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Laser powder bed fusion of high-strength crack-free Al7075 alloy with the in-situ formation of TiB₂/Al₃Ti-reinforced phases and nucleation agents

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Abstract

The existence of solidification cracks caused by columnar grains in precipitation-hardened aluminium alloys limit the applicability of Al7075 components manufactured via laser powder bed fusion (LPBF) additive manufacturing. A novel approach was developed to co-incorporate submicron-sized B and micron-grade Ti6Al4V to eliminate hot cracks and to effectively transform coarse columnar grains into fine equiaxed grains, thus improving the mechanical performance of LPBF-fabricated modified Al7075 material. The grain refinement was mainly attributable to the heterogeneous nucleation promoted by the combination of in-situ-formed L1₂-Al₃Ti and TiB₂ nano-sized phases. After an optimised T6 heat treatment, excellent comprehensive mechanical properties were achieved, with a tensile strength of 460 MPa and

an elongation of 13%. This research provides an efficient and cost-effective path for addressing crack-sensitive metallic materials used for LPBF additive manufacturing processes.

Keywords: laser powder bed fusion; additive manufacturing; Al7075 alloy; cracking suppression; in-situ reaction

1. Introduction

The laser powder bed fusion (LPBF) process, also referred to as selective laser melting (SLM), is a type of metal additive manufacturing [1,2]. LPBF employs a focused, high-energy laser beam that is used to selectively melt and consolidate thin layers of metal powder, enabling the direct fabrication of components from computer-aided design (CAD) models [3,4]. This technology finds extensive applications across various sectors, including aerospace, defense, automotive manufacturing and biomedical engineering [5]. The advantages of this technology include a high degree of customization, rapid production, and decreased waste and energy consumption. High-strength aluminum alloy denotes materials characterized by exceptional strength and hardness, primarily comprising alloys from the second and seventh series [6,7]. Such materials are used in diverse fields, including aircraft fuselage structures, automobile body parts, bicycle frames and marine components, among others [8,9]. Conventional processing methods for high-strength aluminum alloys include casting and plastic processing techniques [10]. Given the steep cooling rates associated with the process, these alloys tend to suffer from micro-cracking during LPBF processing, resulting in parts with poor structural integrity [11].

In contrast to conventional casting processes, metallic materials fabricated through LPBF exhibit non-equilibrium solidification features, with grains tending to grow epitaxially along the forming direction, ultimately resulting in a columnar

crystal structure characterized by a distinct preferred orientation [12-14]. Especially for alloys with high hot-cracking sensitivity, the coarse columnar crystal structure can easily lead to periodic hot cracks [15]. Consequently, among the myriad alloys presently utilized, the range of metal materials suitable for LPBF formation is highly restricted [16]. LPBF can only be used to fabricate limited Al–Si series casting aluminum alloys such as AlSi10Mg [17] and AlSi12[18], which show insufficient strength for aerospace and automotive applications [19]. Due to their limited mechanical properties, these alloys are typically relegated to manufacturing intricate cavities or topological lattice structure components [20,21]. Systematic research on the LPBF of high-strength aluminum alloys is thus required because of the wide freezing range and poor welding performance inherent in high-strength aluminum alloys. The forming process of LPBF often exhibits a pronounced propensity for hot cracking, resulting in the occurrence of hot cracks in the fabricated components that can be challenging to mitigate, thereby weakening various mechanical properties.

Hughes Research Laboratory (HRL) [14] has demonstrated that crack suppression in laser additively manufactured aluminum alloys can be achieved by introducing extra inoculants to control the solidification. Martin et al. [14] introduced 2 wt.% ZrH₂ nanoparticles into Al7075 alloy, and after standard T6 heat treatment, the modified Al7075 alloy showed a reasonable ultimate tensile strength (UTS) of 400 MPa but a poor elongation of 4%. Despite the achievement of grain refinement and crack elimination, the mechanical properties were found to be inferior compared to the conventional wrought parts. This finding could presumably be attributable to several factors, such as the formation of excessive brittle Al₃Zr inclusions. Zhu et al. [22] obtained an Al-Zn-Mg-Cu-Sc-Zr alloy with a UTS value of over 600 MPa and elongation over 10% by using a low scanning speed during LPBF processing and subsequent T6 heat treatment. The high level of the Sc element makes this approach very expensive, however.

In addition to Sc and Zr, the introduction of the Ti element can effectively promote grain refinement and the transformation of columnar crystals to equiaxed crystals in LPBF-fabricated high-strength aluminum alloys, thereby inhibiting hot cracking and improving various mechanical properties [23]. As an alloying element with a strong growth-suppressing effect, the Ti element can suppress the growth of grains, while the Al_3Ti phase generated by the in-situ reaction may act as an efficient nucleation agent, promoting grain refinement. Tan et al. [24] prepared Ti nanoparticle modified 2024 aluminum alloy via LPBF and found that the fabricated specimens offered equiaxed crystals with an average grain size of about 2 μm ; the UTS value and elongation after T6 treatment were determined to be 452 MPa and 10.08 % respectively. Ti powder can burn or explode violently in air at room temperature due to static electricity, sparks or friction, however, making its storage and transportation inconvenient.

However, the nucleation efficiency of carbides or borides introduced into the melt pool could be affected by factors such as particle size distribution, insufficient supercooling, gravity, and agglomeration. Previous studies indicated that less than 1% nucleating particles generally act as effective nucleators, with efficiency declining further when the nucleating agent concentration increases [25]. To achieve a sufficient content of nucleating particles, the introduction of multiple nucleating agents is reasonable. However, this approach may complicate the reactions among elements, potentially leading to counterproductive effects. Ouyang et al. [23] introduced TiB_2 and TiH_2 simultaneously into Al7075 alloy and found that the combination of TiB_2 and TiH_2 could significantly inhibit cracks in LPBF-fabricated specimens. TiH_2 may introduce gas holes that lessen the densification behavior. Zhou et al. [27] prepared modified Al-Zn-Mg-Cu containing 4 wt.% Si and 2 wt.% TiB_2 . The formability of the alloy was improved with complete crack elimination, but a large number of pores occurred. The UTS value of the LPBF-fabricated Al-Zn-Mg-Cu specimen was found to be 556 MPa, and the elongation was approximately 4.5 %. Li et al. [28] used Ti6Al4V (TC4) alloy as the Ti element

to eliminate cracks and improve various mechanical properties.

The addition of all the above refinements requires considerable time to prepare the raw materials [29]. Even the quickest TC4 addition requires two hours of time, which greatly reduces the efficiency for large-scale industrial production. Thus, it is crucial to design an efficient grain refiner to improve the mechanical performance of high-strength aluminum alloy fabricated via LPBF.

