

Research Paper

Microstructural Characterization and Mechanical Properties of AZ31 Magnesium–Titanium Joints Produced by Friction Stir Welding

Mohammad Baghei Shahraki, Mahdi Omid, Mohsen Abbasi-Baharanchi*, Hamid Reza Bakhshi Rad, Saeid Jabbarzare

Department of Materials Engineering, Na.C., Islamic Azad University, Najafabad, Iran

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ABSTRACT

Magnesium alloys possess the lowest density among widely used structural metals and are extensively applied in aerospace and military industries. In this study, the welding behavior and mechanical properties of AZ31 magnesium alloy joints were investigated. After multiple trials, a rotational speed of 1200 rpm and a traverse speed of 8 mm/min were selected for friction stir welding (FSW) using an H13 tool steel pin. To fabricate a composite in the joint region, a thin titanium sheet was incorporated at the interface. Furthermore, to improve the distribution of reinforcing particles within the matrix, the number of welding passes was increased from one to three. Microstructural examinations revealed that friction stir welding not only homogenized the structure but also refined the grain size, leading to improved mechanical properties. The average hardness increased from approximately 70 HV for the as-received AZ31 alloy to 75 HV for the welded non-composite specimen, 78 HV for the one-pass composite specimen, and 82 HV for the three-pass composite specimen. In addition, the ultimate tensile strength increased from approximately 201 MPa for the base alloy to about 280 MPa for the three-pass composite specimen. X-ray diffraction analysis confirmed the presence of α -Mg and α -Ti phases, while no brittle intermetallic compounds were detected at the Mg–Ti interface. To evaluate the wear behavior, reciprocating wear tests were conducted under a load of 30 N over a sliding distance of 1000 m. The weight loss of the three-pass composite sample after 1000 m was approximately 6 mg. The steady-state coefficient of friction values were approximately 0.42, 0.40, 0.37, and 0.34 for the as-received, non-composite welded, one-pass composite, and three-pass composite specimens, respectively. Scanning electron microscopy (SEM) analysis of the worn surfaces showed that with increasing number of welding passes, the abrasive wear mechanism became more pronounced relative to delamination. In the non-composite sample, delamination was dominant, although abrasive wear was also observed. In the as-received sample, adhesive wear and delamination were identified as the dominant wear mechanisms.

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* **Corresponding Author:**

E-Mail: m.abbasi.b@iau.ac.ir

1. Introduction

The use of lightweight materials with high strength has recently attracted significant attention and found widespread applications. In this context, various types of polymers have been increasingly employed. However, due to their relatively low strength and limited service temperature, the use of light metals has become a strategic priority in many industries. Among lightweight high-strength alloys, aluminum, titanium, and magnesium alloys are the most widely used. Aerospace, marine, and transportation industries are among the main consumers of these alloys.

Despite these advantages, one of the major challenges associated with magnesium alloys is the fabrication of components with complex geometries and large dimensions. In such cases, the use of metal joining processes becomes inevitable, and therefore the welding behavior and the resulting joint properties require detailed investigation.

Friction stir welding (FSW), as a solid-state joining process, has been widely employed for joining magnesium alloys. By avoiding melting-related problems associated with fusion welding processes, FSW can produce joints with more favorable microstructural and mechanical characteristics. Nevertheless, the relatively low wear resistance, strength, and ductility of magnesium alloys still restrict their broader use. For this reason, the development of magnesium-based composites has received considerable attention due to their improved strength and wear resistance.

Composites are generally produced through two main approaches. In the first approach, reinforcing particles are directly added to the matrix from the beginning of the fabrication process. In the second approach, the reinforcing particles are formed in situ during processing. The latter method can provide finer and more uniformly distributed particles, which in turn promote microstructural refinement and improved mechanical properties.

Recent studies have demonstrated the potential of FSW for both joining and microstructural modification of AZ31 magnesium alloy. Aydin et al. [1] reported that sound friction stir welded AZ31 joints could be produced and that the joint strength could even exceed that of the base alloy. Liu et al. [2] also showed that increasing the tool rotational speed up to 880 rpm reduced the grain size to about 16 μm and improved the mechanical properties.

Another promising strategy for improving joint performance is the formation of a composite in the

weld region. In this regard, Abdollahzadeh et al. [3] investigated a composite formed in the joint region of AZ31 alloy using friction stir welding, where SiC nanoparticles were introduced into the weld zone. Their results indicated that the hardness of the composite joint region increased significantly compared with the base metal. Gang et al. [4] also studied the effect of welding speed and post-weld hot rolling on friction stir welded AZ31 magnesium alloy and showed that the mechanical properties improved with increasing rolling reduction.

Based on the available literature, composite formation in the joint region of friction stir welded magnesium alloys can significantly enhance joint performance. However, to the best of the authors' knowledge, fabrication of an AZ31-based composite joint reinforced by titanium through friction stir welding has not been sufficiently reported. Therefore, in the present work, AZ31 magnesium alloy was used as the base metal and titanium was employed as the reinforcing phase to produce a composite in the joint region by friction stir welding. The microstructure and mechanical properties, including hardness, tensile behavior, and wear performance, of both non-composite and composite joints were systematically investigated.

2. Materials and Methods

2.1. Materials

In this study, the optimal conditions for friction stir welding of AZ31 magnesium alloy plates were first established. Subsequently, to improve the mechanical properties of the joints, a magnesium-based composite was produced at the interface of two plates by incorporating titanium.

AZ31 magnesium alloy was employed as the base metal, and titanium was used as the reinforcing phase to form a composite in the joint region. The specimens used for the welding process were prepared from AZ31 magnesium alloy sheets with initial dimensions of 100×30×3 mm.

To accurately determine the chemical composition of the alloy, a quantometry test was performed on the prepared plates in accordance with ASTM B954-15. The obtained chemical analysis is presented in Table 1. As can be seen, the measured chemical composition is in good agreement with the compositional specifications of AZ31 magnesium alloy according to ASTM B90M-15. The Ni content was corrected to 0.005 wt%, which is within the maximum allowable limit specified by ASTM B90M-15.

Table 1. Chemical composition of AZ31 magnesium alloy.

Chemical analysis of AZ31 alloy according to ASTM B90M-15 standard		Si	Fe	Cu	Mn	Zn	Ca	Ni	Al	Mg
	Max	0.1	0.005	0.05	0.2	1.4	0.04	0.005	3.5	Re
	Min	—	—	—	—	0.6	—	—	2.5	
AZ31 alloy sheets		0.15	0.004	0.03	0.20	1.1	0.02	0.05	3.17	0.30*

2.2. Friction Stir Welding

The specimens were cut to dimensions of 100×30×3 mm and prepared for welding. Surface preparation included cleaning, degreasing, proper alignment, and clamping of the specimens for the joining process. Friction stir welding was performed using a milling machine and an H13 tool steel pin with a surface hardness of 42 HRC. Tool rotational speeds of 600

and 1200 rpm, traverse speeds of 8 and 16 mm/min, and a dwell time of 5 s at the joint location were employed, as listed in Table 2. These parameters were selected based on experimental trials and previously reported studies [2,4].

Figure 1 shows the specimens produced under different welding conditions. The initial visual inspection of the welded joints indicated that sound joints were successfully obtained.

Table 2. Variation of welding parameters.

