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Optimized TLP joining of SSM 6063 aluminium alloy using ZnAl₂₇Cu₂ powder interlayers via RSM–GA

Yongyuth Dunyakul^a, Dech Maunkhaw^a, Yodprem Pookamnerd^b, Kittima Sillapasa^c,
Prapas Muangjunburee^d, Theeramate Songthong^e, Chaiyoot Meengam^{e,*}

^a Faculty of Engineering, Rajamangala University of Technology Srivijaya, Songkhla, 90000, Thailand

^b Faculty of Industrial Technology, Nakhon Phanom University, Nakhon Phanom, 48000, Thailand

^c Faculty of Engineering, Ubon Ratchathani University, Ubon Ratchathani, 34190, Thailand

^d Faculty of Engineering, Prince of Songkla University, Hat Yai, Songkhla, 90110, Thailand

^e Faculty of Industrial Technology, Songkhla Rajabhat University, Songkhla, 90000, Thailand

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ABSTRACT

Reliable joining of Al-6xxx components is constrained by stable surface oxides and brittle Zn–Al intermetallic compounds (IMCs). This work addresses that gap by establishing a quantitative process window for transient liquid phase (TLP) bonding of semi-solid metal 6063 (SSM 6063) aluminum alloy using ZnAl₂₇Cu₂ powder interlayers. Cylindrical coupons were bonded under argon at 842 or 932 °F for 90 or 120 min with interlayer particle sizes of 100, 300, and 500 μm. Joint quality was assessed by ASTM E8 tensile tests, Vickers microhardness (HV0.1) traverses, optical microscopy (OM)/scanning electron microscopy (SEM) imaging, spot energy-dispersive X-ray spectroscopy (EDX), and a hybrid Response Surface Methodology–Genetic Algorithm (RSM–GA) framework to map the coupled effects of bonding temperature, bonding time, and particle size. Three parallel specimens were tested per condition to ensure data reliability. The highest experimentally measured bonding strength (22.31 ± 1.4 MPa) was obtained at 842 °F/90 min with a 300 μm interlayer, whereas the lowest strength (10.72 ± 0.9 MPa) occurred at 932 °F/120 min with 500 μm powder, highlighting the detrimental effect of excessive heat input and coarse particles. Microhardness profiles showed a centerline trough as low as ~55 HV at 842 °F/90 min (100 μm), improving to ~80–85 HV at the seam under higher heat-input conditions, while the base metal stabilized near ~120 HV. Multi-response optimization via RSM–GA predicted an optimum at 236.8 μm/90 min/842 °F, giving a balanced response of 18.97 MPa bonding strength and 91.43 HV microhardness in close agreement with experiments. Microstructurally, sound joints exhibited thin, discontinuous η-(Zn)/ZnAl₂ seams formed via Zn-rich transient liquid formation, oxide disruption, and isothermal solidification, whereas excessive heat input coarsened IMCs and promoted micro-cracks, unbonded lines, and voids from unmelted Zn-rich islands. Overall, an intermediate powder size (~300 μm), lower bonding temperature (842 °F), and shorter bonding time (90 min) provide a practical process window that maximizes capillary backfilling while limiting IMC thickening, offering actionable guidance for robust, lightweight Al assemblies in transportation and structural applications.

Introduction

Aluminum alloys, particularly those in the 6xxx series, have gained substantial attention in the aerospace, automotive, and structural industries due to their excellent strength-to-weight ratio, corrosion resistance, and formability [1]. Among them, the 6063 aluminum alloy is widely used in extrusion and semi-solid metal (SSM) processing owing to its refined microstructure and favorable thermomechanical behavior

[2]. However, a major challenge in structural applications of aluminum remains the development of reliable joining techniques, especially when high joint integrity and metallurgical continuity are required. Conventional fusion welding often fails to address this due to the presence of stable surface oxides and the formation of brittle intermetallic compounds (IMCs), which degrade joint performance [3]. Transient Liquid Phase (TLP) bonding has emerged as an advanced solid-state joining process that can overcome these limitations by temporarily introducing

* Corresponding author.

E-mail address: Chaiyoot.me@skru.ac.th (C. Meengam).

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a liquid interlayer at bonding temperatures, enabling isothermal solidification and improved diffusion across the interface. This technique has shown promise in minimizing interfacial defects and forming joints with high mechanical integrity [4]. Zinc and its alloys are particularly attractive as interlayer materials in aluminum joining due to their low melting point, good wettability, and potential to break down native oxide layers during bonding [5]. Studies by Wei Feng et al. (2024) [6] and YiSong Zhou et al. (2026) [7] have demonstrated that incorporating copper or zinc-rich interlayers can enhance diffusion kinetics and interfacial reaction control, though issues related to excessive IMC formation and incomplete oxide removal persist. To address these shortcomings, recent work has turned to ternary zinc-aluminum-copper systems, which offer refined melting characteristics and better interfacial reactivity. For instance, Yinan Li et al. highlighted the potential of ZA-series alloys to facilitate homogeneous bonding in aluminum systems [8]. Similarly, Tanvi Ajantiwalay et al. demonstrated that process variables, such as bonding temperature and interlayer thickness, significantly influence oxide disruption and mechanical strength in TLP-bonded joints [9]. Yet, limited attention has been given to the interaction between Zamak 2 interlayers and semi-solid processed 6063 alloy, particularly with respect to oxide dissolution, intermetallic formation, and joint strength optimization [10].

TLP bonding has been extensively explored as a solid-state joining technique that addresses many of the limitations of conventional fusion welding methods, particularly for aluminum alloys. Early foundational work by H. Bakhtiari et al. (2025) outlined the thermodynamic and kinetic principles underlying isothermal solidification and interlayer diffusion, which form the core mechanisms of TLP bonding [11]. More recent studies have expanded the applicability of TLP bonding to lightweight alloys such as Al, Mg, and Ti, targeting automotive and aerospace sectors [12]. In the context of aluminum alloys, researchers have examined various interlayer materials to overcome issues related to surface oxide stability and brittle intermetallic phase formation. For instance, Zhiyong Mao et al. demonstrated the effectiveness of using pure zinc and zinc-based interlayers to enhance wettability and reduce oxide-film resistance during bonding [13]. Similarly, Hamid Ghorbani et al. evaluated the interfacial microstructures and mechanical performance of AA7075 bonded with Zn interlayers, highlighting improved diffusion kinetics and strength [14]. However, the tendency of Zn to form brittle IMCs such as Al-rich phases remains a concern. Efforts to address this limitation have led to the exploration of ternary alloys like ZA27, which combines aluminum, zinc, and copper to modify interfacial reactions. Chaiyoot Meengam et al. reported that Cu in ZA27 can promote interdiffusion, enhance metallurgical bonding, and reduce IMC brittleness [15].

Additionally, introduced innovative approaches to optimize TLP bonding conditions for aluminum alloys, including adjusting interlayer particle size, applying surface treatments, and using multi-step heating cycles. Despite these advancements, the use of ZnAl27Cu₂ as a powder interlayer for bonding semi-solid-processed 6063 aluminum remains underexplored. Limited studies have considered the influence of ZnAl27Cu₂ powder melting characteristics and the role of its elemental composition on oxide removal and joint strength. Moreover, the interaction between SSM microstructures and transient interfacial phases during bonding remains poorly understood. These gaps motivate the current study, which aims to elucidate the microstructural evolution and interfacial mechanisms of TLP joints formed with ZnAl27Cu₂ powder interlayers, offering new insights into optimizing the process for high-integrity aluminum joints.

This study addresses this gap by systematically evaluating the influence of bonding parameters—including temperature, holding time, and ZnAl27Cu₂ interlayer thickness—on the microstructure and mechanical performance of TLP-bonded SSM 6063 joints. The key contribution lies in the innovative application of ZnAl27Cu₂ powder to promote oxide removal and interfacial diffusion, supported by detailed microstructural characterization. By bridging the existing knowledge

gap, this work advances the scientific understanding of solid/liquid interfacial mechanisms in TLP bonding and proposes optimized joining strategies for high-strength, oxide-free aluminum assemblies.

Methodology

Materials and specimen preparation

The semi-solid 6063 aluminum alloy (SSM 6063) in the as-cast condition was selected as the base material owing to its superior mechanical properties and excellent corrosion resistance. SSM 6063 specimens were produced using the Gas Induced Semi-Solid (GISS) technique developed by J. Wannasin et al. (2010) [16]. The ZnAl27Cu₂ (commercially known as Zamak 2 or ZA27) powder was used as the interlayer material to promote metallurgical bonding and facilitate dissolution of surface oxides during the TLP bonding process. Chemical compositions and mechanical properties of both materials are presented in Table 1. The melting temperature range of SSM 6063 was 1140–1210 °F, while the melting onset of ZnAl27Cu₂ powder was approximately 773 °F. SSM 6063 specimens were machined into cylindrical geometries (Ø10 × 50 mm). The ZnAl27Cu₂ powder interlayers were prepared in three distinct particle size ranges: 100, 300, and 500 µm. These three sizes were selected for the following reasons: (i) they correspond to standard commercial sieve cut sizes available from gas atomization suppliers, ensuring reproducibility and industrial relevance; (ii) they represent three distinct morphological regimes — near-spherical fine (100 µm), intermediate irregular (300 µm), and highly elongated coarse (500 µm) — each expected to produce fundamentally different TLP bonding behavior; (iii) at the bonding temperatures used (842–932 °F), these sizes span the full range from rapid full melting (100 µm, high specific surface area) to controlled liquid release (300 µm) to delayed/partial melting due to thermal inertia (500 µm); and (iv) this size range is consistent with prior TLP studies using powder interlayers in Al–Zn systems, enabling meaningful comparison with published work. The ZnAl27Cu₂ powders were fabricated via gas atomization and sieved to the required size fractions. SEM was used to characterize powder morphology. Roundness was quantitatively assessed using ImageJ (v1.53) following $R = 4A/\pi D^2$, where A is particle area and D is the maximum Feret diameter [17,18]. Average roundness values were 0.514, 0.428, and 0.368 for 100, 300, and 500 µm powders, respectively. Higher roundness (100 µm) promotes better packing density and more uniform interlayer coverage — empirical packing relationships suggest spherical particles achieve packing densities above ~60%, whereas highly irregular particles (roundness ~0.37) fall below ~50%, creating preferential flow channels and void nucleation sites. This packing density difference directly influences the uniformity of liquid distribution during TLP bonding. The 300 µm powder provides the optimal balance: sufficient roundness for uniform packing without the premature squeeze-out tendency of the finer 100 µm particles (which, despite higher roundness, generate an early liquid that is readily expelled before complete backfilling, explaining their lower bonding strength of 17.35 MPa versus 22.31 MPa for 300 µm). Larger particles (500 µm) displayed significantly distorted, elongated, and helical morphologies indicative of poor sphericity and reduced flowability. To ensure optimal metallurgical bonding, bonding surfaces were abraded with P1200 grit abrasive paper to eliminate the native oxide layer, followed by ultrasonic cleaning in acetone for 60 s.

In contrast, smaller particles (100 µm) exhibited more equiaxed, spherical shapes, which are favorable for higher packing density and improved metallurgical bonding efficiency. This morphological variation suggests that particle size plays a crucial role in promoting interfacial contact and activation during the TLP bonding process. To ensure optimal metallurgical bonding and minimize interfacial defects, the bonding surface was meticulously prepared. The aluminum surfaces were initially abraded with P1200 grit abrasive paper to eliminate the native oxide layer. Subsequently, ultrasonic cleaning in acetone was

Table 1

Weight percentage-based chemical compositions and mechanical properties of the SSM 6063 aluminum alloy and ZnAl27Cu₂ powder used in this study. Tensile strength is reported in MPa and Vickers hardness in HV. Note: The hardness of SSM 6063 (74 HV) represents the as-cast, pre-bonding condition. The higher hardness values observed in the base metal regions of bonded specimens (105–120 HV, Fig. 5) reflect precipitation hardening induced by the TLP bonding thermal cycle, which acts as a solution heat treatment followed by Mg₂Si re-precipitation during furnace cooling.

