



Targeted defect mitigation via grain refinement at multi-alloy aluminium interfaces in PBF-LB/M via localised TiC particle delivery

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ABSTRACT

Hybrid manufacturing, combining laser powder bed fusion (PBF-LB/M) with conventionally manufactured high-strength aluminium substrates, is hindered by interfacial hot cracking and porosity that compromise interface integrity and mechanical performance. This study investigates two TiC nanoparticle delivery strategies to reduce defects and to improve the mechanical properties of the interface between AA2017 substrates and PBF-LB/M-deposited AlSi10Mg: (i) a TiC nanoparticle interlayer applied directly to the substrate and (ii) TiC nanoparticles blended with the AlSi10Mg feedstock. Quantitative defect analysis combined with EBSD, EDS, micro-hardness, and miniature shear testing was used to correlate particle distribution with microstructural evolution, defect formation, and mechanical performance. Under optimised processing conditions, the TiC interlayer reduced the total defective interface area by 87 %, with crack-related defects reduced by >91 %, while the blended approach achieved a total defect reduction of 77 % and ~84 % crack reduction. The interlayer promoted localised interfacial grain refinement via increased nucleant particle density within the crack-prone interface region, resulting in superior defect suppression. Depositing TiC as an interlayer, rather than blending, also reduced particle agglomeration and sintering. Furthermore, the interlayer approach suppressed interface softening, increased interfacial shear strength with reduced variability, and shifted the fracture path away from the interface into the AlSi10Mg region. The results demonstrate that the effectiveness of TiC modification depends on the spatial distribution of particles, establishing localised interface engineering as a more effective strategy than homogeneous feedstock modification for improving the integrity and mechanical performance of hybrid multi-alloy PBF-LB/M interfaces.

1. Introduction

Laser powder bed fusion (PBF-LB/M) of metallic materials has attracted attention from the automotive, energy, robotics, medical, and aerospace industries [1–3]. However, PBF-LB/M currently faces challenges in manufacturing bulk parts due to extended processing times [4] and limited chamber sizes [5,6]. As a result, bulk structural components are often manufactured using conventional routes and subsequently joined to additively manufactured (AM) sections with complex geometries. These structural bulk regions are often made of high-strength aluminium alloys, which exhibit a large solidification gap, making them prone to cracking under PBF-LB/M's rapid cooling conditions [7–11]. Research has been conducted on modifying conventional alloys to make them suitable for AM [8–10]. Despite some success in this regard, AM applications are still dominated by castable Al–Si alloys [1,12,

13], necessitating post-process joining when combined with high-strength alloys. However, such post-process joining methods introduce undesirable interface characteristics, including joint softening [14–16], porosity and cracks [17–19], reducing the interface integrity and mechanical properties.

To overcome these limitations, an in-situ PBF-LB/M hybrid alloy manufacturing approach has been explored [20], where AM material is directly deposited onto conventionally manufactured substrates. This integration eliminates post-processing joining steps, simplifying production and potentially lowering costs. In our previous work [20], we investigated the feasibility of this hybrid manufacturing approach by printing AlSi10Mg parts onto high-strength alloy (AA2017) substrates and identified the dominant interfacial defects as hot cracking and process-induced porosity. These defects were found to stem from epitaxial grain growth and inherent substrate pore coalescence during

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PBF-LB/M processing and associated thermomechanical phenomena during solidification. Since crack formation at the interface is closely linked to epitaxial grain growth, an equiaxed microstructure at the interface is expected to disrupt this growth and consequently reduce the crack susceptibility of the interface.

Various feedstock material-based modifications have been explored to promote grain refinement in aluminium alloys, including elemental alloying additions [21–26], particle-based inoculants [9,27–29], and eutectic forming elements [30]. Among these, Ti-based additions have been shown to suppress solidification cracking by promoting equiaxed grain structures in high-strength aluminium alloys [8,25,26]. Similarly, TiC particles have been shown to act as effective grain refiners in aluminium alloys processed by PBF-LB/M, contributing to microstructural refinement through the formation of Ti-based phases and heterogeneous nucleation [31,32]. A refined grain structure [33,34], along with the presence of hard, fine dispersed particles within the matrix, has been shown to improve the mechanical performance of materials [27, 29]. However, these approaches predominantly rely on homogeneous incorporation of Ti or TiC within the feedstock and do not address how the spatial distribution of grain-refining particles influences defect formation at critical regions such as multi-alloy interfaces, where cracking and porosity are most pronounced.

In this work, we investigate whether localised TiC nanoparticle delivery at the interface between AA2017 alloy substrates and PBF-LB/M-deposited AlSi10Mg can enable targeted defect mitigation compared to distributed addition via powder blending. Two TiC incorporation strategies are examined: (i) a TiC nanoparticle interlayer applied directly at the substrate surface and (ii) TiC nanoparticles blended with AlSi10Mg powder. By combining quantitative defect analysis with EBSD, EDS and mechanical characterisation, this study tries to find how spatial particle distribution influences grain refinement, agglomeration behaviour, defect suppression at the interface and the resulting mechanical performance of hybrid interfaces, establishing a mechanistic basis for interface-specific defect mitigation in hybrid multi-alloy PBF-LB/M systems.

2. Materials and methods

An AconityMIDI + PBF-LB/M machine from Aconity3D GmbH was used in this study to carry out prints. 20 layers of rectangular geometries of size 6 mm × 2 mm were printed on the substrates in an argon environment with an oxygen level of less than 600 ppm. Process parameter combinations (Table 1) with varying laser powers were employed to evaluate the performance of each feedstock powder case. This parameter range was selected based on the process window identified in our previous work [20], where 350 W was found to produce the fewest interfacial defects for an unmodified interface. The present range was swept around this value to determine if it holds for the TiC-modified interface. The volumetric energy density (VED) imparted was calculated using equation (1), accounting for the beam diameter and the layer thickness [35,36]. The parameters P , v , d , and t represent laser power, scan speed, beam diameter and layer thickness, respectively.

$$VED = \frac{P}{v * d * t} \quad (1)$$

Table 1
Process parameters used in the study.

Laser power (W)			Scan speed (mm/s)	Hatch spacing (mm)	Laser spot diameter (mm)
Min	Max	Step size	1650	0.1	0.08
250	500	50			

2.1. Substrate

AA2017 (Al Cu Mg1/EN AW-2017A/EN AW-Al Cu4 Mg Si) circular substrates of 25 mm diameter and 10 mm thickness were machined from a 500 × 500 mm plate in T451 heat-treated condition. Before installation into the PBF-LB/M machine, the substrates were sandblasted at a pressure of 7 bar. Substrates were mounted on a dedicated holder as discussed in our previous work [20] to ensure a constant first-layer thickness across the substrate. This is done to achieve precise control over the thickness of the first powder layer, which is essential for accurate interface studies.

2.2. Powder

Feedstock powders AlSi10Mg and TiC nano-powder were used in the study. Gas-atomised AlSi10Mg powder was purchased from ECKART TLS GmbH. The AlSi10Mg powder had a particle size distribution (PSD) of 20–63 µm, with particle size values of D_{10} , D_{50} , and D_{90} , respectively, representing particles of 24.35, 40.05, and 60.45 µm in diameter. The TiC powder was purchased from IoLiTec Ionic Liquids Technologies GmbH. The TiC powder had a mean particle size of 50 nm.

2.3. Blending of TiC with AlSi10Mg powder

Blends containing 1, 5, and 10 wt% TiC were prepared by mixing TiC nano-powder with AlSi10Mg powder in 120 mL glass bottles. The mixing was done using a tumble mixer (Turbula T2F, WAB-Group) at a rotational speed of 101 min^{−1} for 15 min. The total mass of each blended batch was 60 g. This blending strategy was inspired by the work of Freundl and Jäggle [31], who reported reduced agglomeration of TiC nanoparticles in the final parts under similar blending conditions. The resulting powder blend was subsequently spread on the substrate (Fig. 1) to fabricate the previously mentioned sample geometry.

2.4. TiC interlayer

To introduce TiC at the interface as an interlayer, TiC nano-powder layers with thicknesses ranging from 1 to 20 µm were applied on the sandblasted substrate surface (Fig. 2a). For this, the substrate was positioned at a depth corresponding to the desired TiC layer thickness with respect to the substrate holder. An initial layer of TiC was spread onto the substrate, and a flat silicone spreader attached to the machine recoater was moved across the substrate holder, while TiC powder was continuously supplied as required, until the powder layer surface was level with the height of the holder surface (Fig. 2a). Subsequently, an additional 30 µm-thick AlSi10Mg powder was deposited on top of the TiC interlayer as shown in Figs. 1 and 2b, before exposing the first layer to the laser radiation. Subsequently, the standard printing procedure with pure AlSi10Mg was continued after the modified first layer.

To assess the surface homogeneity of the deposited TiC interlayer and its reproducibility across independent depositions, confocal laser scanning microscopy using a Keyence VK-X-1100 laser microscope was carried out to measure areal surface roughness (S_a) at five regions (1000 × 1000 µm²) on both the bare substrate and the TiC-covered surface, for three independently prepared substrates using the thinnest (1 µm) interlayer condition, representing the most critical case for coverage uniformity. Mean surface roughness of substrates increased from 2.79 ± 0.48 µm in the sandblasted condition to 3.69 ± 0.94 µm after TiC deposition ($n = 15$ regions per condition, pooled across three substrates), an increase consistent in magnitude with the nominal 1 µm interlayer thickness (see Supplementary Fig. S1). A two-way ANOVA (Condition × Substrate) confirmed that this roughness increase between sandblasted and TiC-deposited samples was highly significant, while the Condition × Substrate interaction was not, indicating that the magnitude of the TiC-induced roughness increase was statistically consistent across all three independently prepared substrates. A Brown–Forsythe

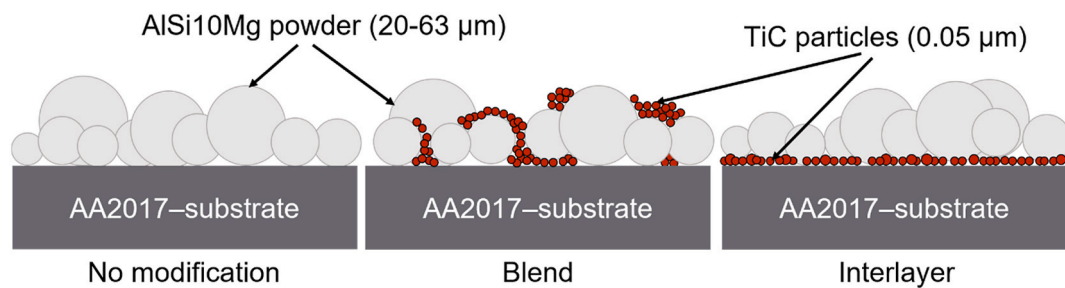


Fig. 1. Schematic of initial layer spread on substrate surfaces in cases of no modification, blended and interlayer approaches.