The aim of the present study is to develop a novel approach to suppress the cracking defect within LPBF of high-strength aluminum alloy. TC4 and B powders were incorporated into the alloy through an efficient double centrifugal powder mixing method, which in-situ generated Al_3Ti and TiB_2 during the LPBF process and offered a strengthening effect on the Al7075 alloy, yielding an alloy with excellent UTS, good ductility and superior formability. The optimal forming process parameters were obtained through a combination of dynamic simulation and experimental studies. The effects of TC4 and B content on the formability, microstructure evolution and mechanical properties of LPBF fabricated modified Al7075 (Al7075-M) aluminum alloy were systematically studied. The lattice matching relationship between the in-situ generated TiB_2 and Al_3Ti phases together with the strengthening mechanism, was also analyzed.

2. Experimental methods

2.1. Materials and equipment

Fig. 1a shows a scanning electron microscopy (SEM) micrograph of spherical Al7075 alloy powder. Particle size characterization was conducted using laser diffraction with a laser particle size analyzer. The particle size of Al7075 exhibited a distinct unimodal distribution, with an average particle size of 32.7 μm , as shown in Fig. 1b. The commercially used TC4 powder (15-53 μm) was employed in the preliminary study, unfortunately, the TC4 powder could not be uniformly dispersed amongst the Al7075 powder under all the mixing conditions

(Fig. S1 and Table S1, the supplementary document). However, when the TC4 powder was reduced to a smaller scale with a D_{50} of 8.74 μm (Fig. 1c), a much more uniform distribution and better flowability of the composite powder were achieved. TC4 powder with diameters in 0-15 μm was thus employed to obtain the uniform composite powder used in this study. The B powders employed were irregularly shaped, with a purity of 99.9% and an average diameter of 500 nm (Fig. 1d). The composite powder was prepared using a high-speed mixer (SpeedMixer DAC 800.1 FVZ, Nordrhein-Westfalen, Germany).

2.3 wt.% TC4 powder and 0.9 wt.% B powder were added to the Al7075 powder to generate 3 wt.% TiB_2 in the modified alloy by in-situ reaction. The specific percentages were determined based on our preliminary experimental work. Assuming that the Ti and B added during the LPBF process react completely, TiB_2 can be generated by the reaction described by the chemical equation $\text{Ti} + 2\text{B} = \text{TiB}_2$. As shown in Fig. S2 (the supplementary document), where varying amounts (1 wt.%, 2 wt.%, and 3 wt.%) were tested. It was found that only the 3 wt.% TiB_2 condition could achieve the microcracks suppression. Therefore, 3 wt.% TiB_2 was determined to be the optimal content for subsequent experiments. Based on this, the corresponding percentages of 2.3 wt.% TC4 and 0.9 wt.% B were determined and employed. It is indeed very challenging to ensure that the added B can fully react with TC4 to produce 3 wt.% TiB_2 , but in terms of the Gibbs free energy calculations (Fig. S3, the supplementary document), B is preferential to react with Ti to form TiB_2 . Our future work will focus on the LPBF process control to evaluate the in-situ reactions.

TC4 and B powder were found to be uniformly distributed within the Al7075 powder, indicating a proper dispersion of the high-speed mixing and acceptable flowability of the mixed feedstock (Fig. 1e). A schematic diagram of the powder mixing process is shown in Fig. 1g. The powder was mixed at 850 rpm for 1.5 min and then repeated after cooling to ambient temperature to prevent excessive heating. Composition analyses of both the initial powder and the

composite powder were conducted using inductively coupled plasma atomic emission spectrometry (ICP-AES); the results are shown in Table 1.

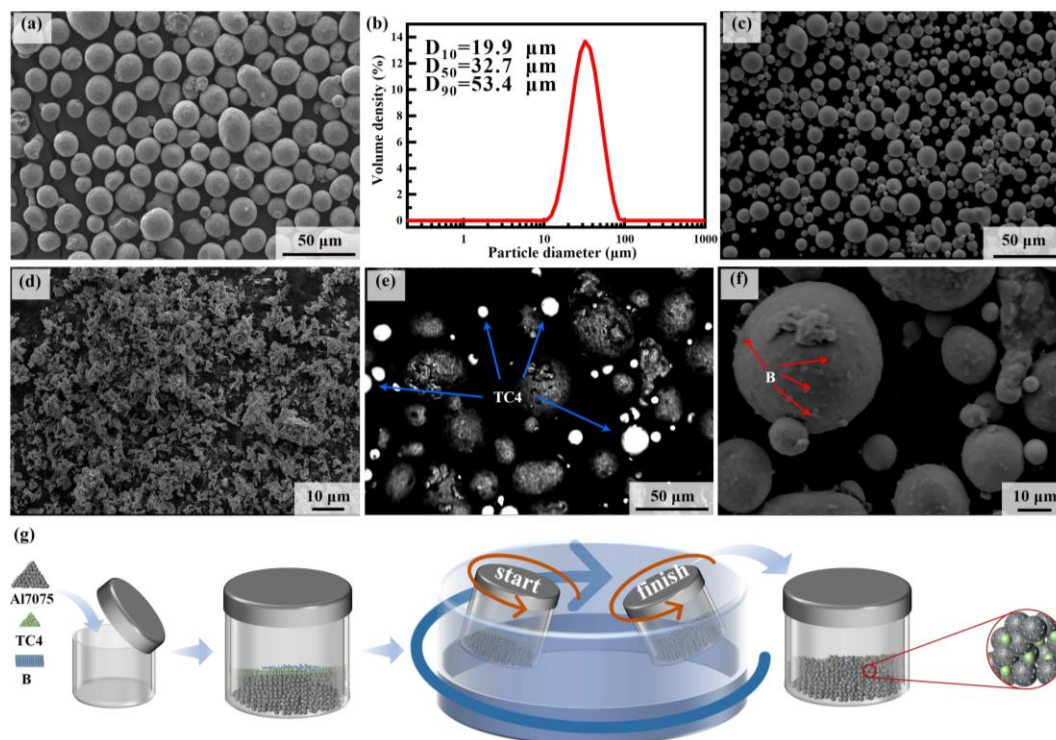


Fig. 1. Scanning electron microscopy (SEM) images and particle size distribution of powders: (a) Al7075 powder; (b) Al7075 powder size distribution; (c) TC4 powder; (d) B powder; (e) Al7075-M composite powder at low magnification; (f) Al7075-M composite powder at high magnification; (g) the powder mixing process

Table 1.

