 Tool Geometry

The design of the tool has a great influence on the distribution pattern of heat and material flow. Flat, concave, and scroll geometries influence surface friction and confinement of the plasticized material. Likewise, the pin profile, which can be cylindrical, tapered, threaded, or polygonal, controls the stirring intensity within the nugget zone. The pin geometry can be complex and plays a useful role in mixing and refining grains, though it can also increase the rate of tool wear, particularly in the presence of hard reinforcements such as SiC particles. A cylindrical pin is the preferred choice of aluminum-magnesium joint because it offers balanced stirring and reduced intermetallic growth.

1.2.5 Interaction of Parameters

These parameters are mutually dependent, which defines the quality of the joints. An example of this is that too high a rotational speed with a high traverse speed may cause the mixture to be incompletely mixed, and the reverse is also true with moderate rotational speeds, resulting in overheating due to too low traverse speeds. Likewise, the performance of the tilt angle and the plunge depth is based on the interaction of tool geometry and material combination. Consequently, the need to optimize all parameters simultaneously, rather than in isolation, is imperative.

1.2.6 Microstructural Development

Finally, the choice of parameters regulates the microstructural development in the weld area. Well-optimized conditions result in fine, dynamically recrystallized grains, evenly dispersed reinforcement particles, and limited IMC at interfaces. These microstructural characteristics enhance the weld's tensile strength, hardness, and ductility. On the contrary, poor parameter selection results in coarse grains, particle agglomeration, or brittle IMC layers, which reduce joint efficiency. In summary, the effects of process parameters in FSW are central to achieving defect-free, high-strength welds. Their optimization is particularly critical for dissimilar systems such as Al-Mg, where the risk of IMC formation is high, and for nanoparticle-reinforced welds, where adequate stirring is required to achieve homogeneous dispersion without degrading the matrix.

1.3 Thermal Aspects in FSW of Dissimilar Aluminium / Magnesium

The thermal properties of FSW are decisive factors in determining the microstructural and mechanical properties of the joints. Although FSW is a solid-state process that does not cause the bulk melting of the base metals, substantial heat generation is required to plasticize the material surrounding the rotating tool, thereby facilitating metallurgical bonding across the interface. In the case of dissimilar systems such as aluminium and magnesium, the differences in melting point, thermal conductivity, specific heat capacity, and diffusivity create a highly asymmetric temperature field. This asymmetry directly affects the extent of material flow, the formation of intermetallic compounds (IMCs), and the development of residual stresses. A clear understanding of heat input, temperature distribution, and their measurement and modelling is therefore essential for controlling weld quality in Al/Mg systems.

1.3.1 Heat Input in FSW

The sources of heat in friction stir welding are primarily due to two causes: the frictional sliding between the rotating tool and the workpiece, and the severe plastic deformation of the material in the stir zone. Of these, the shoulder of the tool, however, produces

most of the heat, and the pin, not only local heating, but also stirring. Process parameters are one of the strongest factors influencing the amount of heat input, with the rotational speed of the tool, traverse speed, axial force applied to it, and tilt angle being the most significant. As the rotational speed increases, the relative sliding speed between the tool and the workpiece surface also rises, resulting in an increase in the quantity of frictional heat. This facilitates successful softening and plasticization of both aluminium and magnesium, enhancing intermixing at the interface. Nevertheless, higher rotational speeds can also cause overheating, and in Al-Mg alloys, this can promote diffusion-based IMC formation, particularly in brittle alloys like $\text{Al}_{12}\text{Mg}_{17}$ and Al_3Mg_2 . On the other hand, low rotational speeds are not always sufficient to remove all the heat needed to consolidate the entire material, and thus, this leads to the formation of voids or poor bonding.

The traverse speed determines how long a tool is engaged with a workpiece at a particular point. Slower traverse speeds result in a greater accumulation of heat over a longer period, leading to higher temperatures and a larger area of heat-affected zones. The higher the traverse rate, the less time is required per unit length to generate heat, which may inhibit IMC growth, but at the cost of insufficient plastic flow. Axial force and tool tilt angle also have their effects on the heat input, variation in the effective contact area, and pressure below the shoulder. In brief, balancing the heat input necessitates a precise proportion between sufficient plasticization and hindrance of the excessive interfacial reactions.

1.3.2 Temperature Distribution

Friction stir welding exhibits non-uniform heat distribution and is highly localized. Maximum temperatures in the case of aluminum dissimilar joints with magnesium tend to range between 400°C and 500 °C. These values are approximately 60-80 percent of the melting points of the alloys, which is sufficiently high to facilitate recrystallization and diffusion but still not high enough to cause bulk melting, thereby retaining the solid-state nature of the process.

Among the most noticeable characteristics of the thermal profile are the seeming asymmetry between the advancing side (AS) and the retreating side (RS). This side has a higher shear rate and, consequently, greater heat accumulation due to the rotating

action of the tool in addition to its forward motion on the AS. The two motions are opposite to one another on the RS, resulting in a decrease in deformation intensity and lower temperatures compared to one another. The result of this imbalance is a shape-deformed nugget zone, having the greater and deeper penetration of thermal energy on the AS.

This unevenness is also enforced by the various thermal conductivities of the base metals. The aluminum alloys, such as AA6061-T6, transfer heat almost twice as fast as magnesium alloys like AZ91D. Consequently, the aluminum side is more likely to conduct heat over a larger area, creating a larger and cooler heat-affected zone (HAZ). When compared to Magnesium, the latter tends to retain most of the heat produced near the weld line due to its low heat transfer. Such local heat generation increases the likelihood of undesired interfacial reactions and the formation of thick intermetallic layers. Additionally, the specific heat capacity of the two alloys differs, and therefore, they absorb and release energy at dissimilar rates. Aluminum also cools comparatively quickly as the tool passes through, thus promoting the development of finer grain structures. In contrast, magnesium cools more slowly and, in many cases, forms coarser microstructures.

1.3.3 Thermal Measurement by Experimental Techniques

Thermal cycles and their effect on microstructural development and joint properties are of high importance in the case of FSW, and proper measurements of weld temperatures are regarded as an essential part of the research. Various techniques have been used, with the most common being thermocouples and infrared thermography. Sophisticated tool-based sensors are being used in current research.

Traditional thermocouples remain the most straightforward and efficient way of measuring temperatures in the workpiece. They may be incorporated at varying depths/distances to the weld line of the aluminum and magnesium plates to monitor the varying temperature of the joint. Their data is accurate and real-time, but the method is intrusive: when creating thermocouples by drilling holes, some level of disturbance is caused by the technique to the thermal field of the joint.

Infrared thermography is an alternative method. It is a non-contact method. Infrared cameras can map surface temperatures at high speed, and the visualization of asymmetry between the moving and the stationary sides can be made. The technique has the advantage of analyzing the temporary behavior along the weld line. Its primary drawback, however, is simply that it only captures the surface data, and the quality of the result is extremely reliant on the emissivity calibration, which varies between magnesium and aluminum alloys.

Researchers have also experimented with sensors being built into the tool itself in recent years. Thermocouples or wireless probes embedded in the shoulder give the information on the situation at the very tool-workpiece interface. This method is technically more complicated and less frequently used, but it provides good information about the mechanism of the actual heat generation below the tool.

Readings of measurements taken under these procedures show that the aluminum side of the weld attains peak temperatures, whereas the magnesium side commonly attains a bit lower temperatures. The fact that the two alloys have quite dissimilar cooling rates also plays a significant role, as it regulates grain refinement and the development of intermetallic compounds (IMCs). This experimental evidence not only records the observed microstructural features but can also serve as benchmarks for validating thermal models and simulations.

1.3.4 Numerical Simulation of Thermal Fields

Numerical simulation is now an essential tool for explaining the thermo-mechanical processes that occur during the friction stir welding (FSW). Because experimental measurements of temperature, strain, and flow of materials are often limited to observations of the surface or discrete points, computational methods provide an option to resolve the three-dimensional, transient nature of the process with greater detail. Specifically, computational fluid dynamics (CFD)-based models have been widely used to study heat transfer, plastic flow, and mixing behavior in the stir zone, particularly in dissimilar material systems, e.g., aluminum and magnesium.

In CFD modeling, the viscous liquid that is under the rotating tool is considered a non-Newtonian, viscoelastic material. The governing equations typically include the

conservation of mass, momentum, and energy, accompanied by appropriate constitutive relations that describe the temperature and strain-rate-dependent flow of material. The heat generation is represented as the product of frictional dissipation at the tool/workpiece contact and primarily plastic deformation in the bulk material. To provide realistic thermal predictions, boundary conditions, including convective heat losses to the surrounding air, conduction through the backing plate, and tool-workpiece contact friction, are included.

1.3.4.1 COMSOL Multiphysics

COMSOL Multiphysics has found extensive application in FSW studies because it can couple heat transfer, fluid flow, and structural mechanics within a single environment. After applying its CFD and Heat Transfer modules, the researchers can model the stir zone as a viscoplastic flow field and solve the governing equations in transient conditions. The capability to add a temperature-dependent material property can be incorporated into COMSOL, thereby achieving a more accurate representation of dissimilar systems, including aluminum and magnesium. The software is also used to enable the study of parameters, where changes in the rotational speed of the tools, traverse speed, or tool geometry can be systematically studied to determine their effect on temperature distributions and nugget formation.

1.3.4.2 ABAQUS

Abaqus, primarily a finite element analysis (FEA) package, has been utilized in CFD-like thermal-mechanical analyses of FSW. Abaqus, with its coupled thermo-mechanical effects and user-defined subroutines (UMAT/UEL), enables the definition of models for viscoplastic material behavior and heat generation at the tool-workpiece interface. This permit modeling the multifaceted interaction between tool geometry, process parameters, and heat flow. Along with temperature fields, Abaqus is often used in conjunction with the investigation of residual stresses, distortion, or structural performance of the welded joint. Despite being more computationally intensive than simplified analytical models, Abaqus provides a more detailed perspective on the thermal cycles, as well as the ensuing stress evolution, in the welding process.

1.3.4.3 ANSYS Fluent

ANSYS Fluent has found numerous applications in the numerical analysis of friction stir welding due to its ability to handle coupled problems of heat transfer and flow with sufficient numerical stability. There are models where the plastically deformed material in the stir zone can be described as a non-Newtonian fluid, and the resulting conservation equations are solved to replicate the material's behavior around the rotating tool. It is also possible to visualize flow structures, such as vortex-like recirculation, and material exchange between the advancing and retreating sides using the software. By characterizing the tool as a moving boundary, its rotation and translational movement can be modeled, resulting in increased realism in predicting temperature fields, cooling rates, and thermal gradients within the weld region. Fluent is also compatible with structural solvers in the ANSYS Workbench platform and can be used to analyze the behavior of post-welded structures, such as stress distribution and potential distortion.

Although COMSOL, Abaqus, and ANSYS are all used in modeling the thermal and flow of FSW, the strengths of each package differ. COMSOL is appreciated in the academic community due to its capability to couple various physical domains, as well as its comparatively easy configuration of parametric analysis. Abaqus is usually preferred when the interests are not limited to heat transfer, but also the stress analysis, residual stresses, or even the deformation of the welded plates. ANSYS Fluent, however, is especially useful in the detailed fluid-like flow of the softened material, and in analyzing the convective heat transport within the stir zone, so that it can be useful in analyzing the asymmetry between the advancing and the retreating side.

Despite these abilities, FSW numerical modeling remains a challenging task. Credible results of a simulation require a meaningful description of the constitutive behavior of the sustaining material at high strain rates and high temperatures, as well as an appropriate modeling of friction and contact conditions at the tool-workpiece interface. It is due to these problems that experimental validation, such as the application of thermocouples or infrared thermography, is necessary. Simulations, however, when cross-checked with experimental findings, can be used as useful guidance for process design. They may be employed in estimating peak weld temperatures, predicting the size of the heat-affected zone, investigating the propensity for the development of

intermetallic compounds in dissimilar joints, and even predicting the likelihood of defect formation. Consequently, CFD-based simulations have also become a key supplement to experimental studies, considerably reducing the need for trial-and-error experiments to be conducted repeatedly.

1.4 Nanoparticle Reinforcement in Aluminum and Magnesium Welds

The addition of nanoparticles to metallic joints has gained significant academic attention as one of the methods of enhancing the microstructural stability and mechanical properties of friction stir welds. This method is especially appealing for dissimilar systems, such as aluminum and magnesium, which exhibit significant differences in thermal conductivity and melting temperatures, and are likely to form brittle intermetallic compounds at the interface. Nanoparticles have a significant effect on grain refinement, dislocation dynamics, thermal stability, and interfacial reactions when added to the stir zone.

One of the most conspicuous benefits of nanoparticle reinforcement is considered to be grain structure refinement. The extreme plastic deformation and high temperatures produced during the welding process promote recrystallization dynamics, whereas the nanoparticles serve as locations for grain boundary movement. This Zener pinning effect inhibits excessive grain coarsening, thus creating a fine equiaxed nugget zone. The hardening of grains, therefore, not only increases the strength following the Hall-Petch relation but also increases ductility and toughness. This type of grain-structure control is of the utmost importance during aluminum-magnesium welds, in which sharp thermal gradients across the joint may cause a high degree of microstructural heterogeneity.

In addition to grain-size control, the nanoparticles also contribute to the strength as they interact with dislocations. Dispersed nanoparticles in the matrix hinder the high dislocation density caused by the intensive deformation of the rotating tool. The strengthening mechanism raises the flow stress of the alloy, resulting in improved tensile and yield strengths. Nanoparticles can also serve as a reinforcing agent in place of conventional strengthening phases in precipitation-hardened aluminum alloys, where

these phases are likely to dissolve during the welding process, thereby helping to maintain or even improve the hardness of the weld region.

The thermal field is also affected by nanoparticles during welding. They can either induce or retard thermal transport by being either intrinsically conductive or insulated. In both events, they alter the distribution of heat in the stir zone, although the sharp gradient between the advancing and retreating sides is minimized. It is a stabilization of the thermal field that slows down grain coarsening and helps maintain the refined microstructure developed during dynamic recrystallization.

The role of nanoparticles in regulating the formation of intermetallic compounds (IMCs) is perhaps the most significant role of nanoparticles in dissimilar aluminum-magnesium welds. Reaction phases such as $\text{Al}_{12}\text{Mg}_{17}$ and Al_3Mg_2 are easily formed at high temperatures and are brittle in nature; therefore, they normally fail the joint prematurely. The presence of nanoparticles at the interface hinders atomic diffusion and the continuity and thickness of IMC layers. This interfacial region is typically thinner and more discontinuous, which minimizes the number of crack initiation sites and enhances the overall ductility of the joint.

Experimental results have consistently shown that welds strengthened with nanoparticles are stronger than those without reinforcement, due to the homogeneous distribution of hardness and improved fracture behavior. The tensile characteristics are generally greater, and the hardness distributions along the weld are more consistent. Fracture surfaces exhibit a ductile appearance, characterized by fine dimples, as opposed to brittle cleavage. It has also been reported to have improved fatigue resistance, as refined grains and dispersion strengthening reduce the chances of crack formation at an early stage.

Despite these benefits, several challenges remain. Nanoparticles tend to agglomerate due to the high surface energy, and these clusters may serve as defects if they are not dispersed properly. To achieve uniform distribution, it is necessary to exercise precise control over welding parameters. The rotational speed should be sufficiently large to encourage stirring, but not too large, as this can add excessive heat to the welding process, which can occur too easily and cause accelerated IMC formation. On the same note, the speed of traverse affects mixing time and thermal exposure. Some of the most

popular methods of attaching nanoparticles are filling surface grooves, applying coated joint edges, or placing powders into shallow holes along the weld line. Although effective, each method has practical shortcomings, requiring the tool to perform a stirring action to homogenize the particles. The other problem is associated with the wear of the tools: hard ceramic particles, including carbides and oxides, may wear the pin and shoulder during welding, influencing tool geometries and process stability over longer runs.

1.4.1 SiC Reinforcement in Dissimilar AZ91D and AA6061-T6

Silicon carbide (SiC) is one of the various nanoparticles studied, and it has been found to be very efficient in enhancing the performance of the aluminum magnesium welds. SiC is hard, chemically inert, and a good thermal conductor, and can be used with either aluminum or magnesium. SiC contributes to the refinement of grains, strengthening, and stability of the interfaces simultaneously in FSW.

SiC nanoparticles are useful in the nugget zone, where they facilitate the growth of grain boundaries during dynamic recrystallization, leading to the formation of ultrafine, equiaxed grains. They are also useful in enhancing dispersion strengthening due to their distribution that hinders the flow of dislocations, leading to quantifiable increments in hardness and tensile strength. Silicon carbide has an exceptionally high thermal conductivity, which contributes to a uniform temperature distribution across the weld area. This helps reduce local overheating, especially on the magnesium side, where thermal conductivity is relatively low, and also helps mitigate softening in the neighboring aluminum area. This causes the even distribution of heat, which results in a more balanced microstructural development across the dissimilar interface.

Nanoparticles of SiC at the Al/Mg interface act as diffusion barriers and thus hinder the progression of intermetallic compound formation. They are thin enough to interrupt the continuity of IMC layers, which increases ductility and inhibits premature fracture of the interface by preventing brittle cleavage. Additionally, these particles serve as mechanical bridges, facilitating the transfer of loads between the magnesium and aluminum matrices.

Empirical studies of friction-stir-welded joints reinforced with SiC nanoparticles on an AZ91D-AA6061-T6 assembly exhibit significantly better mechanical behavior. Fractographic observations indicate ductile fracture with fine dimples, but unreinforced ones fracture mostly by cleavage at the Al-Mg interface.

The intrinsic hardness of SiC nanoparticles is also a significant contributor to wear resistance, which is crucial in applications that involve mechanical abrasion of welded joints. SiC is also chemically stable, which provides it with increased corrosion resistance by preventing galvanic corrosion between aluminum and magnesium alloys.

SiC reinforcement requires precise control of the process parameters. When the Tool Rotational Speed (TRS) is too high, the tendency to form nanoparticle aggregates is enhanced, whereas a lower TRS can lead to poor dispersion. Thus, the Traverse speed (TS) must be optimized to achieve a balance between effective mixing and thermal input. However, the results show that the combined efficiency of the SiC-reinforced AZ91D and AA6061-T6 welds is higher than that of the joints made without such reinforcement, making SiC one of the most promising reinforcements in dissimilar Al-Mg welding.

The nanoparticles play a crucial role in enhancing the structural and mechanical properties of friction-stir-welded Al-Mg joints. Their capability to refine the grain size, enhance the matrix strength, and alleviate brittle intermetallic formation provides a significant enhancement compared to unreinforced welds. Despite the number of nanoparticle systems that have been investigated, such as TiO₂, Al₂O₃, TiC, CNTs, and graphene, SiC stands out because of its synergistic thermal conductivity, hardness, and chemical compatibility with aluminum and magnesium.

The excellence of SiC nanoparticles is reflected in their ability to strengthen, harden, and improve fracture toughness in dissimilar joints between AZ91D and AA6061-T6, as well as to reduce interfacial brittle behavior. SiC is therefore still an ideal reinforcement method to enhance the reliability and performance of Al/Mg dissimilar welds.

1.5 Metallurgical Aspects of FSW

Friction welding (FSW) is a technique for making joints that combines frictional heating, plastic deformation, and diffusion in the solid state. In contrast to the fusion welding processes, which are controlled by the melting and solidification, FSW forms bonds without reaching the melting point of the base materials. This dissimilarity has far-reaching metallurgical implications, especially in dissimilar alloy systems such as aluminum and magnesium, where differences in thermal conductivity, melting points, and chemical affinities make it difficult to form a weld. The microstructure of an FSW joint is not homogeneous; rather, it is separated into distinct areas that represent the local pattern of thermal and mechanical history. These areas, in conjunction with the material circulation in the area of the tool, the processes of the grain refinement, of forming intermetallic compounds, and the influence of reinforcements, altogether determine the metallurgical behaviour of the process.

1.5.1 Weld Zones and Their Microstructural Features

FSW creates a welded joint that can be divided into three regions, each possessing its own distinct microstructural response: the stir zone (SZ), the thermo-mechanically affected zone (TMAZ), and the heat-affected zone (HAZ), as shown in Figure 1.5.

The stir zone, also known as the nugget region, is immediately under the tool pin and shoulder. The maximum temperatures and the most significant plastic deformation occur in this location. The initial grain of the base substance is constantly chipped off and replaced by a new one, which is likewise equiaxed and fine-textured, the outcome of dynamic recrystallization under high shear and heat. The stir zone is even more complex in the case of dissimilar aluminum-magnesium welds, as alloy material in both directions is mixed. In most instances, an onion-ring-like appearance may be observed, which is a layer of successive material deposition with each tool revolution. On the microscopic scale, this area is frequently interspersed with some regions of mixed composition and sometimes with some of the products of these reactions, e.g., Al/Mg intermetallic compounds.

The thermo-mechanically affected zone is located beyond the stir zone. In this case, the material is exposed to high temperature and plastic strain, although not to the level

required to allow recrystallization. As a result, the grains get elongated and distorted due to the shearing effect of the rotating tool. The TMAZ is typically asymmetric: the flow and deformation of the material are more pronounced on the advancing side, where the tool rotation and the direction of traverse are in combination, and on the retreating side, where the two movements are opposite. This dissimilarity is particularly significant in dissimilar welds, where the contrasting flow behaviors of aluminum and magnesium exacerbate the differences.

The outer part of the weld is known as the heat-affected zone. It is not mechanically deformed and undergoes thermal cycles as the heat is conducted through the surrounding areas. In precipitation-strengthened aluminum alloys, such as AA6061-T6, these thermal exposures may either dissolve or coarsen the strengthening precipitates, leading to local softening. The grain size of HAZ is basically the same as the grain size of the base material, although the loss of hardening of the precipitate can render this area the weakest of the weld when tensile loads occur.

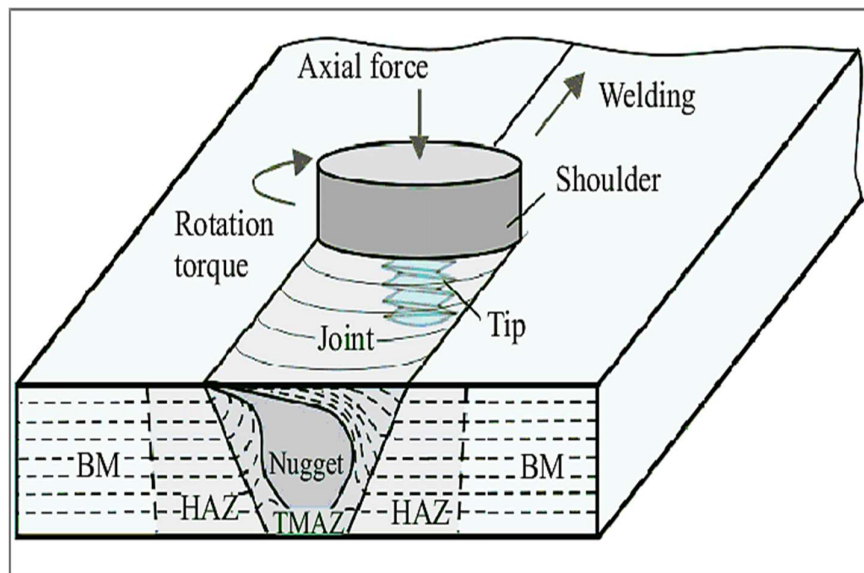


Figure 1.5: FSW weld zones [19]

1.5.2 Material Flow and Its Metallurgical Significance

The manner in which the material circulates through the tool on FSW can have a profound impact on the end microstructure and quality of the weld. During the process

of engaging the rotating tool into the joint line, the shoulder generates frictional heat that warms the surrounding material. At the same time, the pin stirs and moves the plasticized material to form a complex three-dimensional flow pattern. The softened material is extrusion-displaced, i.e., pushed about the tool and solidified behind the tool to create the weld.

Material flow in dissimilar aluminum-magnesium joints is not homogeneous, as both alloys have different plasticity and thermal characteristics. Aluminum is more ductile and has better thermal conductivity, which means that it flows more easily compared to magnesium, which softens more easily but does not deform as easily. This is due to an imbalance in the flow field, which can prove beneficial or harmful, depending on the placement of the alloy. When aluminum is deposited on the advancing side, such as in enhanced plasticity, it has the opportunity to improve the material's mixing. Conversely, an undesirable design can lead to the formation of voids, poor consolidation, or inadequate bonding.

The distribution of reinforcements, such as nanoparticles, is also determined by the flow pattern. A steady and constant flow enables the particles to be spread uniformly in the stir zone, where they can accomplish their functions in refining the grains and strengthening the matrix. On the other hand, unsteady or turbulent flow favors concentration of nanoparticles and forms local heterogeneities that deter mechanical performance. Flow efficiency is thus a major predictor of both metallurgical quality and reinforcement behavior.

Material flow also determines the morphology of intermetallic compounds. In the case of intermixing aluminum and magnesium on an acceptable level, the diffusion rate at the interface is fast, and discrete IMC particles are produced within the stir zone. Conversely, poor flow conditions can result in continuous IMC layers that remain intact in the long run, and such layers are typically more detrimental to mechanical properties. Therefore, flow not only concentrates the weld, but also regulates the extent and form of the metallurgical reactions.

1.5.3 Mechanism of Grain Refinement

Grain refinement of the stir zone is one of the most unique metallurgical products of FSW. This refinement is carried out through dynamic recrystallization, which is guided

by the combined effect of severe plastic deformation and high temperatures attained during welding, and occurs below the solidus temperature (but above the eutectic temperature). It is accomplished by the replacement of coarse grains of base metal by fine grains called equiaxed grains, which are usually an order of magnitude smaller than those of the parent material.

There are two major recrystallization processes, which rely on the alloy system. In aluminum alloys, the predominant process is continuous dynamic recrystallization, in which sub-grains developed by recovery eventually become grains bound together by high-angle boundaries. Magnesium alloys, in contrast, tend to discontinuous dynamic recrystallization, in which new grains nucleate along grain boundaries and expand into the deformed matrix. Both of these mechanisms can be applied to dissimilar Al/Mg welds, resulting in a mixed grain structure within the nugget.

The refined grains tend to increase the hardness and strength, which is in accordance with the Hall-Petch relation. Refinement is highly dependent on the welding parameters, however. The low traverse speed is combined with a high rotational speed to provide a larger heat input, which may lead to grain coarsening even after the initial refinement. Conversely, a low amount of heat causes insufficient deformation, which reduces the recrystallization. The best welding, therefore, must be a balance between temperature and strain to achieve maximum grain refinement without encouraging excessive grain growth.

1.5.4 Intermetallic Compounds

The formation of intermetallic compounds depends on the type of metal and the quantity of that metal in the mixture. Intermetallic compounds are inevitable and essential to the performance of joints in dissimilar aluminum manganese welding. The solubility of aluminum and magnesium is limited by each other, and at high temperatures, they interact to form two distinct phases, $\beta(\text{Al}_3\text{Mg}_2)$ and $\gamma(\text{Al}_{12}\text{Mg}_{17})$. These are hard and brittle, and their presence at the interface may diminish ductility and fracture toughness, as it can act as a favorable location for crack initiation.

These IMCs are morphological and depend on the input of heat and the flow of the material during the welding process. In a high-heat environment, IMCs develop in the

form of thick, continuous layers, which seriously compromise the weld. At lower levels of heat input, the reaction is limited and produces thin, discontinuous films or dispersed IMC particles, which are less damaging. Although some (though very low) concentration of IMCs may occasionally lead to strengthening, there are virtually no other cases where continuous layers at the interface do not worsen mechanical performance.

Therefore, in dissimilar FSW, one of the main metallurgical goals is to control the IMC formation. It is achieved, first of all, by optimizing welding parameters, which influence the maximum temperature and diffusion rate. Gradually, this is further enhanced by the addition of reinforcements, which can impede diffusion and prevent continuity of the layer.

1.5.5 Effect of Nanoparticle Reinforcement

The current trends in friction stir welding have investigated the intentional incorporation of ceramic nanoparticles, including silicon carbide (SiC), into the weld site to improve microstructural stability as well as mechanical activity. The metallurgical activity of these particles is two-fold: they have a direct effect on the grain refinement during recrystallization, and an indirect impact on the diffusion mechanism at the dissimilar interface.

The nucleation of new grains occurs at the nanoparticles during stirring. They lower the critical energy barrier to the recrystallization process, thus favouring the growth of uniformly fine and equiaxed grains in the nugget zone. The particles also inhibit the migration of grain boundaries, in addition to their role in supplying nucleation points. This process, more often referred to as boundary pinning, stabilizes the refined structure and inhibits the coarsening of the structure on thermal cycles. The resulting microstructure is thus finer and more stable as compared to unreinforced welds, which is directly translated to high hardness and better strength.

Nanoparticles have more advantages than grain size control. The distribution of elements along the interface in dissimilar aluminum-magnesium welds may hinder the kinetics of intermetallic compound formation. Given that the development of Al_3Mg_2 and $\text{Al}_{12}\text{Mg}_{17}$ phases is among the primary reasons for joint embrittlement, the capacity

of SiC particles to restrain their thickness is of significant importance. In addition, when properly dispersed, the stiff particles also distribute a portion of the load applied during mechanical testing, which helps reinforce the material through the transfer of loads.

These benefits, however, can only be achieved when the reinforcement is evenly spread throughout the stir zone. Surface energy causes the clustering of nanoparticles, which can result in inhomogeneous regions, leading to stress concentrations that counterbalance the desired improvements. In this way, the introduction of nanoparticles, whether through surface grooves, powder compaction, or coating, should be optimized to achieve uniform dispersion.

1.5.6 Challenges

Despite the few metallurgical benefits of friction stir welding over fusion-based welding techniques, multiple challenges remain that are especially difficult to overcome, particularly when welding nanoparticle-reinforced aluminum and magnesium alloys. The compounding of brittle intermetallic compounds is the initial and most enduring challenge. Despite the optimised parameters, diffusion across the Al/Mg interface is bound to result in reaction phases, and their existence cannot be avoided. The engineering issue is to restrict their extension to thin or discontinuous morphologies that will minimize their adverse effect on joint functioning.

The second issue is associated with material flow. This flow asymmetry between the advancing and withdrawing sides is further emphasized in dissimilar systems, where the responses of aluminum and magnesium to heat and deformation differ. Non-uniform flow may lead to their voids, partial consolidation, or non-homogeneous dispersion of particles. These local defects may determine the general failure mode of the joint.

Further nanoparticle reinforcement leads to increased complexities. Their agglomerate propensity, incompatibility with the thermal expansion of the metallic matrix, and the opportunity for the formation of secondary phases all introduce potential disadvantages. Such drawbacks may, in certain situations, nullify the same advantages that nanoparticles are meant to bring. To overcome these problems, optimization of process parameters is necessary in FSW of dissimilar AZ91D and AA6061-T6 to obtain a nanoparticle-reinforced defect-free weld.

Overall, it can be concluded that the metallurgical characteristics of FSW combine both favorable and unfavorable processes, including the recrystallization-driven reduction of grain size, and intermetallic growth, coarsening of precipitates, and clustering of reinforcements. Although this process already offers benefits compared to fusion welding, its full potential can only be realized through further optimization of processing conditions, enhanced reinforcement strategies, and a deeper understanding of the flow-structure-property relationships.

1.6 Applications of FSW of Aluminum and Magnesium

Aluminum and magnesium alloys have become highly desirable materials in a structural sense due to the growing demand for lightweight and high-performance materials in modern industries. The aluminum alloys, including AA6061-T6, are widely used in aerospace, automotive, and marine applications due to their high strength-to-weight ratio, corrosion resistance, and ease of welding. Alternatively, magnesium alloys like AZ91D are among the lightest structural metals, offering the advantage of high weight reduction while maintaining high castability. The capability to co-exist these two alloys in hybrid alloys may result in components that incorporate the benefits of both, that is, aluminum with excellent strength and fatigue resistance, and magnesium with significant weight reduction.

Friction stir welding (FSW) is a solid-state welding method that is particularly well-suited for welding aluminum-magnesium alloys, which are otherwise challenging to fuse using traditional fusion methods due to the formation of brittle intermetallic compounds. The dissimilar FSW of AZ91D and AA6061-T6 enables the creation of hybrid structures that meet the high demand for weight-sensitive products. Likewise, in the auto industry, such joints may be used in the cross-car beam, engine cradle, and transmission cases, where mass minimization leads to fuel economy and lower emissions. The Al-Mg hybridization can be beneficial in interior panelling, structural ribs, and seat frame supports in aerospace, as the material provides the necessary strength while also reducing weight. There are also comparable opportunities in the marine industry, specifically in deck housings and bulkhead stiffeners, as well as in the consumer electronics sector, including lightweight casings and enclosures.

Therefore, FSW has a wide application potential in AZ91D -AA6061-T6 systems, which are industrially applicable. This process enables the production of advanced, lightweight structures for next-generation engineering solutions by creating defect-free joints that enhance mechanical performance.

Chapter 2 : LITERATURE REVIEW

2.1 Literature Review

The study by Mofid et al. [1] was conducted on the impact of water cooling on submerged friction stir welding (SFSW) on the microstructure, temperature distribution, and intermetallic formation of dissimilar FSW between AA5083 aluminum alloy and AZ31C magnesium alloy. They also applied the same welding parameters to both standard FSW in air and SFSW in water at TRS 300 rpm, TS 50 mm/min, and TTA 3°, using H13 tool steel with a 20 mm concave shoulder and a 7 mm threaded pin of 2.8 mm length. The aluminum plate was placed on the advancing side, and the magnesium plate was placed on the retreating side, which helps in the movement of the metal and reduces the formation of defects. The findings reveal that the highest temperature, 403°C (air) was lowered to 378°C (underwater) through water-induced cooling. This led to the inhibition of the formation of brittle IMCs (Al_3Mg_2 and $\text{Al}_{12}\text{Mg}_{17}$) in the submerged state to a significant extent. Microstructural observation revealed that the air-welded joint exhibited high intermixing, with an intermetallic layer of up to 100 μm in thickness, which led to interface cracking. The submerged welds, on the contrary, had a smooth interface and an intermetallic layer of approximately 3 microns, which is thin. The grains were improved by the submerged condition. The grain size of Mg in air was 14 μm , whereas it decreased to 0.8 μm in water due to dynamic recrystallization. The hardness mapping revealed that the air welds were hard (150-250 HV) due to the presence of IMCs. The hardness of SFSW samples was more uniform (around 80-100 HV) across the joint. The XRD results confirmed the inhibition of Al-Mg intermetallic phases in the submerged welds. Therefore, the research concluded that cooling with water decreases the amount of heat input, restricts the development of IMC, refines grains, and avoids cracking to give a sounder and more structurally sound joint in Al-Mg dissimilar FSW.

Fu et al. [2] studied the interfacial microstructure and mechanical behaviour of FSWed joints of AZ31 magnesium and A6061 aluminum with a relatively small TRS of 300 rpm and TS of 100mm/min. The investigation revealed the formation of a thin intermetallic compound (IMC) layer, with a thickness ranging from approximately 700 nm to 2 μm , comprising Al_3Mg_2 and $\text{Al}_{12}\text{Mg}_{17}$ sublayers, which were created through solid-state diffusion. The layer of IMC was continuous in the upper and mid-thickness

zones of the weld, while discontinuities and micro-cracks were found near the lower interface. The maximum tensile strength of 158 MPa was measured in the upper region, with comparable values in the middle zone. Such findings support the importance of using geometry and continuity of the IMC layer in controlling interfacial bonding and fracture behavior in Mg/Al friction-stir-welded joints.

B. Fu et al. [3] conducted a systematic study on dissimilar friction-stir welding of 6061-T6 aluminum and AZ31B magnesium alloys, examining the effects of material positioning, tool offset, and welding parameters on joint integrity. Defect-free joints were obtained with magnesium located on the advancing side of the joint, and a positive tool offset of 0.3 mm was used, mainly at intermediate rotational speeds of 600-800 rpm and low traverse speeds of 30-60 mm/min. Microstructural examination of the nugget zone showed that there were alternating Al-Mg lamellar features with eutectic phases, and the formation of $\text{Al}_{12}\text{Mg}_{17}$ and Al_3Mg_2 intermetallic phases. Under the optimized conditions, a maximum tensile strength of 175 MPa, approximately 70% of the base strength of the magnesium alloy, was achieved. Torque-based heat input analysis revealed that a moderate heat input resulted in the intermixing of the material, yielding significantly improved mechanical properties of the welds.

Abbasi et al. [4] investigated the effect of particle reinforcement with silicon carbide (SiC) on the microstructure and mechanical properties of AZ31 magnesium alloy friction stir welds. SiC particles with an average size of 55 μm were added to a pre-machined groove before welding to form a composite stir zone. The welding was performed over a broad spectrum of parameters (600-1000 rpm and 25-175 mm/min). The optimum condition, found to be 800 rpm and 75 mm/min, gave a homogeneously distributed particle network and a significantly refined grain structure. Compared with conventional FSW joints, the SiC-reinforced joints showed improved strength, ductility, and hardness. The improvement was attributed to dislocation pinning by SiC particles, a decrease in interparticle spacing, and grain refinement.

Baghdadi et al. [5] investigated the effects of the formation of IMCs on the fracture behavior of dissimilar AZ31 and AA6061-T6 FSWed joints. Welding was carried out at a TRS of 500-600 rpm and a TS of 20-40 mm/min with varying tool offsets (-1, 0, and +1 mm) using an H13 steel tool with a threaded cylindrical pin and no tilt angle. A thin interfacial IMC layer (smaller than 3.5 μm), which consists mainly of $\text{Al}_{12}\text{Mg}_{17}$ and

Al_3Mg_2 phases, was identified in all joints. The thickness of this layer had a substantial effect on the joint strength and fracture location. The maximum tensile strength of 180 MPa, corresponding to approximately 80% joint efficiency, was achieved at 600 rpm and 40 mm/min with a zero-tool offset. Under these conditions, a fracture occurred away from the Mg/Al interface towards the magnesium-sided thermo-mechanically-affected zone, thus indicating the need to control the heat input to limit the IMC growth and improve the reliability of the joint.

Mouria et al. [6] investigated the dissimilar friction stir welding of AZ91D magnesium alloy and AA2024 aluminum alloy, focusing on the effect of TRS on microstructure evolution and mechanical properties. An H13 steel tool, equipped with a square pin (5.5 mm) and a 20 mm shoulder, was used with a 2° tilt angle. The TS was maintained at 35 mm/min, and the TRS ranged from 500 to 800 rpm. The optimum condition was found at 700 rpm, and the balanced heat input favored the occurrence of dynamic recrystallization and creation of fine intercalated lamellar structure. This condition gave a maximum tensile strength of 147.26 MP and a joint efficiency of 68%. Excessive or insufficient heat input at higher or lower speeds resulted in thicker IMC layers with compositions of $\text{Al}_{12}\text{Mg}_{17}$ and Al_3Mg_2 , resulting in brittle fracture. The stir zone hardness achieved was approximately 92 HV, which is higher than that of AZ91D but lower than that of AA2024. Fractographic analysis showed mainly brittle fracture features, although some local ductile dimples were also observed.

Liang et al. [7] investigated the combined effect of tool rotational speed and tool offset on the microstructure and mechanical behavior of dissimilar friction stir welded joints between AA6061 aluminum and AZ31 magnesium alloys. Welding was performed at a rotational speed of 600-1000 rpm, constant traverse speed of 100 mm min⁻¹, and the tool offset of +1 mm (toward aluminum), 0 mm, and -1 mm (toward magnesium) with an H13 tool of 12 mm shoulder, 4 mm pin diameter, 2.5 mm pin length, and 3° tilt angle. Three different areas were found: the bottom interface (BI) zone of limited material mixing; the banded structure (BS) zone of intermetallic compounds; and the stir-affected aluminum region zone. Defect-free, seamless joints with high tensile strength (163 MPa) and considerable elongation (approximately 6%) were achieved at TRS 900-1000 rpm with zero tool offset. Under these conditions, ductile necking fracture was observed on the retreating side, indicating an optimal balance of strength and ductility.

On the other hand, tool offsets towards either base material favored brittle fracture due to the lack of bonding or excessive IMC formation. To increase the rotational speed, the volume fraction of IMC bands and refined grains decreased due to dynamic recrystallization, resulting in a maximum stir zone hardness of 74 HV.

Abdollahzadeh et al. [8] have formed an in-situ nanocomposite during the friction stir welding of dissimilar 6061-T6 aluminum and AZ31 magnesium alloys by using SiC nanoparticles on the magnesium side. A novel two-step procedure has been adopted, where the initial processing involves friction-stir welding (FSW) with a pinless tool for the dispersion of nanoparticles, followed by conventional FSW with a frustum pin tool. Welding parameters used varied from 550 to 650 rpm and 25 to 35 mm/min. The optimized condition (650 rpm and 35 mm/min) yielded an even distribution of nanoparticles, good grain refinement (from 13 μm to 4 μm), and an increased tensile strength of 28%, along with a three times higher elongation compared to the unreinforced joints. The SiC nanoparticles were effective in suppressing grain growth through the Zener pinning mechanism and stimulated recrystallization. Reinforced joints showed improved hardness and ductile fracture behavior, while unreinforced joints behaved in a brittle way due to thick IMC layers.

Kumar. et al. [9] carried out Response-surface optimization of FSW parameters of dissimilar aluminum alloys AA6082 and AA7050. The effects of the tool rotational speed, welding speed, and tool tilt angle on tensile strength, elongation, and hardness were studied. Welding was performed using an H13 steel tool with a square pin on 6 mm thick plates. The optimized parameter combination, with a rotational speed of 900 rpm, a welding speed of 40 mm/min, and an inclination angle of 2° , resulted in a refined microstructure, a maximum tensile strength of 220.32 MPa, a joint efficiency of 80.12%, and a hardness of the stir zone of 104.5 HV. The developed RSM models demonstrated high prediction accuracy, with an R^2 value exceeding 0.92. The results revealed the advantages of increasing rotational speed in dynamic recrystallization and grain refinement, as well as the disadvantages of excessive heat input, including the formation of micro-voids and a decrease in joint strength. Fractographic observations revealed a ductile failure mode, characterized by fine, equiaxed dimples, indicating improved toughness and structural integrity.

Tabasi et al. [10] proposed a new method to dissimilar friction stir join of 7075 aluminium alloy and AZ31 magnesium alloy reinforced with silicon-carbide nanoparticles. A specially designed embedded groove has been incorporated to retain the nanoparticles during the welding process, preventing the scattering of nanoparticles and ensuring uniform distribution within the stir zone. Welding trials were conducted with tool rotational speeds ranging from 450 to 1100 rpm and traverse speeds from 11.2 to 45 mm/min, using a W302 hot-worked steel tool with a threaded triangular pin. The optimum processing condition was found at 710 rpm and 22.4 mm/min. fine and uniform grain structure with an average grain size of 4.36 μm . The silicon carbide nanoparticles were also effective in pinning grain boundaries, thereby enhancing dynamic recrystallization and grain refinement. However, excessive heat input resulted in the formation of brittle $\text{Al}_{12}\text{Mg}_{17}$ and Al_3Mg_2 IMCs, resulting in decreased ductility and brittle fracture. The highest ultimate maximum tensile strength of 106 MPa was achieved at TRS 560 rpm and TS 22.4 mm/min. The authors concluded that the pinning effect of nanoparticles dominated recrystallization and annealing in the microstructural refinement, and lower traverse speeds improved the weld integrity by decreasing particle agglomeration and intermetallic compound formation.

Singh et al. [11] presented a three-dimensional heat transfer and material flow model for dissimilar friction stir welding of Al6061 and AZ31 alloys using experimentally measured thermophysical properties. Unlike previous investigations that used average properties of the base material, this investigation measured the specific heat capacity, thermal diffusivity, thermal conductivity, and density of the stir zone experimentally. Welding was performed at 660 rpm and 30 mm/min using an H13 steel tool with shoulder diameters of 16, 20, and 24 mm. The numerical model successfully predicted the temperature distribution, thermal cycles, and the thermo-mechanically affected zone (TMAZ), showing good agreement with the experimental data. The study demonstrated that utilizing actual thermophysical properties enhanced the accuracy of peak-temperature prediction and symmetry of heat flow between the advancing and retreating sides. Larger shoulder diameters resulted in higher heat input, wider TMAZ, and increased intermixing, whereas the microstructure exhibited a lamellar and intercalated structure due to dynamic recrystallization and material flow. The model successfully captured asymmetric temperature fields and matched experimental observations,

thereby demonstrating that realistic stir zone property data are essential for accurately simulating dissimilar Al-Mg friction stir welding.

Kumar, S., & Wu, X. [12] Effects of ultrasonic vibration assistance on the formation of intermetallic compound (IMC) layer in friction stir welding of dissimilar AA6061-T6 aluminium and AZ31B magnesium alloys. Ultrasonic vibrations (20 kHz, 3000 W) were applied axially via a specially designed horn-tool system and induced a thermo-mechanical-acoustic effect within the stir zone (SZ). Welding was performed at rotational speeds of 400 to 1200 rpm and traverse speeds ranging from 50 to 250 mm/min using an H13 steel tool with a concave, threaded, tapered pin (3.2-2.8 mm) and a 10 mm shoulder. The ultrasonic-assisted joints showed a reduction in the thickness of the IMC layer of up to 57.9% and a significant improvement in joint strength (UTS 160.5 MPa, joint efficiency 71.3%) compared to conventional friction stir welding (UTS 112.3 MPa, 58.4%). Ultrasonic energy helped to improve the material flow, break down IMCs, and improve the interfacial bonding whilst reducing weld defects. The IMCs $\text{Al}_{12}\text{Mg}_{17}$ and Al_3Mg_2 were detected by EDS, and their thickness was minimized at 800 rpm and 100 mm/min. The authors concluded that ultrasonic vibration facilitates dynamic recrystallization, inhibits IMC formation, and results in defect-free Al-Mg welds with excellent mechanical properties.