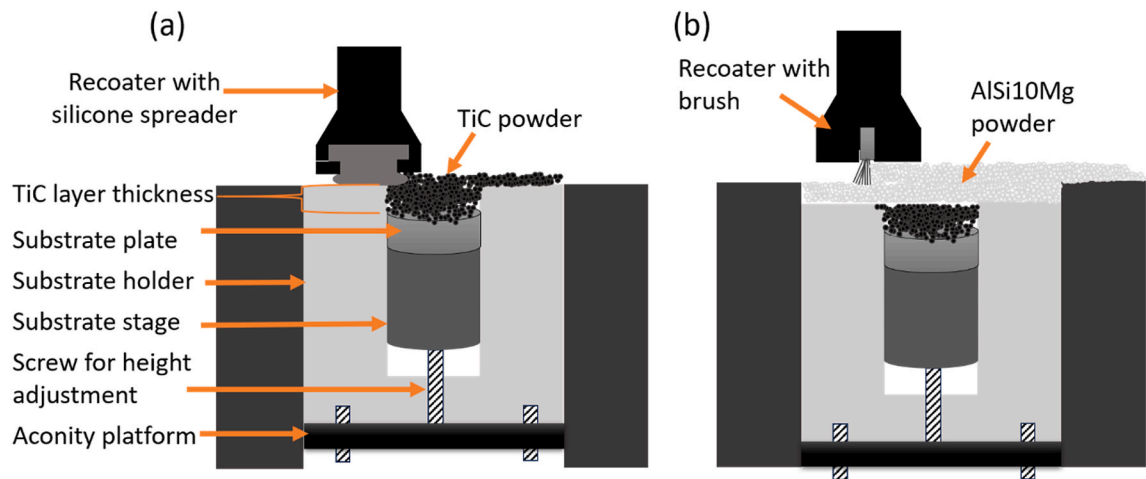


Fig. 2. Interlayer preparation process. (a) Creation of a TiC layer directly on the substrate surface. (b) Spreading of AlSi10Mg powder on top of the TiC layer.

test further showed a significant increase in the variance of S_a following TiC deposition (full statistics in [Supplementary Table S1](#)). Taken together, these results indicate that while local TiC coverage in interlayer preparation is non-uniform at the microscale, its deposition is reproducible in both magnitude and spatial statistics across independent substrate preparations.

Different TiC interlayer thicknesses introduce different amounts of TiC into the first processed layer. To estimate the corresponding TiC weight fraction, the mass of TiC deposited as a 15 μm interlayer was determined by weighing the substrate before and after powder deposition, yielding a deposited TiC mass of 0.006 g. The mass of the overlying 30 μm AlSi10Mg powder layer was estimated as 0.012 g, assuming a powder-bed packing fraction of 0.60. Based on these values, the 15 μm interlayer corresponds to an estimated TiC content of approximately 33 wt% TiC in the first processed layer. TiC masses for the remaining interlayer thicknesses were estimated by assuming linear scaling with interlayer thickness, and the corresponding TiC weight fractions were calculated using the same procedure, as summarised in [Table 2](#). These estimates are intended to provide a comparative measure of TiC addition between the different interlayer configurations and should not be

interpreted as representing the local composition after melting and mixing.

2.5. Microscopy and analyses

Post-manufacturing metallographic preparation was performed to examine interface characteristics. The prepared samples were examined using a Keyence VK-X-1100 laser microscope to measure interface integrity. In the context of this study, the interface region is defined as the initial layers comprising the fusion zone (FZ) and the adjacent partially melted zone (PMZ) within the heat-affected zone (HAZ). For quantitative analysis, four cross-sections for each condition were prepared, and one interface region (1.5 × 0.15 mm) of the samples was specifically cropped out per cross-section to carry out image analysis for quantifying the defect fractions and defect coverage within the interface. Reported values represent the mean ± standard deviation unless otherwise stated. Image analysis was performed using ImageJ software by thresholding defect regions relative to the bulk interface. The raw microscopy images were converted into an 8-bit format, segmented based on grey values, and binarised. The particle analysis function of ImageJ was used to quantify and classify the number, size, and types of interface defects. Defect morphology was classified based on sphericity, with features exhibiting low circularity (0–0.5) identified as cracks and those with higher circularity (0.5–1.0) identified as pores. A threshold of 15 μm² was maintained to separate process-induced pores from pre-existing ones, as the pores in the substrate material were always smaller than this threshold. All remaining features excluded by these criteria were categorised as “other defects”, including substrate pores.

High-resolution microstructural characterisation was performed using a Zeiss Ultra 55 scanning electron microscope (SEM). Unless otherwise specified, all SEM images were acquired at an accelerating voltage of 20 kV. To further assess the effect of feedstock modifications,

Table 2
Estimated TiC weight fraction in the first processed layer for different TiC interlayer thicknesses.

Interlayer thickness (μm)	TiC mass (g)	AlSi10Mg mass (g)	Calculated TiC (wt %)
1	0.0004	0.012	3
5	0.002		14
10	0.004		25
15	0.006 (measured)		33
20	0.008		40

electron backscatter diffraction (EBSD) and energy-dispersive X-ray spectroscopy (EDS) analyses were also conducted. EDS analysis was performed using an EDAX Octane 9 Super Elite detector to evaluate the interface composition and behaviour of TiC particles for both TiC addition strategies, including spatial distribution, agglomeration, and sintering behaviour. In particular, the analysis focused on changes in the count and distribution of Ti-rich particles across the interfaces. All micrographs used for this analysis were acquired at a magnification of 500 $\times$ and an accelerating voltage of 20 kV. The images were cropped into 250 $\times$ 100 μ m strips, and Ti and C enrichment was quantified using EDS. The cropped images were converted to 8-bit format and analysed using ImageJ, where thresholding was applied to selectively isolate TiC particles.

To determine the interface composition of the matrix of each TiC addition strategy, ten particle-free regions of interest within the three characterised interface strips of each condition were selected (in total 3 $\times$ 10 = 30 regions per TiC addition strategy) using the carbon and copper signals as positional references and averages were calculated. Composition values falling below the minimum detection limit (MDL) for a given element were excluded from the mean. Scheil–Gulliver solidification simulations were performed on the measured mean interface composition for each condition, using the “Scheil with solute trapping” formulation with the Aziz model [37] for solute partitioning using Thermo-Calc 2026a (Thermo-Calc Software AB, Sweden) with the TCAL10 database. The solidification front velocity was estimated from the scan speed (1650 mm/s) and an assumed solidification angle (α = 85°, chosen to avoid the singularity at α = 90° while representing the lower region of the melt pool). The growth restriction factor (Q) was extracted as the initial slope of the temperature–mole fraction solid curve within the primary FCC field (following completion of the Ti-bearing peritectic reaction), determined by linear regression over the initial linear region of each curve.

EBSD maps were obtained using an EDAX Velocity Pro detector and used to investigate the dispersion of inoculant elements within the microstructure and to determine whether variations in the amount and type of inoculant addition promote grain refinement, thereby improving the interface integrity. Grain size was quantified as the area-weighted mean grain diameter using OIM Analysis. Area weighting was selected as it emphasises the dominant grain structure within each region. In addition, quantitative pore analysis was performed by evaluating pore size and spatial distribution in 450 $\times$ 1500 μ m regions extracted from microscopy images acquired at multiple depths (n = 4 specimens per condition). This analysis enabled the correlation of pore evolution with TiC modification strategies and processing conditions.

2.6. Mechanical testing

Vickers microhardness measurements were performed across different interface samples using a QATM QNESS 60 A + EVO microhardness tester with a test load of HV0.005 (5 gf). Hardness mapping was conducted across the substrate–AM interface using indentation lines positioned parallel to the interface at fixed distances relative to the interface. For each distance position, five indentations were performed on each of two independent cross-sections, resulting in ten measurements per position (n = 10). The centre-to-centre spacing between adjacent indentations was 0.25 mm (average indentation diagonal length was $\sim$ 5 μ m). Prior to testing, specimens were prepared using standard metallographic grinding and polishing procedures followed by light etching using 5 % NaOH solution to reveal the interface. Indentations affected by pores, cracks, or poorly resolved indentation boundaries were excluded from the analysis and replaced where possible by measurements performed at equivalent distances from the interface.

Preliminary tensile testing was conducted to evaluate the mechanical behaviour of the hybrid specimens. However, all specimens consistently fractured within the AlSi10Mg region rather than at the hybrid interface,

preventing direct assessment of the interfacial mechanical strength. Consequently, miniature shear testing was employed to specifically evaluate the strength of the hybrid interface. The interfacial shear behaviour of the hybrid specimens was evaluated using an in-situ tensile/compression module adapted for miniature shear testing (Kammrath & Weiss GmbH) equipped with a 5 kN load cell. Cuboidal specimens with nominal dimensions of 3 $\times$ 3 $\times$ 6 mm were prepared, with the interface positioned at the mid-plane of the specimen. The shear tests were performed under displacement-controlled loading conditions using a constant crosshead speed of 8 μ m/s. The measured shear plane cross-sectional area was used to calculate the effective shear strength from the maximum recorded load divided by the measured shear area. For each condition, five specimens were tested to evaluate repeatability and statistical variation.

