

Article

Interfacial Stress Development and Cracking Susceptibility during Laser Powder Bed Fusion of Random TiB₂-Particle-Reinforced AlSi10Mg Matrix Composites

Chenglong Ma ^{1,2,*} , Ziyu Fang ^{3,*} , Meiping Wu ^{1,2} , Xiaojin Miao ^{1,2} and Quanlong Wang ^{1,2}

¹ Jiangsu Key Laboratory of Advanced Food Manufacturing Equipment & Technology, School of Mechanical Engineering, Jiangnan University, Wuxi 214122, China

² Jiangsu Province Engineering Research Center of Micro-Nano Additive and Subtractive Manufacturing, Wuxi 214122, China

³ Sino-French Engineer School, Nanjing University of Science and Technology, Nanjing 210094, China

* Correspondence: chenglongma@jiangnan.edu.cn (C.M.); ziyufang@njust.edu.cn (Z.F.)

Abstract: A sequentially coupled multi-phase thermo-mechanical model for laser powder bed fusion (LPBF) of Al-based composites reinforced by 1 wt.% random TiB₂ micron particles was established. Due to the remarkable difference in thermophysical properties, the maximum thermal stress was predicted at the TiB₂/Al matrix interface and formed at the liquid–solid transition stage. Meanwhile, complicated evolution curves of temperature, strain, and strain rate were predicted with the laser moving time during the solidification stage. To evaluate the interface cracking susceptibilities of micron-TiB₂/AlSi10Mg composites, the flow stress of the matrix was calculated, instead of ultimate tensile strength, based on the physical constitutive relationship. From the comparison between the calculated flow stress and the simulated Von Mises equivalent stress, it was found that an increase in TiB₂ particle size was inclined to induce a larger interfacial stress than the calculated flow stress, therefore increasing the interfacial crack tendency, which was also effectively verified by the experimental results.

Keywords: laser powder bed fusion; Al-based composites; interface crack; simulation and modeling



Citation: Ma, C.; Fang, Z.; Wu, M.; Miao, X.; Wang, Q. Interfacial Stress Development and Cracking Susceptibility during Laser Powder Bed Fusion of Random TiB₂-Particle-Reinforced AlSi10Mg Matrix Composites. *Metals* **2023**, *13*, 1405. <https://doi.org/10.3390/met13081405>

Academic Editor: Miguel Cervera

Received: 12 July 2023

Revised: 29 July 2023

Accepted: 4 August 2023

Published: 6 August 2023



Copyright: © 2023 by the authors. Licensee MDPI, Basel, Switzerland. This article is an open access article distributed under the terms and conditions of the Creative Commons Attribution (CC BY) license (<https://creativecommons.org/licenses/by/4.0/>).

1. Introduction

Aluminum matrix composites (AMCs) are a class of promising lightweight structural materials, owing to their low density, high specific strength/stiffness, and low thermal expansion coefficient, and have achieved wide applications in aerospace and automotive fields [1–3]. However, as advanced design concepts, including generative design, topology design, or lightweight design, prevail in the structure design of AMC components [4], the conventional processes are confronted with an increasing challenge due to the significant gap between the design method and the manufacturing technology. Laser powder bed fusion (LPBF), as a mainstream metal-based 3D printing technology that possesses a significant advantage in the shaping of complex components with any geometrical configuration, can well eliminate the above-mentioned gap [5].

Recently, LPBF-processed high-performance AMCs with various reinforcement phases (SiC [6], Fe₂O₃ [7], TiB₂ [8], B₄C [9], ZrO₂ [10], etc.) have been reported widely. For example, Y. Chen et al. [11] studied the microstructural features, mechanical properties, and porosity of 10 vol% oxidized microscale SiC-particle-reinforced AlSi10Mg composites fabricated by LPBF. Results indicated that a transition layer of MgAl₂O₄ was present around the SiC particles and the relative density and ultimate tensile strength (UTS) reached 98.89% and 405.3 MPa, respectively. C.A. Biffi et al. [8] explored the application of pulsed wave laser emission mode in the LPBF process of TiB₂/AlCu composites. They found that the as-built microstructure was characterized by fine equiaxed grains and an even distribution of TiB₂ particles, contributing to an UTS of 395 MPa and an elongation of 13%. G.

Marchese et al. [12] prepared 0.5% MgAl_2O_4 -nanoparticle-reinforced AlSi10Mg composites using the LPBF process. The results showed that the nanocomposite samples presented larger cellular structures and slightly higher ultimate tensile strength than AlSi10Mg samples. Obviously, ceramic-particle-reinforced AMCs have attracted the most attention among these investigations, due to their broadly tunable mechanical or thermal properties and low cost. To obtain superior mechanical or thermal properties, strong interface bonding between the ceramic particle and the metal matrix is the key. However, normally, a remarkable difference in the thermophysical properties of the metal matrix and ceramic particle exists, which is very prone to inducing considerable interfacial stress. Improper material characteristics or LPBF processing parameters can further give rise to the formation of interfacial cracks [13]. Apart from the thermophysical parameter mismatch, the non-equilibrium metallurgical behavior during LPBF also plays an important role in the initiation of interfacial cracks. The extremely rapid heating/cooling rate ($10^{4-6} \text{ }^\circ\text{C/s}$) and ultrahigh temperature gradient ($10^{5-8} \text{ }^\circ\text{C/m}$) within the mesoscale transient molten pool are involved in this metallurgical process [14], thereby leading to a complicated and volatile stress field. Understanding the formation and development of the stress field at the ceramic/metal interface during LPBF is important, and can help to build the crucial approaches to suppress interfacial cracking. Nevertheless, it is almost impossible to study the development of the interfacial stress field during LPBF at the microscale via experimental methods. To date, computational numerical modeling has been used widely to provide deep insights into the physical metallurgical mechanisms during LPBF for the purpose of quality control [15]. Q. Han et al. [16] established a Lagrangian discrete phase model to simulate the dispersion behavior of added sub-micrometer-sized TiC particles within the molten pool. The predictions suggested that the migration of TiC particles was primarily induced by the combination of recoil pressure and Marangoni convection force. G. Meng et al. [17] constructed a pore defect distribution model based on the experimental results to realize the dynamic simulation of the temperature and stress fields in laser additive manufacturing of Inconel 718 alloy with different pore defects. However, to the best of our knowledge, few numerical investigations on the interfacial stress development and interfacial cracking behavior of LPBF-processed AMCs have been conducted.