Sample Code	Rotational Speed (rpm)	Traverse Speed (mm/min)
1	600	16
2 (Base sample)	1200	8
3 (Single-pass composite)	1200	8
4 (Three-pass composite)	1200	8

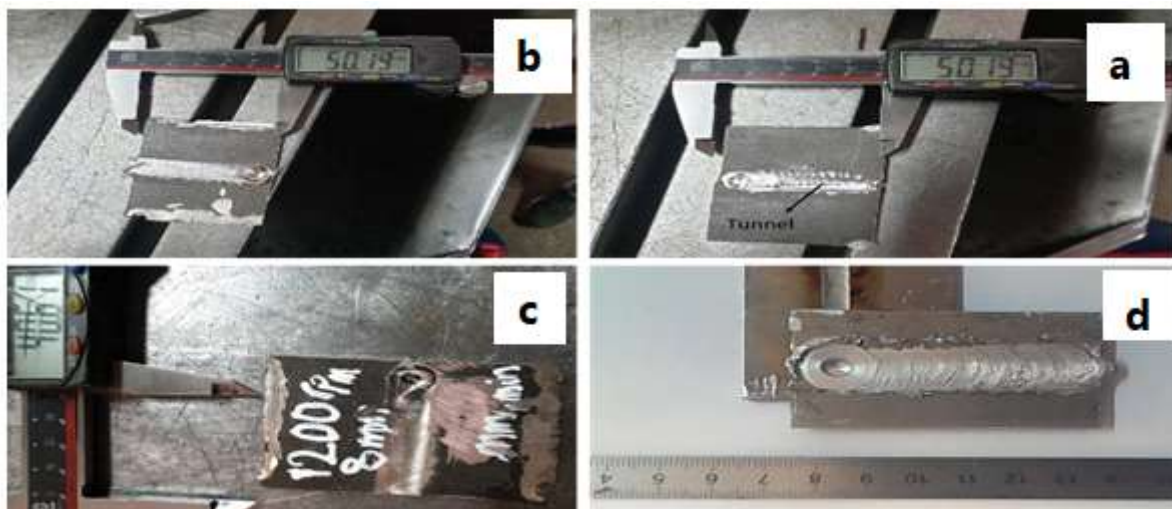


Fig. 1. Specimens obtained from the friction stir welding process at a traverse speed of 8 mm/min: (a) 600 rpm, (b) 1200 rpm (non-composite specimen), (c) 1200 rpm (single-pass composite specimen), and (d) 1200 rpm (three-pass composite specimen).

2.3. Microstructural Characterization

To study the microstructure of the base metals and welded joints, the specimens were first ground using SiC abrasive papers with grit sizes ranging from 100 to 2500, followed by polishing and etching. A solution containing 2.4 g of picric acid, 10 mL of acetic acid, 10 mL of distilled water, and 70 mL of ethanol was used as the etchant to reveal the microstructure of the AZ31 magnesium alloy. Optical micrographs of the samples were obtained and examined at various magnifications using an Epiphot 300 optical microscope.

2.4. Scanning Electron Microscopy (SEM)

To investigate the fracture surfaces of the composite and non-composite samples following tensile testing, as well as to examine the worn surfaces of the specimens, a Philips XL30 scanning electron microscope (SEM) was employed.

2.5. X-Ray Diffraction (XRD) Analysis

X-ray diffraction (XRD) analysis was utilized to investigate the phases formed and their evolution under different welding conditions. A Philips X'Pert diffractometer was used for this purpose. In all measurements, Cu K α radiation with a wavelength of

1.542 Å was employed. A dwell time of 1 s, a step size of 0.05°, and a scanning range of 30° to 100° were selected based on the positions of the peaks. Phase identification was performed based on the

diffraction angles and intensities of the peaks using the X'Pert HighScore Plus software. Figure 2 displays the XRD pattern obtained from the AZ31 magnesium alloy sheet.

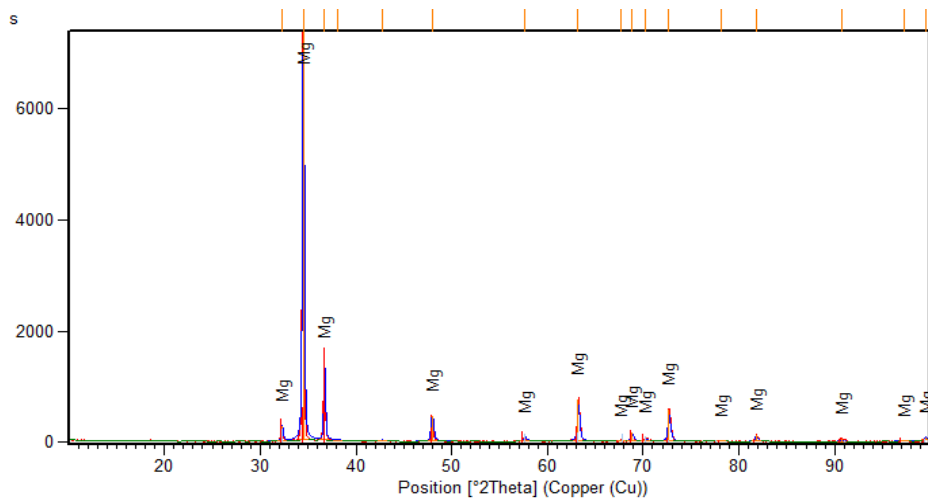


Fig. 2. X-ray diffraction pattern of the AZ31 magnesium alloy sheet.

2.6. Tensile Testing

Tensile tests were conducted on the welded specimens. Specimen preparation for both the welded joints and the as-received AZ31 magnesium alloy was carried out in accordance with ASTM E8, as

illustrated in Figure 3. According to this standard, the minimum specimen length was set to four times its width. Based on preliminary trials and considering the geometry of the specimens, the crosshead speed was set at 0.5 mm/s for the base alloy sheets and 0.2 mm/s for the welded specimens.

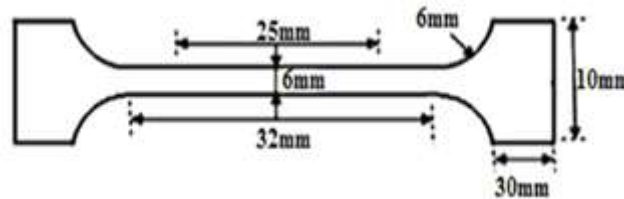


Fig. 3. Dimensions of the ASTM E8 standard tensile test specimen.

2.7. Microhardness Testing

Following the welding process, microhardness measurements were performed across various zones of the specimens. The Vickers microhardness test was conducted using an applied load of 50 g and a dwell time of 10 s. For the hardness profiles presented in Figure 10, all reported values represent the average of at least three measurements, and the error bars indicate the standard deviation.

2.8. Wear Testing

To investigate the tribological behavior of the fabricated joints, a reciprocating wear test was employed. The apparatus measured the frictional force between the contacting surfaces using strain gauges connected to a Wheatstone bridge and an amplifier. The collected data were transferred to software for processing. Since the frictional force and applied load were known, the coefficient of friction (COF) was calculated.