3. Results

For ease of discussion, the samples are named in Table 3 according to their manufacturing method.

3.1. Interface defect reduction using TiC modification strategies

A systematic evaluation of processing parameters and TiC addition strategies was conducted to identify optimal conditions for each configuration (see supplementary sections S2 and S3). For the interlayer approach, intermediate layer thicknesses (5–15 μ m) consistently resulted in lower defect fractions, while both insufficient (1 μ m) and excessive TiC addition (20 μ m) led to increased defect formation (see Supplementary Fig. S2). For the blended feedstock, low TiC contents were found to be more effective, with the minimum defect fraction achieved for the 1 wt% TiC blend processed at 105 J/mm³, whereas higher TiC contents resulted in increased defect formation, largely independent of processing parameters (see Supplementary Fig. S3). Based on this optimisation, the best-performing conditions for each strategy were selected for direct comparison.

Both the TiC interlayer and TiC-blended feedstock showed better interface integrity than the reference sample (Fig. 3). In addition to reducing the area occupied by defects, TiC modification also significantly reduced the defect count. For the best-performing conditions, the defective interface area was reduced by more than 75 % for both TiC modification approaches, with the interlayer strategy achieving a reduction exceeding 86 %. This improvement is further illustrated by the presence of large spherical pores and cracks in the Mix-1-105 interface (Fig. 4c), which are markedly reduced in the Int-5-72 interface (Fig. 4b).

3.2. Dominant role of crack suppression in defect reduction

Defect reduction at the interface is primarily governed by the suppression of cracking rather than porosity (Fig. 5). The reference condition contains approximately 1.08 % crack area, which decreases to 0.09 % for the TiC interlayer and 0.17 % for the blended condition. In comparison, pore area decreases from 0.34 % in the reference condition to 0.1 % and 0.16 % for the interlayer and blended conditions,

Table 3
Nomenclature and details of samples used in the study.

Sample Name	Description
Reference	AlSi10Mg printed on AA2017 without any interface modification
Mix	AlSi10Mg blended with TiC nanoparticles before deposition
Int	TiC interlayer between AA2017 and AlSi10Mg

Note: The numbers, separated by a hyphen (“-”) following the sample name, denote layer thickness/blend ratio and processed energy density values, respectively. Two examples are added below.

Int-5-72: 5 μ m TiC interlayer processed under 72 J/mm³.

Mix-1-105: 1 wt% TiC blend sample processed under 72 J/mm³.

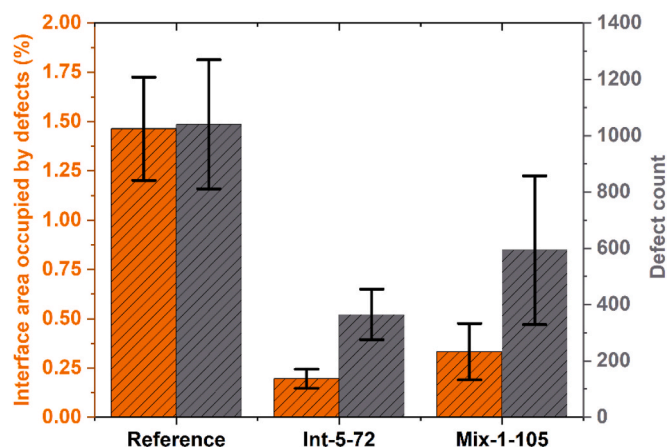


Fig. 3. Defect density and defect count measured at the interface regions of size $150 \times 1500 \mu\text{m}$ for the reference case and optimised TiC modifications. Bars show mean values from $n = 4$ analysed interface regions; error bars denote standard deviation.

respectively. In both TiC-modified approaches, the most pronounced improvement was the significant suppression of cracking at the interface. This crack-dominated behaviour is consistent across the investigated processing window for the interlayer configuration, as shown in supplementary section S4 (Fig. S4).

3.3. Defect reduction mechanisms

EBSD analyses revealed a pronounced reduction in grain size in the FZ and HAZ of the TiC-modified samples compared to the reference sample (Fig. 6). Significant variations in grain size and morphology were observed across AM regions for the different sample types. Epitaxially grown grains are clearly visible in the EBSD image of the reference sample (Fig. 6a). Localised grain refinement is visible for the Int-5-72 sample just above the substrate grains, leading to a triplex grain structure, varying from large substrate grains at the bottom to equiaxed interface grains towards the elongated grains in the AM part (Fig. 6b). However, in the Mix-1-105 sample (Fig. 6c), the AM region is characterised by refined grains, with some large grains observed at the interface; these large grains are smaller than in the reference sample.

The quantitative comparison between samples showed clear variation in grain diameter at the interface (Fig. 6d). Each data point in the graph represents the area-weighted mean grain diameter within a $50 \mu\text{m}$ strip within the EBSD image of the corresponding sample. The reference sample exhibited the largest mean grain diameter at the interface ($30.5 \mu\text{m}$), followed by the blended sample ($17.5 \mu\text{m}$). In the interlayer sample, a reduction in grain diameter is observed immediately before the interface, resulting in a local minimum that limited the interface grain diameter to $10.7 \mu\text{m}$. Notably, the minimum grain size observed in the

interlayer sample is comparable to the lowest grain sizes measured in the core AM region of the blended sample, highlighting a spatially localised grain refinement in the interlayer-modified case. With increasing distance from the interface in the AM build direction, grain sizes in the TiC-interlayer and reference samples converge to similar values. In contrast, the blended sample maintains lower grain sizes throughout the AM region, owing to TiC distributed across the AM region by the blended feedstock. However, the grain size gradually increases and reaches that of the reference sample at around $50 \mu\text{m}$ from the interface into the substrate. Grain size measurements within the substrate region show irregular variations across all samples, reflecting the heterogeneous grain structure of the substrate material.

3.4. Agglomeration and sintering

Throughout this study, sintered TiC micro-particles were observed in samples regardless of the TiC modification strategy (Fig. 7). As an example, an individual TiC-rich particle is shown in Fig. 7a, exhibiting an irregular morphology, size, and appearance consistent with sintered microparticles [31]. Elemental mapping (Fig. 7b) indicates Ti enrichment, and XRD measurements (see Supplementary Fig. S6) confirm the presence of TiC particles. In the Int-5-72 sample, Ti-rich particles are present at a high number density and are preferentially concentrated near the fusion boundary (Fig. 7c–e). In contrast, for the Mix-1-105

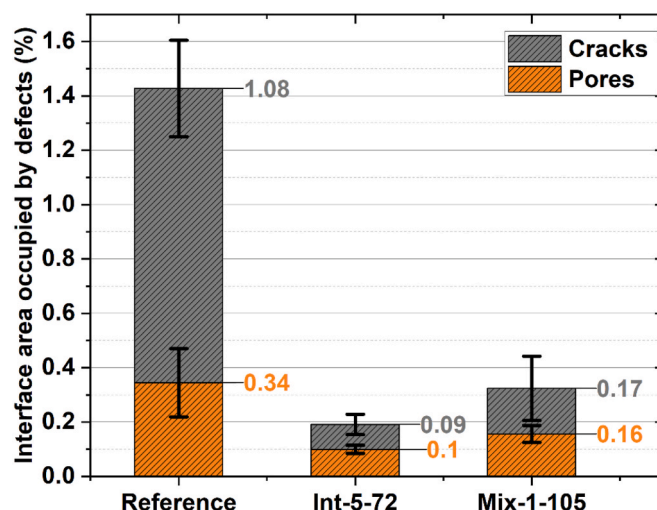


Fig. 5. Breakdown of the interface area occupied by cracks, pores, and other defects for the reference and optimised TiC-modified conditions. The total height of each stacked bar represents the total interface area occupied by defects, while leader lines show the individual contribution of each stacked region. Bars represent the mean defect area measured from four analysed interface regions; error bars indicate one standard deviation.

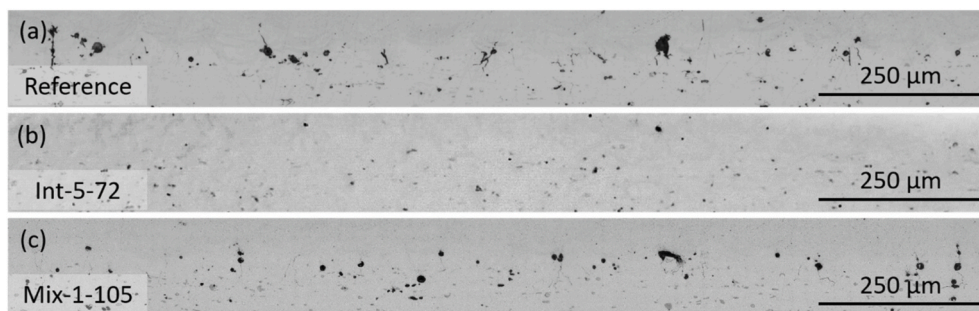


Fig. 4. Micrographs of cross-sections of (a) Reference, (b) Int-5-72, and (c) Mix-1-105. Each image shows a $150 \times 1500 \mu\text{m}$ interface region, with the build direction from bottom to top.