U. Kumar et al. [13] investigated the friction stir welding of AA6061-T6 and AZ31B (6 mm) to map the weldability, microstructure, and microhardness at a wide range of rotation and travel speed. The experiments were conducted using a hardened H13 tool (approximately 56 HRC) with a tapered cylindrical pin (maximum diameter of 6 mm, minimum diameter of 4 mm, and a length of approximately 5.7 mm) and a shoulder diameter of 16 mm. The tool was offset by 3 mm toward the aluminium side and 2 degrees. Trials (TRS: 400-700 rpm; TS: 0.5-0.95 mm/s) yielded defect-free joints at rotation speeds of 500-700 rpm with a travel speed of 0.5 mm/s. In contrast, a higher heat input resulted in flash, the ejection of magnesium particles, and defects. SEM/EDS analysis showed the presence of $\text{Al}_{12}\text{Mg}_{17}$ and Al_3Mg_2 in the stir zone. Microhardness in the SZ increased significantly (up to approximately twice the base metal's hardness, with a range of 51-240 HV0.1) and also varied with the tool speed due to the formation of IMCs and dynamic recrystallization. The authors concluded that careful control of

heat input, positioning of the plates (aluminium on the advancing side, magnesium on the retreating side with aluminium offset), and rotation/travel speeds are necessary to prevent IMC-induced brittleness and obtain defect-free Al-Mg joints.

Shi et al. [14] investigated the formation of intermetallic compounds (IMCs) and their effect on the mechanical properties of dissimilar AA6061-Mg alloy friction stir welds. Welding was carried out at rotational speeds between 600 and 1000 rpm, with a constant traverse speed of 100 mm/min and a 1 mm tool offset towards the aluminium side, using H13 tool steel with a 12 mm shoulder and 4 mm pin diameter. The research revealed a banded structure (BS) zone comprising alternating IMC-rich and Al/Mg lamellar bands adjacent to the advancing side, where $\text{Al}_{12}\text{Mg}_{17}$ was predominant. The morphology, continuity, and density of IMC bands were affected by tool speed and had a significant impact on tensile strength. At 700 rpm, curved and discontinuous bands enhanced the strength (UTS $\sim$ 151 MPa), whereas at 800 rpm, dense, continuous IMC bands caused brittle fracture and low strength (UTS $\sim$ 133 MPa). All joints fractured along the BS region with brittle fracture features with limited elongation ($<1\%$ percent). The authors concluded that the microstructure of the BS can be tailored by controlling the band spacing, IMC density, and curvature to improve the mechanical performance in Al/Mg dissimilar friction stir welding.

Singh and Dubey [20] investigated the friction stir welding behavior of dissimilar AZ31 and AZ91 magnesium alloys using a hybrid cylindrical-threaded-pin tool made from H13 steel. A Box-Behnken Design (BBD) was used to investigate the effects of three factors, namely TRS (700-1000 rpm), TS (30-50 mm/min), and shoulder diameter (15-21 mm), on the effects of surface roughness, grain size, tensile strength, microhardness, and flexural strength. The resulting joints were defect-free in all conditions, and no IMCs were found in the stir zone. Shoulder diameter was found to be the most influential parameter, having a significant effect on surface roughness, grain size, and flexural strength. The best combination (1000 rpm, 40 mm/min, 15 mm shoulder) resulted in the highest tensile strength (234.86 MPa, approximately 89.7% of the AZ31 base metal strength) and an elongation of 11.35%. The micro-hardness in the weld zone ranged from 67.13 to 72.51 HV, with slight crack formation, which is attributed to the formation of MgO. Flexural strength was maximum at 380.06 MPa under the same conditions. The research proved that controlled heat input and tool geometry were

effective in refining grains, inhibiting IMC formation, and creating high-quality, defect-free dissimilar magnesium-magnesium joints.

Das et al. [21] have developed a thermo-mechanical numerical model to predict material defects in friction stir welding of dissimilar aluminium (AA6061) and magnesium (AZ31B) alloys, taking into consideration the often-neglected effect of tool wear. Using the coupled Eulerian-Lagrangian (CEL) approach in ABAQUS 2017, along with a Johnson-Cook plasticity and damage model, the authors simulated the material flow, mixing, and defect formation under varying tool rotations (900-1200 rpm) and traverse speeds (30-90 mm/min). The model was able to accurately predict the surface flash, subsurface tunnels, and exit hole defects with deviations of less than 1.2 mm from the experimental observations. The presence of IMCs ($\text{Al}_{12}\text{Mg}_{17}$ and Al_3Mg_2) and tunnel defects related to the lack of material plasticization at 400°C (60% of the melting point) was experimentally validated. Tool wear analysis revealed that pin length decreased by up to 30 percent and diameter varied by 18-28 percent with an increase in speed; the results were directly related to the formation of defects. Positioning AA6061 on the advancing side and AZ31B on the retreating side resulted in more uniform welds and a lower surface tunnel depth of less than 50% of plate thickness. The study showed that taking tool wear effects into account enables the more accurate prediction of defect initiation and evolution of dissimilar Al-Mg friction stir welding.

Singh and Dubey [22] conducted an in-depth numerical and experimental thermal-mechanical analysis of the dissimilar series alloys of Mg, Al, and Zn (AZ31/AZ91) in the case of friction stir welding. The study investigated the temperature field and residual stress processes using ANSYS Workbench with a finite-element model at process conditions ranging from 700 to 1000 rpm and 30 to 50 mm/min, with a shoulder diameter of 15 to 21mm. The model included heating from frictional and plastic deformation sources, and the maximum weld center temperature was predicted to be 544.04°C. The average von Mises stress was predicted to be 185.07 MPa at 1000 rpm, 40 mm/min, and a 21mm shoulder. The mean difference between predicted temperatures and the measured temperatures was an experimental validation of 11.4 2. Microstructural examination showed that slight cracks were formed as a result of MgO formation, and the tensile strength was 234.86 MPa (about 90% of AZ31 base metal strength). Residual stress analysis showed that higher stresses were shown on the

advancing side (AZ31) and minimum stresses at the weld center. The model successfully captured the effects of tool geometry and process parameters on temperature, stress, and defect evolution, demonstrating good agreement with experimental observations.

Yaduwanshi, Bag, and Pal [23] formulated a detailed three-dimensional transient heat transfer model to simulate the temperature distribution during the friction stir welding of Aluminium alloys AA1100 and AA6061. The model included both sliding and sticking friction conditions at the tool-workpiece interface with a separate heat flux contribution from the tool shoulder, pin side, and pin tip surfaces. A user-defined DFLUX subroutine in ABAQUS was used to calculate the heat generation due to frictional contact, excluding volumetric heat generated by plastic deformation. Experimental validation. Thermocouple measurements were taken to validate the experimental data, showing that the highest temperature (approximately 370°C) was recorded at the location behind the tool pin. Additionally, a temperature asymmetry was observed between the advancing and retreating sides, which could be attributed to different frictional forces. The model was found to be less than 5.5 percent in deviation of simulated and experimental peak temperatures. The study demonstrated that the heat generated from the pin side and tip has a significant impact on the temperature distribution, and that incorporating both sliding and sticking contact conditions enhances the accuracy of predictions, thereby providing a fundamental analytical-numerical framework for transient heat modeling in friction stir welding of aluminum alloys.

Ranjole et al. [24] Coupled thermo-mechanical finite element model (FEM) and experimental validation of dissimilar friction stir welding (FSW) of AA5083 aluminium and AZ31 magnesium alloys. The model, implemented in ABAQUS using a DFLUX user subroutine, was used to predict transient temperature and residual stress distributions for various rotational speeds (900-1400 rpm) and traverse speeds (40-165 mm/min). Simulations showed that the maximum temperature was around 400°C at 1400 rpm and 40mm/min, and this value agreed with the experimental measurements. Longitudinal residual stresses showed an increase with an increase in rotational speed, and traverse speed showed little effect. Experimental joints made with a high-speed steel (HSS) tool (18 mm shoulder, 6 mm pin diameter, 3.8 mm length) agreed with the

numerical results in terms of peak hardness (212.5 HV) and tensile strength (up to 207 MPa) at 1400 rpm and a feed rate of 165 mm/min. The lamellar microstructures and intermetallic compounds ($\text{Al}_{12}\text{Mg}_{17}$ and Al_3Mg_2) at the Al-Mg interface were identified by SEM-EDS and XRD analyses, respectively, which were facilitated by dynamic recrystallization during plastic flow and mixing. The study concluded that both modeling and experimental results show that optimized thermal input improves intermixing, reduces asymmetry, and improves joint integrity in dissimilar Al-Mg FSW.

Babu et al. [25] performed an integrated experimental and thermal simulation study to analyse the effects of different tool pin geometries and rotational speeds on peak temperature, material flow, and joint quality of AZ80A magnesium alloy FSW joints. The goal was to determine a profile of the pin that produces optimum heat without causing defects. Tools made from M35 high-speed steel had a 15 mm shoulder diameter, a 4.75 mm pin length, and three pin shapes: tapered cylindrical, straight cylindrical, and square. Welding trials were conducted at rotational speeds of 750, 818, and 1000 rpm, traverse speeds of 1.48 to 3 mm/s, and axial forces of 3.646 to 4.0 kN. A three-dimensional transient thermal model was developed in COMSOL Multiphysics using the governing heat equations and validated against infrared thermography measurements. The heat generation due to tool-workpiece friction was predicted in the model, considering the convective and radiative losses from the upper and lower surfaces. Simulated peak temperatures (603-633K) were in good agreement with experimental values (less than 1.5%). Results showed that while the straight cylindrical tool had the highest peak temperature (633K), it caused excessive heat, resulting in defects. On the other hand, the tapered cylindrical pin gave an ideal temperature of 616 K (i.e., 80% of the melting point of AZ80A), resulting in defect-free joints with fine equiaxed grains and a homogeneous nugget zone. Microstructural examination showed that 818 rpm with a tapered cylindrical pin produced the most refined grain structure because of balancing frictional heat and material flow. The brittle fracture behavior was confirmed by SEM fractography, which showed well-fused grains, thus high-quality bonding. Lower (750 rpm) and higher (1000 rpm) speeds resulted in insufficient or excessive plasticization, respectively, leading to voids or tunnel defects. The study concluded that the combined optimisation of rotational speed and pin geometry is

critical in achieving defect-free magnesium joints with a tapered cylindrical pin at 818 rpm, demonstrating optimal heat flow and strength for AZ80A FSW.

Moradi et al. [26] investigated the influence of tool pin geometry and multiple-pass welding on the microstructure, natural aging, and mechanical behavior of SiC-incorporated dissimilar FSW joints between AA6061-T6 and AA2024-T351 aluminum alloys. SiC microparticles (11-18 μm) were introduced into the joint line through a pre-machined groove, and welding was performed using square and triangular frustum pins made of H13 steel under the same process parameters. The single-pass and double-pass welds were compared with the same or inverse tool rotation directions. Results showed that the square frustum pin resulted in a finer and more homogeneous distribution of SiC and a smaller stir zone (SZ) grain size (8.1 μm) than the triangular pin (10.2 μm). Increasing the number of passes resulted in better particle uniformity, but also led to particle agglomeration bands and a decrease in hardness in the thermo-mechanically affected zone (TMAZ). Natural ageing for 70 days allowed the samples to regain hardness, especially those made with triangular pins, due to faster cooling and a higher precipitation driving force. The highest ultimate tensile strength (UTS 231 MPa) was obtained with the triangular pin, which was attributed to balanced hardness and refined grains, and double pass welds had greater ductility (toughness) and dimple fracture morphology. The study demonstrated that the tool geometry and pass strategy control SiC dispersion, grain refinement, and natural aging response, which in turn determine the mechanical performance in nanoparticle-reinforced dissimilar Al-Al joints.

Mouria et al. [27] examined the microstructural and mechanical responses of SiC and TiO_2 nanoparticle-reinforced dissimilar FSW joints between AA7075 and AA2024 aluminium alloys. Joints were manufactured with SiC, TiO_2 , and without reinforcement, all under the same parameters (1000 rpm, 30 mm/min, double pass). Grooves (1.5 mm diameter, 4.5 mm depth) were filled with nanoparticles in a zigzag pattern to ensure uniform dispersion. An H13 steel cylindrical tool (20 mm shoulder diameter, 5.5 mm pin diameter, and 5.7 mm length) was used. SEM and EDX showed a homogeneous distribution of nanoparticles and defect-free welds. XRD analysis revealed the presence of Al_2Cu and MgZn_2 IMCs, which enhanced the strengthening but kept ductility. The SiC-reinforced joint yielded the highest UTS (206 MPa), followed by the TiO_2 -reinforced (199 MPa) and unreinforced (177 MPa) samples. Hardness in the stir zone

has been found to increase to 172 HV (SiC) and 164 HV (TiO₂) compared to 147 HV for unreinforced specimens. Fractography revealed a ductile mode of fracture, characterized by finer dimples and limited cleavage facets, indicating improved strength-ductility. The results of the present study concluded that SiC and TiO₂ nanoparticles improved load transfer, refined grains through dynamic recrystallization, and reduced crack propagation, resulting in good joint strength and hardness in dissimilar Al-Al FSW.

Abdolahzadeh et al. [28] Effects of rotational and traverse speeds on microstructure and mechanical behaviour of SiC nanoparticle reinforced AZ31 magnesium alloy joints using friction stir welding. Nanosized particles of SiC (45-65 nm) were added into the joint line via a groove to create an in-situ composite in the stir zone (SZ). Welding was carried out at a rotational speed of 600-1000 rpm and a traverse speed of 25-75 mm/min using an H13 steel tool (20 mm shoulder, 5.8 mm threaded, tapered pin). The optimum parameters were found to be 800 rpm and 75 mm/min, which resulted in defect-free joints with homogeneous SiC dispersion and fine equiaxed grains (~5 µm, compared with 44 µm in the base metal). The SiC-rich and SiC-free onion-ring bands observed in the SZ promoted the grain refinement by DRX. The best sample obtained the highest UTS (239 MPa), yield strength (131 MPa), elongation (8.2 %), and average hardness (68.6 HV), and showed the improvement of ductility and strength compared with unreinforced AZ31. Fractography revealed that ductile dimples were present in the optimised specimens, while brittle cleavage was observed in the defective specimens. The study concluded that SiC nanoparticles simultaneously improve strength, hardness, and ductility by inducing DRX, dislocation pinning, and strong interfacial bonding, and the particle distribution was the most critical factor determining the mechanical performance.

Kaya et al. [29] systematically investigated the effect of theoretical volume fraction (TVF) of reinforcements on the microstructure, hardness, tensile strength, and wear behavior of Al5083/SiC and Al5083/ZrB₂ surface composites (SCs) fabricated by friction stir processing. Reinforcement TVFs of 5, 10, 15, and 20% were studied with micron-sized SiC (less than 15 µm) and ZrB₂ (5.5 µm) particles. FSP was carried out at 900 rpm, 20 mm/min, 8000 N axial load, and 3.5° tilt angle using an H13 tool with 20 mm shoulder and threaded M8 × 1.25 pin (4 mm length). The study introduced the

concept of comparing theoretical volume fractions (TVF) and actual volume fractions (AVF). It demonstrated that AVF values were consistently lower due to the effects of voids and thermal conductivity. Optimum mechanical performance was achieved at 15 (% SiC) and 5 (% ZrB₂ TVF) with a 55 (% increase in hardness (119.6 HV), a 3.3 (% increase in UTS (305.86 MPa)), and a 309 (% increase in wear resistance) when compared to a base metal (Al5083-H111). XRD showed the presence of Al, SiC, and ZrB₂, but no other intermetallic phases. Microstructural analysis of the material revealed the presence of fine equiaxed grains and defect-free stir zones at optimal TVF; however, excessive reinforcement (>15%) resulted in the formation of tunnelling and agglomeration defects due to reduced plasticization and heat flow. Fractography revealed a mixed ductile-brittle fracture mode, confirming the strong particle-matrix bonding. Overall, the performance of ZrB₂ reinforcement showed better strengthening and wear resistance compared to SiC, and 5% volume fraction of ZrB₂ was found to be the optimal volume fraction for high-performance surface composites.

Sharifi Asl et al. [30] investigated the effect of TiO₂ nanoparticles addition on microstructure and mechanical properties of dissimilar AA6063-T4 aluminium alloy and AZ31B-O magnesium alloy joints produced by friction stir welding. TiO₂ nanoparticles (20-70 nm) were incorporated through a novel pair groove interlocking design to avoid nanoparticle loss. FSW was conducted at a speed of 700-900 rpm and a feed rate of 40-60 mm/min, with the optimum joint achieved at 800 rpm and 40 mm/min. The tool was made of H13 steel (18mm shoulder, tapered threaded pin with a 6mm top, 4mm bottom, and a height of 4.65mm). Dynamic recrystallisation and pinning of nanoparticles resulted in a high refinement (12-16 µm) in the stir zone. The optimum joint had a UTS of 132 MPa (approximately 76% of the aluminum base metal), a hardness of 196 HV, and a mixed brittle-ductile fracture. TiO₂ nanoparticles enhanced dislocation density and hardness; however, when the heat input was too high (more than the optimum heat input), agglomeration and grain coarsening occurred. The major IMCs identified were Al₁₂Mg₁₇ and Al₃Mg₂, with the intermetallic layers at the interface controlling the initiation of fracture. An optimum heat input resulted in a defect-free interface, homogeneous TiO₂ distribution, and maximum joint strength, confirming that the addition of nanoparticles and controlling the heat are key to improving Al-Mg dissimilar welds.

Azizieh et al. [31] investigated the impact of rotational speed, probe geometry, and multi-pass friction stir processing (FSP) on the microstructure and hardness of AZ31 Mg/Al₂O₃ nanocomposites. Using H13 steel tools with non-threaded, threaded, and three-fluted probes, the study aimed at optimizing particle dispersion and matrix refinement. Nanocomposites were produced with Al₂O₃ reinforcement sizes of 35 nm, 350 nm, and 1000 nm at rotational speeds (800-1200 rpm) and travel speed of 45 mm min⁻¹ using 2 to 4 passes and tool tilt of 2 °. Comparisons of the three different types of tools revealed that the threaded probe produced the most uniform stir zone, containing no voids. In contrast, the non-threaded and fluted tools resulted in particle clustering with cavities. Increasing the number of passes reduced the Al₂O₃ cluster size from 490 nm to 230 nm and refined the grain size from 2.9 µm to 2.2 µm, confirming the progressive grain-boundary pinning effect. The hardness of the unreinforced AZ31 was 49 HV, which was doubled for the optimally processed nanocomposites, increasing from 49 to 92 HV, thanks to the combination of grain refinement and Orowan strengthening. However, the maximum rotational speed (1200 rpm) promoted fine cluster dispersion, but also resulted in grain coarsening due to excessive heat input, which limited further improvement in hardness. Consequently, the authors concluded that multi-pass FSP using a threaded pin at moderate speeds (800-1000 rpm) yields optimum particle dispersion, an ultrafine particle size (2-3 µm), and improved hardness in Mg-based nanocomposites.

Abdolazadeh et al. [32] Investigation of the microstructural and mechanical properties of SiC nanoparticle reinforced AZ31 magnesium alloy joints by friction stir welding (FSW). SiC nanoparticles (45-65 nm) were added along the joint line in a groove to prepare AZ31/SiC nanocomposites in the stir zone (SZ). Welding was performed at rotational speeds ranging from 600 to 1000 rpm and traveling speeds of 25 to 75 mm/min with an H13 steel tool (20 mm shoulder, threaded tapered pin, 5.8 mm length). The optimum condition was found to be 800rpm and 75 mm/min, which yielded the best combination of particle dispersion, microstructure, and tensile strength. The resultant joint exhibited fine equiaxed grains (5 µm) compared to the coarse base metal (44 µm) and displayed onion ring patterns with SiC-rich and SiC-free bands. The ultimate tensile strength (UTS) and yield strength (YS) obtained were 239 MPa and 131 MPa, respectively, with an elongation of 8.2% and an average hardness of 68.6 HV,

representing significant enhancements compared to unreinforced AZ31. Ductile fracture was observed in the optimized samples, while brittle fracture was observed in the defect-laden ones, as determined by fractographic analysis. The study concluded that a suitable SiC dispersion enhances grain refinement through DRX and composite strengthening through Orowan pinning and dislocation accumulation. In contrast, inappropriate heat input or particle agglomeration results in poor mechanical performance.

Bahrami et al. [33] Microstructural and mechanical behavior of nano-SiC reinforced AA7075-O aluminum alloy joints using two-pass friction stir welding (FSW) using different tool pin geometries. SiC nanoparticles (45-65 nm in size) were added to a closed groove between two adjacent plates before welding. Two pin designs (threaded tapered and square) were used, made from H13 steel (58 HRC), for single and dual pass conditions, as well as reverse rotational direction in the second pass. Optimum joining parameters were achieved at 1250 rpm and 40 mm/min, resulting in homogeneous dispersion, small grains (3.6 μm compared to 73 μm of the base metal), and low particle agglomeration. The square pin in the second pass enhanced the SiC distribution, and the cluster size decreased from 256 nm to 219 nm. The best mechanical properties were obtained in the hybrid pass (threaded and square), with a UTS of 270.7 MPa, an elongation of 11.1%, and a hardness of 123.7 HV. Fracture analysis revealed that the SiC-reinforced joints fractured from the base metal, exhibiting ductile dimples. In contrast, the unreinforced joints were fractured with quasi-cleavage characteristics in the stir zone. Enhanced performance was attributed to Orowan strengthening, Zener pinning, and grain refinement by DRX. The authors determined that multi-pass FSW with different pin geometries is very effective in achieving uniform dispersion and better mechanical integrity of nanoparticles in Al-SiC nanocomposite welds.

Karthikeyan and Mahadevan [34] investigated the influence of SiC particle addition in the weld zone during friction stir welding (FSW) of Al 6351 alloy on its mechanical performance in both as-welded and annealed conditions. SiC particles (12 μm) were manually dispersed along the joint line with weight fractions of 1.0wt% and 1.5wt%. Welding was carried out on a vertical milling machine following a Taguchi L18 orthogonal array, based on four parameters, namely SiC wt (%), axial load (1.5-2.5 kN), rotational speed (710-1400 rpm), and welding speed (14-29 mm min⁻¹). Mechanical

properties were tested with tensile test, Brinell hardness test, and impact test in both conditions. The as-welded state had a 33% increase in tensile strength and hardness compared to unreinforced joints, while annealed joints had only a 25 to 29% decrease compared to the base alloy, which indicates the pinning effect of SiC to grain coarsening. Uniform dispersion of SiC and pinning of grains at the boundaries during annealing was verified by SEM-EDS analysis, and this was found to preserve hardness and strength. The best parameter combination, i.e., 1.5 wt% SiC, 2 kN loading, 1400 rpm, and 21 mm/min, resulted in a uniform particle distribution and improved impact resistance (1.26 kg·m as compared to 1.12 kg·m for the base metal). The authors concluded that the addition of SiC reduces grain growth in post-weld heat treatment and significantly improves the overall quality and mechanical integrity of the Al 6351 welds.

Moradi et al. [35] investigated the impact of SiC particle size (nano vs. micro) on the microstructural evolution, particle dispersion, and mechanical response of dissimilar friction stir-welded joints between AA6061-T6 and AA2024-T351. SiC particles in two size ranges, nano (25-50 nm) and micro (11-18 μm), were added along the joint interface by the grooves of different widths (0.2-0.4 mm) and tool offsets (0-1 mm) towards the AA6061 side. Experiments were done using an H13 tool steel with an 18 mm shoulder (4° concave cavity), a square frustum pin (3.5-7 mm diameter, 5.9 mm length), a 2° tilt with 800 rpm, and 31.5 mm/min TS. Results showed that the best configuration, i.e., 0.3 mm groove width and 0.5 mm tool offset with nano-SiC particles, resulted in a defect-free weld with a homogeneous stir zone. Nano-SiC improved the flow of materials and reduced tunnel defect formation compared to micro-SiC, due to its rounded shape and smaller size, which increased the plastic flow of the material. Microstructural analysis (OM and SEM-EDS) revealed evident onion-ring structures with refined grains. Dynamic recrystallization and Zener pinning were found to be the dominating mechanisms of grain refinement. Microhardness results showed an overall increase in stir zone hardness (up to 130 HV) compared to the base materials (AA6061, 107 HV; AA2024, 137 HV). The enhancement was more pronounced in the nano-SiC joints due to uniform dispersion and strain accommodation at the particle-matrix interface. Tensile testing revealed that the nano-SiC-reinforced joint exhibited a UTS of 237 MPa and YS of 177 MPa, representing 76% and 63% of the UTS and YS

of the AA6061 base metal, respectively. Tensile toughness increased from 16.9 (MJ m⁻³) (no SiC) to 27.4 (MJ m⁻³) (nano-SiC), indicating that the energy absorption and ductility were enhanced. Fracture analysis revealed fine dimples in nano- and mixed cleavage, as well as dimples in micro-SiC joints (ductile mode and mixed cleavage), confirming the enhanced interfacial bonding with nanoparticles. The study concluded that nano-SiC reinforcement is more advantageous in terms of particle dispersion, grain refinement, hardness uniformity, and tensile toughness compared to micro-SiC. Additionally, it found that a groove width of 0.3 mm and a tool offset of 0.5 mm are optimal for creating defect-free, high-performance dissimilar Al-Al welds.

Verma et al. [36] investigated the effect of rotational and traverse speeds of the tool on the microstructure, corrosion behavior, and mechanical performance of dissimilar AZ31B magnesium and AA6061 aluminum alloy FSWed joints. Welding was performed using an H13 steel tool with a concave shoulder of 18 mm and a tapered pin with a 5 mm root diameter and a 3 mm tip diameter, measuring 3.5 mm in length. The tool was rotated at speeds ranging from 560 to 1010 rpm, and the traverse speed ranged from 16 to 25 mm/min. Defect-free welds were obtained at intermediate speeds; excessive TRS (greater than 1010 rpm) led to liquation, oxide formation, and intermetallic (IMC) generation (Al₁₂Mg₁₇ and Al₃Mg₂). The stir zone showed alternating lamellar shear bands of Al and Mg because of dynamic recrystallization (DRX), resulting in a characteristic onion ring structure. The maximum tensile strength (118.56 MPa) and hardness (165.8 HV) were achieved at a rotation speed of 1010 rpm and a feed rate of 25 mm/min. The refined grains (3 µm) are interlocking at the interface. Joints used with low rotation/high traverse speeds showed incomplete mixing and low strength (55-63 MPa). Fractography showed a brittle cleavage pattern in all joints and confirmed the presence of brittleness induced by IMC. The authors concluded that both higher rotational and moderate traverse speeds optimize strength, hardness, and corrosion resistance in Al-Mg FSW joints.

Azizieh et al. [37] investigated the microstructural and mechanical properties of dissimilar friction stir-welded (FSW) joints between AA1100 aluminum alloy and AZ31 magnesium alloy, examining the effects of rotational speed, traverse speed, and pin position. Welding was performed using a tool made of steel H13 with a threaded pin of 5 mm in diameter, 2.7 mm in height, and 1 mm in pitch, featuring a shoulder

diameter of 18 mm. The tool was operated at a rotational speed of 420-1000 rpm and a traverse speed of 15-30 mm/min. Varied pin offsets (1.5-2.5 mm) were made towards either the Al or the Mg side. Sound, crack-free joints were obtained only when the tool was offset to the AZ31 side, which improved material flow and decreased liquation. The optimum rotational-to-traverse speed ratio (about 28-32 rev mm⁻¹) resulted in maximum tensile strength (122 MPa) and elongation (9%) that corresponds to about 70% and 53% of the strengths of AA1100 and AZ31 base metals, respectively. XRD and EDS analyses confirmed the formation of Al₁₂Mg₁₇ and Al₃Mg₂ IMCs in the stir zone, and the IMC layer thickness increased with increasing peak temperature (400-500 °C). Maximum hardness (110HV) was obtained at 1000rpm as a result of fine Al₁₂Mg₁₇ dispersion. A refinement of the grains on both sides was observed (4.1 and 5.9 micrometers in Mg and Al, respectively). The results of this study indicate that an optimum heat input in the vicinity of the Al-Mg eutectic temperature (430-460°C) yields the best joint quality, while excessive or insufficient heat results in liquation cracking or non-mixing. A linear correlation was established between IMC volume fraction, peak temperature, and microhardness, highlighting the importance of temperature control in improving Al-Mg dissimilar welds.

Mehdi et al. [38] investigated the effect of friction stir welding (FSW) process parameters on the microstructural and mechanical properties of dissimilar AZ91D magnesium and AA7075 aluminum alloys in a comprehensive investigation. The study examined the effects of the rotational tool speed (RTS: 400-800 rpm) and traverse speed (TS: 20-50 mm/min) using an H13 steel tool with a 20 mm shoulder diameter and a square pin (5.8 mm width, 5.5 mm length tilted 3°). The AA7075 plate was used on the advancing side (AS) and the AZ91D on the retreating side (RS). The results showed that both process parameters and heat input had a significant effect on IMC formation, grain refinement, and joint efficiency. Defect-free joints were obtained at 800 rpm and 20 mm/min with a maximum UTS of 116.64 MPa and a joint efficiency of 59.8 % and a minimum UTS (68.32 MPa) was obtained at 400 rpm and 20 mm/min due to dense IMC layers (Al₁₂Mg₁₇ and Al₃Mg₂). Microstructural observations (OM, SEM, TEM, EDS) showed that the material has a lamellar intercalated structure, DRX-induced fine grains, and well-bonded Al-Mg interfacial regions with nanosized Al₃Mg₂ precipitates (~200 nm). XRD showed the presence of Al₁₂Mg₁₇ close to AZ91D side and Al₃Mg₂

close to AA7075 side, and the presence of constitutional liquation near eutectic temperatures (435-450 °C). Hardness varied from 84 to 95 HV in the stir zone, reaching a maximum value of 93 HV at 400 rpm (50 mm/min) due to the formation of DRX and IMC. Fractography showed brittle cleavage failure with an overwhelming IMC layer, small ductile areas with swirl patterns, and localized plastic deformation. The authors concluded that optimized heat input (moderate RTS and low TS) improves intermixing, controls IMC thickness, and improves the mechanical properties of AZ91D-AA7075 joints.

Karakizis et al. [39] studied the effect of SiC and TiC nanoparticles reinforcement on the microstructural evolution, dislocation behavior, microhardness, and tensile strength of AA6082-T6 FSWed joints. Nanoparticles were introduced in a machined groove ($180 \times 1 \times 2.5 \text{ mm}^3$) between the faying surfaces before welding, using a flat-shouldered, left-hand threaded tool with a shoulder diameter of 23 mm, pin diameter of 6 mm, pin height of 2.8 mm, and tilt angle of 3° made of SVERKER-21 tool steel (61HRC). Welding was carried out at rotational speeds ranging from 750 to 1500 rpm and traverse speeds from 19 to 85 mm/min, with the addition of up to three passes to improve the dispersion of nanoparticles. Both SiC (20-30 nm) and TiC (150-200 nm) particles produced by CVD were uniformly distributed throughout the weld nugget after three passes, with minor localized agglomeration in the case of SiC. TEM and SEM-EDS analyses revealed the presence of dynamic recrystallization (DRX) in all welds, resulting in a drastic decrease in grain size from 40 micrometers (in the base metal) to ~4-5 micrometers (in the nugget) and an increase in dislocation multiplication. The addition of nanoparticles slightly improved boundary pinning, further enhancing grain refinement and stabilizing the sub-grain structure. Three intermetallic phases with the compositions Mg_2Si , $\text{Al}_9\text{Mn}_3\text{Si}$, and $\alpha\text{-Al}_{12}(\text{FeMn})_3\text{Si}$ were found, indicating partial dissolution and reprecipitation during FSW, irrespective of the reinforcement type. No interfacial reaction was observed between the nanoparticles and the matrix. The SiC reinforced joint was found to have a UTS of 167.6 MPa, elongation of 5.97, and hardness of 105-110 HV, while the TiC reinforced joint showed a UTS of 190.6 MPa, elongation of 3.95, and hardness of 118-122 HV. The unreinforced AA6082-T6 joint exhibited a UTS value of 199 MPa and an elongation of 3.79%. All failures occurred in the heat-affected zone, consistent with the minimum hardness area resulting from

precipitate coarsening. SiC-reinforced joints exhibited higher ductility and toughness (913 Jm⁻³), while TiC exhibited higher hardness and wear potential. Ductility dimple rupture was confirmed by fractography in all joints. The study concluded that SiC contributes to ductility and energy absorption. In contrast, TiC enhances the hardness, while both reinforcements result in defect-free welds, uniform particle distribution, and microstructural refinement due to the coupled effects of DRX and nanoparticle pinning.

Singh et al. [40] examined the effects of the volume fraction of Al₂O₃ nanoparticle (0.2-0.4 vol%) on the microstructural, mechanical, and tribological properties of friction-stir welded (FSW) AA6061-T6 aluminum alloy nanocomposites. The nanocomposites were prepared by loading machined holes on the faying surfaces of 2.5-mm-thick AA6061-T6 sheets with Al₂O₃ particles (15nm diameter, 99.99% high purity) in a zig-zag pattern before welding. Welding was performed on a vertical milling machine using a cylindrical H13 steel FSW tool with a 15mm shoulder and a 5mm pin of rounded base, at a rotational speed of 2000rpm, a traverse speed of 70mm/min, and a zero-tilt angle. Optical microscopy and scanning electron microscopy revealed a uniform distribution of nanoparticles at a volume fraction of 0.3%, resulting in a nugget zone (NZ) depth of 3286 μ m. Well-shaped fine equiaxed grains were formed through continuous dynamic recrystallization (CDRX) and Zener pinning. The microhardness of the micrograin was measured in the NZ. It was 125 HV, which is significantly higher than the base alloy, and this can be attributed to the refinement of the grains and the even distribution of particles. The tensile testing revealed that 0.3 vol% Al₂O₃ was the optimal proportion for achieving both strength and ductility, while 0.2 and 0.4 vol% resulted in nanoparticles agglomeration and interfacial debonding, respectively, and reduced tensile strength. Fractographic data were used to verify fractographic evidence of ductile failure in thin dimples of 0.3 vol% and brittle fracture of 0.4 vol% in the specimen, which had particles pulled out and voids. Wear studies demonstrated the lowest wear rate and coefficient of friction at 0.3vol%, which was a result of the creation of a stable tribolayer that served as a solid lubricant. Therefore, the authors inferred that the addition of 0.3 vol% of Al₂O₃ is the most effective in the dispersion of particles, microhardness, wear resistance, and tensile performance in FSWed AA6061-T6 nanocomposites.

Suman et al. [41] investigated the effect of SiC particle volume fraction on the temperature distribution, microstructure, and mechanical behavior of AA6061-T6 plates processed by friction stir processing (FSP). The experiments were conducted using an H13 steel tool with a concave shoulder and a taper-threaded pin (5.8mm) under constant process conditions: rotational speed, 1100 rpm; traverse speed, 1.5mm/s; and tilt angle, 2 degrees. A groove was used to add SiC particles (15mm) in different volume ratios (0%, 0.2%, 0.25%, and 0.3%). K-type thermocouples were used to record temperature profiles at various positions on the advancing side of the sample. As the volume fraction of SiC increased, the peak temperature rose to a maximum of 353.45 °C with 0.25 % volume fraction, after which the temperature began to fall at higher concentrations (0.3 % volume fraction) as a result of the heat-absorbing property of the SiC. The microstructural examination showed that equiaxed grains were in the nugget region, and the grain refinement was as high as 0.25-0.5 Vf, followed by coarsening due to the excessive heat and particle agglomeration. At 0.25% Vf, uniform dispersion of SiC was obtained, which favored dynamic recrystallization (DRX) and Zener pinning. The highest microhardness (127 HV 0.1) and tensile strength (341 MPa) were achieved at 0.25 percent VF, which is a significant increase from the unreinforced sample (UTS 267 MPa, HV 107). Still, the ductility reduced with the addition of particles (17.8 to 8.1). Fractography revealed that dimples and micro-void coalescence dominated ductile failure modes at lower volume fractions, while partial cleavage prevailed at higher fractions. The researchers found that a 0.25 percent volume of reinforcers (SiC) had the best strength, hardness, and homogeneity of thermal conductivity; however, above that, there was poor mixing of the reinforcers and also poor mechanical performance.

The study conducted by Jain and Mishra [42] is based on a multi-response optimisation of dissimilar friction-stir welded (FSW) joints of AA6061 and AA7075 aluminum alloys, which were prepared using micro-sized particles of Al₂O₃ (4 vol%). The aim was to optimize the TRS (700 to 1100 rpm), TS (40 to 60 mm/min), and the volume fraction of reinforcement using Response Surface Methodology (RSM) on a Central Composite Design (CCD) with 20 experimental runs. A tool used was an H13 steel, which features a threaded cylindrical pin (6mm diameter, 3mm length) with a shoulder of 18mm and a 2° tilt angle on the retreating side, made of AA7075. The findings

revealed that rotational speed had the most significant effect, followed by the Al_2O_3 content, and lastly, the traverse speed. The best condition (971 rpm, 40 mm/min, 10% Al_2O_3) yielded an ultimate tensile strength of 241.4 MPa, a microhardness of 157.5 HV, and an elongation of 19.4%, corresponding to an approximate joint efficiency of 83%. Microstructural observations also revealed uniform dispersion of Al_2O_3 and fine grains (7.2 μm) in the stir zone, attributed to dynamic recrystallization and particle pinning. Reduced rotational speeds (700 rpm) resulted in clustering of Al_2O_3 and partial bonding, which impacted the mechanical performance. Analysis of variance revealed that rotational speed accounted for more than 84% of the variance in strength. The quadratic model of all responses demonstrated excellent predictive ability ($r^2 > 0.99$), with confirmatory tests indicating an error rate of less than 5% in predictions. Under ideal conditions, the fracture exhibited ductile characteristics, characterized by fine dimples and micro-void coalescence adjacent to the AA6061 side. The experiment found that Al_2O_3 reinforcement of 10 percent, combined with a moderate rotational speed and a low traverse speed, yields an optimum balance of strength, hardness, and ductility in dissimilar AA6061-AA7075 joints.

In their study, Anil Kumar et al. [43] experimentally investigated dissimilar friction-stir welded joints between AA2024-T351 and AA7075-T651 in the presence of SiC nanoparticles (4050nm) at different volume fractions (0, 5, 8, 13 percent). The study investigated the impact of nanoparticle concentration and multi-pass FSW (first and second passes) on the microstructure, mechanical properties, and strengthening behavior of the metal-matrix nanocomposite (MMNC) in the weld nugget zone (WNZ). Experiments involved a taper-threaded HDS tool (55 HRC) with a pin length of 6.2-6.3mm, a 2° Tilt angle, a rotational speed of 1000 rpm, and a traverse speed of 40 mm/min. SiC nanoparticles were placed in advance in rectangular edge grooves (0.2 to 0.5mm), at the interface of the plate, with AA2024-T351 in place on the advancing side. The second FSW pass in reverse rotation had a significant effect on SiC dispersion and the removal of defects left by the first pass. The analysis of the images revealed that a smaller actual SiC volume fraction (39.3%) was used in the second pass, as the WNZ area was larger. Microstructural analysis has revealed fine equiaxed grains (2-4 μm) with evenly distributed SiC in the WNZ with 5 percent reinforcement; however, high fractions (8-13 μm) led to agglomeration, porosity, and debonding. The optimum

condition (5% SiC, second pass) was recorded to have an ultimate tensile strength of 411 MPa, yield strength of 241 MPa, and elongation of 14.3%, which is higher than the unreinforced joint (UTS= 287 MPa, YS= 176 MPa, and elongation= 4.7%). At 13 per cent SiC, maximum hardness (150 HV) was attained. The primary improvement in strength was attributed to Orowan strengthening and Hall-Petch grain refinement, with minimal contributions from thermal expansion mismatch or load-bearing mechanism. Fractography revealed that ductile failure was characterized by fine dimples and strong bonding of SiC to the SiC matrix at 5% SiC content. However, at higher contents, interfacial cracking was observed. The research proved beyond doubt that an optimum nanoparticle fraction (approximately 5 percent) and dual-pass stirring can produce a homogeneous dispersion, a refined microstructure, and a high tensile-hardness interaction in dissimilar Al–Al nanocomposite joints.

Subramani et al. [44] investigated the effect of nano-SiC reinforcement volume fraction on the microstructure and mechanical behavior of friction-stir-welded (FSW) AZ31B magnesium alloy welds. SiC nanoparticles (~50nm) were added along the weld line using grooves of different widths (0.3-1.5mm), corresponding to volume fractions of 4-20% (V4-V20). The welding was performed using an H13 steel tool (56 HRC) with a cylindrically threaded pin (5.7mm long) and a D/d ratio of 3, at a rotational speed of 1250 rpm and a traverse speed of 25 mm/min. Macroscopic and microscopic analyses revealed defect-free joints in all conditions. However, at the increased fractions (V16, V20), agglomeration of the particles and clustering were observed due to inadequate heat production. Specimens V4-V12 exhibited uniform SiC dispersion and clear onion-ring structures in the nugget zone, whereas V16 and V20 showed poor plasticization and homogeneous flow. SEM-EDS analysis confirmed that there were no voids or reaction products in the stir zone and that there were SiC particles. The optimum joint (V12, 12% Vf) had a final tensile strength of 235.81 MPa, microhardness of 84 HV, and elongation of 6.6%. Dynamic recrystallization and Zener pinning of SiC particles, as well as excessive reinforcement, were credited with mechanical enhancement and ductility, respectively. Fractography V4-V12: dimples, V16-V20: brittle cleavage. It was concluded that the optimal level of heat input, the strength-ductility balance, and microstructural uniformity of MgSiC FSW joints are enhanced by the addition of a 12 0 SiC nanoparticle.

In Bhatnagar et al. [45], an optimization study is presented on dissimilar friction-stir welded joints between AA7050 and AA6061 aluminum alloys, using the Response Surface Methodology (RSM) to maximize tensile strength and hardness. The design employed was a Central Composite Design (CCD), with input factors including rotational speed (1100-1500 rpm), welding speed (120-150 mm/min), and tool tilt angle (1-2 degrees). The joints were made using a non-consumable tool with a threaded cylindrical pin. Analysis of variance revealed that rotational speed (TRS) and welding speed (WS) were the most significant variables influencing ultimate tensile strength (UTS), hardness, and strain. The highest UTS (255.83 MPa) and strain (21.02%) were obtained at the optimal condition, which was 1500 rpm, 150 mm/min, and a 2° tilt angle. The lowest UTS (148.58 MPa) was recorded at a feed rate of 120 mm/min and a rotation speed of 1100 rpm. The high accuracy of the statistical model ($R^2 = 0.9829$ with UTS, 0.9827 with strain, and 0.9723 with hardness) indicated the sufficiency of the regression equations developed. The research developed empirical equations to forecast tensile, strain, and hardness responses, and confirmed them through experimental verification within a 95% confidence interval. The findings have shown that faster rotational rates increase the consolidation of material, refinement of the grains, and strength of the joints. However, excessive or insufficient heat input results in an inhomogeneous microstructure and degradation of properties. Therefore, optimisation using RSM can be successfully employed to enhance the integrity of dissimilar AA7050-AA6061 welds.

Ravi Kumar et al. (2021) [46] optimised the parameters of friction-stir welding (FSW) between dissimilar AA6061 and AA6082 aluminium alloys over Response Surface Methodology (RSM) to determine the effects of tool rotational speed (TRS: 800-1400 rpm), and traverse speed (TS: 40-60 mm/min) and tilt angle (0-2°) on tensile strength, hardness and strain. A H13 tool steel with a 6mm diameter pin, a 5.5mm pin length, and a 19.5mm shoulder diameter was used. All welds were made on a CNC FSW machine with a dwell time of 45 seconds and at a perpendicular angle to the rolling direction. Maximum tensile strength (225 MPa) and hardness (104 HV) were recorded at 1400 rpm, 40 mm/min, and 2° tilt, and minimum tensile strength (135 MPa) and hardness (78 HV) were recorded at 800 rpm, 60 mm/min, and 2° tilt. The ANOVA test was used to determine the statistical significance of the empirical models developed ($F = 13.11$, $p <$

0.05, $R^2 = 0.9219$, tensile strength). The models forecasted the optimum outputs of 205.6 MPa UTS, 18.97 percent strain, and 105.35 HV hardness at 1178 rpm, 45.9mm/min, and a 1.89-degree inclination. Microstructural observations revealed that fine equiaxed grains were present in the stir zone, resulting from dynamic recrystallization, while coarse, elongated grains were present in the heat-affected zone, which was associated with lower hardness and tensile failure on the AA6082 side. Strain gauging. The asymmetrical profile of hardness was attributed to the different heat production and dissolution of precipitates on the advancing (AA6061) and retreating (AA6082) sides. The research concluded that RSM-based models are effective in predicting mechanical behavior with high accuracy and optimality of the heat input at 1400 rpm and 40 mm/min to 60 mm/min, yielding defect-free welds with a homogeneous microstructure and strength.

