

Research paper

Interface evolution during refill friction stir spot welding and post-weld heat treatment of AA6013/Ti6Al4V with varying Mg and Si content

Lasse Malaske^{a,b,*}, Ting Chen^b, André M. Neves^{c,d}, Elizabeth Mathew^b, Emad Maawad^e, Henry Ovri^c, Uceu Suhuddin^b, Benjamin Klusemann^{a,b}

^a Institute for Production Technology and Systems, Leuphana University Lüneburg, Universitätsallee 1 21335, Lüneburg, Germany

^b Solid State Materials Processing, Institute of Material and Process Design, Helmholtz-Zentrum Hereon, Max-Planck-Straße 1, 21502, Geesthacht, Germany

^c Institute for Hydrogen Technology, Helmholtz-Zentrum Hereon, Max-Planck-Straße 1 21502, Geesthacht, Germany

^d Professorship for Applied Materials Engineering, Helmut Schmidt University, University of the Federal Armed Forces Hamburg, Holstenhofweg 85 22043, Hamburg, Germany

^e Institute of Materials Physics, Helmholtz-Zentrum Hereon 21502, Geesthacht, Germany

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ABSTRACT

The welding of dissimilar material combinations such as aluminum and titanium alloys is challenging due to their distinct mechanical and thermal properties. In particular, the formation of brittle intermetallic compounds can negatively affect the mechanical properties. Refill friction stir spot welding is a solid-state joining method that can reduce the thermal influence compared to fusion welding. This study analyzes the effect of systematic Mg and Si additions in AA6013 on the intermetallic compound formation for Al-Mg-Si / Ti6Al4V welds after processing and artificial aging heat treatments. The thermal cycles during the refill friction stir spot welding of five Al-Mg-Si variations are analyzed. Microscopic investigations showed fragments up to discontinuous layers of intermetallic compounds (IMCs) after welding and highly columnar Ti(Al,Si)_3 and $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$ IMCs after the post-weld heat treatment, depending on the specific alloy composition. $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$ is formed as mixed structure with Ti(Al,Si)_3 or as separate layer between the Ti(Al,Si)_3 and the aluminum, depending on the amount of solute Mg which did not form Mg_2Si . The IMCs thickness after heat treatment showed significant differences depending on the Mg and Si amount and element accumulations were found in the interfaces. Thermodynamic calculations revealed the phases during heat treatment and proved a liquid phase for alloys containing high amount of solute Mg. This liquid phase solidified as $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$ after the heat treatment, showing a fragile characteristic, including cracks. To minimize the liquid and therefore $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$ formation, a limited amount of Mg in solid solution must be respected by an appropriate ratio of Mg to Si.

1. Introduction

The increasing demand for lightweight and high-performance structures in the transportation and aerospace industries has led to the widespread application of aluminum (Al) and titanium (Ti) alloys. These materials are favored for their excellent specific strength, with aluminum offering a cost-effective solution and titanium providing exceptional corrosion resistance and mechanical performance at elevated temperatures [1]. However, the welding of Al to Ti remains a significant challenge due to their distinct physical and chemical properties. Traditional fusion welding methods often result for this material combination in poor mechanical properties due to the formation of

brittle intermetallic compounds (IMCs), high residual stresses, and porosity [1–3]. In particular differences in the coefficients of thermal expansion present challenges regarding residual stress formations around the interface [4].

To overcome these limitations, solid-state welding techniques have emerged as a promising solution for joining dissimilar Al/Ti combinations. Since the process temperature in solid-state welding remains below the melting point of the base materials, it significantly reduces the formation of detrimental IMCs and avoids the problems associated with liquid-phase reactions [1]. Besides, processing times for these techniques are typically low, reducing even further the time for the formation of IMCs [5]. Techniques such as friction stir welding (FSW) [6,7],

* Corresponding author.

E-mail address: lasse.malaske@leuphana.de (L. Malaske).

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friction stir lap welding [8] and refill friction stir spot welding (refill FSSW) have been successfully applied to create sound Al/Ti joints [9, 10]. The mechanical properties of as-welded joints using the material combinations of the present study, varying the parameters rotational speed (RS) and dwell time (DT), were previously published [9]. These methods promote dynamic recrystallization and the formation of fine-grained microstructures, which enhance atomic diffusion and bonding at the interface without inducing excessive heat [11]. Refill FSSW was shown to be a suitable joining technique to weld Ti to 5xxx and 6xxx aluminum series, which are mainly based on Mg and Si alloying elements [5,9]. For a comprehensive overview about the different processing techniques for Al/Ti joints, the interested reader is referred to the review by Grassel et al. [12].

Despite these advantages, the formation of IMCs at the Al/Ti interface during and after welding remains a key factor, influencing joint strength and long-term reliability. The most commonly reported IMC is TiAl_3 , which forms readily due to its high thermodynamic stability [2, 13]. However, other phases such as TiAl , TiAl_2 , and Ti_3Al may also form under certain conditions [14]. Post-weld heat treatment (HT) is often used to investigate IMC growth kinetics and their effect on joint properties. It has been reported that HT at temperatures between 400 °C and the solidus temperature (T_{sol}) of the aluminum alloy can influence both the thickness and morphology of the interfacial layers [15], and in some cases, improve joint strength [16]. The thickness of the IMC layer increases at high HT temperatures and typically deteriorates the joint integrity [14,15]. Additionally, an unnecessary high heat-input during welding would increase the size of the heat affected zone, diminishing the mechanical properties of the weld [17]. Low temperature HT, as in this case 120 °C, can increase the mechanical properties due to re-aging effects on the Al heat affected zone [15].

In addition to thermal effects, alloying elements, such as magnesium (Mg) and silicon (Si), commonly found in Al 5xxx and 6xxx series alloys, have shown significant influence on the formation and evolution of IMCs [15]. The enrichment of the interface with Mg, and/or Si from the Al side and V from Ti side was reported to decelerate the growth kinetics of IMCs due to the consumption of enthalpy energy and a reduced amount of crystal defects [18]. Due to similar atomic radius, Si can substitute for Al in TiAl_3 , forming the compound $\text{Ti}(\text{Al},\text{Si})_3$, and even leading to the formation of Ti_5Si_3 under certain conditions [18,19]. It was found that up to 15 at. % Al within the TiAl_3 can be substituted by Si [20]. Mg has also been implicated in altering diffusion behavior and, in some cases, forming distinct phases like $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$ [21], while in other studies, no local accumulation was found [22]. While the individual effects of Mg and Si on IMC growth have been studied, the combined influence of these two elements on the welding process and the IMC development has not been comprehensively understood.

Therefore, this study focuses on welding Ti6Al4V to five custom-cast and extruded Al-Mg-Si aluminum alloys, based on AA6013, with systematically varied Mg and Si contents using refill FSSW. The custom materials were chosen over standard industrial alloys, to keep additional element amounts besides Mg and Si constant to rule out secondary side effects. The objective is to evaluate how these alloying elements influence the thermal profile during welding and the growth of IMC during post-weld heat treatment. Understanding the synergistic effects of Mg and Si on IMC formation will provide critical insights for optimizing Al/Ti welding processes and enhancing joint performance.

2. Materials and methods

2.1. Base materials

For this study the titanium alloy Ti6Al4V (ASTM Grade 5), an alpha-beta titanium alloy, in the following called Ti64, was used with the sheet dimensions of $100 \times 25.4 \times 2 \text{ mm}^3$. The Ti64 was annealed under air at 790 °C for 62 minutes in a pre-heated oven and left to cool down in air. The aluminum alloys were custom cast and extruded, based on the alloy

AA6013 with systematic variations in the Mg and Si content. The compositions of all five alloys were obtained via optical emission spectroscopy, and the average of five measurements are shown in Table 1. Standard deviation of the measurements was calculated as 0.09 % for Si and 0.11 % for Mg. The composition of the alloy Al1Mg1Si fulfills the requirements for the chemical composition of AA6013 [23], so this alloy is considered as reference case during this study.¹ All five alloys were cast via gravity die casting at 700 °C and immediately water quenched afterwards to maintain the homogenized state. The casting skin was lathed off and the billets with a diameter of 100 mm were conventionally extruded using a flat profile matrix. For this purpose, the billets were pre-heated to 450 °C and extruded at a speed of 1 mm/s to the final sheet thickness of 1.8 mm at TU Dortmund University. Since the Al5Mg1Si showed hot cracking during the extrusion process the speed of extrusion was reduced to 0.3 mm/s for this alloy. The resulting material was cut in strips of $100 \times 25.4 \times 1.8 \text{ mm}^3$.

2.2. Welding process

Refill Friction Stir Spot Welding (refill FSSW) is a friction-based welding technique developed at Helmholtz-Zentrum Hereon to join similar, but also dissimilar material combinations [24]. The sheets to be welded are placed in overlap configuration and exit hole-free welds are generated by the process. The tool made from tungsten carbide consists of three components: clamping ring (CR), shoulder and probe (Fig. 1). The CR as only stationary part does apply the clamping force and keeps the sheets in position, clamped between the CR and the steel backing. Shoulder and probe, on the other hand, rotate concentrically, causing friction between workpiece and tool as well as imposing shear forces, which plasticize the material. During the plunging phase, the shoulder is moved downwards into the material while the probe is moved upwards, creating a cavity within the tool. Plasticized material is pressed into the cavity. Subsequently, shoulder and probe retract and push the material back, finishing the weld. In dissimilar refill FSSW, it is usual that the shoulder only plunges within the upper sheet and does not reach the lower sheet material to avoid excessive heat-input and minimize wear to the shoulder. In most cases the plunge depth plays an important role, so that the distance between interface and shoulder tip is minimized to around 0.1 or 0.2 mm in Al/Ti welds [10,11]. If not stated differently, a constant parameter set of RS 2200 rpm, DT of 10 s, plunge depth (PD) of 1.7 mm, plunge/retraction time (PT) of 2 s and clamping force (CF) of 5 kN was used in this study. To show the effect of longer welding times, additional joints were welded using longer DT. Oxide layers of both aluminum and titanium sheets were carefully ground off by hand prior welding, using P1200 SiC abrasive paper and the sheets were afterwards cleaned with ethanol.

2.3. In-Process temperature measurement

During the welding procedure the temperature was measured using type K thermocouples with a diameter of 0.5 mm. The thermocouples were placed in the interface by pressing them into milled slots of $0.6 \times 0.6 \text{ mm}^2$ in the Ti64 sheet surface as schematically illustrated in Fig. 2a) and shown via a cross-section of the joint after welding, see Fig. 2b). Peak process temperatures were measured in the center of the spot weld and right below the shoulder, the area of highest plastic deformation rates.

2.4. Microstructural characterization

The melting characteristics of the Al alloys were evaluated using differential scanning calorimetry (DSC) to study the differences in T_{sol} .

¹ To be precise, Al1Mg0Si represents more a AA5xxx alloy. All other variations represent AA6xxx alloys.

Table 1

Chemical composition of the alloys measured by optical emission spectroscopy, based on AA6013; Al1Mg1Si fulfills the requirements for the chemical composition of AA6013. For the other four alloys, the amount of Mg and Si was systematically varied. All values are given in wt. %.

Alloy	Mg	Si	Cu	Fe	Mn	Zn	Cr	Ti	Al
Al1Mg1Si	1.10	0.98	0.6 – 1.1	Max. 0.5	0.2 – 0.8	Max. 0.25	Max. 0.1	Max. 0.1	Bal.
Al1Mg0Si	1.06	0.01							Bal.
Al1Mg1.5Si	1.03	1.52							Bal.
Al3Mg1Si	3.13	1.10							Bal.
Al5Mg1Si	5.30	0.78							Bal.

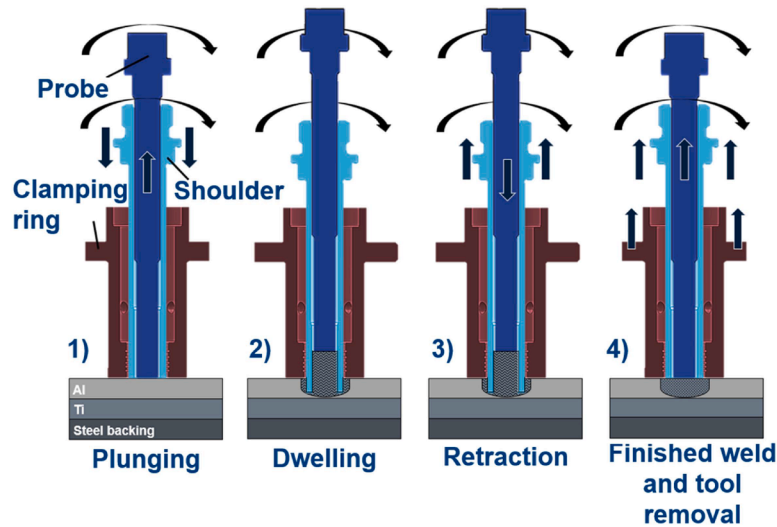


Fig. 1. Schematic representation of the refill Friction Stir Spot Welding (refill FSSW) process, showing the tool assembly and process stages.