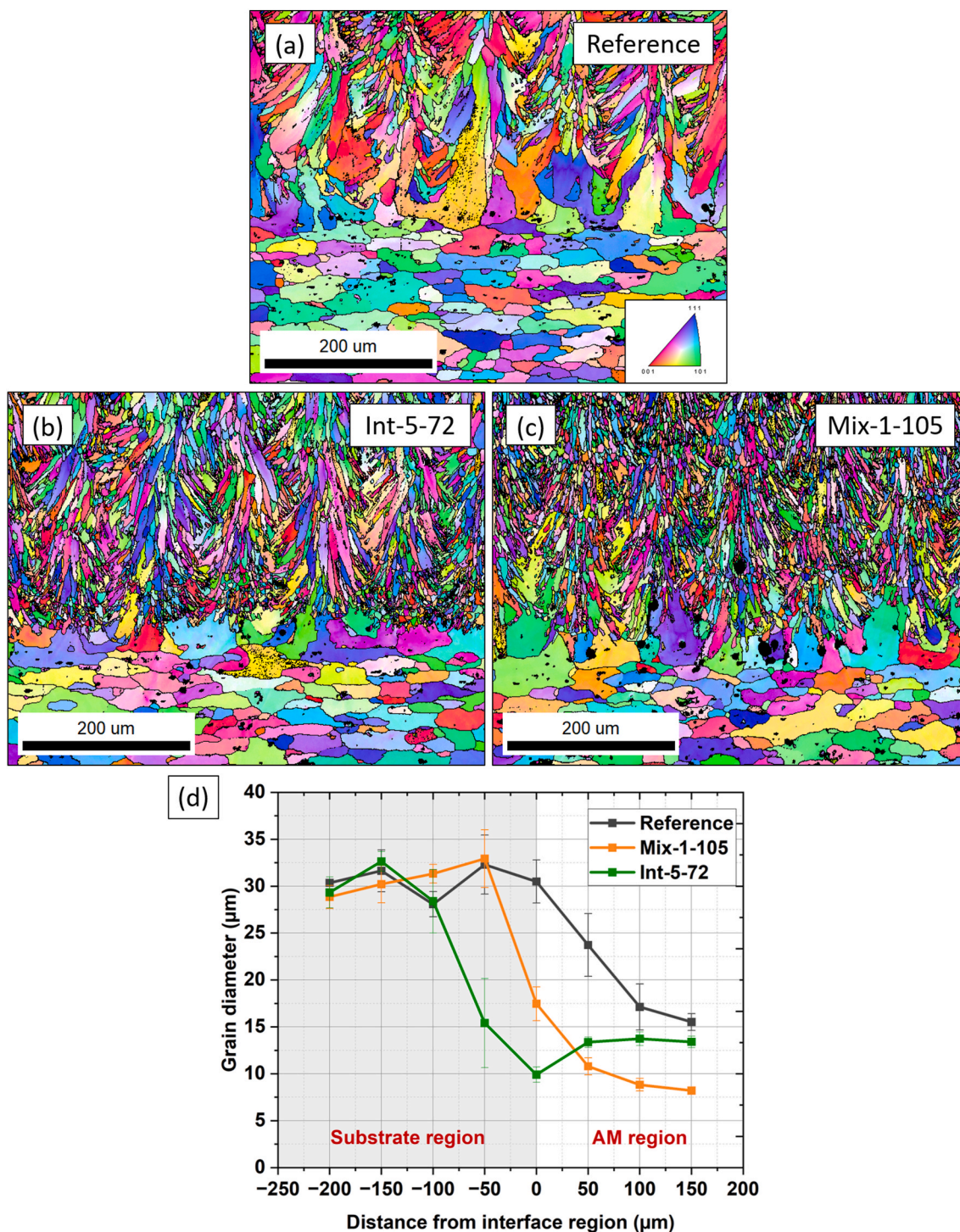


Fig. 6. IPF maps of the (a) Reference, (b) Int-5-72, (c) Mix-1-105 interfaces. (d) Area-weighted mean grain size variation of different sample types across the interface and the nearby region. Each point represents the area-weighted mean grain diameter within a 50 μm strip averaged across three different ($n = 3$) cross sections.

sample, Ti-rich particles are distributed throughout the entire AM region (Fig. 7d–f). Quantitative analysis (Fig. 7g) shows a clear difference between the interlayer and blended strategies, with the blended feedstock exhibiting a lower particle count but a substantially larger average particle size, indicating greater susceptibility to sintering in the blended approach. A similar trend is observed in Fig. S5 (see Supplementary Section S5), where increasing the TiC blend ratio from 1 to 10 wt% leads to increased particle coarsening and defect formation.

3.5. Crack suppression via TiC-induced grain nucleation

Correlative EDS-EBSD analysis identified that TiC nano-agglomerates lead to nucleation of grains (Fig. 8). Higher-magnification investigations show how grain refinement resulted in defect reduction. Fig. 8a presents a low-magnification image of the interface region of the TiC interlayer sample. A magnified view of a crack segment originating from the substrate side of the interface is shown in Fig. 8b. The crack is arrested, and a particle in the close vicinity of the crack tip can also be observed. In Fig. 8c and d, a distinct grain is

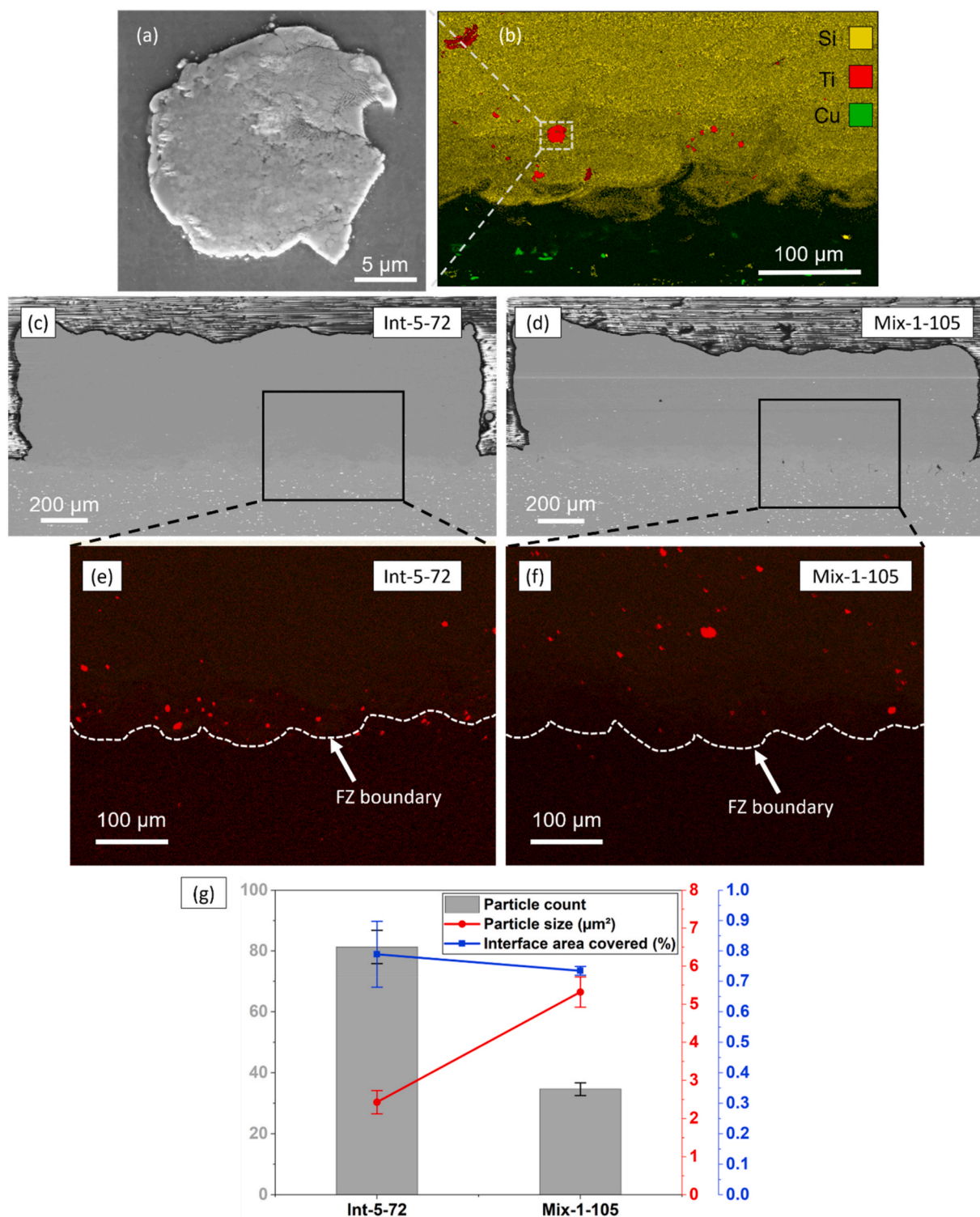


Fig. 7. (a) A sintered TiC microparticle measured at 10 kV at a magnification of 9kx using a secondary electron detector. (b) Low-magnification EDS image of the interface region of a Mix-1-105 cross-section, where the particle is observed. Similar sintered TiC microparticles (red) can be observed. (c) Slice of an Int-5-72 sample. (d) Slice of a Mix-1-105 sample. Both (c) and (d) are taken at 125 $\times$ magnification at 20 kV using an Angle-selective Backscatter detector. (e) EDS image showing Ti enrichment in a 500 $\times$ magnified region within the Int-5-72 sample. (f) EDS image showing Ti enrichment in a 500 $\times$ magnified region within the Mix-1-105 sample. (g) Agglomeration at the interface for the best-performing interlayer and blended approaches. Bars and points show mean values from $n = 4$ analysed interface regions; error bars denote standard deviation.

observed surrounding the particle, which suppresses the epitaxial growth of the substrate grain, thus effectively stopping the progress of the crack. It can be clearly observed that the crack progression is interrupted by the additional grain boundaries introduced through TiC-

induced grain nucleation. While EBSD analysis reveals the presence of a distinct grain surrounding the particle, EDS mapping confirms Ti enrichment at the particle location. A high-resolution EDS analysis and microscopy (Fig. 8e–j) of the particle show that the feature does not