Actual chemical composition of powders used in the laser powder bed fusion (LPBF) process

Element(wt.%)	Al	Zn	Mg	Cu	Cr	Ti	B	Si
Al7075 powder	Bal.	4.6	2.71	1.51	0.21	0.02	-	0.09
Al7075-M powder	Bal.	4.6	2.75	1.53	0.21	1.94	0.77	0.08

All the specimens were manufactured under an argon atmosphere using a Mlab cusing 200R (Concept Laser GmbH, Germany) LPBF machine. The processing parameters were optimised by changing the hatch spacing and the scanning speed values, while the layer thickness and laser power were fixed at 30 μm and 190 W, respectively. Cubic specimens ($7 \times 7 \times 7 \text{ mm}^3$) were constructed using an interlayer rotation sequence of 67° on the Al7075 substrate without pre-heating. The densities of the additively manufactured

specimens were measured using the Archimedes method, with each specimen being measured three times and then averaged. Relative densities were then calculated based on the measured density and theoretical density, the latter of which is 2.73 g/mm³ for this material. The optimal LPBF process parameters were determined by the relative density and microstructure observation of the cubic specimens. Tensile specimens of Al7075 and Al7075-M were then manufactured using the optimal parameters. An optimised heat treatment, based on the T6 heat treatment (which consists of solution, quenching and ageing), was applied to the specimens in this study. The ageing temperature was fixed at 120 °C, while the solution time, solution temperature and ageing time were optimised using the orthogonal experimental method. The heat treatment scenario using orthogonal experimentation is shown in Table 2.

Table 2.

Orthogonal experimental table of heat treatment process parameters

Group	Solution temp. (°C)	Solution time (h)	Ageing time (h)
H1	460	1	24
H2	460	2	48
H3	460	3	36
H4	490	1	48
H5	490	2	36
H6	490	3	24
H7	520	1	36
H8	520	2	24
H9	520	3	48

2.2 Numerical model establishment

In this study, to better understand the thermodynamics of the melt pool and the added TC4 particles, numerical simulation of fluid dynamics was conducted

based on the developed single-track three-dimensional transient model that was developed. As mentioned in section 2.1, the B powder was irregular in shape and had an average diameter of only 500 nm, which was difficult to model and mesh. As a result, only the Al7075 and TC4 powders were taken into consideration during the simulation. This model was implemented in a computational fluid dynamics (CFD) framework using the commercial software Ansys Fluent, which operates on the finite volume method (FVM) [30]. The geometric model used in this simulation is shown in Fig. 2. In this model, the substrate was an Al7075 aluminium alloy that was consistent with the matrix material. The Al7075 powder particle size was normally distributed between 15 and 53 μm , and the TC4 powder particle size was normally distributed between 1 and 15 μm .

The underlying assumptions of our model are crucial for its accurate representation: (i) the flow is incompressible and viscous; (ii) the flow is assumed to be laminar and the fluid is Newtonian; (iii) the spattering of material is not considered; (iv) composition changes due to evaporation are ignored; (v) the absorptivity of the laser is set as a constant [30].

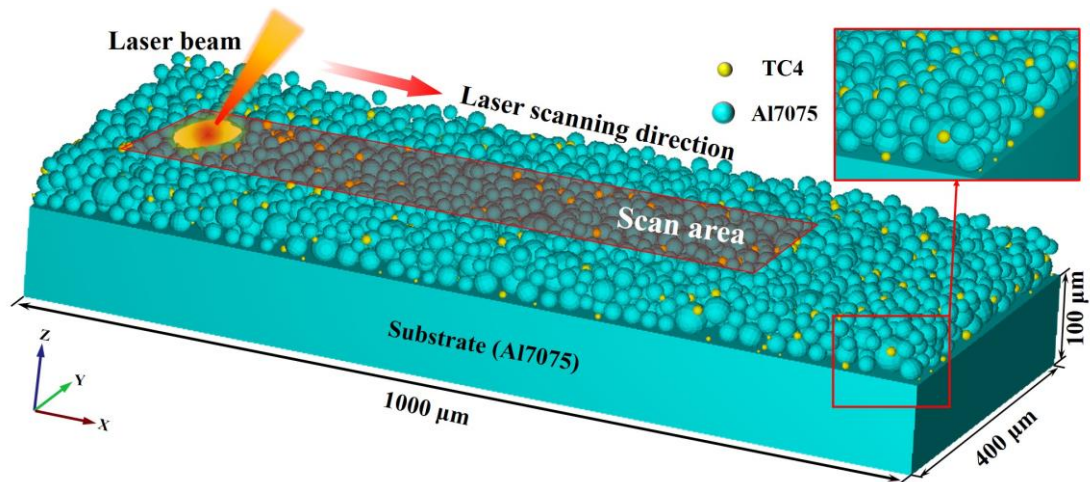


Fig. 2. Schematic of the model used in simulation

The volume of fluid (VOF) algorithm was employed to capture the free surface of the metallic phase. The conservation of volume fraction is described by Equation 1 [31]:

$$\frac{\partial \alpha_1}{\partial t} + \nabla \cdot (\vec{u} \alpha_1) = - \frac{\alpha_1}{\rho_{metal}} \frac{D \rho_{metal}}{Dt} \quad (1)$$

where $\vec{u}$ is the velocity vector field, $\frac{D}{Dt}$ represents the Lagrangian time derivative, and ρ_{metal} and α_1 denote the density and volume fraction of the metallic phase, respectively.

Mass conservation, as described by the Mass Conservation Equation, ensures the preservation of mass within the system.

$$\frac{\partial \bar{\rho}}{\partial t} + \nabla \cdot (\bar{\rho} \vec{u}) = 0 \quad (2)$$

The momentum conservation equation is as follows [32]:

$$\frac{\partial (\bar{\rho} \vec{u})}{\partial t} + \nabla \cdot (\bar{\rho} \vec{u} \vec{u}) = -\nabla P + \bar{\rho} \vec{g} + \nabla \cdot \hat{\tau} + \vec{f}_{st} + \vec{f}_M + \vec{P}_r + \vec{f}_{damp} \quad (3)$$

where the right-hand side's terms stand for (respectively) forces due to pressure gradient, gravity, viscous stress, surface tension, Marangoni force, recoil pressure and a damping force term. The Marangoni shear force is computed as [30]:

$$\vec{f}_M = \frac{d\sigma}{dT} [\nabla T - (\vec{n} \cdot \nabla T) \vec{n}] |\nabla \alpha_1| \frac{2\bar{\rho}}{\rho_{metal} + \rho_{gas}} \quad (4)$$

where $\frac{d\sigma}{dT}$ is the surface tension gradient of Al7075. The recoil pressure is [30]:

$$\vec{P}_r = 0.54 P_0 \exp \left[\frac{L_v M (T - T_v)}{R T T_v} \right] \vec{n} |\nabla \alpha_1| \frac{2\bar{\rho}}{\rho_{metal} + \rho_{gas}} \quad (5)$$

where P_0 is the ambient pressure, L_v is the latent heat of evaporation, T_v is the evaporation temperature, M is the molar mass and R is the universal gas constant.