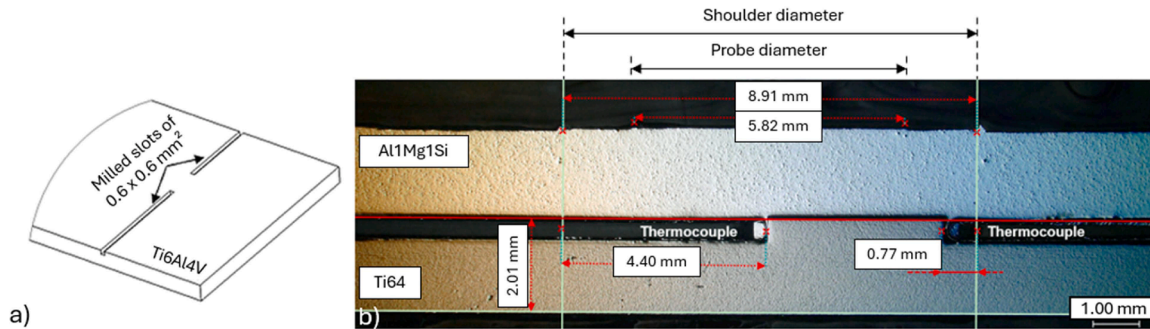


Fig. 2. Positioning of the thermocouples during the welding process; a) Schematic visualization of the Ti6Al4V sheet presenting the milled slots on the surface; b) Illustration of thermocouple position within the welded joint.

as a function of the chemical composition. 10 mg samples were placed in alumina crucibles and subjected to two heating cycles under Argon flow rate using a Netsch DSC 200 F3 Maia® equipment, starting from room temperature until 600 °C. Heating and cooling rates were set to 10 K/min and the first heating cycle was utilized to eliminate the thermo-mechanical history of the extrusion process. The second heating cycle was then analyzed to compare the alloys.

Samples for energy dispersive X-ray spectroscopy (EDS) and electron backscattered diffraction (EBSD) measurement were prepared by standard grinding and polishing procedure, with a final step of chemical-mechanical polishing step using a (100/20/5 ml) mixture of colloidal silica (OP-S), hydrogen peroxide and ammonia. Hydrogen peroxide helps to minimize the effect of mechanical deformation of the Ti generated by a purely mechanical polishing procedure [25] and potentially also the IMCs in the interface. Optical micrographs were collected using a Keyence VHX6000 series. For the electron microscopy, a FEI Quanta 650 scanning electron microscope (SEM) equipped with a EDAX

scanning transmission electron microscopy (STEM) detector, an EDAX Apollo X energy-dispersive X-ray spectrometer detector and an EDAX EBSD detector was employed. The EBSD investigation was carried out using a step size of 0.05 µm and the data was analyzed using the TSL OIM 7.3 software.

Conventional X-ray diffraction (XRD) and high-energy (HE-XRD) were performed to identify the IMCs in the joints' interface. Conventional XRD was performed with a Bruker D8 Discover (Bruker, Germany) in the Bragg–Brentano geometry, equipped with a Cu K-α source (wavelength $\lambda = 0.154184$ nm) operating at 50 kV and 1000 mA and a VANTEC-500 2-D detector for data acquisition. Diffraction patterns were collected in the 2θ range of 15° to 95° in steps of 10° with 480 s of data collection, without rotation. For the alignment of the incident beam, a laser and an in-built digital camera were used to illuminate the interface of the weld. The obtained 2-D patterns were processed using the software DIFFRAC.SUITE. High-energy XRD measurements were performed at the Hereon high-energy materials science beamline P07 at

PETRA III at Deutsches Elektronen-Synchrotron (DESY) [26]. The measurements were performed in transmission mode with a monochromatic X-ray beam with an energy of 87.1 keV. The data collection was performed with a PerkinElmer XRD detector. As calibrant, LaB_6 was used and the sample to detector distance was 1.557 m. After data collection, the scattering signal was analyzed using the Fit2D software [27]. The processed diffractograms peaks were analyzed individually with crystallographic cards from international centre for diffraction data (ICDD) database.

2.5. Post-Weld heat treatment

Post-weld HT was performed to artificially grow the IMCs after the welding procedure. The welded joints were wrapped in aluminum foil, embedded in sand and introduced in a pre-heated oven at a temperature of 525 °C for periods between 1 h and 192 h. The aluminum foil served to protect against contamination, while the surrounding sand slowed down the heating and cooling rates, see Fig. 3. Pre-trials at higher cooling rates, e.g. in air or quenching in water, led to noticeable audible signs of cracking and, in some cases, even delamination. The temperature of 525 °C was selected as approximate balance between a reasonable time frame for IMC growth and the specification to avoid exceeding the T_{sol} of the aluminum alloys.

2.6. Phase equilibrium calculation

A correlation between material thermodynamics, kinetics, chemistry, crystallography, microstructure, and processing is possible by computational thermodynamics modelling [28]. The phase fractions when varying the Mg and Si composition in the Al-alloy as a function of Ti content, are calculated to simulate a simplified description of the IMC formation in the joint interfaces during the post-weld HT. The initial situation of Al/Ti interface is set up and the formation of stable phases as more Ti is diffusing into the Al alloy is studied at 525 °C, which is the HT temperature. The thermodynamic modelling was done using Thermo-Calc software 2022a [29] and the thermodynamic database adapted from Ni-alloy databases TCNI10 [30].

2.7. Nanoindentation testing

Mechanical properties, such as hardness and elastic modulus, of the

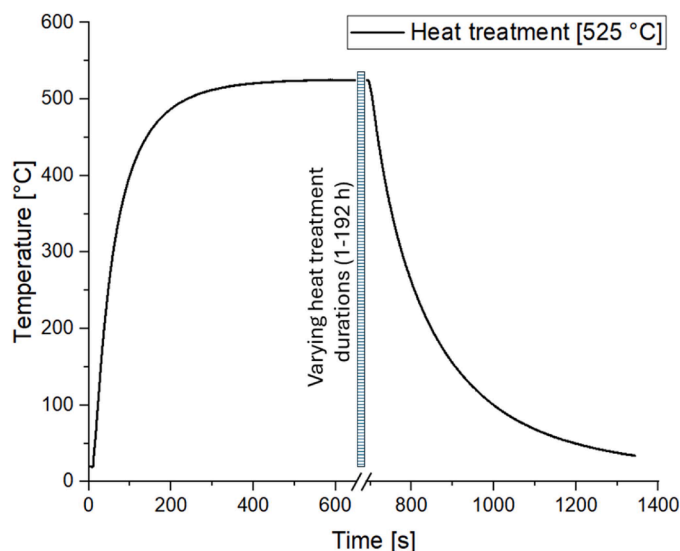


Fig. 3. Visualization of the heating and cooling curve of the post weld heat treatment at 525 °C, where the specific holding time varies between 1–192 h, depending on the specific heat treatment duration.

individual matrix and IMC layers in the Al1Mg1.5Si/Ti64 and the Al3Mg1Si/Ti64 joints HT for 72 h were assessed using nanoindentation method.

Nanoindentation, which is an extension of the conventional indentation methods to the sub-micron scale, is particularly well suited for this kind of investigation because of its nN (load) and sub nm (displacement) resolutions and its site-specific testing capabilities. Unlike in conventional indentation tests, the hardness can be determined with analytical models (Oliver-Pharr model) that obviate the need for a direct measurement of the contact area of the indents [31].

In this work, the nanoindentation experiments were performed with a Bruker Hysitron TI 980 Triboindenter equipped with a Berkovich indenter. The indents were made at room temperature at a strain rate of 0.1 s^{-1} to depths between 70 and 190 nm in dynamic mode. The dynamic mode allows for the continuous measurement of stiffness and consequently, hardness and modulus. The depth limit was dictated by the width of the individual phases. Precise identification and positioning of the indents in the matrix phases and in the different IMCs was achieved with the aid of the in-situ scanning probe microscopy (SPM) imaging option incorporated in the Triboindenter. The in-situ SPM option allows for high resolution imaging and nanometer-scale positioning accuracy.

The sample preparation resulted in a small but inevitable height step that arises from differing erosion rates between the Ti and the Al regions, resulting in a slightly tilted surface. As surface planarity is critical for accurate hardness measurements, the absolute values obtained are expected to deviate slightly and systematically from the true values. Moreover, as shown in the in-situ SPM image in Fig. 4, variations in the topography of the individual phases and the presence of $\text{Ti}(\text{Al},\text{Si})_3$ fragments within the $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$ phase in the Al3Mg1Si/Ti64 joint make it difficult to completely avoid overlap between adjacent phases. Consequently, the results should be regarded as approximations that provide a comparative indication of the hardness difference between the different phases in the materials.

3. Results and discussion

3.1. Joint formation and characteristics prior to heat treatment

Sound Al-Ti hybrid joints were successfully processed for all five compositions. Micrographs of the welded samples in the interfacial region show that in all cases the boundary layer between the materials are not homogeneous, see Fig. 5 for typical examples of these interfaces. The IMCs formed during the welding process and appeared as discontinuous

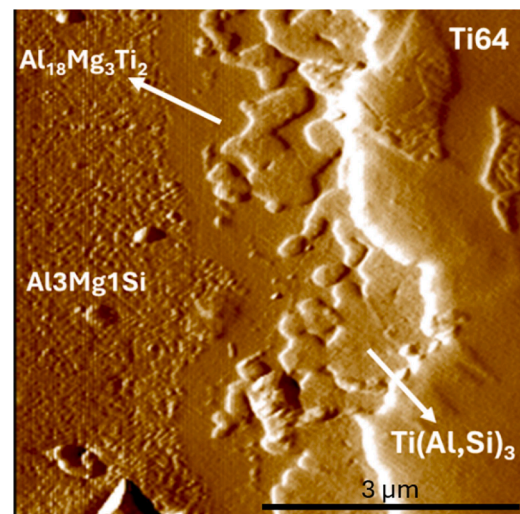


Fig. 4. In-situ SPM image showing the different phases in the Al3Mg1Si/Ti joint after 72 h HT. The image was acquired prior to indentation.

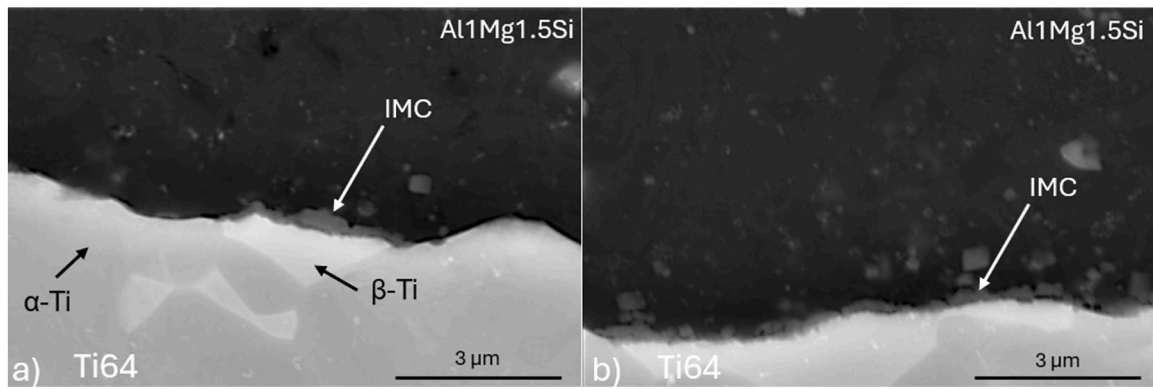


Fig. 5. SEM micrograph of an Al1Mg1.5Si/Ti64 joint interface processed with 2000 rpm RS, a DT of 10 s and a PD of 1.7 mm, highlighting the IMC formation in a) the shoulder region and b) the center of the spot weld.

fragments, i.e. the formation of a homogeneous layer could not be noticed. Fig. 5a) shows the region below the shoulder and Fig. 5b) presents the center of the weld for a comparatively long process of 10 s of DT, where regions with pronounced IMCs could be identified. For most of the other welding conditions, the interfaces are characterized by less pronounced presence of IMCs, so that especially in the shoulder stir zone hardly any IMC particles were observable. For Al1Mg1Si, Al1Mg0Si and Al1Mg1.5Si similar results were obtained, although compositions with higher quantities of Mg led to an even reduced formation of IMCs. For the Al5Mg1Si/Ti64 joints the IMCs were too small to detect at highest magnification. In general, there seemed to be a tendency for the IMCs to more frequently nucleate from the β -Ti phase, compared to the α -Ti phase, as indicated in Fig. 5a). While this tendency for higher nucleation rates on the β -interface was observed for the majority of the interfaces investigated, no systematic evaluation could be performed and the underlying reasons for this occurrence need further research, which is out of scope of the current manuscript.