In this work, we focused on the interfacial stress evolution during LPBF of TiB_2 /AlSi10Mg composites. In addition, the effects of TiB_2 particle size on the interfacial stress and cracking susceptibility of the LPBF-processed TiB_2 /AlSi10Mg composites were studied and are discussed herein. The in-depth understanding of the interfacial stress development, as well as the role of the reinforcement particle size in influencing cracking susceptibility, can be expected to provide guidance for the material processing performance-integrated controlling principle of LPBF-fabricated AMCs.

2. Materials and Methods

2.1. Experiment Investigation

Spherical gas atomized AlSi10Mg powder with an average particle diameter of $26.85 \text{ }\mu\text{m}$ and 1 wt.% TiB_2 particles with a mean particle size of $\sim 5.5 \text{ }\mu\text{m}$ were used in this work. Two batches of bulk samples (corresponding to AlSi10Mg and TiB_2 /AlSi10Mg composites) with a size of $10 \text{ mm} \times 10 \text{ mm} \times 5 \text{ mm}$ and a batch of tensile samples (TiB_2 /AlSi10Mg) were fabricated by LPBF using the following processing parameters: laser power P of 350–450 W, scanning speed v of 1500–3000 mm/s, hatch spacing h of 50 μm , layer thickness d of 50 μm , and island scanning strategy, as shown in Figure 1a–c. Based on the Archimedes method, the relative density of these bulk samples was assessed and the optimized parameters were basically determined as $P = 350 \text{ W}$ and $v = 1500 \text{ mm/s}$. In this case, the highest relative densities of AlSi10Mg and TiB_2 /AlSi10Mg samples were 98.51% and 98.72%, respectively. The molten pool morphologies of the LPBF-processed samples were observed by a DM-2700M optical microscope (OM). The microstructure was characterized by a transmission electron microscope (TEM) using a Tecnai G2 F20 S-TWIN (operated at 200 kV) (FEI. Co., Hillsboro, USA). The tensile experiment was conducted on

a CMT5205 testing machine (MTS Industrial Systems, Shanghai, China) in displacement control mode with a cross-head velocity of 1 mm/min at room temperature. The tensile sample had a full length of 70 mm, a width of 6 mm, and a thickness of 4.3 mm, and the detailed morphology can be seen in Figure 1d.

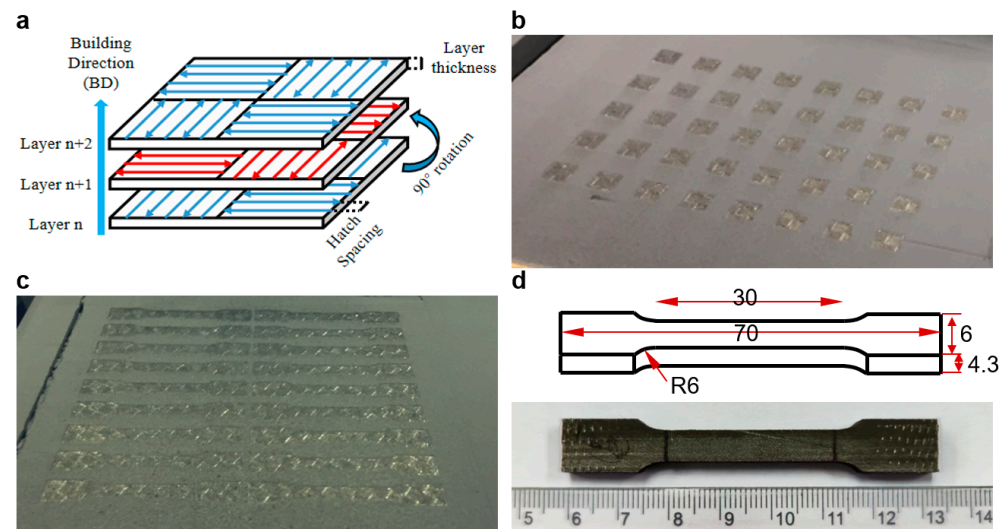


Figure 1. The LPBF process of bulk and tensile samples: (a) as-used scanning strategy; (b) as-fabricated bulk samples; (c) as-fabricated tensile samples; (d) the shape and size of a tensile sample.

2.2. Modeling Approach

For the $\text{TiB}_2/\text{AlSi10Mg}$ composite powder, TiB_2 particles with higher laser absorptivity can lead to a considerable enhancement in laser energy input into the powder bed. Then, the fusion of AlSi10Mg matrix powder occurs firstly as the temperature of the powder bed is raised to the melting point of AlSi10Mg , after the laser irradiation with high energy density. Due to high melting point of TiB_2 , its fusion behavior is confined to the edge zone of TiB_2 particles. Within the transit molten pool formed by the fusion of the matrix powder, TiB_2 particles are rearranged under the effect of the Marangoni flow. As the laser beam moves forward, the molten pool is cooled rapidly with a speed of $10^{3-5} \text{ }^\circ\text{C/s}$ and then the solidification follows, with the formation of ultrafine grains and non-equilibrium phases.

In consideration of the complexity of the LPBF process and the diversity/variability of powder material properties, some physical assumptions had to be made, in order to ensure the laser–powder interaction was mathematically tractable and the transit thermodynamic characteristics could be described quantitatively. These assumptions included the following: (1) the powder bed consisted of the AlSi10Mg matrix material and TiB_2 ceramic particles; each TiB_2 ceramic particle was assumed to be a regular sphere and the rest was filled with AlSi10Mg , which was isotropic and continuous; (2) the heat flux from the laser beam followed the Gaussian distribution and the laser absorptivity of the powder bed was considered to be a constant; (3) the radiation of laser energy was neglected; (4) the coefficient of convection between the powder bed and the surrounding environment was set as a constant of $12.5 \text{ W/(m}^2\cdot^\circ\text{C)}$; (5) the effect of argon gas on the metallurgical behavior was neglected; and (6) the evaporation behavior and recoil pressure within the laser-induced molten pool were not considered in this model. The physical principle of the LPBF process of particle-reinforced metal matrix composites has been reported heavily, and is thus not repeated here. Then, a sequentially coupled multi-phase thermo-mechanical model was established based on the above assumptions, to give more insight into the thermodynamics characteristics and stress distribution/evolution at the $\text{TiB}_2/\text{AlSi10Mg}$ interface. The nonlinear relationships of material thermophysical properties with temperature were considered in this model, as shown in Figure 2, including thermal conductivity κ , special heat capacity C_p , elasticity modulus E , and coefficient of thermal expansion α of AlSi10Mg .

and TiB_2 [18,19]. Figure 3a shows the thermo-mechanical coupling model of the LPBF-processed $\text{TiB}_2/\text{AlSi10Mg}$ composite. In this model, the powder bed size was set as $300 \times 100 \times 50 \mu\text{m}^3$, and contained AlSi10Mg matrix material with a loose density of 1.340 g/cm^3 and randomly generated TiB_2 sphere particles with an average diameter of $5.66 \mu\text{m}$. The size and position of TiB_2 particles were determined by a random function. The density of TiB_2 particles was 4.52 g/cm^3 . By calculation, the weight fraction of TiB_2 was 0.96 wt.%, which was basically consistent with the experimental parameter.