Based on preliminary load-bearing capacity evaluations, an applied load of 30 N was selected over a total sliding distance of 1000 m. The wear tests were conducted at an ambient temperature range of 20–30 °C under dry sliding conditions. Weight loss was measured every 100 m using a digital balance with a precision of 0.0001 g. The counterface used was a 52100 tool steel pin with a hardness of 62 HRC and a diameter of 5 mm. The wear stroke length was set at 4 cm, with a linear pin speed of 0.1 m/s. The hardened 52100 steel pin was selected to provide a stable and reproducible laboratory counterface; therefore, the wear results are intended for comparative evaluation of the as-received, welded, and composite specimens under identical conditions rather than direct simulation of a specific service condition. Furthermore, the morphologies of the worn surfaces were characterized using SEM. For the COF curves shown in Figure 14, all plotted values

represent the average of at least three measurements, and the error bars indicate the standard deviation.

3. Results and Discussion

3.1. Microstructural Investigation of the Base Metal and Weld Metal

In this research, a magnesium-based composite was fabricated using titanium as the reinforcement phase. To this end, ultra-thin titanium foils with a thickness of less than 1 mm were placed at the joint interface of two AZ31 magnesium specimens, and the joining was performed via friction stir welding. Initially, the parameters influencing the process, such as rotational speed, traverse speed, and number of passes, were investigated to determine the optimal conditions for composite formation. Subsequently, the microstructures of the base metal, welded, and composite specimens were evaluated. These investigations focused on the presence of reinforcement particles and the evolution of their distribution under various welding conditions. The mechanical properties were also analyzed through Vickers microhardness profiles, tensile testing, and wear testing. In addition, the fracture surfaces of the tensile specimens and the worn surfaces were studied

using SEM to identify the prevailing deformation and wear mechanisms.

Figure 4 displays the optical microscopy images of the AZ31 magnesium alloy before and after etching. As observed in Figure 4a, no distinct phases, structures, or primary precipitates are visible in the prepared specimen prior to etching. Following the etching process (Figure 4b), the microstructure of the AZ31 alloy reveals equiaxed but non-homogeneous grains. One of the primary phenomena occurring during the FSW process is microstructural refinement and homogenization. Tool geometry is one of the most critical factors influencing this process, as it plays a vital role in material flow and microstructural modification. The rotation of the tool causes stirring and mixing of the material around the rotating pin, while the translation of the tool transfers the stirred material from the front to the back of the pin. This mechanism leads to a more homogeneous and refined microstructure within the welded zone. The refinement and increased homogeneity of the microstructure are expected to enhance both the mechanical and tribological properties of the specimens [5–7].

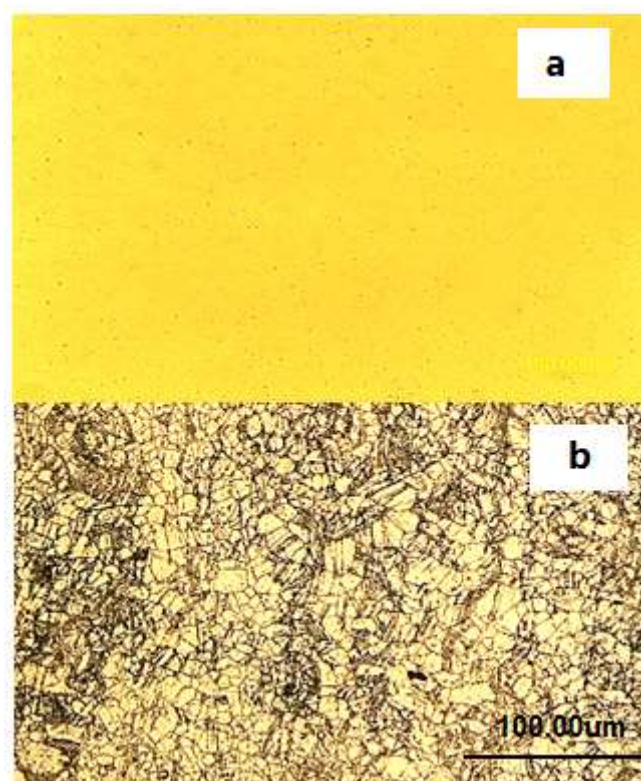


Fig. 4. Optical microscopy image of the AZ31 magnesium alloy base metal microstructure: (a) before etching and (b) after etching.

To investigate the microstructure of the welded specimens, cross-sectional surfaces were prepared and analyzed by optical microscopy. Figures 5–7 present the macroscopic and microscopic images of

the welded specimens. As observed in Figure 5, there is no evidence of blowholes, porosity, or common defects typically associated with poor-quality friction stir welded specimens. Although the heat generated

by the rotational and translational motion of the welding pin can promote diffusion and grain growth, the intense material flow and severe plastic deformation (SPD) inherent to the process result in a more uniform and refined grain structure, as reported in the literature and confirmed in the present study [8–12].

The mechanism of grain refinement during friction stir processing, particularly in magnesium-based alloys, is mainly attributed to the activation of dynamic recrystallization (DRX). DRX is triggered by an increase in dislocation density and the

subsequent interaction and entanglement of dislocation networks. Therefore, severe plastic deformation increases the dislocation density within the weld zone and provides the driving force for DRX, leading to a significantly finer microstructure [13]. Quantitative grain size measurements were not performed in the present study; however, the optical micrographs clearly indicate grain refinement within the stir zone compared with the base metal. Therefore, the Hall–Petch-based discussion is used qualitatively to explain the observed strengthening trend.

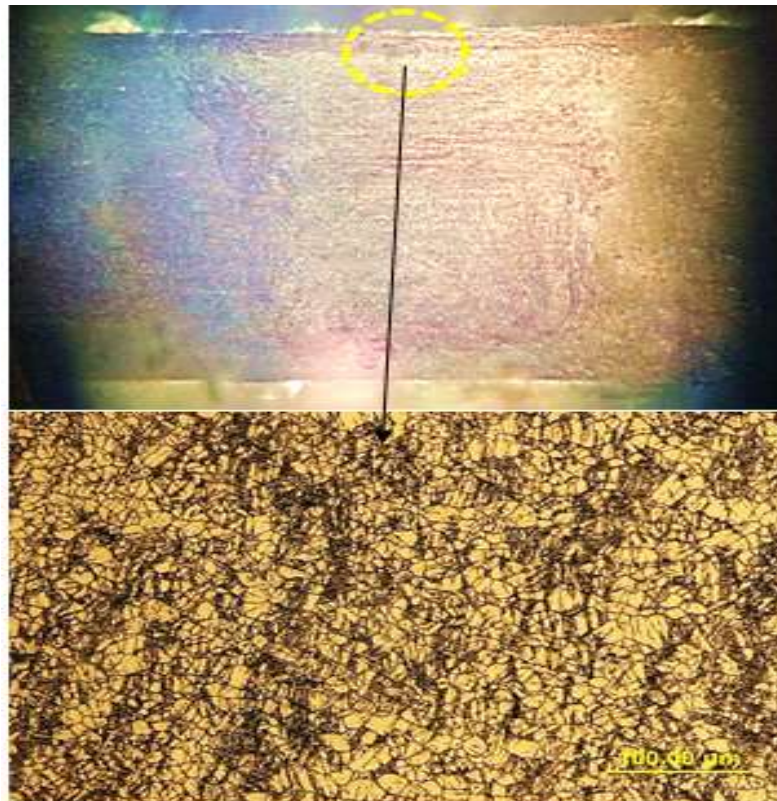


Fig. 5. Optical micrograph of the friction stir welded AZ31 magnesium alloy specimen (non-composite), showing the stir zone (SZ), thermo-mechanically affected zone (TMAZ), and heat-affected zone (HAZ).