A fully consistent IMC layer was not achieved for the studied materials at the chosen process parameters. Fig. 6 presents an extreme condition of 60 s DT using the Al1Mg0Si alloy. The IMC layer is predominantly continuous, but still thin in the range of 50 to 300 nm and cracks can be seen along the interface between IMC and the Al alloy. Based on this observation, it seems that the Ti64 surface topography structure plays a significant role. Elevated areas of the Ti surface tend to show reduced IMC layer thickness. As the position of this micrograph is in the center of the spot weld, the influence of material flow should be minimal.

Fig. 7 presents the peak process temperatures in the interface a) directly in the center of the weld and b) right in the middle between the inner and outer radius of the shoulder (3.75 mm offset to the center). The peak temperature during the process varies between the different Al alloys welded to Ti64, whereby the trend correlates with the T_{sol} of the Al alloys, simplified derived from the binary Al-Mg phase diagram. For the compositions studied, lower amount of alloying elements, like in the case Al1Mg0Si, presented the highest maximum temperature and the composition with the most alloying elements, Al5Mg1Si/Ti64, presented the lowest temperature during welding. The measured peak temperatures in the center are almost identical for Al1Mg1Si, Al1Mg1.5Si and Al3Mg1Si. In the region below the shoulder an even

stronger correlation between the amount of alloying elements and maximum temperatures could be observed showing the order of $1Mg0Si > 1Mg1Si > 1Mg1.5Si > 3Mg1Si > 5Mg1Si$.

This trend was also seen in the DSC measurements of the base materials presented in Fig. 8, showing that the order of endothermic peaks of partial melting correlates with the observed peak process temperatures. Although the maximum operation temperature of the equipment was limited to 600 °C and only two peaks were found, the onsets of the peaks corroborate the in-process temperature results. The order, starting with highest temperature, was found to be $Al1Mg0Si > Al1Mg1Si > Al1Mg1.5Si > Al3Mg1Si > Al5Mg1Si$. In summary, increases in both Mg and Si content in Al-Mg-Si aluminum alloys lead to a reduction of T_{sol} . This reduction of temperature with increasing amounts of alloying elements is likely associated with a downward slope of the solidus and liquidus temperatures (that approaches the eutectic composition) of the ternary Al-Mg-Si diagram. Considering that possible liquefaction during welding, as will be shown later absorbs a significant amount of heat and simultaneously reduces internal frictional heat generation, the processing temperature tends to stabilize toward a dynamic equilibrium. Consequently, higher Mg and Si contents result in a lower peak temperature during the welding process.

3.2. IMC formation and phase identification after HT

Fig. 9 presents an overview of interfaces for Al1Mg0Si/Ti64 and Al3Mg1Si/Ti64 after 96 h of HT. In both cases, two IMCs are clearly visible in the interface. Depending on the matrix from which they grow, different IMCs dominate, which will be specified later. For the Al1Mg0Si/Ti64 sample, however, a mixed region is visible where both phases are alternating next to each other. The grown IMCs are, in general, homogeneous over the interface. Especially in the Al1Mg0Si/Ti64 weld, only minor local differences can be observed, and these differences are mainly caused by the Ti rough surface characteristics, as the IMC/Al interface is comparatively flatter. In the case of the Al3Mg1Si/Ti64, a similar trend is observed. However, since the overall IMC thickness is lower and two IMC layers grow above each other in the interface, the flattening effect is less pronounced. Especially, the IMC layer on the Ti side is relatively thin, while at the same time, a rough transition to the IMC on the Al side is generated, resulting in an irregular

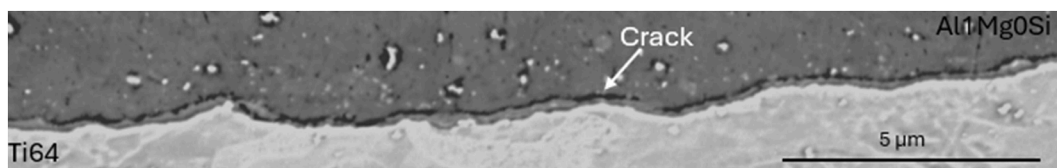


Fig. 6. SEM micrograph of an Al1Mg0Si/Ti64 spot weld, after 60 s of DT at 2000 rpm, showing the interfacial region within the center region of the spot.

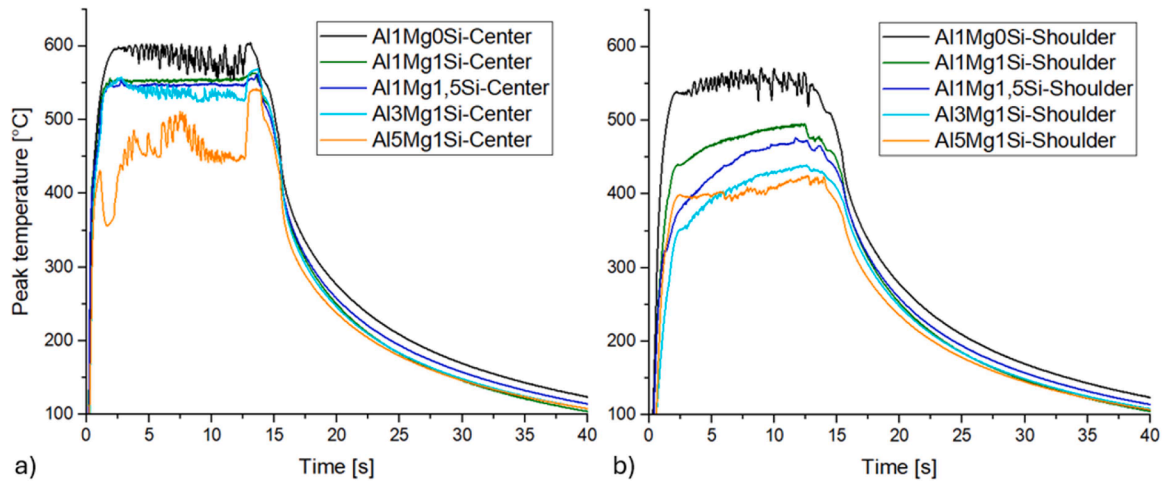


Fig. 7. Temperature evolution during the welding procedure, for measurement position see Fig. 2, a) directly in the center of the interface and b) in the interface, centered between the outer and the inner radius of the shoulder. The joints with higher alloyed aluminum alloys show lower peak temperatures.

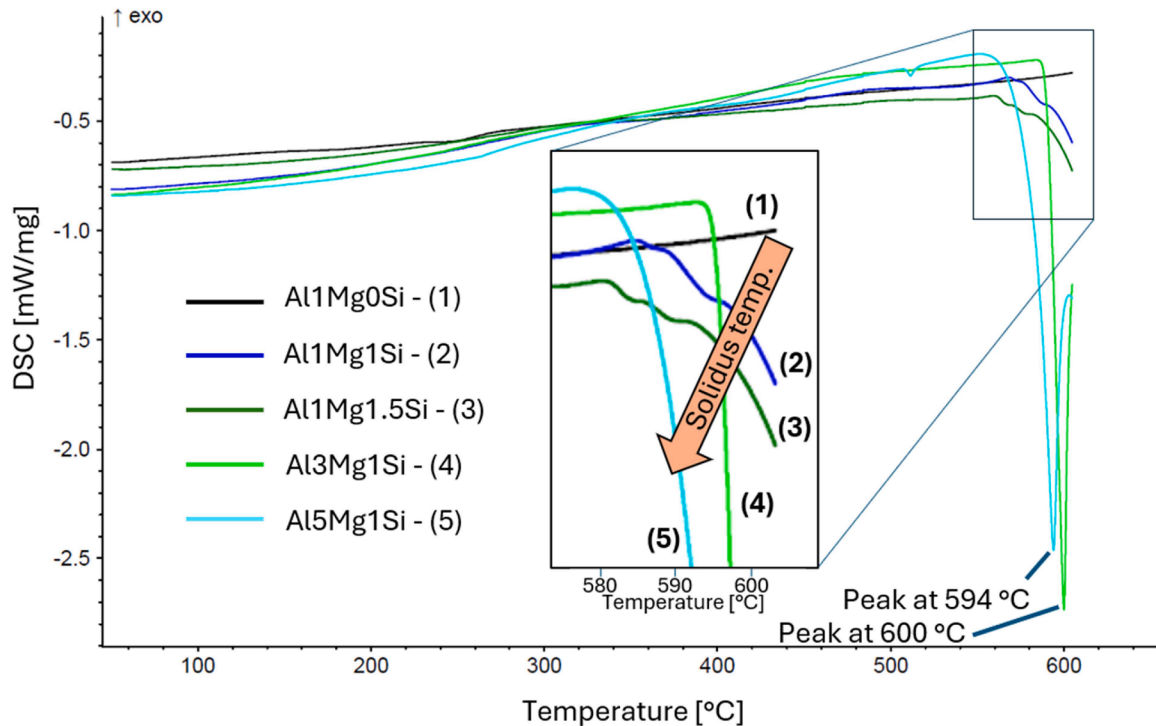


Fig. 8. DSC measurements of the five Al alloys, indicating the order of melting points depending on the chemical composition.

thickness. Locally, the IMC layer on the Ti side is still particularly thin, indicating the possibility for IMC discontinuities even after 96 h of HT.

This discontinuous growth characteristics of the IMC growing on the Ti side can be associated with an increase in the amount of Mg that can accumulate in the interface. Consequently, an even higher Mg amount is assumed to increase this effect and could give a potential explanation for the delamination of the Al5Mg1Si/Ti64 welds.

Several cracks were found running through the IMC layer close to the Al side, as highlighted in Fig. 10a). These cracks are associated with the brittle characteristics of the IMC layer and are expected to confer the joint poorer mechanical properties. In comparison, Fig. 10b) shows the interface of an Al1Mg1.5Si/Ti64 joint, displaying only one IMC layer with significantly lower amount of cracks, despite its much larger layer thickness. The Mg content is as high as in Al1Mg0Si alloy, presented in Fig. 9a), but no Mg accumulations are found in the interface. It can be

correlated that the additional 1.5 % of Si led to precipitation formation, which prevented the Mg from migrating into the interface.

The weld with the highest amount of Mg, Al5Mg1Si/Ti64, delaminated after the HT. The thick IMC visible in Fig. 11 is most likely the same IMC, as in the Al3Mg1Si/Ti64 interface, Fig. 10a). However, it has grown significantly bigger, up to an approximate thickness of 70 μm , partially attached to each of the sides. In these areas, it is possible to see the region where delamination occurred. On the inset figure, the small IMC layer is shown between the thick IMC layer and the Ti64 surface, similar to the Al3Mg1Si/Ti64 joint in Fig. 10a).

To identify the phases present in the IMCs layers formed in the different joints, EBSD analysis was employed. This technique enabled detailed crystallographic characterization, providing valuable insights into the phase composition and microstructural features of the IMC layers. Fig. 12a) presents an EBSD mapping of the Al1Mg1.5Si/Ti64

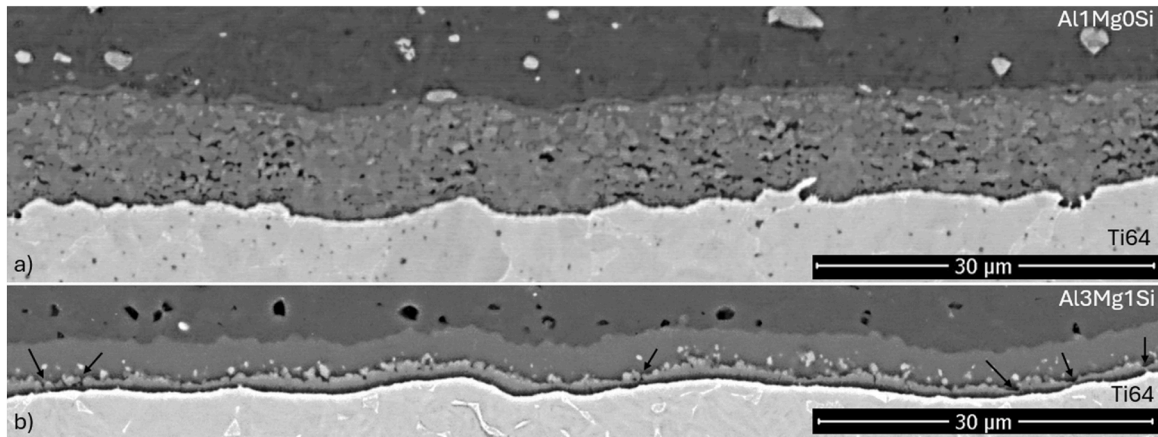


Fig. 9. SEM micrographs of the interfaces for a) Al1Mg0Si/Ti64 and b) Al3Mg1Si/Ti64 welds after 96 h of HT. Regions where the lower IMC layer is very thin are highlighted by black arrows in b).