The energy balance equation is applied to the computational domain and subsequently coupled to the velocity field, to find the temperature field and the fluid's fusion condition (liquid or solid state) [30]:

$$\frac{\partial (\rho H)}{\partial t} + \frac{\partial (\rho u H)}{\partial x} + \frac{\partial (\rho v H)}{\partial y} + \frac{\partial (\rho w H)}{\partial z} = \frac{\partial}{\partial x} \left(k \frac{\partial T}{\partial x} \right) + \frac{\partial}{\partial y} \left(k \frac{\partial T}{\partial y} \right) + \frac{\partial}{\partial z} \left(k \frac{\partial T}{\partial z} \right) - \dot{q}_{evap}''' \quad (6)$$

where $\dot{q}_{evap}'''$ ($W \cdot m^{-3}$) is the volumetric heat sink due to the evaporative cooling of the fluid (metal), defined as [30]:

$$\dot{q}_{evap}''' = \Delta H_{lv} \cdot A_{accom} \cdot \sqrt{\frac{M}{2\pi \bar{R} T}} \cdot P_{atm} \exp \left[\frac{\Delta H_{lv}}{C_v(\gamma-1)} \left(\frac{1}{T_g} - \frac{1}{T} \right) \right] \quad (7)$$

where A_{accom} and $\bar{R}$ ($\text{J}\cdot\text{kmole}^{-1}\cdot\text{K}^{-1}$) are the accommodation coefficient and the universal gas constant, respectively. P_{atm} and T_g in Equation 7 are the atmospheric pressure and the fluid's saturation temperature (atmospheric boiling temperature) at this pressure, respectively. ΔH_{lv} ($\text{J}\cdot\text{kg}^{-1}$) is the latent heat of vaporisation and C ($\text{J}\cdot\text{kg}^{-1}\cdot\text{K}^{-1}$) is the specific heat capacity of the vaporised fluid at a constant volume.

The heat generation amount used in the model is described as follows [33]:

$$Q = \frac{3nP}{\pi r^2} \exp \left(-\frac{3d^2}{r^2} \right) \quad (8)$$

where P is laser energy (W), r is the radius of laser beam (μm), d is the distance between the powder surface and the centre of the laser (μm), and n is the laser absorptivity of metal powders. By setting boundary conditions, the laser absorptivity varies when the laser irradiates different materials. Table 3 shows the physical properties of the Al7075 alloy and TC4 alloy used in the study.

Table 3.

Material properties and related coefficients used in the simulation

Properties	Al7075 alloy [34]	TC4
Density ($\text{kg}\cdot\text{m}^{-3}$)	2368	4407
Solidus temperature (K)	805	1877
Liquidus temperature (K)	901	1923
Specific heat capacity ($\text{J}\cdot\text{kg}^{-1}\cdot\text{K}^{-1}$)	1135	628
Heat conductivity coefficient ($\text{W}\cdot\text{m}^{-1}\cdot\text{K}^{-1}$)	220	9.47
Surface tension ($\text{N}\cdot\text{m}^{-1}$)	0.868	1.538
Dynamics viscosity ($\text{Pa}\cdot\text{s}$)	1.257×10^{-3}	2.347×10^{-3}
Latent heat for solidification ($\text{J}\cdot\text{kg}^{-1}$)	397500	286000
Laser-absorption rate (%)	48.5	75.4

2.3 Material characterisation techniques

The YOZ surfaces of the cubic specimens, which were parallel to the build direction (BD), were ground and polished to analyse the microstructural evolution of LPBF-fabricated Al7075 and Al7075-M materials. To quantitatively analyse the density of defects and cracks, six regions (with a total area of about 8 mm²) on the polished surface were selected and photographed using an optical microscope on each specimen. The polished surfaces were etched using a mixed acid solution (H₂O: HNO₃: HCl: HF = 95: 2.5: 1.5: 1.0, volume fraction) at room temperature (25 ± 2 °C) for 10 s and then rinsed with water before observation. An EM-30 Plus (COXEM, South Korea) SEM was employed to observe the microstructure and cracking morphology of the specimens. Backscattered electron (BSE) and secondary electron (SE) images were taken at 20 and 10 kV accelerated voltages, respectively. Electron backscattered diffraction (EBSD) characterisation was performed using a JSM-7800 (Jeol, Japan) SEM equipped with a Nordlys Max3 (Oxford Instruments, UK) EBSD probe. The EBSD data were analysed using the HKL Channel 5 (Oxford Instruments, UK) software package. A transmission electron microscope (TEM) (Talos F200F, FEI, USA) and high-angle annular dark field detector (HAADF) were utilized to analyze the distribution of precipitated phases. To simulate the effect of TC4 addition on the LPBF processing of Al7075, the laser absorptivity of TC4 and Al7075 powders was measured using a high-performance ultraviolet-visible (UV-vis) and near-infrared (NIR) spectrophotometer (PerkinElmer, Lambda750 (S), USA).

The mechanical properties of the Al7075 and Al7075-M alloy were determined through tensile tests conducted using an electromechanical universal testing machine (ME 50, BLOWAS, Shenzhen, China). Three tests were conducted for each type of specimen. A standard extensometer with a gauge length of 25 mm was used to measure the strain. A tensile speed of 1 mm/min was used in all tests.

3. Results

3.1. Cracking elimination

3D optical micrographs of the Al7075 and Al7075-M alloys prepared by LPBF are shown in Fig. 3a-b, respectively. Macrocracks and pores were observed in the Al7075 alloy, the cracks were large in size and propagated through the melt pool. As expected, cracks were not found on either the XOZ or XOY planes of the Al7075-M specimen. A small number of open pores, however, were still observed in the Al7075-M specimen, possibly due to the adsorption of water vapor from the air. Hydrogen pores were generated in the melt pool, and the burning of alloying elements with low melting points may also cause the pores formation, which is unavoidable. These findings are consistent with the literature [35].