3.2. Fabrication of the Magnesium-Based Composite at the Joint Interface

Figures 6 and 7 show the optical microscopy images obtained from the cross-sections and various regions of the composite weld zone after one and three FSW passes, respectively.

As seen in Figure 6, defects such as tunneling, cavities, or voids are absent in the cross-section. These images reveal that while the reinforcement phase particles are embedded within the matrix, their distribution throughout the stir zone is not fully satisfactory after a single pass, and large particle clusters are still observed in some regions. This indicates that although a single FSW pass was effective in bonding the base metal substrates, it was not sufficient to produce a homogeneous composite with a uniform distribution of reinforcing particles.

EDS mapping was not performed in the present study. However, since the reinforcing phase was intentionally introduced as a titanium sheet at the joint interface before welding, the particles observed in the weld zone are attributed to the fragmented and mechanically dispersed titanium phase generated during friction stir welding.

The presence of reinforcement particles at the joint interface increases the interaction between the welding pin and the metals and thus intensifies the magnitude of plastic deformation. During severe plastic deformation, a high density of structural defects, including dislocations, is generated. Furthermore, the frictional interaction and associated strain increase atomic mobility and promote microstructural modification. Collectively, these phenomena result in a more refined and homogeneous

structure in the single-pass composite specimen compared with the non-composite welded sample [14].

To further refine the weld zone microstructure and achieve a more uniform distribution of the reinforcement phase within the AZ31 magnesium alloy matrix, the number of FSW passes was increased to three. As illustrated in Figure 7, the distribution of reinforcement particles in this specimen is significantly more uniform than in the single-pass specimen. Moreover, the grain size in the three-pass composite joint is finer and more homogeneous than in the other specimens. In general,

two primary competing factors influence grain size within the stir zone (SZ) of welded specimens [15]: The heat generated by the friction stir process acts similarly to an annealing treatment and may promote grain growth.

Dynamic recrystallization, resulting from the induced stirring and plastic deformation, increases the number of nucleation sites and leads to grain refinement.

Consequently, as demonstrated by these micrographs, increasing the number of welding passes enhances the degree of stirring and plastic deformation and ultimately yields a structure with significantly finer grains.

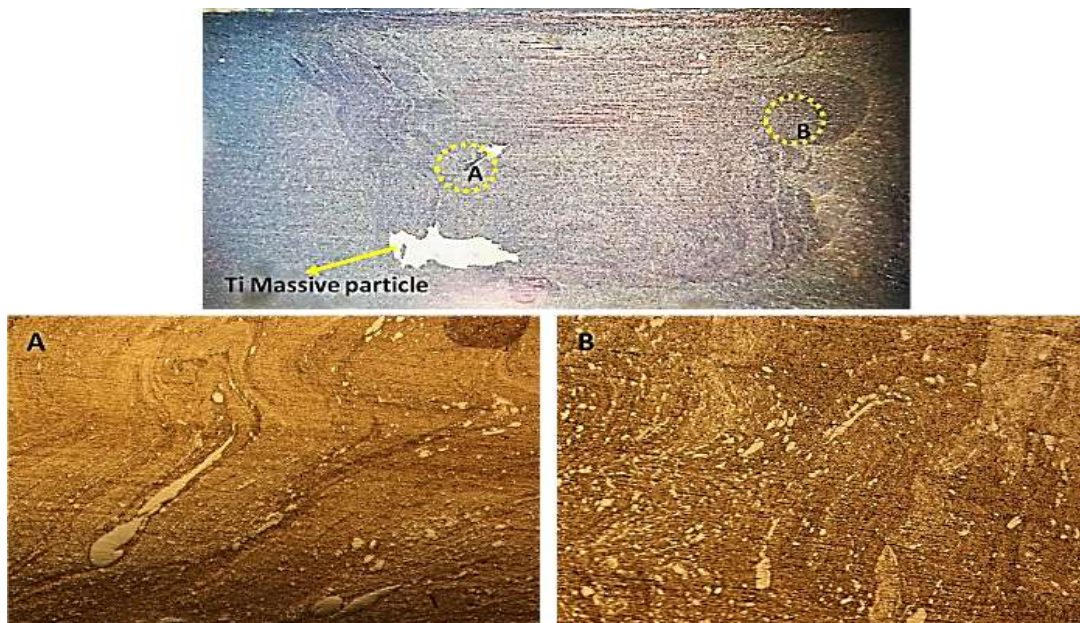


Fig. 6. Optical microscopy images obtained from the stir zone (SZ) and adjacent regions of the composite weld after a single friction stir welding pass.



Fig. 7. Optical microscopy images obtained from the stir zone (SZ) and adjacent regions of the composite weld after three friction stir welding passes.

3.3. Phase Analysis

Figure 8 displays the X-ray diffraction patterns corresponding to the magnesium base metal and the titanium sheet used as reinforcement. As observed, the base metal and titanium sheet contain only the α -Mg and α -Ti phases, respectively, with no peaks indicating secondary or detrimental phases. Subsequently, to fabricate the in-situ composite at the joint interface, the titanium sheet was placed at the joint and friction stir welding was carried out under the optimized parameters of 1200 rpm rotational speed and 8 mm/min traverse speed for one and three passes.

After applying the friction stir welding process for three passes, the local stress and strain increased significantly. This promoted the fragmentation of the titanium sheet and improved the dispersion and distribution of titanium particles in the magnesium

matrix. As observed in Figure 9, the peaks corresponding to titanium are clearly visible. Based on the XRD results, it can be concluded that increasing the number of welding passes improves the dispersion of the reinforcing phase because of severe plastic deformation and the accompanying mixing and stirring of materials.

Importantly, no diffraction peaks corresponding to brittle intermetallic compounds such as Mg–Ti phases, Mg_2Ti , TiH_2 , or Al–Ti phases were detected within the detection limit of XRD. This indicates that the bonding in the Mg–Ti joint region occurred primarily through physical mixing, mechanical interlocking, and intimate interfacial contact rather than through the formation of brittle reaction layers. The absence of such intermetallic compounds is beneficial for preserving the integrity and mechanical stability of the joint.

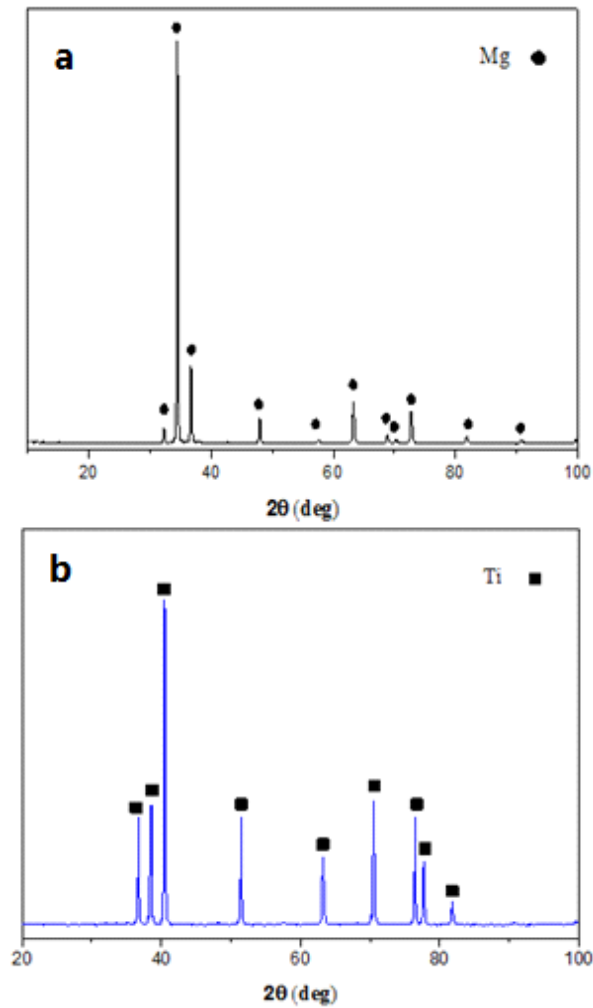


Fig. 8. X-ray diffraction pattern of (a) magnesium base metal and (b) titanium sheet used in this study.