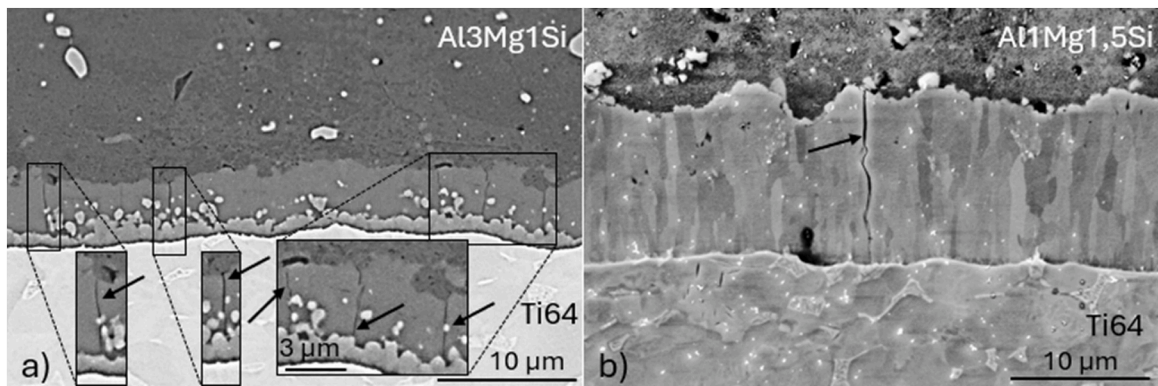


Fig. 10. SEM micrographs of a) the Al3Mg1Si/Ti64 weld and b) the Al1Mg1.5Si/Ti64 welds. The inset figures show details of the interface in higher detail. The arrows indicate fractures on the IMC layer.

interface after 120 h of HT. The diffraction patterns, shown in Fig. 12d) were used to successfully identify the IMC as TiAl_3 . The grains of the IMC growing from the Ti side have an elongated, columnar morphology, whereas near the Al side, the grains tend to an equiaxial morphology, as presented in Fig. 12c). These differences in grain structure can be attributed to the reduced diffusion rates with growing IMC layer thickness. A higher number of TiAl_3 grains near the Ti/ TiAl_3 interface has been reported in heat-treated AA6063/Ti64 samples from compression bonding tests [18]. α -Ti and β -Ti particles are visible in the Al part, which have not yet formed an intermetallic phase and TiAl_3 particles are present in both Al and Ti matrices, whereby the distribution on the Al side is denser. No α -Al was found within the Ti matrix. The quality of the data acquisition in the transition area from Ti to IMC was inferior and was not indexed in the mapping, so that the grain structure shown here might not be as fine as the mapping suggests. This reduction in quality is believed to originate from a slight step in the sample's height and an associated tilt angle issue.

To verify the grain morphology, a STEM sample of the Al1Mg1.5Si/Ti64 interface was produced by focused ion beam. The highly elongated and columnar grain structure in the IMC on the Ti side and rather equiaxed morphology on the Al side was also observed using STEM. Fig. 13 presents the grain structure at the Ti64/ $\text{Ti}(\text{Al},\text{Si})_3$ interface, within the center of the $\text{Ti}(\text{Al},\text{Si})_3$ and at the $\text{Ti}(\text{Al},\text{Si})_3/\text{Al1Mg1.5Si}$ interface.

The IMC layer close to the Al side could not be identified by EBSD, as the Kikuchi pattern was of insufficient quality. To aid phase identification, XRD scans with Cu $K\alpha$ radiation were performed. The results taken for the base materials were congruent with what was expected, but the

phases present in the interface in the welds, both for samples as processed and after heat treatment, could not be determined. Not only that the volume of these phases is comparatively small to the illuminated volume (making their signal very weak when compared to the background), but also, there was some difficulty in guaranteeing that the interfacial region was being consistently illuminated. While a baseline for the expected phases could be provided by these results (see Supplementary Material, Figures S1 to S9), a proper identification of the phases could only be performed with HE-XRD. Fig. 14 shows the results for the Al1Mg0Si/Ti64, Al1Mg1.5Si/Ti64 and Al3Mg1Si/Ti64 welds. The diffractograms of the data acquired in the interface and the data for regions with 30 μm distance from the interface in Al and Ti64 are shown as reference. The phase TiAl_3 was verified and the most prominent peaks that indicate its presence are highlighted. Some of the TiAl_3 peaks in Al1Mg1.5Si/Ti64 and Al3Mg1Si/Ti64, as e.g. at 2.32° , are shifted to slightly higher 2θ angles. This is assumed to either originate from the Si substitution or could be attributed to texture, as the IMC grows highly oriented and the Ti64 was produced by a rolling process. The second IMC was identified as $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$, whereby most representative peak positions are highlighted. Those can be seen in the interfaces of Al3Mg1Si/Ti64 and Al1Mg0Si/Ti64 but could not be confirmed for Al1Mg1.5Si/Ti64. This correlates with the finding of the formation of either one or two IMC layers, depending on the aluminum alloy as observed in the SEM investigations, Figs. 9 and 10.

To validate the findings and reveal the chemical composition of the IMC layers, including potential accumulation of local elements, EDS was used. Fig. 15 presents line scans over the interfaces for the Al1Mg0Si/Ti64, Al1Mg1.5Si/Ti64 and Al3Mg1Si/Ti64 welds. The Al1Mg0Si/Ti64

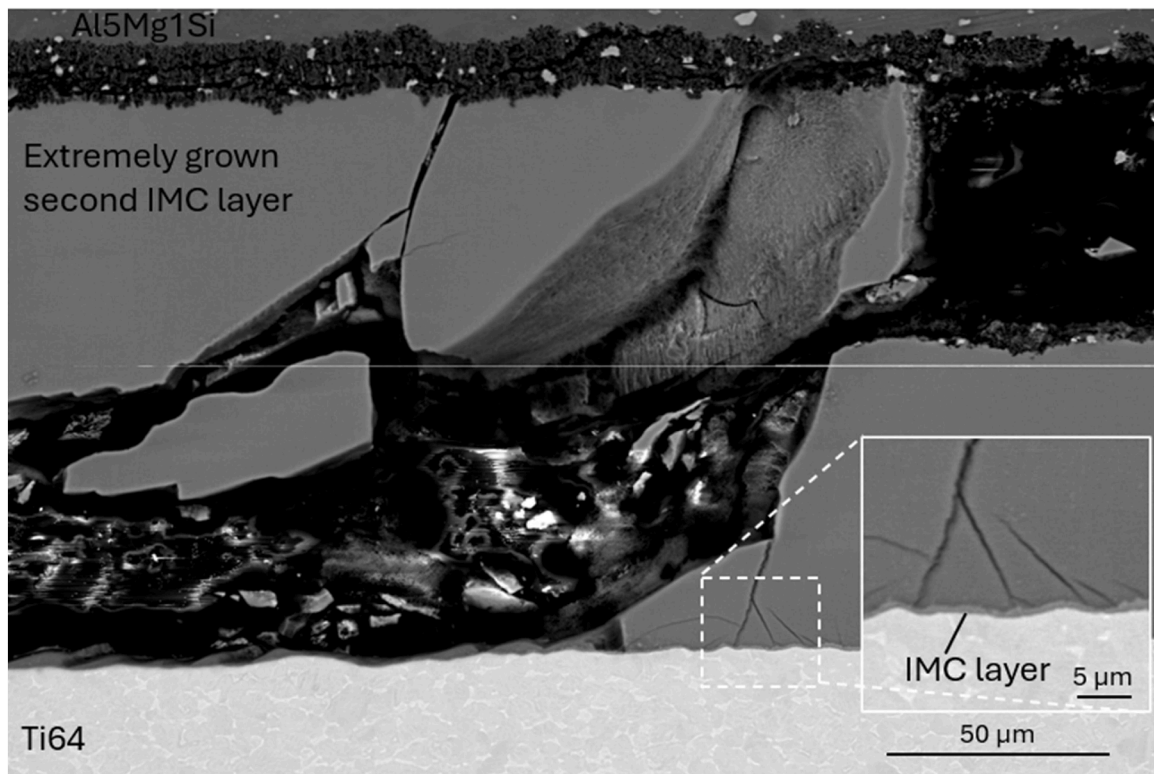


Fig. 11. SEM micrograph of a cross-section of the Al5Mg1Si/Ti64 weld, showing the formation of a thick IMC layer and delamination. Inset figure presents the interfacial region of the first IMC layer.

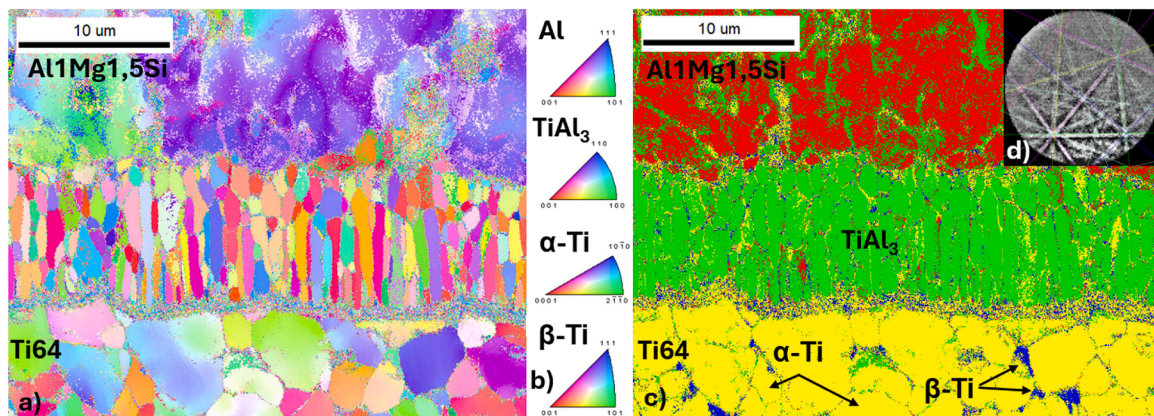


Fig. 12. EBSD mapping of the IMC in the interface of an Al1Mg1.5Si/Ti64 joint. a) Grain orientation map, b) inverse pole figures; c) phase distribution map; d) diffraction pattern of TiAl₃, showing excellent match with the simulated pattern.

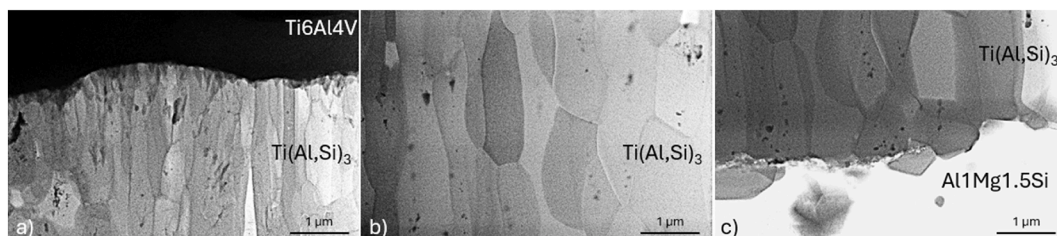


Fig. 13. STEM micrographs of the IMC; a) transition from Ti64 to IMC, b) center of IMC, c) transition from IMC to Al1Mg1.5Si.

interface shows a widely constant amount of ~ 76 % Al, ~ 21 % of Ti and a small amount of ~ 1 % V. The amount of Mg is highest between the Al

side and the mixed region. As both IMCs coexist, the presence of Mg, Al and Ti appears to be linked locally with the regions with reduced Mg

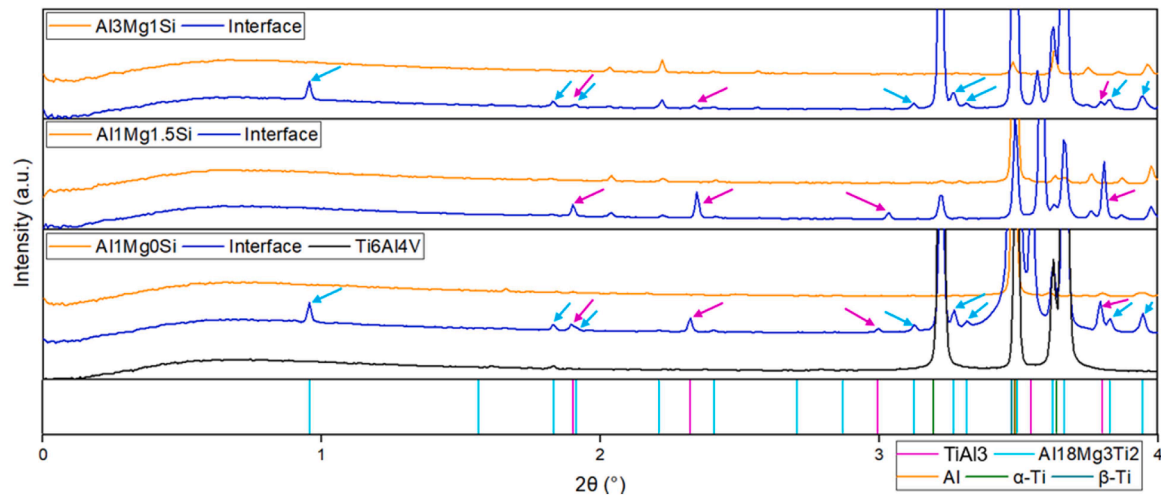


Fig. 14. XRD pattern of Al1Mg0Si/Ti64, Al1Mg1.5Si/Ti64 and Al3Mg1Si/Ti64 joint scans in the interface (blue lines) and the parent material (orange lines for the Al-alloys, and black for the Ti64) using HE-XRD at beamline P07 at PETRA III at Deutsches Elektronen-Synchrotron (DESY). The most prominent peaks of the phase TiAl_3 and $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$ are highlighted by magenta and blue arrows, respectively.