Elangovan et al. [47] developed a mathematical model to predict the tensile strength of friction-stir welded (FSW) AA6061 aluminium alloy joints using Response Surface Methodology (RSM) and Design of Experiments (DOE) in terms of process parameters and tool geometry. The experiment employed a four-factor, five-level central composite design (CCD) that included tool pin profile, rotational speed (800-1600 rpm), welding speed (0.25-2.25 mm/s), and axial force ($F = 59$ kN). The joints were made using high-carbon steel tools, which had a shoulder diameter of 18 mm and a profile that included a pin (straight cylindrical, threaded cylindrical, square, triangular, and tapered cylindrical). The developed regression model was tested using analysis of variance (ANOVA) at a 95% confidence level. The highest tensile strength (192 MPa) was achieved with joints welded using the square pin tool at 1200 rpm, 1.25 mm/s, and a 7 KN axial force, demonstrating the importance of tool geometry and optimal heat input on joint performance. Observation under microscopes showed that moderate parameters revealed defect-free welds, whereas low heat input (≤ 800 rpm, 0.25 mm/s - 1) and excessive input (1600 rpm, 2.25 mm/s) resulted in tunnel and pinhole defects, respectively. The square and triangular pin tools were superior in stirring. The good agreement between the predictive and experimental results was affirmed by the R^2 value (0.91). The report found that the square pin profile with medium parameter values produces the best tensile strength and defect-free joints, confirming the consistency of RSM-based predictive modelling in predicting the AA6061 alloy FSW.

Mallieswaran et al. (2018) [48] used Response Surface Methodology (RSM) to optimize the friction-stir welding (FSW) parameters of dissimilar AA1100 and AA6061 aluminum alloy welded blanks in the form of tailored welded blanks (TWB). The aim was to develop an empirically validated model that correlates process parameters with tensile strength. The chosen factors included tool rotational speed (650850 rpm), weld speed (1050 mm/min), tool tilt angle (1-3 °), and tool pin profile (cylindrical, taper cylindrical, taper threaded, cylindrical threaded, and square). It used a four-factor and five-level Central Composite Design (CCD) of 30 experimental runs. A super-high-speed steel (SHSS) tool with a 1mm pin diameter and a length of 1mm was used for welding, where AA1100 was attached to the advancing side and AA6061 to the retreating side. The predictive reliability of the model was confirmed by analysis of variance ($F = 210.51$, $p < 0.0001$, $R^2 = 0.995$, $\text{adj. } R^2 = 0.991$), indicating the model's adequacy. The highest tensile strength of 234 MPa was obtained with the optimum combination of 725 rpm, 30 mm/min, a 2-threaded taper pin, and a prediction error of less than 1.5%. The microstructural analysis revealed that fine equiaxed grains and onion-ring structures were present in the stir zone, resulting from dynamic recrystallization. In contrast, the TMAZ exhibited elongated, deformed grains, and no coarsening occurred in the HAZ. The pin profile was found to have the most significant impact on the joint's strength, followed by welding speed, rotational speed, and tilt angle. The taper-threaded pin enhanced the flow and intermixing of plastics, resulting in flawless joints. The paper has shown that empirical modelling is based on RSM.CCD is a good predictor and optimizer of the tensile behaviour of dissimilar TWB FSW joints.

2.2 Mechanism of Joining Aluminium (Al) and Magnesium (Mg) by FSW and its Incompatibility Issues

Mechanical interlocking due to continuous material deformation and the formation of weld bonds resulting from the existence of Intermetallic Compounds (IMCs) are the two main factors on which the FSW weld strength of dissimilar Al-Mg alloys (heterogeneous structures) depends [13,14]. Limitations in Al-Mg solubility constitute a significant issue during the FSW of Al-Mg alloys because they promote the formation

of IMCs. The Al-Mg Equilibrium Phase Diagram is shown in Figure 2.1, which discloses various facts about Al-Mg solubility [49].

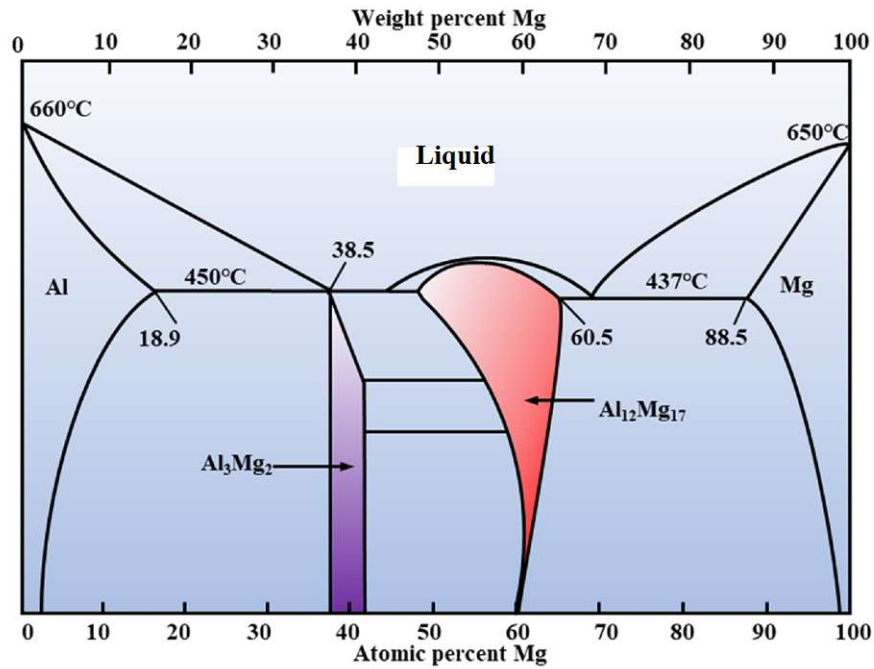


Figure 2.1: Al-Mg Equilibrium phase diagram system [30].

According to the Al-Mg phase diagram, intermetallic compounds such as Al_3Mg_2 and $\text{Al}_{12}\text{Mg}_{17}$ are formed in the weld after solidification. As we know, welding is a non-equilibrium process, even though the equilibrium phase diagram is widely used as a reference for predicting the stability of phases and various chemical reactions throughout FSW. Based on the trends of solvus-line (not shown), the solubility of Al & Mg at room temperature is exceptionally low. Furthermore, two eutectic streaks at 437°C and 450°C, in the binary framework, are significantly lower than the liquefaction temperatures of unalloyed Aluminium (660°C) and Magnesium (650°C). Three intermetallic phases, i.e., (i) Al_3Mg_2 (intricate cubic β) phase, (ii) $\text{Al}_{12}\text{Mg}_{17}$ (cubic of α -Mn type γ) phase, and (iii) rhombohedral R phase (ϵ -phase) make up the outline. During the cooling phase of welding, Al_3Mg_2 and $\text{Al}_{12}\text{Mg}_{17}$ phases, which are unique in their composition, structure, and properties, stabilize. This is because friction stir welding is a solid-state welding process. The Aluminum side layer is termed Al_3Mg_2 (Liquid $>$ Al $\pm$ Al_3Mg_2 at 450°C), while the Magnesium side layer consists of $\text{Al}_{12}\text{Mg}_{17}$ and Mg-rich phase (Liquid $>$ Mg $\pm$ $\text{Al}_{12}\text{Mg}_{17}$ at 437°C) [49,50]. Figure 2.2 shows micrographs

of the different IMCs phases of the investigated alloys AZ91 and AA5454. AA5454 and its microstructures are made up of $Al_x(Si, Fe, Mn)$, a type of IMC precipitate, integrated into an aluminum solid solution α -phase [51].

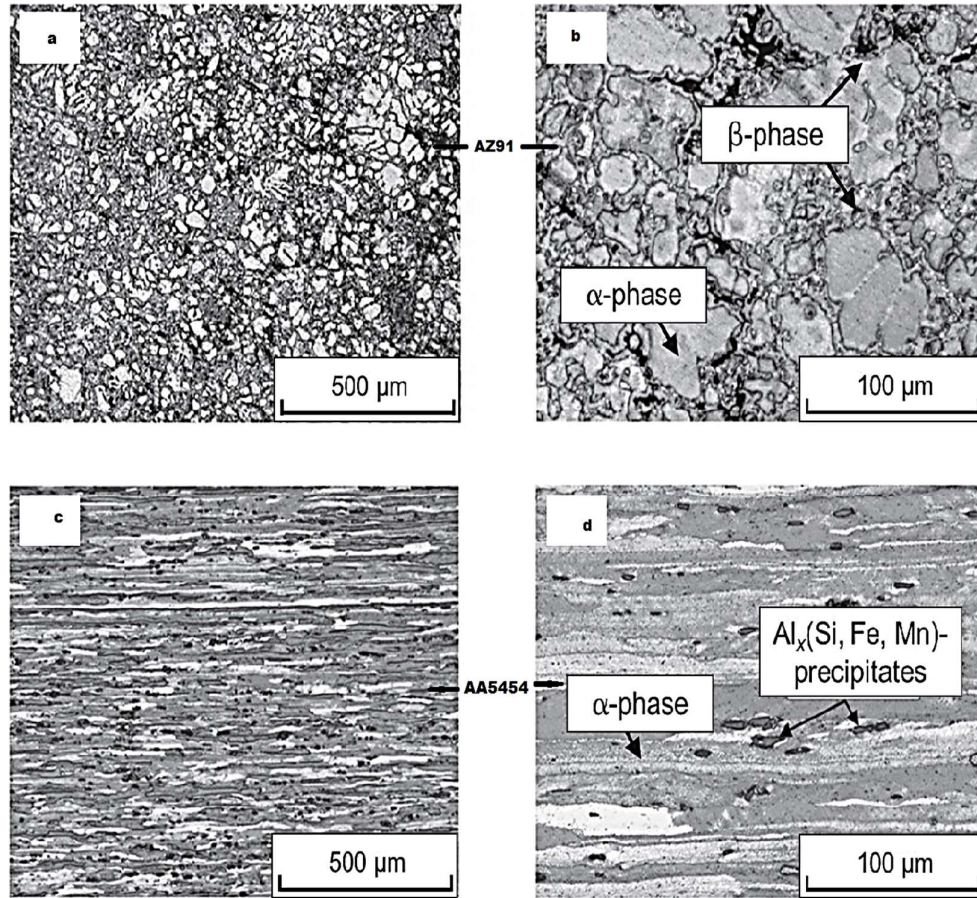


Figure 2.2:(a, b) Micrographs of various phases of AZ91, (c, d) AA5454 [51].

The IMCs show different precipitation behavior in the weld nugget zone (WNZ) along the thickness direction. Because the temperature in the upper WNZ stays relatively high, large-sized IMCs (Al_3Mg_2 , $Al_{12}Mg_{17}$, and $AlMg$) precipitate during solidification in this zone, whereas, in the lower WNZ, simple formation of small-sized Al_3Mg_2 particles takes place. This shows that various metallurgical reactions occur in the WNZ [52]. Figure 2.3 shows the steps involved in the precipitation of Intermetallic compounds in the upper portions of the weld nugget zone [52]. The formation of IMCs mainly depends on the local temperature and material composition during FSW, irrespective of the sample's elemental composition (i.e., chemical composition). Many

researchers have made numerous efforts to reduce or eliminate the formation of these IMCs, but success has not yet been achieved.

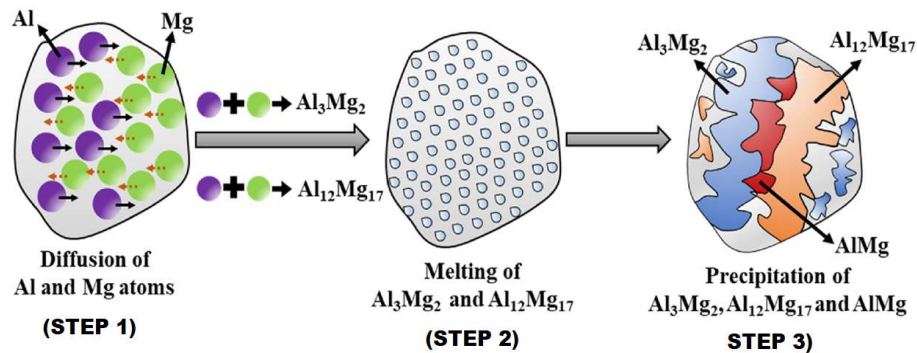


Figure 2.3: Steps involved in the precipitation of IMCs [52].

2.2.1 IMCs Formation Mechanism

The biggest challenge is limiting the formation of brittle and hard IMCs layers at the weld joint interface during dissimilar Al & Mg FSW. It deteriorates the strength and ductility of Al & Mg joints. This problem is not only associated with FSW but is also associated with many other welding processes, such as laser welding, diffusion welding, resistance spot welding, etc. [53,54]. Weld joints of Al and Mg have hard and brittle IMCs that promote microcracks and stress concentrations. IMCs formation is unavoidable and nearly impossible to prevent; hence, the best approach is to minimize such compounds. Different researchers have proposed various theories for the evolution of IMCs in dissimilar FSW of Al & Mg. Still, the central ideas discussed in multiple works of literature to clarify the formation of IMCs are mainly based on diffusion bonding, mechanical/material interlocking, and constitutional liquation or eutectic reaction mechanisms [55, 56, 57]. The primary discussion centers on whether the peak temperature during FSW welding surpasses Al-Mg eutectic lines of 450°C (Aluminium dominant side) and 437°C (Magnesium dominant side). Yamamoto et al. attributed the structure and growth of such mixtures to a diffusion phenomenon caused by the temperatures below the eutectic line [58].

On the other hand, the majority of researchers believe that the eutectic reaction is the most crucial element in the creation of IMCs between Al-Mg FSW. Constitutional liquation is considered to be the leading cause of excessive IMCs and the weakening of Al-Mg welds[59]. Firouzdor et al. noted constitutional liquation as the primary cause of solidification cracks in welds [56]. Figure 2.4 presents the eutectic microstructure in Al1050 and AZ31weld [60].

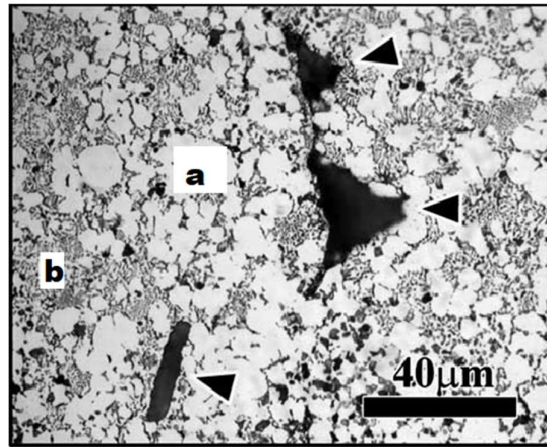


Figure 2.4: Eutectic microstructure in Al1050 and AZ31 welded joints [60].

Because the eutectic temperature is substantially below the melting points of both Al and Mg, it is most probable that the temperature during frictional heat frequently passes the limiting line, triggering localized melting that solidifies into layers of such impermeable compounds.

2.2.2 Material Interlocking and Al-Mg Joint Interface

Material/Mechanical interlocking is considered one of the most effective methods for enhancing the weld strength of dissimilar materials using the FSW technique [61]. The joint interface primarily determines the complex material flow behavior. When both Aluminum and Magnesium combine in solid form in the stir zone, mechanical interlocking occurs as shown in Figure 2.5 [62].

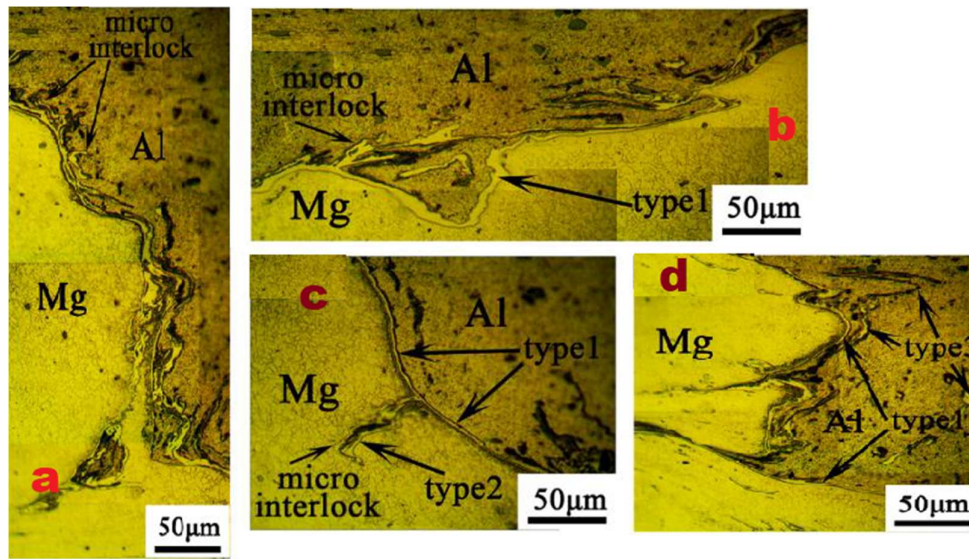


Figure 2.5:(a-d) Micro-interlocks and microstructures obtained during dissimilar FSW (at 50mm/min and 700rpm) [62].

Mechanical interlocking is a bonding process that enhances mechanical strength by producing sophisticated geometries at the bonded surface. The mechanical interlocking phenomenon during FSW improves weld strength [56]. Creating these material interlocks between the interpenetrating layers is essential and advantageous for joint stability. A familiar heat source is required to ensure a low temperature and inhibit IMC growth. Due to material mixing and stress dispersion in the nugget zone, the presence of IMCs at such interfaces can also promote bonding. Tool design and workpiece positioning are two criteria that must be considered to provide a better weld through mechanical/material interlocking [56,58]. Based on the preceding reviews, arguments, and discussions, a perfect joint must have thin and inter-layered IMCs with complicated mechanical/material interlocking. A thin, brittle IMC layer reduces the occurrence of flaws at the weld interface, and a complex mechanical bonding promotes better flow stress circulation. According to Lv et al., the creation of IMCs results in the establishment of a separate layer at the weld junction. Based on the position of IMCs in the vicinity of the Al-Mg weld interface, they are classified into the three main types: (i) Type I: coherent IMCs layer along with the Al-Mg joint interface, (ii) Type II: fragments of IMCs located in Mg, and (iii) Type III: fragments of IMCs inserted into Al [62]. In this literature review, an attempt is made to compare the wide variety of weld interfaces with process parameters. However, due to the intricacy of microstructures in

the nugget zone, difficulty arises in determining the total strength in the presence of interpenetrating layers and mechanical interlocking .

2.3 Welding Parameters and Their Influence on FSW of Al-Mg

Welding parameters play a crucial role in producing high-quality, zero-defect welds. Some parameters contribute to the quality of welds, while others have a negligible effect on the FSW of dissimilar Al-Mg. To achieve superior-quality weld joints, understanding the parametric influences during FSW is essential.

2.3.1 Effect of Base Metal Positioning on Al-Mg Weld

The positioning of workpieces on the tool plays a critical role during the FSW of Al & Mg. It helps determine the formation, material flow, and mixing of IMCs during FSW. Base metal positioning is of two types: AS (Advancing side) and RS (Retreating side). When the tangential direction of the rotating tool and weld feed directions are the same, this position is referred to as the Advancing side (AS); on the other hand, if both directions are opposite to each other, this position is known as the Retreating side (RS) [56]. Due to incorrect workpiece alignment during FSW, abnormalities in material flow occur, resulting in voids and tunnel defects. Keeping the Mg plate at the top and the Al plate at the bottom of the FSW lap joint is always recommended to achieve adequate mixing and high weld quality. Because Mg has a lower thermal conductivity than Al, it allows for improved material flow and heat dissipation in the nugget zone [56].

In various similar materials, butt joint FSW, it was observed that maximum heating occurs on the advancing side (AS) compared to the retreating side [32, 65, 66]. As the stirring process occurs, the material on the advancing side experiences greater shearing and heating than the retreating side (RS) [56]. Zettler et al.[63] and Firouzdor et al.[56] claimed that aluminum (Al) alloys of the 6xxx series produce higher temperatures than those of Magnesium (Mg) counterparts. Gerlich et al. also claimed similar results [64].

Facts obtained from the literature above clearly show that the position of material (either AS or RS) in FSW is responsible for producing a variety of mechanical and

microstructural properties. Furthermore, higher heat generation is expected when aluminum is placed on the Advancing side. In contrast, higher shearing and stirring activities are expected on the AS; hence, placing the magnesium alloys on this side would be a better option, as their liquation tendency is significantly higher compared to that of Al alloys [56, 3]. Sato et al. successfully fabricated an FSW butt joint of 1050 (Al alloy) and AZ31 (Mg alloy) [60] by placing the Mg alloy AZ31 on the advancing side and the Al alloy 1050 on the retreating side; however, changing the positions of both alloys resulted in weld failure. Along with Zettler et al. [63], many other researchers have also supported the same material positioning as suggested by Sato et al. for successful fabrication [65, 66]. However, many other available reports suggested placing Al alloy on the Advancing side and Mg alloy on the Retreating side for better fabrication and optimal metallurgical properties [53, 1]. Such an arrangement promotes higher welding process temperatures and enables effective plasticization [1]. If someone wants to use specific Aluminum alloys, it would be better to use the 2xxx and 5xxx aluminum series on the advancing side and the 1xxx and 6xxx aluminum series on the retreating side, because the 1xxx and 6xxx series possess lower flow stress compared to the 2xxx and 5xxx series, even at rising or higher temperatures [57,63].

2.3.2 Effect of Rotational Speed & Welding Speed (Traverse) on Al-Mg Weld

The most critical and necessary FSW parameters are the rotational tool speed, ω (rpm), and welding speed or tool linear speed, v (mm/sec), as these parameters significantly impact material input and heat flow. Higher rotating speeds result in higher heat input. Still, higher welding rates or welding speeds result in lower heat input because heat input is directly proportional to rotational speed and inversely proportional to traverse speed. To achieve the best weld quality, optimization of these parameters is necessary. Rotational tool speed affects the fracture location of the weld by widening the strained area and shifting the largest strain zone to the advancing side (AS) from the original retreating side (RS) [67]. Many researchers have demonstrated the phrase “Revolutionary Pitch (RP)” as one of the primary and critical parameters for generating quality welds, based on the compounding influence of both these parameters. The ratio of welding speed and rotational speed (v/ω) is the revolutionary pitch. A high “RP” represents fast welding (cold welding), while a lower “RP” represents slow welding

(hot welding). Higher pitch ratios are responsible for lower peak temperatures and insufficient material flow. In contrast, too low pitch ratios are responsible for poor material flow and higher liquation. It further encourages the formation of IMCs and deteriorates weld quality [38]. When the rotational speed is low, it promotes low-temperature generation in the nugget zone, i.e., the strengthening of particles is increased, as shown in Figures 2.6(a) and (b). When this rotational speed is high, it leads to particle separation in the TMAZ, as shown in Figures 2.6(c) and (d) [67].

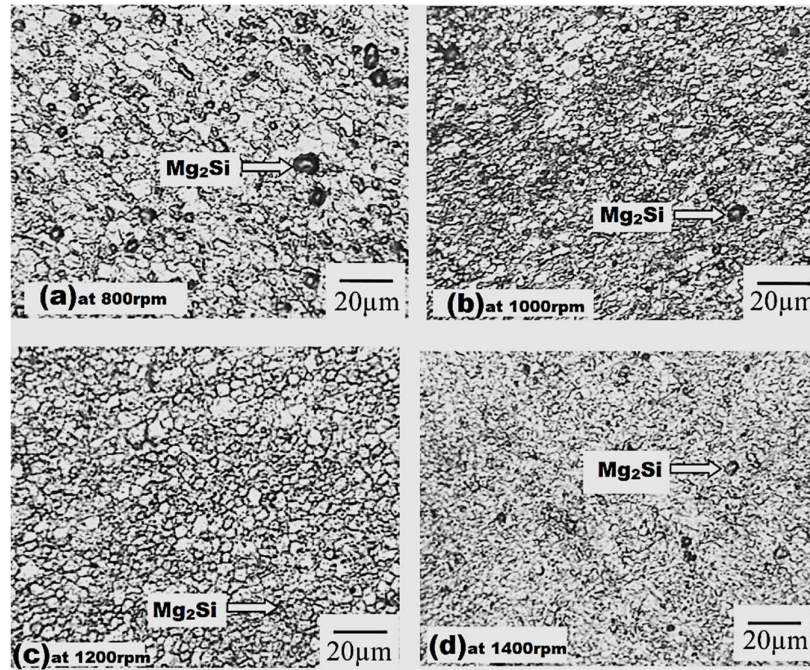


Figure 2.6(a-d) Microstructural variation in FSWed zones at varying TRS [67].

According to different research articles described here, the “RP” ranging from 0.02 to 0.38mm/rev is considered favorable for dissimilar FSW of Al & Mg. The parameter that defines the joint quality is known as Joint efficiency. Joint efficiency is the fraction of the ultimate tensile strength (UTS) and base metal UTS with the lowest value (i.e., higher efficiency) in dissimilar welding. Rotational speed and traverse speed also affect the macrostructure and weld appearance in dissimilar Al-Mg FSW [3]. Fu B et al. further claimed that to obtain a quality and defect-free weld, the rotational speed and welding speed should be between 600rpm and 800rpm, and between 30mm/min and 60mm/min, respectively. Any value outside the given range results in defect formation,

such as cavity formation (an area less than 0.02 mm²) and tunnel formation (an area greater than 0.05 mm²), as shown in Figure 2.7 [3].

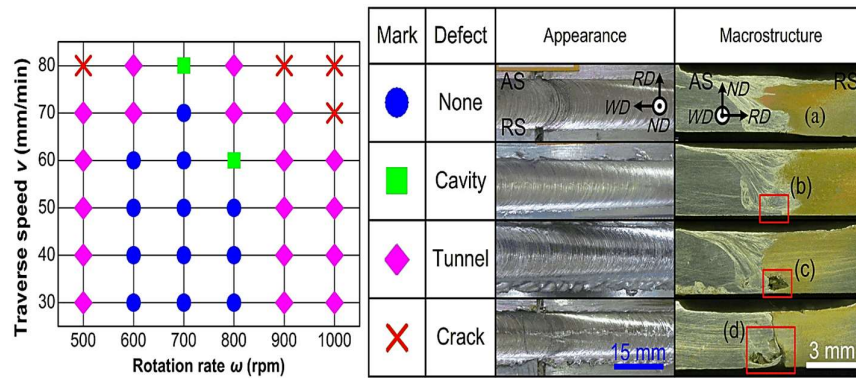


Figure 2.7: Defect, weld appearance, and macrostructure of dissimilar Al-Mg FSW joints at varying TRS and TS [3].

The Ultimate tensile strength (UTS) of the welds increases with increasing welding speed (at constant rotational speed), as shown in Figure 2.8 (a,b)[3,68]. Furthermore, Yang et al. reported an increase in elongation percentage with increasing traverse speed, as shown in Figure 2.8(b) [68]. Elongation takes place due to the formation of IMCs and heat generation. The generated heat is inversely proportional to welding speed; hence, at higher welding speeds, a lesser amount of heat is produced. As a result, the formation of IMCs is significantly reduced, thereby making the SZ less brittle [67-69].

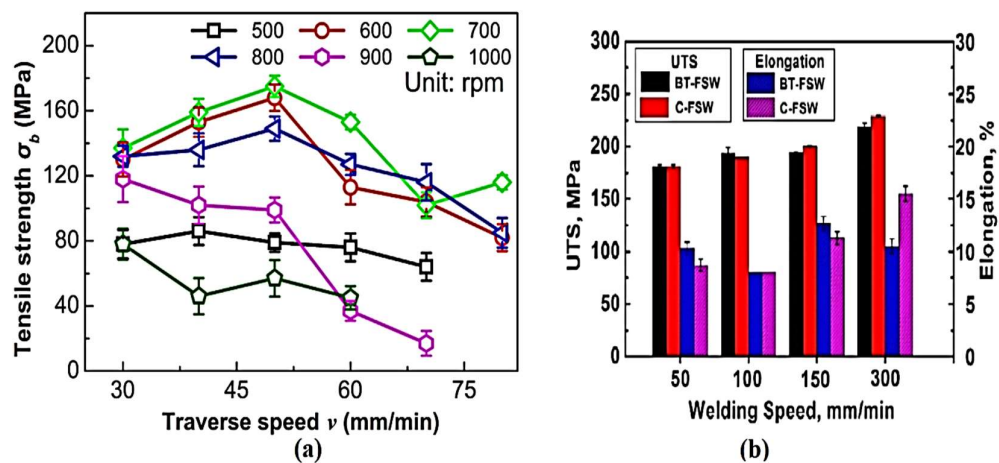


Figure 2.8: (a) Variations in tensile strength with increasing TS (at constant TRS) [3], (b) percentage elongation and variation in UTS with increasing TS [68].

2.3.3 Effect of Tool: Design, Material, Geometry, Positioning, or Offset & Tilt on Al-Mg Weld

Tool design, geometry & tool positioning are also responsible for producing quality welds, unlike other parameters. The most common tool material used for Al-Mg dissimilar welding is H13 steel. The majority of research work highlighted here used H13 Steel. A few other tool materials reported include high-strength steel, SKD51 Steel, an H13 steel shoulder with an MP159 cobalt-base superalloy pin, and SKD61 Steel. As far as tool geometry is concerned, threaded and cylindrical shapes are utilized on a large scale. Some research work producing type 3 interfaces used a pin with flutes or threads. During FSW, threads and flutes help improve material flow behavior and also increase the heat generation rate, as their surface area is significantly larger [3, 56, 63, 66].

Different types of FSW tools are used for various purposes. Fixed-type FSW tools are used for uniform-thickness workpieces, whereas adjustable-type tools are used for varying pin lengths during welding. Its two components are the pin and the shoulder. Bobbin-type or self-reacting tools are used when processed zones in the workpiece are required during welding. This tool consists of a pin, top shoulder, and bottom shoulder. Different types of tool shoulders and tool pins, along with their corresponding geometries, are illustrated in Figures 1.3 and 1.4. The shoulder geometry helps to improve the microstructural and mechanical properties of the weld. The shoulder effect pushes the material very close to the top surface towards AS. The effects of tool rotational speed and tool profile on UTS, % elongation, and joint efficiency are shown in Figures 2.9, 2.10, and 2.11, respectively [67]. Furthermore, variation in microhardness was observed at different TRS with the same tool profile (Square pin), as shown in Figure 2.12 [67].

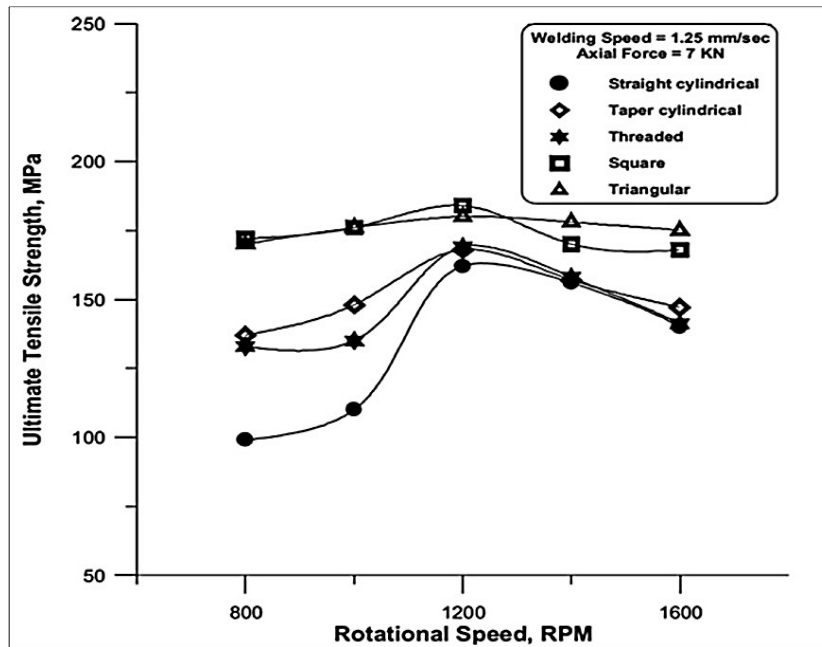


Figure 2.9: Effect of TRS and tool profile on Tensile strength [67]

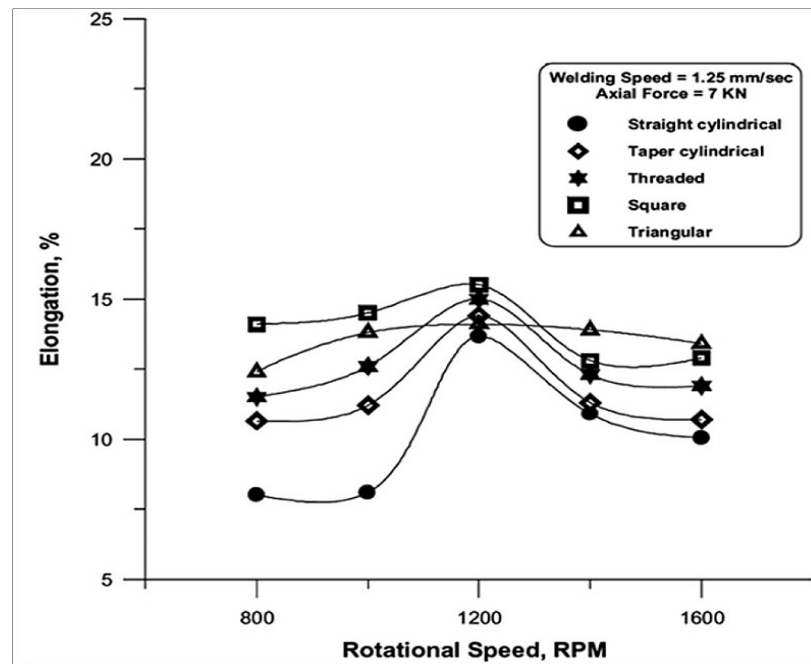


Figure 2.10: Effect of TRS and tool profile on % elongation [67]

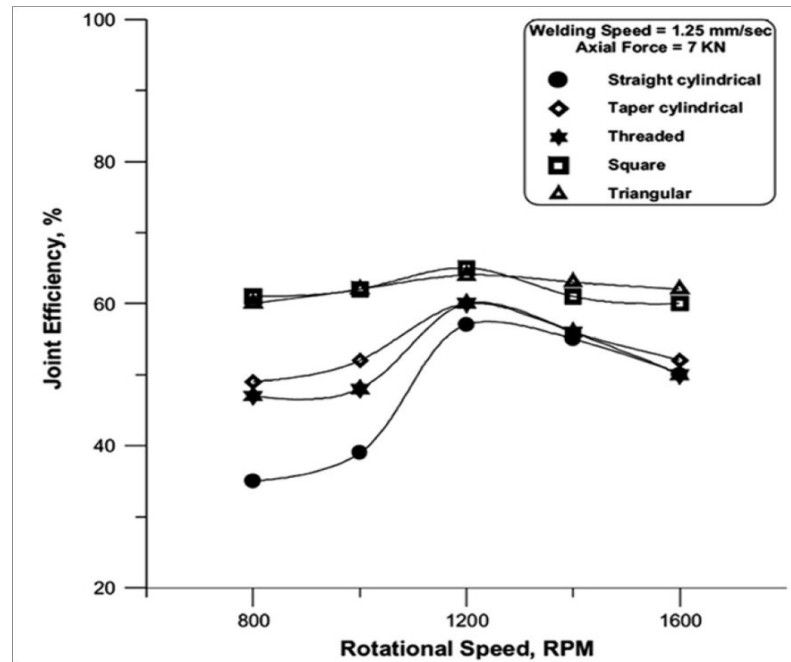


Figure 2.11: Effect of TRS and tool profile on Joint efficiency [67]

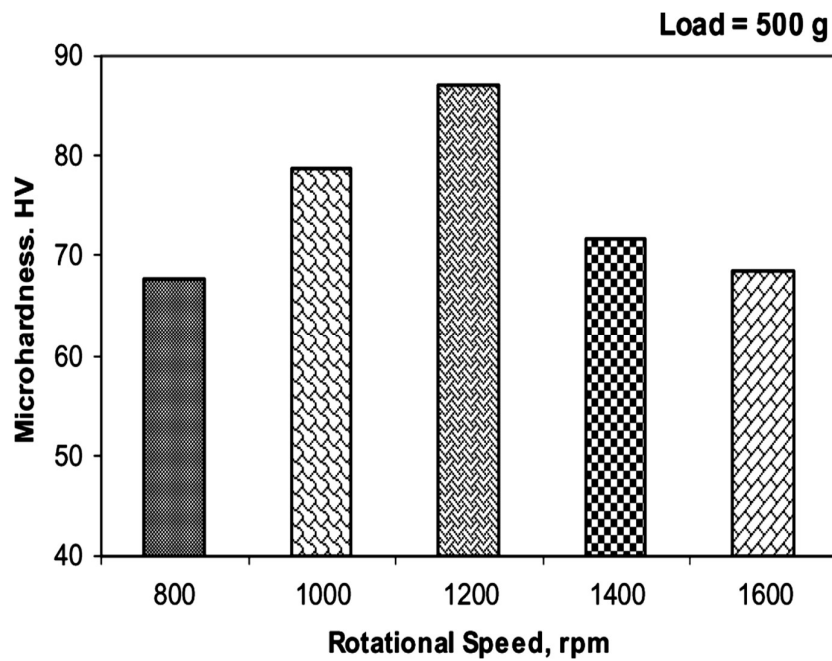


Figure 2.12: Variation in the microhardness at different TRS having the same tool profile (Square pin) [67]

Tool shoulder geometry plays a significant role in welding because it alone contributes to 88%(approximately) of total heat accumulation during the stirring process [17,18]. Material flow, stir zone size, and microstructural behavior are directly affected by the pin or probe configuration during FSW. It shears the material in the tool's to-and-fro direction, resulting in plastic deformation. Various studies have shown a slight variation in pin length, suggesting that there should be a minimal difference ($\Delta t \leq 0.5\text{mm}$) between pin length and workpiece thickness. But, at the same time, Yan et al. [59] and McLean et al. [57] used a larger Δt of 1mm and 1.4mm, respectively. McLean et al. proved in their work that a vast difference leads to premature weld failure [57]. However, Yan et al. [59] achieved a joint efficiency of 67%.

Many studies have claimed that tool offset is a significant consideration in the FSW of Al & Mg. High-strength alloys require a higher heat input during FSW; therefore, a correct offset is essential during such welding procedures to achieve better, defect-free joints. The tool shifting from the centreline of the workpiece either towards the advancing side or the retreating side is called tool offset. If the position of the pin is perfectly aligned with the weld centerline, there is no tool offset. In FSW, more heat is required to plasticize the high-strength alloys. As a result, an appropriate offset is required during dissimilar material FSW. Because various thermal conductivities of materials cause heat dispersion to be unbalanced in the case of dissimilar material production, it can be balanced by giving an offset. The tool will successfully agitate both materials if a substantial amount of tool offset is applied to the softer material (i.e., towards the more rigid material). During the welding of different materials, an optimal and acceptable tool offset becomes critical. Tool pin offsets contribute to higher-quality joints during FSW by reducing material flow behavior and the volume fraction of IMCs. The pin offset further results in heat distribution in the weld zone. The composition of the sample, its thickness, and the tool's design all play a crucial role in determining the optimal pin offset value. Therefore, selecting the best tool offset value during FSW of dissimilar Al and Mg is necessary.

Traversing the tool along the faying surface centered between two metals is the most common approach in tool positioning. [57,63]. Yamamoto et al. reported this position as the most favorable because offsetting the tool either towards the Mg side or towards the Al side would result in internal cavities and surface flash defects [58]. Most studies

reported tunnel defects and kissing bond defects as the most common defects, due to very little heat input. An optimum pin offset value is responsible for generating the necessary heat in the weld zone, and it primarily depends on the tool design, composition, and thickness. The tool shoulder holds the stirred material via a threaded pin and efficiently moves the material from the front to the back of the pin when the spindle is tilted in the trailing direction [3]. Most studies have described the effect of the tool tilt angle on material movement, with an angle ranging from 0 to 3 degrees [3]. Meshram et al. found that certain defects were absent between 0° to 0.5° , and surface defects were present beyond 0.5° tilt angle. However, they found internal kissing bond defects in the range of 2.5° to 3° [70]. However, other claimed that with increasing tilt angle, the hardness value at the weld zone also increases, which gives rise to IMCs formation due to rising temperatures [71,72].

Furthermore, Shah et al. described the effect of the tool tilt angle on fracture mode during tensile testing and found that the best mechanical properties were obtained at a tilt angle of 2° [73]. Hence, for Al-Mg dissimilar FSW, an optimum tool offset and the proper tilt angle are essential. Srinivasa et al. reported that weld strength in FSW depends on pin configurations, rotating tool speed, and feed rate. Therefore, to obtain a defect-free weld, these parameters must not be ignored [74].

2.4 Welding Parameters' Effect on Power and Torque

Firouzdor et al. stated that (as shown in Figure 2.13a,b), significantly very high torque and power are produced while fabricating similar Al 6061 by the FSW technique as compared to that of similar magnesium AZ31 under identical conditions (i.e., at similar parameters) [56]. On the other hand, fabricating different Al 6061 and Mg AZ31 using the same technique resulted in increased power and torque with increasing travel speed, as shown in Figure 2.14a and b. When Al 6061 was positioned on the advancing side, both power and torque were much higher, indicating an increase in heat input. Furthermore, the reduced torque and power in welding Al-Mg resulted in decreased heat generation. This supports the earlier-described liquation behavior in the Al-Mg eutectic reaction, which produces tool slippage and represents the higher amplitude of the 'slip-stick' process during Al-Mg FSW [55,56].

The parametric effect of different FSW of Al alloy 6063-T5 and Mg alloy AZ31B-H24 on power and torque was reported by Venkateswaran et al. [61]. Power was calculated by multiplying the spindle torque by the rotational speed. An increase in tool rotational speed from 900 to 2700 rpm decreased torque, indicating lesser resistance to material flow at higher rates. This clearly shows that tool rotation is directly related to power and torque.

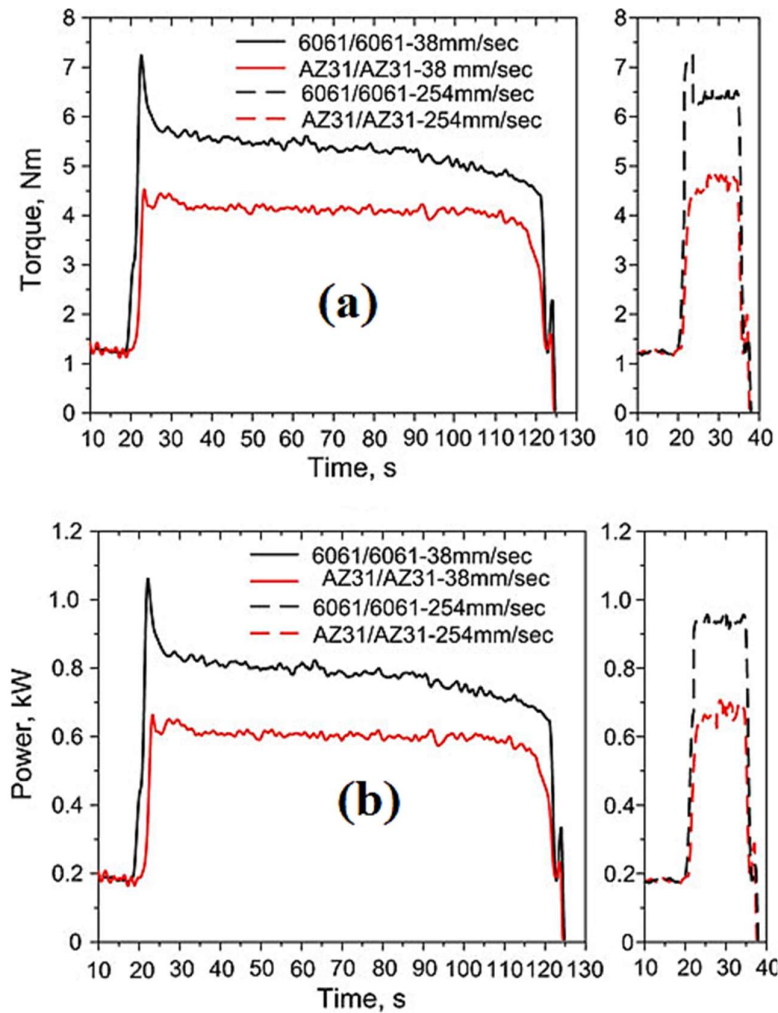


Figure 2.13. (a) Torque-time and (b) power-time curves for similar FSW of AA6061 and AZ31B alloys at different TS and a TRS of 1400 rpm [56].