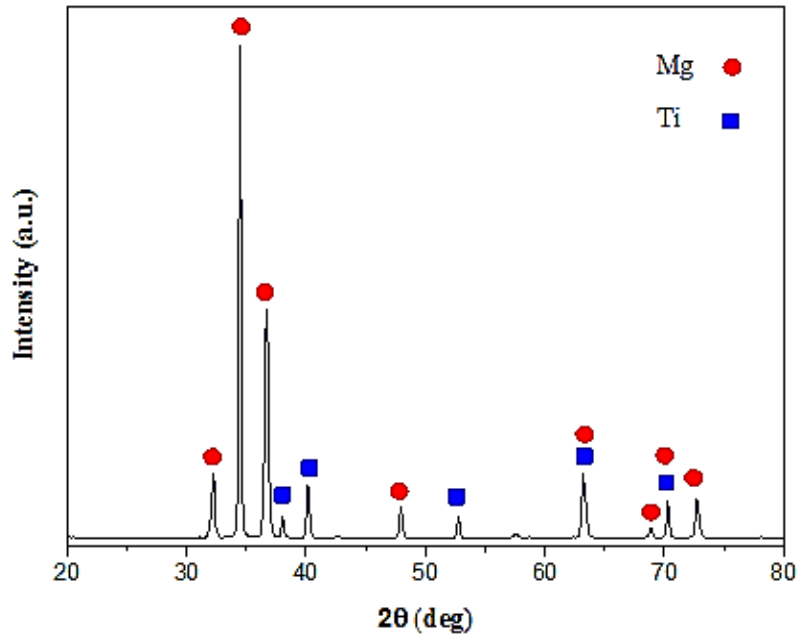


Fig. 9. X-ray diffraction pattern of the composite fabricated through friction stir welding after three passes, with indexed α -Mg and α -Ti diffraction peaks.

3.4. Microhardness Evaluation

To assess the hardness of the as-received specimens and the hardness variations across different regions of the welded specimens, microhardness testing was performed on the cross-sectional surfaces. Figure 10 illustrates the variations in Vickers hardness as a function of distance from the center of the joint area. In the welded specimens, the hardness profile can be related to the characteristic FSW regions, including the stir zone (SZ), thermo-mechanically affected zone (TMAZ), and heat-affected zone (HAZ). The highest hardness values were generally observed near the central weld zone/SZ, where severe plastic deformation and dynamic recrystallization occurred. The adjacent TMAZ experienced both deformation and thermal exposure, while the HAZ was mainly affected by the thermal cycle.

As observed, the average hardness of the as-received specimen is approximately 70 HV. In contrast, the average hardness values of the FSW-welded AZ31 specimen and the one-pass and three-pass composite specimens are approximately 75, 78, and 82 HV, respectively. Further examination of the hardness profiles obtained from the three welded samples reveals that in the non-composite welded specimen, the hardness in the central weld zone increased compared with that of the as-received specimen. As discussed above, friction stir welding produces a finer and more homogeneous microstructure because of

intense material flow and severe plastic deformation. According to the Hall–Petch relationship, hardness increases as the grain size decreases.

For composite materials, the hardness improvement can be attributed to several factors, including grain refinement, the presence of reinforcing particles in the matrix, and the mismatch in thermal expansion and deformation behavior between the matrix and the reinforcement. As shown in Figure 10, the hardness of the composite specimens is significantly higher than that of the as-received and non-composite welded specimens. With the distribution of the titanium reinforcing phase in the magnesium matrix, the higher intrinsic hardness of titanium and the enhanced work hardening in the joint area lead to an increase in hardness. In addition, dislocation strengthening mechanisms, including Orowan strengthening, also contribute to the observed hardness increase. By increasing the number of welding passes, the distribution of reinforcing particles becomes more uniform and the grain size is further reduced; consequently, the average hardness at the center of the joint area increases.

Another notable feature of the hardness profiles is the relative decrease in hardness in the heat-affected zone (HAZ) of the welded specimens. Heat transferred from the weld zone to this region provides favorable conditions for grain growth, and, according to the Hall–Petch relationship, the increase in grain size results in a reduction in hardness [16].

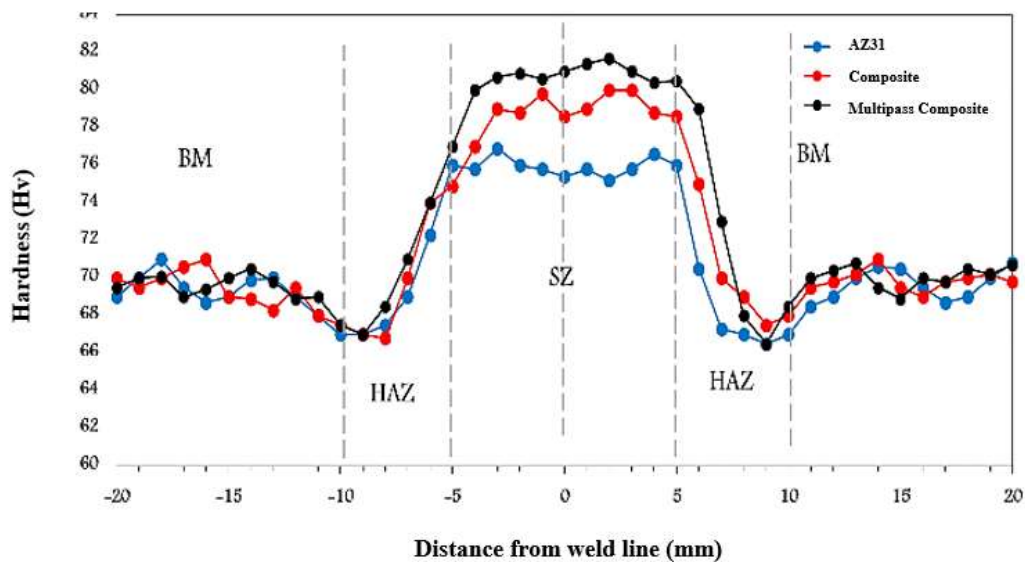


Fig. 10. Hardness profile of the base metal, welded, and composite specimens as a function of distance from the weld line.

3.5. Mechanical Properties Evaluation

3.5.1. Tensile Behavior

The engineering stress–strain curves obtained from the tensile tests of the base metal and the fabricated composites are presented in Figure 11. As can be

seen, composite formation increased the ultimate tensile strength while reducing the ductility of the specimens. The base metal exhibited an ultimate tensile strength of approximately 201 MPa and an elongation of 10.5%. After one FSW pass, the

ultimate tensile strength of the welded non-composite specimen and the composite specimen increased to approximately 213 MPa and 249 MPa, respectively. The results also show that applying three FSW passes increased the ultimate tensile strength from 201 MPa for the AZ31 sheet to about 280 MPa for the composite specimen.