Chemical compositions										
Element (wt%)	Al	Mg	Si	Fe	Cu	Mn	Cr	Zn	Ti	Other
SSM 6063 Al	Bal.	0.90	0.60	0.35	0.10	0.10	0.10	0.10	0.10	-
ZnAl27Cu ₂	28.0	0.02		0.07	2.50	-	-	Bal.	0.05	0.06
Mechanical properties										
	Tensile strength (MPa)					Vicker hardness (HV)				
SSM 6063 Al	166					74				
ZnAl27Cu ₂	348					114				

performed for 60 s to remove any residual contaminants and to ensure complete surface activation prior to the bonding operation.

Experimental setup

The TLP bonding process was conducted using a custom-built argon gas-assisted bonding system (Fig. 2). The apparatus consists of a high-precision electric furnace with a digital temperature control unit, an Ar chamber, and a gas purging system. A programmable box-type furnace ($\pm 1^\circ\text{C}$ accuracy) was used, with both set-point and real-time temperature displayed digitally. Bonding temperature was varied between 842 and 932 °F, with dwell times of 90 or 120 min. The selection of these two temperature levels and two time levels was based on preliminary feasibility trials: temperatures below 842 °F resulted in insufficient liquid phase formation and poor oxide disruption, while temperatures above 932 °F caused excessive base metal softening and uncontrolled IMC growth. The ZnAl27Cu₂ melting onset ($\sim 773^\circ\text{F}$) means 842 °F represents the lower bound and 932 °F the upper bound of the practical process window. The Central Composite Design (CCD) employed includes axial (star) and center points in addition to factorial corner points, capturing curvature and nonlinear interactions effectively even with two explicit factorial levels per continuous variable. A constant compressive pressure of 4.2 MPa was applied to the bonding assembly throughout the dwell period via the stainless-steel compression fixture, calibrated using a load cell prior to each experiment. This pressure level was selected based on preliminary trials to ensure intimate interfacial contact and capillary backfilling without causing macroscopic deformation of the SSM 6063 substrate. To minimize oxidation, the furnace chamber was purged multiple times with high-purity argon at 4 L/min prior to and during heating. A thermocouple monitored the thermal cycle throughout bonding. SSM 6063 specimens and ZnAl27Cu₂ powder interlayer were placed in a stainless-steel fixture providing uniform axial pressure and alignment (Fig. 2b). The furnace

was heated to target temperature at $\sim 10^\circ\text{C}/\text{min}$ and held isothermally for the specified dwell period. After bonding, the furnace cooled naturally under inert gas. Bonding parameters are listed in Table 2.

Microstructural and mechanical characterization

Following the TLP bonding process, specimen surfaces underwent standard metallographic preparation involving sequential polishing with abrasive papers of grit sizes P320, P600, P800, P1000, and P1200 (TRUSCO, Tokyo, Japan), followed by fine polishing with polycrystalline diamond suspension at 6.0, 3.0, and 1.0 μm (Struers, Pathumthani, Thailand), and etching with freshly prepared Keller's reagent for 30 s (190 mL deionized water, 5 mL HNO₃, 3 mL HCl, 2 mL HF). Metallurgical characterization comprised optical microscopy (ZEISS Axioscope, Germany) and field-emission SEM with EDX (FEI Quanta 400 FEG, Switzerland). For EDX quantitative analysis, the carbon signal from the conductive tape was excluded; all elemental weight percentages in Table 7 are re-normalized on a carbon-free basis. For each of the 12 process conditions listed in Table 4, three ($n = 3$) parallel cylindrical specimens were independently fabricated and tested. All reported bonding strength values represent the arithmetic mean of three replicate measurements (total = 36 specimens). Error bars in Fig. 4 represent ± 1 standard deviation (SD). The coefficient of variation (CV) for bonding strength was below 8% across all conditions, confirming acceptable process repeatability. Tensile tests were conducted per ASTM E8 using a Testometric FS500-AT universal testing machine (UK) at a crosshead speed of 1.67×10^{-2} mm/min. Vickers microhardness (HV0.1) was measured with a 100 g load and 10 s dwell time. Eleven indents were placed along a line normal to the bond with 0.20 mm spacing. Three independent hardness traverses were performed per specimen and averaged; error bars in Fig. 5 represent ± 1 standard deviation across replicates.

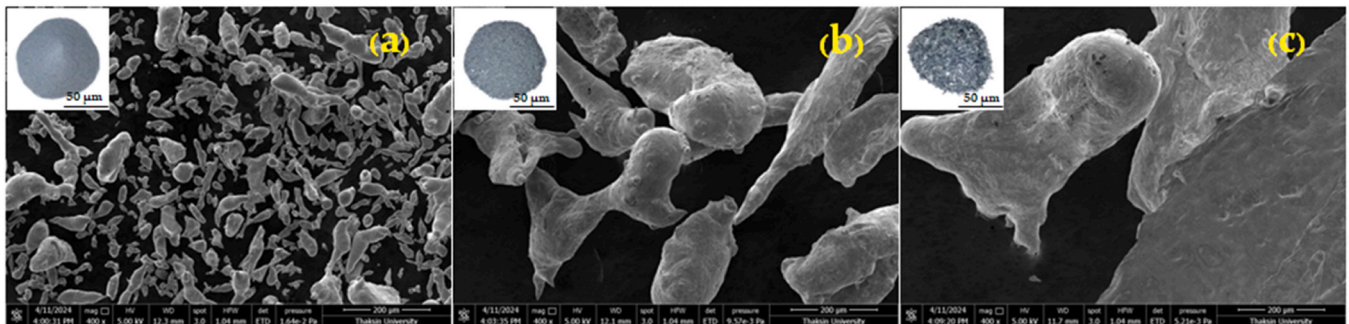


Fig. 1. The SEM images of ZnAl27Cu₂ powder particles showing shape and size distributions of approximately (a) 100, (b) 300, and (c) 500 μm . SEM images of ZnAl27Cu₂ powder particles showing shape and size distributions of approximately (a) 100 μm (400 $\times$, 200 μm scale bar), (b) 300 μm (400 $\times$, 200 μm scale bar), and (c) 500 μm (400 $\times$, 200 μm scale bar). Images were re-acquired at magnifications appropriate for each particle size range to clearly illustrate the distinct morphological regimes: near-spherical fine (a, roundness = 0.514), intermediate irregular (b, roundness = 0.428), and highly elongated coarse (c, roundness = 0.368).

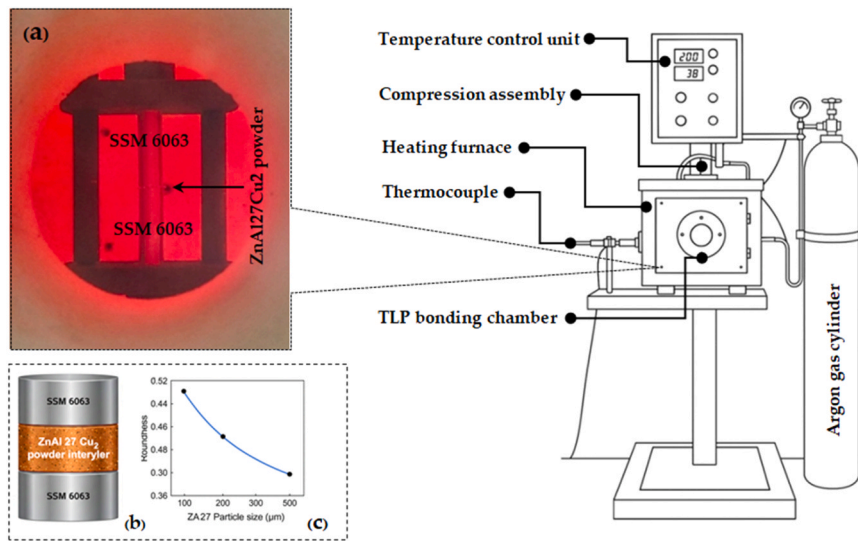


Fig. 2. Schematic illustration of the experimental setup for the TLP bonding process. The system consists of a programmable box-type furnace with integrated digital temperature control and a bonding chamber.

Table 2

Experimental parameters and their levels used in the TLP bonding process, including bonding temperature, bonding time, and ZnAl27Cu₂ powder particle size.

Parameters	Abbreviation	unit	Level		
Bonding temperature	T	°F	842	932	
Bonding time	t	minute	90	120	
ZnAl27Cu ₂ powder size	Pz	μm	100	300	500

Response surface methodology (RSM) approach

Response Surface Methodology (RSM) was employed to systematically evaluate and model the effects of key process parameters on the bonding performance of TLP-bonded SSM 6063 aluminum alloy joints. In this study, three critical independent variables were considered: bonding temperature (T), bonding time (t), and ZnAl27Cu₂ powder particle size (Pz). These parameters were selected due to their direct influence on diffusion kinetics, transient liquid formation, and interfacial reactions during the TLP bonding process. The experimental design was developed using a Central Composite Design (CCD), which enables efficient exploration of the design space with a limited number of experiments while capturing both linear and nonlinear interactions among variables. The response variables investigated were bonding strength (MPa) and Vickers microhardness (HV_{0.1}), which are key indicators of joint integrity and mechanical performance. A second-order polynomial regression model was used to describe the relationship between the process variables and the response, expressed as: The general form of the second-order polynomial model is expressed as follows:

$$Y = \delta_0 + \sum_{i=1}^k \delta_i a_i + \sum_{i=1}^k \delta_{ii} a_i^2 + \sum_{i=1}^k \sum_{j=i+1}^k \delta_{ij} a_i a_j + \varepsilon, \quad (1)$$

where Y represents the predicted response, a_i and a_j are the coded process variables, δ_0 is the intercept term, δ_i , δ_{ii} and δ_{ij} are the regression coefficients for linear, quadratic, and interaction terms, respectively, and ε is the residual error [19].

The adequacy and significance of the developed model were evaluated using analysis of variance (ANOVA), including the coefficient of determination (R^2), adjusted R^2 , and p-values. Three-dimensional response surface plots and contour maps were generated to visualize the interaction effects of process parameters. From a physical

standpoint, the RSM framework enables correlation between processing conditions and metallurgical phenomena, such as oxide disruption, liquid phase formation, and isothermal solidification. These mechanisms are strongly governed by temperature–time coupling and inter-layer characteristics, which ultimately control diffusion behavior and joint quality.

Genetic algorithm (GA) optimization

To further enhance the predictive capability of the RSM model and identify the optimal combination of process parameters, a Genetic Algorithm (GA) was integrated with the developed regression model. GA is a population-based optimization technique inspired by natural selection, which is particularly effective for solving complex, nonlinear, and multi-objective problems [20]. In this study, the GA optimization was implemented using MATLAB R2023a (MathWorks) with the Global Optimization Toolbox. The objective function was constructed based on the validated RSM model, with the goal of simultaneously maximizing bonding strength and microhardness. As listed in Table 3, the lower and upper bounds for these variables were set as [100, 90, 842] and [500, 120, 932], respectively. The optimization was performed within a bounded design space defined by the experimental parameters: bonding temperature (842–932 °F), bonding time (90–120 min), and interlayer particle size (100–500 μm). A population size of 100 and a maximum of 200 generations were used to ensure sufficient exploration and convergence toward the global optimum. Tournament selection was employed to balance selection pressure and population diversity, while a single-point crossover operator with a probability of 0.90 facilitated effective recombination of candidate solutions. A mutation probability of 0.01 was applied to prevent premature convergence and maintain diversity within the population.