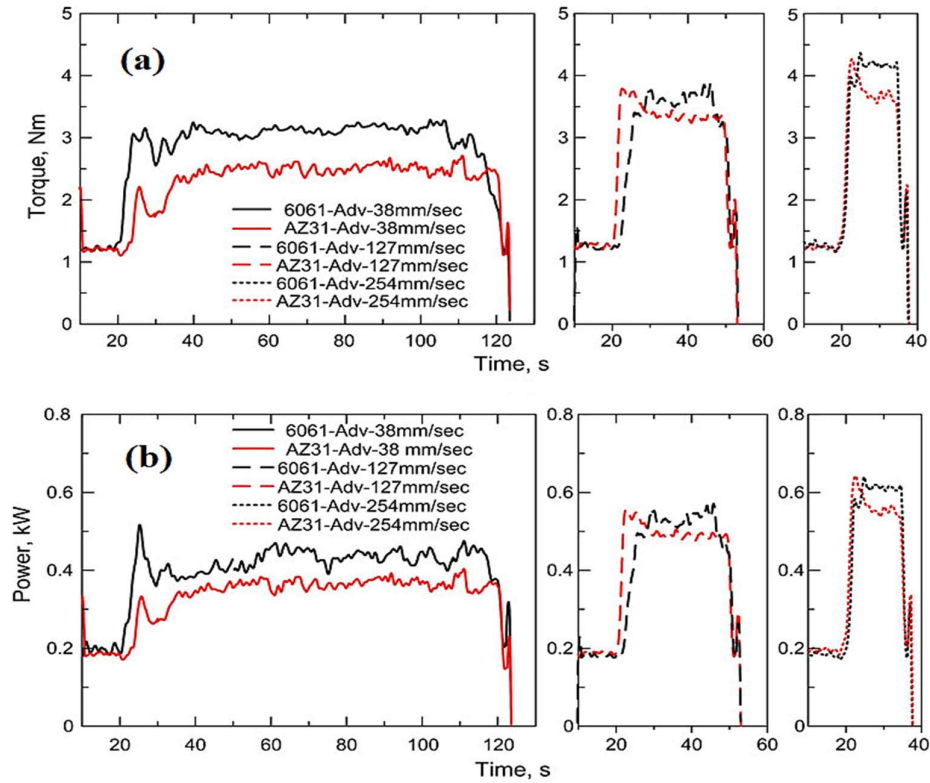


Figure 2.14. ((a) Torque-time and (b) power-time curves for dissimilar FSW of AA6061-AZ31B alloys at different travel speeds and a rotational speed of 1400 rpm [56].

2.5 Peak Temperature Effect and Cooling Effect on Friction Stir Welding

Many studies have reported the effects of peak temperature and cooling in various ways. The peak temperatures are measured using several types of thermocouples, which are either located inside the tool pin or inserted into the sample. Azizieh et al. [75] used a thermocouple (K-type) at the bottom side of the job piece to measure peak temperatures. The best strength was reported at the eutectic temperature or peak temperature, between 430°C and 460°C. Further, they reported the effect of (a) peak temperature on grain size in the stir zone, (b) IMCs, and (c) hardness of the weld joint, as shown in Figure 2.15 [75]. Firouzdor et al. stated that to obtain higher peak temperatures, the Al must be placed at the advancing side and offsetting the tool towards the Al side [56].

Mofid et al. reported using external water-cooling methods during FSW of Al 5083 and Mg AZ31C-O to reduce the peak temperature and thereby achieve better weld quality [1]. Zhao et al. found similar results when an underwater Friction stir welding (UFSW) method was developed and implemented on Mg alloy AZ31 and Al alloy 6013 [76]. During UFSW, the cooling rate is faster than in air because the specific heat of water is roughly four times that of air. A faster cooling rate suggests that grain expansion in the heat-affected and weld nugget zone is constrained. It's also worth noting that the rate at which the weld cools varies from surface to surface, depending upon the thickness of the weld. According to Zettler et al. [63], the bottom portion of Al 6040 and Mg AZ31 alloy manufactured via FSW has a higher hardness value.

Feedback control is another technique that can control the peak temperature. Ross and Sorensen proposed this technique. An integrated thermostat guarantees that the necessary temperature is maintained by controlling the tool rotation speed in real-time. It maintains a constant temperature, reduces weld property fluctuations throughout the joint contact, extends tool life, and improves repeatability. Feedback control is tremendously advantageous, but it is also complicated and expensive. According to the prior literature, relatively few investigations have been conducted on temperature measurement systems for Al-Mg Friction stir welding. As a result, research on FSW of dissimilar Al & Mg becomes a fascinating and much-needed topic.

2.6 Mechanical & Microstructural Characterization in FSW of Al-Mg

The Base Material (BM), heat-affected zone (HAZ), thermomechanically affected zone (TMAZ), and stir zone (SZ), also known as the nugget zone (NZ), are formed by plastic deformation and friction heating caused by the rotating and stirring impact of FSW tools. The quality and characteristics of the welded junction are determined by the microstructural features of each weld zone. The weld nugget zone and its surroundings are the subjects of microstructural characterization. Dynamic recrystallization (DRX) of Al and Mg grains has been observed in the nugget zone, where a reduction in grain size below 18 μ m compared to the parent metal is often observed due to the tool's stirring action [63,75]. Grain size refinement and the existence of IMCs increase the hardness and decrease the ductility in the welding zone compared to the workpiece.

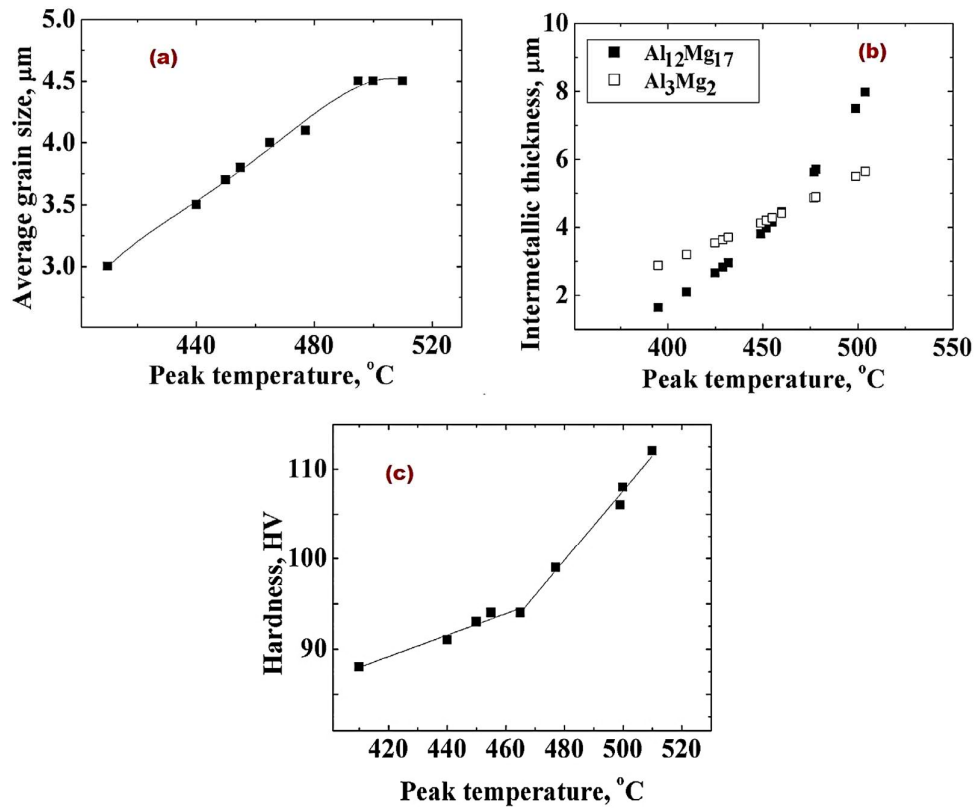


Figure 2.15. Effects of peak temperature on: (a) grain size in SZ, (b) IMCs, & (c) weld joint hardness [75].

2.6.1 IMCs Observation

IMCs occur either at the joint interface in the weld nugget zone or at lamellar shear bands. IMCs so formed are brittle, and their thickness ranges from 1 μm to 3 μm [3]. However, many studies have claimed that welds with IMCs between 1.5 μm and 10 μm produce good, defect-free welds. The development of nano-sized grains of the Al_3Mg_2 phase in immediate proximity to the $\text{Al}_{12}\text{Mg}_{17}$ IMC phase layer shows the presence of both IMCs' brittle phases. Many other studies also claimed the same. X-ray diffraction (XRD) patterns (see Figures 2.16 and 2.17) obtained from the joint surfaces of the Mg alloy and Al alloy clearly show the existence of Al_3Mg_2 and $\text{Al}_{12}\text{Mg}_{17}$ compounds at these surfaces, which adversely affect the joint strength [58,77].

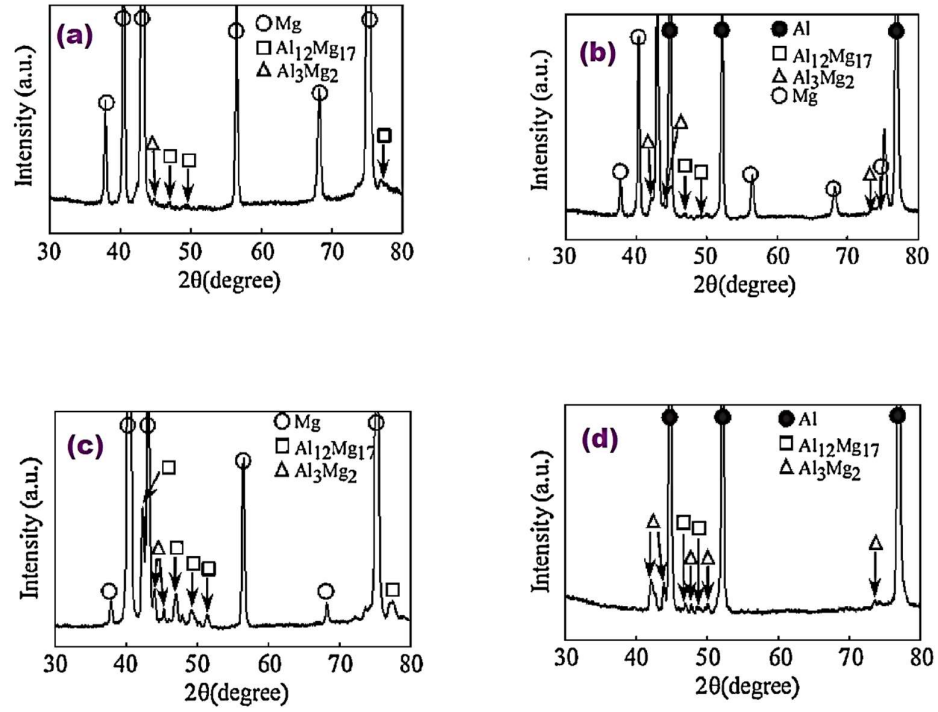


Figure 2.16: XRD patterns obtained from the Mg (RCPAZ31) and Al (A5083) sides at : (a,b) 0 mm offset, (c,d) 1.5 mm offset [58].

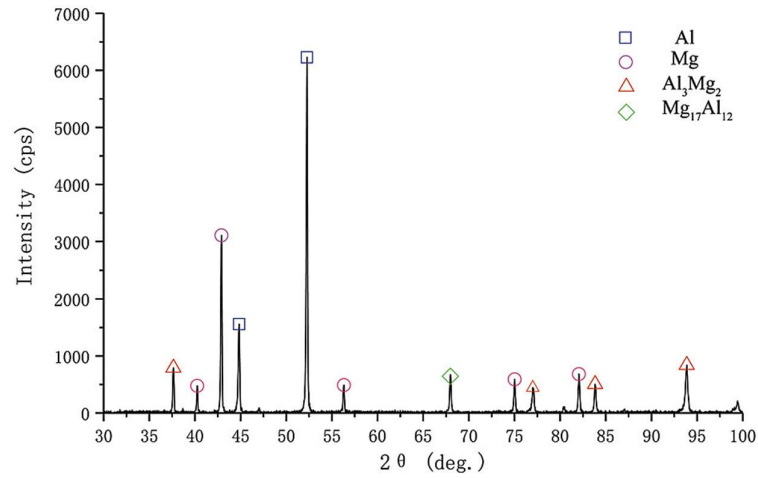


Figure 2.17: XRD patterns of 5083 and AM60B, showing the availability of Al_3Mg_2 , $Al_{12}Mg_{17}$, Al, and Mg [77].

However, compared to the Al_3Mg_2 phase, the $\text{Al}_{12}\text{Mg}_{17}$ phase is more prevalent. This contradicts the earlier assertion by Yamamoto et al. [58]; however, research using energy-dispersive X-ray spectroscopy demonstrates that when Al-Mg FSW joints are cooled, they undergo constitutional liquation, resulting in the formation of intermetallic phases (IMCs) [55, 63, 78]. Azizieh et al. [75] reported the formation of $\text{Al}_{12}\text{Mg}_{17}$ phase IMCs due to their lower eutectic temperature. Therefore, to prevent the development of IMCs, it is necessary to maintain peak temperatures to some extent, which is a challenging task. As a result, choosing the best set of parameters to solve the IMCs problem is a fascinating and important research topic. Figures 2.18 to 2.21 depict the production of IMCs at the interface of Al and Mg alloys during FSW [2,58,71,79].

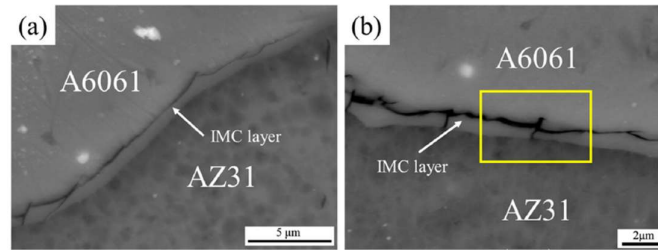


Figure 2.18:(a, b) IMCs at the interface of dissimilar metals [2]

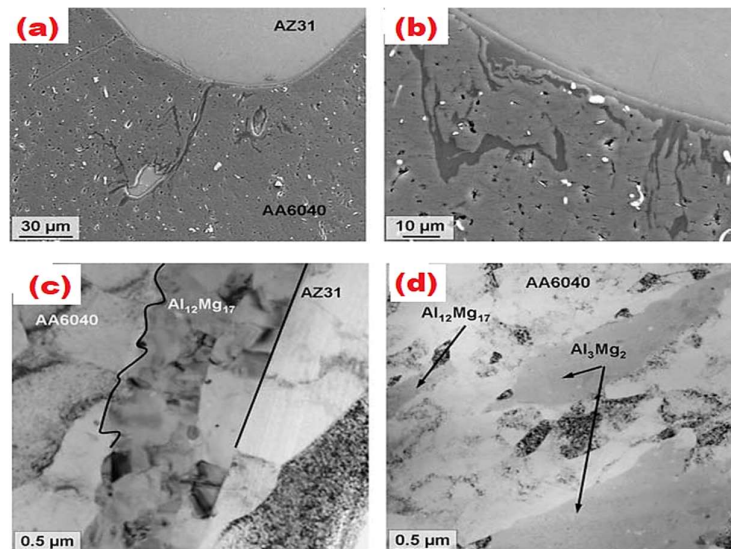


Figure2.19. (a,b) SEM, and (c,d) TEM images showing IMC formation at the AA6040-AZ31 interface [79].

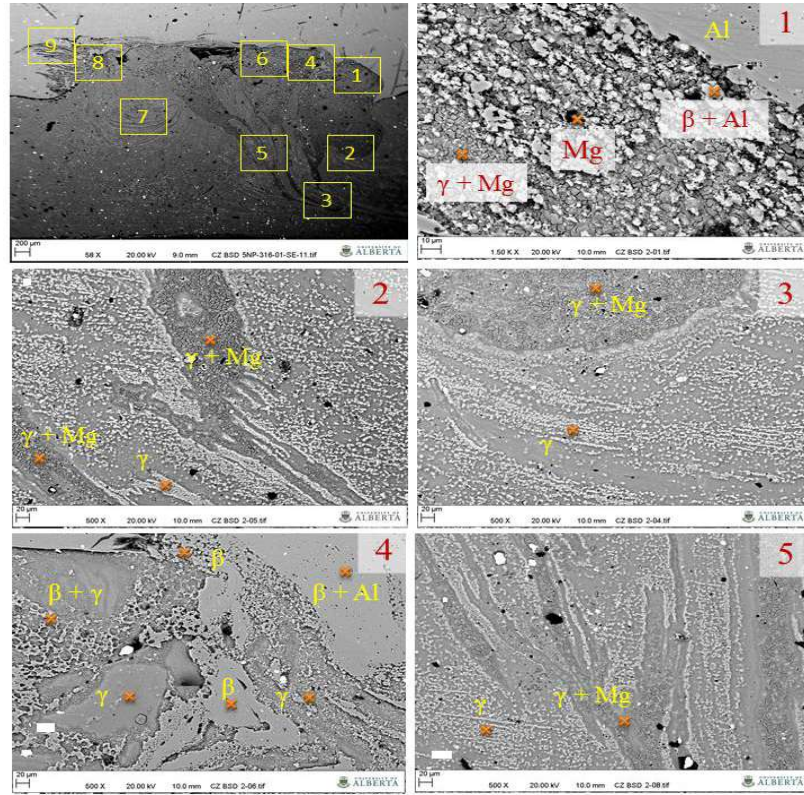


Figure 2.20. SEM images of distinct locations in the Mg-Al lap joint [71].

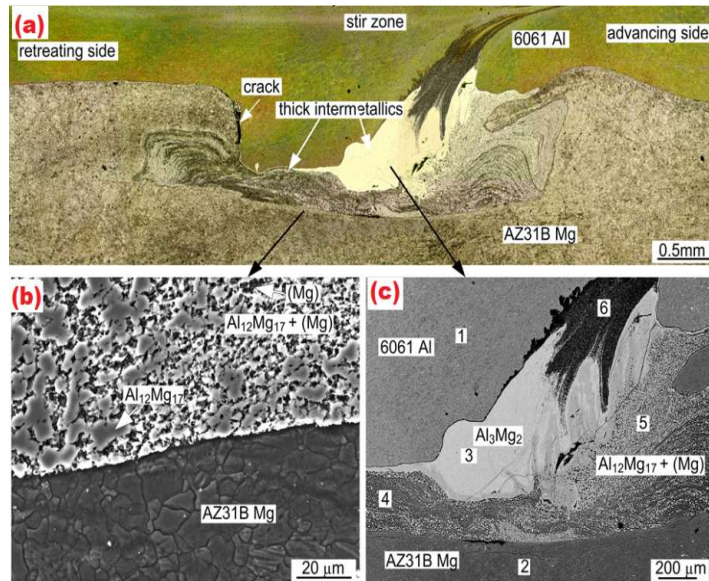


Figure 2.21. (a) Optical micrograph, (b,c) SEM images of IMCs transverse cross-section of a lap weld [58].

2.6.2 Joint Efficiency and Tensile Strength

Mechanical characterization, including hardness and tensile tests, is the most common method for assessing weld quality. Joint efficiency is defined as the ratio of a joint's tensile strength to that of the base metal and is used to describe the quality of a joint. Many studies have reported that optimal welding parameters can be used to achieve higher joint efficiencies, even when workpieces are dissimilar. Fractures near the joint interface, on the other hand, are prevalent, and the fracture surface clearly shows the presence of IMCs, which have a negative impact on joint quality. Furthermore, when compared to base metals, the percentage elongation of the weld is significantly lower at the weld interface, as shown in Figure 2.22 [76].

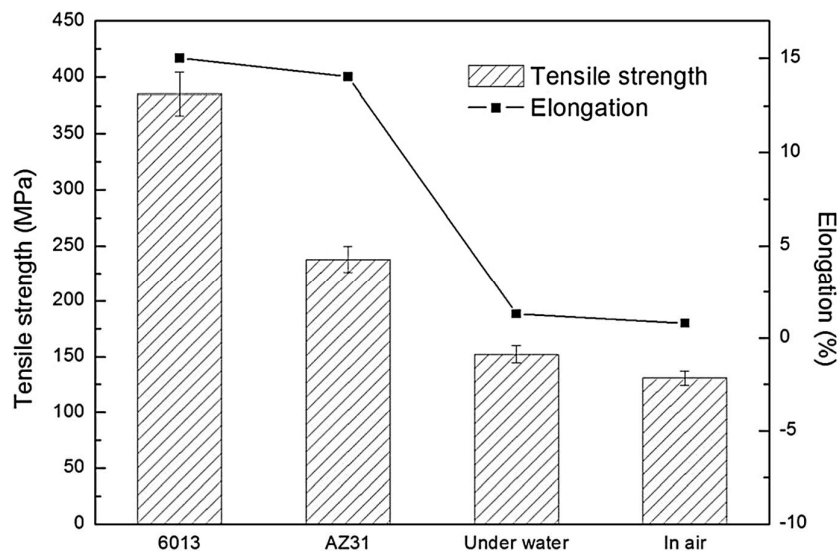


Figure 2.22: Tensile Strength of base metals vs. welded joints [76]

Furthermore, a better elongation was observed as the fracture area shifted from the weld bond contact to the softened HAZ of the aluminum alloy, where necking occurred prior to fracture [66]. This change is unknown; however, macrographs of a sample cross-section reveal a Type 3 weld contact. During the tensile testing, the formation of dimples is very common on the fracture surface. Bai et al. claimed that a brittle fracture pattern with tear ridges occupies most of the fracture surface. They further observed the formation of large and small dimples in certain regions of the fracture surface, as

illustrated in the fractographs (see Figures 2.23 and 2.24). To avoid the formation of such dimples, Bai et al. suggested carrying out Ultrasonic vibration Friction Stir welding instead of Friction stir Welding [80]. Variation in tensile strength and % elongation with Simple FSW and Ultrasonic vibration-assisted FSW at 800 rpm and varying welding speeds is shown in Figure 2.25. Obtaining high tensile strength as a base metal while minimizing the development of IMCs remains a difficult issue and a fascinating area of research.

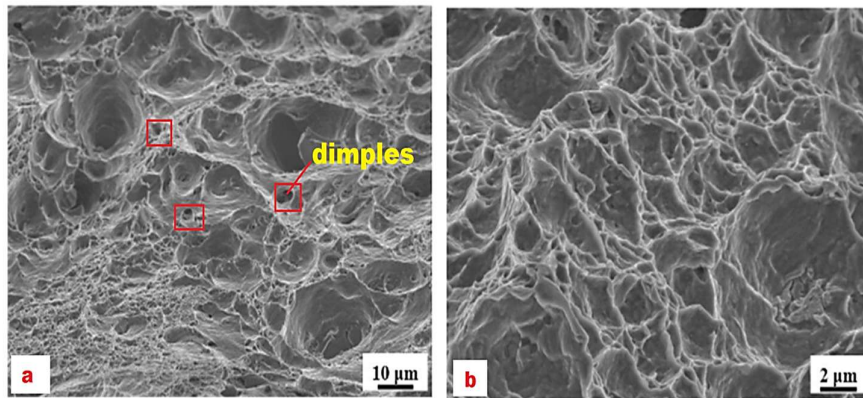


Figure 2.23: SEM fractographs of FSW joints at 800 rpm and 50 mm/min showing dimples [63].

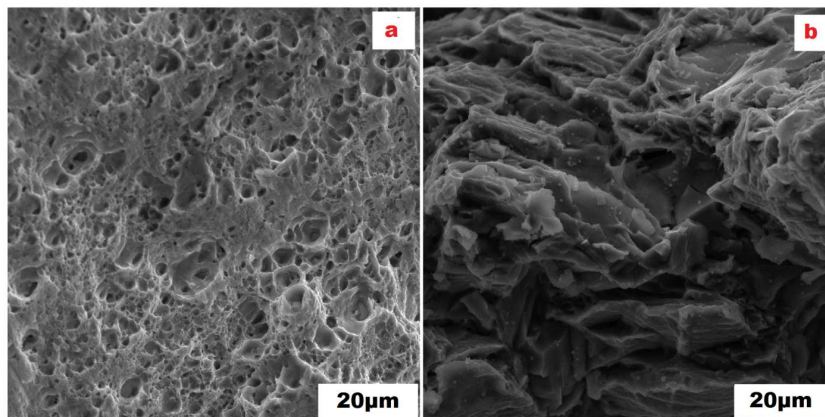


Figure 2.24: SEM Fractographs of fracture surface of tensile specimen (a) Al-Al weld (b) Al-Mg weld (Mg on Advancing side)[66].

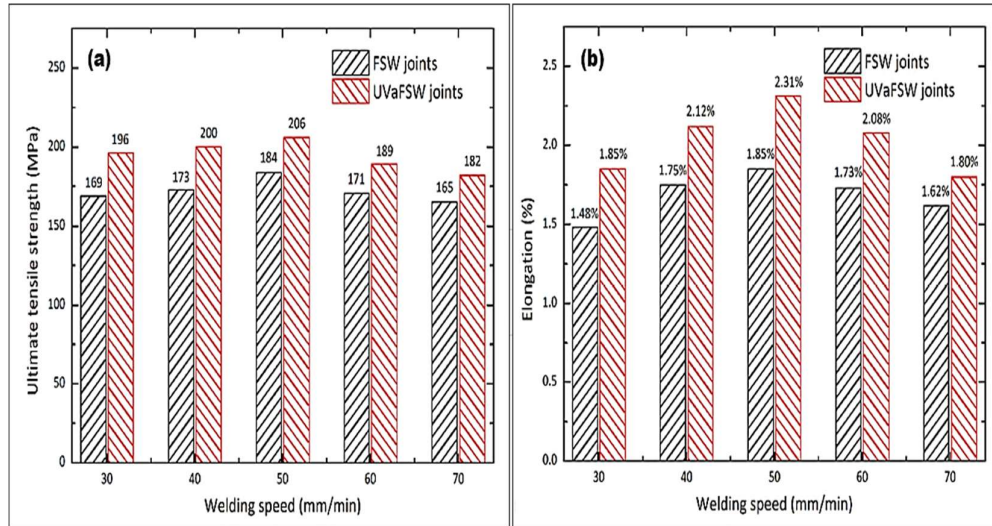


Figure 2.25: Tensile strength and elongation variation in conventional and ultrasonic-assisted FSW at 800 rpm and varying TS [80].

2.6.3 Hardness and Its Variation

The hardness value primarily depends on the strain rate, process temperature, and material flow. The dispersion of magnesium particles in the aluminum matrix is also important. According to numerous research studies, hardness valuations are indeed a comprehensive characterization method for determining the properties of tensile strength, specifically the ultimate and yield tensile strengths, in HAZ. The hardness value substantially increases during the strengthening phenomenon due to the formation of IMCs [81]. Figures 2.26 (a-d) present microhardness profiles of various dissimilar Al-Mg systems, including Al-A383/Mg-AZ91 [82], Al-6013/Mg-AZ31 [83], Al-6063/Mg-AZ31B [61], and AA6061-T4/Mg-AZ31B [62]. The Al-Mg weld line had the highest hardness value, most likely due to mechanical twinning, grain refinement, recrystallization, and solid solution strengthening. The presence of the $Al_{12}Mg_{17}$ phase IMC contributes to the increased hardness. Because nano-sized grains (recrystallized grains) were predominantly found in the weld nugget zone (WNZ), the maximum hardness value was obtained there, as shown in Figure 2.26 (a-d).

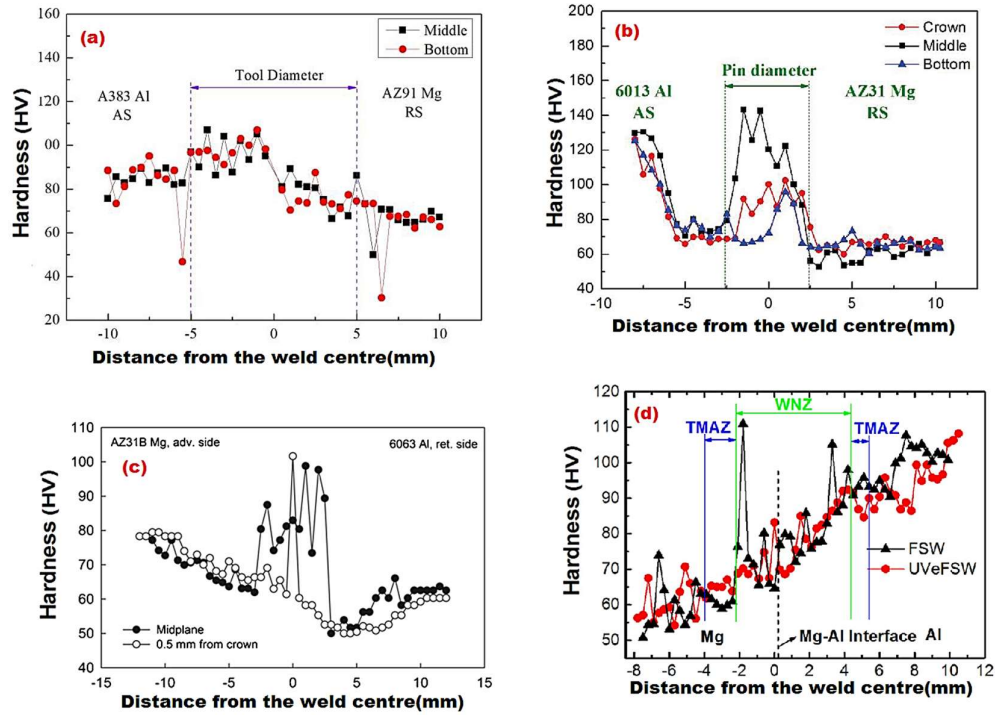


Figure 2.26: Microhardness distribution across dissimilar Al-Mg-type FSW joints [61,62,82,83]

The complex material behavior was observed by Singh et al. [84], which provided the fundamental explanation for the highest microhardness values in the weld nugget zone. The hardness of the weld nugget zone is always higher than that of the base metals due to brittle IMCs and microscopic grain size after severe plastic deformation[14]. The hardness degree of a weld varies not just between weld zones but also from top to bottom along its length. Welding dissimilar alloys using the FSW method is an interesting and challenging research topic that warrants thorough investigation.

2.7 Effect of Nanoparticles on Microstructure and Mechanical Properties in FSW

Furthermore, metals and alloys welded by the FSW technique often fail to meet modern industrial demands due to metallurgical constraints, such as coarse nugget grains, brittle intermetallic compounds (IMCs), and limited fatigue performance in dissimilar joints. These limitations become more pronounced in Al-Mg welding, where strong chemical affinity promotes thick Al_3Mg_2 and $\text{Al}_{12}\text{Mg}_{17}$ layers and asymmetric softening

between the Al and Mg sides, even when macro-defects are eliminated. To further enhance mechanical and metallurgical properties, researchers have increasingly focused on nanoparticle-reinforced FSW and friction stir processing (FSP). These approaches aim to deliberately incorporate nanoparticles into the stir zone (SZ) to enhance grain refinement, inhibit IMC growth, and improve strength, toughness, and wear resistance beyond what is achievable with unreinforced joints. Nanoparticle-reinforced FSW thus converts the weld nugget into an in-situ metal matrix composite (MMC), in which the microstructure and properties are controlled not only by the thermal cycle and deformation history but also by particle matrix interactions at multiple length scales.

Several types of nanoparticles, including SiC, Al₂O₃, and TiO₂, have been employed as reinforcements [27-31]. Their effectiveness stems from mechanisms such as Zener pinning, where nanoparticles hinder grain boundary migration, leading to finer grains and enhanced strength [31]. Their effectiveness stems from several well-established mechanisms:

- Zener pinning, where nanoparticles hinder grain boundary migration, limiting grain growth during and after dynamic recrystallization, and generating ultra-fine, thermally stable grain structures [85].
- Heterogeneous nucleation, where particle-matrix interfaces act as preferential sites for nucleation of new recrystallized grains, increasing nucleation density and refining the nugget microstructure.
- Dispersion strengthening, where hard particles obstruct dislocation motion, increasing yield strength and work-hardening capacity.
- Diffusion blocking, where particles disrupt continuous diffusion paths at dissimilar interfaces, suppressing excessive IMC layer growth and improving fracture resistance.

Figure 2.27 illustrates the pinning of grain growth by reinforcing particles schematically, showing how nanoparticles positioned along grain boundaries generate a retarding pinning pressure that balances the driving pressure for boundary migration, thereby limiting the grain size [86]. This classical Zener mechanism has been experimentally confirmed in multiple nanoparticle-reinforced FSW systems, where

grain size reductions from approximately 50 μm to 2-5 μm have been reported as the nanoparticle volume fraction increases and the particle size decreases [87].

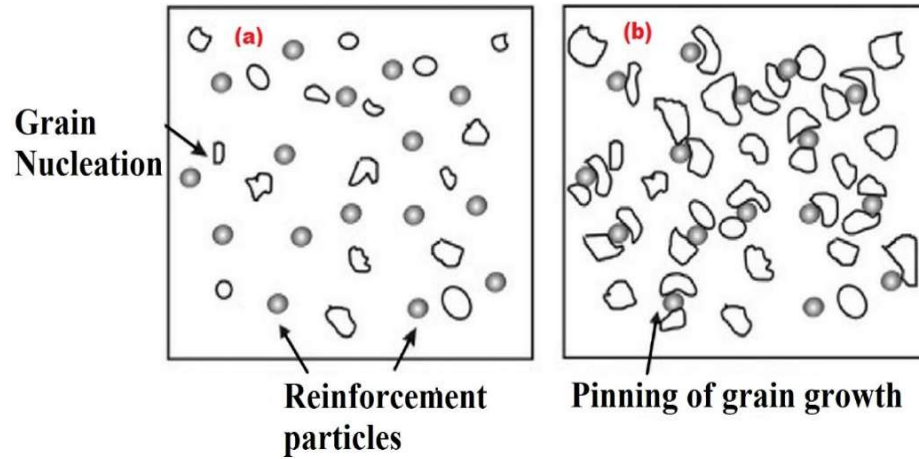


Figure 2.27: Schematic of grain refinement mechanisms due to reinforcement particles [86]

Among available reinforcements, Silicon Carbide (SiC) nanoparticles are especially attractive because of their high hardness, modulus, low thermal expansion, outstanding corrosion resistance, and excellent wear resistance [32]. Numerous studies confirm the beneficial effects of SiC reinforcement in Al- and Mg-based FSW joints. In dissimilar AA7075-AZ31 joints, the addition of nano-SiC in the nugget led to improved integration of the Al and Mg sides and marked grain refinement; however, at higher rotational speeds and lower traverse speeds, extensive Al-Mg IMC growth still limited tensile strength, indicating that nanoparticles cannot fully compensate for excessive heat input [10]. AA7075-O reinforced with SiC using a threaded taper pin at 1250 rpm and 40 mm/min showed notable improvements in fatigue life, tensile strength, and toughness [33].

In AA6351-SiC reinforced joints, optimum conditions near 710 rpm yielded maximum tensile strength, as grain boundary pinning by SiC restricted post-DRX grain coarsening and stabilized a fine-grained nugget structure [34]. AA6061-T6 and AA2024-T351 joints with SiC reinforcement have demonstrated improved particle dispersion, a more

uniform hardness distribution, and enhanced tensile toughness, indicating that nanoparticle addition can mitigate the softening often observed in heat-treatable Al alloys after FSW [35]. The microhardness profile is shown in Figure 2.28.

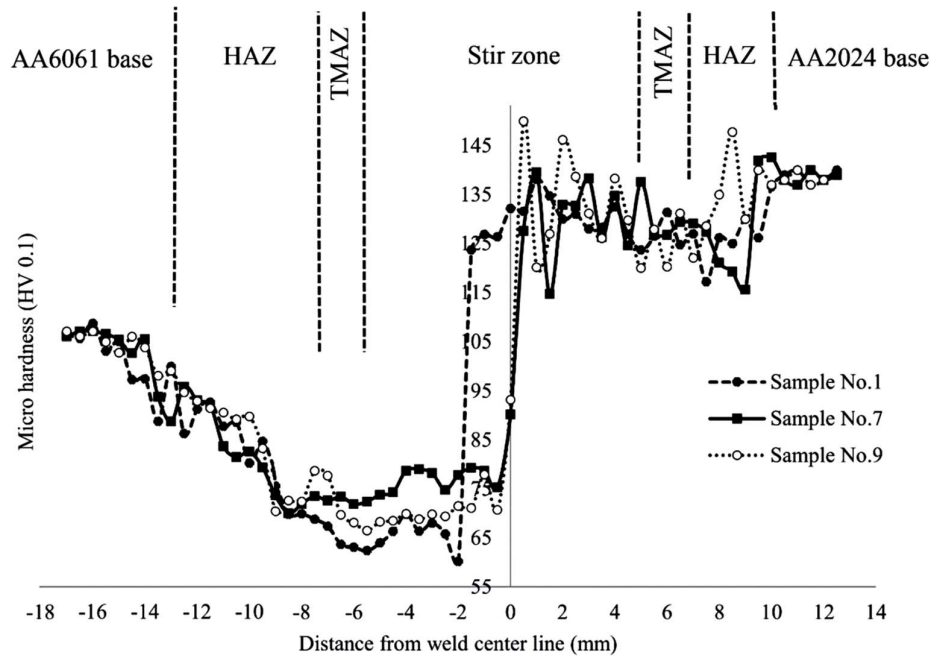


Figure 2.28: The effect of SiC nanoparticles on the microhardness profile [35]

Conversely, Al6061-Mg AZ31B joints (3.6 mm thick) at high rotational speed of 1010 rpm (at 16 and 25 mm/min) caused liquation, creating disbanding on the surface of joints, and resulting in the formation of many surface cracks despite the presence of SiC, highlighting the fundamental requirement to maintain heat input within a safe window, even when nanoparticles are present [36].

Several studies have focused solely on optimizing process parameters for unreinforced Al-Mg systems. For example, AA1100-AZ31 joints welded at 750 rpm and 20 mm/min yielded a tensile strength of 100 MPa with optimal efficiency [37]. AA7075-AZ91D joints achieved a tensile strength of 110 MPa at 800 rpm and 35 mm/min, but lower TRS and TS led to dense, continuous IMC growth and severe strength loss, demonstrating the delicate balance between heat input, IMC morphology, and mechanical performance [38]. These results collectively indicate that while optimized

FSW parameters can yield reasonable properties in unreinforced dissimilar joints, the addition of nanoparticles offers a route to refine the microstructure further and stabilize properties against parameter fluctuations.

In addition to SiC, Al-Mg joints reinforced with TiO₂ nanoparticles have shown that nanoparticle addition can significantly increase tensile strength, reduce grain size, and shift fracture mode from brittle to predominantly ductile by breaking up continuous IMC layers and homogenizing the nugget [30]. AZ31 Mg alloy reinforced with SiC nanoparticles exhibited clear onion-ring structures with alternating particle-rich and particle-lean bands, but with overall improved microstructural uniformity and better mechanical properties compared with unreinforced welds [28]. In AA6082-T6, comparative reinforcement with SiC and TiC showed that SiC predominantly enhanced ductility and energy absorption, while TiC produced higher hardness; both particles promoted defect-free welds, uniform particle distribution, and pronounced microstructural refinement driven by the combined effects of DRX and nanoparticle pinning [39]. AA2024-AA6061 joints with SiC reinforcement exhibited reduced grain size and enhanced toughness [26]. AZ31 Mg and 7075 Al dissimilar joints processed with SiC nanoparticles showed uniform MMC formation in the weld nugget [10]. In addition to strengthening, SiC particles also influence fracture behavior, with many studies reporting a transition from brittle to ductile fracture due to suppression of thick IMCs and improved load transfer [28,30].

The method used to incorporate nanoparticles into the joint plays a critical role in determining final microstructure and properties. Groove-based addition (filling a pre-machined groove along the joint line) generally yields better and more repeatable particle dispersion than surface spraying or pasting, which tend to suffer from non-uniform particle feeding and surface loss during tool engagement. Studies comparing these methods have shown that groove-filled joints exhibit fewer agglomerates, more continuous particle trails within the SZ, and higher joint efficiency than those produced by surface sprinkling of powders [30]. The volume fraction of nanoparticles plays a decisive role in determining joint properties. With Al₂O₃ nanoparticle volume fraction (0.2-0.4 vol%), uniform dispersion and finer grains were obtained [40]. AA6061-T6 FSP joints reinforced with 0.25 and 0.30 Vf SiC showed decreased mechanical properties at higher Vf due to particle clustering [41]. AA7075-AA6061 joints

reinforced with 4-10% SiC displayed significant improvements in tensile strength and joint efficiency [42]. AA2024-AA7075 joints reinforced with 5%, 8%, and 13% SiC were identified as optimal, with homogeneous dispersion and the best weld quality [43].

Silicon carbide nanoparticles incorporated into AZ31B magnesium alloy joints at different stir-zone volume fractions revealed that very high reinforcement levels (15-20% Vf) frequently caused agglomeration, porosity, and a reduction in strength and toughness, despite continued grain refinement. This behavior is consistent with the Zener limit, where, beyond a certain fraction, the detrimental effects of clustering and stress concentration outweigh the benefits of additional pinning and dispersion strengthening. Overall, the literature suggests an optimal window for nanoparticle volume fraction (typically 5-15% for SiC and similar reinforcements in FSW of Al/Mg systems), within which grain refinement, IMC control, and mechanical property enhancement can be maximized while avoiding clustering and excessive tool wear [44].

Despite these advances, no systematic investigation exists on the effect of SiC nanoparticle reinforcement in AZ91D-AA6061-T6 dissimilar FSW joints. Previous work has focused mainly on similar alloys (Al-Al, Mg-Mg), unreinforced dissimilar joints. Existing studies rarely combine a complete statistical design, multi-response optimization, and detailed thermal, microstructural, and mechanical correlation for nanoparticle-reinforced AZ91D-AA6061 joints. To fill this gap, the present study considers SiC volume fractions of 5%, 10%, and 15%, selected based on literature reports and preliminary trials, as a practical range to balance strengthening and avoid clustering.

Furthermore, experimental optimization of parameters such as TRS, TS, and TTA through trial and error is time-consuming and costly, particularly when multiple reinforcement levels and dissimilar materials are involved. To address this, statistical approaches such as the Design of Experiments (DOE) and Response Surface Methodology (RSM) are increasingly employed in welding research to model and optimize complex multi-factor processes [45].

Among these, RSM combined with Central Composite Design (CCD) offers a robust framework for efficiently exploring the parameter space, capturing linear, interaction, and quadratic effects, and building predictive models for key responses such as tensile

strength, hardness, and grain size. Numerous recent FSW studies have demonstrated that RSM-CCD can accurately predict joint strength and hardness, identify optimal combinations of rotational speed, traverse speed, tilt angle, and reinforcement fraction, and significantly reduce the number of required experiments.

RSM enables the construction of response surfaces and contour plots that visualize how responses vary with process parameters, supports the development of regression models that quantify parameter sensitivity, and allows ANOVA-based significance testing to identify the most influential variables and their interactions [46]. Compared to Taguchi designs, which primarily emphasize main effects and are limited in their ability to detect curvature and higher-order interactions in the response surface, RSM provides a more comprehensive understanding of the nonlinear coupling among TRS, TS, TTA, and reinforcement volume fraction. This comprehensive modeling capability is crucial in nanoparticle-reinforced FSW, where thermal, mechanical, and particle-distribution effects interact strongly and nonlinearly [47, 48]. Consequently, the present work combines SiC reinforcement in AZ91D and AA6061-T6 dissimilar joints with an RSM-based CCD design, enabling a systematic evaluation of microstructural refinement, IMC control, and mechanical performance as functions of both process parameters and nanoparticle volume fraction, and providing a statistically validated basis for optimization and industrial implementation.

2.8 Research Gaps

Based on the literature review, the following key research gaps are identified:

- Limited studies on AZ91D-AA6061-T6 dissimilar joints, either with or without SiC reinforcement. Previous works have broadly examined either similar Al-Al or Mg-Mg or unreinforced Al-Mg dissimilar joints. Systematic studies on the reinforcement of SiC nanoparticles in AZ91D and AA6061-T6 are lacking.
- Limited focus on systematic process parameter optimization. While many studies report experimental results, very few employ statistical design frameworks such as RSM with CCD to model interactions between TRS, TS, and

nanoparticle volume fraction (Vf). Optimization approaches tailored for AZ91D-AA6061, both reinforced and unreinforced, are lacking.

- Absence of thermal analysis of both reinforced and unreinforced AZ91D-AA6061-T6 joints. Although several researchers have modeled and measured the temperature distribution in unreinforced Al-Al, Mg-Mg, and some Al-Mg systems, no dedicated thermal study has been reported for the AZ91D-AA6061 system. Establishing a baseline thermal profile is crucial for understanding the heat input and its impact on microstructural evolution.
- Insufficient comparative evaluation of reinforced vs. unreinforced joints. Most studies have treated nanoparticle-reinforced and unreinforced FSW joints separately. A unified investigation that considers both conditions under the same process window to establish comparative performance is missing.

2.9 Research Objectives

The objectives can be summarized as follows:

- To optimize the input and output responses of the welded joints.
- Determination of process parameters' effects on the mechanical properties of the friction stir-welded joints of dissimilar metals.
- The metallurgical characterization of the friction stir-welded joints of dissimilar metals with or without nanoparticles.
- Determination of temperature distribution during friction stir welding of dissimilar metals.

Chapter 3 : RESEARCH METHODOLOGY

3.1 Research Framework

This research work is conducted within a systematic and integrated framework to investigate the mechanical, metallurgical, and thermal behavior of FSWed dissimilar AA6061-T6 and AZ91D joints reinforced with SiC nanoparticles. The methodology is designed to comprehensively address the research objectives by combining statistically planned experiments, welding trials, detailed characterization, and numerical modeling. Such an integrated approach ensures reliability, repeatability, and scientific rigor in establishing correlations between process parameters, microstructural evolution, and joint performance.