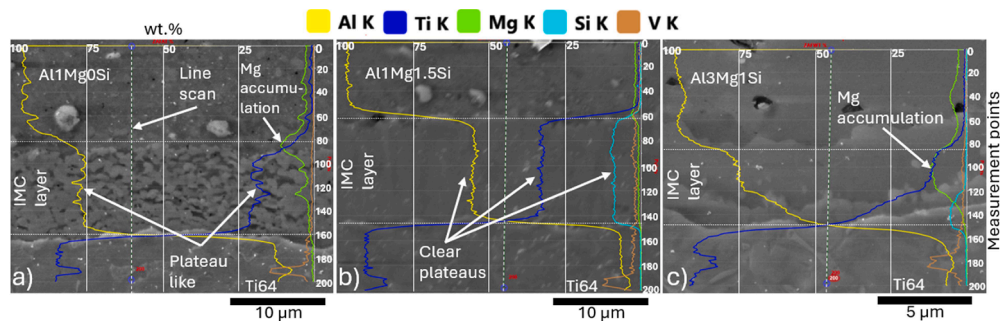


Fig. 15. EDS line scans through the interface of the welds, including Al1Mg0Si, Al1Mg1.5Si and Al3Mg1Si. All values in wt. %.

content in the direction of the Ti side. The Al1Mg1.5Si/Ti64 interface region is characterized by a different, but highly constant composition. Here, the amount of Al is reduced to $\sim 63\%$, Ti increased to $\sim 24\%$ and $\sim 13\%$ of Si were detected. Mg content was minimal in the IMC, only showing a small peak of $\sim 3\%$ at the IMC/Al transition and V was again $\sim 1\%$. At the Al3Mg1Si/Ti64 interface, it is visible that the compositions do not show clear plateaus. However, the trend is very similar to the Al1Mg0Si/Ti64 interface measurement, and the values support the finding of $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$. It is evident that the second IMC layer forming predominantly on the Al side is $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$. It develops either in a separate way as presented in the Al3Mg1Si/Ti64 interface, Fig. 9b), or in a mixed structure together with the TiAl_3 , as can be seen in the

Al1Mg0Si/Ti64 interface, Fig. 9a).

Fig. 16 presents Mg and Si mappings of the Al1Mg0Si/Ti64 and Al1Mg1.5Si/Ti64 interfaces to highlight specific regions of the interface. In the Al1Mg0Si/Ti64 interface, the clear trend in Mg content towards the IMC/Al transition is visible, including a dense region in the transition between the mixed region of $\text{TiAl}_3/\text{Al}_{18}\text{Mg}_3\text{Ti}_2$ and the Al, as seen in Fig. 16a). A similar effect, but significantly less pronounced, is shown in the Al1Mg1.5Si/Ti64 interface, Fig. 16b) although the Mg alloying content was identical. In terms of Si, Fig. 16c) illustrates a clear accumulation directly at the Ti-IMC transition. Besides this and a few small areas of slight increase in concentration, the Si distribution within the IMC is very homogenous. This suggests the substitution of Al by Si in the

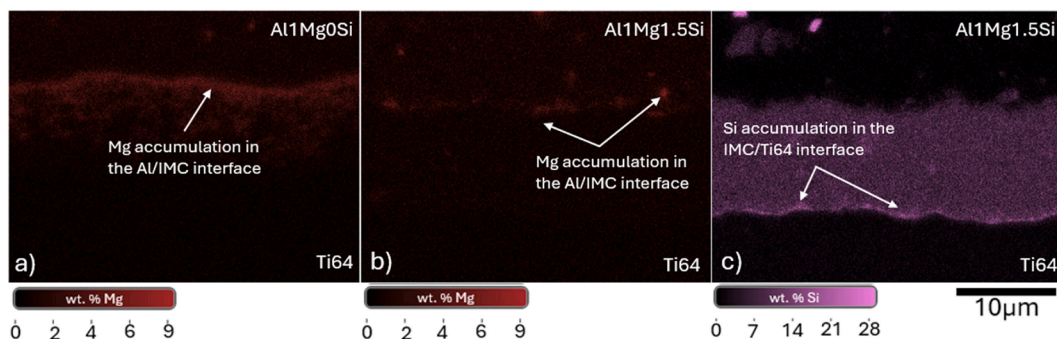


Fig. 16. EDS mappings of a) Al1Mg0Si for Mg, b) Al1Mg1.5Si for Mg and c) Al1Mg1.5Si for Si. The arrows indicate local accumulations of the alloying elements within the interfacial region.

TiAl₃ phase and the formation of Ti(Al,Si)₃.

The elemental mappings of Mg, Si and Ti for the Al₃Mg₁Si/Ti64 interface are presented in Fig. 17. In this sample the interface between the two IMCs can be identified by the rapid change in the content of Mg and Si, see dashed lines in Fig. 17a), b), c). On the Al side, disconnected from the IMC layer, a few Mg₂Si particles can be identified. In terms of Ti, the transition is not as clear as for Mg and Si, mainly caused by the strong contrast of the parent Ti64 and the fact that both IMCs contain Ti. Still, it is visible that the IMC on the Ti side contains more Ti compared to the IMC on the Al side.

3.3. Intermetallic compound growth characteristics

Fig. 18 presents a comparison between the center of the interfaces in the Al₁Mg₁Si/Ti64, Al₁Mg₀Si/Ti64, Al₁Mg_{1.5}Si/Ti64 and Al₃Mg₁Si joints at varying HT durations at 525 °C. In the Al₁Mg₁Si/Ti64 and Al₁Mg_{1.5}Si/Ti64 interfaces, only one IMC, the Ti(Al,Si)₃, is present for all durations. In the transition between this IMC and the aluminum matrix, several particles that are only partially connected to the layer are identified. The Al₁Mg₀Si/Ti64 and Al₃Mg₁Si/Ti64 interfaces show two different IMCs layers as described before, showing the clear differences in structure, especially for long HT durations. In this context, in the case of a continuous and sufficiently thick Al₁₈Mg₃Ti₂ layer, the TiAl₃ growth characteristics are significantly reduced, leading to overall thinner IMC layers (TiAl₃ + Al₁₈Mg₃Ti₂) in the interface. Also, the structure of Ti(Al,Si)₃ appears different, as the grains are clearly visible in Al₁Mg₁Si/Ti64 and Al₁Mg_{1.5}Si/Ti64, revealing a well-defined columnar grain structure. For Al₁Mg₀Si/Ti64 a rather globular TiAl₃ structure was found, whereby the Al₁₈Mg₃Ti₂ filled the volume in between the TiAl₃ areas. The ratio of TiAl₃ to Al₁₈Mg₃Ti₂ noticeably decreases from the Ti to the Al side, and a thin but dense Al₁₈Mg₃Ti₂ layer is present in between the mixed region and the Al side. In the Al₃Mg₁Si/Ti64 interface, a similar structure developed in the transition between both IMC layers. When comparing the IMC layer appearance of Al₃Mg₁Si/Ti64 after HT from 48 to 192 h, it can be clearly seen that the continuous layers grow simultaneously, while exhibiting the characteristic of a clear interface in between. Grain boundaries could not be observed in both IMCs.

Fig. 19 compares the IMCs layer thicknesses, showing that the IMC in Al₁Mg_{1.5}Si/Ti64 has the highest thickness of approximately 12 µm after 192 h of HT, evidencing ideal diffusion characteristics among the tested material combinations. The IMC layers in the Al₁Mg₀Si/Ti64 and Al₁Mg₁Si/Ti64 interfaces show slightly lower thicknesses of around 8 to 9 µm, followed by the Al₃Mg₁Si/Ti64 IMC of approximately 5 µm (~4 µm of Al₁₈Mg₃Ti₂ and ~1 µm of Ti(Al,Si)₃). The results obtained also suggest that the growth rate tends to slow down as the thickness increases. This can be explained by the increase in the mean path that has to be travelled by the diffusing species.

It seems that a clear correlation between alloying elements and growth kinetics was found, as higher amount of Si increased the

diffusion rates while increasing amount of Mg decreases the diffusion. However, while the layer thickness until 72 h of HT supports this finding, the results of Al₁Mg₁Si/Ti64 and Al₁Mg₀Si/Ti64 need further examination. The addition of Si might have a synergistic effect with Mg that not only changes the phases that are preferentially formed. Also, the dynamics of the formation of the IMC layers may be changed and could therefore influence the layer growth rate over time. However, since the IMC layer thickness in the Al₃Mg₁Si/Ti64 interface showed only 5.2 µm of both IMCs combined, clear evidence is found that the Al₁₈Mg₃Ti₂ formation suppresses the formation of Ti(Al,Si)₃. Similar results were found by Fuji et al. [21], who reported that mainly Al₁₈Mg₃Ti₂ formed in the interface of a Ti-Al4.6 %Mg joint. According to these authors, the higher tendency to the formation of Al₁₈Mg₃Ti₂, compared to TiAl₃, is ascribed to its higher thermodynamic stability, i.e. lower free energy [21].

As already shown in Fig. 9, within the spot area the IMC layers form homogeneously over a large width. Fig. 20 presents the transition from the outer area of the spot weld to the unbonded periphery. The thickness of the Ti(Al,Si)₃ increases almost instantaneously to a thickness above 5 µm, very close to the value measured also in the center of the spot weld (5.15 µm).

Although the majority of investigated samples (>50) have presented homogeneous IMC layer growth and morphology, Fig. 21 presents a case in which there is a discontinuous layer formation for a Al₁Mg₁Si/Ti64 joint was observed. Even after 48 h of HT, a significantly reduced Ti(Al,Si)₃ IMC at these positions was observed. A fundamental cause, such as contamination or delamination after welding, could not be found. However, a missing bond after welding over a range of a few µm in width, e.g., due to a scratch in the Al or Ti surface, is assumed to be the reason. This gives a good demonstration of the susceptibility of homogeneous IMC layer growth, even after 48 h of HT. Investigating the causes of such occurrence is important to avoid microstructural features that might negatively affect mechanical properties.

Fig. 22 highlights areas where the Ti(Al,Si)₃ grows on the V-stabilized β-Ti, indicating that the diffusion of the V into the IMC or the distribution within the Ti(Al,Si)₃ takes place more slowly compared to the Al and Ti. Especially in terms of reduced diffusion rates it is noticeable that typically the areas consisting of β-Ti are the peaks in the Ti surface structure after a given HT time. However, this stands in direct contrast to the finding that IMC nucleation is favored on the β-Ti during welding as indicated in Fig. 5. The fact that differences in the nucleation and growth behavior of IMCs were observed on the respective Ti phases implies that further research is needed to fully understand the mechanisms during welding of Al/Ti64.

3.4. Local mechanical properties

Fig. 23a) shows an in-situ SPM map with nanoindents located in the IMC at the joint interface as well as in the parent materials of the

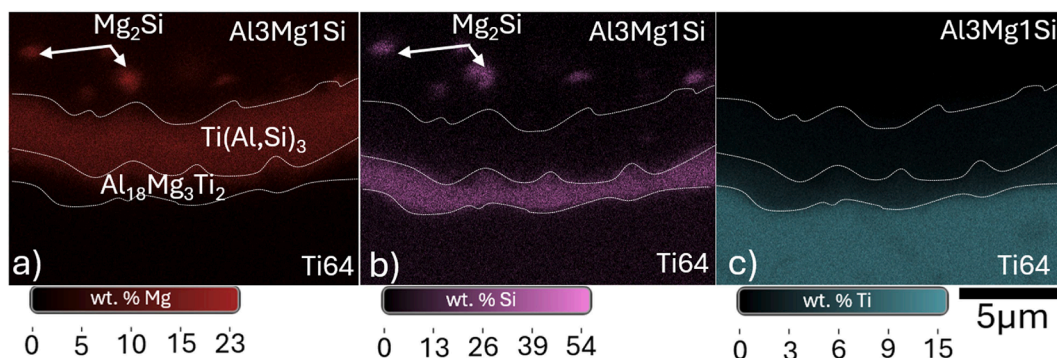


Fig. 17. EDS mappings for the Al₃Mg₁Si/Ti64 joint interface for the elements a) Mg, b) Si and c) Ti. The dashed lines indicate the Al₃Mg₁Si/Ti(Al,Si)₃, Ti(Al,Si)₃/Al₁₈Mg₃Ti₂ and Al₁₈Mg₃Ti₂/Ti64 transitions.