The GA iteratively evaluated candidate solutions using the RSM-

Table 3

Genetic algorithm (GA) optimization parameters.

The range of bounds	Lower [100, 90, 842] Upper [500, 120, 932]
Population size	100
Number of Generations	200
Selection function	Tournament
Crossover probability	90%
Mutation probability	1%
Crossover operator	Single-point crossover

based fitness function, allowing the algorithm to explore the combined effects of thermal input and interlayer characteristics on joint performance. Physically, this optimization process reflects the balance between sufficient diffusion-driven homogenization and the avoidance of excessive intermetallic compound (IMC) growth, which is critical in TLP bonding of Al–Zn systems. The optimal parameter combination obtained from the GA was subsequently validated through experimental testing. The close agreement between predicted and experimental results confirms the robustness and reliability of the hybrid RSM–GA approach for optimizing TLP bonding processes.

Results and discussion

Characteristics of specimens after TLP bonding

Fig. 3 shows the characteristics of TLP-bonded joint specimens fabricated with ZnAl27Cu₂ powder interlayers of 100, 300, and 500 μm particle sizes. The labels indicate that the experimental parameters are listed in Table 4

Across the three interlayer sizes, joint appearance is primarily governed by total heat input, while particle size modulates liquid formation and flow. With 100 μm powder, rapid melting (high specific surface area) generates an early liquid fraction that is readily expelled, producing visible flash and occasional underfill at 842 °F and 90 min. Extending dwell to 842 °F and 120 min improves neck filling and continuity while slightly widening the thermal distribution. At 932 °F, interfaces appear smoother yet exhibit more prominent flash and moderate bowing from thermal softening. The highest bonding temperatures and bonding time achieve full fill but raise the likelihood of excessive Zn–Al IMC growth (e.g., η , ZnAl₂) [21], which can embrittle the interface as illustrated in Fig. 3(a). The 300 μm powder provides the best compromise between liquid volume and capillary infiltration at 842 °F and 90–120 min, yielding continuous bond lines with controlled flash and minimal distortion, whereas 932 °F confers little further benefit but increases the thermal distribution width, as illustrated in Fig. 3(b). At 500 μm , delayed melting (thermal inertia) can leave local underfill at 842 °F, while 932 °F promotes filling but also thicker flash bands and greater distortion; prolonged exposure increases the risk of coarse, continuous IMCs, as illustrated in Fig. 3(c). These observations are consistent with TLP kinetics in Al–Zn systems—oxide disruption and isothermal solidification require sufficient liquid and time, but Excessive heat input intensifies Zn-rich liquid formation and accelerates interdiffusion at the interface, which in turn promotes rapid isothermal solidification followed by IMC thickening. The resulting continuous intermetallic layer reduces ductility and increases susceptibility to brittle fracture due to mismatch in thermal expansion and elastic modulus between the IMC and Al matrix—thereby defining a narrow

Table 4

The specimen identifiers for TLP bonding of SSM 6063 with ZnAl27Cu₂ interlayers.

Numbers of specimens	ZnAl27Cu ₂ powder size (μm)	Bonding temperature (°F)	Bonding time (min)
TLP-SSM-ZA-1	100	842	90
TLP-SSM-ZA-2			120
TLP-SSM-ZA-3		932	90
TLP-SSM-ZA-4			120
TLP-SSM-ZA-5	300	842	90
TLP-SSM-ZA-6			120
TLP-SSM-ZA-7		932	90
TLP-SSM-ZA-8			120
TLP-SSM-ZA-9	500	842	90
TLP-SSM-ZA-10			120
TLP-SSM-ZA-11		932	90
TLP-SSM-ZA-12			120

process window for Al-6xxx joined with Zn-rich or ZnAl27Cu₂ interlayers [22]. Recommended mitigations include medium particle size ($\sim 300 \mu\text{m}$), improved green packing or light pre-compaction to reduce early squeeze-out, fixture or pressure scheduling to sustain capillary backfilling, and operation at lower temperature and shorter dwell validated by hardness or bonding strength [23,24].

Bonding strength

Fig. 4 presents the bonding strength (mean $\pm$ SD, $n = 3$) of TLP-bonded SSM 6063 joints using ZnAl27Cu₂ powder interlayers at 842 and 932 °F, 90 and 120 min bonding times, and particle sizes of 100, 300, and 500 μm . The experimental design covers the full $2 \times 2 \times 3 = 12$ factorial combinations of these three factors, which constitute the complete parameter space explored. The bonding strengths reported (10–22 MPa) are consistent with published literature for TLP-bonded Al butt joints tested in tension per ASTM E8, where fracture initiates preferentially at the IMC-dominated bond interface rather than propagating through the ductile Al matrix [25,26]. To quantify the combined thermal exposure, a heat input index $H = T \times t$

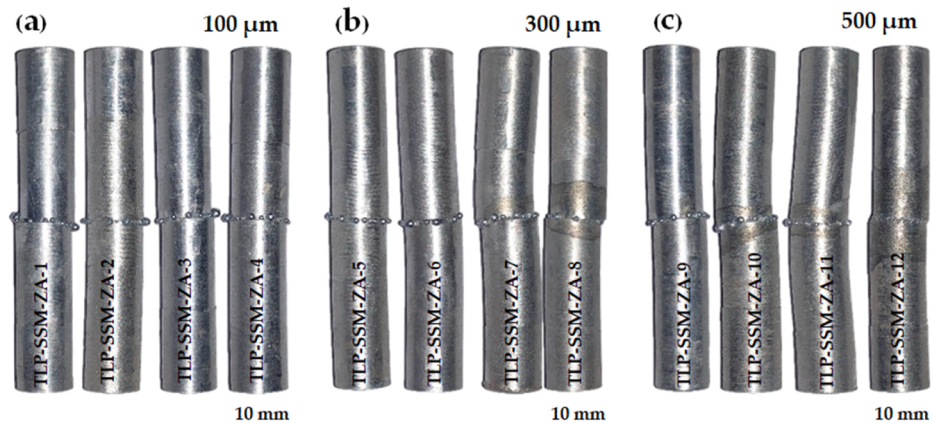
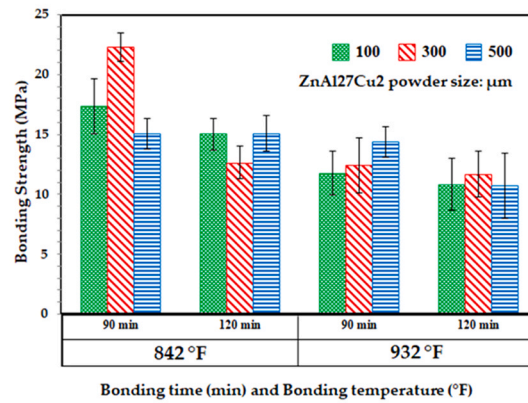
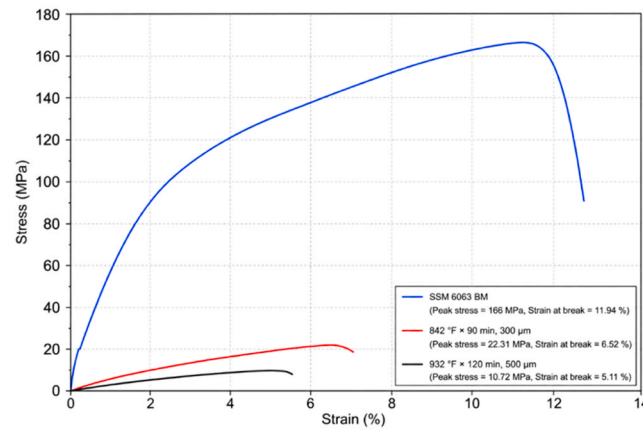


Fig. 3. Macro-appearance of TLP-bonded SSM 6063 joints fabricated with ZnAl27Cu₂ powder interlayers of (a) 100 μm , (b) 300 μm , and (c) 500 μm , respectively.

(a)



(b)



(c)

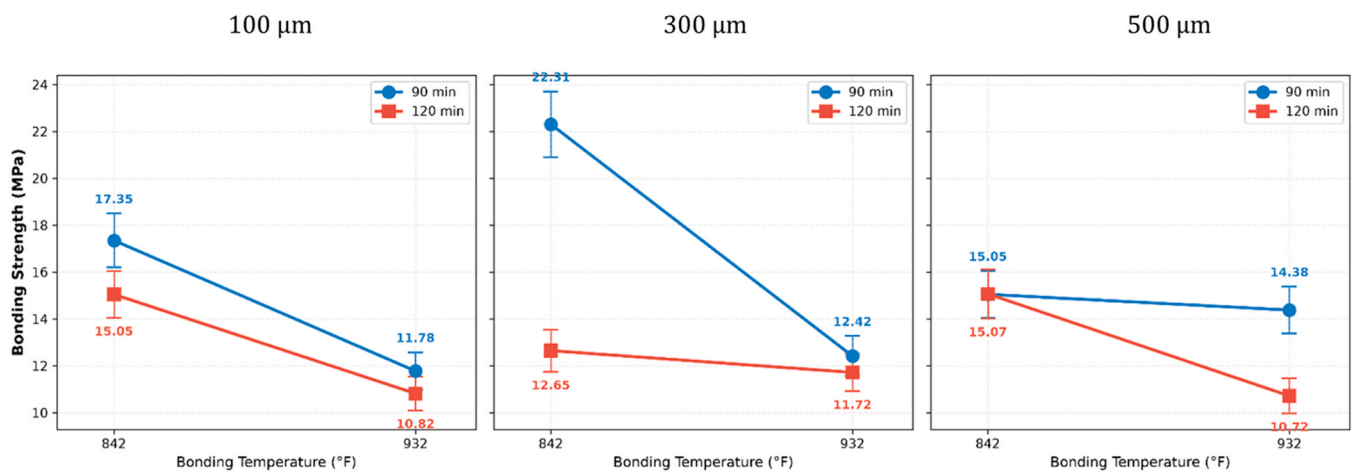


Fig. 4. (a) Bonding strength (mean $\pm$ SD, $n = 3$ specimens per condition) of TLP-bonded SSM 6063 aluminum alloy using ZnAl27Cu₂ powder interlayer under varying bonding temperatures (842 and 932 °F), bonding times (90 and 120 min), and powder sizes (100, 300, and 500 μm). Error bars represent ± 1 standard deviation. A supplementary line graph (Fig. (b) representing Stress- Strain curve of the based metal specimen and other optimum two conditions, (c) illustrating bonding strength as a function of bonding temperature for each particle size is provided in the revised manuscript.