The tensile result corresponding to the 600 rpm and 16 mm/min condition was used during the preliminary optimization stage and was not included in the final tensile comparison because the optimized non-composite and composite joints were produced at 1200 rpm and 8 mm/min. Therefore, Figure 11 and Table 3 focus on the as-received alloy and the final selected welded/composite conditions.

The increase in ultimate tensile strength and yield strength of the composite specimen compared with the base metal can be attributed to grain refinement and the presence of titanium reinforcing particles in the magnesium matrix. In polycrystalline metals, hardness and strength are strongly dependent on

grain size; as the grain size decreases, hardness and yield stress increase. In general, if the particles are uniformly distributed and well bonded to the matrix, they can enhance the strength through the following mechanisms [21]:

- Grain-refinement strengthening (Hall–Petch strengthening)

- Strengthening due to thermal expansion coefficient mismatch

- Orowan strengthening

As noted above, finer grains are often obtained in particle-reinforced welds, mainly because of heterogeneous nucleation on the particles and the inhibition of grain-boundary migration during grain growth. Moreover, finer particles act more effectively as obstacles to the movement of subgrain and grain boundaries, thereby delaying grain growth in the composite even at elevated temperatures [21]. Therefore, an increase in the tensile strength of particle-reinforced composites is expected.

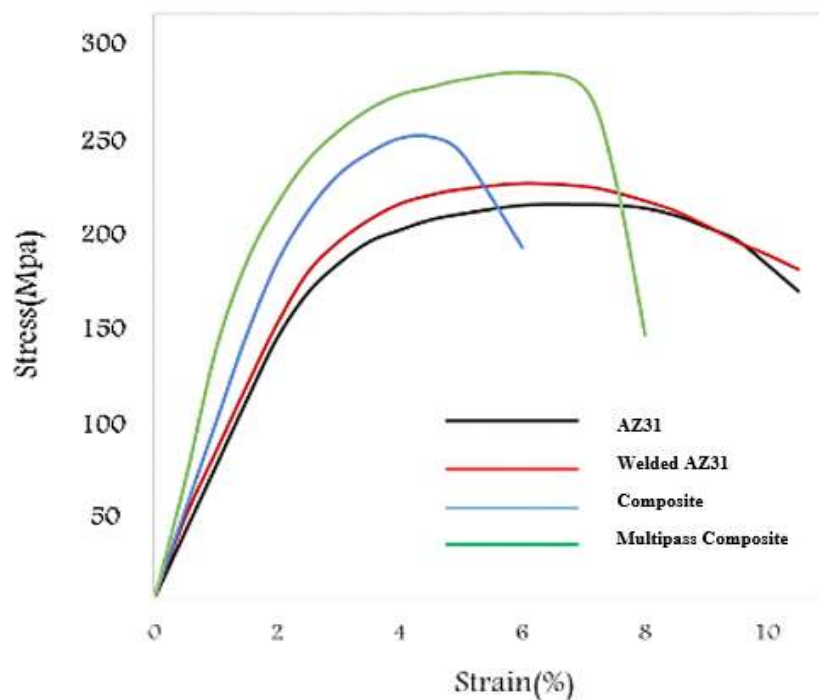


Fig. 11. Engineering stress–strain curves of the base metal and the fabricated composites.

3.5.2. Fracture Surface Examination

Figure 12(a–d) shows SEM images of the AZ31 magnesium base metal, FSW-welded magnesium, and the composites produced after the first and third passes, respectively. As shown in Figures 12a and 12b, both fine and coarse dimples are observed in the as-received and welded specimens.

In general, during tensile testing, localized deformation leads to the nucleation and growth of cracks, and these cracks may propagate along the particle–matrix interface. In the composite

specimens, the presence of titanium reinforcing particles provides additional sites for microvoid and microcrack nucleation. Therefore, the fracture mechanism in these specimens involves the nucleation, growth, and coalescence of microvoids and microcracks. The coalescence of these defects depends strongly on their spacing and distribution.

On the other hand, the presence of hard titanium particles, which are resistant to deformation, within a relatively soft and ductile magnesium matrix imposes constraints on plastic flow. As a result, greater

applied stress is required to initiate plastic deformation in the vicinity of these particles. This is consistent with the observed increase in yield strength and elastic response of the composite compared with the AZ31 magnesium alloy.

The fracture surfaces of the composite specimens (Figures 12c and 12d) show deep and equiaxed dimples, which are characteristic of ductile fracture. Crack growth in ductile fracture requires considerable deformation and therefore consumes a substantial amount of energy. Another important feature of ductile fracture is that it proceeds through slow tearing and rupture accompanied by extensive plastic flow. Initially, plastic strain leads to the formation of microvoids, often at second-phase particles or impurities. With continued tensile loading, these microvoids grow and eventually coalesce. Final fracture occurs when the remaining ligaments between growing voids rupture.

The first stage of ductile fracture involves void nucleation through particle cracking or decohesion at the particle–matrix interface, arising from the mismatch in mechanical properties between the matrix and the particles. Following void nucleation, void growth occurs until coalescence takes place. It should be noted that the number of void nucleation sites is higher in the composite specimens than in pure magnesium. Because of the significant mismatch in the load-bearing capacity between brittle particles and the soft magnesium matrix, severe stress concentration is generated within and around the particles, leading to local damage in the adjacent matrix. The debonding and fracture of particles indicate that localized plastic strain involving a considerable fraction of particles is partly responsible for the reduced ductility of the composite compared with the base alloy.

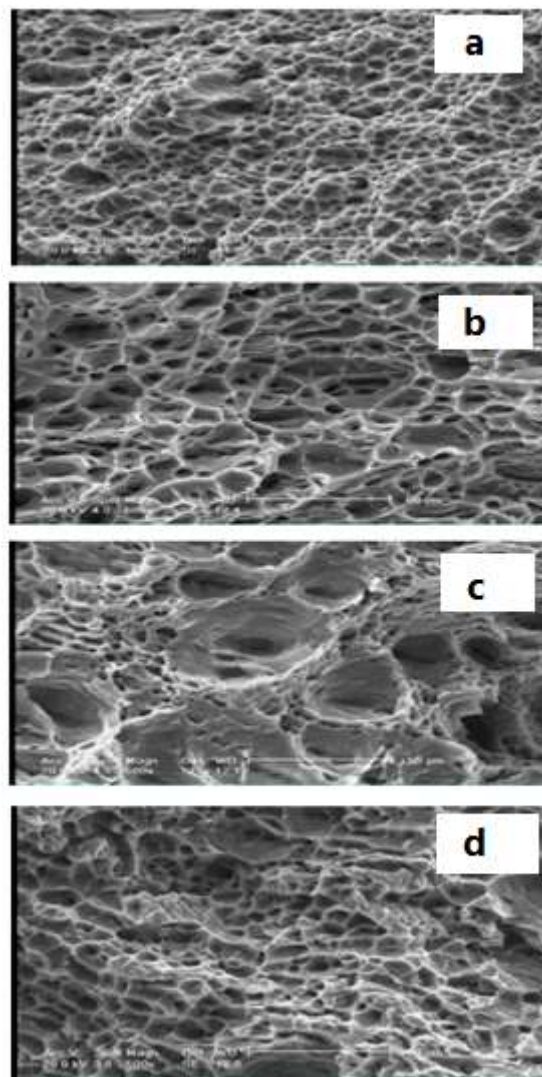


Fig. 12. Scanning electron microscopy images of (a) AZ31 magnesium base metal, (b) FSW-welded magnesium, (c) composite produced after the first pass, and (d) composite produced after the third pass.