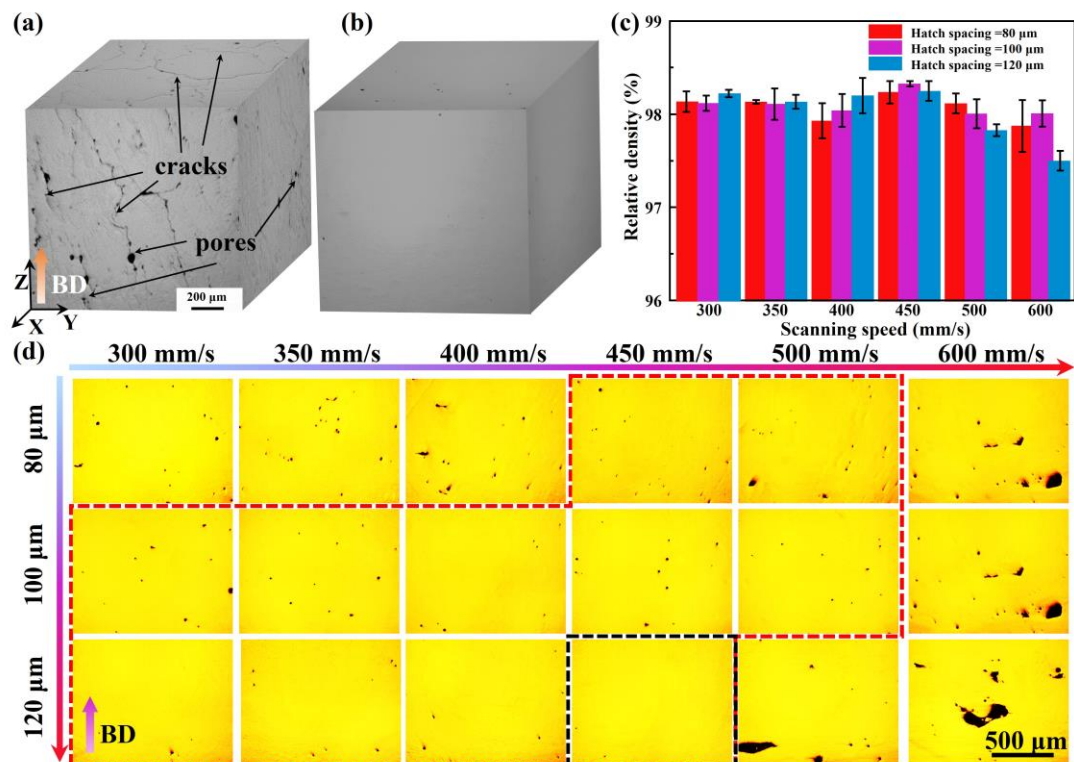


Fig. 3. Optical micrographs (OM) of the alloys prepared by LPBF: (a) Al7075 alloy; (b) Al7075-M alloy; (c) Effects of scanning speed and hatch spacing on relative density of Al7075-M; (d) cross-sectional optical micrographs of Al7075-M alloy

The cross-sectional OM images of the Al7075-M alloy are shown in Fig. 3d. The Z-axis was parallel to the build direction in this study. Hot cracks were not observed under any parameter. The main defects at lower scanning speeds (between 300 mm/s and 450 mm/s) were open pores, while lack of fusion became the main defect when the speed was higher than 500 mm/s. Unfused pores under low energy density were usually caused by adjacent pores among the powders, where the melt pools were insufficiently overlapped. The overall defects of the specimens were evaluated by measuring the relative density, as shown in Fig. 3c. Considering both efficiency and density comprehensively, the 120 μm –450 mm/s condition was selected as the optimal process parameters for subsequent specimens' fabrication.

Generally, laser power (p) and scanning speed (v) are considered to be the two most important factors related to the melt pool's state and morphology. Since previous studies have shown that a higher laser power is required in the LPBF processing of high-strength aluminium alloys, the laser power was fixed at 190 W during numerical simulations in the present study [36]. Only the effect of scanning speed on the continuity and stability of the melting trajectory was explored in this study. Fig. 4 shows the simulated melt pool temperature field in single-track scanning at different scanning speeds. At the 600 mm/s scanning speed, the laser energy density was too low to completely melt the powder bed (Fig. 4a), resulting in some TC4 particles not melting. Reducing the scanning speed to 450 mm/s increased the accumulated heat, melting more powder particles and eliminating spherical defects. Unmelted TC4 particles were not observed, as shown in Fig. 4b.

As shown in Fig. 4c-d, the maximum temperature of the melt pool was relatively stable, which was near the TC4 liquidus temperature at 450 mm/s and 600 mm/s. Some of the TC4 particles remained unmelted when the scanning speed was 600 mm/s, however. The midpoints A and B of the single-track simulation trajectory were selected at 450 mm/s and 600 mm/s (respectively) to observe their temperature evolution history; the results are shown in Fig. 4c-

d. As Fig. 4c-d shown, the temperature at point A in the 600 mm/s reached the liquidus temperature for a short period, which was much shorter than that at point B, at 450 mm/s. This finding indicates that a scanning speed of 600 mm/s was insufficient to completely melt the TC4 powder, which in turn means that a longer duration at a higher temperature of 450 mm/s could result in more complete melting and better bonding of the TC4 particles with the Al7075 matrix.

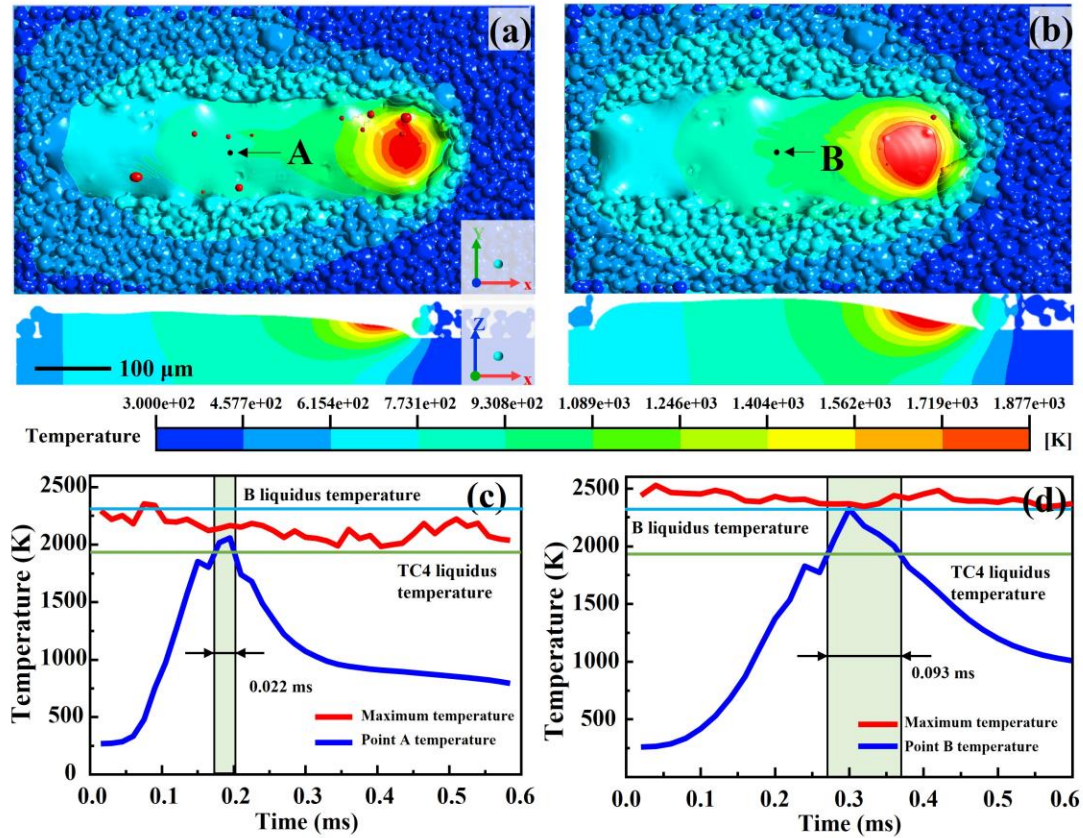


Fig. 4. (a-b) Effect of scanning speed on melt pool temperature: (a) 600 mm/s; (b) 450 mm/s; (c-d) maximum temperature in the melt pool and the temperature change at (c) 600 mm/s, point A; (d) 450 mm/s, point B