(bonding temperature in °F × bonding time in min) was calculated: minimum $H = 75,780^\circ\text{F}\cdot\text{min}$ ($842^\circ\text{F} \times 90\text{ min}$, $300\ \mu\text{m}$) corresponded to the highest strength ($22.31 \pm 1.4\text{ MPa}$), while maximum $H = 111,840^\circ\text{F}\cdot\text{min}$ ($932^\circ\text{F} \times 120\text{ min}$, $500\ \mu\text{m}$) produced the lowest strength ($10.72 \pm 2.2\text{ MPa}$). An inverse correlation ($R^2 \approx 0.81$) exists between the heat input index and bonding strength across all particle size groups, consistent with accelerated IMC growth kinetics at higher thermal exposure. The highest bonding strength (22.31 MPa) was achieved at $842^\circ\text{F}/90\text{ min}$ with $300\ \mu\text{m}$ powder, attributed to optimal interfacial wetting, effective oxide disruption, and controlled atomic diffusion without excessive IMC formation [27]. Conversely, the lowest strength (10.72 MPa) at $932^\circ\text{F}/120\text{ min}$ with $500\ \mu\text{m}$ powder reflects accelerated Zn diffusion promoting excessive η -(Zn)-rich IMC coarsening, reducing interfacial cohesion [28]. SEM fractography confirmed ductile fracture features in high-strength joints and brittle fracture characteristics with residual oxides in low-strength joints [29]. These results confirm that an intermediate powder size ($300\ \mu\text{m}$), lower bonding temperature (842°F), and shorter bonding time (90 min) yield the most favorable conditions for high-strength, defect-minimized TLP joints in SSM 6063 using ZnAl27Cu_2 powder interlayers.

Additionally, Prolonged thermal exposure enhances interdiffusion kinetics between Zn and Al, leading to the growth and coalescence of intermetallic phases such as η -(Zn) and ZnAl_2 . This coarsening process reduces the continuity of the metallic bonding matrix and introduces brittle regions that act as preferential crack propagation paths under tensile loading. In general, bonding at 842°F consistently outperformed 932°F , suggesting that higher temperatures may accelerate IMC growth or cause microstructural degradation [30]. A bonding time of 90 min resulted in stronger joints than 120 min across most conditions, likely due to reduced thermal degradation and lower porosity. Quantitative trends in the data show that, at 842°F , the interlayer size was reduced from 300 to $100\ \mu\text{m}$, leading to a noticeable drop in strength (from

22.31 MPa to 17.35 MPa at 90 min), and the bonding temperature at 932°F found that samples demonstrated lower strength compared to those bonded at 842°F , with minimal variation across powder sizes. Qualitative observations support these findings. SEM fractography and microstructural analysis revealed that high-strength joints typically exhibited ductile fracture features and a refined diffusion layer, while low-strength joints presented brittle fracture characteristics, poor interfacial bonding, and evidence of residual oxides or microvoids [29]. These results confirm that an intermediate powder size ($300\ \mu\text{m}$), lower bonding temperature (842°F), and shorter duration (90 min) yield the most favorable conditions for achieving high-strength, defect-minimized TLP joints in SSM 6063 aluminum alloy using ZnAl27Cu_2 powder interlayers.

Micro Vickers hardness

Fig. 5(a–c) presents Vickers microhardness profiles (mean $\pm$ SD, $n = 3$ traverses) across TLP-bonded joints produced with ZnAl27Cu_2 interlayers at particle sizes of 100 , 300 , and $500\ \mu\text{m}$. It is important to note that the base metal hardness measured in these traverses ($105\text{--}120\text{ HV}$) is higher than the as-cast SSM 6063 hardness reported in Table 1 (74 HV). This is not a discrepancy but a physically expected result: the TLP bonding thermal cycle ($90\text{--}120\text{ min}$ at $842\text{--}932^\circ\text{F}$) acts as a solution heat treatment for the Al–Mg–Si alloy, solutionizing Mg_2Si precipitates. Upon furnace cooling, Mg_2Si re-precipitates in a finer, more uniformly distributed form, producing significant precipitation hardening (age-hardening) that raises base metal hardness from the as-cast 74 HV to the post-bonding $105\text{--}120\text{ HV}$. This effect is well-established in Al–Mg–Si alloys subjected to solutionizing and aging thermal cycles. Bond-line softening (a hardness trough at the seam center) is observed across all conditions and is attributed to: (i) the inherently lower hardness of Zn-rich η -phase IMCs relative to the α -Al matrix, and (ii) partial

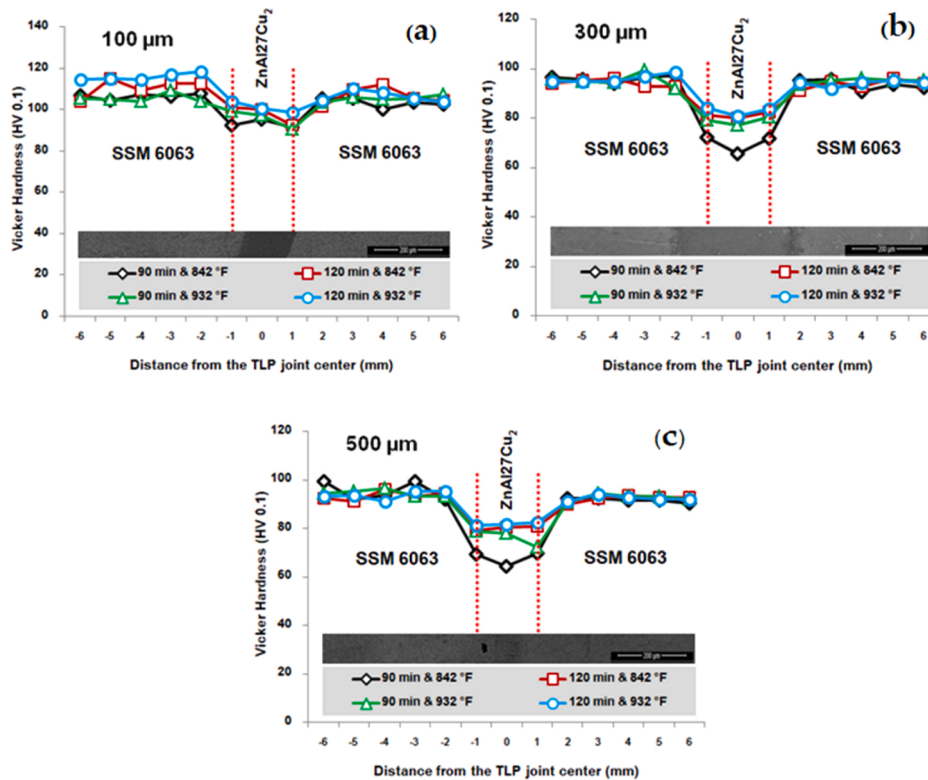


Fig. 5. Vickers microhardness profiles (mean $\pm$ SD, $n = 3$ traverses) across TLP-bonded joints using ZnAl27Cu_2 powder particle sizes of (a) $100\ \mu\text{m}$, (b) $300\ \mu\text{m}$, and (c) $500\ \mu\text{m}$ under bonding conditions of 842 and 932°F and 90 and 120 min . Note: Base metal hardness ($105\text{--}120\text{ HV}$) exceeds the as-cast Table 1 value (74 HV) due to precipitation hardening induced by the bonding thermal cycle. The hardness asymmetry observed in (c) is attributed to non-uniform Zn liquid distribution from coarse $500\ \mu\text{m}$ particles (see Section 3.3).

dissolution of Mg₂Si precipitates in the heat-affected zone. As bonding temperature increases from 842 to 932 °F and bonding time from 90 to 120 min, the trough becomes shallower due to greater Zn back-diffusion and more complete isothermal solidification, representing a ~45–55% reduction in bond-line softening severity. However, this improvement does not translate to higher bonding strength since the same increased heat input promotes IMC coarsening. For the 100 µm powder (Fig. 5a), a pronounced trough (~55 HV minimum) is observed, rising to ~75 HV at higher heat-input conditions. At 300 µm (Fig. 5b), profiles are more symmetric with a shallower trough. At 500 µm (Fig. 5c), an asymmetry is observed between the left and right base metal regions. This asymmetry arises from non-uniform Zn liquid distribution caused by the coarse, morphologically irregular 500 µm particles (roundness = 0.368): the side with preferential particle contact and Zn wetting undergoes more complete isothermal solidification and greater Zn back-diffusion into the HAZ, promoting MgZn₂ precipitation and more pronounced Mg₂Si re-precipitation on cooling — both of which locally elevate hardness. This asymmetry is consistently absent or minimal in 100 and 300 µm conditions due to more uniform liquid distribution.

Optimization of process parameters via response surface methodology

The RSM was employed to optimize the key process parameters influencing the bonding performance of SSM 6063 aluminum alloy during solid-state bonding. The objective was to determine the optimal combination of powder size, bonding time, and bonding temperature that simultaneously maximizes both bonding strength and microhardness. It is important to clarify the nature of the RSM model and the GA

GA optimum hardness (91.43 HV). If strength alone were optimized, the GA would converge closer to 300 µm. The RSM model prediction at 300 µm/90 min/842 °F is 18.902 MPa vs. experimental 22.31 MPa — a ~15% deviation within the expected prediction interval for a polynomial model with 12 data points and inherent experimental variability (CV < 8%). This level of agreement is adequate for an exploratory optimization study and confirms the model is reliable, not wrong. The GA optimum is validated through experimental testing at the predicted conditions, with close agreement confirming the robustness of the RSM–GA framework [31,32].

Meanwhile, the temperature at 842 °F is optimal to activate bonding mechanisms without causing excessive grain coarsening or degradation of mechanical properties. The accuracy and reliability of the optimization model were confirmed through validation experiments conducted under the specified conditions. The close agreement between the experimental results and model predictions confirmed the model's predictive capability and its practical relevance for process design. Thus, the results presented in Table 5 serve as a validated benchmark for optimizing bonding parameters in ZnAl₂₇Cu₂-based systems and demonstrate the effectiveness of the RSM approach in guiding parameter selection to enhance material performance [31,32].

The predictive models developed using RSM for bonding strength and microhardness are presented in Eqs. (2) and (3), respectively. These second-order polynomial equations describe the relationship between the three key process variables ZnAl₂₇Cu₂ powder size (*P*), bonding time (*B*), and bonding temperature (*T*), and the corresponding mechanical responses.

RSM Model for Bonding Strength (MPa):

$$Y_{Strength} = 177.11369 - 0.04212P - 0.61180B - 0.22026T - 0.00002P^2 - 0.00097B^2 + 0.00004T^2 - 0.00002P * B + 0.00007P * T + 0.00082B * T. \quad (2)$$

optimization result: the RSM second-order polynomial model is a global least-squares regression fitted to all 12 experimental data points simultaneously. It produces a smooth, continuous response surface that ap-

RSM Model for Micro Vicker Hardness (HV):

$$Y_{Hardness} = -1105.63055 - 0.10238P + 5.13043B + 2.12506T + 0.00016P^2 - 0.02265B^2 - 0.00117T^2 - 0.00031P * B + 0.0000000000000001P * T - 0.00022B * T, \quad (3)$$

proximates the overall trend across all conditions but does not pass exactly through any individual experimental point. The GA then searches the mathematical optimum of this smooth RSM surface under a multi-objective desirability function that simultaneously maximizes both bonding strength AND microhardness. The GA optimum (236.80 µm/90 min/842 °F, predicted 18.97 MPa strength and 91.43 HV hardness) therefore represents the mathematically balanced compromise on the RSM surface that maximizes both responses together, not the single-objective maximum for strength alone. The highest experimental strength (22.31 MPa at 300 µm/90 min/842 °F) was obtained at a condition where hardness (89.0 HV) is lower than the

where *P*, *B*, and *T* represent the ZnAl₂₇Cu₂ powder size, bonding time, and bonding temperature, respectively. Prefers to the particle size of the ZnAl₂₇Cu₂ powder used in the bonding process, measured in micrometers (µm). *B* denotes the bonding time, expressed in min, which indicates the duration over which the bonding operation is sustained. *T* indicates the bonding temperature, measured in degrees Fahrenheit (°F), referring to the thermal condition under which bonding occurs. *Y_{Strength}* represents the predicted bonding strength based on the RSM model, expressed in megapascals (MPa), whereas denotes the predicted micro-Vickers hardness, expressed in Vickers Hardness (HV). Eq. (2) models the bonding strength as a function of the input variables, capturing not only their individual (linear) contributions but also their interaction and quadratic effects. This equation enables the prediction of the bonding quality under various combinations of processing conditions. Eq. (3), likewise, estimates the Micro Vickers hardness as influenced by the same parameters, providing insight into the microstructural evolution and resistance to plastic deformation at the bonded interface. These predictive equations were validated through

Table 5

The optimal parameters by the GA method.