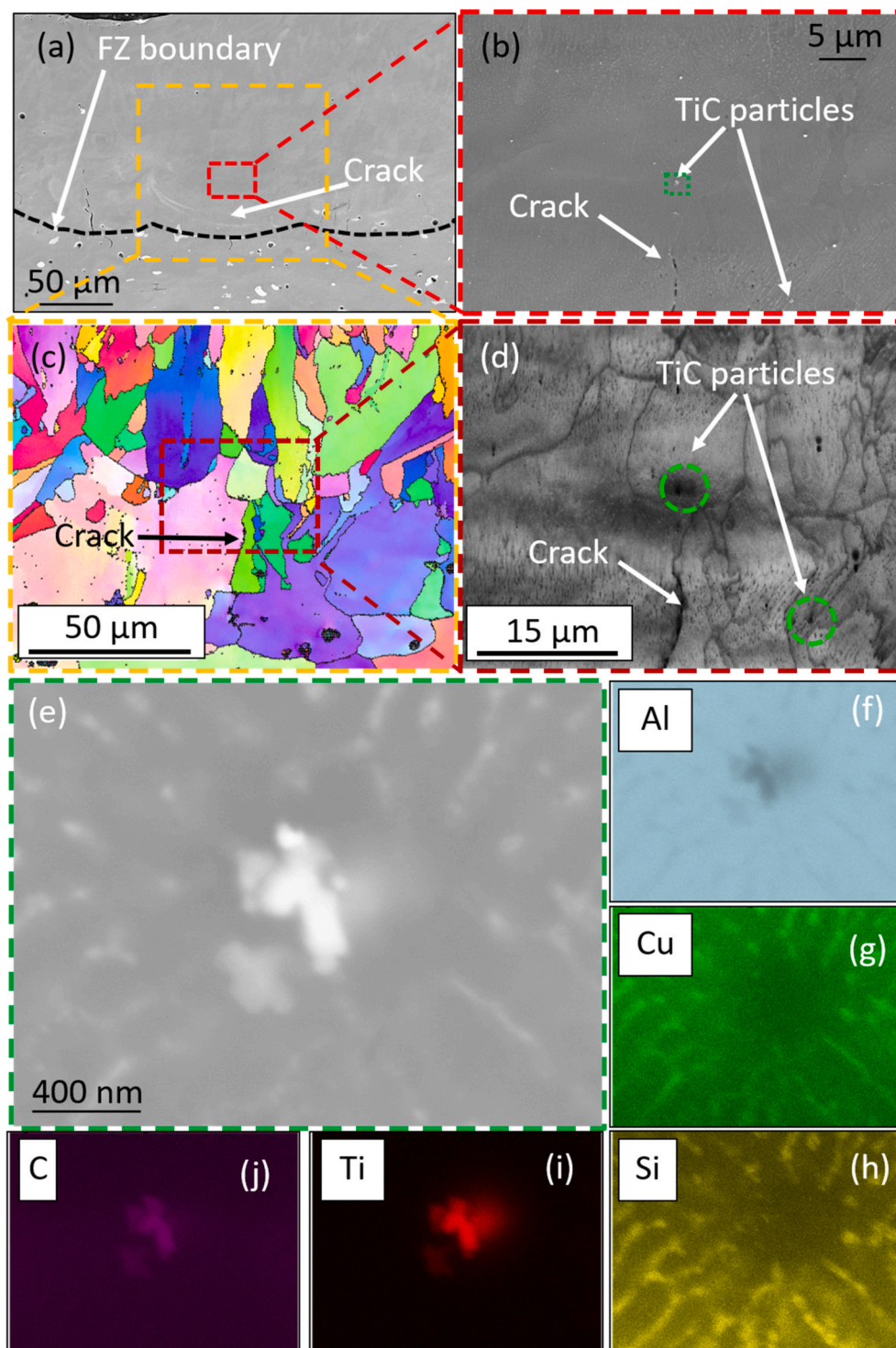


Fig. 8. Crack suppression by a TiC particle. (a) Low-magnification image of the interface. (b) The crack tip at the interface region micrograph with particles. (c) EBSD IPF map of the same region shown in (b). (d) TiC particles were observed within the grains in the grain boundary map. (e–j) High-magnification image and EDS mapping of the TiC.

correspond to a sintered TiC microparticle or to tri-aluminide precipitates, but instead consists of agglomerated TiC nanoparticles. Such nanoparticle agglomerates were predominantly observed in interlayer samples. In these cases, the agglomerates are located within the region corresponding to the AlSi10Mg layer deposited above the TiC interlayer and retain a nanoscale, clustered morphology rather than forming consolidated microparticles.

3.6. Growth restriction factor and interfacial nucleant density analysis

The growth restriction factor (Q) for α -Al grains was calculated for Mix-1-105 and Int-5-72 interface compositions (Table 4). Scheil simulation predicts compositions of both TiC-modified interfaces to solidify via primary Al_3Ti before α -Al nucleation (see Supplementary Section S7).

Int-5-72 and Mix-1-105 compositions resulted in Q values of 35.4 K

Table 4

Mean interface composition determined from 30 ROI averaging for both TiC-modified conditions.

Element (wt%)	Int-5-72	Mix-1-105
Al	93.25 ± 0.42	93.75 ± 0.46
Mg	1.27 ± 0.10	1.28 ± 0.06
Si	1.70 ± 0.82	1.59 ± 0.48
Cu	3.24 ± 0.40	3.12 ± 0.23
Ti	0.51 ± 0.18	0.25 ± 0.08

and 34.7 K, respectively (Table 5, Supplementary Fig. S7). The growth restriction factor is effectively invariant between the compositions within the combined uncertainty of the linear fits, with a spread of merely 2 %. At the same time, the volume fraction of primary Al₃Ti predicted at the onset of α-Al nucleation was ~2x higher in Int-5-72 than in Mix-1-105 (Table 5), similar to the ~2x difference in measured interface Ti content between the two conditions (Table 4).

Primary Al₃Ti could not be resolved by SEM-EDS even at high magnification in the present samples. The minimum particle diameter to act as free-growing nuclei in the system was estimated from the free growth criterion [38], as given in equation (2).

$$d = \frac{4\sigma}{\Delta S_v \cdot \Delta T_n} \quad (2)$$

where ΔT_n is the undercooling required for a particle of diameter d to begin free growth, here approximated by the growth restriction factor, which characterises the buildup of constitutional undercooling ahead of the interface [39]. σ is the solid-liquid interfacial energy, and ΔS_v is the volumetric entropy of fusion. Values for pure Al were used for both the solid-liquid interfacial energy ($\sigma = 0.10 \text{ J/m}^2$) [40] and the volumetric entropy of fusion ($\Delta S_v = 1.148 \times 10^6 \text{ J/m}^3\text{K}$) was calculated from the latent heat of fusion and melting point of pure Al in the absence of composition-specific data for the present alloy. Plugging in the mean Q value (35 K) for ΔT_n in equation (2) for the system results in critical diameter of ~10 nm. Previous studies [41,42] have reported L1₂-Al₃X primary phases of sizes of 2–15 nm in similar rapidly solidified Al alloy systems. Therefore, a working particle diameter d of 15 nm was adopted as a conservative estimate to calculate the Al₃Ti number density. Subsequently, the Al₃Ti number density ($n_0^{\text{Al}_3\text{Ti}}$) was calculated from the ratio of the total phase volume to the single-particle volume, resulting in $4.92 \times 10^{21} \text{ m}^{-3}$ for the interlayer and $2.34 \times 10^{21} \text{ m}^{-3}$ for blended cases. Additionally, the SEM-EDS-measured TiC particle population (Fig. 7g) was converted to an approximate volumetric number density using Fullman's stereological relation [43], leading to a TiC particle density (n_0^{TiC}) of $1.67 \times 10^{15} \text{ m}^{-3}$ for the interlayer and $5.50 \times 10^{14} \text{ m}^{-3}$ for blended cases. The combined nucleant density ($n_0 = n_0^{\text{Al}_3\text{Ti}} + n_0^{\text{TiC}}$) is therefore dominated by the Al₃Ti contribution in both TiC-containing conditions.

3.7. Sensitivity of interface defects to processing conditions: role of crack size

To further examine the origin of this crack-dominated behaviour, the sensitivity of defect size and count to processing energy was evaluated in detail for the 5 μm TiC interlayer (Fig. 9b). The pore size and count exhibited limited sensitivity to processing energy. In contrast, while the

crack count showed relatively low sensitivity, the crack size varied significantly with processing energy. The crack size is largest at the lowest energy density (Int-5-54), decreases markedly at intermediate energy densities (Int-5-72 and Int-5-90), and increases again at the highest energy density (Int-5-108). Although the variability is relatively large, as reflected by the error bars, the same non-monotonic trend is consistently observed. This behaviour is not observed for pores, whose average size remains comparatively low and varies only slightly across the processing window.

For the interlayer samples with a thickness of 5 μm, variations in processing energy also result in distinct changes in sintered microparticle characteristics (Fig. 9b). At the lowest processing energy (Int-5-54), the particle count of the particles was low, and the average size was high, indicating agglomeration/sintering. At 72 J/mm³ (Int-5-72), the particle count increases markedly, and the average particle size reaches its lowest value. With further increases in processing energy (Int-5-90 and Int-5-108), both the particle count and the TiC-covered interface area decrease, while the average particle size increases, indicating renewed particle coarsening.

3.8. Pore reduction by TiC modification

The spatial distribution of pores associated with TiC modification was assessed using two-dimensional microscopy images and quantitative analysis of pore area and count (Fig. 10). An increase in pore count is observed in the interface region (120–200 μm) for the optimal blended (Mix-1-105) and interlayer (Int-5-72) approaches (Fig. 10a). In contrast, the average pore size does not increase within the interface region (Fig. 10b), with pore sizes remaining relatively uniform across the fusion

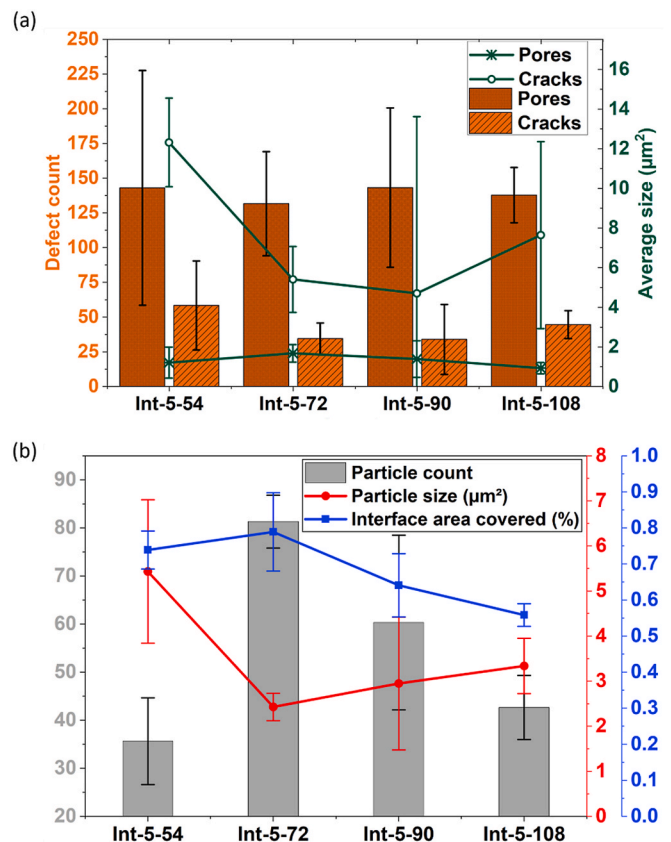


Fig. 9. (a) Variation of count and average size of pores and defects with respect to the 5 μm TiC interlayer. (b) Effect of laser energy on agglomeration behaviour in the interlayer approach at constant layer thickness. Bars and points show mean values from three analysed interface regions; error bars denote standard deviation.