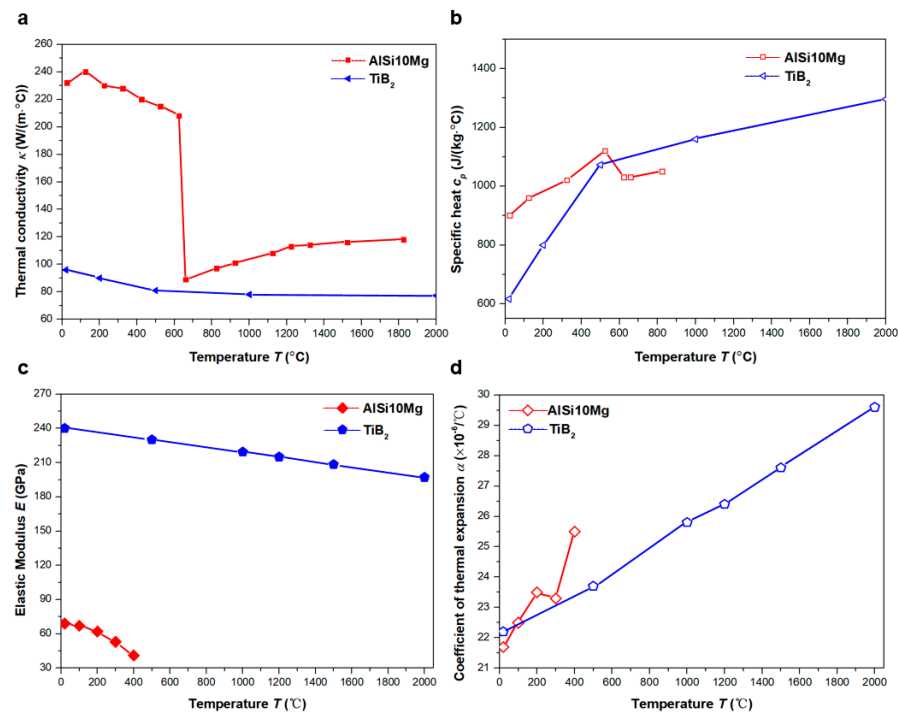


Figure 2. The thermophysical parameters of AlSi10Mg and TiB_2 at different temperatures: (a) thermal conductivity; (b) special heat capacity; (c) elasticity modulus; (d) coefficient of thermal expansion.

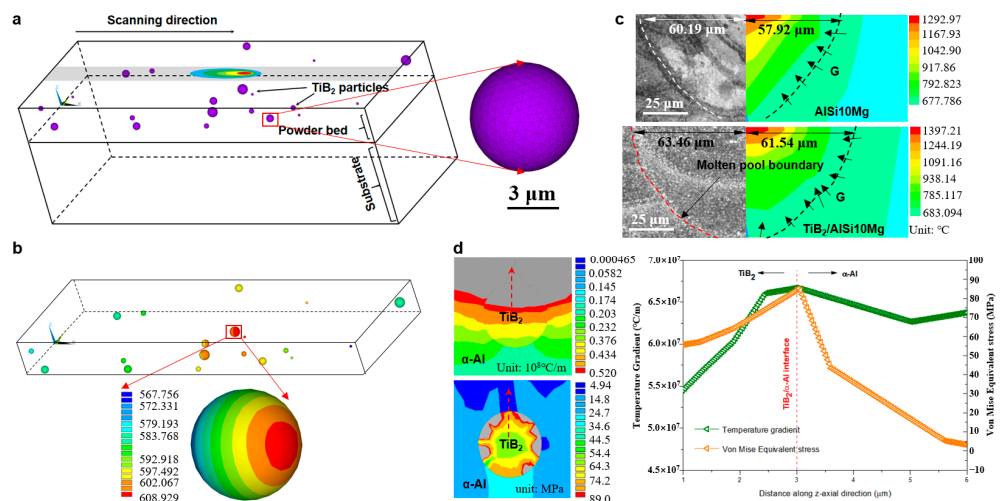


Figure 3. The finite element simulation of the temperature field during the LPBF of $\text{TiB}_2/\text{AlSi10Mg}$ composites: (a) the multi-phase thermo-mechanical model; (b) the temperature contours within the randomly distributed TiB_2 particles; (c) the cross-sectional temperature profiles of the molten pools with and without TiB_2 particles; (d) the temperature gradient and stress distribution features for the cases with and without TiB_2 particles.

With regard to the heat source model, the laser beam was modeled as a volume heat source with a Gaussian energy distribution, which could be described by the following equation [20]:

$$q(x, y, z) = \frac{3c_s AP}{\pi H(1 - \frac{1}{e^3})} \exp\left(-\frac{3c_s}{\log(\frac{H}{z})}(x^2 + y^2)\right) \quad (1)$$

where P is the laser power, A is the laser absorptivity of powder material, H is the height of the heat source, c_s is the shape coefficient, equal to $3/(R_0)^2$, and R_0 is the laser beam radius (here, 70 μm).