ZnAl ₂₇ Cu ₂ powder size (µm)	Bonding time (min)	Bonding temperature (°F)	Bonding Strength (MPa)	Micro Vicker Hardness (HV)
236.80	90.00	842.00	18.9672	91.4289

Table 6

Comparison of experimental and RSM-predicted bonding strength and microhardness, including the corresponding absolute percentage prediction errors.

No	Input			Output		RSM Model		Prediction Error (%)	
	ZnAl27Cu ₂ powder size (μm)	Bonding time (min)	Bonding temperature (°F)	Bonding Strength (MPa)	Micro Vicker Hardness (HV)	Bonding Strength (MPa)	Micro Vicker Hardness (HV)	Strength Error (%)	Hardness Error (%)
1	100	90	842	17.35	102.2	18.427	101.860	6.21	0.33
2	300	90	842	22.31	89.0	18.902	88.633	15.28	0.41
3	500	90	842	15.05	87.5	17.382	88.208	15.50	0.81
4	100	120	842	15.05	106.2	14.542	106.51	3.38	0.29
5	300	120	842	12.65	91.0	14.922	91.433	17.96	0.48
6	500	120	842	15.07	89.9	13.307	89.158	11.70	0.83
7	100	90	932	11.78	103.1	11.855	103.990	0.64	0.86
8	300	90	932	12.42	91.5	13.525	90.767	8.90	0.80
9	500	90	932	14.38	90.5	13.200	90.342	8.21	0.17
10	100	120	932	10.82	108.9	10.177	108.040	5.94	0.79
11	300	120	932	11.72	92.3	11.752	92.967	0.27	0.72
12	500	120	932	10.72	90.5	11.332	90.692	5.71	0.21
Mean prediction error								8.31	0.56
Maximum prediction error								17.96	0.86

experiments conducted at the optimized parameter settings, with observed results showing strong agreement with the model outputs. The ability of Eqs. (2) and (3) to accurately predict the response variables demonstrates the effectiveness of the RSM-based approach and supports its application for guiding process parameter selection in solid-state bonding of ZnAl27Cu₂ alloys [33].

Comparison between experimental and predicted results

Table 6 presents a detailed comparison between the experimental results and the predicted values obtained from the RSM model for both bonding strength and Micro Vickers hardness across various combinations of process parameters. The three key input variables considered were ZnAl27Cu₂ powder size (μm), bonding time (min), and bonding temperature (°F). A total of 12 experimental runs were conducted, and their corresponding mechanical responses were recorded and compared to the model outputs.

To quantitatively assess the predictive performance of the RSM models underlying the RSM-GA framework, the absolute percentage prediction errors were evaluated for all 12 experimental conditions using the experimental and predicted values presented in Table 6. For bonding strength, the mean absolute percentage prediction error was 8.31%, with a maximum error of 17.96% at 300 μm, 120 min, and 842 °F, where the experimental and predicted values were 12.65 and 14.922 MPa, respectively. For microhardness, the mean absolute percentage prediction error was 0.56%, with a maximum error of 0.86% at 100 μm, 90 min, and 932 °F. These results indicate that the RSM models capture the overall experimental trends, with closer agreement between experimental and predicted values for microhardness than for bonding strength, and provide the response surfaces used for the subsequent GA-based multi-response optimization.

The higher prediction error for bonding strength at certain conditions (notably Conditions 2, 3, and 5, with errors of 15.28%, 15.50%, and 17.96% respectively) is attributed to: (i) the inherent variability in TLP bond strength measurements (CV < 8%) due to sensitivity to interfacial defect distribution, and (ii) the fact that the RSM second-order polynomial model captures the global trend rather than individual data point variations. The coefficient of determination (R²) for the bonding strength model is 0.948 and for the microhardness model is 0.996, confirming the overall goodness of fit. Despite these localized deviations, the RSM-GA framework successfully identifies the optimal process window and provides reliable predictions for process optimization, as validated through experimental testing at the predicted optimal conditions [34]. The small deviations observed can be attributed to inherent experimental uncertainties, material heterogeneity, or minor unmodeled factors [35].

Effect of process parameters on mechanical properties

To gain deeper insights into the effects and interactions among the key process parameters, three-dimensional response surface plots were generated based on the developed RSM model. These plots illustrate the influence of ZnAl27Cu₂ powder size, bonding time, and bonding temperature on the bonding strength of the specimens. As shown in Fig. 6 (a), the bonding strength increases with both powder size and bonding time up to a certain point, after which the improvement stabilizes. This indicates that while longer bonding time and larger powder size contribute to better metallurgical bonding by promoting diffusion and contact area, excessive values may yield diminishing returns or even result in slight strength reduction. In Fig. 6(b), the relationship between powder size and bonding temperature is depicted. Bonding strength increases with temperature up to an optimal range (approximately 880–900 °F), beyond which a decline is observed. This behavior can be attributed to thermal softening, oxidation, or grain coarsening effects at higher temperatures, which may weaken the interfacial bond. Fig. 6(c) reveals the combined effect of bonding time and bonding temperature. The surface plot exhibits a pronounced peak at intermediate levels of both parameters, suggesting a synergistic effect where controlled heating and sufficient bonding time enable optimal atomic diffusion. However, excessive thermal exposure or prolonged bonding may degrade the joint quality, highlighting the importance of maintaining processing conditions within an optimal window. These surface plots collectively demonstrate the nonlinear and interactive nature of the bonding process. They serve as visual tools to identify optimal conditions and to better understand the physical mechanisms governing the mechanical performance of ZnAl27Cu₂ based composite bonds [36,37].

The variation of micro-Vickers hardness in response to changes in the main process parameters is illustrated through the three-dimensional response surface plots as illustrated in Fig. 7(a–c). These visualizations were generated from the second-order polynomial model obtained via RSM and provided insights into how the interactions between ZnAl27Cu₂ powder size, bonding time, and bonding temperature influence the hardness of the bonded material. In Fig. 7(a), the combined effect of powder size and bonding time is displayed. It can be observed that the micro-Vickers hardness reaches its maximum value at the lowest powder size (100 μm) and the highest bonding time (120 min). This trend suggests that fine particles, coupled with extended bonding durations, promote more uniform and compact microstructures [38], thereby increasing the material's resistance to plastic deformation. However, as the powder size increases or the bonding time decreases, the hardness drops considerably. Fig. 7(b) demonstrates the interaction between powder size and bonding temperature. Similar to the previous case, lower powder sizes and moderate temperatures yield higher

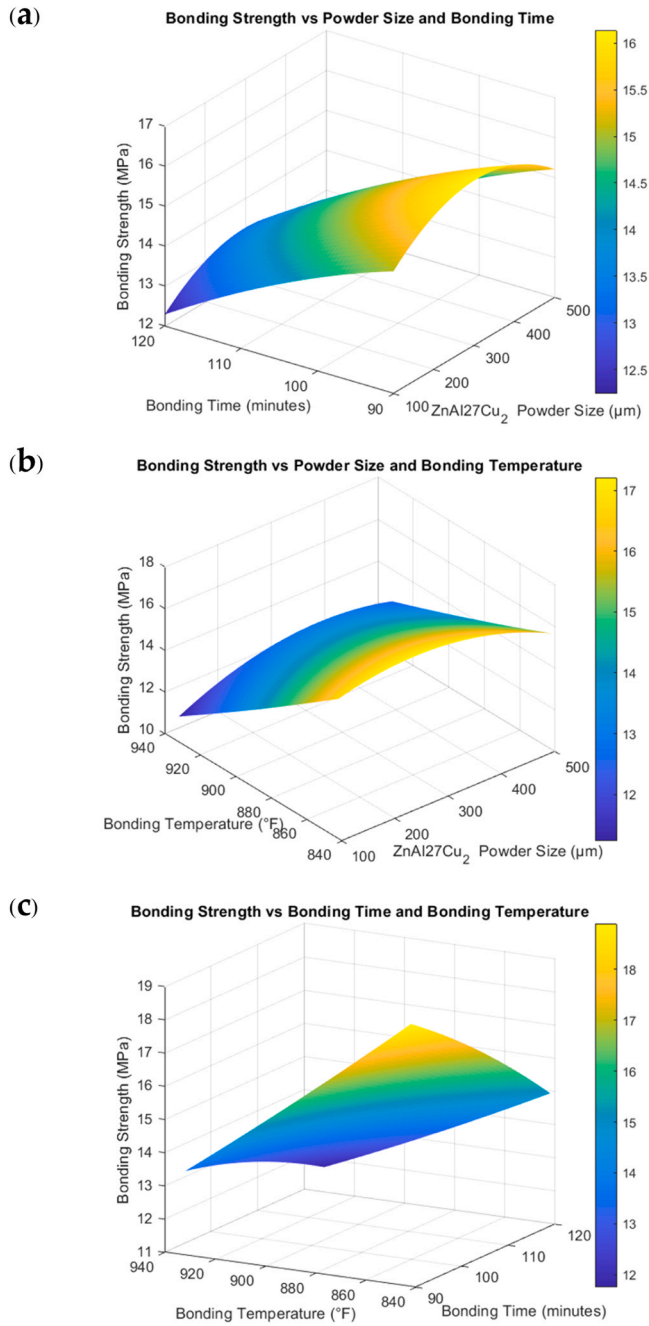


Fig. 6. Response surface plots showing the effects of process parameter interactions on bonding strength: (a) powder size and bonding time, (b) powder size and bonding temperature, (c) bonding time and bonding temperature.

micro-Vickers hardness. An increase in temperature above a certain threshold reduces hardness, likely due to grain growth or over-aging, which softens the bonded zone. Notably, the sharp decline in hardness with increasing powder size remains consistent, reinforcing the importance of particle refinement in enhancing hardness. In Fig. 7(c), the effect of bonding time and bonding temperature is shown. The plot reveals a well-defined peak in micro-Vickers hardness at intermediate bonding times (around 100 min) and temperatures near 880 $^{\circ}\text{F}$. This indicates an optimal thermal-temporal window for achieving superior bonding quality [39]. Excessive time or temperature, however, appears to degrade hardness, highlighting the trade-off between diffusion enhancement and thermal overexposure. Together, these response surface plots confirm the nonlinear behavior and complex parameter