Table 5

Growth restriction factor (Q) and linear fit quality for the interlayer and blended interface compositions.

	Int-5-72	Mix-1-105
Q	35.4 ± 0.2	34.7 ± 0.1
FCC onset f_s	~0.01	~0.005
Al ₃ Ti vol% at FCC onset	0.87	0.41

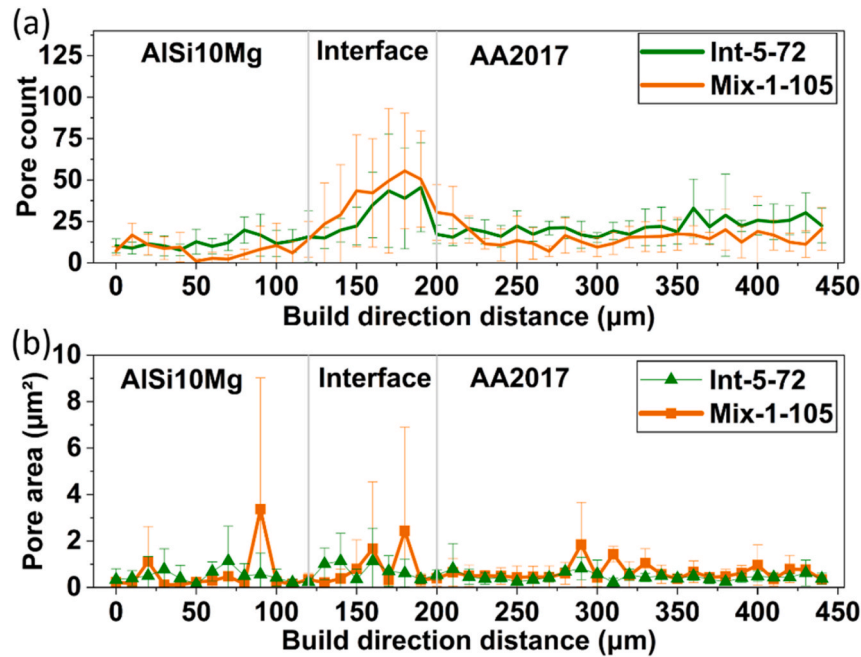


Fig. 10. (a) Variation in mean pore count across the interface for Reference, Int-5-72 and Mix-1-105 samples. (b) The corresponding mean pore diameter is measured for the samples shown in (a). Points show mean values from 3 analysed interface regions; error bars denote standard deviation.

zone and substrate for both processing approaches.

3.9. Effect of TiC modification on interface melt pool depth

The interface melt pool depth for the reference and TiC-modified samples is compared in Fig. 11. Under the respective optimised processing conditions (Fig. 11a), both TiC modification strategies resulted in slightly deeper melt pools than those of the reference condition. The TiC interlayer (Int-5-72) and blended (Mix-1-105) samples exhibited similar interface melt pool depths of approximately 81 μm, compared with approximately 75 μm for the reference sample. Furthermore, to isolate the influence of the TiC modification strategy from differences in processing parameters, additional samples processed at a constant laser power of 350 W were compared (Fig. 11b). Under these conditions, the

TiC interlayer sample (Int-1-86) exhibited the greatest interface melt pool depth (~80 μm), whereas the blended sample (Mix-1-88) exhibited the lowest melt pool depth (~65 μm), representing a reduction of approximately 19 % relative to the interlayer configuration.

3.10. Interface hardness

Fig. 12 compares the Vickers microhardness profiles measured across the substrate-AM interface for the reference case, Int-5-72, and Mix-1-105. The reference condition exhibited a reduction in hardness in the interface region (approximately 167 HV0.005). In contrast, the Int-5-72 condition showed a higher hardness near the interface than the reference condition, with a mean hardness of approximately 184 HV0.005. The Mix-1-105 condition displayed intermediate behaviour with a peak

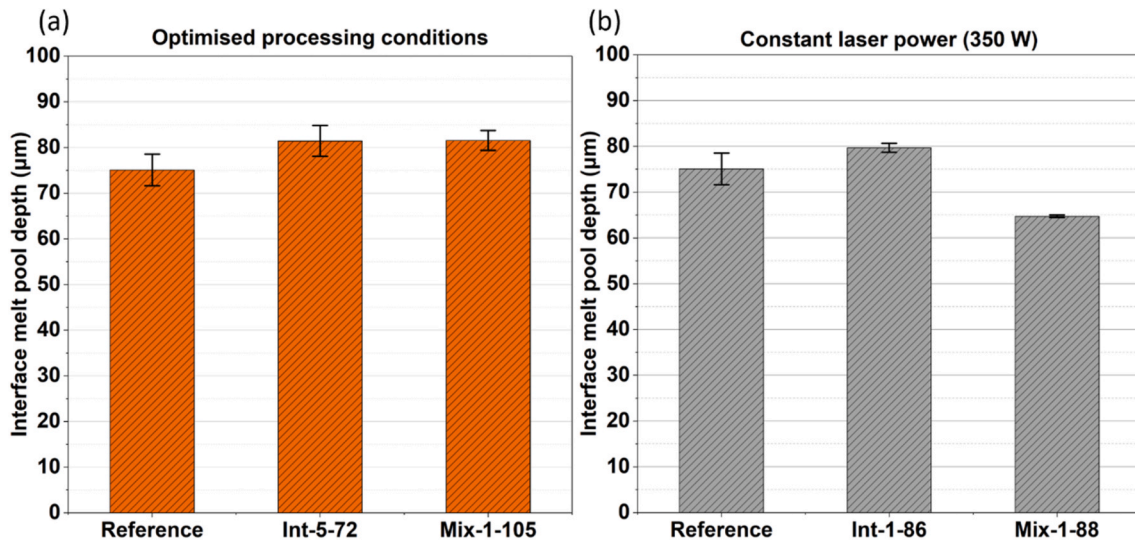


Fig. 11. Comparison of interface melt pool depth for (a) the reference and optimised TiC-modified conditions and (b) samples processed at a constant laser power of 350 W. Owing to differences in the TiC interlayer configuration, the corresponding nominal VED values differ slightly between conditions. Bars represent the mean values measured from three interface cross-sections (approximately 20 melt pool depth measurements per cross-section); error bars indicate one standard deviation.

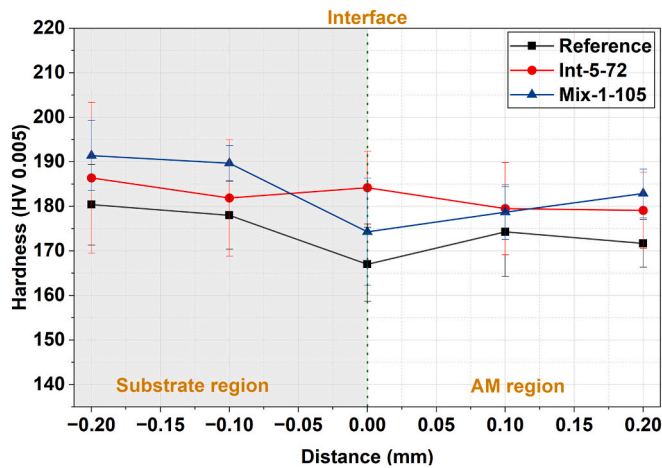


Fig. 12. Spatial evolution of Vickers microhardness (HV0.005) across the substrate-am interface. The interface position is defined as $x = 0$ mm. Hardness values represent the mean of ten measurements obtained from two different cross-sections, with error bars indicating ± 1 standard deviation.

hardness of approximately 174 HV0.005 at the interface.

3.11. Interface shear strength

Fig. 13 compares the nominal shear strength of the parent materials and hybrid interface configurations. AA2017 exhibited the highest shear strength, followed by AlSi10Mg. The hybrid interfaces showed lower average shear strength than both parent materials. The unmodified hybrid reference interface showed the lowest average shear strength and exhibited the largest variability among the investigated conditions. Among the TiC-modified conditions, the TiC interlayer (Int-5-72) exhibited higher median and mean shear strength than the TiC-blended feedstock condition (Mix-1-105). Relative to the reference interface, the TiC interlayer increased the median nominal shear strength from 178.9 MPa to 186.0 MPa while reducing the coefficient of variation from 7.6 % to 2.8 %. In contrast, the TiC-blended condition exhibited a median nominal shear strength (179.2 MPa) comparable to that of the reference interface, despite the reduced variability (7.6 % to 3.1 %). Both TiC-modified interfaces displayed lower scatter in measured strength

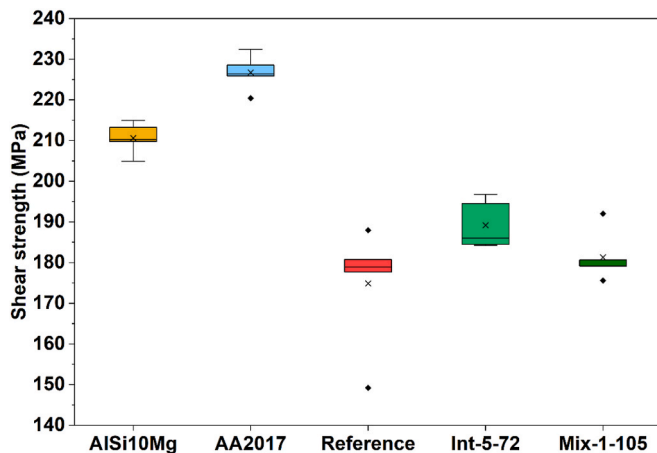


Fig. 13. Box plots showing the nominal shear strength of AlSi10Mg, AA2017, the unmodified AlSi10Mg/AA2017 hybrid interface (Reference), and best cases of TiC-modified interlayer (Int-5-72) and blended feedstock (Mix-1-105) interfaces. Boxes indicate the interquartile range, the central line represents the median, whiskers denote the minimum and maximum values excluding outliers, and symbols indicate the mean and outlier values. Five specimens were tested for each condition.

values, indicating a more consistent mechanical response.