During the LPBF process, the transient spatial temperature distribution $T(x, y, z, t)$ satisfied the differential equation:

$$\rho \frac{\partial(C_p(T) \times T)}{\partial t} = \frac{\partial}{\partial x}(\kappa(T) \times \frac{\partial T}{\partial x}) + \frac{\partial}{\partial y}(\kappa(T) \times \frac{\partial T}{\partial y}) + \frac{\partial}{\partial z}(\kappa(T) \times \frac{\partial T}{\partial z}) \quad (2)$$

where ρ is the density of powder material. Subsequently, the thermal stress analysis was performed based on the predicted temperature data of all nodes and the thermal-stress coupling relationship. In the thermal-stress coupling model, the following equilibrium equation exists within any element:

$$\{dF\}^e + \{dR\}^e = [K]^e \{d\delta\}^e \quad (3)$$

$$\{dR\}^e = \int [B]^T \{C\} dT dV, [K]^e = \int [B]^T [D] [B] dV \quad (4)$$

where $\{dF\}^e$ is the nodal force increment vector of a mesh-cell, $\{dR\}^e$ is the thermal-induced equivalent nodal force increment vector, $\{d\delta\}^e$ is the nodal displacement increment vector, $[K]^e$ is the stiffness matrix of element, $[B]$ represents the matrix related to the element strain vector, $\{C\}$ represents the vector related to temperature, and $[D]$ is the elastic or elastoplastic matrix. By integration of the stiffness matrix of all elements and all nodal force vectors, the equilibrium equation of the whole model could be obtained:

$$\sum (\{dF\}^e + \{dR\}^e) = \sum \{dR\}^e = \sum ([K]^e) \{d\delta\}^e \quad (5)$$

Note that, in consideration of no loading force being imposed during LPBF, the stress field of the powder bed and substrate were in self-balance, as a result of which $\sum \{dF\}^e$ was equal to zero. In addition, the thermal elastoplastic stress could be known from the following equation:

$$\{d\sigma\} = [D] \{d\varepsilon\} - \{C\} dT \quad (6)$$

$$\{d\varepsilon\}^e = [B] \{d\delta\}^e \quad (7)$$

where $\{d\varepsilon\}^e$ is the strain increment. By solving the above equations, the stress or strain evolution process and deformation behavior of all nodes during the LPBF of the $\text{TiB}_2/\text{AlSi10Mg}$ composite powder bed could be obtained.

3. Results and Discussion

Figure 3b shows the temperature profiles of randomly generated TiB_2 particles during the LPBF of $\text{TiB}_2/\text{AlSi10Mg}$ composites. Furthermore, to demonstrate the perturbation effect of TiB_2 particles on the molten pool, the temperature distribution contours with and without TiB_2 particles are displayed in Figure 3c. Meanwhile, to verify the simulated results, the morphologies of the molten pools in the as-fabricated samples were further demonstrated. It was seen that, regardless of the use of AlSi10Mg or $\text{TiB}_2/\text{AlSi10Mg}$, the predicted molten pool sizes were similar to the measured values, thus validating the reliability of the simulated results. Furthermore, the predicted molten pool exhibited a remarkably extended geometrical configuration with apparently higher T_{max} in the case

with TiB_2 , by comparison with the case without TiB_2 , due to the relatively higher laser absorptivity and lower thermal conductivity of TiB_2 . It was also predicted that the addition of TiB_2 particles could significantly influence the direction of the temperature gradient G , and was thus expected to interrupt the continuously epitaxial growth along the building direction (BD) and facilitate the columnar-to-equiaxed transition. Figure 3d demonstrates the G distribution and stress distribution contours as well as the corresponding evolution curves along the red arrow direction. It was seen that the TiB_2 particle disturbed the G distribution in the surrounding matrix. Additionally, the maximum G and maximum Von Mises equivalent stress were predicted at the $\text{TiB}_2/\text{AlSi10Mg}$ interface.

Figure 4a shows the heat flow distribution within the molten pool during the LPBF of $\text{TiB}_2/\text{AlSi10Mg}$ composites. Due to the lower thermal conductivity, the heat transfer within the TiB_2 particles was apparently suppressed, thereby causing the heat accumulation and resultant high temperature gradient at the particle/matrix interface. To quantitatively describe the interfacial thermal behavior, some characteristic points located at the interface were chosen, including top pole point (TP), bottom pole point (BP), left pole point (LP), and right pole point (RP), as shown in the inserted schematic in Figure 4b. Then, the temperature–time curves of these characteristic points and their first derivatives were plotted (Figure 4b). Three typical stages, namely, the pre-heating stage, melting stage, and solidification stage could be recognized clearly. Both the $T_{\max}$ and highest cooling rate $R_{\max}$ were predicted at TP, reaching 1037.72°C and $7.79 \times 10^7^\circ\text{C/s}$, respectively. Furthermore, based on the thermo-mechanical coupling, the evolution of the Von Mises equivalent stress σ_{eq} of these characteristic points with the irradiation time were predicted (Figure 4c). In the pre-heating stage, as the laser beam moved forward, the matrix around TiB_2 particles was gradually heated, thus inducing an expansion of the matrix material, in which case the interfacial stress formed due to the mutual constraint between the matrix and TiB_2 particle. When the matrix around the TiB_2 particle was fused completely (corresponding to the melting stage), the interfacial stress declined rapidly. In the solidification stage, owing to the $R_{\max}$ existing at TP, the predicted maximum σ_{eq} also emerged at TP, reaching 323.74 MPa , which was much higher than that at other characteristic points. Thus, among the studied positions, TP was the most prone to becoming a crack source. In fact, any position at the interface possibly became a crack source, in consideration of the particle rotation induced by the Marangoni flow [21]. Here, to simplify the complicated LPBF process, we made an assumption that the particle was fixed. Figure 4d further shows the evolution of the elastic strain ε_e and plastic strain ε_p at TP with the irradiation time. It shows that ε_e dominated in the pre-heating stage, while considerable ε_p formed in the solidification stage, due to the large contraction stress and shrinkage stress induced by the ultrahigh cooling rate.