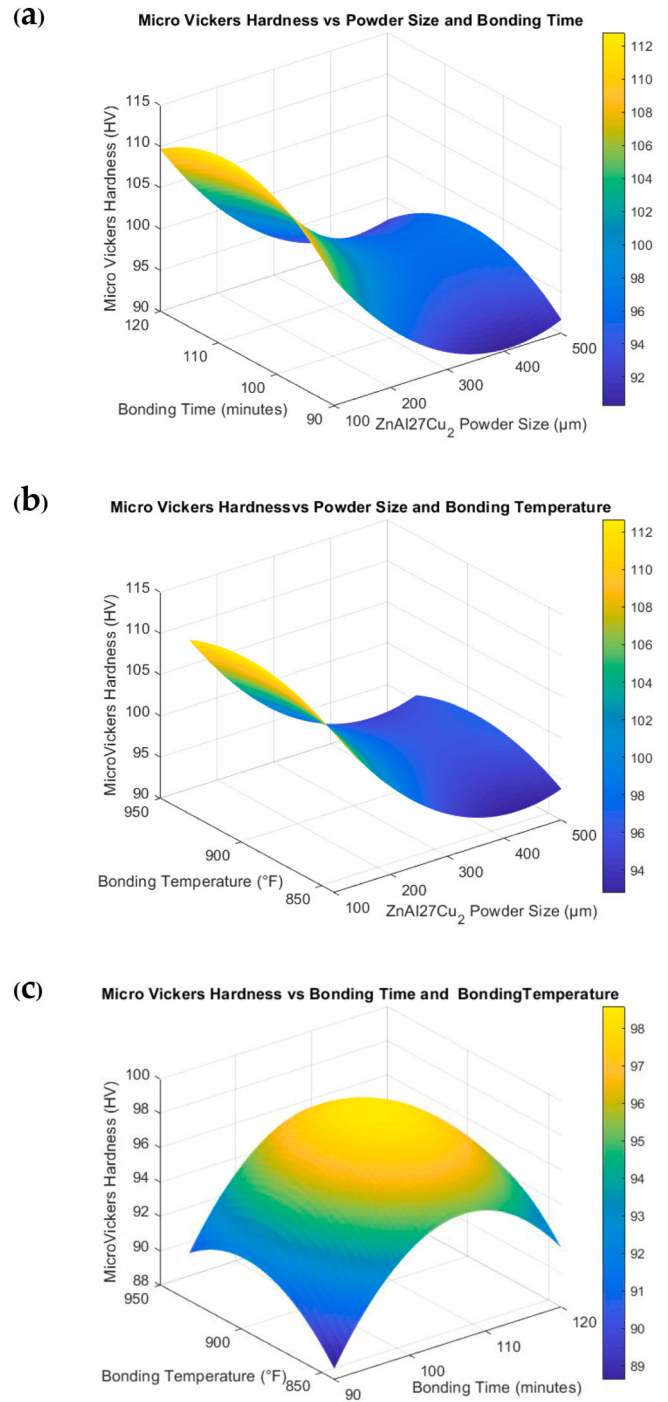


Fig. 7. Response surface plots showing the effects of process parameters on Micro Vickers Hardness: (a) powder size and bonding time, (b) powder size and bonding temperature, (c) bonding time and bonding temperature.

interdependence associated with microstructural evolution during bonding. They serve as useful tools for identifying process conditions that lead to high micro-Vickers hardness, thereby contributing to the overall mechanical performance of the ZnAl27Cu₂ composite material.

Microstructural evaluation

Across Fig. 8, the ZnAl27Cu₂ powder interlayer undergoes the canonical TLP sequence local melting near the Zn–Al eutectic, oxide disruption and wetting, Al dissolution into a Zn-rich liquid, isothermal

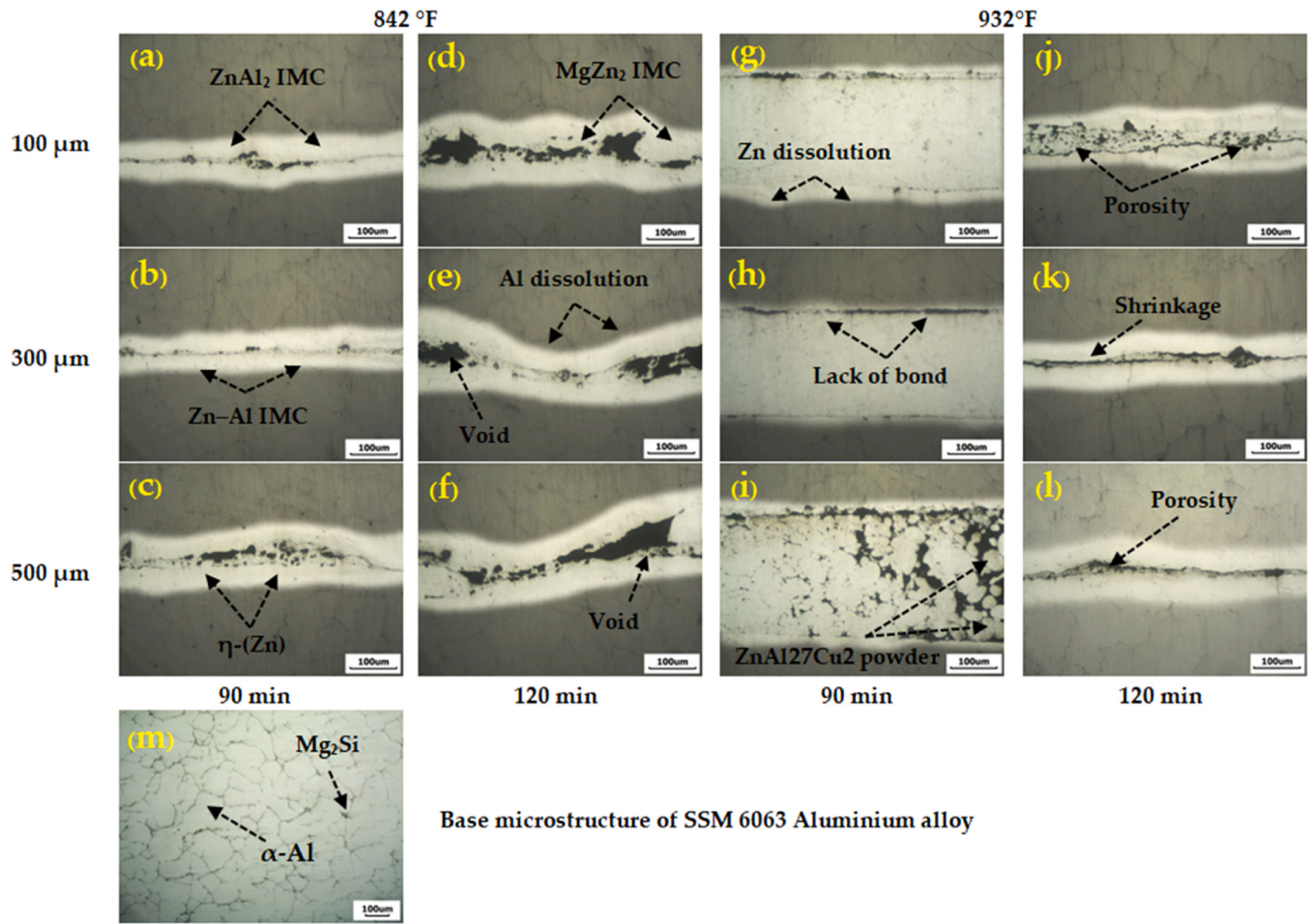


Fig. 8. Optical microstructures of SSM 6063 joints produced by TLP bonding with a ZnAl27Cu2 powder interlayer: (a–l) microstructures of the TLP-bonded joints under different bonding conditions and (m) base microstructure of the SSM 6063 aluminium alloy.

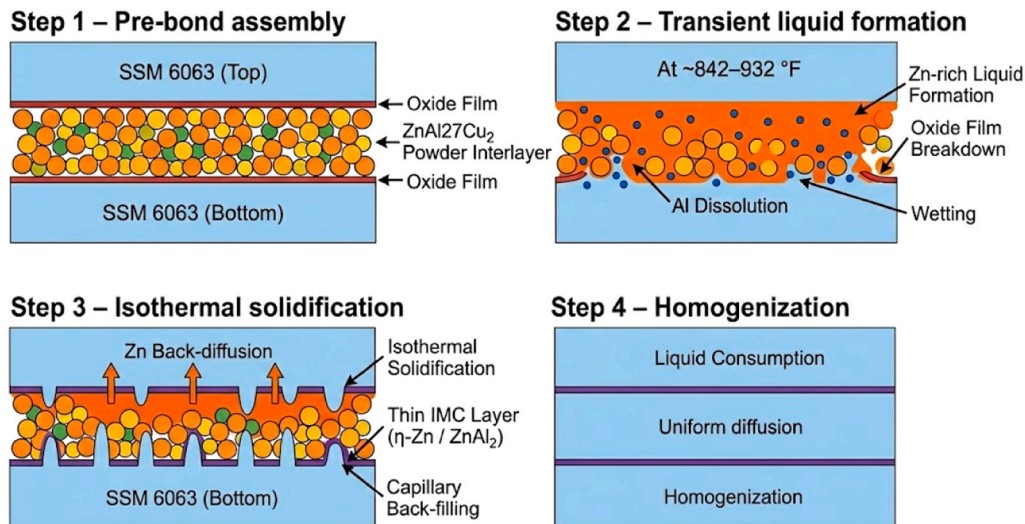


Fig. 9. Schematic of the TLP bonding mechanism for SSM 6063 using a ZnAl27Cu2 powder interlayer.

solidification via Zn back-diffusion, and subsequent homogenization, whose kinetics are jointly governed by bonding temperature (842/932 °F), bonding time (90/120 min), and particle size at 100, 300 and 500 μm. At 842 °F with 100 μm powder as illustrated in Fig. 8(a, d), the high specific-area particles melt early, generating a thin liquid that wets

and breaks the native Al₂O₃ but is prone to premature squeeze out, the seam shows intermittent porosity or underfill, and a relatively narrow thermal affected band. Increasing the bonding time to 120 min at the same temperature shown in Fig. 8(d) promotes capillary backfilling and smoother interfaces, consistent with extended diffusion time. With