Representative nominal shear stress–strain curves for all tested conditions are provided in the Supplementary Fig. S8. The parent materials exhibited highly reproducible responses, whereas the hybrid interface conditions reached lower maximum shear stresses and lower apparent strain before fracture. Fractographic examination of the reference specimens coupled with EDS analyses (Fig. 14) revealed a fracture occurring predominantly along the interface with limited traces of Si visible, indicating no propagation into the AlSi10Mg. In contrast, the TiC interlayer condition exhibited fracture initiation at the hybrid interface, followed by crack propagation into the AlSi10Mg. The TiC-blended condition exhibited a fracture morphology comparable to the reference interface, with fracture initiating and predominantly propagating through the interface-adjacent substrate region, with a negligible shift into AlSi10Mg towards the final stages. The mechanical implications of these observations are discussed in relation to the interface microstructure, grain refinement, and defect mitigation in Section 4.5.

4. Discussion

4.1. Grain refinement mechanisms due to TiC addition

Both the TiC-modification strategies employed in this study resulted in interfacial grain refinement. The invariance of Q across the blended and interlayer conditions (Table 5) indicates that solute-driven constitutional supercooling does not differ meaningfully with TiC delivery strategy, despite the $\sim 2\times$ difference in local interface Ti content between the interlayer and blended conditions. Together with the closely matched interface melt pool depths of the two conditions (Fig. 11), this indicates that neither compositional nor local thermal conditions differ significantly between the two TiC modifications.

According to the interdependence model by StJohn et al. [39], the grain size (d_{gs}) depends on three major contributions,

$$d_{gs} = x_{cs} + x'_{dl} + x_{sd} \quad (3)$$

x_{cs} is the distance a nucleated grain must grow to generate constitutional supercooling sufficient to nucleate the next grain. x'_{dl} is the distance from the solid–liquid interface to the point where this supercooling first becomes sufficient. Together, x_{cs} and x'_{dl} define the nucleation-free zone, within which no further nucleation can occur. The third, x_{sd} is the distance from the edge of this zone to the next most potent nucleant particle. The contributions from the three terms indicate whether the given system is limited by growth restriction, by solute diffusion, or by the availability of nucleant particles. The nucleation-free zone is governed by the growth restriction factor and the solidification conditions, which are effectively invariant between the interlayer and blended conditions, such that the entire predicted difference arises from x_{sd} and hence from nucleant particle density. The enhanced interfacial grain refinement achieved by the interlayer route relative to blending is therefore attributed to primary Al_3Ti nucleant particles [26,31,42, 44–46], and TiC nanoparticle agglomerates [47,48] that can act directly as heterogeneous nucleation sites for new α -Al grains, as observed in Fig. 8.

4.2. Crack reduction by refined interfacial grain structure

Together, the mechanisms mentioned in the previous section lead to pronounced grain refinement observed in the TiC-modified samples. Such refinement alters the solidification microstructure at the interface, which in turn reduces interfacial hot cracking (Figs. 3 and 5). Grain refinement reduces hot cracking susceptibility by reducing segregation, accommodating local strain, and maintaining adequate liquid films, thereby improving feeding along shortened grain boundaries [9,49–51].

Another important outcome of the present work is the reduction in liquation cracking associated with interfacial grain refinement. While

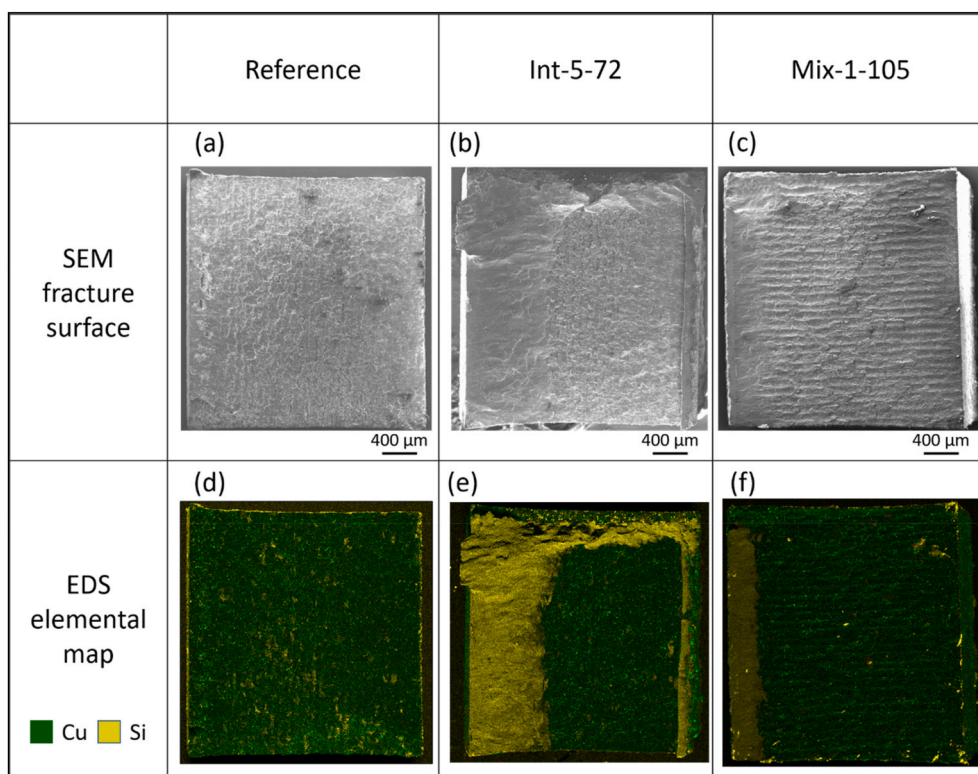


Fig. 14. Representative SEM fracture surfaces ((a)–(c)) and corresponding Cu–Si EDS elemental maps ((d)–(f)) of the shear-tested hybrid interfaces for the Reference, TiC interlayer (Int-5-72), and TiC-blended (Mix-1-105) conditions. The fracture morphology and elemental distribution illustrate the differences in the fracture path among the investigated interface conditions.

grain refinement within the FZ alone is known to reduce solidification cracking, the TiC modifications employed here also lead to a reduction in liquation cracking within the HAZ. The refinement of FZ grains limits the formation and propagation of cracked FZ grain boundaries, reducing the likelihood of crack extension into partially melted HAZ grain boundaries [20]. Additionally, the presence of TiC at or near the interface modifies the local thermal conditions during processing [52], reducing heat transfer into the substrate and limiting localised melting within the HAZ. The influence of TiC on local thermal conditions is detailed in Section 4.3.

Although both TiC-modification strategies reduce cracking, the interlayer approach is more effective, achieving a reduction exceeding 90 % compared to approximately 85 % for the blended approach. This difference arises from the spatial distribution of TiC. The interlayer configuration enables targeted TiC delivery to the bottom region of the first melt pool, where crack initiation is most critical, whereas the blended approach distributes TiC throughout the AM region and is unable to deliver comparable Ti concentrations specifically at the interface. As a result, the interlayer strategy provides more effective suppression of crack formation and improved interface integrity.

4.3. Agglomeration and sintering

The final microstructure of all modified interfaces evaluated in this study resulted in an Al-rich matrix containing TiC nano-agglomerates and TiC sintered microparticles (Fig. 7). Agglomeration behaviour of ceramic nanoparticles, such as TiC, is primarily driven by the high surface energy [48]. Independent areal surface roughness (S_a) measurements performed across three independent substrates with and without TiC addition after sand-blasting showed a 5–8-fold increase in local S_a variability following TiC addition, indicating a spatially non-uniform, locally agglomerated particle distribution. Furthermore, Brown–Forsythe test from ANOVA also indicated spatially non-uniform,

agglomerated coverage of the TiC nanopowder layer rather than a continuous film. Such agglomerates, when exposed to the laser path, result in sintering and a transformation into individual microparticles [31,52].

An increased number of sintered microparticles was observed in the blended sample compared to the interlayer samples (Fig. 7). In the blended approach, ex-situ mixing promotes both agglomeration between TiC nanoparticles and agglomeration of TiC on the surfaces of AlSi10Mg powder particles (Fig. 15). Additionally, an increase in the number and size of TiC microparticles was observed with increasing TiC blend ratio (supplementary section S4). In contrast, the absence of a blending stage in the interlayer approach reduces initial agglomeration, and the overlying AlSi10Mg powder layer (Fig. 1) partially shields TiC particles from direct laser irradiation, limiting sintering.