3.6. Wear Properties Evaluation

To investigate the wear behavior of the as-received, welded, and composite specimens, reciprocating wear tests were conducted under a load of 30 N. Figure 13 shows the weight loss as a function of sliding distance for the as-received, welded, one-pass composite, and three-pass composite specimens. In all specimens, the weight loss increased with sliding distance. However, the weight loss of the composite specimen was lower than that of the as-received and non-composite welded specimens. It is also evident that the welded specimen exhibited a lower rate of weight loss than the as-received specimen.

Furthermore, increasing the number of welding passes reduced the rate of weight loss during the wear process.

As discussed earlier, friction stir welding refines and homogenizes the structure. The reduction in grain size, increase in grain-boundary area, and increase in structural defects provide additional barriers to dislocation motion and therefore enhance hardness and wear resistance. In the composite specimens, the presence of reinforcing particles also acts as an obstacle to dislocation movement, leading to improved wear resistance and reduced mass loss [16,17].

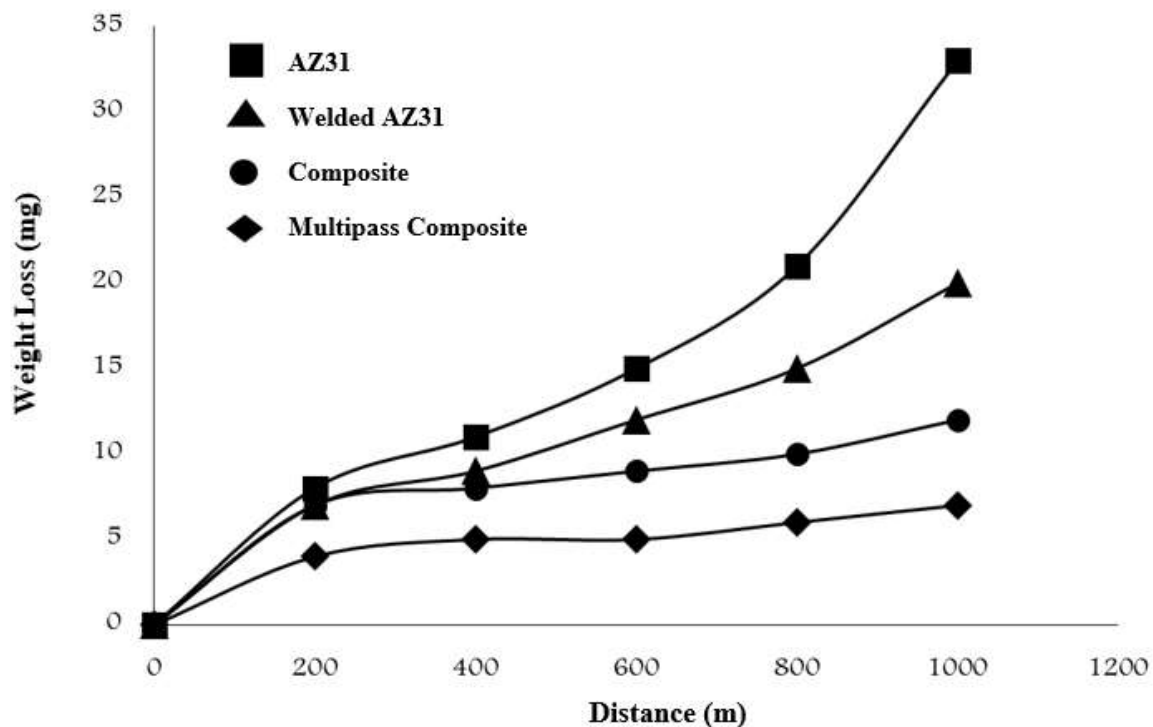


Fig. 13. Weight loss of the as-received, welded, and composite specimens as a function of sliding distance.

Figure 14 shows the coefficient of friction as a function of sliding distance for the tested specimens. As observed, in all specimens the coefficient of friction reached its highest value after approximately 100 m of sliding. At the beginning of the test, the surface asperities of the specimen and the counterface interact, deform, and are gradually removed. After this initial running-in stage, a steady-state condition is established and the coefficient of friction becomes relatively stable. The steady-state coefficient of friction values were approximately 0.42 for the as-received specimen, 0.40 for the friction stir welded non-composite specimen, 0.37 for the one-pass composite specimen, and 0.34 for the three-pass composite specimen. The difference in the coefficient of friction between the as-received specimen and the friction stir welded non-composite specimen is relatively small. In contrast, the

coefficient of friction of the composite specimens, especially that of the three-pass specimen, is lower than those of the as-received and non-composite welded specimens.

When the pin starts moving on the sample surface, plastic deformation occurs in the surface layers, and dislocations begin to multiply. In the composite specimens, the presence of reinforcing particles increases hardness and, according to the Archard relation, this increase in hardness improves wear resistance. Previous studies have shown that reinforcing particles also intensify work hardening by activating dislocation sources more effectively. In other words, the reinforcing phase particles act as load-bearing constituents and, because of their higher hardness compared with the magnesium matrix, reduce the effective load carried by the matrix and improve wear behavior [30].

The use of a 52100 tool steel pin with a hardness of 62 HRC provides a severe and stable counterface condition. Because the hardness of the counterface is considerably higher than that of the AZ31 magnesium matrix, the wear response is influenced

by ploughing and abrasion caused by the hard counterface and detached oxide debris. Therefore, the obtained wear mechanisms should be interpreted as comparative laboratory results under identical dry sliding conditions.

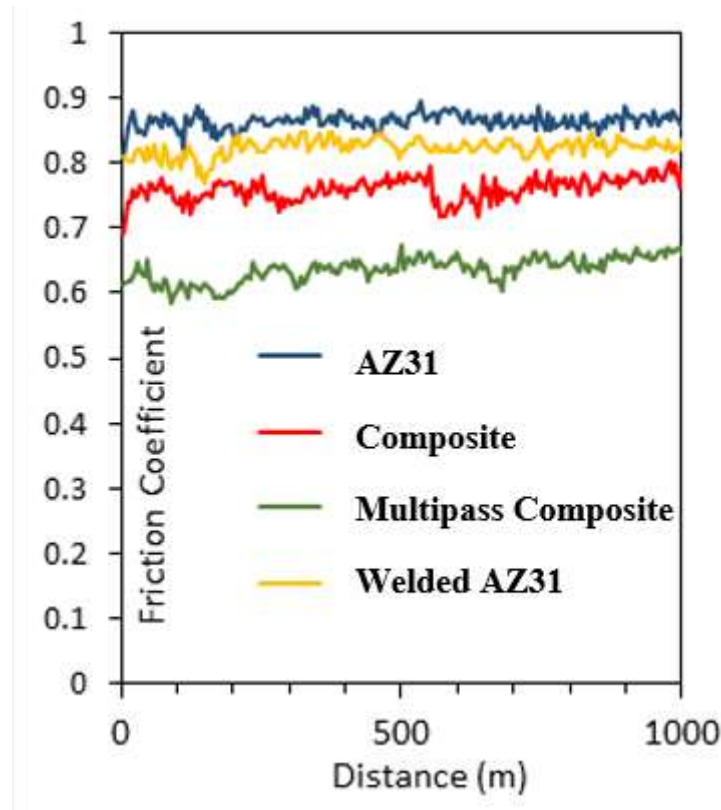


Fig. 14. Coefficient of friction of the as-received, welded, and composite specimens as a function of sliding distance.