In the first stage of the investigation, Design of Experiments (DOE) is employed to efficiently plan the experimental trials while minimizing the number of experiments without compromising the quality of results. Response Surface Methodology (RSM) based on the Central Composite Design (CCD) is employed to investigate the combined effect of key friction stir welding parameters, specifically TRS, TS, and SiC Vf. This statistical approach enables the development of predictive models for output responses and facilitates the identification of optimal process conditions.

Experimental FSW trials are conducted according to the design matrix obtained from the DOE framework. During welding, thermal measurements were carried out using K-type thermocouples to record temperature variations across the weld region. These measurements provide critical insight into the heat input and thermal cycles experienced by the material, which directly influence microstructural transformations and mechanical properties. Post-weld sample preparation is performed following standard metallographic procedures to ensure accurate and consistent characterization.

Subsequently, the welded joints are subjected to comprehensive testing and characterization to evaluate their mechanical and metallurgical performance. Tensile testing, microhardness measurements, and fracture analysis are conducted to assess joint strength, hardness distribution, and failure mechanisms. Microstructural characterization using optical microscopy and scanning electron microscopy is performed to examine grain refinement, material flow behavior, and intermetallic

compound formation across different weld zones. These experimental observations form the basis for understanding the structure-property relationships in dissimilar AA6061-T6 and AZ91D joints.

To complement the experimental investigation, an analytical model of the heat source is developed to describe the thermal behavior of the friction stir welding process. Numerical simulations are then performed using ANSYS to predict temperature distribution and validate the experimentally measured thermal profiles. The simulation results are systematically compared with experimental data to ensure model accuracy and to strengthen the reliability of the proposed approach.

Finally, statistical analysis using analysis of variance (ANOVA) is employed to identify the most significant process parameters and to validate the developed response surface models. The integration of experimental results, statistical analysis, and numerical simulation enables a comprehensive evaluation of the welding process, supporting the derivation of meaningful conclusions. The overall research workflow adopted in the present study is schematically illustrated in Figure 3.1.

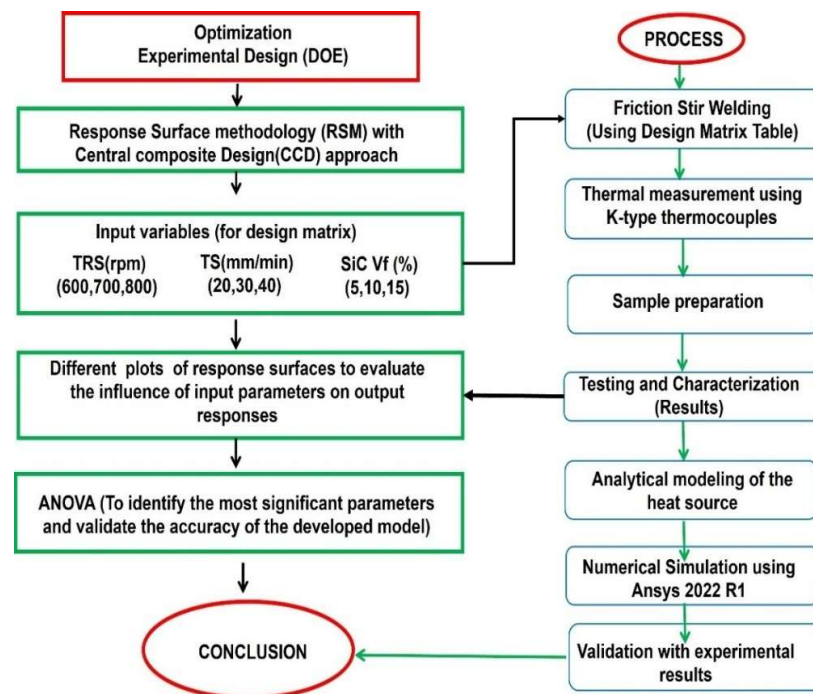


Figure 3.1: Schematic representation of Research Framework

3.2 Base Materials and Nanoparticles

The AA6061-T6 and AZ91D plates (dimensions: 140 mm × 40 mm × 6mm) were used as base metals for producing dissimilar weld joints by FSW. SiC nanoparticles with an average particle size of 90-100 nm and 99% purity, procured from Vedayukt India Pvt. Ltd., Jharkhand, India, were used as the reinforcing material. The selected base materials, AZ91D and AA6061-T6, are widely used in aerospace and automotive applications due to their favorable properties. AZ91D is appreciated for its low density and machinability, while AA6061-T6 is a high-strength aluminum with excellent corrosion resistance [38]. Unlike earlier works that primarily explore FSW of either similar combinations or unreinforced dissimilar combinations, this research examines the complex interaction between SiC reinforcement, heat input, and metallurgical bonding mechanisms in friction stir-welded (FSWed) joints of AZ91D and AA6061-T6. The distinctive characteristics of AZ91D and AA6061-T6 metals make them promising subjects for exploring the FSW process and finding their impact on the welding of these dissimilar metals.

Further, SiC was chosen for its high thermal and chemical stability, low thermal expansion, and strong matrix bonding, which help reduce cracking and enhance joint properties in dissimilar FSW [88]. Tables 3.1 and 3.2 present the elemental composition and mechanical properties of the base materials, respectively. The grain structure and elemental composition of the base metals and SiC nanoparticles are presented in Figure 3.2 (a-f). The elemental composition presented in Table 3.1 and Figure 3.2 was obtained using energy-dispersive spectroscopy (EDS) analysis.

Table 3.1 : Elemental composition of base metals

Materials	Al	Si	Mn	Zn	Cu	Fe	Ti	Cr	Mg
AA6061-T6	97.89	0.16	0.19	0.13	0.31	0.19	0.11	0.17	0.8
Mg AZ91D	8.42	0.06	0.28	0.26	0.01	0.06	0.15	0.39	90.38

Table 3.2: Mechanical properties of base metals

Metals	UTS (Mpa)	Elongation (%)	Hardness (HV0.1)
AA6061-T6	320	11	105
Mg AZ91D	219	9.5	66

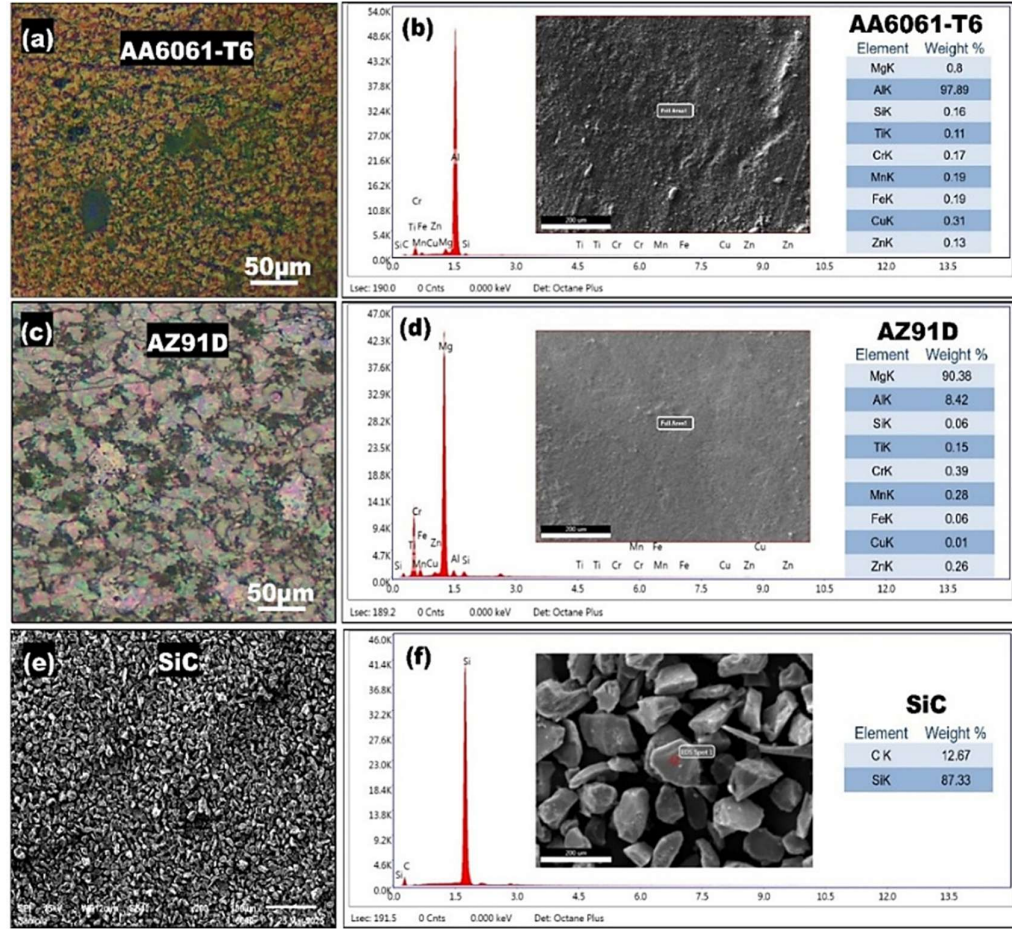


Figure 3.2: Optical micrographs showing (a, c) grain structure of base metals, (b, d, f) EDS analysis with elemental composition, (e) SEM morphology of SiC

A 2 mm deep groove was excavated along the joint interface of the base metal plates to integrate the SiC into the weld zone. Figure 3.3 shows the SiC nanoparticles, base metal plates, and grooves used in this research work. Moreover, SiC at different Volume fractions (Vf) of 5%, 10%, and 15% was introduced by excavating grooves with varying widths (0.71, 1.43, and 2.14 mm). To ensure that SiC was adequately filled into the grooves, a thick slurry of SiC nanoparticles and 99% pure ethanol was made. For each sample, the Silicon Carbide (SiC) volume fraction (Vf) was calculated as the ratio of the groove volume to the stirred volume displaced by the tool pin along the weld length (L), which was 120 mm.

The groove volume (V_G) is expressed as:

$$V_G = (W_G \times D_G) \times L \quad (3.1)$$

$$\text{or, } V_G = (A_G) \times L \quad (3.2)$$

where, A_G , W_G and D_G are the groove's cross-sectional area, width, and depth, respectively.

The stirred volume (V_S) displaced by the tool pin is expressed as:

$$V_S = (D_P \times H_P) \times L \quad (3.3)$$

$$\text{or, } V_S = (A_P) \times L \quad (3.4)$$

where, A_P , D_P and H_P are the tool pin's projected area, diameter, and length, respectively. Finally, Silicon Carbide Volume fraction (Vf) is calculated as:

$$\text{Silicon carbide Volume fraction (Vf)} = \left(\frac{\text{Groove Volume (} V_G \text{)}}{\text{Stirred Volume (} V_S \text{)}} \right) \times 100 \quad (3.5)$$

$$\text{or, Silicon carbide Volume fraction (Vf)} = \left(\frac{\text{Cross sectional Area of the groove}}{\text{Projected area of the tool pin}} \right) \times 100 \quad (3.6)$$

Using expressions 3.2, 3.4, and 3.6, the volume fraction was calculated.

3.3 Tool Material and Geometry

All FSW experiments were conducted using an H13 tool steel with a cylindrical pin profile, characterized by a hardness of 40-55 HRC, as shown in Figure 3.4. The tool material and geometry were chosen based on preliminary trials and established as suitable for dissimilar Al-Mg welding. H13 steel was selected because of its thermal stability, high hardness, and wear resistance, which ensure durability under the elevated temperatures and stresses of FSW of AZ91D and AA6061-T6. The cylindrical pin profile was chosen for its ability to provide stable material flow, uniform heat generation, homogeneous mixing in the SZ, and ease of fabrication. This combination ensured consistent weld quality, extended tool life, and dependable performance across the investigated process parameters [7,22].

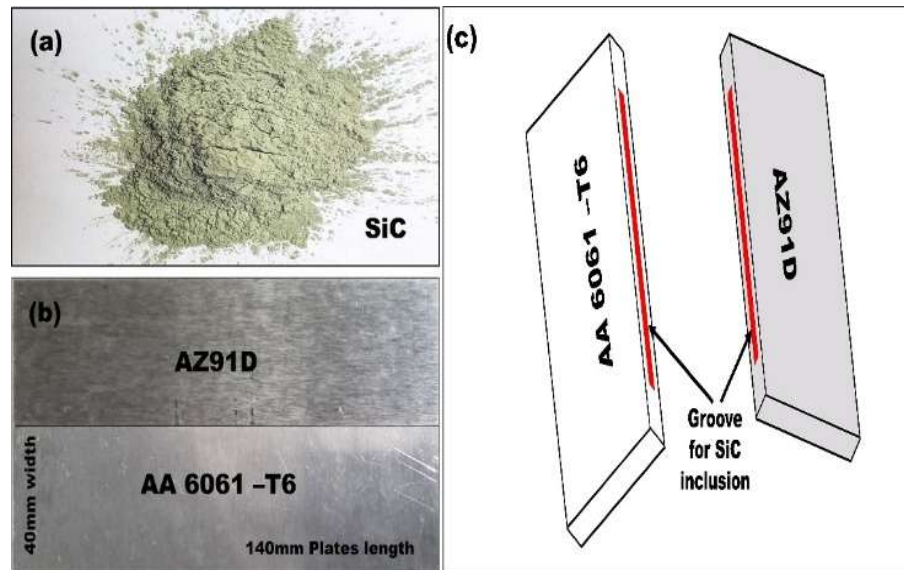


Figure 3.3:(a) SiC nanoparticles (b) Base metals used in FSW(c) Groove for SiC inclusion

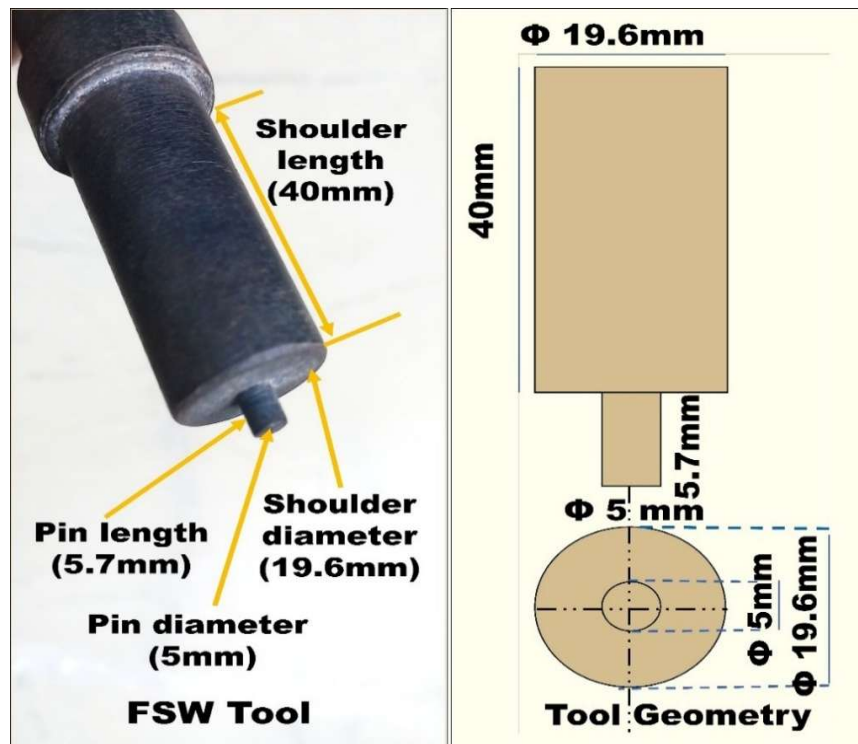


Figure 3.4: FSW tool and Geometry

3.4 Welding Configuration and Setup

Friction stir welding was performed under rigid clamping conditions using an advanced FSW machine (HYD-40KN), as shown in Figure 3.5. Based on preliminary trials, the configuration of AA6061-T6 on the advancing side (AS) and AZ91D on the retreating side (RS) yielded consistent, defect-free welds, and this configuration was adopted for all subsequent experiments. The tool was plunged to a depth of 0.1 mm and held for a dwell time of 3 to 5 seconds before welding was initiated.

3.4.1 Process Parameters

The main process parameters that varied were:

- Tool rotational speed (TRS): 600, 700, and 800 rpm
- Traverse speed (TS): 20, 30, and 40 mm/min
- SiC volume fraction (Vf): 5%, 10%, and 15%

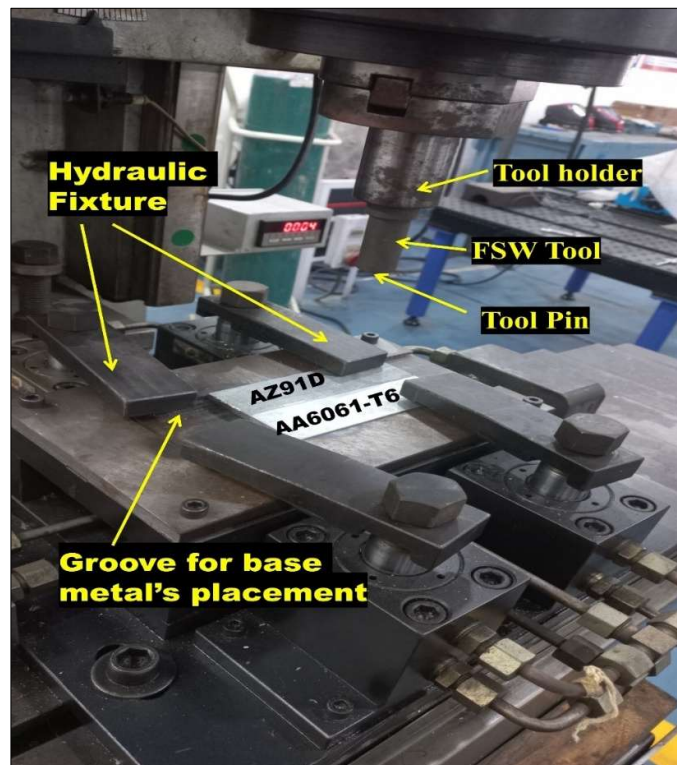


Figure 3.5: Experimental Setup of FSW of dissimilar AZ91D and AA6061-T6



Figure 3.6: Top view of SiC reinforced FSWed samples of AA6061-T6 and AZ91D showing surface appearance

3.5 Experimental Design

To systematically study the effect of process parameters on the joint properties, a Central Composite Design (CCD) approach was employed. CCD enables the evaluation of linear, quadratic, and interaction effects of parameters while minimizing the number of experiments. The selected input variables for the design were Tool Rotational Speed (600, 700, 800 rpm), Traverse Speed (20, 30, 40 mm/min), and SiC Volume Fraction (Vf) (5%, 10%, 15%). The tool tilt angle was maintained at 2° for all experiments, as preliminary trials indicated that this angle provided adequate forging pressure, stable material flow, and defect-free joints; and hence, it was not included as an optimization variable in the CCD.

Each factor was investigated at three levels, and the design generated a set of experimental runs to cover the entire parameter space. The CCD matrix used for

experimentation is presented in Table 4.2. The relationship between the response variable and the input parameters was modeled using a second-order polynomial regression equation. This study introduces a novel framework that combines experimental and statistical optimization using RSM with CCD to model and improve the mechanical and metallurgical characteristics. Regression analysis was performed using Design Expert software, and graphical representations of the output responses were generated. Contour plots of the response surfaces were examined to evaluate the influence of input parameters on the output responses. ANOVA was conducted to identify the most significant parameters and validate the accuracy of the developed model. Furthermore, model validation through multiple experimental confirmatory tests at non-optimal settings ensures robustness and generalizability across the design space. Figure 3.6 displays photographs of top views of SiC-reinforced friction stir-welded (FSW) samples produced according to the design matrix.

3.6 Sample Preparation

The welded plates were sectioned using a wire-cut EDM machine. Different types of specimens were prepared for mechanical and microstructural evaluation:

3.6.1 Tensile Specimens

Prepared according to ASTM E-8 standard dimensions as shown in Figure 3.7.

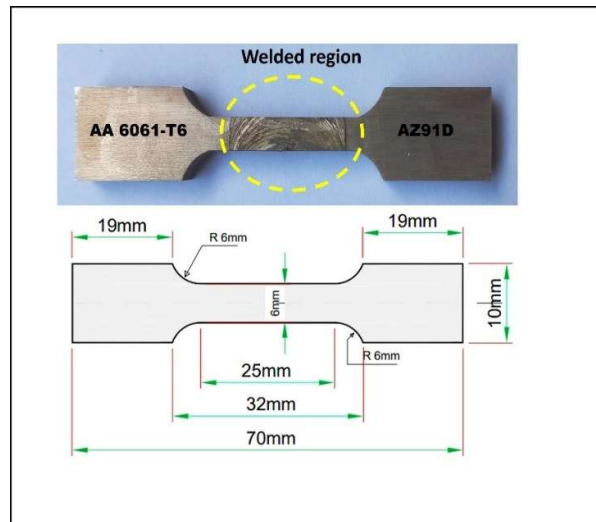


Figure 3.7: Tensile specimen as per ASTM E8 standards

3.6.2 Metallographic Samples

Samples were extracted and prepared from the welded specimens for microstructural examination by polishing and etching with suitable reagents for aluminum and magnesium sides. Figure 3.8 illustrates the metallographic samples and the single-disc polisher. Keller's solution, which consists of 6 mL of hydrofluoric acid (HF), 6 mL of hydrochloric acid (HCl), 3 mL of nitric acid (HNO₃), and 150 mL of distilled water, was used to etch the AA6061-T6 side for 20 to 30 seconds. A picric solution, comprising 6g picric acid, 100 mL ethanol (C₂H₆O), 10 mL acetic acid (CH₃COOH), and 10 mL distilled water, was used to etch the AZ91D side for 5-10 seconds. Additionally, 20g NaOH and 100 mL distilled water were used to etch the stir zone for 30-40 seconds. The surfaces of the polished specimen were subsequently analyzed using optical microscopy (Olympus OM microscope), which facilitated visual inspection and evaluation of the intermetallic compounds in the welded joints. A scanning electron microscope (SEM) was also employed to examine the particle distribution in the weldment and the microstructure of the stir zone. X-ray diffraction (XRD) analysis was conducted to confirm the existence of the intermetallic phases within the weld specimens. Further, the Energy dispersive spectroscopy (EDS) technique was utilized to determine the elemental composition of localized regions (especially SiC nanoparticles) in friction-welded joints. Finally, the fractography of the fractured specimens was also analyzed using the above-mentioned scanning techniques, SEM.



Figure 3.8:(a) Metallographic Samples (b) Single Disc Polisher

3.6.3 Hardness samples

Mounted and polished using standard metallographic procedures. Before the indentation, samples were polished using 300-3000 grade emery sheets to ensure surface quality. The sample preparations and testing procedures for the micro-hardness measurements were followed according to the guidelines of the Standard Test Method for Micro-indentation Hardness of Materials.

3.7 Testing and Characterization

3.7.1 Tensile Testing

Tensile tests were conducted on a Zwick Rowell Universal Testing Machine (UTM) at a crosshead speed of 1 mm/min, as shown in Figure 3.9.



Figure 3.9: Universal Testing Machine (UTM)

3.7.2 Microhardness Testing

Vickers microhardness measurements were performed at 1 mm intervals across the weld zone under a 100-gf load with a dwell time of 15 s. The hardness profile was used

to evaluate variations across the stir zone, TMAZ, HAZ, and base materials. The Microhardness Testing machine is shown in Figure 3.10.



Figure 3.10: Microhardness Testing Machine

3.7.3 Optical Microscopy (OM)

Metallographic samples were polished and etched appropriately for aluminum and magnesium alloys. OM was used to observe grain distribution, nugget zone structure, and the presence of defects. The Optical Microscope is shown in Figure 3.11.



Figure 3.11: Optical Microscope

3.7.4 Scanning Electron Microscopy (SEM)

SEM was employed to investigate the weld microstructure at high magnification. It was used to identify features such as grain refinement, particle dispersion, and interfacial bonding. All the samples were gold-coated on a coating Machine prior to SEM analysis, as shown in Figure 3.12.



Figure 3.12: Gold coating Machine, (b) SEM Machine

3.7.5 Energy-Dispersive Spectroscopy (EDS)

EDS analysis was performed in conjunction with SEM to determine the elemental composition and validate the distribution of alloying elements and SiC nanoparticles within the stir zone and interfacial regions. The EDS Machine is shown in Figure 3.13.

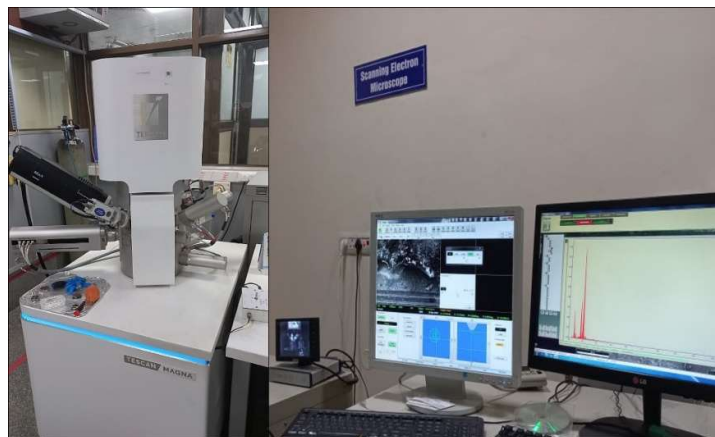


Figure 3.13: EDS Machine

3.7.6 X-ray Diffraction (XRD)

XRD analysis was performed on selected welded specimens to identify the phases and intermetallic compounds (IMCs) present in the weld zone. The obtained diffraction patterns were compared with standard reference data for phase confirmation. The XRD Machine is shown in Figure 3.14.

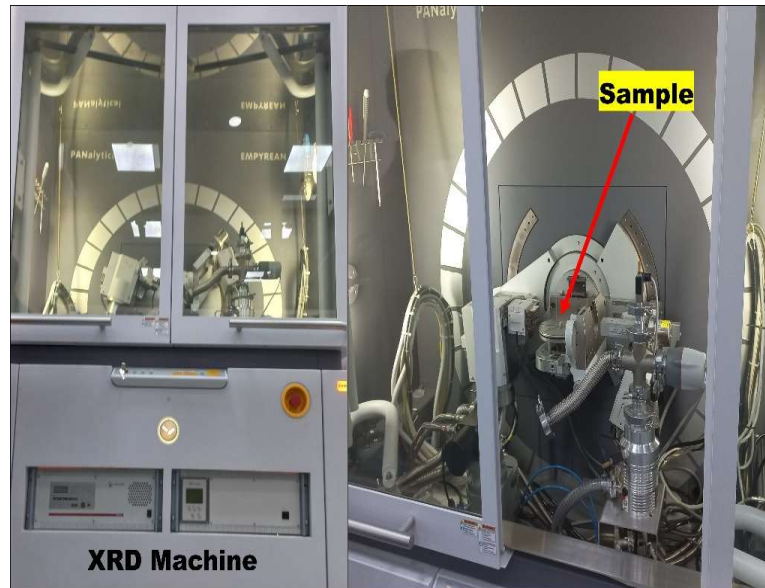


Figure 3.14: XRD Machine

3.7.7 Fractography

Fractured tensile specimens were examined under SEM to study the failure modes and identify ductile, brittle, or mixed fracture characteristics in relation to the welding parameters and SiC reinforcement levels.

3.8 Thermal Measurement by Thermocouples

To monitor the thermal cycles during welding, K-type thermocouples were inserted at selected positions in the plates. The thermocouples were placed at varying distances from the weld centerline and embedded in small drilled holes to ensure intimate contact with the base material, as shown in Figure 3.15 (a-c). Peak temperatures at

thermocouple positions were continuously recorded using a temperature indicator system (Figure 3.15b) during the welding process. The recorded peak temperatures were later used to validate the numerical model.

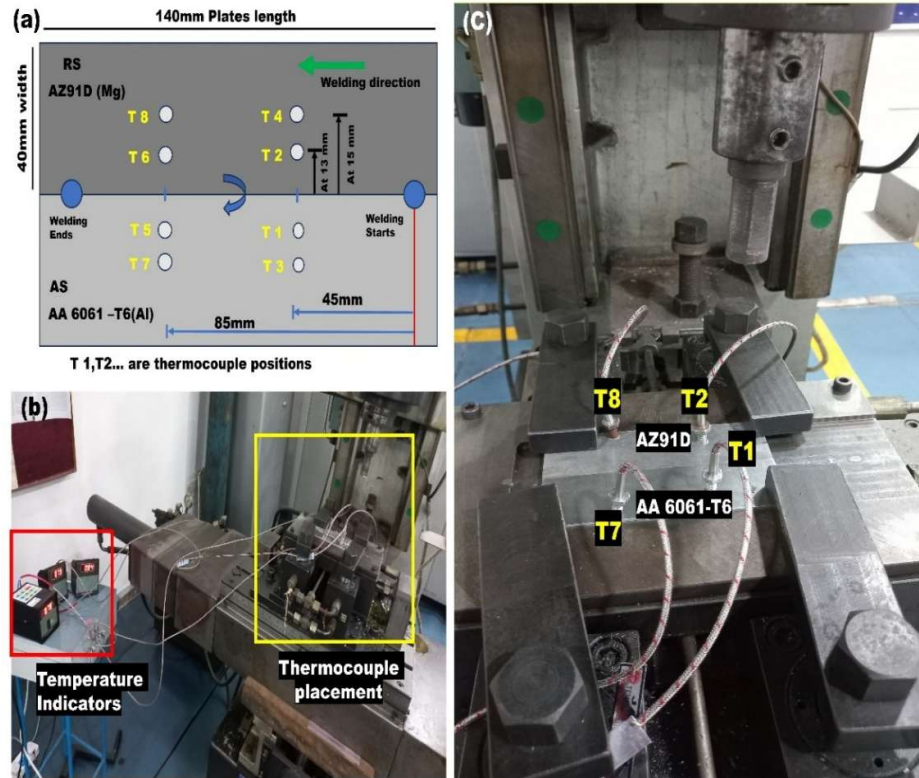


Figure 3.15: (a) Schematic representation of thermocouple locations, (b) overall experimental setup with temperature indicators, and (c) close-up view of thermocouple placement on AZ91D and AA6061-T6 plates.

3.8.1 Modeling of the Heat Source

The total heat generated by the tool during friction stir welding primarily arises from the frictional sliding and plastic deformation beneath the tool shoulder and around the pin. The total heat generated (Q_T) was estimated using the classical analytical formulation expressed as:

$$Q_T = Q_S + Q_P \quad (3.7)$$

where Q_S and Q_P are the heat generated by the shoulder and the pin, respectively.

3.8.1.1 Heat Generated by the Tool Shoulder (Q_S)

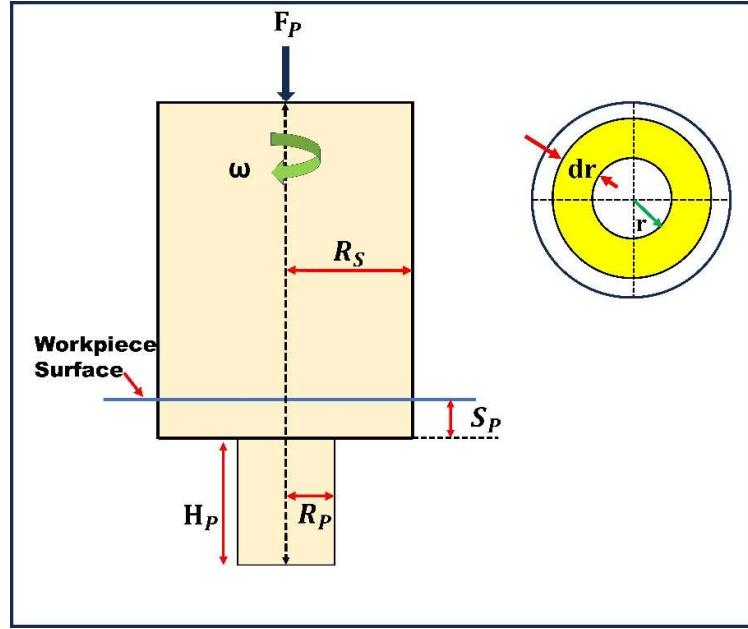


Figure 3.16: Schematic of the plunged tool and Analytical modeling assumptions

Consider a small ring element at a distance ‘r’ from the tool center with thickness ‘dr’ as shown in Figure 3.16. Then, the Area can be calculated as

$$dA = 2\pi r dr \quad (3.8)$$

The heat generated from this ring is

$$dQ_S = \omega r \tau_{\text{contact}} dA \quad (3.9)$$

$$dQ_S = \omega r \tau_{\text{contact}} 2\pi r dr \quad (3.10)$$

Where ‘ω’ is TRS, τ_{contact} is the contact shear stress. Then, the net heat generated through the flat shoulder is

$$\int_0^{Q_S} dQ_S = \int_{R_P}^{R_S} \omega r \tau_{\text{contact}} 2\pi r dr \quad (3.11)$$

$$Q_S = 2\pi \tau_{\text{contact}} \omega \int_{R_P}^{R_S} r^2 dr \quad (3.12)$$

$$Q_S = \frac{2\pi}{3} \tau_{\text{contact}} \omega (R_S^3 - R_P^3) \quad (3.13)$$

Where, R_p pin radius and R_s is the shoulder radius, respectively.

Now, the term S_p denotes the depth to which the tool shoulder penetrates the workpiece under the applied axial force F . During welding, the tool traverses at a velocity V_w , representing the welding speed [22,23]. The friction produced by the tool's rotation causes a rise in temperature within the material, which in turn lowers its yield strength. As the tool begins its forward movement, it encounters a temperature-dependent yield strength σ_{YS} , corresponding to the yield strength of the material at the prevailing welding temperature. Only the front portion of the shoulder's curved surface (contact surface) interacts effectively with the plasticized material and contributes predominantly to heat generation during tool movement [22,23].

Now, the half-curved surface (contact surface) area of the shoulder is given by

$$A_{CS} = \pi R_s S_p \quad (3.14)$$

where, R_s is the tool shoulder radius and S_p is the shoulder plunge depth.

Now, the force acting on this surface area will be given by

$$F = \mu \sigma_{YS} A_{CS} = \mu \pi \sigma_{YS} R_s S_p \quad (3.15)$$

where F is the frictional shear force on the curved surface, μ is the friction coefficient and σ_{YS} is the effective yield (flow) stress of the workpiece at welding temperature.

Now, the Heat generated by the front half of the shoulder curved surface (contact surface) is given by

$$Q_{Sc} = \eta F V_w = \eta \mu \pi \sigma_{YS} R_s S_p V_w \quad (3.16)$$

where Q_{Sc} is the heat generation rate (W) on the front half of the shoulder curved surface (or curved surface), η is the heat-transfer efficiency (fraction of frictional energy absorbed by the workpiece), and $V_w = \omega R_s$ is the tangent speed at radius R_s (ω in $\text{rad} \cdot \text{s}^{-1}$).

3.8.1.2 Heat Generated from Cylindrical Pin Probe (Q_p)

Similar to the shoulder surface, heat generated by the probe's bottom face is obtained by integrating the local frictional power $\tau_{\text{contact}}(\omega r)$ over the pin's circular area at a distance 'r' from the tool center with thickness 'dr'.

$$Q_{P_b} = \int_0^{R_p} \omega r \tau_{\text{contact}} 2\pi r dr \quad (3.17)$$

$$Q_{P_b} = 2\pi\omega\tau_{\text{contact}} \int_0^{R_p} r^2 dr = 2\pi\omega\tau_{\text{contact}} \left[\frac{R_p^3}{3} \right] \quad (3.18)$$

$$Q_{P_b} = \frac{2}{3} \pi\omega\tau_{\text{contact}} R_p^3 \quad (3.19)$$

Now, Heat generation from the side surface of the probe due to rotation is

$$\text{Area}(dA) = \int_0^{2\pi} R_p dH d\theta \quad (3.20)$$

$$\int_0^{Q_{P_s}} dQ_{P_s} = \int_0^{H_p} \omega R_p \tau_{\text{contact}} 2\pi R_p dH \quad (3.21)$$

$$Q_{P_s} = \int_0^{H_p} \omega R_p \tau_{\text{contact}} 2\pi R_p dH \quad (3.22)$$

$$Q_{P_s} = 2\pi\omega R_p^2 \tau_{\text{contact}} \int_0^{H_p} dH \quad (3.23)$$

$$Q_{P_s} = 2\pi\omega R_p^2 \tau_{\text{contact}} H_p \quad (3.24)$$

where, H_p is the height of the pin.

Heat generation due to the half-curved surface area of the probe (Q_{P_c}) occurs as the tool probe travels along the weld line. The material exhibits yield strength σ_{YS} at the elevated welding temperature, as reported in [22,23]. During the tool's movement, only the front half of the probe's curved surface area (A_{P_c}) is subjected to force and is responsible for producing heat.

The half-curved surface area of the probe (A_{P_c}) is given by

$$A_{P_C} = \pi R_P H_P \quad (3.25)$$

Now, the force acting on this surface area will be given by

$$F = \mu \sigma_{YS} A_{CS} = \mu \pi \sigma_{YS} R_P H_P \quad (3.26)$$

Then, the heat generated by the front half of the probe's curved surface area is given by

$$Q_{P_C} = \eta F V_w = \mu \pi \sigma_{YS} R_P H_P V_w \quad (3.27)$$

where Q_{P_C} is the heat generation rate (W) on the front half of the probe's curved surface, η is the heat-transfer efficiency (fraction of frictional energy absorbed by the workpiece), and $V_w = \omega R_P$ is the tangent speed at radius R_P (ω in $\text{rad} \cdot \text{s}^{-1}$).

So, the total heat generated from the tool in the FSW process will be

$$Q_T = Q_S + Q_{S_C} + Q_{P_b} + Q_{P_s} + Q_{P_C} \quad (3.28)$$

However, heat generation due to the half-curved surface areas of the shoulder and probe as the tool probe travels along the weld line is negligible as compared to the heat generated due to rotation. Hence, Q_T becomes

$$Q_T = Q_S + Q_{P_b} + Q_{P_s} \quad (3.29)$$

$$Q_T = \frac{2\pi}{3} \tau_{\text{contact}} \omega (R_S^3 - R_P^3) + \frac{2}{3} \pi \omega \tau_{\text{contact}} R_P^3 + 2\pi \omega R_P^2 \tau_{\text{contact}} H_P \quad (3.30)$$

Assuming sliding contact [22]

$$\tau_{\text{contact}} = \tau_{\text{friction}} = \mu P \quad (3.31)$$

$$Q_T = \frac{2\pi}{3} \mu P \omega (R_S^3 - R_P^3) + \frac{2}{3} \pi \omega \mu P R_P^3 + 2\pi \omega R_P^2 \mu P H_P \quad (3.32)$$

$$Q_T = \frac{2}{3} \pi \omega \mu P (R_S^3) + 2\pi \omega R_P^2 \mu P H_P \quad (3.33)$$

Where, P = average contact pressure, which can be calculated as

$$P = \frac{F_P}{A_{\text{contact}}}, \text{ with } A_{\text{contact}} = \pi(R_S^2 - R_P^2) \quad (3.34)$$

Also F_P is Plunge Force and can be calculated as

$$F_P = P \pi (R_S^2 - R_P^2) \quad (3.35)$$

Since the shoulder radius dominates, neglecting the pin radius,

$$F_p = P\pi R_s^2 \quad (3.36)$$

Now, in terms of plunging force Q_T can be written as:

$$Q_T = \frac{2}{3} \omega \mu F_p \left[R_s + \frac{3R_p^2 H_p}{R_s^2} \right] \quad (3.37)$$

This equation represents the total theoretical maximum heat generated by the tool during FSW, assuming all mechanical work is converted to heat within the workpiece. In practice, some heat is lost through the tool, backing plate, and environment. To account for this, a thermal efficiency factor η is introduced.

The effective heat transferred to the workpiece is given by:

$$Q_{\text{effective}} = \eta Q_T = \eta \frac{2}{3} \omega \mu F_p \left[R_s + \frac{3R_p^2 H_p}{R_s^2} \right] \quad (3.38)$$

In the present study, $\eta = 0.8$ was adopted based on literature for Al-Mg dissimilar FSW systems, which accounts for typical losses through conduction and convection. The heat input per unit weld length is calculated as:

$$Q_\ell = \frac{Q_{\text{eff}}}{v} \quad (3.39)$$

where v is the traverse speed ($\text{mm} \cdot \text{min}^{-1}$).

These analytical heat estimations provide $Q_{\text{effective}}$ (Eq. 3.38) and the heat input per unit length (Eq. 3.39), which were used to correlate experimental and simulated temperature data, as the total heat governs the temperature field.

3.9 Numerical Simulation of Temperature Distribution

A three-dimensional finite element model of the non-reinforced AA6061-AZ91D friction stir-welded joint was developed in ANSYS to simulate the temperature distribution during welding. The model incorporated the tool shoulder and pin as heat sources, while assigning the thermal properties of aluminum and magnesium alloys to the respective regions. Heat transfer was considered through conduction within the plates and convection at exposed surfaces.

The simulation was performed at various tool rotational and traverse speeds to determine the peak temperatures and thermal gradients in the stir zone. The predicted temperature histories were validated against experimental thermocouple data obtained from the non-reinforced welds.

This analysis provided a reliable understanding of the thermal field in FSW, helping to explain the observed material flow, microstructural evolution, and IMCs formation while ensuring confidence in the experimental observations.

3.9.1 Governing Equations

The transient heat transfer during FSW of dissimilar AZ91D and AA6061-T6 alloys was modeled using the general form of Fourier's heat conduction equation with internal heat generation:

$$k \left(\frac{\partial^2 T}{\partial x^2} + \frac{\partial^2 T}{\partial y^2} + \frac{\partial^2 T}{\partial z^2} \right) + Q_{\text{gen}} = \rho C_p \frac{\partial T}{\partial t} \quad (3.40)$$

where:

T : temperature (K)

k : thermal conductivity ($\text{W} \cdot \text{m}^{-1} \cdot \text{K}^{-1}$)

Q_{gen} : volumetric heat generation rate ($\text{W} \cdot \text{m}^{-3}$)

ρ : density ($\text{kg} \cdot \text{m}^{-3}$)

C_p : specific heat capacity ($\text{J} \cdot \text{kg}^{-1} \cdot \text{K}^{-1}$)

The first term on the left-hand side represents heat conduction in the x, y, and z directions, while the second term accounts for the volumetric heat generated due to frictional sliding and plastic deformation between the tool and the workpiece. The right-hand side expresses the rate of transient energy storage in the material.

This governing equation is mathematically equivalent to the compact vector form solved in ANSYS:

$$\rho C_p \frac{\partial T}{\partial t} = \nabla \cdot (k \nabla T) + \dot{Q}_{\text{vol}} \quad (3.41)$$

Both expressions describe transient heat conduction with internal generation, and the latter form was used by the numerical solver during simulation.

3.9.2 Thermal Boundary Conditions

The initial and boundary conditions were defined as follows:

1. Initial Condition:

$$T(x, y, z, 0) = T_i \quad (3.42)$$

This represents the ambient temperature of the workpiece before the tool plunge, where T_i is the ambient temperature at time $t = 0$.

2. Convective Boundary Condition:

Convective heat loss from the top and side surfaces of the workpiece exposed to air is modeled as:

$$q = hA(T - T_0) \quad (3.43)$$

where, h is the convective heat transfer coefficient, and T_0 is the ambient temperature (300K).

3. Conductive Boundary Condition at the Bottom Surface:

Heat loss from the bottom of the workpiece to the backing plate is expressed as:

$$q = h_b A(T - T_0) \quad (3.44)$$

Where, h_b is the effective heat transfer coefficient at the bottom surface (workpiece to backing plate)

4. Radiative losses:

Although small compared with conduction and convection, surface radiation was considered using:

$$q = \varepsilon \sigma (T^4 - T_0^4) \quad (3.45)$$

where, $\varepsilon = 0.3$ is the surface emissivity, σ is the Stefan-Boltzmann constant, and its value is $5.67 \times 10^{-8} \text{W/m}^2 \text{K}^4$

A surface emissivity of 0.3 was assigned to the workpiece surface, corresponding to the typical emissivity value for aluminum-magnesium at elevated temperatures. However, radiative losses are relatively small compared with convective and conductive heat transfer under FSW conditions (below 500 °C). Therefore, while emissivity was defined for completeness in the simulation setup, its influence on the overall temperature distribution is negligible.

3.9.3 Material properties

The following material properties (Table 3.3) were considered for the Simulation.

Table 3.3: Material properties used in simulation

Material	Density (Kg/m³)	Specific Heat C_p (J/Kg K)	Thermal conductivity (W/m K)
AZ91D	2719	871	202.4
AA6061-T6	1810	1050	72

3.9.4 Assumptions Used in Thermal Simulation

The following assumptions were adopted to simplify the numerical modeling without compromising the accuracy of temperature prediction.

- The process was modeled as a transient heat transfer problem to capture the time-dependent temperature evolution during tool movement.
- The heat generated by frictional sliding and plastic deformation at the tool-workpiece interface was represented as an effective uniform surface heat load over the shoulder contact area.
- The material properties (thermal conductivity, density, and specific heat) of AZ91D and AA6061 were considered temperature-independent constants to maintain computational efficiency.
- The tool and workpiece contact was assumed to be perfect, ensuring uniform heat transfer without interfacial gaps.