To determine whether the interface crack occurs or there is susceptibility to cracking at the interface during the simulation, the general criterion is to see whether the condition ' $UTS < \sigma_{eq}$ ' is met, where UTS is the ultimate tensile strength at ambient temperature. Nevertheless, this criterion would lead to a narrower process window or material optimization range, in consideration of the strong dependence of UTS on T and $\dot{\varepsilon}$ (the UTS at ambient temperature is usually the largest). According to the results displayed in Figure 4b–d, LPBF involves a non-equilibrium solidification process with an ultrahigh cooling rate and complicated elastic–plastic strain evolution, consequently leading to a nonlinear change in UTS with the solidification time. Due to the absence of the relationship among UTS , T , and $\dot{\varepsilon}$, the plastic flow stress σ_s is paid more attention, and is defined as the yield limit at a certain T , ε , and $\dot{\varepsilon}$. In the case of the Al-based alloy, K. Mercier et al. developed the physical constitutive relationship of a commercial Al-Mg-Si alloy. It could be expressed by the following equation [22,23]:

$$\sigma_s = \sigma_a + \left\{ 1 - \left[\frac{kT}{\mu b^3 g_{0i}} \ln\left(\frac{\dot{\varepsilon}_0}{\dot{\varepsilon}}\right) \right]^{2/3} \right\}^2 \frac{\hat{\sigma}_i \mu}{\mu_0} \quad (8)$$

where k is the Boltzmann constant, b is the Burgers vector (0.286 nm), μ is the temperature-dependent shear modulus ($\mu(T) = \mu_0 - 3440/(\exp(215/T) - 1)$, MPa), g_{0i} is the normalized activation energy (1.18 here [24]), $\dot{\epsilon}_0$ is a constant that equals $\rho_D \cdot b \cdot l \cdot v_0$, where ρ_D denotes the dislocation density, l is the distance between adjacent obstacles on the slip plane, v_0 is the atom vibrational frequency (10^{11} /s), and $\hat{\sigma}_i$ represents the mechanical threshold associated with the interaction of glide dislocations with precipitates and solute atoms. σ_a is the athermal stress dependent on the grain size, which can be described by the Hall–Petch relationship:

$$\sigma_a = \sigma_0 + \hat{k}d^{-1/2} \quad (9)$$

where σ_0 is the stress due to initial defects and $\hat{k}$ is a constant ($50 \text{ MPa} \cdot \mu\text{m}^{1/2}$ [25]). To estimate σ_a , ρ_D , and l of the LPBF-processed $\text{TiB}_2/\text{AlSi10Mg}$ composite in this work, the corresponding TEM characterization was performed, as shown in Figure 5. It was observed that the average grain size was around $0.3 \mu\text{m}$ (Figure 5a), which is far smaller than that in the conventional Al alloy. M. Zamani et al. found that σ_a was only 30.32 MPa by calculation for the as-cast Al–Si alloy with the initial grain size of 1.32 mm [26]. Hence, σ_a in this work could be estimated as 120.23 MPa according to Equation (9). For ρ_D , by measuring and calculating the dislocation line length per unit area (Figure 5a), it could be estimated as $1.89 \times 10^{15}/\text{m}^2$. Due to the cyclic thermal behavior during the LPBF, a large number of secondary Si particles precipitated from the supersaturated $\alpha\text{-Al}$ matrix (Figure 5b), with an orientation relationship of $[0-11]_{\text{Si}} // [-110]_{\text{Al}}$. These secondary Si particles could be regarded as the obstacles on the main slip plane $\{111\}_{\text{Al}}$ of the matrix. By measurement, l was estimated as 6.12 nm. Therefore, $\dot{\epsilon}_0$ was equal to $3.3 \times 10^8/\text{s}$. Furthermore, the tensile test was carried out (Figure 5c) at ambient temperature (298 K). The true stress–strain curve was obtained via the following equations: $\sigma_T = \sigma(\epsilon + 1)$ and $\epsilon_T = \ln(\epsilon + 1)$. The yield strength and the corresponding strain rate were measured to be $\sim 245 \text{ MPa}$ and $2.5 \times 10^{-4}/\text{s}$, respectively. According to Equation (8), $\hat{\sigma}_i$ could be calculated as 285.34 MPa. Hence, the flow stress σ_s for the LPBF-processed $\text{TiB}_2/\text{AlSi10Mg}$ composite in this work could be reformulated as follows:

$$\sigma_s = 120.23 + \left\{ 1 - \left[\frac{0.50T}{\mu} \ln \left(\frac{3.3 \times 10^8}{\dot{\epsilon}} \right) \right]^{2/3} \right\} \frac{\mu}{100.98} (\text{MPa}) \quad (10)$$

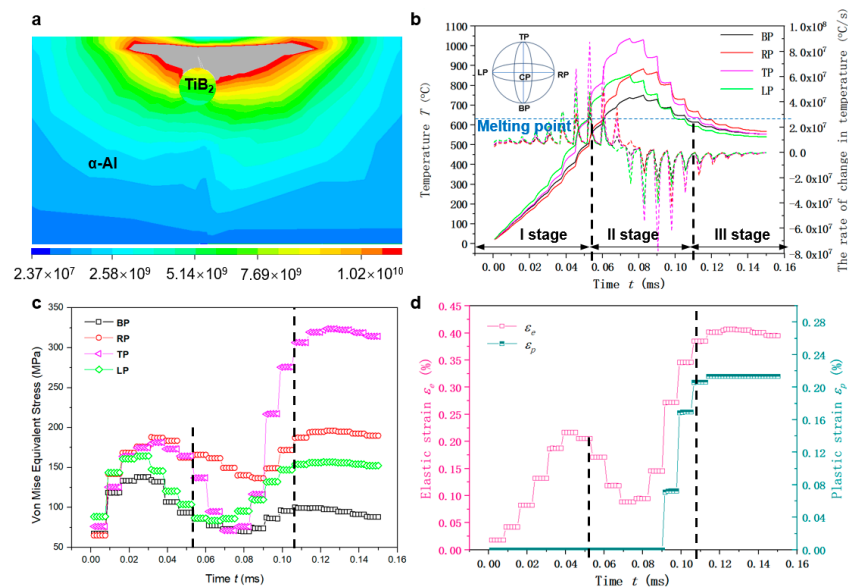


Figure 4. The changes of the thermal parameters and thermal-related mechanical parameters of the selected points with time: (a) the heat flow distribution contour; (b) the temperature–time curves and their first derivative; (c) the change in σ_{eq} with time; (d) the changes in ϵ_e and ϵ_p with time.