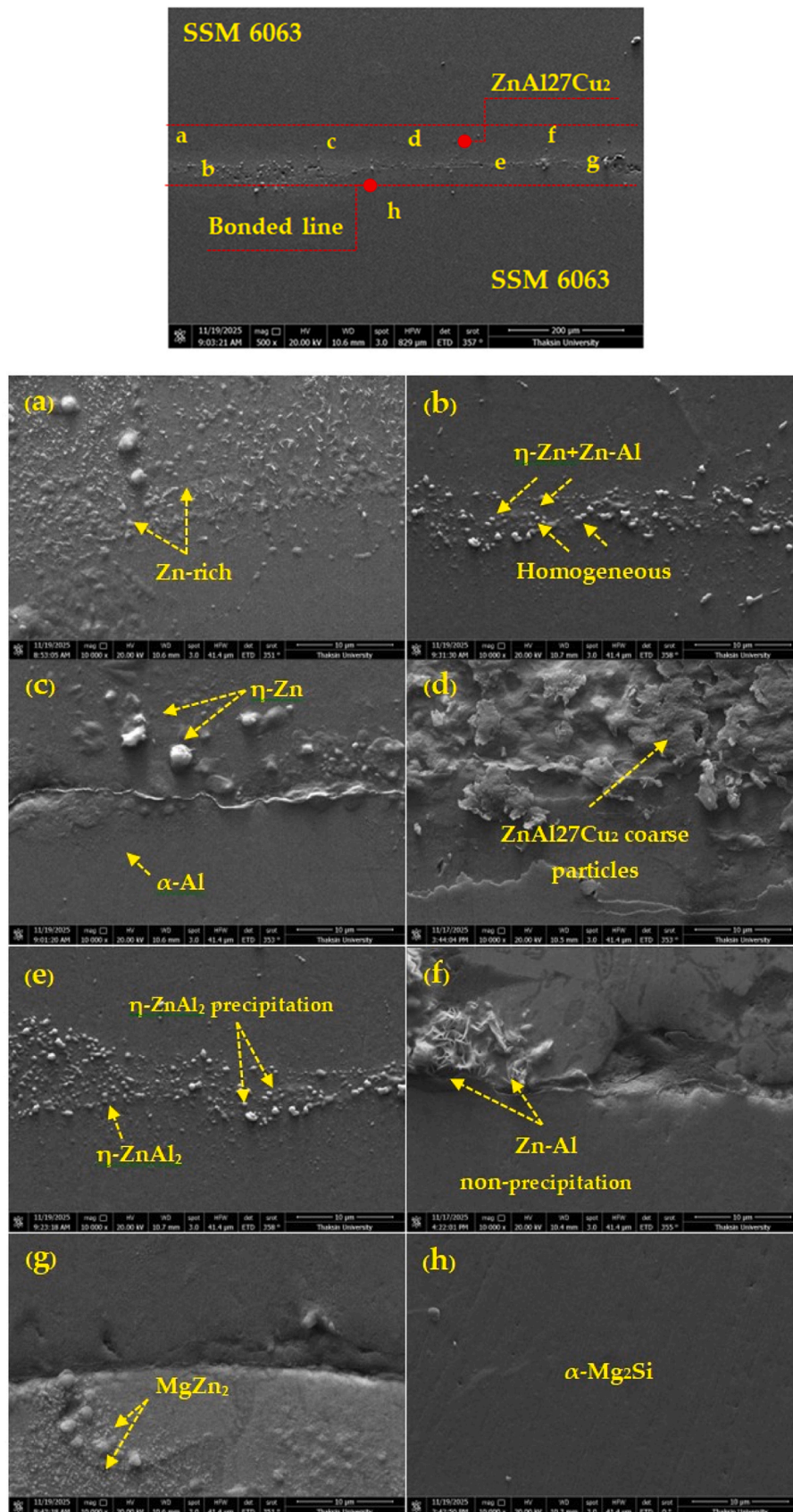
300 μm powder, as illustrated in Fig. 8(b, e), the liquid supply is more balanced: fewer flash beads, better neck filling, and a continuous seam, indicating improved coupling between dissolution and isothermal solidification [40]. Raising the temperature to 932 $^{\circ}\text{F}$, as illustrated in Fig. 8(g, h), enlarges the transient liquid fraction and accelerates oxide disruption, bond lines become straight and featureless at 90–120 min for 100–300 μm , but the wider thermal band implies greater softening and an elevated risk of η -(Zn) and ZnAl_2 IMC thickening if over-held. For 500 μm powder as illustrated in Fig. 8(c, f, i, l) thermal inertia delays core melting; at 842 $^{\circ}\text{F}$ islands of partially melted particles and voids persist as illustrated in Fig. 8(c, f), whereas at 932 $^{\circ}\text{F}/90$ min as illustrated in Fig. 8(i) excessive liquid coalesces into porosity clusters, extending to 120 min as illustrated in Fig. 8(l) restores continuity yet still leaves a thin IMC seam. These trends align with established TLP models showing a narrow process window for Al–Zn systems, moderate particle size (300 μm), lower temperature (842 $^{\circ}\text{F}$), and controlled dwell (90 and 120 min) to maximize capillary infiltration while limiting brittle Zn–Al IMC growth and thermal distortion [41]. During TLP bonding of SSM 6063, shrinkage arises primarily from solidification contraction of the Zn-rich transient liquid and partial dissolution–reprecipitation at the seam; if capillary backfilling is incomplete, this manifests as interdendritic voids or micro-cracks at the bond line. Optimizing the liquid supply and solidification kinetics as illustrated in Fig. 8(k). The as-processed SSM 6063 produced by GISS exhibits a typical semi-solid microstructure comprising globular or equiaxed primary α -Al grains embedded in a fine intergranular eutectic network (α -Al+Mg–Si constituents). The grains show pronounced spheroidization and broken dendrite arms, evidence of partial remelting and shear during slurry preparation, along with low porosity and clean grain boundaries, indicating effective degassing and a controlled solid fraction. The pale matrix corresponds to α -Al, while the thin, darker intergranular films mark eutectic regions that locally enrich in Mg/Si/Si and minor Fe-bearing particles; these phases act as reservoirs for Mg and Si that later participate in interfacial reactions (e.g., MgZn_2) during TLP bonding [25]. Such a globular SSM morphology is advantageous for joining because it reduces anisotropy and thermal distortion in the near-bonded line, provides smoother, more continuous contact surfaces that aid wetting [26], capillary infiltration of the $\text{ZnAl}_{12}\text{Cu}_2$ liquid, and limits grain-boundary cracking by distributing strains more uniformly during heating or cooling. Overall, confirms a stable, low-porosity, globular α -Al matrix appropriate as a baseline for interpreting the interfacial transformations observed as illustrated in Fig. 8(m).

During TLP bonding of SSM 6063 with a $\text{ZnAl}_{12}\text{Cu}_2$ powder interlayer, the sequence begins with a pre-bond assembly in which oxide-covered SSM 6063 coupons are separated by a loosely packed Zn–Al–Cu powder bed (Step 1). Upon heating to ~ 842 – 932 $^{\circ}\text{F}$, the Zn-rich constituents in the interlayer reach the Zn–Al eutectic and a transient liquid forms; this liquid wets the freshly disrupted Al-oxide, dissolves Al from both faying surfaces, and breaks up the native oxide film (Step 2). The amount and stability of this liquid depend strongly on both dwell time and particle size: finer powder (100 μm) generates liquid rapidly (high specific surface area), promoting early wetting but also squeeze-out or flash, coarser powder (500 μm) shows thermal inertia, delaying core melting and fostering local areas that are non-melted islands unless the dwell is extended. During isothermal holding, back-diffusion of Zn into the Al substrate raises the local liquidus, consuming the liquid and driving isothermal solidification from the Al sides toward the midline (Step 3). A thin, discontinuous intermetallic seam—dominated by η -(Zn) and ZnAl_2 —can form at the interface; its thickness increases with total heat input. Capillary backfilling closes interfacial pores when liquid supply and pressure are balanced; insufficient time or pressure leaves a lack of bond or voids, whereas excessive liquid generation (high bonding temperature, long bonding time, or very fine powder) increases flash and distortion. Finally, continued holding at the bonding temperature (or a brief post-bond homogenization) promotes solute redistribution, thinning of the transient IMC layer, and microstructural leveling across

the diffusion zone (Step 4). Overall, the observations align with canonical TLP kinetics, which are dissolution, widening, isothermal solidification, and homogenization, respectively [42–44].

In Fig. 10 (a–h), the sequence of interfacial phases observed along the TLP-bonded seam of SSM 6063 joined with a $\text{ZnAl}_{12}\text{Cu}_2$ powder interlayer can be rationalized by classic TLP kinetics: transient liquid formation at the Zn–Al eutectic, dissolution of Al into a Zn-rich melt, back-diffusion, controlled isothermal solidification, and subsequent homogenization. Phase assignments in this study are based on compositional evidence from spot EDX and EDX line scans, corroborated by the Al–Zn binary and Al–Zn–Mg ternary phase diagrams at the bonding temperatures used, and are consistent with XRD-confirmed phase identifications reported in prior TLP studies of Al–Zn systems [45,46]. Definitive crystallographic confirmation via XRD of the bond interface or EBSD phase mapping is identified as a priority for future work. Zn-rich regions consistent with η -(Zn) phase, as illustrated in Fig. 10(a), mark early wetting of the Al surface after oxide disruption; with increasing interfacial reaction, the melt becomes compositionally more uniform and the seam exhibits a fine η -Zn + Zn–Al type mixture consistent with partial back-diffusion, as illustrated in Fig. 10(b). Local α -Al reprecipitation adjacent to the seam, as illustrated in Fig. 10(c), indicates Al enrichment during back-diffusion. Coarse interlayer fragments, as illustrated in Fig. 10(d), reflect incomplete melting of 500 μm powder agglomerates at the selected bonding conditions, which locally impedes capillary flow [47]. Discrete η - ZnAl_2 -type precipitates decorating the interface, as illustrated in Fig. 10(e), are consistent with IMC thickening at high heat input. Filamentary non-precipitation debris, as illustrated in Fig. 10(f), corresponds to re-solidified or fragmented oxide surface films. In the Mg-bearing SSM 6063 substrate, Zn enrichment promotes MgZn_2 -type phase formation, as illustrated in Fig. 10(g), while the base-metal equilibrium α -Mg–Si is retained away from the seam, as illustrated in Fig. 10(h). EDX line scan profiles across the bond interface (Fig. 10, inset) reveal a Zn-enriched seam region (~ 5 – 10 μm wide) with gradual Zn depletion moving away from the seam consistent with back-diffusion, and localized Mg depletion adjacent to the seam consistent with MgZn_2 -type phase formation. These observations collectively imply a narrow process window: sufficient liquid and time are required for oxide break-up and capillary backfilling, but excessive bonding temperature and time coarsen IMC layers and leave unmelted interlayer islands, degrading toughness [48]. (Table 7)

The spot EDX taken on the mid-bonded line (inset) shows a Zn–Al reaction layer with Al ~ 41.9 wt% and Zn ~ 25.8 wt%, trace Cu ~ 0.34 wt%, minor O ~ 3.0 wt% and Na ~ 1.6 wt%, plus a high carbon signal (27.4 wt%) from the conductive tape/embedding resin (artefact). In atomic terms (Al ~ 34.6 at%, Zn ~ 8.8 at%), the spectrum indicates an Al-rich interface containing a Zn-enriched film, particles rather than a fully Zn-dominated layer; this is consistent with partial back-diffusion of Zn into the Al matrix during isothermal holding and the formation of a thin η -(Zn+Al) and/or η - ZnAl_2 reaction zone dispersed in α -Al. The very low Cu at the spot reflects both the modest Cu content of $\text{ZnAl}_{12}\text{Cu}_2$ and dilution into the Al substrate; Cu is known to partition into Zn–Al IMCs at low levels and can be below EDX detectability in localized analyses. The oxygen peak indicates residual alumina fragments at the seam, but its low magnitude suggests limited oxide entrapment, consistent with the relatively continuous macro-bond observed for this condition [49]. The absence of detectable Mg at this spot (no Mg peak) implies that Mg-bearing phases (e.g., MgZn_2) are not pervasive at the selected location, agreeing with reports that MgZn_2 is spatially localized and sensitive to local Zn activity and thermal history. Quantitatively, the Al–Zn ratio and the narrow reaction layer match prior TLP studies on Al–Zn systems, where sound joints exhibit thin Zn–Al intermetallics (η/δ) after partial Zn back-diffusion, while excessive Zn retention or thick IMCs at higher heat input degrade properties. The present composition, therefore, supports a balanced reaction layer expected to favor bond integrity at 842 $^{\circ}\text{F}/90$ min; further reduction of oxygen by longer Ar purge and careful surface activation would minimize the residual oxide signature.



(caption on next page)

Fig. 10. SEM overview and high-magnification features across the TLP-bonded SSM 6063/ZnAl27Cu₂ joint (scale bar: 200 μ m). The map (top) locates points a–h along the bonded line. Panels (a–h) show phases consistent with: (a) Zn-rich region, (b) homogeneous η -Zn/Zn–Al layer, (c) η -Zn on α -Al, (d) coarse ZnAl27Cu₂ remnants, (e) η -ZnAl₂-type precipitates, (f) re-solidified Zn–Al interfacial film, (g) MgZn₂-type compounds, and (h) α -Mg₂Si in the base metal. Phase assignments are based on EDX composition and thermodynamic consistency; crystallographic confirmation via XRD/EBSD is identified as future work. Scale bar in the overview image is 200 μ m (high contrast).

Table 7

Quantitative EDX results at the mid-bond line of specimen TLP-SSM-ZA-5.