- Heat loss occurred only through convection and radiation at the exposed surfaces; internal heat sources other than the tool were neglected.
- The mechanical effects of material flow and strain hardening were not modeled explicitly; their impact was implicitly accounted for in the effective heat input term.
- These assumptions allowed the simulation to predict reliable temperature fields consistent with the experimental thermocouple data while maintaining a manageable computational load.

3.9.5 Simulation Details

A transient thermal analysis was performed in ANSYS 2022 R1. The solver inherently applies the above heat conduction equation, which is discretized using the finite element method. Temperature-dependent material properties were assigned for both AZ91D (magnesium) and AA6061-T6 (aluminum) regions. The applied surface heat flux was distributed over the tool shoulder contact area, while convective and conductive losses were imposed on the exposed and bottom surfaces, respectively.

The simulated temperatures at the thermocouple locations were compared with the experimentally measured peak temperatures for validation. A close agreement was obtained, confirming that the heat input model and selected boundary conditions accurately represent the thermal behavior of the dissimilar AZ91D-AA6061 system.

Chapter 4 : RESULTS AND DISCUSSION

4.1 Development of Mathematical Models and Design Matrix

RSM embodies a convergence of mathematical and statistical techniques that optimize processes and augment the value of products. The methodology entails the DOE and the investigation of the influence of various parameters on responses of interest. The findings from the experiments are analyzed using statistical models to identify the optimal parameter configurations that yield the intended response. The models created with Response Surface Methodology predict results for new process parameter configurations, thereby minimizing resources and time by reducing the need for additional experimentation. Contour plots and response surfaces are essential for clarifying the link between output and input responses, enabling the rapid and direct determination of optimal process parameter settings.

A series of experiments involving base metals delineated the suitable working boundaries for the defined parameters. The trial runs helped to determine the ideal and most appropriate range of process parameters to obtain SiC-reinforced weldments free of defects. The effectiveness of the CCD developed by Box and Wilson [89] in reducing the number of experimental runs while maintaining reliable data for building second-order polynomial models encouraged me to utilize this design in my research work. The Central Composite Design is particularly useful when resource constraints render complete factorial designs impractical in experimental settings. Additionally, its rotatability guarantees constant variance at every design point, which is essential for accurate analysis using response surface methods.

CCD is a widely used statistical approach in RSM for designing experiments. This approach analyses and enhances a reaction that relies on multiple numerical variables. This design facilitates understanding of the relationships between various factors and their responses, as well as identifying the optimal conditions for the process under investigation. Furthermore, this design can manage numerous variables, thereby enhancing its adaptability for various experimental configurations. It incorporates replicated center points to improve understanding of the process's variability, refine

prediction precision, and enhance the design by incorporating points along each axis at a defined distance from the center, effectively capturing the curvature in the response surface. It employs a comprehensive or partial factorial design as the primary experimental structure. Every sample underwent welding in compliance with the design matrix's requirements. It involves regression-based analysis to evaluate the impact of key process variables, enabling accurate predictions with minimal experimental effort.

Table 4.1 presents the selected parameter ranges, and Table 4.2 outlines the design matrix. The expressions that follow represent the various factors selected and their corresponding interaction effects. Regression models were developed using RSM to establish the relationship between input variables (TRS (A), TS (B), and SiC (Vf) (C)) and output responses (UTS, % strain, and microhardness), consistent with prior methodology. This can be explained well by using the following equations [47].

$$\text{Ultimate Tensile Strength (UTS)} = f(A, B, C) \quad (4.1)$$

$$\% \text{ Strain} = f(A, B, C) \quad (4.2)$$

$$\text{Microhardness} = f(A, B, C) \quad (4.3)$$

The equation for the second-order polynomial regression representing the response surface 'Y' is expressed as

$$y = n_0 + \sum n_i x_i + \sum n_{ii} x_i^2 + \sum n_{ij} x_{ij} \quad (4.4)$$

Where, n_0 = Mean value of response

n_i = Linear coefficient

n_{ii}, n_{ij} = Quadratic and interaction coefficient

All coefficients were evaluated for significance at a 95% confidence interval using Design Expert software. The final model incorporated only the significant coefficients. The resulting mathematical model, developed to predict the tensile strength, % strain, and microhardness of SiC-reinforced FSWed AA6061-T6 and AZ91D joints, based on this methodology, is presented as follows:

$$\text{UTS} = 112.35 + 10.62A - 2.38B + 3.51C + 1.17AB - 0.75AC + 0.53BC - 18.38A^2 - 1.46B^2 - 1.29C^2 \quad (4.5)$$

$$\text{Strain (\%)} = 6.02 + 0.47A - 0.27B + 0.38C + 0.29AB - 0.21AC - 0.20BC - 1.57A^2 - 0.30B^2 + 0.05C^2 \quad (4.6)$$

$$\text{Microhardness} = 87.23 + 3.47A - 1.44B + 1.54C - 0.60AB - 0.44AC + 0.48BC - 11.80A^2 - 0.43B^2 - 0.32C^2 \quad (4.7)$$

In Equations (4.5, 4.6, and 4.7), AB, BC, and AC represent the interaction effects between two variables, while A, B, and C denote the individual influence of each parameter. Table 4.2 presents the central composite design matrix, detailing input process parameters and corresponding output responses. Twenty different samples were welded at these process parameters using SiC nanoparticles. Also, to assess the impact of SiC nanoparticles, a separate sample was welded without using SiC at the optimized RSM parameters. These optimized parameters are shown in the ramp chart.

Table 4.1: FSW process parameters' range and their level

Parameter			Range		Levels		
Factor		Units	Minimum	Maximum	-1	0	1
A	Tool Rotational Speed (TRS)	rpm	600	800	600	700	800
B	Traverse Speed (TS)	mm/minute	20	40	20	30	40
C	Volume fraction (Vf)	%	5	15	5	10	15

4.2 Adequacy of the Developed Model

A mathematical model based on RSM and ANOVA was developed to predict the UTS, % strain, and microhardness of dissimilar AZ91D and AA6061-T6 FSWed joints. The model has been assessed as adequate within a 95% confidence interval. Table 4.3-4.5 further illustrates the models' efficacy in predicting output responses. All regression models exhibited $p < 0.0001$. In particular, the model terms A, B, C, AB, AC, BC, A^2 , B^2 , and C^2 were statistically significant ($p < 0.05$) for UTS and % strain. The lack of fit for all responses was not statistically significant, reaffirming that the developed models accurately represent the experimental data.

Model reliability was further supported by strong regression metrics (Table 4.6), with R^2 , adjusted R^2 , and predicted R^2 values for UTS, % strain, and microhardness

indicating strong model fits. Adequate precision values for all responses exceeded the recommended threshold of 4.0, confirming model adequacy. The predicted vs. actual plots (Figure 4.1a-c) revealed a close alignment of data points along the 45° line, demonstrating accurate model predictions with minimal deviation.

Table 4.2: CCD Design Matrix with Input and Output Responses

FSW welded specimen number	Process (Input) Parameters			Output Responses		
	Factor 1	Factor 2	Factor 3	Response 1	Response 2	Response 3
	A: Tool Rotational Speed (rpm)	B: Traverse Speed (mm/min)	C: SiC (Volume Fraction) (%)	UTS (MPa)	Strain (%)	Microhardness (HV0.1)
1	700	30	10	113.33	6.30	88.00
2	700	40	10	108.78	5.45	85.67
3	800	30	10	104.74	4.70	78.56
4	600	20	15	88.12	4.44	73.45
5	600	40	15	81.27	4.00	72.47
6	700	30	15	114.98	6.67	88.90
7	700	30	10	111.49	6.11	86.89
8	600	40	5	71.46	2.22	67.87
9	700	30	10	112.00	5.90	87.01
10	700	30	10	111.12	5.67	86.50
11	700	30	10	112.41	6.23	87.44
12	800	20	15	105.89	4.80	80.78
13	700	30	5	107.89	5.38	85.10
14	700	20	10	113.75	5.88	88.12
15	800	40	5	96.90	4.60	74.56
16	600	30	10	83.96	4.11	72.50
17	800	40	15	103.21	4.89	77.69
18	800	20	5	101.23	4.68	79.89
19	600	20	5	80.94	4.12	70.45
20	700	30	10	112.23	6.15	87.12

Furthermore, the graph depicting the residual probability for UTS, % strain, and microhardness of the dissimilar metals AA6061 and AZ91D is presented in Figure 4.2 (a-c). The alignment of points along a straight line suggests that the errors conform to

a normal distribution [89]. The observations presented here lend credence to the viability of models concerning UTS, % strain, and microhardness in joints, affirming their applicability in welding processes. A confirmation analysis (Table 4.7) further validated the models. Predicted UTS, strain, and microhardness values were in close agreement with the observed values and fell within the 95% prediction intervals. Standard prediction error (SE Pred) remained within acceptable limits, confirming the statistical soundness of the developed models. Including standard deviation, standard error, and prediction intervals provides statistical depth, supporting the adequacy and predictive accuracy of the response surface models developed using Design-Expert software. These results demonstrate that the RSM-based models can reliably predict and optimize the mechanical characteristics of dissimilar AA6061-T6 and AZ91D FSWed joints reinforced with SiC particles.

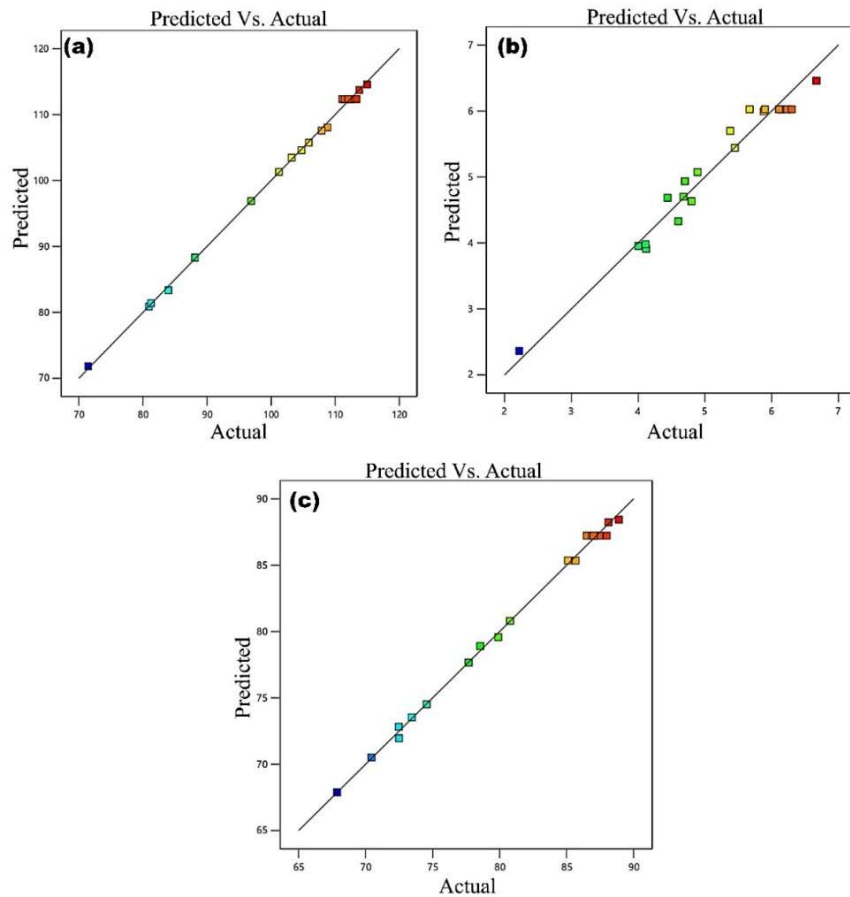


Figure 4.1: Predicted Vs Actual, (a) UTS, (b) % strain, (c) microhardness

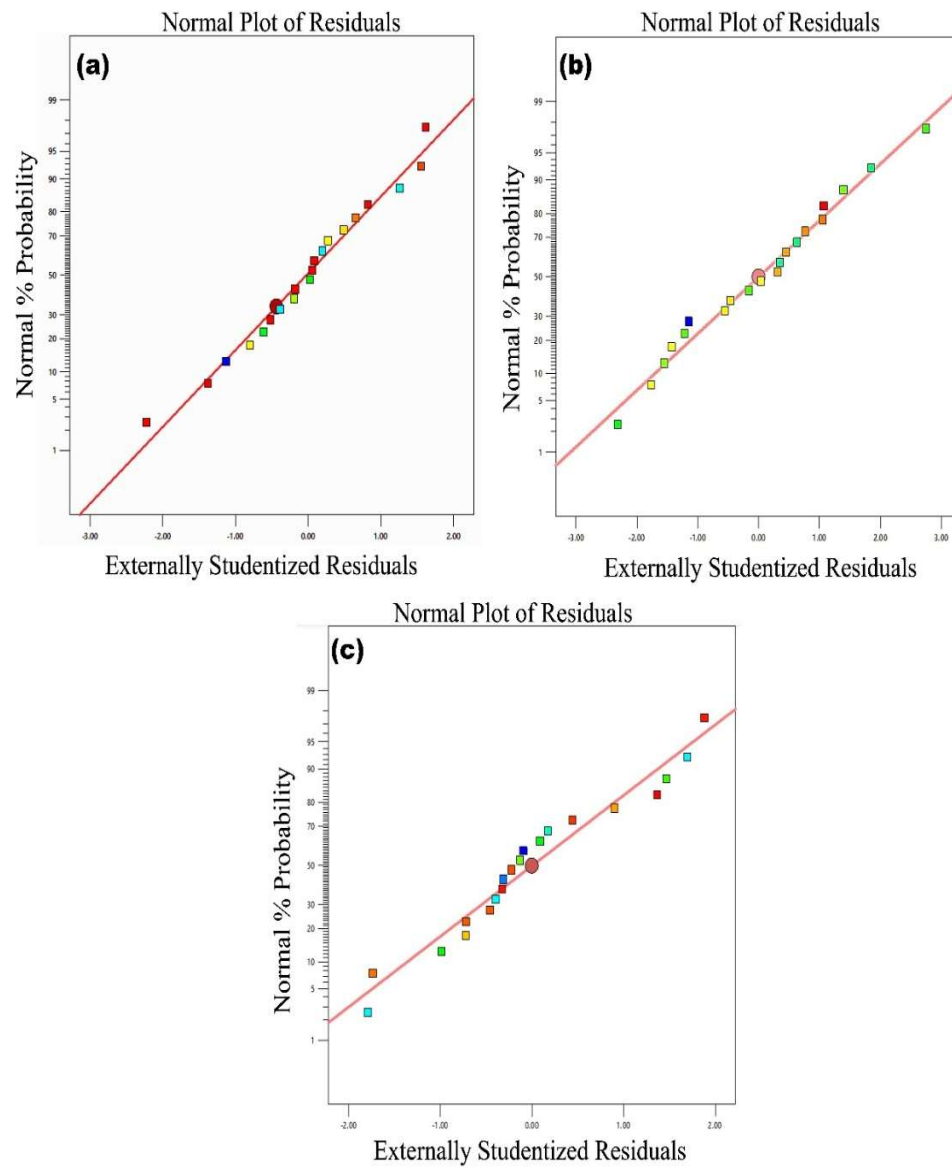


Figure 4.2: Normal probability plots of residual for (a) UTS, (b) % strain, (c) microhardness

Table 4.3: ANOVA results for UTS

Source	Sum of Squares	DOF	Mean Square	F-value	p-value	
Model	3370.82	9	374.54	775.28	< 0.0001	significant
A-TRS	1128.27	1	1128.27	2335.49	< 0.0001	
B-TS	80.15	1	80.15	165.9	< 0.0001	
C-TTA	122.85	1	122.85	254.3	< 0.0001	
AB	10.86	1	10.86	22.48	0.0008	
AC	4.53	1	4.53	9.38	0.012	
BC	2.29	1	2.29	4.74	0.0345	
A ²	928.56	1	928.56	1922.09	< 0.0001	
B ²	5.87	1	5.87	12.14	0.0059	
C ²	4.58	1	4.58	9.48	0.0117	
Residual	4.83	10	0.4831			
Lack of Fit	1.86	5	0.3725	0.6275	0.6892	Not significant
Pure Error	2.97	5	0.5937			
Cor Total	3375.65	19	374.54	775.28		

Table 4.4: ANOVA results for % strain

Source	Sum of Squares	DOF	Mean Square	F-value	p-value	
Model	20.9	9	2.32	29.87	< 0.0001	significant
A-TRS	2.28	1	2.28	29.39	0.0003	
B-TS	0.7618	1	0.7618	9.8	0.0107	
C-TTA	1.44	1	1.44	18.58	0.0015	
AB	0.6903	1	0.6903	8.88	0.0138	
AC	0.357	1	0.357	4.59	0.0377	
BC	0.3321	1	0.3321	4.27	0.0386	
A ²	6.75	1	6.75	86.8	< 0.0001	
B ²	0.2581	1	0.2581	3.32	0.0284	
C ²	0.0079	1	0.0079	0.1018	0.0436	
Residual	0.7773	10	0.0777			
Lack of Fit	0.5025	5	0.1005	1.83	0.2619	Not significant
Pure Error	0.2748	5	0.055			
Cor Total	21.68	19				

Table 4.5: ANOVA results for microhardness

Source	Sum of Squares	DOF	Mean Square	F-value	p-value	
Model	923.64	9	102.63	429.01	< 0.0001	significant
A-TRS	120.69	1	120.69	504.5	< 0.0001	
B-TS	20.82	1	20.82	87.04	< 0.0001	
C-TTA	23.78	1	23.78	99.4	< 0.0001	
AB	2.95	1	2.95	12.34	0.0056	
AC	1.6	1	1.6	6.7	0.027	
BC	1.84	1	1.84	7.71	0.0196	
A ²	382.73	1	382.73	1599.92	< 0.0001	
B ²	0.5139	1	0.5139	2.15	0.0435	
C ²	0.2945	1	0.2945	1.23	0.0231	
Residual	2.39	10	0.2392			
Lack of Fit	1.08	5	0.2151	0.8169	0.5851	Not significant
Pure Error	1.32	5	0.2633			
Cor Total	926.04	19				

Table 4.6: Regression adequacy statistics for UTS, % strain, and microhardness

Parameters	Std. Dev.	Mean	C.V. %	R ²	Adjusted R ²	Predicted R ²	Adeq Precision
UTS	0.6951	101.79	0.68	0.9986	0.9973	0.9954	86.99
% Strain	0.2788	5.12	5.45	0.9641	0.9319	0.6582	20.77
Microhardness	0.4891	80.95	0.60	0.9974	0.9951	0.9886	59.42

Table 4.7: Confirmation results at optimal process parameters with 95% prediction interval

Response	Predicted Mean	Predicted Median	Observed	Standard deviation	Standard Error of Prediction	95% PI low	95% PI high
UTS	114.563	114.563	114.98	0.695052	0.848677	112.672	116.454
% Strain	6.45818	6.45818	6.67	0.278806	0.34043	5.69966	7.21671
Microhardness	88.416	88.416	88.9	0.489101	0.597206	87.111	89.7723

4.3 Optimization of FSW Process Parameters and their Implications on Output Responses of FSWed Joints

The RSM-based model optimized the process parameters and output responses by generating 3D surface and 2D contour plots. The contour plots represented in Figures 4.3(d-f), 4.4(d-f), and 4.5(d-f) reveal optimal regions for second-order response behavior, used to identify and evaluate stationary points (maxima, minima, or saddle points) [46]. Leveraging the experimental findings illustrated in Figures 4.1(a-c) and 4.2(a-c), the model effectively predicted the influence of process parameters on microhardness, % strain, and UTS of the dissimilar AA6061-T6/AZ91D FSW joints. Figures 4.3(a-c), 4.4(a-c), and 4.5(a-c) show 3D response surfaces for parameter interactions, reflecting output variability across the design space.

Perturbation plots, which highlight the sensitivity of responses to individual variables, are shown in Figure 4.6(a-c). TRS shows pronounced curvature, confirming its dominant effect on mechanical properties. Based on the quadratic model, these plots enable evaluation of both main and interaction effects, aiding FSW optimization [47]. During the FSW of dissimilar AA6061-T6 and AZ91D metals, the multi-response optimization method simultaneously optimized various objective functions. Here, the UTS, % strain, and microhardness at the SZ served as the objective functions. According to the optimization results, the microhardness at the SZ was 88.4 HV0.1, the optimal strain value was 6.45%, and the optimal value of UTS was 114.56 MPa, as shown in the ramp chart, Figure 4.7. These optimal values have refined the FSW procedure to produce the required mechanical characteristics at the SZ.

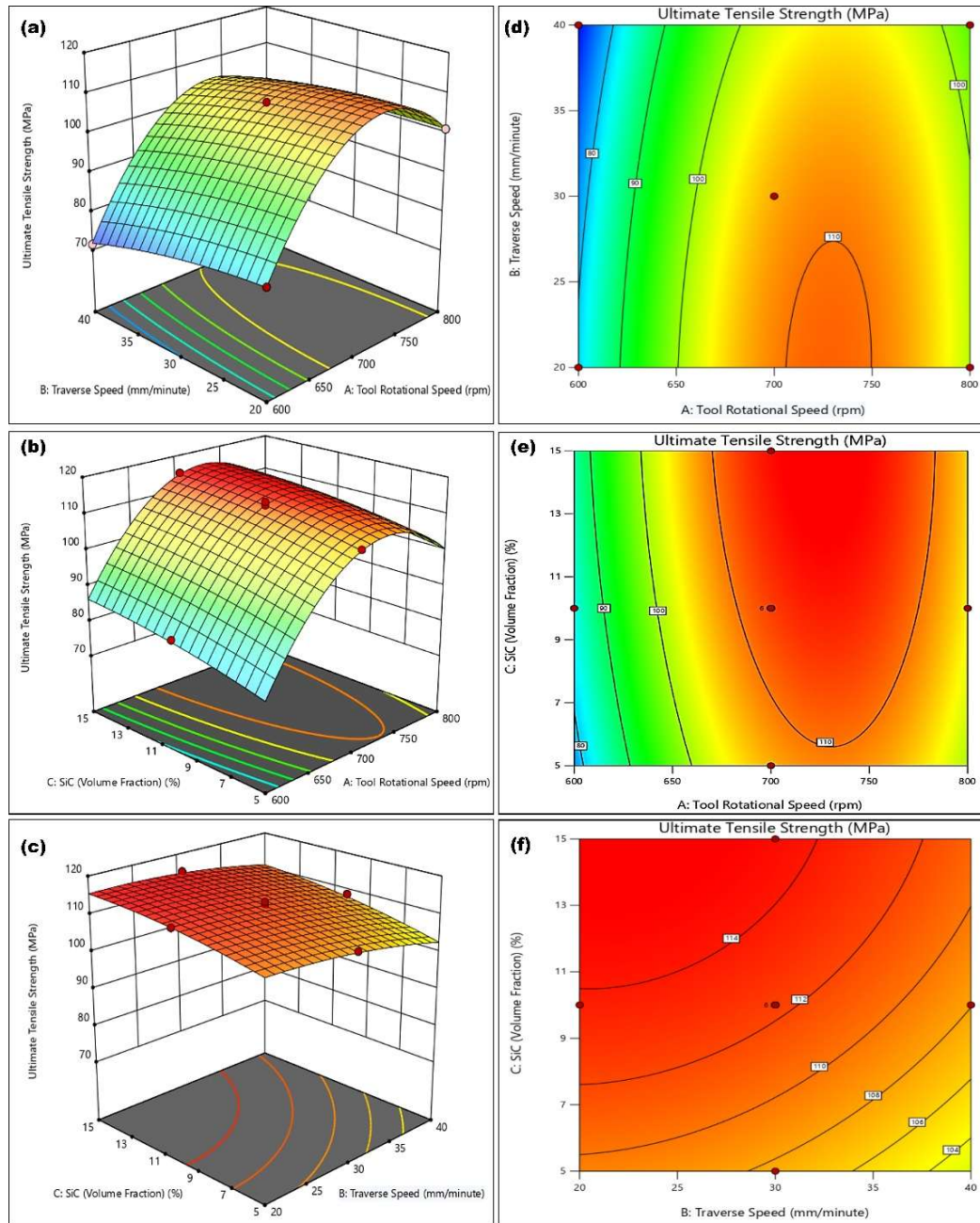


Figure 4.3:UTS :(a-c) 3D response surface plots, (d-f) 2D contour plots

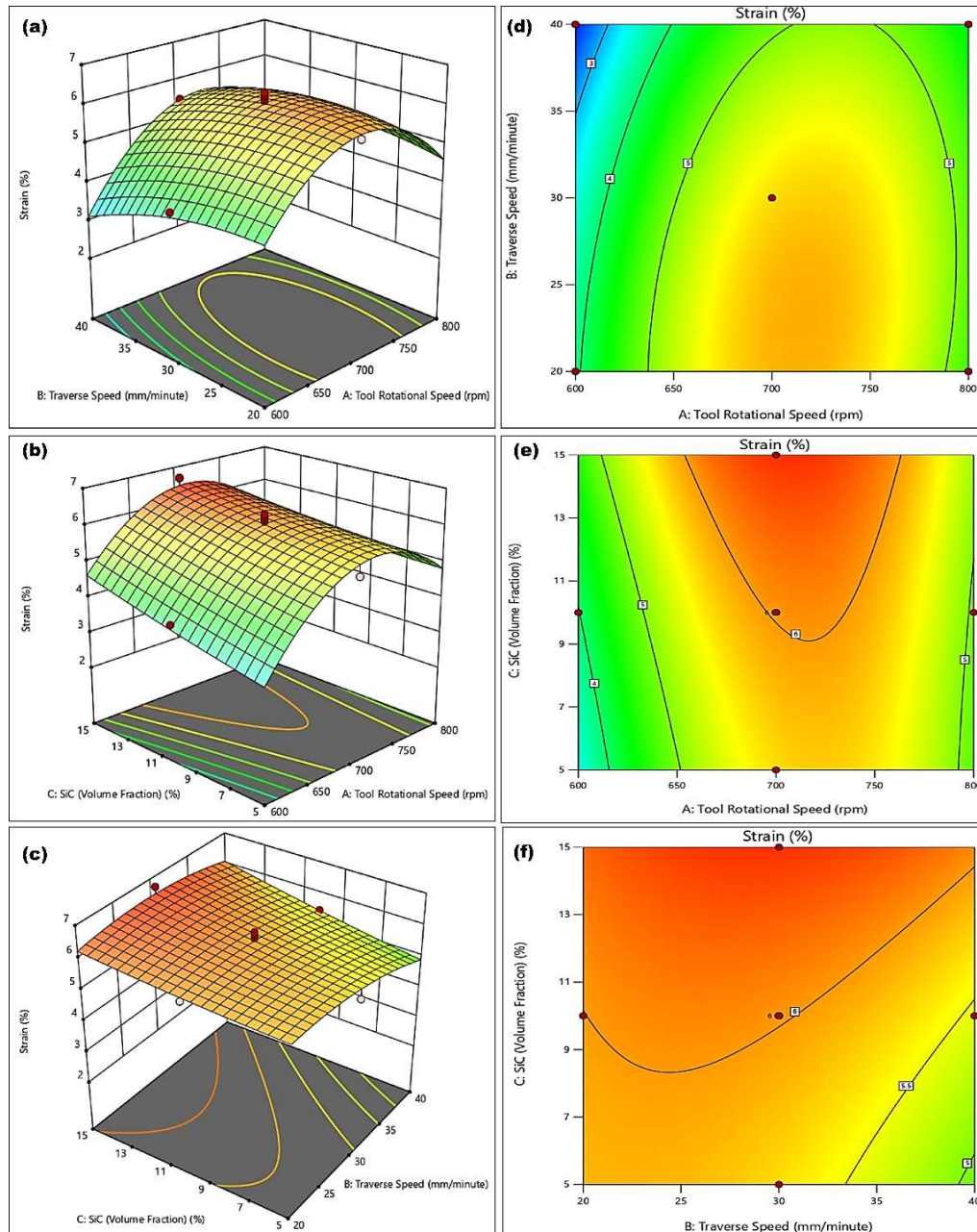


Figure 4.4: %Strain :(a-c) 3D response surface plots, (d-f) 2D contour plots

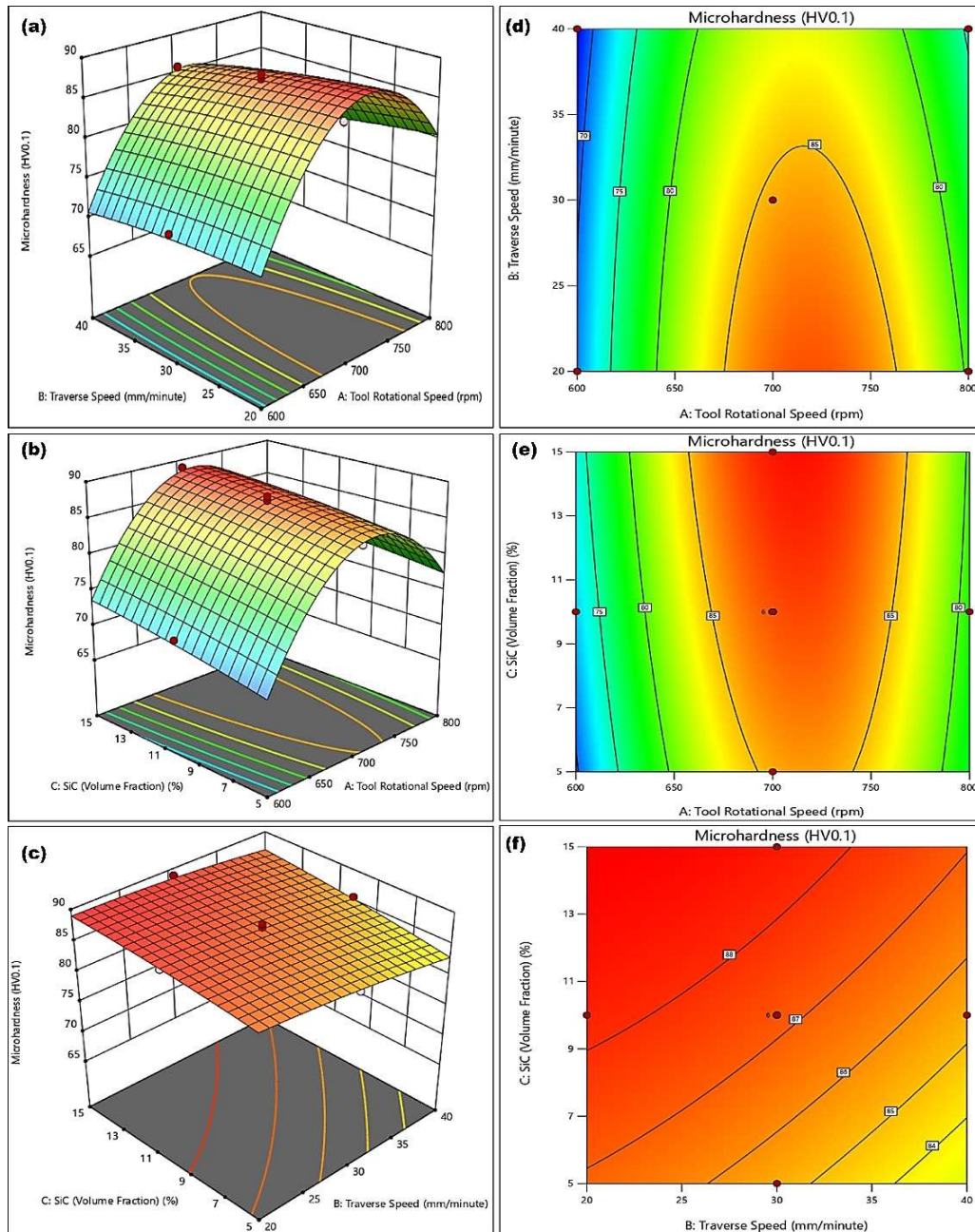


Figure 4.5. Microhardness : (a-c) 3D response surface plots, (d-f) 2D contour plots

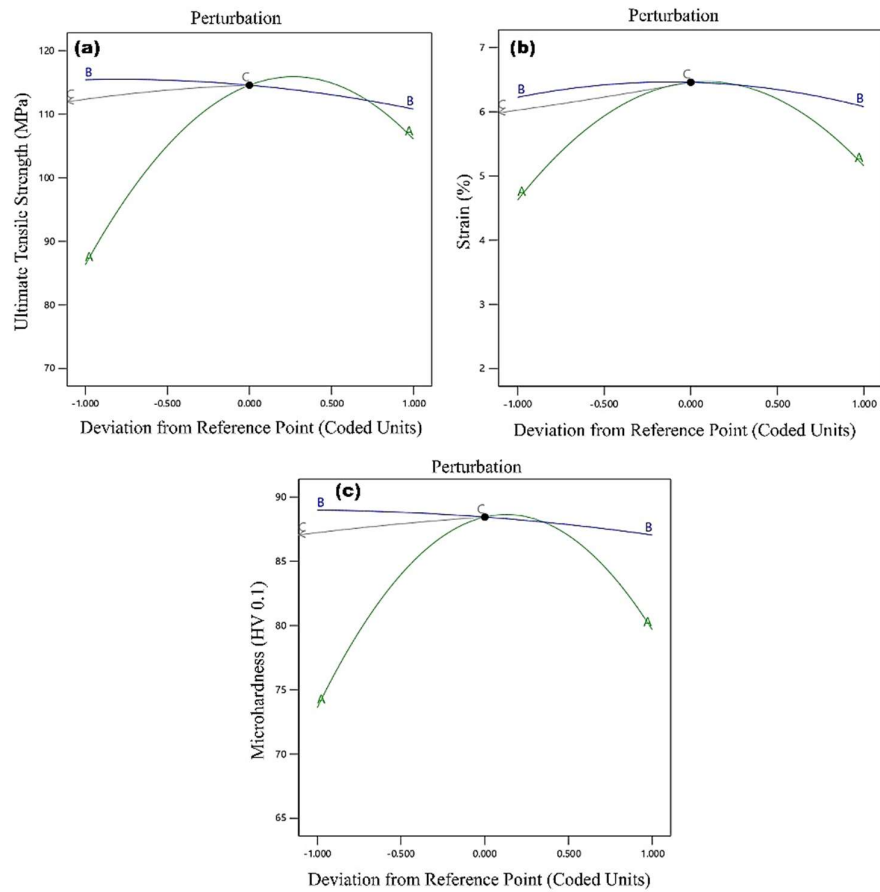


Figure 4.6: Perturbation Graphs (a) UTS, (b) % strain, (c) microhardness

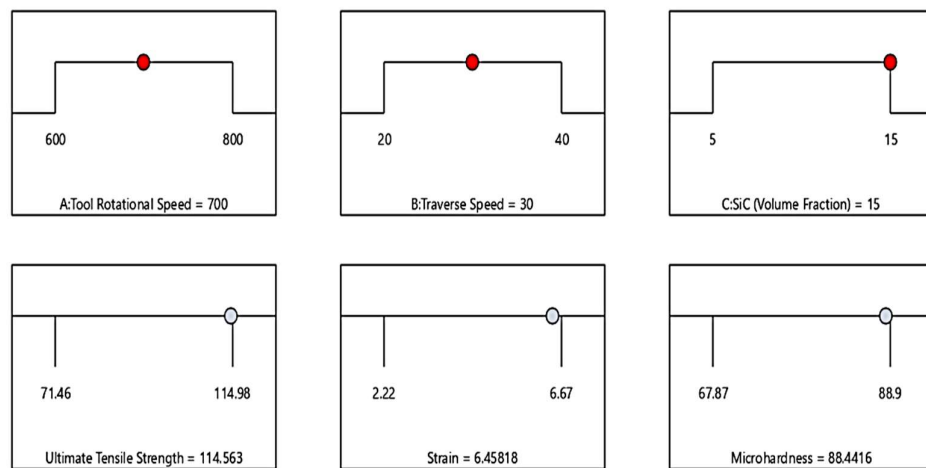


Figure 4.7: Ramp chart showing optimized values of input parameters and output responses

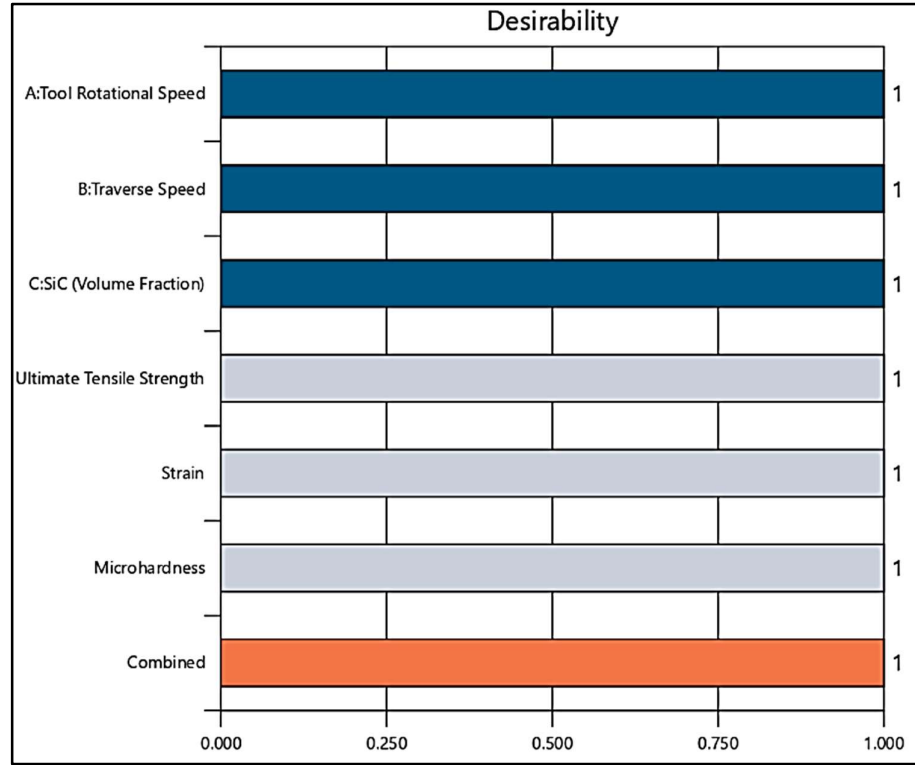


Figure 4.8: Desirability analysis showing individual responses and overall optimization

A desirability-based multi-response optimization was performed using Design-Expert software to identify the most favorable combination of FSW parameters. The desirability function (DF) method converts each response into a desirability score ranging from 0 (completely undesirable) to 1 (highly desirable), depending on the specific optimization goal. These individual desirabilities are then combined into a metric known as the overall (or composite) desirability index [90]. The composite desirability (D) is calculated as the geometric mean of the individual desirability, as shown in the following equation:

$$D = (d_1 \times d_2 \times d_3 \times \dots \times d_n)^{\frac{1}{n}} = (\prod_{i=1}^n d_i)^{\frac{1}{n}} \quad (4.8)$$

Where D is the overall or composite desirability and serves as the optimum value for various responses, d_i is the individual desirability of the i -th response, and n is the total number of responses considered. In the present investigation, a multi-objective desirability function approach was employed to optimize FSW parameters (TRS, TS, SiC Vf) and achieve a high desirability of the output responses. A desirability score of 1 (Figure 4.8) confirmed the method's effectiveness in identifying optimal conditions

for producing sound SiC-reinforced AZ91D/AA6061-T6 dissimilar joints. Additionally, the experimental results demonstrate that FSWed joints of dissimilar metals, welded at TRS 700 rpm, TS of 30mm/minute, and 15% SiC Vf, showed a higher UTS of 114.98 MPa, a % strain of 6.67%, and a microhardness of 88.90 HV0.1, respectively, at SZ. The results indicate a significant relationship between the experimental and predicted optimized conditions of the FSW parameters for dissimilar AA606-T6 and AZ91D joints.

Furthermore, the statistical analysis conducted in this study extends beyond traditional empirical approaches by establishing a predictive and optimization-driven model using RSM and CCD. This approach quantified the individual and interactive effects of key process parameters, enabling high-confidence predictions of mechanical responses. The model's strong adequacy metrics and ANOVA validation confirm its reliability, offering a structured pathway to optimize dissimilar FSW involving nanoparticle reinforcement. This area remains relatively underexplored in the existing studies.

4.4 Microstructural Analysis

The microstructural evolution of the base metals and the welded region, together with the distribution of SiC nanoparticles, was examined in detail to evaluate how the reinforcement influences grain refinement within the stir zone (SZ). Particular attention was given to changes in grain size, grain morphology, and second-phase distribution on both the AZ91D and AA6061-T6 sides, as these features strongly govern the mechanical response of dissimilar FSW joints. The dispersion of SiC within the SZ was found to be non-uniform under certain parameter combinations, reflecting the complex material flow patterns and the tendency of nanoparticles to agglomerate in regions of insufficient stirring or excessive local viscosity. Figures 3.2(a) and 3.2(c) present the optical microstructures of the as-received AZ91D and AA6061-T6 metals, respectively, revealing the initial grain morphology and providing a baseline for comparison with the welded condition. Figure 3.2(e) shows the SEM morphology of the SiC nanoparticles, while Figures 3.2(b), 3.2(d), and 3.2(f) display the EDS elemental analysis of base materials Al, Mg, and Si nanoparticles. Quantitative grain size measurements were performed using ImageJ software on multiple micrographs taken at representative

locations, and the average grain sizes of AZ91D and AA6061-T6 in the base metal condition were determined to be 57 μm and 38 μm , respectively.

During the friction stir welding (FSW) process, the rotating tool shoulder and pin impose intense mechanical stirring on the adjoining AA6061-T6 and AZ91D plates, generating both severe plastic deformation and localized heating within the stir zone. This combined thermo-mechanical action activates dynamic recrystallization (DRX), whereby the high dislocation density produced by plastic shear is progressively rearranged into subgrains and, eventually, into fine equiaxed recrystallized grains. As deformation continues, existing high-angle grain boundaries migrate and new boundaries form, leading to a significant breakup of the original coarse-grain structure of the base metals and the development of a refined microstructure in the SZ. The refinement effect is further enhanced in regions with well-distributed SiC nanoparticles, since these particles interact with moving grain boundaries and impede their growth, helping to stabilize a fine recrystallized grain size. Overall, the severe plastic deformation imposed by the tool, together with the elevated temperatures characteristic of FSW, promotes DRX as the dominant mechanism responsible for grain refinement and the observed transformation from heterogeneous base-metal microstructures to a more homogeneous, equiaxed grain structure in the stir zone [91,92].

Three essential components are required for the development of FSWed joints. TS, TRS, and TTA are some aspects that can affect the accumulation of layered plasticized substances behind a rotating device. These process parameters govern the balance between heat generation and material flow, thereby controlling how effectively the plastically softened material is transported from the advancing side to the retreating side and consolidated behind the tool. This is the key aspect of the material disturbance that occurs with plastic materials. When the tool rotational speed (TRS) is sufficiently high, adequate frictional heat is produced, and the material in the vicinity of the pin and shoulder is softened and extruded, forming characteristic onion-ring structures in the stir zone. Conversely, the traverse speed (TS) determines the time available for this softened material to be stirred and consolidated; excessively high TS can cause incomplete mixing and void formation, whereas very low TS may lead to overheating and grain coarsening. The tool tilt angle (TTA) facilitates the downward and backward forging of the plasticized material, promoting intimate contact between the plastically

deformed layers and enhancing metallurgical bonding. Together, these three parameters control the thickness, continuity, and uniformity of the deposited plasticized layers, which in turn dictate the final microstructure and integrity of the FSWed joint.

Firstly, understanding the differences in microstructure between the stir zone (SZ), heat-affected zone (HAZ), and thermomechanically affected zone (TMAZ) is crucial for evaluating the material's properties and performance. Three separate zones comprise FSW weld joints: the SZ (or nugget zone), where permanent deformation occurs; the TMAZ (or transition zone between the SZ and HAZ); and the HAZ (or heat cycle zone), which experiences heat due to the welding process. In this area, the material's microstructure and properties are susceptible to rapid temperature changes. Within the SZ, the combined effect of severe plastic deformation and elevated temperature results in intense dynamic recrystallization, producing a fine, equiaxed grain structure that is markedly different from the coarse grains observed in the base metals. In the TMAZ, deformation is significant but insufficient to fully recrystallize the microstructure, so the grains appear elongated and bent, reflecting partial recovery and subgrain formation. The HAZ is primarily influenced by the thermal cycle of FSW without substantial plastic strain, leading to microstructural changes dominated by grain growth, precipitate coarsening, and, in some cases, overaging, particularly on the AA6061-T6 side. As a consequence, each zone exhibits distinct grain morphology, phase distribution, and defect content, which collectively control local hardness, tensile response, and fracture behaviour. A clear understanding of these zonal differences is therefore essential for interpreting the mechanical performance of the dissimilar AZ91D/AA6061-T6 joints reinforced with SiC nanoparticles [93,94].

The second factor is the rise in temperature within the welding zone, which ultimately makes the base plate in the SZ more pliable. As the rotating tool advances, friction between the shoulder and the workpiece surface, combined with plastic deformation around the pin, generates heat that elevates the local temperature to a substantial fraction of the alloy's melting point. This controlled temperature increase reduces the flow stress of both AA6061-T6 and AZ91D, allowing them to be stirred and extruded without actual melting, which is a hallmark of the solid-state nature of FSW. However, the heat input must remain within an optimal window: insufficient heating leads to poor

plasticization and a lack of bonding, whereas excessive heating promotes grain coarsening and excessive growth of intermetallic compounds at the Al-Mg interface.