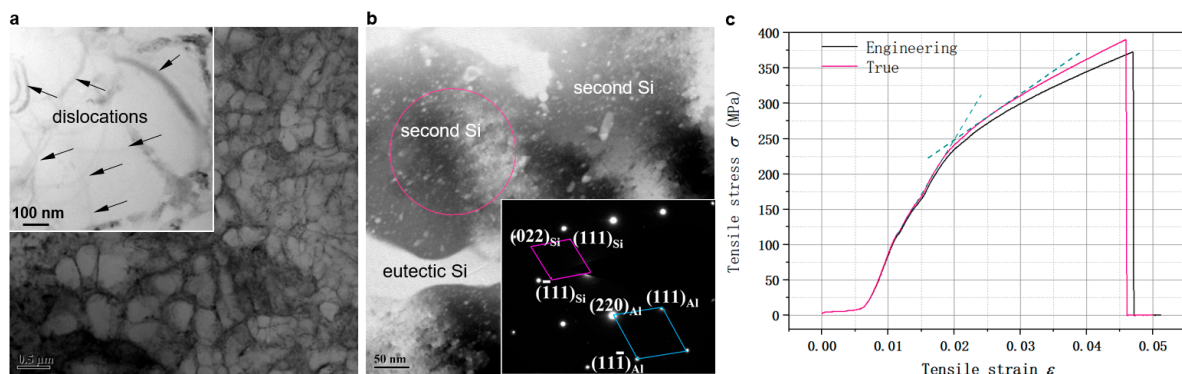


Figure 5. Microstructure and tensile curve of the LPBF-processed $\text{TiB}_2/\text{AlSi10Mg}$ composite: (a) TEM bright image showing the fine cellular α -Al grains and considerable dislocations (as the arrows pointed); (b) the tiny nanoscale secondary Si particles within the α -Al grain (the inserted image is the selective electron diffraction pattern of the region enclosed by the red circle); (c) the tensile stress–strain curve.

Figure 6 depicts the effects of TiB_2 particle size on interfacial temperature evolution, strain rate change, and stress development. It was predicted that the interface position of the larger particles tended to experience higher temperature and strain rate during the melting stage, but relatively smaller strain rate during the solidification stage (Figure 6a,b). According to Equation (10), the corresponding flow stress in the case with the larger particles was thereby lower. On the other hand, the increase in particle size caused more heat flow to be accumulated in the interface, thus inducing a larger interface temperature gradient and resultant interfacial stress. Figure 6c shows the surface stress contour of TiB_2 particles with a diameter of 6 μm , and that the maximum surface stress reached 361 MPa. Furthermore, Figure 6d–f compares the simulated Von Mises equivalent stress and the calculated flow stress for three cases with different particle sizes. It was found that the gap between the simulated value and the calculated value apparently became smaller during the solidification stage with the increase in the particle size. In particular, when the particle size was increased to 6 μm , the calculated flow stress was apparently lower than the simulated Von Mises equivalent stress, which meant the interfacial cracks might occur. To verify whether the method used to describe the cracking susceptibility of the $\text{TiB}_2/\text{AlSi10Mg}$ composite during the LPBF process was accurate, the corresponding experimental characterization was performed. As shown in Figure 7a,b, it was observed that cracks indeed only formed at the interface between the matrix and the large particles ($>3 \mu\text{m}$). Figure 7c further displays the X-component stress profile of the surrounding matrix of the TiB_2 particle. The predicted results indicate that considerable compressive stress formed at two sides of the particle along the scanning direction, while high tensile stress emerged at two sides of the particle along the building direction. Because the tensile stress is more prone to inducing the formation of cracks, the top and bottom positions of the particle could be therefore regarded to be the crucial regions. This was also basically consistent with the experimental observation (Figure 7a). Regarding the formation mechanism of compressive and tensile stress, the corresponding schematic is shown in Figure 7d. Due to the difference in the thermophysical properties between the matrix and the particle, considerable heat flow accumulated at the top region of the particle. As a result, the solidification stage at the top region lagged behind that at the other regions (Figure 4b). Due to the contraction stress effect, the top region suffered from the tensile stress while the other regions mostly experienced compressive stress due to the mechanical restraint of the particle.

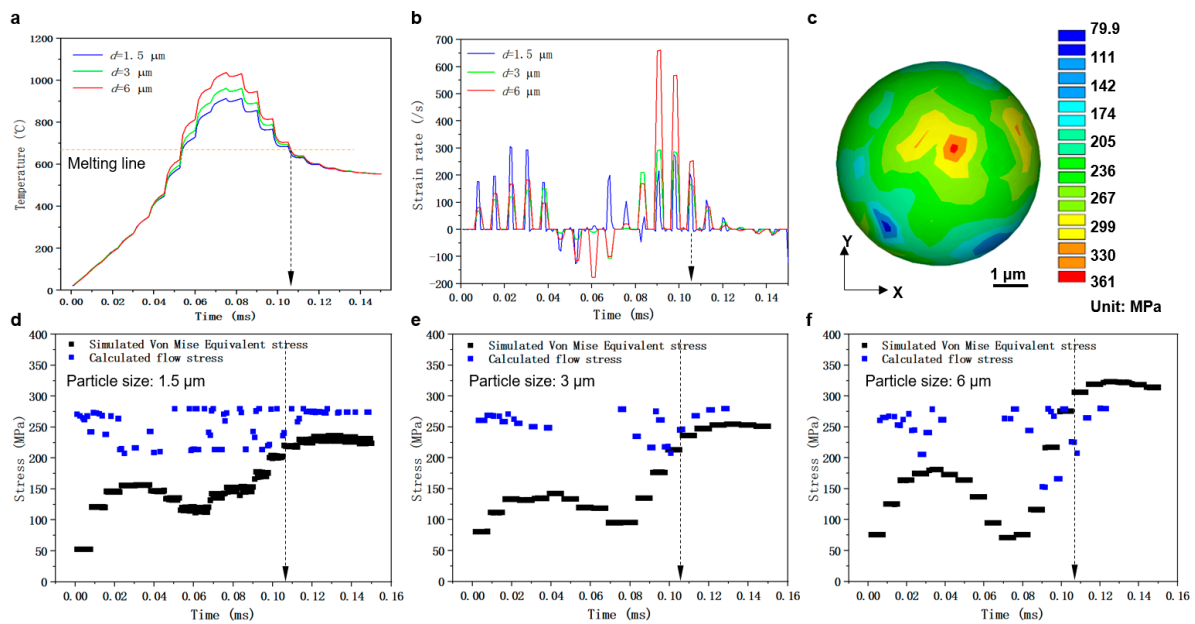


Figure 6. The effect of particle size on the thermal stress development and interface cracking susceptibility: (a) the temperature and its change rate evolutions at the TPs of TiB_2 particles with different sizes; (b) the strain rate evolutions at the TPs of TiB_2 particles with different sizes; (c) the stress contour of TiB_2 particles with a size of $6 \mu\text{m}$; (d–f) comparison between the calculated flow stress and simulated Von Mises equivalent stress for three cases with different particle sizes. The arrow pointed to the moment when the solidification period starts.