Specimens no.	Element	Line Type	Apparent Concentration	k Ratio	wt%	wt% Sigma	Atomic %	Standard Label
TLP-SSM-ZA-5	C	K series	0.61	0.00605	27.43	0.80	50.84	C Vit
	O	K series	0.42	0.00143	2.99	0.21	4.16	SiO ₂
	Na	K series	0.41	0.00173	1.60	0.21	1.55	Albite
	Al	K series	9.94	0.07136	41.87	0.51	34.55	Al ₂ O ₃
	Cu	K series	0.10	0.00098	0.34	0.11	0.12	Cu
	Zn	K series	7.50	0.07497	25.78	0.38	8.78	Zn
	Total:				100		100	

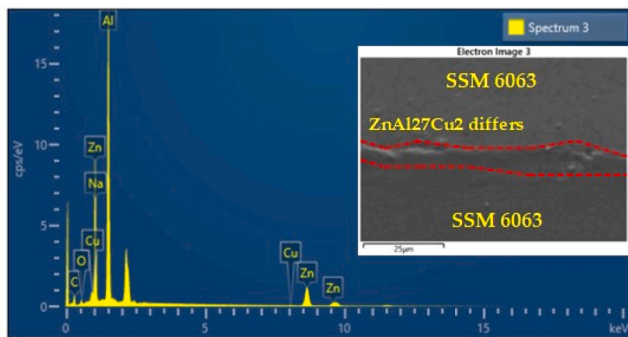


Fig. 11. Point EDX spectrum acquired at the mid-seam of specimen TLP-SSM-ZA-5 (300 μ m ZnAl27Cu₂ interlayer, 842 $^{\circ}$ F/90 min).

These observations are consistent reports on Zn-based interlayers in Al TLP bonding, which document Zn-rich liquids wetting and oxide disruption followed by Zn back-diffusion and isothermal solidification, the benefit of thin η/η' layers for strength [50], and the detrimental effect of thickened Zn–Al IMCs at higher bonding temperature and bonding time, as illustrated in Fig. 11.

The mechanistic relationship between bonding parameters and microstructural evolution in ZnAl27Cu₂-mediated TLP bonding of SSM 6063 proceeds through four sequential stages: Stage 1 — Heating (RT $\rightarrow$ $\sim$ 773 $^{\circ}$ F): Powder particles remain solid; minor native Al₂O₃ thickening occurs on the aluminum substrate surface. Stage 2 — Transient Liquid Formation ($\sim$ 773–842 $^{\circ}$ F): The Zn-rich eutectic melts first, generating a low-viscosity liquid that wets the Al surface and mechanically disrupts and fragments the Al₂O₃ film via viscous flow. Cu in the ZnAl27Cu₂ interlayer reduces surface tension and promotes spreading. Particle size critically governs this stage: 100 μ m particles melt rapidly (high specific surface area), generating early liquid prone to premature squeeze-out; 300 μ m particles provide controlled liquid release supporting complete capillary backfilling; 500 μ m particles exhibit thermal inertia, risking unmelted cores and incomplete wetting. Stage 3 — Isothermal Solidification (hold at 842/932 $^{\circ}$ F): Zn back-diffuses into the Al matrix; Al dissolves into the liquid, progressively raising the local liquidus; the interface resolidifies as α -Al with thin η -(Zn)/ZnAl₂ IMC layers. The rate is governed by Zn diffusivity in α -Al and follows a parabolic law: $d = k\sqrt{t} \cdot \exp(-Q/RT)$, where $Q \approx 59$ kJ/mol. At 842 $^{\circ}$ F (723 K)/90 min, estimated IMC thickness is $\sim$ 2–4 μ m (thin, discontinuous $\rightarrow$ sound bond); at 932 $^{\circ}$ F (773 K)/120 min, estimated IMC thickness rises to $\sim$ 8–15 μ m (thick, continuous $\rightarrow$ preferential brittle fracture path). Stage 4 —

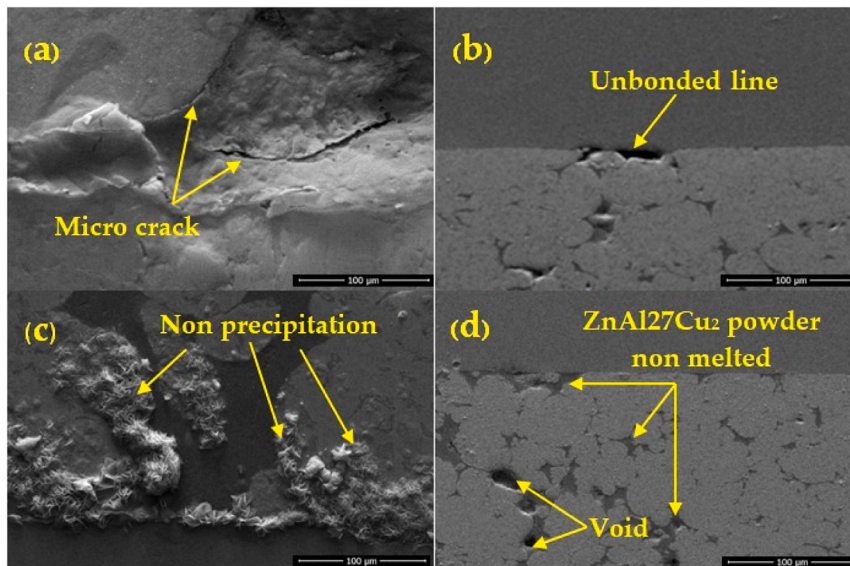


Fig. 12. Representative SEM micrographs of defect morphologies in TLP-bonded SSM 6063 joints produced using a ZnAl27Cu₂ powder interlayer: (a) micro-cracks, (b) interfacial unbonded line, (c) oxide debris (non-precipitation filamentary features), and (d) unmelted ZnAl27Cu₂ particles and associated voids.

Cooling and Homogenization: Residual Zn-rich pockets solidify as η -(Zn) or ZnAl_2 ; Mg_2Si re-precipitates in the heat-affected zone; thermal contraction generates residual stresses across IMC/Al boundaries, particularly where CTE mismatch between η - ZnAl_2 ($\sim 22 \times 10^{-6} / ^\circ\text{C}$) and α -Al ($\sim 23.6 \times 10^{-6} / ^\circ\text{C}$) drives stress concentrations at IMC–matrix interfaces. Particle size acts as a first-order design variable — on par with bonding temperature and time — for balancing liquid generation and minimizing continuous IMCs, flash, and porosity [51]. As shown in Fig. 12 (a–d), the following defect modes were identified and their mitigation strategies defined: (a) Micro-cracks originate from solidification shrinkage and CTE mismatch stresses during cooling when thick, continuous ZnAl_2 IMC layers form under high heat input. Mitigation: reduce bonding temperature/time, use $\sim 300 \mu\text{m}$ powder, and apply post-bond homogenization aging [52]. (b) Unbonded lines indicate incomplete disruption of the native Al_2O_3 or insufficient wetting time [53]. Mitigation: extend argon purge time, improve surface parallelism and pressure uniformity, and increase dwell time modestly at 842°F to complete capillary backfilling. (c) Oxide debris (non-precipitation filamentary features) results from fragmented oxide skins swept by the liquid and frozen in place when the liquid supply is exhausted. Mitigation: use $300 \mu\text{m}$ powder, maintain lower bonding temperature, and increase contact pressure slightly [54]. (d) Unmelted $\text{ZnAl}_{12}\text{Cu}_2$ particles and voids occur with $500 \mu\text{m}$ powder at 842°F due to thermal inertia leaving particle cores solid, blocking flow paths and trapping gas [55]. Mitigation: use tighter sieve fractions, raise temperature slightly, or use pre-compacted powder preforms [56].

Conclusions

The following key conclusions were drawn from this study on the TLP bonding of SSM 6063 aluminum alloy using $\text{ZnAl}_{12}\text{Cu}_2$ powder as an interlayer material:

- (1) Transient liquid phase bonding of SSM 6063 aluminum alloy using $\text{ZnAl}_{12}\text{Cu}_2$ powder interlayers is strongly governed by bonding temperature, bonding time, and powder particle size, which jointly control liquid formation, oxide disruption, capillary backfilling, and Zn–Al IMC growth.
- (2) The highest experimentally measured bonding strength, 22.31 MPa, was achieved at 842°F for 90 min with a $300 \mu\text{m}$ $\text{ZnAl}_{12}\text{Cu}_2$ interlayer, while the lowest strength, 10.72 MPa, occurred at 932°F for 120 min with $500 \mu\text{m}$ powder, demonstrating that excessive heat input and coarse particles significantly degrade joint performance.
- (3) Vickers microhardness traverses revealed a central hardness trough at the bond line (as low as ~ 55 HV for $100 \mu\text{m}$ powder at $842^\circ\text{F}/90$ min), which became shallower and more symmetric with increased temperature, dwell time, and particle size, reaching ~ 80 – 85 HV at the seam and ~ 105 – 120 HV in the adjacent base metal under the highest heat-input conditions.
- (4) Quantitative validation of the RSM models showed mean absolute percentage prediction errors of 8.31% for bonding strength and 0.56% for microhardness, with corresponding maximum errors of 17.96% and 0.86%, respectively. The microhardness predictions showed closer agreement with the experimental results than the bonding-strength predictions, supporting the use of the RSM response surfaces for subsequent GA-based optimization. The GA identified a multi-response optimum at $236.80 \mu\text{m}$, 90 min, and 842°F , with predicted bonding strength and microhardness of 18.97 MPa and 91.43 HV, respectively.
- (5) Microstructural analysis confirmed the canonical TLP sequence: Zn-rich transient liquid formation near the Zn–Al eutectic, oxide break-up and wetting, Al dissolution, Zn back-diffusion, isothermal solidification, and partial homogenization, with thin or discontinuous η -(Zn)/ ZnAl_2 seams associated with sound joints and excessive bonding temperature/time promoting coarse IMCs,

micro-cracks, unbonded lines, oxidic debris, and voids around unmelted Zn-rich islands.

- (6) Overall, an intermediate particle size ($\sim 300 \mu\text{m}$), lower bonding temperature (842°F), and shorter bonding time (90 min) form a practical process window that maximizes capillary infiltration and limits IMC thickening, while RSM–GA refinement around this region ($\approx 237 \mu\text{m}/842^\circ\text{F}/90$ min) offers quantitatively guided parameter selection for producing defect-lean, mechanically robust SSM 6063 joints suitable for lightweight structural applications. From an industrial perspective, the identified process window is compatible with existing batch furnace infrastructure and argon-purged brazing systems widely deployed in automotive and aerospace joining operations. The $\text{ZnAl}_{12}\text{Cu}_2$ interlayer material is commercially available as Zamak 2 (ZA27) with established supply chains. Key scale-up considerations include: (i) fixture design to maintain uniform contact pressure across larger bond areas; (ii) atmosphere control (dew point $< -40^\circ\text{C}$ Ar) to replicate laboratory oxide suppression at production scale; and (iii) powder pre-forming via tape casting or cold compaction for improved handling and green density uniformity in high-throughput settings. The RSM–GA optimization framework is transferable and can be recalibrated for different Al-6xxx variants or interlayer compositions with a minimal experimental dataset. Future work should include shear strength characterization (e.g., lap shear per AWS C3.2 or ASTM D1002) and fatigue testing under cyclic loading to fully assess joint suitability for transportation structural applications.

CRedit authorship contribution statement

Yongyuth Dunyakul: Writing – original draft, Investigation, Formal analysis, Data curation. **Dech Maunkhaw:** Validation, Methodology. **Yodprem Pookamnerd:** Software, Formal analysis. **Kittima Sillapasa:** Resources, Investigation. **Prapas Muangjunburee:** Supervision, Project administration. **Theeramate Songthong:** Writing – review & editing, Validation. **Chaiyoot Meengam:** Writing – review & editing, Visualization, Supervision, Methodology, Funding acquisition, Conceptualization.

Declaration of Competing Interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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