To identify the dominant wear mechanisms, SEM images of the worn surfaces were analyzed. Figure 15 shows SEM micrographs of the worn surfaces of all four specimens after a sliding distance of 1000 m.

As observed in Figure 15a, clear evidence of material removal is present on the worn surface of the as-received specimen. Subsequently, severe plastic deformation occurs, and the accumulation of dislocations on the surface alters the wear mechanism. Under these conditions, the formation of shear bands during sliding leads to detachment of material from the specimen surface, indicating an adhesive wear mechanism. In this specimen, localized cold welding between wear debris can also occur; therefore, delamination features are observed in some regions. With continued sliding at longer distances, the detached fine particles are transferred and redistributed over the surface by the counterface, which promotes surface fatigue. The propagation and

coalescence of fatigue-induced cracks result in deep spallation pits and consequently a higher mass-loss rate [29]. Therefore, the dominant wear mechanisms in the as-received specimen are adhesive wear and delamination.

For the non-composite welded specimen (Figure 15b), delamination and abrasive wear mechanisms are observed. Hard oxide debris can also produce grooves on the worn surface [31]. As shown in Figure 15c, composite formation reduces the depth and width of the worn track as well as the severity of the grooves; the prevailing wear mechanisms are abrasive wear and delamination. For the composite specimen produced using three welding passes (Figure 15d), the depth and width of the worn track reach their minimum values, and abrasive wear becomes clearly dominant, while delamination is still present to some extent.

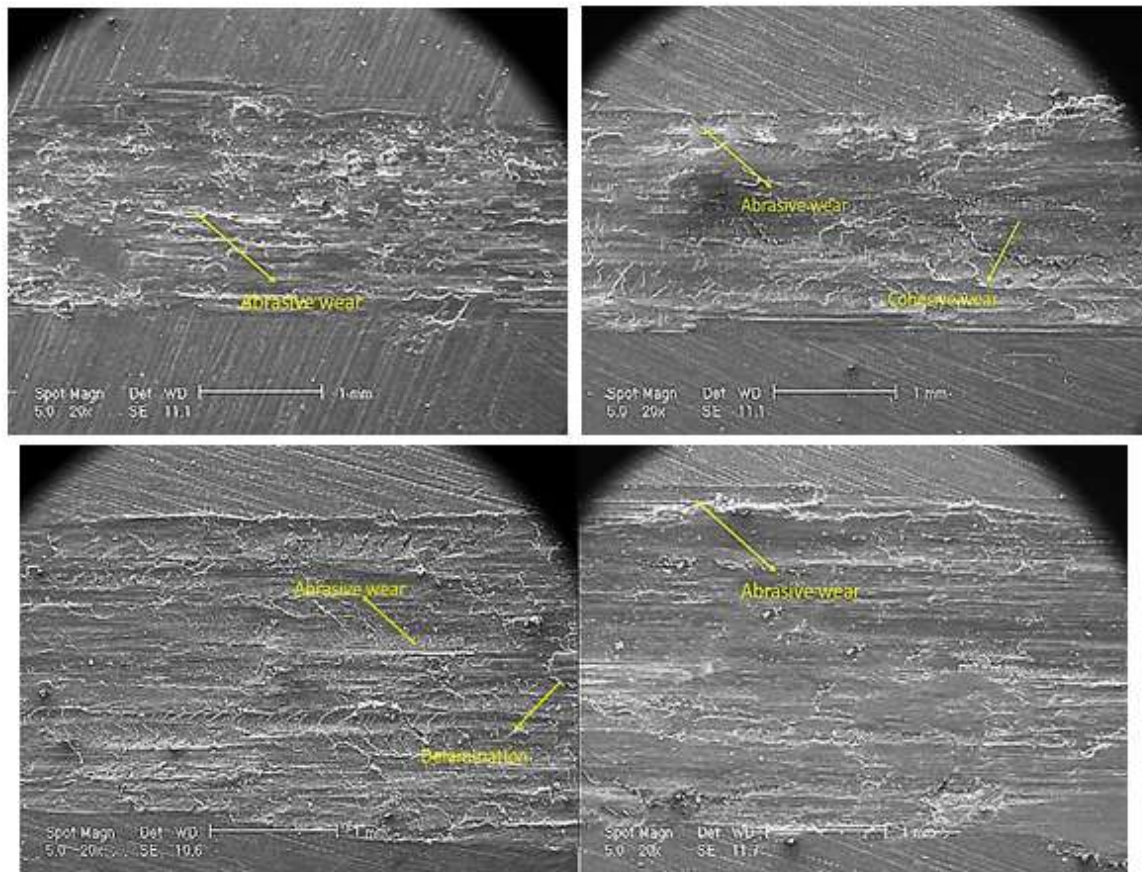


Fig. 15. Scanning electron microscopy images of the worn surfaces of the tested specimens after a sliding distance of 1000 m.

4. Conclusions

In this study, AZ31 magnesium alloy sheets were joined by friction stir welding. Based on extensive preliminary trials, a tool rotational speed of 1200 rpm and a traverse speed of 8 mm/min were selected as the optimum welding parameters. To improve the microstructure and mechanical properties of the welded specimens, a titanium sheet was inserted at the joint interface, and an in-situ composite was produced in the weld region through the stirring action of the rotating tool. One specimen was processed using a single pass, while another was processed using three passes. The microstructural changes and the distribution of the reinforcing phase within the matrix were investigated, and the mechanical and tribological properties were evaluated. The main findings are summarized as follows:

FSW resulted in a more homogeneous and refined microstructure in the AZ31 base alloy.

An in-situ composite was formed in the joint region, and increasing the number of welding passes from one to three led to a more uniform distribution of the reinforcing phase within the matrix.

XRD analysis confirmed the presence of α -Ti reinforcement in the joint region, while no brittle intermetallic compounds such as Mg-Ti, Mg₂Ti,

TiH₂, or Al-Ti were detected at the interface, indicating that the joint was formed mainly through physical/mechanical bonding rather than through formation of brittle reaction layers.

The average hardness increased from approximately 70 HV for the as-received alloy to 75 HV for the non-composite welded specimen, 78 HV for the one-pass composite specimen, and 82 HV for the three-pass composite specimen.

The ultimate tensile strength of the composite specimen increased compared with the non-composite specimen and reached approximately 280 MPa for the three-pass condition.

For all specimens, the coefficient of friction reached a steady state after a sliding distance of about 100 m; the three-pass composite specimen exhibited the lowest coefficient of friction and the lowest mass loss.

With increasing number of passes in the composite specimens, delamination features became less pronounced, while abrasive wear features became more dominant. In the non-composite welded specimen, delamination was dominant, although abrasive wear was also observed. In the as-received specimen, adhesive wear and delamination were the primary wear mechanisms.

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