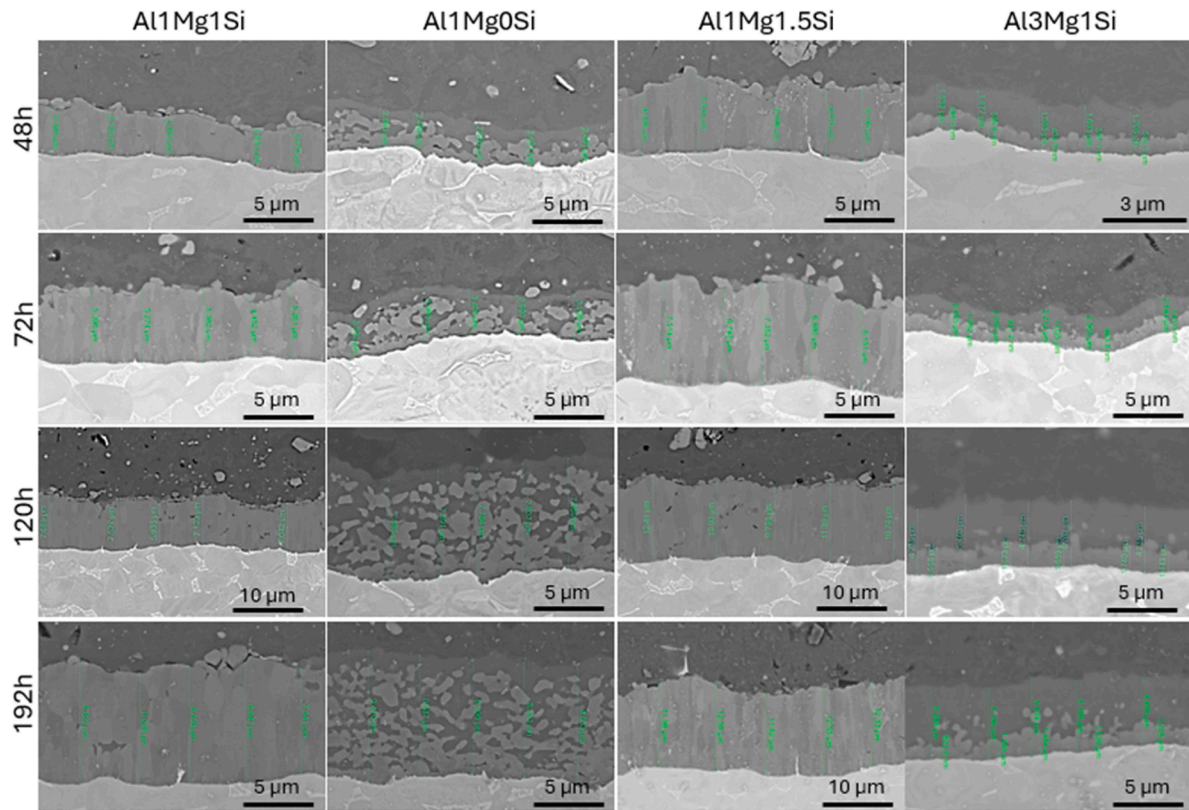


Fig. 18. SEM images of the interfacial regions of Al1Mg1Si/Ti, Al1Mg0Si/Ti, Al1Mg1.5Si/Ti and Al3Mg1Si/Ti after HT at 525 °C for varying HT times.

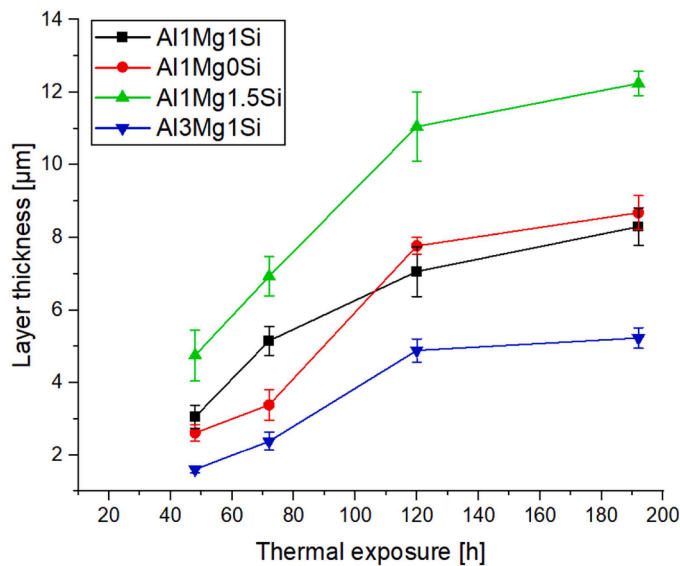


Fig. 19. Layer thickness for the different alloys after post-weld HT as a function of HT time.

Al1Mg1.5Si, while Fig. 23b) shows a higher magnification image of the indents on the interface of the Al3Mg1Si/Ti64. The indents in the individual IMCs are highlighted in the image.

Fig. 24a) shows the average hardness of the different phases as a function of indentation depth. The hardness values, particularly of the IMCs, show pronounced indentation size effect that most likely arises from sample preparation induced surface hardening effects. For the matrix phases and the relatively large Ti(Al,Si)₃ IMC in the Al1Mg1.5Si/Ti64 joint, this effect becomes negligible beyond ~125 nm, where

depth-independent hardness values are observed. Once the hardness reaches a horizontal plateau, surface influences can be considered insignificant. In the Ti(Al,Si)₃ IMC of the Al1Mg1.5Si/Ti64 joint, a depth-independent hardness of approximately 15 GPa is obtained beyond this depth. This phase exhibited the highest hardness, approximately two times the hardness of the Ti64 matrix. Conversely, a depth-independent hardness value could not be reached in the IMCs (TiAl₃ and Al₁₈Mg₃Ti₂) in the Al3Mg1Si/Ti64 joint. The indentation depth in this case was limited to very shallow depths (<80 nm) because of the narrow width of the IMCs. Nevertheless, the load-displacement curves in Fig. 24b) do indicate that these phases would exhibit a slightly lower hardness than that of the Ti(Al,Si)₃ phase in the Al1Mg1.5Si/Ti64 joint since the indentation load typically scales with hardness. The lower hardness is likely the result of the mixture of an overlap between Ti(Al,Si)₃ and the softer Al₁₈Mg₃Ti₂ phase in the Al3Mg1Si/Ti64 joint. The lower hardness of Al₁₈Mg₃Ti₂ compared to Ti(Al,Si)₃ correlates well with the finding of cracks mainly through the Al₁₈Mg₃Ti₂. As the Al₁₈Mg₃Ti₂ forms from liquid after the thermal exposure, particularly during and shortly after the solidification, the relatively soft Al₁₈Mg₃Ti₂ does fail due to the different thermal expansion of Al and Ti. To avoid Al₁₈Mg₃Ti₂ formation it can be noted that a relevant amount was not found in Al1Mg1Si/Ti64 but a significant amount in Al3Mg1Si/Ti64, so the threshold is between 1 : 1 and 3 : 1 Mg : Si, potentially close to the stoichiometric ratio of Mg₂Si of 1.73 : 1 Mg : Si.

3.4. Discussion of intermetallic compound formation mechanisms

The formation of the Mg-rich IMC layer, particularly Al₁₈Mg₃Ti₂, at the Al/Ti interface is strongly dependent on the amount of “free” or diffusible Mg present in the Al alloys. The EDS analysis of Mg enrichment at the Al/Ti interface, see Figs. 16 and 17, suggests that the relative amount of free Mg in the aluminum alloys increases in the following order: Al1Mg1.5Si < Al1Mg0Si < Al3Mg1Si. Although Al1Mg1.5Si and Al1Mg0Si have the same total Mg content (1 wt. %), the presence of Si in

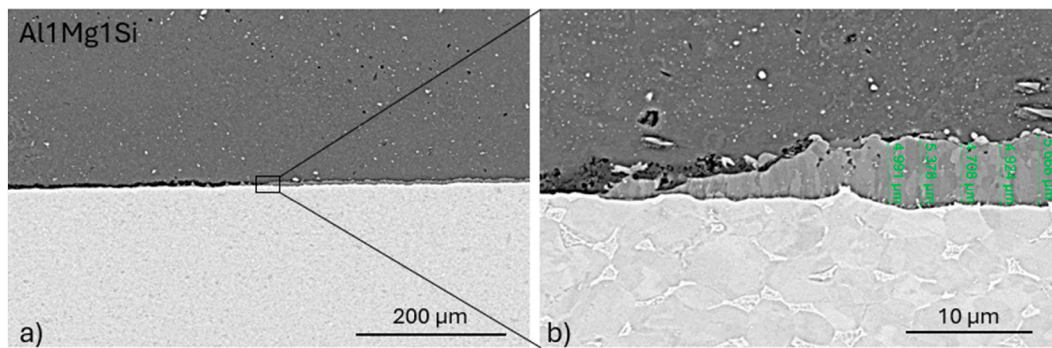


Fig. 20. SEM micrograph of the outer spot weld region, adjacent to the unbonded region, highlighting the IMC layer thickness of an Al1Mg1Si/Ti64 joint after 72 h of HT at 525 °C, showing a) the overview micrograph and b) the region in the transition from unbonded to welded area.

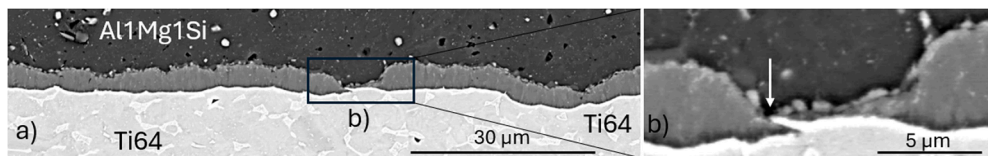


Fig. 21. a) SEM micrograph of a Al1Mg1Si/Ti64 joint interface presenting a region without bonding after 48 h of HT at 525 °C; b) Highlighted area indicated by rectangle in a); Arrow indicates the potentially unbonded region.

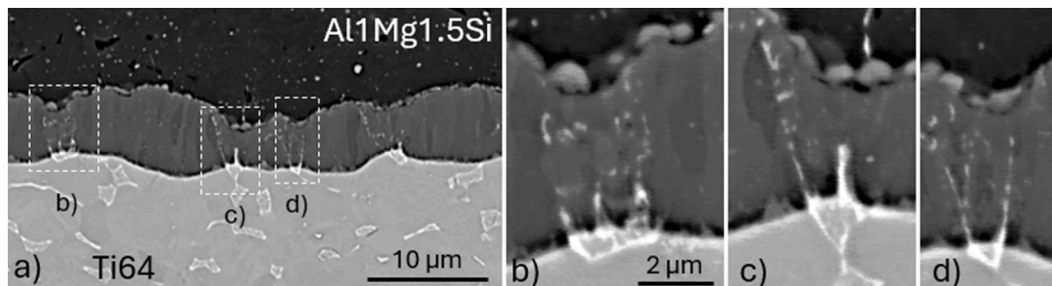


Fig. 22. Micrograph of an Al1Mg1.5Si/Ti64 joint interface heat-treated for 48 h, highlighting the differences between α - and β -Ti on the IMC layer thickness. The areas of α -Ti in contact with the IMC layer show homogeneous growth. The β -Ti regions are characterized by a reduced layer thickness, including brighter areas in the IMC which probably is an indicator of slowly distributing V.

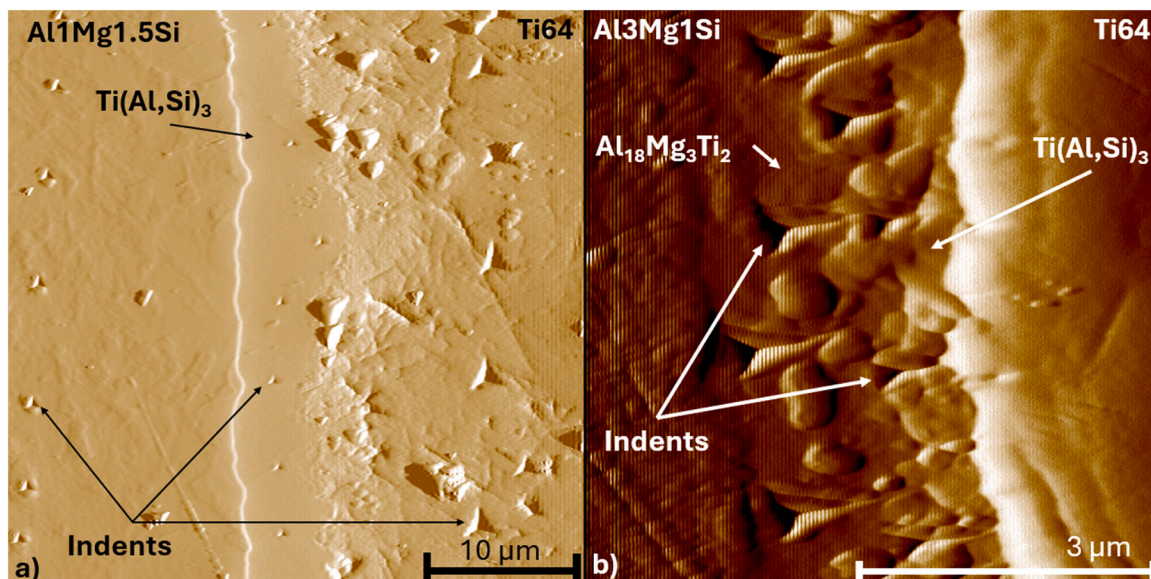


Fig. 23. Nano-indentations over the interface of the a) Al1Mg1.5Si/Ti64 joint and b) Al3Mg1Si/Ti64 joint after HT for 72 h.