In addition, the maximum temperature of the melt pool was around 2400 K, while the liquidus temperature of B is around 2300 K. Since the particle size of the B powder was much smaller than that of the TC4 powder, 450 mm/s was sufficient to melt the B powder. The simulation results were consistent with the LPBF-forming experiments. Obvious defects were not observed at 450 mm/s, and the specimen was highly dense, while many unmelted fusions were observed in the 600 mm/s specimen (Fig. 3d). Additionally, slower speeds (450

mm/s) typically result in slower cooling rates, which is beneficial in minimising thermal stress and promoting a more uniform microstructure.

Fig. 5a-b shows the volume fraction of TC4 in the melt pool at two different scanning speeds. Most of the TC4 observed in the melt pool at 450 mm/s was irregularly shaped, indicating that the TC4 powder was well melted and diffused into the melt pool. At 600 mm/s, however, most TC4 in the melt pool remained spherical. This finding that the 450 mm/s scanning speed provided a more favorable environment for melting and diffusing. Fig. 5c-d shows the speed vectors at different positions inside the melt pool, which indicate the movement and mixing behavior of the melted material. The temperature field is shown as well. The laser produced a steep temperature gradient in the melt pool, with the highest temperature located in the center, near the laser spot, and lower temperatures found at the edge. The temperature gradient drove the Marangoni convection, which in general is mainly caused by the different surface tensions of a liquid at different temperatures [37].

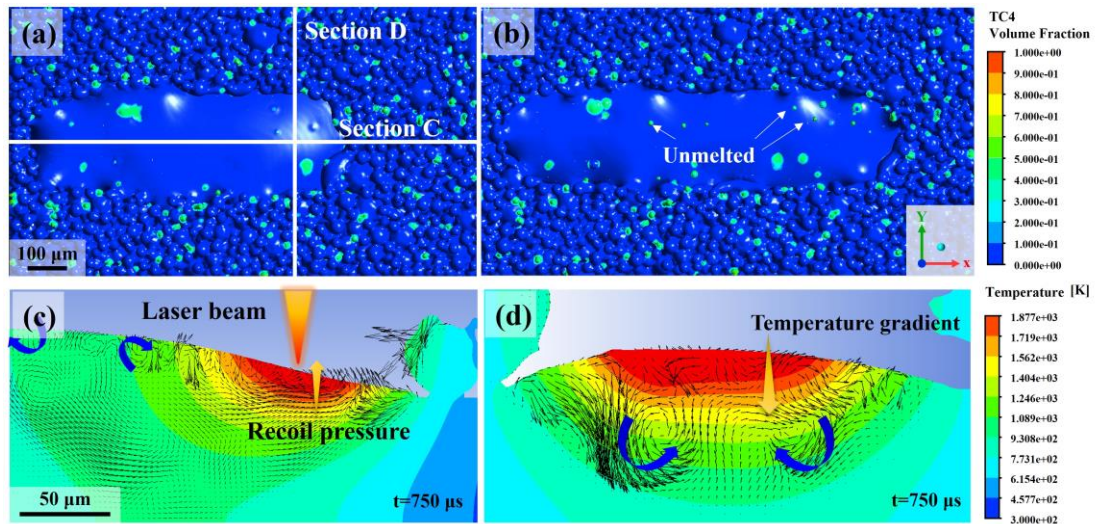


Fig. 5. (a-b) Effect of scanning speed on unmelted TC4 volume fraction: (a) 450 mm/s; (b) 600 mm/s; (c-d) cross-sectional view of the evolution of the melt pool at $t = 750 \mu\text{s}$ during single-track scanning at 450 mm/s: (c) section C; (d) section D

The surface tension of a metal liquid is generally lower at higher temperatures, which may lead to a force in the same direction of temperature gradient. Driving the melted material flows from hotter areas to colder areas.

Due to the surface tension effect, the melted material is influenced by capillary forces, causing the liquid metal to flow and become redistributed within the melt pool. The density of the molten material changes with temperature; hotter areas are less dense, and vice versa. These density differences create buoyancy, driving hotter material to rise and cooler material to sink, further enhancing the convection within the melt pool.

In addition, the direct impact of the laser beam on the melt pool creates local pressure and vaporization effects. This pressure can push the melted material away from the laser spot, promoting flow and mixing. In general, the vaporization of the material at the laser spot creates recoil pressure, pushing the melted material outward and enhancing movement within the melt pool [38]. These factors work together to create a dynamic environment within the melt pool, promoting effective mixing of the Ti element and ensuring uniformity. The results of the simulation also support the hypothesis that Al_3Ti is generated by in-situ reactions between TC4 and Al7075. It should be noted that the Al vaporization is not included in the present simulation and this is mainly because both the TC4 and Al7075 alloys contain the Al element, offering a huge computing challenge in simulating its vaporization during the multi-physics LPBF simulation. In addition, the goal of the numerical simulation is to predict the temperature distribution within the melt pool under different LPBF parameters in order to determine whether the TC4 can be fully melted to provide the necessary Ti for subsequent in-situ reactions. Based on the two factors, the vaporization of Al element and Al-B-Ti alloy process were not considered in the present modelling.

3.2. Microstructure evolution

Obvious cracks parallel to the build direction and fish-scale melt pools were observed in the Al7075 alloy (Fig. 6a). The cracks extended along the coarse columnar grains, with lengths up to 300 μm . A few open pores were also observed in the specimen. The cracks were completely suppressed after adding TC4 and B, and the grains were transformed from coarse columnar

grains to fine equiaxed grains (Fig. 6b). The results show that the combination of TC4 and B managed to eliminate cracks in Al7075 alloy and promote equiaxialization during LPBF processing. Fig. 6c shows the microstructure of the LPBF-fabricated Al7075-M specimen at higher magnification on a specific area (highlighted by the yellow box in Fig. 6b). The sub-grains were about 1 μm in size and small irregularly shaped precipitates were found at the boundaries. Energy dispersive x-ray (EDX) mapping was then carried out to investigate the element distribution (Fig. 6c). The precipitants revealed a deficiency of Al and an accumulation of Ti and B, suggesting the possible presence of TiB_2 . Further analysis was later conducted via TEM.