Differences in melt pool depths under identical processing conditions for both interlayer and blended TiC modifications (Fig. 11b) reinforce the argument that the blended approach leads to increased formation of

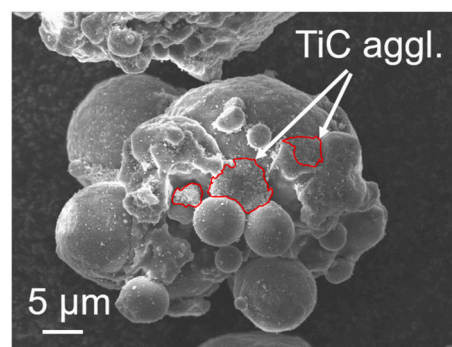


Fig. 15. Agglomeration of TiC nanoparticles forming shells on AlSi10Mg particles in 1 wt% blended powder.

sintered microparticles. The reduction in melt pool depth exhibited by the TiC-blended sample relative to the interlayer configuration indicates that a reduced portion of energy is transferred into the substrate in the blended approach, while the rest of the input energy is used within the powder bed, potentially for the heating and sintering of TiC agglomerates. The processing energy further influences the sintering behaviour of TiC particles, where increasing energy density beyond the optimum leads to a reduction in sintered particle count and an increase in average particle size, indicating particle coarsening (Fig. 9).

The coarsening of TiC particles is expected to reduce particle dispersibility within the melt pool [53], thereby decreasing the availability of effective heterogeneous nucleation sites and consequently the grain refinement, leading to reduced interface integrity. This is reinforced by the stronger local grain refinement and reduced defect density in interlayer samples, which have a higher fraction of smaller TiC agglomerates (Fig. 8). In contrast, fewer but larger sintered particles in blended samples result in less uniform grain refinement and higher defect densities. This trend was very consistent, where samples with larger TiC sintered particles exhibited increased interfacial defects (see Supplementary Fig. S4). The above observations suggest that, in addition to the TiC addition strategy, the effectiveness of grain refinement depends on maintaining a fine and well-dispersed particle population throughout processing.

4.4. Effect of TiC modification on pore evolution

The pore characteristics observed at the interface indicate that TiC modification alters pore evolution behaviour. Although the pore count increases near the interface for both blended and interlayer samples (Fig. 10) compared to the pure AM and substrate regions of the sample, no corresponding increase in pore size is detected. This behaviour is in contrast to the reference case reported in our previous work [20], where significant pore coalescence resulted in a pronounced increase in both pore number and size at the interface. This observation suggests that, although pores continue to form, their growth through coalescence is affected by the presence of TiC.

The melt pools of the best-performing TiC-modified samples are deeper (~5 μm) than the reference case (Fig. 11). A larger melt pool is expected to result in interfaces with increased pore count and size due to the melting of a larger substrate volume, which releases its inherent porosity into the melt pool [20,54]. However, no increase in pore size could be detected for TiC-modified samples (Fig. 10), indicating a variation in the pore evolution mechanism. This suppression of pore size increase in TiC-modified samples is likely facilitated by more grain boundaries offered by finer grains and TiC particles at the interface that restrict bubble mobility within the melt pool, thereby limiting pore coalescence and growth [55,56].

4.5. Influence of TiC-induced microstructural modification on mechanical properties

The microstructural modifications introduced by TiC addition directly influence the mechanical response of the hybrid interface. Tensile specimens fractured within the AlSi10Mg irrespective of interfacial condition, and therefore, miniature shear testing was the only available means of characterising the interface directly. Under shear loading, the reference condition interface fails at much lower loads than both the substrate and the AM-produced parts, indicating that the interface is the weakest area of the hybrid specimen (Fig. 12), due to its high defect density (Fig. 5) and epitaxially grown coarse grains (Fig. 6). This is consistent with the previously reported defective interface region with coarsened grains exhibiting softening [18]. Both TiC modification strategies mitigate this interface softening behaviour to some extent (Fig. 12).

The behaviour of spatial variation of hardness across different interface types is consistent with the size, spatial distribution and

consequential microstructural changes generated due to TiC particles. The observed increase in interface hardness can be primarily attributed to grain boundary strengthening [27,29,57] arising from the localised grain refinement revealed by EBSD (Fig. 6). In addition, the substantial reduction in defect density (Fig. 3) and contributions from particle-related strengthening associated with retained TiC particles (Fig. 7) and possible precipitated phases are expected to contribute to the measured local mechanical response. However, the present work does not distinguish between the individual particle- and precipitate-related-strengthening mechanisms.

In addition to the improvement in hardness, TiC-modified interfaces resulted in improved shear strengthening of the interface (Fig. 13). The reference condition exhibited the highest variability (CV = 7.6 %) in the case of shear strength, indicating less consistent mechanical behaviour, suggesting greater interfacial heterogeneity than the TiC-modified conditions. The reduced variability exhibited by TiC-interlayer (2.8 %) and blended conditions (3.1 %) was consistent with the respective reductions in the defect ratios and grain refinement.

Another interesting observation was the difference in the crack progression across the specimen. The progression of fracture from the substrate-adjacent region at the early stages of shearing towards the AlSi10Mg during the latter stages (Fig. 14b–e) indicates that the TiC-modified strengthened interlayer interface no longer represents the mechanically weakest region of the hybrid structure compared to AlSi10Mg, which offers an easier path for the crack to propagate. This is consistent with the initial tensile test results, which clearly indicated that AM-deposited AlSi10Mg is a favourable region for crack propagation within the hybrid specimens. Additionally, finer interface grains within the TiC interlayer-interface lead to increased grain boundary density, which may act as a Hall-Petch barrier against dislocation movement [33,34]. Together, a less defective interface along with the above-mentioned effects contributes to a more consistent interfacial mechanical response and increases the resistance to crack propagation along the interface during shear loading. Future work should quantify the relationship between TiC particle distribution, grain refinement, and local mechanical properties to further distinguish the individual strengthening contributions. Since the tensile specimens behaved differently, the shear improvements reported here would not be expected to translate into higher tensile strength or ductility, which are governed by the AlSi10Mg rather than by the interface. Fatigue characterisation was not carried out in this study, and the discussion is therefore limited to the improvement in interfacial shear behaviour.

Beyond its role in grain refinement, it must be noted that the introduction of TiC could also impact the long-term stability of the interface. The dissolution of TiC that supplies Ti for Al_3Ti formation also releases carbon to the region in stoichiometric proportion. The approximately two-fold higher dissolved Ti at the interlayer interface (0.51 vs 0.25 wt %) therefore implies a correspondingly higher dissolved carbon content and a greater potential for Al_4C_3 formation. The Al_4C_3 phase readily hydrolyses by atmospheric moisture and can initiate localised degradation. The finer, more widely dispersed particle population of the interlayer condition acts in the same direction by presenting a greater TiC-Al interfacial area. No Al_4C_3 was observed, and no exposure testing was performed in the present work. However, with the current understanding, it can be hypothesised that the interlayer configuration could be more susceptible to such degradation compared to the blended approach.

5. Conclusions

This study investigated feedstock material-based interface modification strategies using TiC nanoparticles to improve the interface integrity between AA2017 substrates and PBF-LB/M-deposited AlSi10Mg. Two TiC delivery approaches were investigated and compared: (i) blending TiC nanoparticles with AlSi10Mg powder and (ii) introducing TiC nanopowder as an interlayer between the substrate and

the AlSi10Mg powder. Major conclusions from the study can be summarised as follows:

- (1) Both TiC modification strategies improved interface integrity compared to the unmodified reference condition. Under optimised parameters, the interlayer approach achieved a total defect reduction of 86.6 %, with crack-related defects reduced by over 90 %, while the blended approach achieved a reduction of 77.2 % in the case of all defects and ~84 % in crack-related defects.
- (2) The superior performance of the TiC interlayer approach arises from enhanced TiC delivery at the interface. Along with a less agglomerated, fine TiC particle distribution, Scheil simulations predicted an approximately two-fold higher primary Al_3Ti fraction at the interlayer interface. The higher nucleant particle density at the interface led to interface-focused grain refinement and thereby enabled more effective suppression of interfacial defects.
- (3) Thermodynamic analysis showed the growth restriction factor to be effectively invariant, and measured melt pool depths were closely matched across interlayer and blended conditions, excluding both solute-driven growth restriction and local thermal conditions as the origin of the grain refinement difference.
- (4) The crack reduction in TiC-modified samples was achieved mainly through restricting epitaxial growth by interfacial grain refinement. Additionally, this finer-grain structure presumably hindered pore mobility and thereby limited pore coalescence and growth, resulting in lower mean pore size and count for both TiC modifications.
- (5) All TiC-modified samples contained sintered TiC microparticles, whose size and spatial distribution depend on the TiC delivery strategy and processing conditions. Finer, more uniformly distributed particles are associated with improved interfacial grain refinement and reduced defect formation. The interlayer approach limits agglomeration and sintering compared to the blended approach, resulting in a more effective particle distribution at the interface.
- (6) The improved interface integrity obtained through the TiC interlayer approach translated directly into enhanced mechanical performance. Localised TiC delivery suppressed interface softening, increased the interfacial shear strength, and reduced the variability in shear response, demonstrating that targeted interface modification provides a more mechanically reliable hybrid interface than homogeneous TiC distribution through powder blending.

CRedit authorship contribution statement

T.M. Nandakishor: Writing – original draft, Visualisation, Validation, Methodology, Investigation, Software, Data curation, Conceptualisation. V. Budinski: Investigation, Software, Data curation. P. Hegele: Software, Data curation. E.A. Jägle: Writing – review & editing, Supervision, Methodology, Funding acquisition. T. Stoll: Writing – review & editing, Supervision.

Data availability

Data will be made available on request.

Declaration of generative AI and AI-assisted technologies in the writing process

During the preparation of this work, the authors used generative AI and AI-assisted technologies developed by **OpenAI** (ChatGPT) to enhance the clarity and readability of the language. These tools were used to optimise and rephrase the text. After using this tool/service, the authors reviewed and edited the content as needed and take full

responsibility for the content of the publication.

Declaration of competing interest

The authors declare the following financial interests/personal relationships which may be considered as potential competing interests: T. M. Nandakishor reports financial support was provided by Munich Aerospace eV. T. M. Nandakishor reports a relationship with Munich Aerospace eV that includes: funding grants. If there are other authors, they 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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Appendix A. Supplementary data

Supplementary data to this article can be found online at <https://doi.org/10.1016/j.jmrt.2026.10.022>.

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