As a final point of consideration, the third component involves the plasticized hot forging made possible by the tool shoulder. The tool shoulder applies a downward forging force that consolidates the plasticized material behind the pin, eliminating voids and promoting intimate contact between the freshly formed layers. The combined actions of stirring by the pin and forging by the shoulder generate a characteristic layered structure in the stir zone, often referred to as onion rings, which reflects the periodic deposition of material as the tool advances. Proper control of the shoulder plunge depth and tilt angle ensures that sufficient hydrostatic pressure is maintained during welding, thereby improving metallurgical bonding and reducing the risk of internal defects. The culmination of integrating these components is a welded connection that is both robust and dependable. In the present dissimilar joints, achieving the correct balance of heat generation, material plasticization, and forging action is especially critical due to the different thermal conductivities, flow stresses, and melting ranges of AZ91D and AA6061-T6, which could otherwise lead to asymmetric flow and defect formation.

Utilizing the DRX process during FSW made it possible to create a sophisticated and uniform grain structure at increased TRS. Higher tool rotational speeds intensify the stirring action and raise the peak temperature in the SZ, which in turn accelerates the nucleation and growth of dynamically recrystallized grains. Nevertheless, the beneficial effect of increasing TRS is only realized when it is combined with an appropriate traverse speed, so that the strain rate and thermal cycle remain within the range that promotes continuous DRX without causing overheating or abnormal grain growth. The dynamic movement of the instrument, in conjunction with the elevated heat conditions, causes plastic deformation, which in turn encourages the DRX of grains. As a result, the original coarse grains of the base metals are fragmented into smaller subgrains, and repeated cycling of deformation and recovery gradually transforms these subgrains into fine, equiaxed recrystallized grains. This refined microstructure, particularly when stabilized by well-dispersed SiC nanoparticles, forms the basis for the improved mechanical properties observed in optimally processed joints [10,30].

The microstructure exhibits larger grains due to the moderate strain rate within the HAZ. The development of recrystallized grains in the SZ, coupled with DRX, is promoted by the mechanical agitation of the materials via the tool pin and the frictional heat produced from the tool shoulder's contact. In the present joints, the contrast between the fine, equiaxed grains in the SZ and the relatively coarser grains in the HAZ is especially pronounced due to the dissimilar thermal properties and initial microstructures of AZ91D and AA6061-T6. The magnesium alloy, with its cast microstructure and second-phase β -Mg₁₇Al₁₂ particles, is more prone to local melting and grain coarsening in the HAZ, whereas the precipitation-strengthened aluminium alloy is susceptible to precipitate dissolution and overaging, which reduces hardness despite limited grain growth. Within the SZ, repeated stirring by the rotating tool continuously fragments the original grains, distributing second-phase particles and SiC nanoparticles throughout the nugget, thereby facilitating continuous dynamic recrystallization. The synergy between frictional heating from the shoulder, plastic deformation around the pin, and the presence of SiC particles, therefore, leads to a highly refined microstructure in the SZ, while only partial microstructural modification occurs in the surrounding TMAZ and HAZ.

The thermo-mechanically affected zone (TMAZ) and heat-affected zone (HAZ) exhibit a coarser grain morphology compared to the stir zone (SZ), which is attributed to their exposure to lower temperatures and strain rates during friction stir welding (FSW) [10,30]. In these regions, the imposed deformation and thermal cycle are insufficient to produce the same level of grain subdivision observed in the SZ, and therefore, only partial recovery and limited recrystallization occur. As a result, grains tend to be elongated or mildly coarsened, especially near the interface between AZ91D and AA6061-T6, where the thermal and mechanical fields are strongly asymmetric. This microstructural gradient from fine, equiaxed grains in the SZ to coarser, deformed grains in the TMAZ and HAZ has a direct influence on local mechanical properties, promoting hardness peaks in the nugget and softer bands in the adjacent regions.

The FSW results in a heterogeneous microstructure because it causes the temperature and plastic strain that the material experiences to vary within the SZ. Even inside the nugget, the material near the tool pin, shoulder interface, advancing side, and retreating side is subjected to different combinations of strain rate and peak temperature.

Consequently, distinct sub-regions can form, such as a fine-grained central nugget, slightly coarser peripheral zones, and locally deformed bands associated with the characteristic onion-ring pattern. The presence of SiC nanoparticles further enhances this heterogeneity, as areas with high particle density experience stronger grain boundary pinning and more pronounced refinement than particle-lean regions. The material surrounding the SZ undergoes dynamic recrystallization (DRX) due to the high maximum temperature and plastic deformation present in the material. Observations made in SZ revealed a finer-grained structure. This refined structure, stabilized by dispersed SiC, provides a favourable microstructural basis for improved strength and hardness compared with the surrounding zones.

SEM micrographs of the SZ, shown in Figure 4.9(a-d), illustrate how different parameters influence particle dispersion. These images systematically capture the evolution of SiC distribution across the critical parameter space tested, revealing distinct morphological transitions from a clustered to a homogeneous state. At 600 rpm and 40 mm/min (Specimen 5, 15% SiC), SEM (Figure 4.9a) revealed SiC agglomeration and weak bonding from low heat input and inadequate material mixing. The clustered nanoparticles create micron-scale regions of high particle density surrounded by particle-free matrix, which not only disrupts uniform load transfer but also serves as preferential sites for void nucleation during subsequent mechanical testing. Insufficient heat leads to elevated material viscosity in the SZ, hindering consistent mixing. Under these conditions, the plasticized material exhibits limited flowability, which prevents the effective breakup and redistribution of SiC agglomerates formed during initial groove filling. TRS and TS values significantly impact the heat input in FSW, with larger TRS and lower TS resulting in increased heat generation [8, 10, 25, 32, 33, 19].

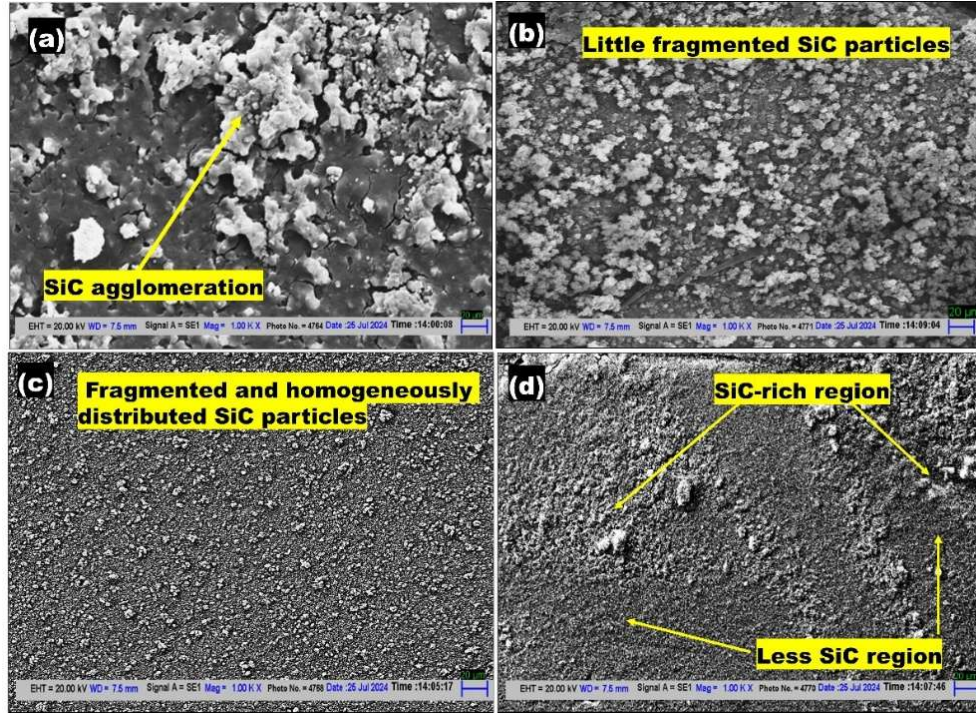


Figure 4.9: SEM micrographs showing SiC dispersion in SZ at various process parameters (a) Specimen 5, (b) Specimen 1, (c) Specimen 6, (d) Specimen 12

At elevated tool rotational speeds (TRS) and reduced traverse speeds (TS), increased heat generation and plastic strain promote uniform SiC dispersion within the stir zone (SZ) [10]. This optimal thermal-mechanical regime lowers the material's viscosity sufficiently to allow the fragmentation of initial SiC agglomerates and their redistribution throughout the stirred volume via vortex-like flow patterns. Adequate heat input ensures effective material flow and integration of reinforcement particles. The enhanced fluidity facilitates particle breakup against the tool pin surface and subsequent entrainment into the plasticized matrix, preventing re-agglomeration. Increasing the TRS from 600 to 700 rpm and reducing the TS from 40 to 30 mm/min enhanced particle fragmentation and dispersion, preventing agglomeration and resulting in strong matrix bonding (Figure 4.9b, c). Consequently, increased TRS and reduced TS improved particle mobility [10]. The resulting uniform dispersion eliminates stress concentration sites and enables effective load partitioning between the soft matrix and hard SiC reinforcement during deformation. Compared with joints containing 5% (Specimen 13) and 10% (Specimen 1) SiC Vf, the 15% SiC Vf joint (Specimen 6) exhibited denser, homogeneous distribution under identical

parameters, achieving optimal dispersion at 700 rpm, 30 mm/min, and 15% SiC Vf (Figure 4.9c) due to effective strain and SiC-induced grain pinning [31,34,43]. Higher volume fractions increase the probability of particle-particle interactions during flow, which paradoxically aids dispersion by creating a self-stirring network that resists settling. The 15% Vf condition represents the limit where pinning pressure exceeds flow resistance without causing excessive tool wear or flow restriction. These observations highlight the role of SiC nanoparticles in the FSW of dissimilar metals. Beyond mechanical reinforcement, SiC acts as a heterogeneous nucleation site for recrystallized grains, modifying local flow stress gradients across the Al-Mg interface and promoting more uniform material intermixing.

At 800 rpm and 20 mm/min with 15% SiC (Specimen 12), excessive heat caused turbulent flow and coarser SZ grains, showing particle-rich and particle-less regions (Figure 4.9d) [32,43]. The combination of high rotational speed and very low traverse speed generates peak temperatures approaching the solidus limits of both alloys, leading to flow instabilities characterized by vortex shedding and material expulsion from the stir zone. This results in banded microstructures with alternating high-SiC and SiC-depleted layers, which compromise uniform load transfer and create preferential crack paths under tensile loading. Increasing SiC Vf from 5% to 15% consistently refined grains, whereas increasing 20 to 25% Vf weakened composites by restricting flow, lowering plasticization, and accelerating tool wear [27,43]. At volume fractions above 15%, the high particle density creates a slurry-like consistency that resists shear thinning, leading to incomplete consolidation and increased frictional forces on the tool surfaces. Preliminary trials confirmed that 20-25% Vf conditions produced surface galling and pin breakage, justifying the selection of 15% as the maximum tested volume fraction. Following the review of the results from these researchers and based on preliminary trials, our volume fraction was restricted to a maximum of only 15%. This conservative approach ensured process stability while maximizing the benefits of nanoparticle reinforcement for microstructural control and mechanical enhancement.

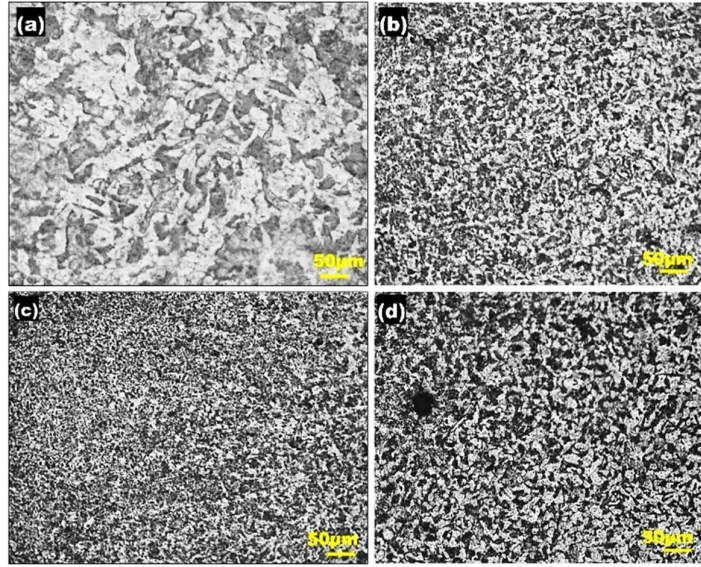


Figure 4.10: Optical microstructure showing grain formation at various process parameters in SZ, (a) Specimen 5, (b) Specimen 1, (c) Specimen 6, (d) Specimen 12

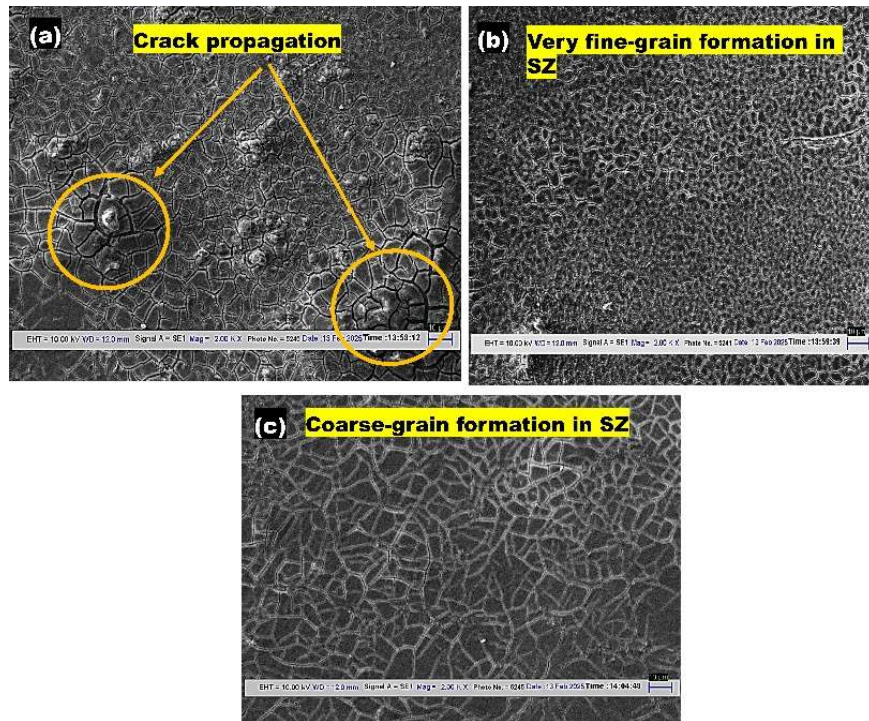


Figure 4.11: SEM micrographs showing grain formation at various process parameters in SZ, (a) Specimen 5, (b) Specimen 6, (c) Specimen 12

These optical (Figure 4.10) and SEM micrographs (Figure 4.11) systematically document the progressive refinement of the stir zone microstructure across the parameter matrix, providing direct visual evidence of the interplay between process conditions and SiC reinforcement on grain morphology and size distribution.

FSW with SiC nanoparticle reinforcement resulted in a refined grain structure in the stir zone (SZ) compared to the base materials, as shown in optical micrographs (Figure 4.10a-d) and SEM images (Figure 4.11a-c). The systematic progression from coarse, heterogeneous base metal grains (57 μm AZ91D, 38 μm AA6061-T6) to sub-equiaxed structures demonstrates the effectiveness of combined thermo-mechanical processing and particle reinforcement. Grain refinement resulted from dynamic recrystallization (DRX) and the pinning effect of dispersed and fractured SiC. While DRX provides the primary mechanism for grain subdivision through dislocation rearrangement and boundary migration, the SiC nanoparticles exert a secondary stabilizing effect by anchoring high-angle grain boundaries, preventing post-deformation coarsening even under prolonged elevated temperatures.

The uniform dispersion of second-phase particles inside the matrix is a popular technique for controlling grain size. This approach leverages the fundamental interaction between migrating grain boundaries and incoherent second-phase particles, where the particles exert a drag force proportional to their surface area density within the boundary plane. According to Smith [95], Zener defined the limiting grain size as the balance between the drag force exerted by particles and the pushing force required for grain expansion. The classical Zener model posits that a stable grain size is attained when the boundary curvature driving pressure precisely balances the particle pinning pressure, resulting in a self-limiting microstructure that remains independent of further deformation. Additional research has developed this idea [96,97]. Modern modifications incorporate particle spacing statistics and boundary-particle disengagement mechanisms, providing more accurate predictions for high-volume fraction reinforcements, such as the 15% SiC used in this study. Humphreys et al. [98] state that Equation (4.9) represents the computations made using second-phase dispersed particles to determine the grain formation rate during the recrystallization of metals.

$$\frac{dR_g}{dt} = M_b(P_d - P_{zp}) = M_b \left(\frac{g_c \gamma_{be}}{R_g} - \frac{3V_f \gamma_{be}}{2r_p} \right) \quad (4.9)$$

Where R_g represents grain's growth, M_b represents grain's boundary mobility, P_d represents driving pressure originating from the curvature of boundaries of grain, P_{zp} represents Zener pinning pressure, g_c represents geometrical constant, γ_{be} represents energy associated with grain boundary, V_f and r_p represent volume fraction and radius of second phase particles, respectively.

In the end, grain growth ceases when $P_d = P_{zp}$.

If the average mean granular radius D equals the curvature radius (R_g), then the Zener limiting grain size is determined ($g_c = 1$).

$$\frac{g_c \gamma_{be}}{R_g} - \frac{3V_f \gamma_{be}}{2r_p} \quad (4.10)$$

Then,

$$D_{zl} = \frac{4r_p}{3V_f} \quad (4.11)$$

Specimen 5 (Figures 4.11a) and Specimen 12 (Figure 4.11c) exhibited average SZ grain sizes of 16.67 μm and 10.3 μm , respectively, whereas Specimen 6 (Figure 4.11b) showed finer grains of 4.9 μm due to uniform SiC dispersion and increased volume fraction. The progressive refinement from 16.67 μm to 10.3 μm to 4.9 μm across these specimens directly correlates with the increasing effectiveness of particle pinning pressure ($P_{zp} \propto V_f/r_p$), where specimen 6 benefits from both the maximum V_f of 15% and optimal dispersion that minimizes ineffective clustering. Grain refinement occurs through particle-induced grain boundary pinning, consistent with previous studies [32, 43].

Moreover, the FSWed joints without SiC nanoparticles, processed at a TRS of 700 rpm, a TS of 30 mm/min, and a tool tilt angle of 2° , exhibited coarser grains in the stir zone compared to the SiC joints (Figure 4.12(a)). The non-reinforced sample exhibited a larger mean grain size (18.67 μm), compared to the refined grains in the silicon carbide-reinforced samples, highlighting the significant role of nanoparticle addition.

Without pinning particles, the recrystallized grains undergo continuous growth during the final stages of cooling, eliminating the Zener limiting mechanism and resulting in coarser structures that correlate with the reduction in joint efficiency.

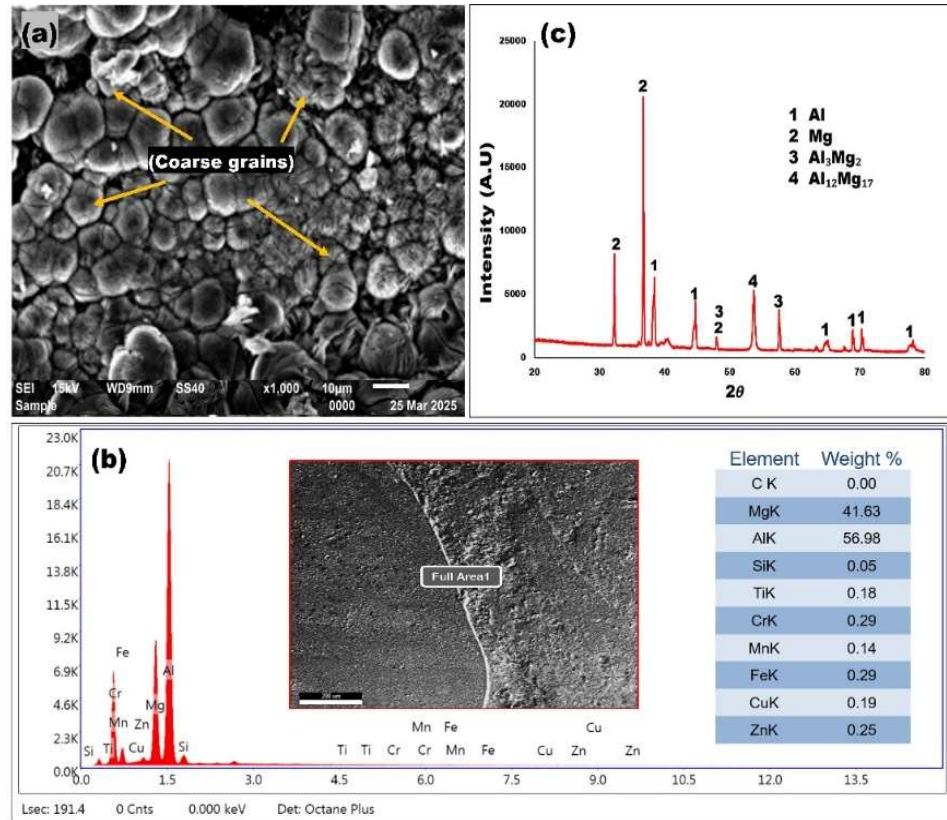


Figure 4.12: Non-SiC reinforced samples at 700 rpm, TS of mm/min, and a tool tilt angle of 2° (a) SEM image (b) EDS (c) XRD analysis

The EDS maps and spectra provide quantitative verification of SiC incorporation and elemental interdiffusion across the Al-Mg interface, revealing spatial variations in Si, Al, and Mg concentrations that correlate directly with the observed microstructural features and mechanical performance differences between specimens. EDS analysis (Figures 4.13 and 4.14) confirms the presence of SiC, with a higher Si percentage in specimen no. 6 (Figure 4.13) compared to specimen no. 1 (Figure 4.14), indicating varying SiC incorporation. It also confirms the presence of aluminum and magnesium in all samples, both with and without nanoparticles (Figures 4.12b, 4.13, and 4.14). In contrast, silicon was detected only in the nanoparticle-reinforced samples produced by

FSW. This verifies the elemental composition and reinforces the significance of processing parameters and silicon carbide (SiC) inclusion in controlling microstructure and mechanical properties.

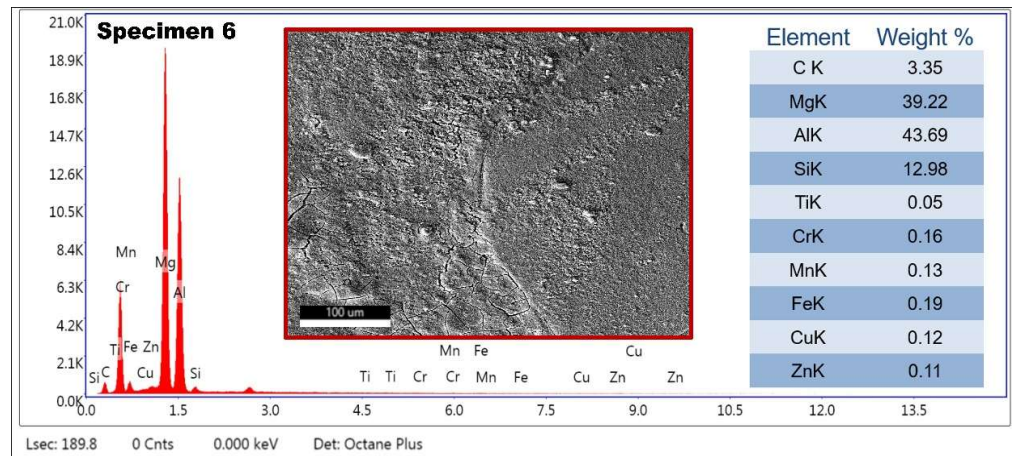


Figure 4.13: Figure showing EDS elemental composition of FSWed Specimen 6 in SZ

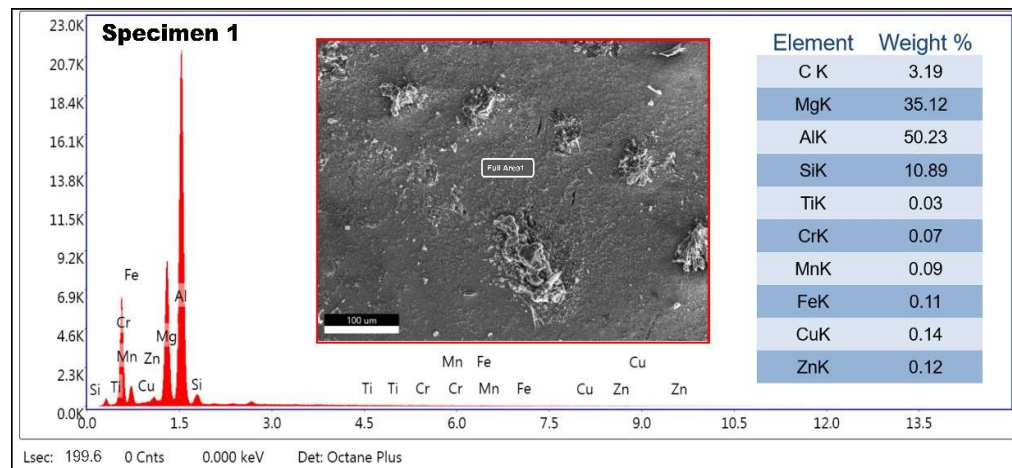


Figure 4.14: Figure showing EDS elemental composition of FSWed Specimen 1 in SZ

The X-ray diffraction analysis (Figure 4.15) revealed multiple reflections corresponding to aluminum and magnesium phases across the 2θ range of 30° to 80° . Prominent peaks of Al were observed at approximately 38.5° , 44.7° , 65.1° , and 78.2° , while Mg exhibited strong reflections near 32.2° , 34.4° , 36.6° , and 57.3° . Intermetallic compounds Al_3Mg_2 and $\text{Al}_{12}\text{Mg}_{17}$ were identified within the 34° - 36° and 44° - 46° 2θ regions. IMCs peaks were most pronounced in the non-reinforced FSWed samples,

indicating extensive aluminum-magnesium diffusion and growth of intermetallic compounds in the absence of nanoparticles, as shown in Figure 4.12(c). In SiC-reinforced welds, additional peaks were detected around 35.6°, 41.4°, 60°, and 71.8°, confirming the incorporation of β -SiC particles. SiC-reinforced welds displayed additional reflections at 35.6°, 41.4°, 60°, and 71.8°, corresponding to a 2θ interval, verifying the incorporation of SiC. The weak intensity of SiC peaks is attributed to their partial encapsulation within the aluminum-magnesium matrix and the overlap of their diffraction signals with the stronger reflections of aluminum and magnesium. Reinforced welds, regardless of parameter variation, exhibited a balanced distribution of aluminum, magnesium, Silicon Carbide, and intermetallic compound phases, consistent with microstructural refinement. Similar observations have been reported by several other researchers as well [27, 29].

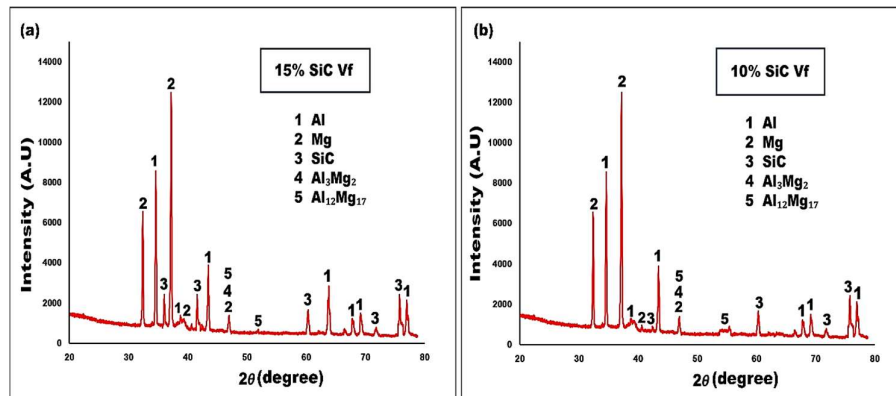


Figure 4.15: XRD peaks at TRS 700, TS 30 mm/minute with 15% and 10% SiC Vf (specimens 6 and 1)

Weld characteristics were governed by particle dispersion, dislocation density, and IMC morphology. These three microstructural features interact synergistically to determine load partitioning, crack propagation resistance, and overall joint efficiency under service conditions. IMCs such as Al_3Mg_2 and $\text{Al}_{12}\text{Mg}_{17}$ formed via Al-Mg interdiffusion under thermal cycles, appearing as dark interfacial bands in optical micrographs (Figure 4.16a-d) [2,12]. The layered morphology reflects the sequential deposition of interdiffused material during tool passage, with varying band thicknesses depending on the local temperature history and diffusion time. SiC nanoparticles suppressed but did not eliminate IMC formation [10]. Particle dispersion creates physical barriers to diffusional flux and locally modifies the thermal field, reducing the continuity of the

IMC layer from fully lamellar (unreinforced) to a fragmented, island-like morphology in optimally reinforced joints. Specimen 5 (600 rpm, 40 mm/min, 15% Vf) showed severe particle clustering and thick, continuous IMCs; Specimen 1 (700 rpm, 30 mm/min, 10% Vf) displayed thinner, discontinuous IMCs with higher strength; and Specimen 6 (700 rpm, 30 mm/min, 15% Vf) achieved uniform SiC dispersion, fragmented IMCs, and optimal mechanical response. At 800 rpm, 20 mm/min, and 15% Vf (Specimen 12), overheating resulted in continuous IMCs and particle-rich/lean zones. XRD (Figure 4.15a-b) identified Mg, Al, Al_3Mg_2 , and $\text{Al}_{12}\text{Mg}_{17}$ phases, corresponding to eutectic points near 435 °C ($\text{Mg} + \text{Al}_{12}\text{Mg}_{17}$) and 450 °C ($\text{Al} + \text{Al}_3\text{Mg}_2$) [38]. Typically, Al_3Mg_2 formed near AA6061-T6, and $\text{Al}_{12}\text{Mg}_{17}$ near AZ91D. Increasing Vf from 5% to 15% reduced IMC thickness, though agglomeration (Specimen 5) or overheating (Specimen 12) negated this benefit. Higher TRS or lower TS enhanced diffusion, thickening IMCs, whereas balanced parameters and moderate IMC formation improved metallurgical bonding. Hence, a 10-15% Vf with optimized TRS/TS produced a refined microstructure, controlled IMC growth, and superior joint integrity.

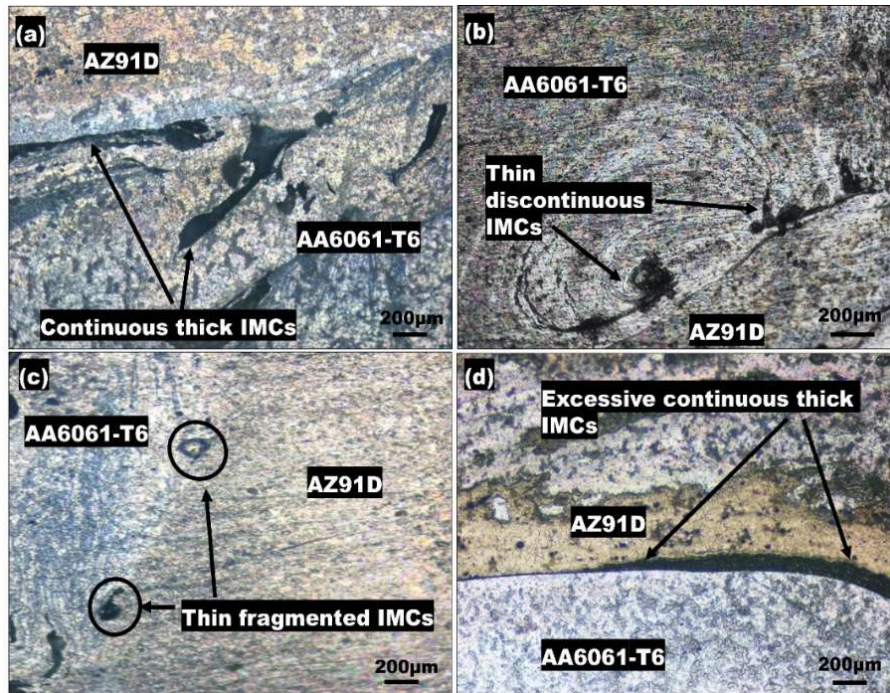


Figure 4.16: Cross-sectional optical micrographs showing representative IMC layers at the Al-Mg interface in SiC-reinforced FSW joints, (a) Specimen 5, (b) Specimen 1, (c) Specimen 6, (d) Specimen 12

4.5 Effect of SiC Nanoparticles on Tensile Strength

The mechanical performance of FSWed AA6061-T6/AZ91D joints without nanoparticles primarily depends on grain size, whereas SiC-reinforced joints are influenced by particle size, distribution, and interfacial bonding [10]. Mechanical results, including tensile strength, strain, microhardness, and joint efficiency, are presented in Table 4.8.

Table 4.8: Mechanical properties of FSWed joints

FSW welded specimen number	Process Parameters			Mechanical properties			
	A: Tool Rotational Speed (rpm)	B: Traverse Speed (mm/min)	C: SiC (Volume Fraction) (%)	UTS (MPa)	Strain (%)	Microhardness (HV0.1)	Joint Efficiency (%)
1	700	30	10	113.33	6.3	88	51.75
2	700	40	10	108.78	5.45	85.67	49.67
3	800	30	10	104.74	4.7	78.56	47.83
4	600	20	15	88.12	4.44	73.45	40.24
5	600	40	15	81.27	4	72.47	37.11
6	700	30	15	114.98	6.67	88.9	52.50
7	700	30	10	111.49	6.11	86.89	50.91
8	600	40	5	71.46	2.22	67.87	32.63
9	700	30	10	112	5.9	87.01	51.14
10	700	30	10	111.12	5.67	86.5	50.74
11	700	30	10	112.41	6.23	87.44	51.33
12	800	20	15	105.89	4.8	80.78	48.35
13	700	30	5	107.89	5.38	85.1	49.26
14	700	20	10	113.75	5.88	88.12	51.94
15	800	40	5	96.9	4.6	74.56	44.25
16	600	30	10	83.96	4.11	72.5	38.34
17	800	40	15	103.21	4.89	77.69	47.13
18	800	20	5	101.23	4.68	79.89	46.22
19	600	20	5	80.94	4.12	70.45	36.96
20	700	30	10	112.23	6.15	87.12	51.25

The base metals showed tensile strengths of 320 MPa (AA6061-T6) and 219 MPa (AZ91D) with strain rates of 11% and 9.5%, respectively. The FSWed joints exhibited UTS values ranging from 71.46 MPa to 114.98 MPa depending on process conditions. The stress-strain behavior under tensile loading is shown in Figures 4.17(a-c). Grain refinement, SiC pinning, and interfacial bonding primarily determined the tensile strength. The highest UTS of 114.98 MPa was achieved for Specimen 6 (700 rpm TRS, 30 mm/min TS, 15 % Vf SiC) due to improved SiC dispersion and pinning, which restricted dislocation motion and grain growth.

In contrast, the agglomerated SiC particles in Specimen 5 acted as crack nucleation sites (Figure 4.11a), resulting in a UTS of 81.27 MPa, which is consistent with the trends reported by Bahrami et al. [33]. The optimized parameters (700 rpm, 30 mm/min, 15% Vf) produced a higher UTS than several reported Mg/Al FSW systems. Tabasi et al. [10] (86 MPa, AA7075/AZ31 + SiC), Azizieh et al. [37] (96 MPa, AA1100/AZ31 without reinforcement), Sharifi Asl et al. [30] (107 MPa, AA6063-T4/AZ31B-O + TiO₂), and Mehdi et al. [38] (110.32 MPa, AZ91D/AA7075 without SiC). Within the tested TRS range (600-800 rpm), 700 rpm consistently yielded the highest strength, confirming SiC's beneficial effect under moderate, industrially viable conditions.

SiC nanoparticles also suppressed brittle Mg-Al IMC formation, promoting finer, uniform Al-Mg IMCs and improved metallurgical bonding [33]. Smaller SiC particles refined the microstructure and reduced porosity, enhancing compressive strength, while larger ones promoted defects [10]. Thus, SiC incorporation significantly enhanced the tensile strength, microstructural uniformity, and joint integrity of dissimilar AZ91D/AA6061-T6 FSW joints. Additionally, at 30 mm/min TS and 700 rpm TRS and 2° TTA, the unreinforced joint exhibited a UTS of 80.12 MPa and a strain of 3.37%, which is lower than that of the nanoparticle-reinforced joints under these parameters, indicating that SiC enhances both mechanical and metallurgical properties.

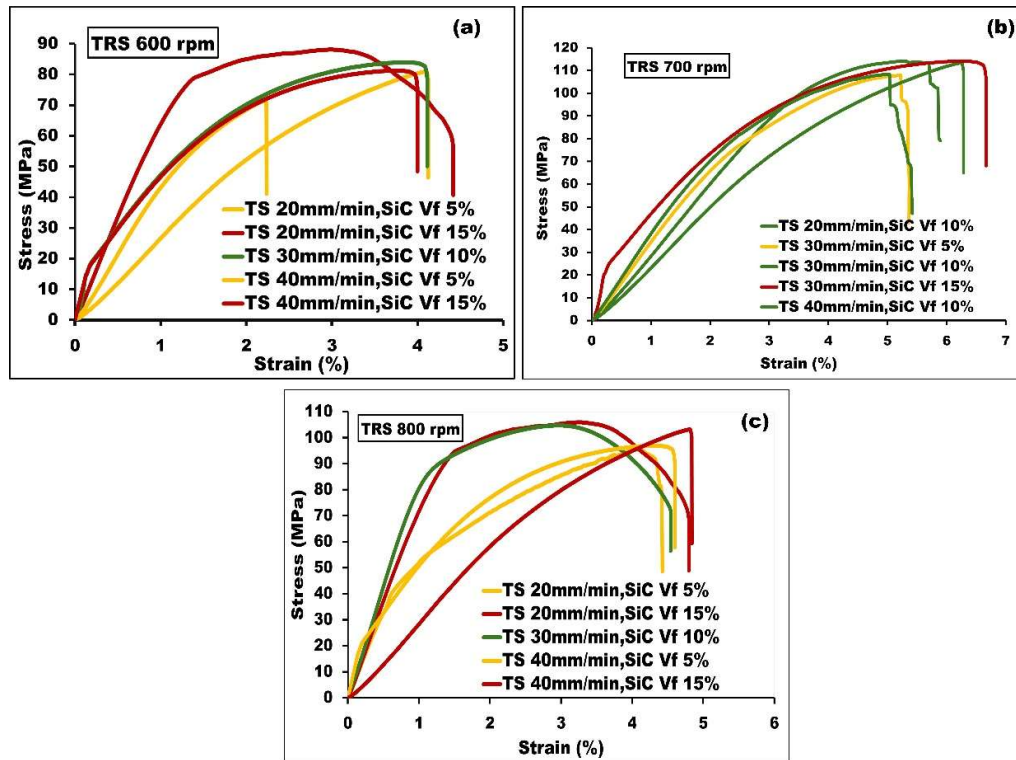


Figure 4.17: (a-c) Stress-strain diagram of FSWed joints at varying SiC Vf

4.6 Effect of SiC Nanoparticles on Microhardness

The microstructural characteristics of the materials are closely related to their hardness. A material's microstructure refers to the distribution, size, and arrangement of its constituent particles, such as grains or phases. A substance's hardness indicates its resistance to deformation, particularly when it comes to scratches or indentations. A material with a fine-grained microstructure typically demonstrates superior hardness compared to a coarse-grained one. This phenomenon occurs as tiny grains further impede dislocation movement, aggravating the deformation process.

During FSW, the DRX facilitates the development of refined grains, potentially resulting in increased hardness values within the SZ. Plastic deformation, which affects hardness values, is amplified by DRX, grain boundaries, and dislocations in the SZ. Hardness in the SZ is increased due to these structural characteristics, which intensify plastic deformation. The welded joints' microhardness can be impacted by the creation

of intermetallic compounds (IMCs) during FSW of dissimilar metals. Grain size, dislocation density, and phase precipitations impacted the SZ's microhardness [99].

The dispersion of hardness was not uniform across the TMAZ and HAZ. Stir Zones' microhardness was lower than that of AA6061-T6 but more significant than that of AZ91D. Variations in the hardness of specific sites may be associated with SiC-rich zones or the development of IMCs. The difference in hardness in the welded area is attributed to the presence of intercalated microstructures. The hardness variation is a result of these microstructures. In contrast, the HAZ, experiencing localized thermal softening, exhibited reduced hardness compared to the base metals.

The presence of these robust phases serves as an impediment, obstructing dislocation movement and enhancing the material's ability to withstand deformation. Consequently, understanding the material's microstructure is essential for predicting and controlling its microhardness characteristics. The hardness of joints produced by FSW is influenced by many factors, including welding parameters, tool design, material properties, etc.

The base metals AA6061-T6 and AZ91D had mean hardness values of 105 HV0.1 and 66 HV0.1, respectively. Specimen 6 exhibited a peak hardness of 88.9 HV0.1 at 700 rpm RS, 30 mm/min TS, and 15% SiC Vf. Hardness variation across the weld zone is influenced by grain refinement, SiC dispersion, and volume fraction [36]. Figure 4.18(a-c) illustrates the microhardness at various parameters, confirming an increase in hardness with the optimal SiC content. Smaller grains in well-reinforced specimens lead to higher hardness.

Lower TS increased heat input, resulting in reduced TMAZ hardness, which is consistent with prior studies. The SiC strengthening effect was most pronounced in specimen 6 due to its complex nature, grain refinement, and intermixing. Microhardness was highest in SZ due to finer grains, while TMAZ hardness dropped due to precipitate dissolution. In HAZ, precipitate coarsening further reduced hardness [4, 41, 42]. Grain refinement, high SiC hardness, and dispersion strengthening enhanced SZ hardness.

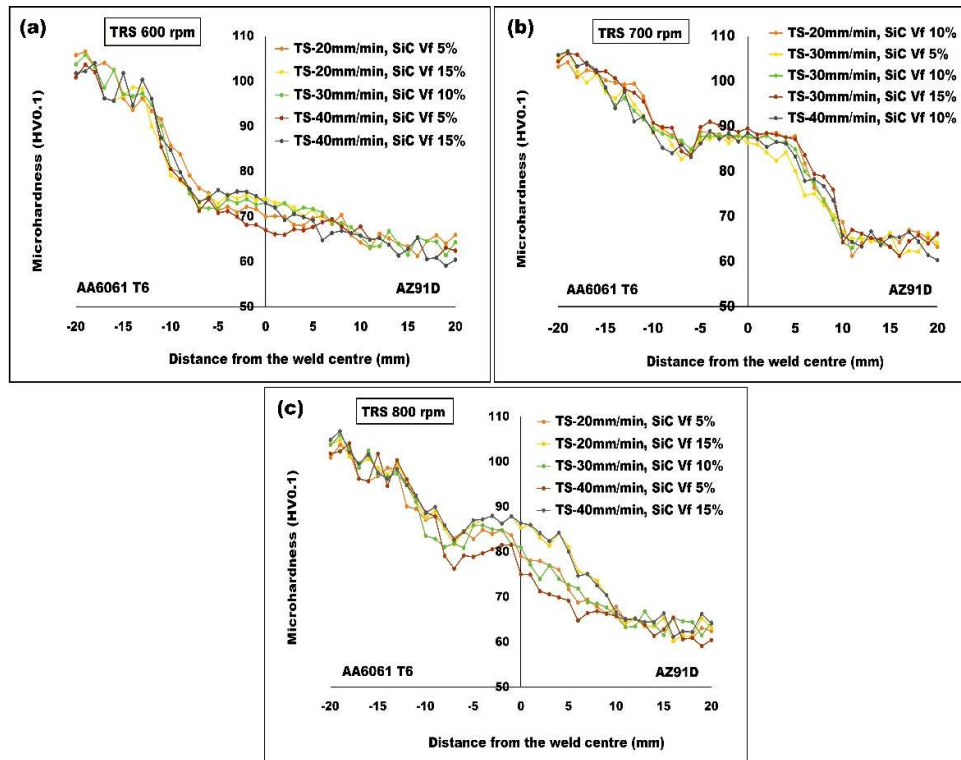


Figure 4.18: Microhardness variation in FSWed joints at varying SiC Vf

4.7 Fractography

Fractographic analysis of SiC-reinforced AZ91D/AA6061-T6 joints revealed key features such as dimples, cleavage facets, and voids, which directly influenced mechanical behavior. The fracture surfaces showed ductile tearing in the matrix and localized brittle regions near SiC particles. Features such as particle pull-out and cleavage were standard, indicating that SiC reinforcement enhances strength and introduces potential sites for fracture initiation. Fracture morphology was influenced by SiC size, dispersion, and volume fraction, as well as by welding parameters such as TRS and TS. At 600 rpm (Figure 4.19a), the fracture surface showed shallow and uneven dimples, pronounced cleavage facets, and several deep voids formed by SiC particle pull-out and agglomeration. These defects acted as stress concentrators, promoting premature crack initiation, which is consistent with the reduced elongation and lower strength observed under this condition [27]. At 700 rpm (Figure 4.19b), the fracture morphology was dominated by fine, dimples with only a few small, well-distributed voids. This indicates a predominantly mixed fracture mode supported by uniform particle dispersion. A predominance of dimples and the mixed fracture mode

were directly associated with enhanced joint strength. At 800 rpm (Figure 4.19c), the fracture surface displayed large, irregular honeycomb-like dimples interspersed with large voids and occasional cleavage facets. These features suggest a mixed ductile-brittle fracture, arising from excessive heat input and turbulent material flow that created particle-rich and particle-lean zones. As a result, joint strength and ductility were lower than at 700 rpm.