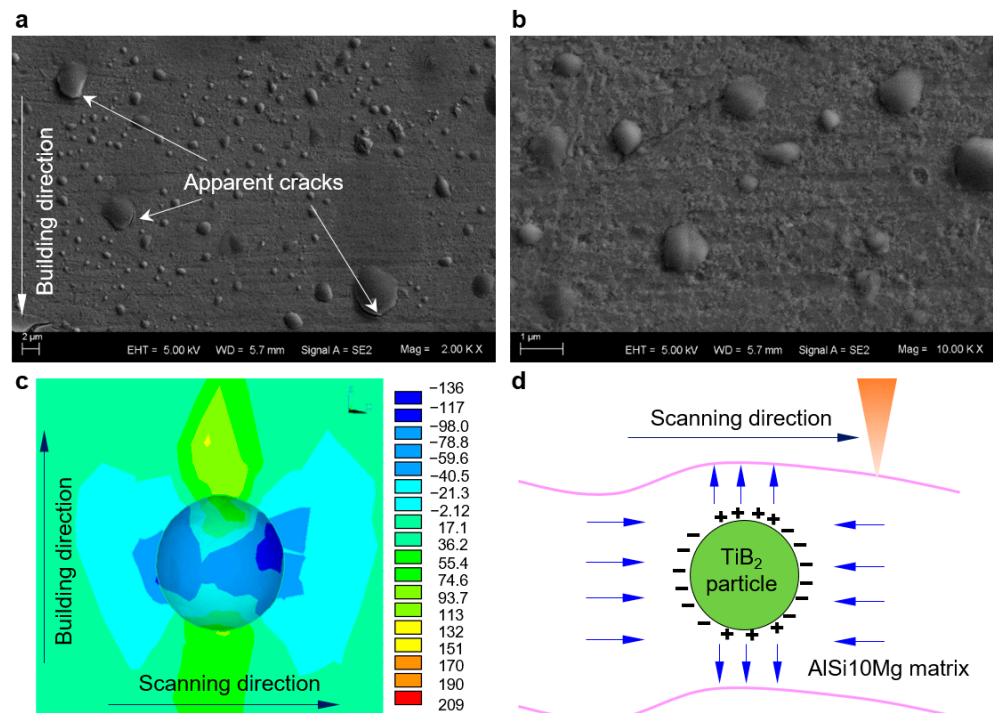


Figure 7. (a,b) The SEM images showing the interface features of TiB_2 particles with different sizes; (c) the X-component stress contour of the matrix surrounding the particle (unit: MPa); (d) the schematic for the formation of compressive and tensile stress.

4. Conclusions

A sequentially coupled multi-phase thermo-mechanical model was established. It was predicted that the addition of TiB₂ particles contributed to an increase in the molten pool temperature and size, and to a more chaotic temperature gradient distribution that would be able to facilitate the columnar-to-equiaxed transition. Furthermore, the maximum thermal stress was predicted at the TiB₂/Al matrix interface and formed at the liquid–solid transition stage. The matrix surrounding the particle showed considerable compressive stress along the scanning direction and remarkable tensile stress along the building direction. To determine the cracking susceptibility of the TiB₂/Al interface, the flow stress was then calculated based on the physical constitutive relationship. It was found that the calculated flow stress was higher than the simulated Von Mises equivalent stress for TiB₂ particles with a relatively smaller size (<3 µm), while a lower flow stress was obtained for larger particles (i.e., 6 µm), which was further verified by the experimental results. This work can assist in determining the optimized particle size range of the reinforcement phase and the corresponding processing windows for the LPBF of particle-reinforced metal matrix composites.

Author Contributions: C.M.: Conceptualization, Writing—Original Draft, Funding acquisition. Z.F.: Investigation, Methodology, Writing—Reviewing and Editing. M.W.: Supervision, Writing—Reviewing and Editing. X.M.: Visualization, Investigation. Q.W.: Investigation, Writing—Reviewing and Editing. All authors have read and agreed to the published version of the manuscript.

Funding: This work was supported by the Fundamental Research Funds for the Central Universities (No. JUSRP122028).

Data Availability Statement: The data presented in this study are available on request from the corresponding author.

Acknowledgments: We sincerely appreciate the supported by the Fundamental Research Funds for the Central Universities (No. JUSRP122028). Besides, we also thank Director Nengwu Zhang of the mechanical experiment center for kind help in the tensile tests.

Conflicts of Interest: The authors declare no conflict of interest.

References

1. Ramnath, B.V.; Parswajinan, C.; Dharmaseelan, R.; Thileepan, K.; Krishna, K.N. A review on aluminium metal matrix composites. *Mater. Today Proc.* **2021**, *46*, 4341–4343. [\[CrossRef\]](#)
2. Gill, R.S.; Samra, P.S.; Kumar, A. Effect of different types of reinforcement on tribological properties of aluminium metal matrix composites (MMCs)—A review of recent studies. *Mater. Today Proc.* **2022**, *56*, 3094–3101. [\[CrossRef\]](#)
3. Imran, M.; Khan, A.A. Characterization of Al-7075 metal matrix composites: A review. *J. Mater. Res. Technol.* **2019**, *8*, 3347–3356. [\[CrossRef\]](#)
4. Liu, M.; Zhan, J.; Zhu, B.; Zhang, X. Topology optimization of flexure hinges with a prescribed compliance matrix based on the adaptive spring model and stress constraint. *Precis. Eng.* **2021**, *72*, 397–408. [\[CrossRef\]](#)
5. Dejene, N.D.; Lemu, H.G. Current Status and Challenges of Powder Bed Fusion-based Metal Additive Manufacturing: Literature Review. *Metals* **2023**, *13*, 424.
6. Liu, X.; Zhuo, L.; Wang, Z.; Zhao, Z.; Zhang, H.; Yuan, B.; An, Z.; Li, C.; Yin, E.; Xu, T. Effects of processing parameters and post-process heat treatment on selective laser melted SiC/AlSi₁₀Mg composites. *Mater. Lett.* **2022**, *308*, 131254. [\[CrossRef\]](#)
7. Shi, Q.; Mertens, R.; Dadbakhsh, S.; Li, G.; Yang, S. In-situ formation of particle reinforced Aluminium matrix composites by laser powder bed fusion of Fe₂O₃/AlSi₁₂ powder mixture using laser melting/remelting strategy. *J. Mater. Process Technol.* **2022**, *299*, 117357.
8. Biffi, C.A.; Bassani, P.; Fiocchi, J.; Albu, M.; Tuissi, A. Selective laser melting of AlCu-TiB₂ alloy using pulsed wave laser emission mode: Processability, microstructure and mechanical properties. *Mater. Des.* **2021**, *204*, 109628.
9. Prabhakar, S.; Radhika, N.; Raghu, R. Analysis of Tribological Behavior of Aluminium/B₄C Composite Under Dry Sliding Motion. *Procedia Eng.* **2014**, *97*, 994–1003.
10. Sun, M.; Yang, Z.; Zhang, J.; Zhang, S.; Yang, Q.; Song, S.; Lu, B. Effects of ZrO₂ nanoparticles on the microstructure and mechanical properties of ZrO₂/AlSi₁₀Mg composites manufactured by laser powder bed fusion. *Ceram. Int.* **2023**, *49*, 19673–19681.
11. Chen, Y.; Ren, Y.; Li, K.; Dang, B.; Jian, Z. Laser powder bed fusion of oxidized microscale SiC-particle-reinforced AlSi₁₀Mg matrix composites: Microstructure, porosity, and mechanical properties. *Mater. Sci. Eng. A* **2023**, *870*, 144860.