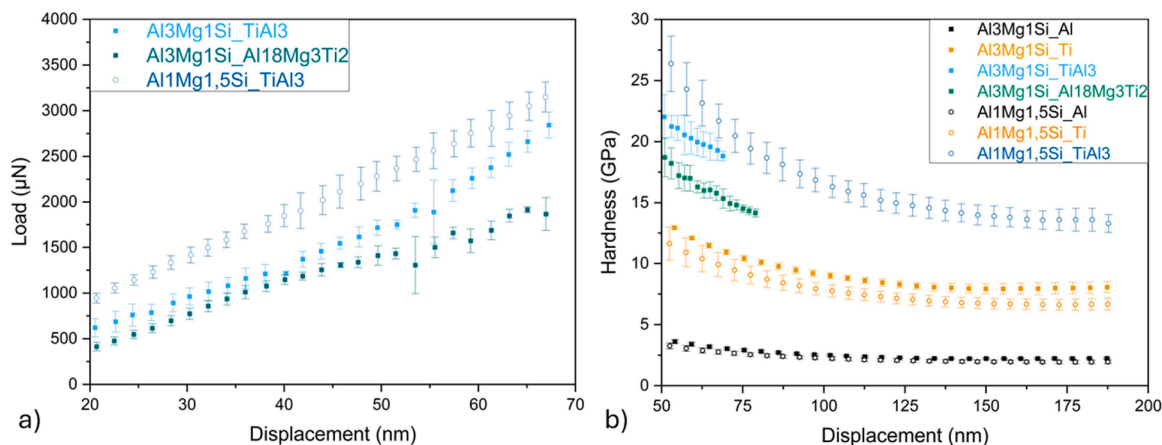


Fig. 24. a) Load over displacement plots for the IMC layers in the Al3Mg1Si/Ti64 and Al1Mg1.5Si/Ti64 joints after HT for 72 h; b) Hardness over displacement plots for the Al1Mg1.5Si/Ti64 and Al3Mg1Si/Ti64 joints after HT for 72 h for the Al side, Ti side and the IMC layers.

Al1Mg1.5Si leads to the formation of Mg_2Si , as confirmed by the HE-XRD results, see Fig. 25. Mg_2Si is a thermodynamically stable compound with a high melting temperature (1085 °C [32]) and does not decompose during welding or post-weld HT. This means that a large portion of Mg atoms are immobilized and do not contribute to diffusion toward the interface. In contrast, in the Si-free Al1Mg0Si alloy, Mg remains primarily in solid solution, enabling significant Mg diffusion toward the interface. In Al3Mg1Si, although 3 wt. % Mg and 1 wt. % Si are present, not all of the Mg react to form Mg_2Si . As a result, <2 wt. % of Mg is bound in Mg_2Si , leaving >1 wt. % of Mg unbound in solid solution. As a result, both Al1Mg0Si and Al3Mg1Si exhibit a higher degree of Mg enrichment at the Al/Ti interface compared to Al1Mg1.5Si.

The formation mechanism of IMCs at the interface can be conceptually divided into two stages. First, atomic diffusion between the Al and Ti sides alters the local chemical composition near the interface. Second, depending on the local composition and temperature, new equilibrium phases are formed. To explore these equilibrium phase formations at 525 °C during the diffusion process, ThermoCalc calculations were performed for the three Al alloys with varying Ti additions, and the results are shown in Fig. 26. As the Ti content increases, the fraction of $TiAl_3$ increases and reaches nearly 100 % at around 37 wt. % Ti. Additionally, a liquid phase begins to form with increasing Ti content and reaches a peak at about 33 wt. % Ti. Compared to Al1Mg1.5Si, both

Al1Mg0Si and Al3Mg1Si show a significantly higher fraction of liquid phase at the same temperature. These predictions are consistent with experimental observations. In Al1Mg1.5Si, only one IMC layer was observed, while in Al1Mg0Si and Al3Mg1Si, two distinct IMC layers formed. The presence of $Al_{18}Mg_3Ti_2$ in the interface is thus attributed to the liquid phase that forms during diffusion at elevated temperatures. Upon cooling, this liquid phase solidifies into the observed $Al_{18}Mg_3Ti_2$ IMC layer. Notably, $Al_{18}Mg_3Ti_2$, which has a melting point just below 480 °C [33], was also found in these regions, supporting the hypothesis that the Mg-rich layer was indeed liquid during heat treatment.

To summarize these findings, Fig. 27 presents a schematic illustration of the IMC formation mechanism. In the presence of only a small amount of diffusible Mg, as in Al1Mg1.5Si, the interfacial reaction is dominated by the interdiffusion of Al and Ti, leading to the formation of a columnar $Ti(Al,Si)_3$ layer. The limited availability of free Mg in this alloy is attributed to the formation of stable Mg_2Si precipitates, which effectively reduce the amount of Mg available for diffusion toward the interface. In the absence of Si (as for Al1Mg0Si/Ti64), more Mg is available to diffuse and react with $Ti(Al,Si)_3$ to form the liquid phase, which can grow alongside $Ti(Al,Si)_3$ in a mixed morphology. In Al3Mg1Si, the excess of free Mg leads to the formation of a continuous liquid phase layer, which is separated from the $Ti(Al,Si)_3$ layer, with only a thin transitional mixed region in between.

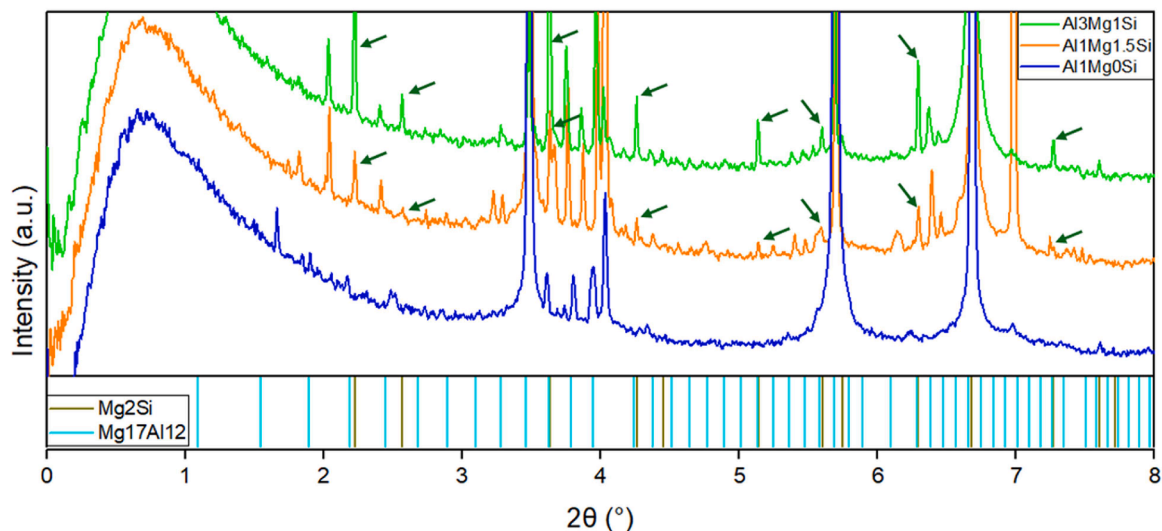


Fig. 25. XRD patterns of Al1Mg0Si/Ti64, Al1Mg1.5Si/Ti64 and Al3Mg1Si/Ti64 joints in the interface using HE-XRD at beamline P07 at PETRA III at Deutsches Elektronen-Synchrotron (DESY).

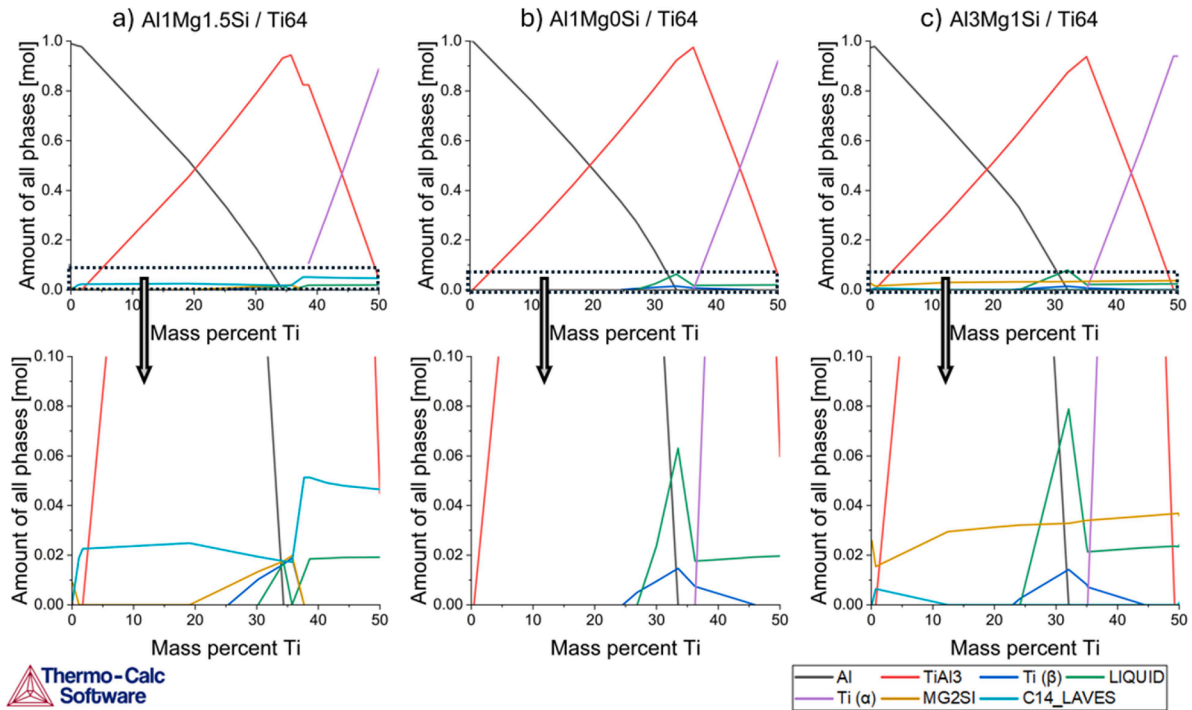


Fig. 26. Phase calculations in the joint interface assuming equilibrium conditions at 525 °C for the alloys a) Al1Mg1.5Si, b) Al1Mg0Si and c) Al3Mg1Si.

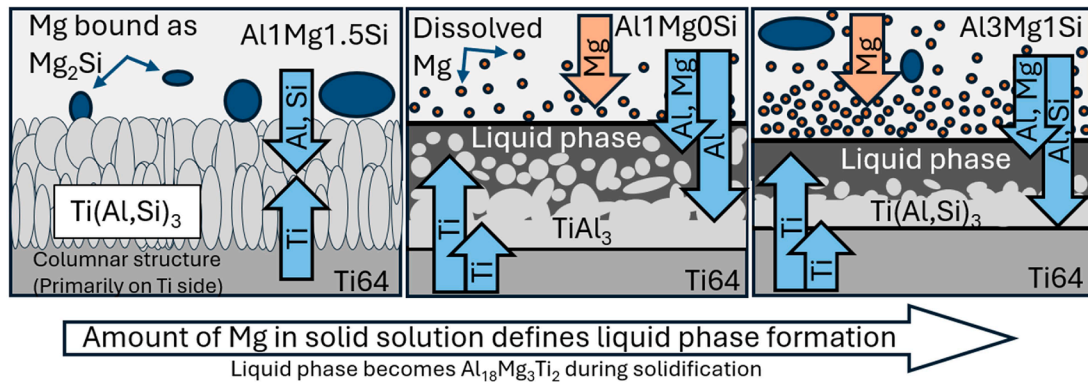


Fig. 27. Schematic illustration of the mechanisms for IMC formation in Al/Ti welds depending on the amount of dissolved Mg within the Al matrix.