Overall, the fractography confirms that welds produced at either low or high TRS are prone to brittle-dominated fracture due to void clustering and cleavage. In contrast, an intermediate TRS (700 rpm) promotes the formation of fine dimples and limited voids. This balance suppresses brittle features and enhances ductility, highlighting the critical role of process parameters and SiC dispersion in governing fracture behavior and joint integrity [27].

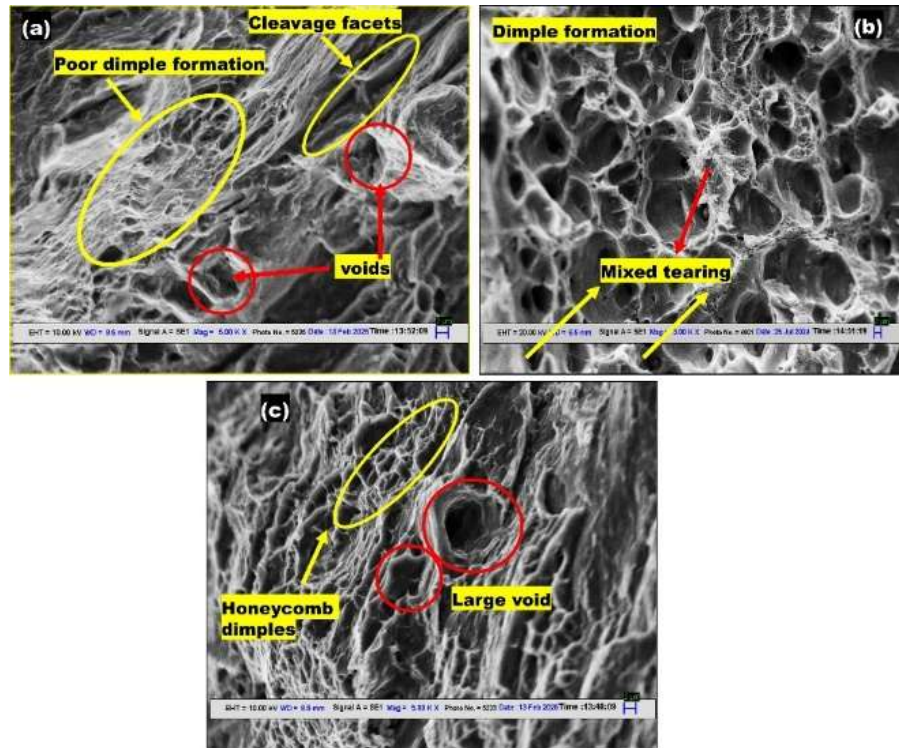


Figure 4.19: Fractured surface morphology of FSWelded joints (a) Specimen 5, (b) Specimen 6, (c) Specimen 8

4.8 Thermal Aspects

The thermal field generated during friction stir welding (FSW) governs almost every physical event within the weld zone, including material softening, plastic flow, and the evolution of microstructure and mechanical properties. In dissimilar systems such as AZ91D-AA6061-T6, the thermal cycle is especially critical because the alloys exhibit different melting ranges, thermal conductivities, and flow stresses; thus, even modest variations in peak temperature or heating rate can alter the dominant deformation and diffusion mechanisms on each side of the interface. Although the earlier sections presented the optimization, microstructural, and mechanical findings for SiC-reinforced AZ91D-AA6061 joints, the present thermal analysis correlates the observed experimental responses with the underlying temperature distribution and heat input characteristics, thereby strengthening the physical understanding of the process. By linking analytically estimated heat input, experimentally measured temperature histories, and numerically simulated thermal fields, a consistent picture is developed that explains the observed trends in grain refinement, SiC dispersion, intermetallic compound formation, and mechanical properties across the range of process parameters.

The optimized process parameters for the SiC-reinforced AZ91D/AA6061-T6 joints (TRS, TS, and Vf) were used as the reference condition for thermal analysis. This reference case (700 rpm, 30 mm/min, 15% Vf) corresponds to the joint exhibiting the finest stir-zone grain size, uniform SiC distribution, and highest ultimate tensile strength. It therefore provides a meaningful target for thermal characterization and correlation of process, structure, and properties. Because the explicit modeling of SiC nanoparticles within the weld nugget is computationally impractical, a transient thermal simulation was performed for the corresponding non-reinforced AZ91D/AA6061-T6 matrix to obtain the baseline temperature field. The results were then validated against experimental thermocouple data. A good agreement between the simulated and measured peak temperatures at multiple locations along the weld line confirms that the simplified thermal model accurately reproduces the actual thermal environment experienced during welding.

The analytical and simulated thermal analysis was conducted on unreinforced joints because the inclusion of SiC nanoparticles (5 to 15 vol%) does not significantly modify

the heat generation mechanism, which is primarily governed by tool geometry, friction, and plastic deformation, as confirmed by the peak temperatures obtained at various thermocouples during experimental analysis. The measured temperature histories for reinforced and unreinforced welds at the same TRS and TS conditions showed nearly identical peak and average values, indicating that any changes in heat generation due to the presence of hard particles or local viscosity variations are negligible compared to the dominant contributions from shoulder-workpiece friction and bulk plastic work. Furthermore, these experimental thermocouple data confirm that the temperature deviations between reinforced and unreinforced joints are less than 2%, validating the applicability of the unreinforced thermal model. Consequently, the unreinforced AZ91D/AA6061-T6 thermal response can be used as a robust baseline to interpret how SiC particles interact with the temperature field through pinning, diffusion blocking, and local flow modification without significantly altering the global thermal input to the system.

Thus, the combined simulated and experimental thermal data for the unreinforced system serve as a dependable reference for interpreting the particle behavior, microstructural changes, and mechanical performance of SiC-reinforced welds. By superimposing the microstructural maps and mechanical property distributions obtained for reinforced joints onto the validated thermal field of the unreinforced matrix, it becomes possible to attribute changes in grain size, IMC thickness, and fracture mode to either local thermal gradients or the presence and distribution of SiC nanoparticles. This integrated approach ensures that conclusions drawn regarding the role of heat input and temperature distribution in controlling weld quality are firmly anchored in both experimental evidence and physically consistent thermal simulations.

4.9 Analytical Heat Input results

Parameters and constants used (exact values)

- Shoulder diameter = 19.6 mm $\rightarrow R_s = 9.8 \text{ mm} = 9.8 \times 10^{-3} \text{ m}$
- Pin diameter = 5.0 mm $\rightarrow R_p = 2.5 \text{ mm} = 2.5 \times 10^{-3} \text{ m}$
- Pin length = $H_p = 5.7 \text{ mm} = 5.7 \times 10^{-3} \text{ m}$

- Friction coefficient = $\mu = 0.4$
- Axial (plunge) force = $F_P = 4000 \text{ N}$
- Heat-transfer efficiency = $\eta = 0.8$
- Angular speed: $\omega = 2\pi n$ where n is $\text{rev}\cdot\text{s}^{-1}$ ($n = \text{RPM}/60$).
- Q_T and Q_{eff} are in watts ($\text{W} = \text{J/s}$). Q_ℓ is reported in J/mm

Also, from equations 3.37, 3.38 and 3.39

$$Q_T = \frac{2}{3} \omega \mu F_P \left[R_S + \frac{3R_P^2 H_P}{R_S^2} \right]$$

$$Q_{\text{effective}} = \eta Q_T = \eta \frac{2}{3} \omega \mu F_P \left[R_S + \frac{3R_P^2 H_P}{R_S^2} \right]$$

$$Q_\ell = \frac{Q_{\text{eff}}}{v}, \text{ where } v \text{ is the traverse speed } (\text{mm}\cdot\text{min}^{-1}).$$

4.9.1 Calculations

Table 4.9: Analytical estimation of heat input of non-reinforced joints for selected parameters

Rotational speed (RPM)	Rotational Speed ω (rad/s)	Traverse speed (mm/min)	Traverse speed (v) (mm/sec)	Q (total) (W)	Q (effective) = η Q (total) (W)	Q/v (J/mm)
600	62.83	40	0.66	731.38	585.11	877.66
700	73.3	30	0.5	853.28	682.63	1365.25
800	83.78	20	0.33	975.18	780.14	2340.43

These values were obtained using standard FSW heat input relations, where total heat is proportional to the product of torque and angular speed, and the adequate heat per unit length accounts for the welding speed and heat-transfer efficiency. These analytical heat estimations were used to correlate experimental and simulated temperature data, as the total heat governs the temperature field. In particular, heat per unit length acts as a convenient scalar index that directly compares different parameter combinations in

terms of thermal severity, providing a physical basis for understanding the observed changes in microstructure and mechanical response.

It is evident from Table 4.9 that the total and effective heat generation increase with higher tool rotational speed due to greater frictional power input. The increase in ω from 62.83 to 83.78 rad/s raises Q (total) by approximately 33%, reflecting the strong sensitivity of frictional heating to rotational speed at constant axial force and friction coefficient. Conversely, increasing traverse speed reduces the net heat input per unit length, as the tool spends less time over a given point. As Q/v scales inversely with v , halving the traverse speed from 40 to 20 mm/min nearly triples the energy deposited per unit length, pushing the process from a low-heat to a high-heat welding regime. These variations directly explain the observed mechanical performance trends described in the preceding sections.

At 600 rpm and 40 mm/min, the lower heat input resulted in incomplete plasticization and insufficient material flow, producing discontinuities at the interface and a lower UTS of 71.46 MPa. The relatively low Q/v (877.66 J/mm) limits both peak temperature and the spatial extent of the high-temperature zone, thereby constraining dynamic recrystallization and promoting defects and regions of lack of bonding in the stir zone. At 700 rpm and 30 mm/min, the heat input reached an optimum level, ensuring sufficient softening for sound material flow and defect-free joint formation (UTS of 114.98 MPa). Here, Q/v increases to 1365.25 J/mm, which is sufficient to fully plasticize the interface, sustain stable flow, and promote fine equiaxed grain formation without excessive grain growth or IMC thickening. At 800 rpm and 20 mm/min, the excessive heat led to turbulent flow and extended thermal exposure, promoting grain coarsening and reducing hardness and strength despite complete mixing. The highest Q/v value (2340.43 J/mm) broadens the high-temperature region and prolongs the time spent in a range, accelerating diffusion-controlled IMC growth and driving abnormal grain growth in the stir zone and the HAZ. Thus, the analytical results provide a quantitative foundation for selecting 700 rpm and 30 mm/min as the optimal parameter combination for reinforced joints. This parameter set represents a thermal compromise where the heat input is high enough to achieve complete consolidation and favorable microstructural refinement, yet low enough to avoid deleterious overaging, IMC thickening, and flow instabilities.

4.10 Experimental Thermocouple Data

K-type thermocouples were placed at eight distinct positions along the weld line during the experiment, and temperatures were recorded for each case. Four sensors were embedded on the advancing side and four on the retreating side, at two distances (45 mm and 85 mm) from the weld start, to capture the thermal behavior along the weld length. Due to equipment channel limitations, four thermocouples were used per run, and the same parameters were repeated to record all eight locations. This approach ensured that all positions experienced identical process conditions, enabling direct comparison between different runs and eliminating uncertainty arising from parameter drift. In the numerical simulation, all eight thermocouple points were monitored simultaneously. This difference does not affect the accuracy, since the parameters and heat source remained identical. The close agreement between separate experimental runs and the single simulated pass supports the assumption of repeatable thermal behavior and validates the use of a single heat source model for all positions. The experimental comparison of SiC-reinforced and non-SiC-reinforced is presented in Table 4.10.

Table 4.10: Experimental temperature of SiC-reinforced vs non-SiC-reinforced (700 rpm, 30 mm/min)

Thermocouple Position	Position from weldstart (mm)	Side	700 rpm,30mm/min		
			Experimental Temperature Non-reinforced (°C)	Experimental Temperature SiC reinforced (15% Vf) (°C)	Temperature deviation (%)
T1	45	Advancing	295	303	2.71
T2	45	Retreating	290	292	0.69
T3	45	Advancing	281	284	1.07
T4	45	Retreating	278	283	1.80
T5	85	Advancing	329	336	2.13
T6	85	Retreating	315	320	1.59
T7	85	Advancing	318	324	1.89
T8	85	Retreating	326	329	0.92

The results showed excellent agreement, with deviations of less than 2%. The slightly higher temperatures measured in the reinforced joints fall well within experimental scatter and do not indicate any fundamental change in heat generation. These results validate that the developed model accurately reproduces the real thermal field and can be reliably used to interpret microstructural and mechanical outcomes. In particular, the close match at both 45 mm and 85 mm positions, as well as on the advancing and retreating sides, demonstrates that the model captures not only the peak temperatures but also the spatial temperature gradients that drive asymmetric microstructural evolution.

The thermal field correlates directly with the microstructural and mechanical performance observed in the preceding sections. Peak temperature and cooling rate control the extent of recovery and recrystallization in the stir zone, as well as the thickness and morphology of Al-Mg intermetallic compounds at the interface, and ultimately determine the final distribution and stability of SiC nanoparticles within the matrix. The peak temperature governs the extent of dynamic recrystallization, particle distribution, and intermetallic compound formation at the interface.

At 600 rpm and 40 mm/min, the low heat input per unit length produced insufficient stirring, leading to discontinuous nugget formation and lower UTS. Thermocouple readings in this condition show peak temperatures 20-40 °C lower than those at 700 rpm/30 mm/min, which restricts the width of the region exceeding the critical recrystallization temperature and causes incomplete mixing, cracks, and non-uniform SiC dispersion in reinforced welds. At 700 rpm and 30 mm/min, moderate heat input promoted adequate plastic flow and uniform stirring, resulting in fine equiaxed grains in the stir zone and the highest UTS (114.98 MPa). Here, the thermal field is sufficiently intense and broad to ensure that the entire nugget region experiences temperatures conducive to full softening and grain refinement. At the same time, the time-at-temperature remains short enough to prevent excessive IMC growth or overaging in the HAZ. At 800 rpm and 20 mm/min, excessive heat per unit length resulted in prolonged exposure to high temperatures, leading to grain growth, softening, and a slight reduction in tensile strength. The thermocouple data for this condition exhibit peak temperatures approaching or exceeding near the advancing side, with longer dwell times above critical diffusion thresholds, which explains the observed thickening of IMCs,

coarsening of stir-zone grains, and reduction in hardness, despite the elimination of macroscopic defects.

4.11 Validation with Simulation Results: Thermal Profile and Temperature

Contours

The simulation was conducted using Ansys 22 R1. The comparison between experimental and simulated temperature histories at different thermocouple positions is presented in Table 4.11.

Table 4.11: Peak Temperature Experimental vs Peak Temperature Simulation of Non-Reinforced Specimens

Thermocouple Position	Position from weldstart (mm)	Side	600 rpm,40mm/min			700 rpm,30mm/min			800 rpm,20mm/min		
			Peak Temp Experiment (°C)	Peak Temp Simulation (°C)	Error (%)	Peak Temp Experiment (°C)	Peak Temp Simulation (°C)	Error (%)	Peak Temp Experiment (°C)	Peak Temp Simulation (°C)	Error (%)
T1	45	Advancing	264	274.55	4.00	295	301.63	2.25	336	343.77	2.31
T2	45	Retreating	262	271.58	3.66	290	298.75	3.02	331	341.11	3.05
T3	45	Advancing	255	265.80	4.24	281	292.02	3.92	327	332.62	1.72
T4	45	Retreating	251	264.29	5.29	278	290.01	4.32	324	331.15	2.21
T5	85	Advancing	308	315.00	2.27	329	335.74	2.05	355	362.03	1.98
T6	85	Retreating	298	303.52	1.85	315	323.93	2.83	339	349.53	3.11
T7	85	Advancing	295	305.46	3.55	319	325.25	1.96	341	350.46	2.77
T8	85	Retreating	298	312.74	4.95	326	333.31	2.24	347	359.82	3.69

Across all parameter combinations and thermocouple locations, the simulation predicts peak temperatures within approximately 2 to 5% of the experimental values, demonstrating excellent agreement and confirming the validity of the applied heat source and boundary conditions. The most significant deviations occur at T4 and T8 at 600 rpm, where an underestimation of local convection or radiation losses near the plate edges may slightly influence the measured temperatures; however, the trends remain fully consistent. Importantly, the model accurately reproduces the relative temperature differences between advancing and retreating sides, as well as between the 45 mm and 85 mm positions, which is essential for reliable prediction of asymmetric microstructure and phase evolution. Further, the temperature profiles and temperature contours obtained at selected parameter combinations are presented in Figures 4.20-4.25.

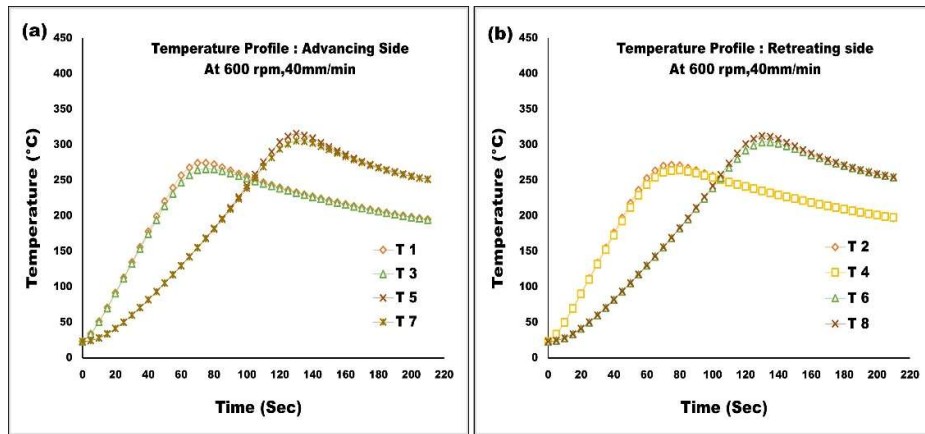


Figure 4.20: (a,b) Temperature profile at different thermocouple positions at 600 rpm, 40mm/min.

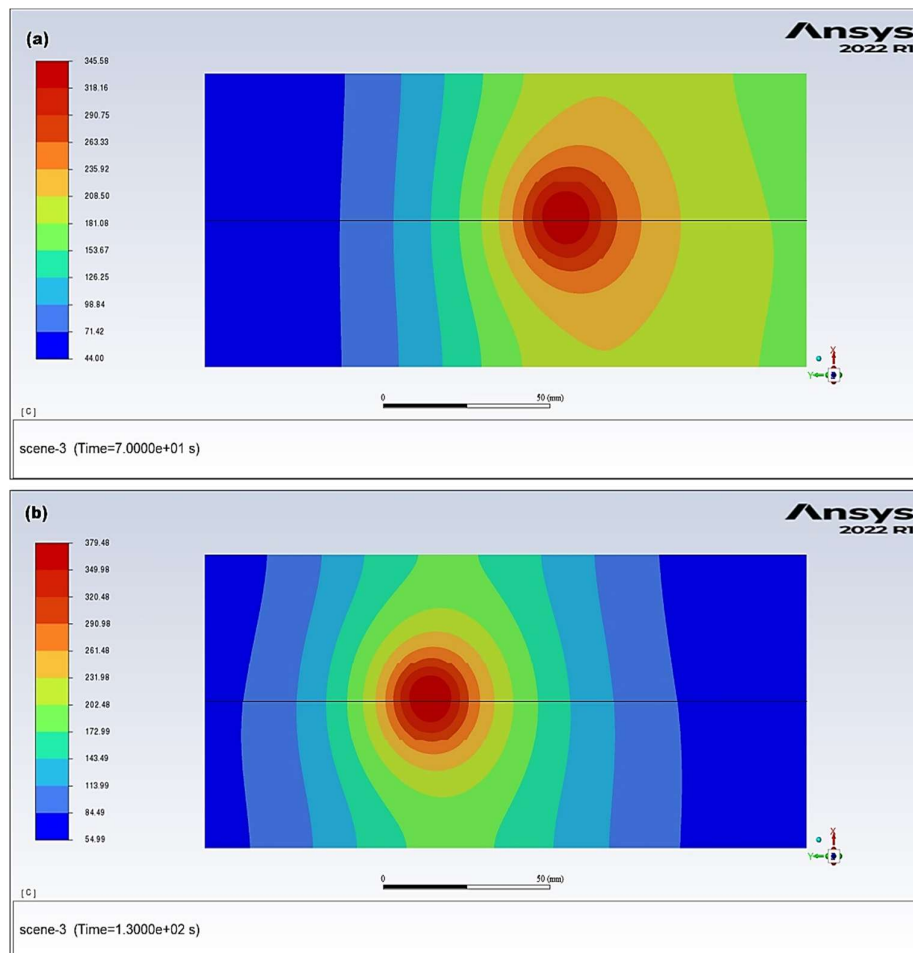


Figure 4.21: (a,b) Temperature Contours at 600 rpm, 40mm/min.

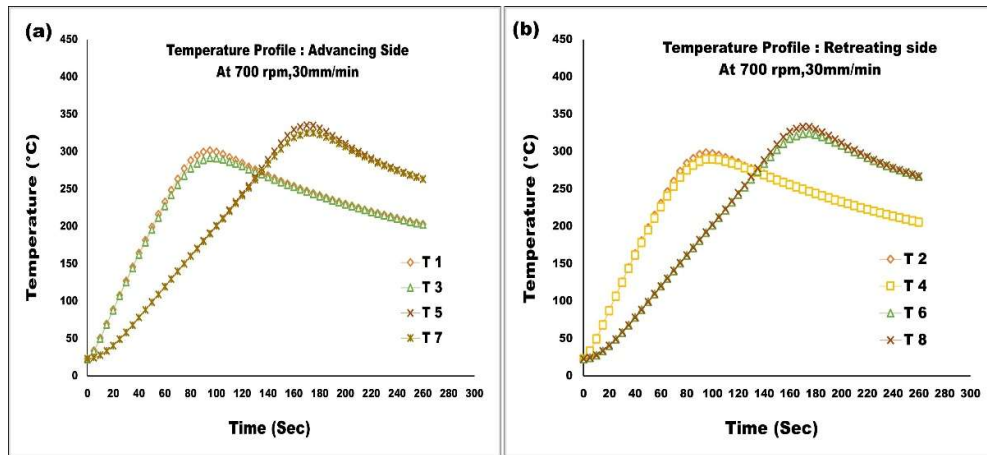


Figure 4.22:(a,b) Temperature profile at different thermocouple positions at 700 rpm,30mm/min.

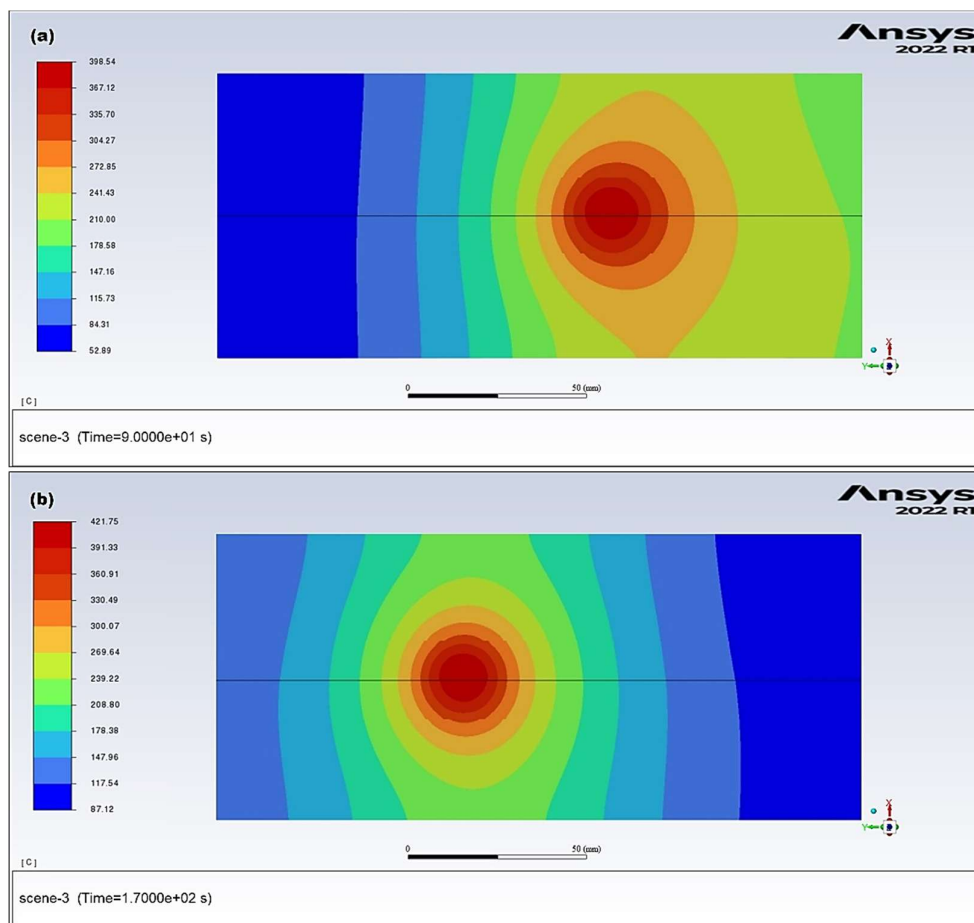


Figure 4.23: (a,b) Temperature contours at 700 rpm,30mm/min.

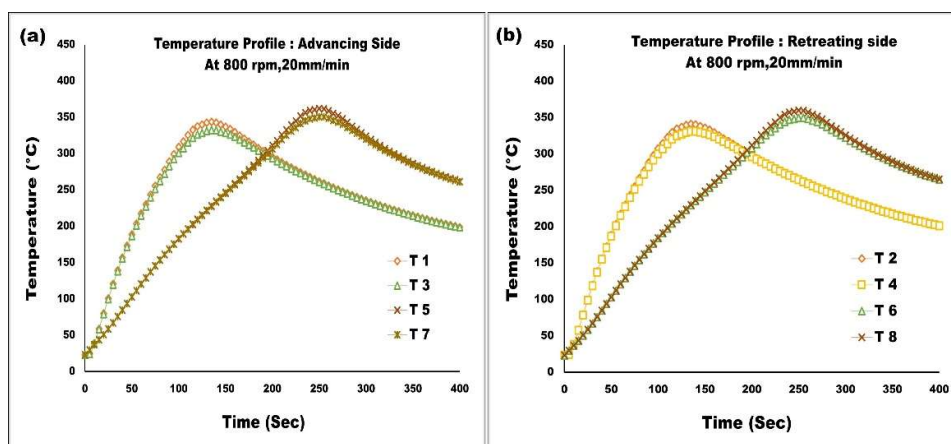


Figure 4.24: (a,b) Temperature profile at different thermocouple positions at 800 rpm, 20mm/min.

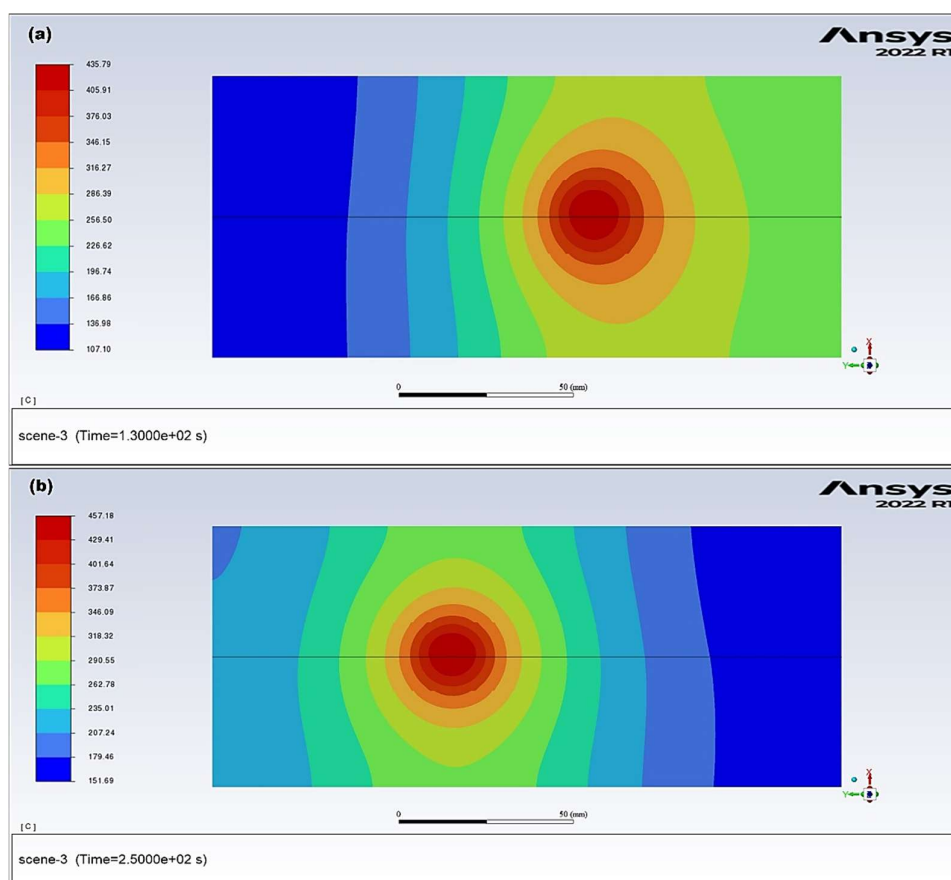


Figure 4.25: (a,b) Temperature contours at 800 rpm, 20mm/min.

The time-temperature profiles (Figures 4.20, 4.22, and 4.24) show that peak temperatures and heating/cooling rates increase systematically with rotational speed and decreasing traverse speed, with the 700 rpm/30 mm/min case providing the most balanced thermal cycle: sufficiently high peak temperature for full softening and recrystallization, but with a relatively steep cooling rate that limits grain growth and IMC thickening. The corresponding contour plots (Figures 4.21, 4.23, and 4.25) reveal that the isothermal regions expand both in depth and width as the heat input increases, particularly on the advancing side, where the tool motion enhances friction and plastic deformation, consistent with the observed asymmetry in microstructure and hardness profiles.

The simulation also indicated that the temperature gradient across the advancing and retreating sides increased with tool speed, consistent with the asymmetric microstructure observed in optical and SEM images. At 800 rpm, the advancing side exhibits significantly higher peak temperatures and broader high-temperature zones than the retreating side, which explains the more pronounced grain coarsening, thicker IMC layers, and higher incidence of flow-related defects observed on that side in microstructural examinations. These correlations confirm that the mechanical and metallurgical responses of the joints are thermally driven phenomena. By integrating analytical heat input estimates, thermocouple measurements, and numerical thermal simulations, a coherent thermal framework is established that explains the parameter-dependent evolution of grain size, SiC dispersion, IMC morphology, hardness, and tensile strength in SiC-reinforced AZ91D/AA6061-T6 friction stir welds.

4.12 Industrial Implications and Practical Considerations

While the optimized process parameters identified in this study (700 rpm TRS, 30 mm/min TS, and 15% SiC Vf) demonstrated improved tensile strength, strain, and microhardness under controlled laboratory conditions, their application in industrial environments requires consideration of practical manufacturing factors such as tool durability, scalability, repeatability, and cost-effectiveness.

4.12.1 Tool Wear and Durability

The incorporation of SiC nanoparticles enhances mechanical performance through grain refinement and dispersion strengthening. However, since SiC is a rigid ceramic reinforcement, its presence may influence tool wear behavior during prolonged welding operations. At higher rotational speeds, increased frictional interaction between the tool and reinforced material may accelerate abrasive wear. Under the optimized 700 rpm condition, excessive heat input was avoided, which may help moderate severe tool degradation. For industrial-scale production, the use of wear-resistant tool materials, such as hardened tool steels or polycrystalline cubic boron nitride (PCBN), may improve tool longevity and process stability.

4.12.2 Scalability of Optimized Parameters

The optimized parameter combination is governed by the balance between TRS and TS, which controls heat input per unit length. Since FSW is a solid-state process driven by frictional heating and plastic deformation, similar thermal trends may be maintained over extended weld lengths provided that process parameters are consistently controlled. However, during continuous industrial welding, thermal accumulation may occur; therefore, real-time temperature monitoring or adaptive parameter control may be beneficial to maintain uniform microstructure and prevent excessive growth of intermetallic compounds (IMCs).

4.12.3 Repeatability and Process Stability

The RSM and CCD optimization framework established in this study defines a stable processing window rather than a single operating point. This enhances repeatability by allowing operation within a defined parameter range while maintaining acceptable mechanical properties. Such a defined window may improve manufacturing robustness and reduce the likelihood of defects such as tunnel formation or incomplete consolidation during industrial production.

4.12.4 Cost-Performance Considerations

The addition of 15% SiC nanoparticles introduces additional material preparation and processing costs. However, the experimental results clearly demonstrate improved

grain refinement, controlled IMC morphology, increased tensile strength, and enhanced hardness under optimized conditions. In applications where strength-to-weight ratio and structural reliability are critical, the performance benefits may justify the added processing cost. The overall economic feasibility would depend on the specific industrial application and required service performance.

4.12.5 Practical Manufacturing Outlook

Overall, the optimized laboratory parameters identified in this research are technically adaptable to industrial conditions, provided that appropriate tool materials and process monitoring strategies are implemented. The findings suggest that controlled heat input and uniform SiC dispersion are key factors in achieving mechanically reliable dissimilar AZ91D/AA6061-T6 joints. Future investigations may include extended weld trials, systematic tool life evaluation, and detailed cost-benefit analysis to further bridge laboratory optimization with large-scale manufacturing practice.

Chapter 5 : CONCLUSIONS AND FUTURE SCOPE

5.1 Conclusions

The present investigation comprehensively examined the feasibility and performance of SiC-reinforced dissimilar FSWed joints of aluminum AA6061-T6 and magnesium AZ91D. By systematically varying TRS, TS, and SiC volume fraction, this work established robust correlations among welding parameters, thermal history, microstructural evolution, intermetallic compound (IMC) formation, and mechanical response. The outcomes clearly demonstrate that FSW, when combined with controlled nanoparticle reinforcement, is a viable solid-state route for producing structurally sound and mechanically reliable Al-Mg dissimilar joints. Careful evaluation of IMC morphology, hardness, tensile strength, and fracture features revealed the following key findings:

- FSW successfully produced defect-free joints across a wide range of processing conditions, confirming the suitability of the technique for joining AA6061-T6 and AZ91D, which are otherwise difficult to weld using conventional fusion processes due to their significant differences in melting temperatures, thermal conductivities, and chemical affinities. The incorporation of SiC nanoparticles significantly altered the stir zone (SZ) characteristics by influencing heat flow, plastic deformation behavior, and microstructural stability during welding. Compared to unreinforced joints, the reinforced welds exhibited refined grain structures, modified IMC morphology, and enhanced mechanical performance when optimal processing conditions were employed.
- A key outcome of this study is the identification of an optimal welding condition corresponding to a TRS of 700 rpm, a TS of 30 mm/min, and a SiC volume fraction of 15%. Under this condition, a balanced heat input was achieved, which played a decisive role in governing material flow, nanoparticle dispersion, and interfacial reactions. The stir zone formed under these parameters exhibited uniform dispersion of SiC nanoparticles, significant grain refinement on both the AA6061-T6 and AZ91D sides, and suppression of excessive IMC growth at the interface. The refined microstructure promoted effective load transfer across the joint. It delayed

the onset of crack initiation, resulting in the highest observed UTS of 114.98 MPa, a microhardness of 88.9 HV0.1, and a joint efficiency of 52.5%.

- In contrast, joints fabricated at a lower tool rotational speed (600 rpm) combined with a higher traverse speed (40 mm/min), even at the same reinforcement fraction of 15%, suffered from insufficient heat input. The inadequate thermal energy restricted plastic flow and limited effective stirring, resulting in a non-uniform nanoparticle distribution and pronounced SiC agglomeration within the stir zone. These agglomerates acted as stress concentrators, promoting the formation of thick, continuous, and brittle IMCs layers at the Al-Mg interface. Consequently, premature crack initiation and propagation were observed, resulting in significantly reduced mechanical performance, limiting the UTS to 81.27MPa and microhardness to 72.47 HV0.1.
- At the other extreme, excessive heat input generated under high TRS (800 rpm) and low TS (20 mm/min) adversely affected joint performance despite the presence of SiC (15% Vf) reinforcement. Overheating led to abnormal material flow behavior, producing particle-rich and particle-lean zones within the stir region. This heterogeneity disrupted the continuity of load-bearing paths and facilitated the formation of irregular and locally thickened IMCs. The resulting microstructure exhibited reduced ductility and inferior fracture resistance compared to joints welded under intermediate heat input conditions, underscoring that higher thermal energy does not necessarily translate to improved joint properties.
- Microstructural and phase analyses provided further insight into the governing mechanisms. XRD confirmed the presence of Al_3Mg_2 and $\text{Al}_{12}\text{Mg}_{17}$ IMC phases. Increasing the SiC volume fraction from 5% to 15% progressively reduced the thickness and continuity of these IMC layers by acting as diffusion barriers and promoting heterogeneous nucleation of finer grains. However, this beneficial effect was strongly dependent on proper thermal control; excessive agglomeration or overheating counteracted the advantages of higher reinforcement content, reaffirming the critical role of balanced process parameters.
- Hardness profiling across the weld region revealed a strong dependence on both nanoparticle distribution and thermal exposure. The enhanced microhardness observed in optimally welded joints was attributed to combined grain refinement, dispersion strengthening by SiC nanoparticles, and moderated IMC formation.

Conversely, joints produced under non-optimal conditions showed localized hardness fluctuations associated with particle clustering, softening due to overaging, or brittle IMC dominance.

- Fractographic analysis corroborated the mechanical test results and provided clear evidence of the influence of heat input on failure mechanisms. Joints welded at the intermediate TRS of 700 rpm exhibited mixed fracture features, characterized by fine, equiaxed dimples and limited void coalescence. In comparison, joints fabricated at either low (600 rpm) or high (800 rpm) TRS displayed brittle-dominated fracture behavior, marked by cleavage facets, void clustering, and crack propagation along IMC-rich regions. These observations confirm that fracture behavior in dissimilar Al-Mg FSW joints is intrinsically linked to thermal history, microstructural stability, and reinforcement dispersion.
- Building upon the experimentally obtained data generated through the CCD design matrix, the post-experimental application of RSM enabled a quantitative interpretation of how tool rotational speed, traverse speed, and SiC volume fraction interact to govern the performance of dissimilar AZ91D-AA6061-T6 friction stir welded joints. The statistical models confirmed that joint strength and hardness in this Mg-Al system are governed by coupled thermal-mechanical effects rather than by any single process parameter in isolation. The optimization outcomes reinforced the experimentally observed optimum by identifying an intermediate heat input regime in which stable material flow, uniform SiC nanoparticle dispersion within the stir zone, and restrained Al-Mg intermetallic compound formation coexist. The close correspondence between optimized predictions and experimental responses demonstrates that the statistical framework effectively bridges empirical observations and underlying physical mechanisms, thereby providing a reliable and system-specific methodology for designing and controlling SiC-reinforced dissimilar AA6061-T6 / AZ91D FSW joints.
- To complement the experimental and statistical analyses, a numerical simulation was developed to predict temperature evolution during welding. The model accounted for heat generation from both frictional sliding and plastic deformation at the tool-workpiece interface and was validated using thermocouple measurements on the advancing and retreating sides of the joint. The predicted thermal profiles showed a deviation of less than 5% from the experimental data,

confirming the accuracy of the numerical approach. Simulation results revealed that heat input increased with TRS and decreased with TS, directly correlating with observed material flow behavior, grain refinement, and IMC morphology. This strong agreement validates the reliability of the simulation framework and supports its use as a predictive tool for process optimization. Notably, the simulation results reinforced the experimental finding that the thermal field serves as the root cause governing microstructural development, IMC formation, and mechanical response in SiC-reinforced dissimilar FSW joints.

- Overall, this study establishes that controlled SiC nanoparticle dispersion, when combined with carefully selected welding parameters, is essential for suppressing deleterious IMCs growth, refining fracture morphology, and achieving reliable mechanical performance in AA6061-T6/AZ91D dissimilar joints. The integrated experimental and numerical approach adopted herein provides a comprehensive understanding of the relationships between process, structure, and properties in SiC-reinforced FSW, offering practical guidelines for designing mechanically reliable dissimilar joints for lightweight structural applications.

5.2 Future Scope

- While the present work provides a detailed understanding of SiC-reinforced dissimilar FSW joints, several avenues remain open for further investigation to expand both scientific knowledge and industrial applicability. Future studies may explore alternative reinforcement systems such as Al_2O_3 , TiC, or hybrid combinations (e.g., SiC + Al_2O_3) to compare their effectiveness in controlling IMCs, grain refinement mechanisms, and load transfer efficiency. Such comparisons would help identify reinforcement-specific strengthening mechanisms and optimize the design of composite joints.
- The numerical simulation performed in this work was performed using ANSYS 22R1. Various other simulation software can be used to carry out the numerical simulation of dissimilar materials.
- Further improvements in joint quality may be achieved by investigating alternative tool pin geometries, including tapered, threaded, or fluted designs, to enhance material flow, improve nanoparticle distribution, and strengthen interfacial

bonding. The combined influence of tool geometry and reinforcement on thermal and mechanical behavior warrants a systematic study.

- Post-weld treatments such as friction stir processing (FSP) or heat treatment cycles can be examined to refine grain structure further, homogenize particle distribution, and relieve residual stresses, potentially leading to improved mechanical performance.
- The long-term performance of the optimized joints, including fatigue, corrosion, and thermal stability under service conditions, can be assessed for industrial applications.

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LIST OF PUBLICATIONS

➤ List of papers published in SCIE journals: -

1. Gaurav, S., Zunaid, M. & Mishra, R.S. Process Optimization and Mechanical Characterization of SiC-Reinforced Friction Stir Welded AZ91D/AA6061-T6 Joints Using Response Surface Methodology. *J. of Materi Eng and Perform* **35**, 7050–7075 (2026). <https://doi.org/10.1007/s11665-025-12132-2>
Springer Nature (Impact factor 2.0)
2. Gaurav, S., Zunaid, M. & Mishra, R.S. Role of SiC nanoparticle reinforcement in the mechanical and metallurgical characterization of friction stir-welded dissimilar AZ91D and AA6061-T6 joints. *Sāadhanā* **51**, 27 (2026). <https://doi.org/10.1007/s12046-025-03041-0>
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3. Mechanical Gaurav, S., Zunaid, M., & Mishra, R.S. Mechanical and metallurgical characterization of SiC-reinforced friction stir-welded joints of dissimilar AZ91D and AA6061-T6 metals. *Materialwissenschaft und Werkstofftechnik (Materials Science and Engineering Technology)*. (2025) <https://doi.org/10.1002/mawe.70063>
Wiley (Impact factor 1.1)
4. Shwetanshu Gaurav, R. S. Mishra & M. Zunaid (2023): A critical review on mechanical and microstructural properties of dissimilar aluminum (Al)-magnesium (Mg) alloys, *Journal of Adhesion Science and Technology*, DOI: 10.1080/01694243.2022.2067437
Taylor & Francis (Impact factor 3.7)
5. Shwetanshu Gaurav, R.S Mishra & M. Zunaid (2023): Enhancement of microstructure and mechanical properties of similar and dissimilar aluminium alloy by friction stir welding/processing using nanoparticles: a review, *Journal of Adhesion Science and Technology*, DOI: 10.1080/01694243.2023.2186756
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List of papers presented at international conferences:

1. Effect of friction stir welding parameters on microstructure and mechanical properties of nanoparticle reinforced magnesium AZ91D and aluminum AA6061-T6. Shwetanshu Gaurav, M. Zunaid & R.S Mishra
(Conference -ICAMSF 2025, Chitkara University, Punjab)

2. Mechanical and metallurgical characterization of SiC reinforced friction stir welded joints of dissimilar AZ91D and AA6061-T6 metals. Shwetanshu Gaurav, M. Zunaid & R.S Mishra
(Conference -ICAMSF 2025, Chitkara University, Punjab)

CURRICULUM VITAE

Shwetanshu Gaurav

Research Scholar | Mechanical Engineering | Friction Stir Welding | CFD

Shwetanshu Gaurav is a research scholar in the Department of Mechanical Engineering at Delhi Technological University (DTU), Delhi, with over eight years of teaching experience. His research focuses on friction stir welding (FSW) of dissimilar alloys, with particular emphasis on microstructural evolution, mechanical characterization, and process optimization. His work also involves CFD-based numerical analysis to study material flow and heat transfer phenomena in FSW. His academic interests include welding metallurgy, advanced manufacturing processes, and computational modeling of joining technologies.

Research Interests: Friction Stir Welding, Dissimilar Metal Joining, CFD Analysis, Heat Transfer, and Material Flow Modeling.