12. Marchese, G.; Aversa, A.; Lorusso, M.; Manfredi, D.; Calignano, F.; Lombardi, M.; Biamino, S.; Pavese, M. Development and Characterisation of Aluminium Matrix Nanocomposites AlSi₁₀Mg/MgAl₂O₄ by Laser Powder Bed Fusion. *Metals* **2018**, *8*, 175. [\[CrossRef\]](#)
13. Lee, C.; Park, H.; Yoo, J.; Lee, C.; Woo, W.; Park, S. Residual stress and crack initiation in laser clad composite layer with Co-based alloy and WC + NiCr. *Appl. Surf. Sci.* **2015**, *345*, 286–294. [\[CrossRef\]](#)
14. DebRoy, T.; Wei, H.L.; Zuback, J.S.; Mukherjee, T.; Elmer, J.W.; Milewski, J.O.; Beese, A.M.; Wilson-Heid, A.; De, A.; Zhang, W. Additive manufacturing of metallic components—Process, structure and properties. *Prog. Mater. Sci.* **2018**, *92*, 112–224.
15. Chowdhury, S.; Yadaiah, N.; Prakash, C.; Ramakrishna, S.; Dixit, S.; Gupta, L.R.; Buddhi, D. Laser powder bed fusion: A state-of-the-art review of the technology, materials, properties & defects, and numerical modelling. *J. Mater. Res. Technol.* **2022**, *20*, 2109–2172. [\[CrossRef\]](#)
16. Han, Q.; Low, K.W.; Gu, Y.; Wang, X.; Wang, L.; Song, B.; Huang, C.; Setchi, R. The dynamics of reinforced particle migration in laser powder bed fusion of Ni-based composite. *Powder Technol.* **2021**, *394*, 714–723. [\[CrossRef\]](#)
17. Meng, G.; Zhang, J.; Li, J.; Jiang, Z.; Gong, Y.; Zhao, J. Impact of pore defects on laser additive manufacturing of Inconel 718 alloy based on a novel finite element model: Thermal and stress evaluation. *Opt. Laser Technol.* **2023**, *167*, 109782. [\[CrossRef\]](#)
18. Touloukian, Y.S.; Buyco, E.H. *Thermophysical Properties of Matter: Specific Heat-Metallic Elements and Alloy*, 1st ed.; IFI/Plenum: New York, NY, USA, 1970.
19. Munro, R.G. Material properties of titanium diboride. *J. Res. Natl. Inst. Stand. Technol.* **2000**, *105*, 709–720. [\[CrossRef\]](#) [\[PubMed\]](#)
20. Bag, S.; Trivedi, A.; De, A. Development of a finite element based heat transfer model for conduction mode laser spot welding process using an adaptive volumetric heat source. *Int. J. Therm. Sci.* **2009**, *48*, 1923–1931. [\[CrossRef\]](#)
21. Hansel, A.; Spittel, T. *Kraft und Arbeitsbedarf bildsamer Formgebungsverfahren*, 1st ed.; Deutscher Verlag für Grund-Stoffindustrie: Leipzig, Germany, 1978.
22. Le Mercier, K.; Guerin, J.-D.; Dubar, M.; Dubar, L.; Puchi-Cabrera, E. Physically-based constitutive description of a commercial Al-Mg-Si alloy deformed under cold-warm working conditions. *J. Alloys Compd.* **2019**, *790*, 1177–1191. [\[CrossRef\]](#)
23. Gao, C.; Zhang, L. A constitutive model for dynamic plasticity of FCC metals. *Mater. Sci. Eng. A* **2010**, *527*, 3138–3143. [\[CrossRef\]](#)
24. Newman, M.L.; Robinson, B.J.; Sehitoglu, H.; Dantzig, J.A. Deformation, residual stress, and constitutive relations for quenched W319 aluminum. *Met. Mater. Trans. A* **2003**, *34*, 1483–1491. [\[CrossRef\]](#)
25. Gutierrez-Urrutia, I.; Muñoz-Morris, M.A.; Morris, D.G. Contribution of microstructural parameters to strengthening in an ultrafine-grained Al–7% Si alloy processed by severe deformation. *Acta Mater.* **2007**, *55*, 1319–1330.
26. Zamani, M.; Dini, H.; Svoboda, A.; Lindgren, L.-E.; Seifeddine, S.; Andersson, N.-E.; Jarfors, A.E. A dislocation density based constitutive model for as-cast Al-Si alloys: Effect of temperature and microstructure. *Int. J. Mech. Sci.* **2017**, *121*, 164–170. [\[CrossRef\]](#)

Disclaimer/Publisher’s Note: The statements, opinions and data contained in all publications are solely those of the individual author(s) and contributor(s) and not of MDPI and/or the editor(s). MDPI and/or the editor(s) disclaim responsibility for any injury to people or property resulting from any ideas, methods, instructions or products referred to in the content.