4. Conclusions

In the present work, five Al-Mg-Si aluminum alloys, based on AA6013 with varied Mg and Si contents, were successfully joined to Ti6Al4V using refill FSSW. The effects of Mg and Si on the thermal profile during welding and the growth kinetics of IMCs during post-weld heat treatment were systematically investigated. The conclusions are as follows:

- (1) Although the dwell time was set constant to 10 s and the total process time was 14 s, no continuous, homogenous IMC layers were found after welding. In the area below the shoulder, disperse IMC fragments were identified, while in the weld center a higher number of fragments were found, forming discontinuous layers in the joints.
- (2) Two IMCs formed during post-weld heat treatment. For all alloys, Ti(Al,Si)_3 was formed and in the Al1Mg1.5Si/Ti64 and Al1Mg1Si/Ti64 joints, a highly columnar Ti(Al,Si)_3 grain structure was detected. Depending on the amount of Mg that was not bound to Mg_2Si , and therefore directly correlated to the ratio of Mg and Si in the aluminum alloy, the phase $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$ was found. In the Al1Mg0Si/

Ti64 joint, both phases grew in a mixed structure, showing only a tendency of the $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$ being more pronounced on the Al side and TiAl_3 on the Ti64 side. In the Al3Mg1Si/Ti64 joint the phases grew predominantly separated from each other in Ti64-Ti(Al,Si)₃- $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$ -Al3Mg1Si sequence.

- (3) The formation of $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$ hinders strongly the directional grain growth, leading to a significant reduction of Ti(Al,Si)_3 . In Al1Mg0Si/Ti64 the mixed IMCs structure only slightly hindered the directional growth of Ti(Al,Si)_3 , as the directional growth only needs to be shifted in case of simultaneously forming and partially blocking $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$. In contrast, the two individual layers in the Al3Mg1Si/Ti64 joint led to a significant reduction in IMCs, and particularly Ti (Al,Si)₃ layer thickness.
- (4) Despite the difficult circumstances of the thin layers in interfaces of dissimilar welded joints, a trend was found by nano-indentation measurements. Ti(Al,Si)_3 has the highest hardness, approximately two times the hardness of Ti6Al4V, followed by the $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$ in the middle between Ti(Al,Si)_3 and Ti6Al4V. Micrographs revealed multiple cracks in the $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$ which can be correlated with the lower hardness compared to the TiAl_3 . These differences, particularly at

elevated temperatures, as the $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$ first solidifies after the thermal exposure, cause the crack formations in the $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$ layer.

The formation of $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$, showing severe cracks in the interfaces proves that the material selection plays a highly important role in the welding of Al-Mg-Si aluminum alloys to Ti64. To avoid or at least minimize liquid phase formation in the joint interface, an appropriate Mg to Si ratio needs to be chosen for the Al alloy to limit the amount of Mg in solid solution. A clear threshold was not investigated but is assumed to be close to the stoichiometric Mg_2Si ratio of 1.73 : 1 Mg : Si. Otherwise, a continuous $\text{Al}_{18}\text{Mg}_3\text{Ti}_2$ layer can significantly reduce the mechanical properties of the weld due to its lower hardness compared to TiAl_3 and liquid state during heat treatment and potentially welding.

CRedit authorship contribution statement

Lasse Malaske: Writing – original draft, Visualization, Methodology, Investigation, Formal analysis, Conceptualization. **Ting Chen:** Writing – review & editing, Validation, Methodology. **André M. Neves:** Writing – review & editing, Investigation. **Elizabeth Mathew:** Writing – review & editing, Investigation. **Emad Maawad:** Writing – review & editing, Methodology. **Henry Ovri:** Writing – review & editing, Investigation. **Uceu Suhuddin:** Writing – review & editing, Investigation. **Benjamin Klusemann:** Writing – review & editing, Supervision, 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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Supplementary materials

Supplementary material associated with this article can be found, in the online version, at [doi:10.1016/j.rineng.2026.110355](https://doi.org/10.1016/j.rineng.2026.110355).

Data availability

The data obtained from this research is available on ZENODO (DOI: <https://doi.org/10.5281/zenodo.19471500>).

References

- [1] V.S. Gadakh, V.J. Badheka, A.S. Mulay, Solid-state joining of aluminum to titanium: a review, *Proc. Inst. Mech. Eng. L: J. Mater.* 235 (8) (2021) 1757–1799, <https://doi.org/10.1177/14644207211010839>.
- [2] F.J.J. van Loo, G.D. Rieck, Diffusion in the titanium-aluminium system—I. Interdiffusion between solid Al and Ti or Ti-Al alloys, *Acta Metall.* 21 (1) (1973) 61–71, [https://doi.org/10.1016/0001-6160\(73\)90220-4](https://doi.org/10.1016/0001-6160(73)90220-4).
- [3] Y.C. Kim, A. Fuji, Factors dominating joint characteristics in Ti – Al friction welds, *Sci. Technol. Weld. Join.* 7 (3) (2002) 149–154, <https://doi.org/10.1179/136217102225004185>.
- [4] D. Abolhasani, J.H. Kim, W.J. Lee, Y.H. Moon, Asymmetric interlocking and inclined remelting in laser powder bed fusion of horizontally joined AISI 316L–alumina parts, *J. Mater. Res. Technol.* 39 (2) (2025) 3566–3584, <https://doi.org/10.1016/j.jmrt.2025.10.057>.
- [5] A.H. Plaine, U.F.H. Suhuddin, N.G. Alcântara, J.F. dos Santos, Microstructure and mechanical behavior of friction spot welded AA6181-T4/Ti6Al4V dissimilar joints, *Int. J. Adv. Manuf. Technol.* 92 (9–12) (2017) 3703–3714, <https://doi.org/10.1007/s00170-017-0439-2>.
- [6] A. Esmaili, C. Sbarufatti, A.M.S. Hamouda, Characteristics of intermetallic compounds in dissimilar friction stir welding: a review, *Met., Microstruct. Anal.* 8 (4) (2019) 445–461, <https://doi.org/10.1007/s13632-019-00557-w>.
- [7] H. Zhao, M. Yu, Z. Jiang, L. Zhou, X. Song, Interfacial microstructure and mechanical properties of Al/Ti dissimilar joints fabricated via friction stir welding, *J. Alloys. Compd.* 789 (1) (2019) 139–149, <https://doi.org/10.1016/j.jallcom.2019.03.043>.
- [8] B. Li, Y. Shen, L. Luo, W. Hu, Effects of processing variables and heat treatments on Al/Ti-6Al-4V interface microstructure of bimetal clad-plate fabricated via a novel route employing friction stir lap welding, *J. Alloys. Compd.* 658 (2016) 904–913, <https://doi.org/10.1016/j.jallcom.2015.10.288>.
- [9] L. Malaske, U.F.H. Suhuddin, B. Klusemann, Effect of magnesium and silicon on temperature and mechanical properties in refill friction stir spot welding of aluminum to titanium, *Sci. Technol. Weld. Join.* 1040 (8) (2026) 1–8, <https://doi.org/10.4028/p-gj7osM>.
- [10] A.H. Plaine, A.R. Gonzalez, U.F.H. Suhuddin, J.F. dos Santos, N.G. Alcântara, The optimization of friction spot welding process parameters in AA6181-T4 and Ti6Al4V dissimilar joints, *Mater. Des.* 83 (2015) 36–41, <https://doi.org/10.1016/j.matdes.2015.05.082>.
- [11] X. Nan, et al., Interface characterization and formation mechanism of Al/Ti dissimilar joints of refill friction stir spot welding, *Int. J. Adv. Manuf. Technol.* 126 (3–4) (2023) 1539–1551, <https://doi.org/10.1007/s00170-023-11226-2>.
- [12] F. Grassel, L. Malaske, M. Hoffmann, B. Klusemann, (Semi-)Solid-state joining of aluminium and titanium alloys – A critical review, *J. Mater. Res. Technol.* 39 (2025) 3270–3291, <https://doi.org/10.1016/j.jmrt.2025.09.239>.
- [13] V.B. Rao, C.R. Houska, Reactions and diffusion between an Al film and a Ti substrate, *Met. Trans. A* 14 (1) (1983) 61–66, <https://doi.org/10.1007/BF02643738>.
- [14] L. Zhou, et al., Interface microstructure and formation mechanism of ultrasonic spot welding for Al-Ti dissimilar metals, *Int. J. Min. Met. Mater.* 28 (9) (2021) 1506–1514, <https://doi.org/10.1007/s12613-020-2043-y>.
- [15] A. Fuji, In situ observation of interlayer growth during heat treatment of friction weld joint between pure titanium and pure aluminium, *Sci. Technol. Weld. Join.* 7 (6) (2002) 413–416, <https://doi.org/10.1179/136217102225006903>.
- [16] J.A. Österreicher, et al., Dissimilar friction stir welding and post-weld heat treatment of Ti-6Al-4V and AA7075 producing joints of unprecedented strength, *J. Adv. Join. Process.* 9 (5) (2024) 100213, <https://doi.org/10.1016/j.jajp.2024.100213>.
- [17] D. Bandhu, Multiobjective optimization of RMD welding parameters for ASTM A387 steel using Taguchi method and artificial neural networks, *Measurement* 262 (2026) 120114, <https://doi.org/10.1016/j.measurement.2025.120114>.
- [18] M. Liu, C. Zhang, Z. Meng, G. Zhao, L. Chen, TiAl₃ nucleation mechanism and atomic-scale interface features in the Al/Ti composite structures, *Compos. B* 226 (8) (2021) 109331, <https://doi.org/10.1016/j.compositesb.2021.109331>.
- [19] S.P. Gupta, Intermetallic compounds in diffusion couples of Ti with an Al-Si eutectic alloy, *Mater. Charact.* 49 (4) (2002) 321–330, [https://doi.org/10.1016/S1044-5803\(02\)00342-X](https://doi.org/10.1016/S1044-5803(02)00342-X).
- [20] L.F. Mondolfo, Aluminum alloys: structure & properties, Butterworth-Heinemann (1976), <https://doi.org/10.1016/C2013-0-04239-9>.
- [21] A. Fuji, K. Ikeuchi, Y.S. Sato, H. Kokawa, Interlayer growth at interfaces of Ti/Al-1%Mn, Ti/Al-4-6%Mg and Ti/pure Al friction weld joints by post-weld heat treatment, *Sci. Technol. Weld. Join.* 9 (6) (2004) 507–512, <https://doi.org/10.1179/136217104225021797>.
- [22] K.S. Bang, K.J. Lee, H.S. Bang, H.S. Bang, Interfacial microstructure and mechanical properties of dissimilar friction stir welds between 6061-T6 aluminum and Ti-6%Al-4%V alloys, *Mater. Trans.* 52 (5) (2011) 974–978, <https://doi.org/10.2320/matertrans.L-MZ201114>.
- [23] M.p. LIU, et al., Aging behavior and mechanical properties of 6013 aluminum alloy processed by severe plastic deformation, *Trans. Nonferrous. Met. Soc. China* 24 (12) (2014) 3858–3865, [https://doi.org/10.1016/S1003-6326\(14\)63543-3](https://doi.org/10.1016/S1003-6326(14)63543-3).
- [24] C. Schilling, J. Dos Santos, Method and device for linking at least two adjoining work pieces by friction welding, EP1230062B1, 2000.
- [25] G.F. Vander Voort, Metallographic specimen preparation for electron backscattered diffraction part II*, *Pract. Metallogr.* 48 (10) (2011) 527–543, <https://doi.org/10.3139/147.110151>.

- [26] N. Schell, R.V. Martins, F. Beckmann, H.U. Ruhnau, R. Kiehn, A. Schreyer, The high energy materials science beamline at PETRA III, *MSF* 571-572 (2008) 261–266, <https://doi.org/10.4028/www.scientific.net/MSF.571-572.261>.
- [27] A. Hammersley, *FIT2D: An Introduction And Overview*, ESRF Internal Report, ESRF97HA02T, 1997. Grenoble, France.
- [28] F. Zhang, et al., In situ structural characterization of ageing kinetics in aluminum alloy 2024 across angstrom-to-micrometer length scales, *Acta Mater.* (2016), <https://doi.org/10.1016/j.actamat.2016.03.058>.
- [29] J.O. Andersson, T. Helander, L. Höglund, P. Shi, B. Sundman, Thermo-Calc & DICTRA, computational tools for materials science, *CALPHAD* 26 (2) (2002) 273–312, [https://doi.org/10.1016/S0364-5916\(02\)00037-8](https://doi.org/10.1016/S0364-5916(02)00037-8).
- [30] *TCS Ni-Based Superalloys Database, Version 10*, S, Thermo-Calc Software AB, Sweden, 2020.
- [31] W.C. Oliver, G.M. Pharr, An improved technique for determining hardness and elastic modulus using load and displacement sensing indentation experiments, *J. Mater. Res.* 7 (6) (1992) 1564–1583, <https://doi.org/10.1557/JMR.1992.1564>.
- [32] G. Frommeyer, S. Beer, K. Oldenburg, Microstructure and mechanical properties of mechanically alloyed intermetallic Me₂Si-Al alloys, *Int. J. Mater. Res.* 85 (5) (1994) 372–378, <https://doi.org/10.1515/ijmr-1994-850516>.
- [33] S.H. Lee, et al., Crystallographic characterization of Al₁₈Mg₃Ti₂ intermetallic phase in Al–Zn–Mg–Cu–Zr–Ti alloy, *J. Alloys. Compd.* 844 (2020) 156173, <https://doi.org/10.1016/j.jallcom.2020.156173>.