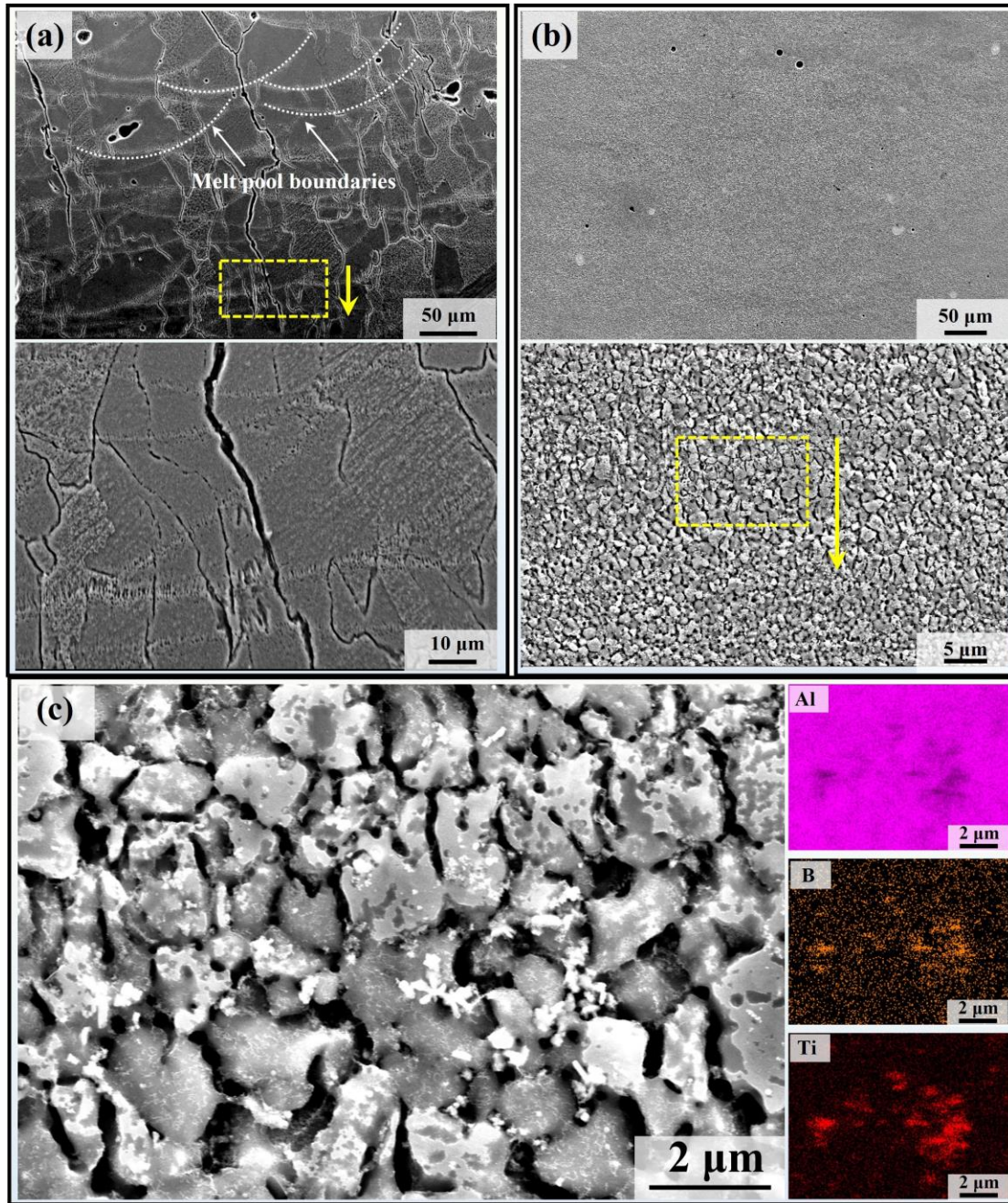


Fig. 6. SEM micrographs of as-fabricated specimens: (a) Al7075 specimen; (b) Al7075-M specimen; (c) high-resolution images and energy dispersive spectrometry (EDS) mapping of elements Al, Ti and B of the area indicated by the yellow rectangles in (b)

Fig. 7 illustrates the EBSD results of the as-fabricated Al7075 and Al7075-M alloys along the BD. Cracks were formed along the grain boundaries between the coarse columnar grains in the Al7075 specimen (Fig. 7a). 70% of the grains were within 10–80 μm in diameter, and the average grain size was determined to be 20.25 μm . In contrast, the as-fabricated Al7075-M alloy was shown to be crack-free (Fig. 7b), and to have an equiaxial-grained microstructure.

Importantly, the average grain size measured 1.95 μm , which was about 9.6% of the Al7075 specimen. All grains were found to be less than 8 μm in diameter. Additionally, the grain sizes showed a normal distribution within a narrow range of 1–8 μm . Fig. 7c shows the EBSD results of T6-treated Al7075-M. The grains remained equiaxial, and the average grain size was determined to be 1.88 μm , which was similar to the as-fabricated specimen.

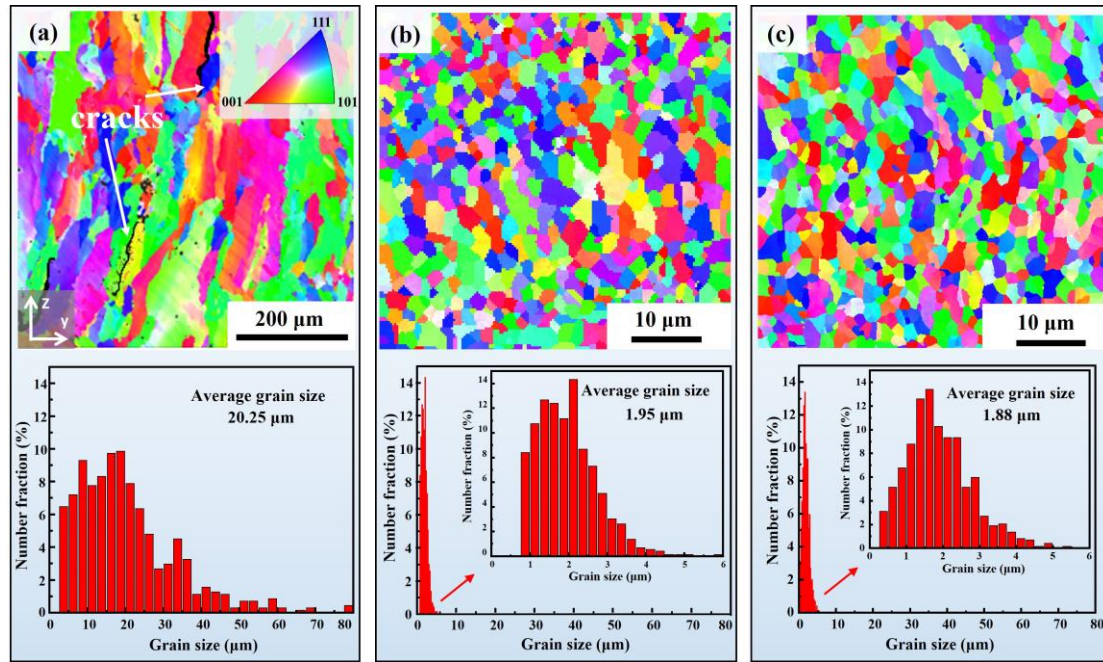


Fig. 7. Inverse pole figure (IPF)-Y maps and grain size distribution images: (a) Al7075 specimen; (b) Al7075-M specimen; (c) T6-Al7075-M specimen

Interestingly, an apparent grain growth in the Al7075-M specimen after T6 heat treatment was not detected and the absence of coarsening may be attributed to the following factors. First, the TiB_2 particles at the grain boundaries of the sedimentary samples may act as pinning agents and restrict grain growth [39]. Second, during the T6 heat treatment, fine precipitates such as MgZn_2 and Al_2Cu could contribute to stabilizing the grain size by effectively pinning the grain boundaries and preventing significant grain growth [40]. In addition, TiB_2 may continue to precipitate during the solution treatment, further impeding the movement of grain boundaries. The future work will conduct a more detailed examination regarding the microstructure evolution of the Al7075-M after the heat treatment.

High-resolution TEM inspection was used to reveal the strengthening and grain refinement mechanism of TC4 and B in the Al7075-M alloy. Fig. 8a shows an HAADF image of the Al7075-M specimen. Dark precipitates were found on the grain boundaries and inside the grains. EDX mapping was then conducted to investigate the element distribution in the area shown in Fig. 8a. An accumulation of Zn, Mg and Cu was noted at the grain boundaries. EDX analysis was then carried out at point A to investigate the element composition of the precipitate, shown in Fig. 8b. The results indicated that the precipitate was η -Mg (Zn, Cu, Al)₂. EDX mapping also showed an accumulation of Ti at the position of precipitates inside the grains. Fig. 8c shows the precipitates at a higher magnification. Selected area electron diffraction (SAED) was then conducted to analyse the precipitates, which indicated that the precipitate was L1₂-Al₃Ti; its spatial structure was consistent with the matrix, as shown in Fig. 8d.

Fig. 8e shows a high-resolution transmission electron microscopy (HRTEM) image at the L1₂-Al₃Ti/Al interface, revealing the coherent interface between the two phases. The corresponding inverse fast Fourier transform (IFFT) images of the interface marked by the circular frame in Fig. 8d are shown in Fig. 8(f-g). The fully coherent interface between the L1₂-Al₃Ti and α -Al matrix indicates an excellent crystallographic match. The measured interplanar spacings of α -Al and L1₂-Al₃Ti were 0.250 and 0.252 nm respectively, as shown in Fig. 8f-g; the mismatch was about 0.8 %. According to the crystal structure and lattice parameters of the α -Al matrix and L1₂-Al₃Ti, the two planes were determined to be [101] α -Al and $[\bar{1}00]$ L1₂-Al₃Ti.

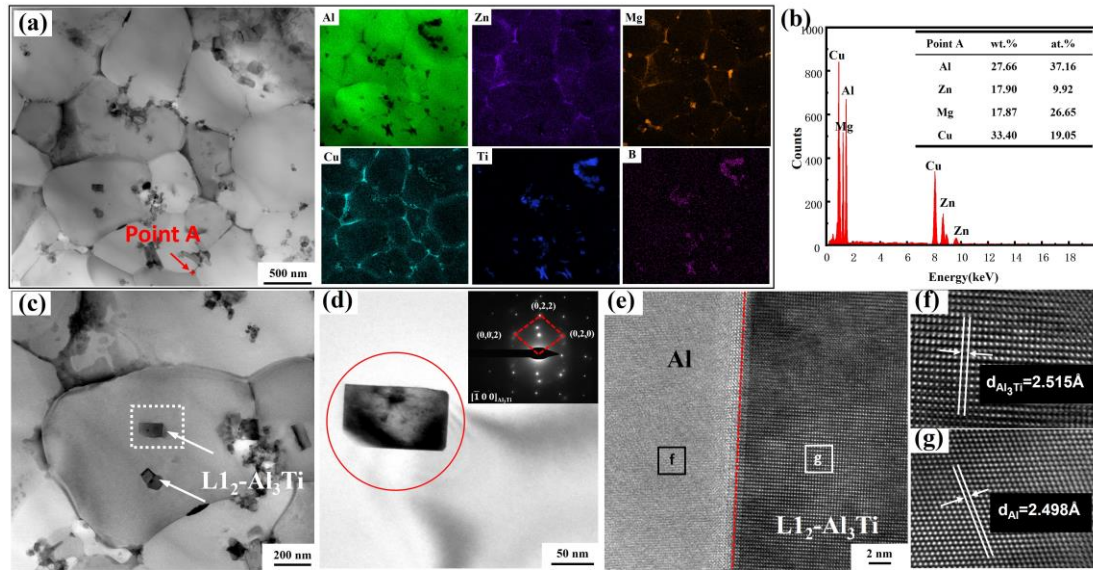


Fig. 8. (a) High-angle annular dark field detector (HAADF) micrographs and EDS mapping of as-built Al7075-M specimen; (b) EDS results of point A, marked in (a); (c) TEM bright-field image of as-built Al7075-M specimen; (d) selected area electron diffraction (SAED) pattern; (e) high-resolution transmission electron microscopy (HRTEM) image of as-built Al7075-M specimen; (f-g) inverse fast Fourier transform (IFFT) image marked by the box in (e)

Further TEM study of the Al7075-M specimen, showed that, in addition to the Al_3Ti with a tetragonal shape, some irregularly shaped precipitated phases appeared at the grain boundaries and within grains. According to the EDX mapping shown in Fig. 8a, the precipitates were rich in Ti and B. Fig. 9a shows a TEM image of a single precipitate inside a grain. The SAED pattern shown in Fig. 9b proves that the precipitate was TiB_2 . The HRTEM image in Fig. 9c highlights the TiB_2/Al interface, showing good interface bonding between the two phases. Figs. 8 and 9 show that the TC4 powder had melted and reacted in-situ with Al7075 and B to form TiB_2 and Al_3Ti .

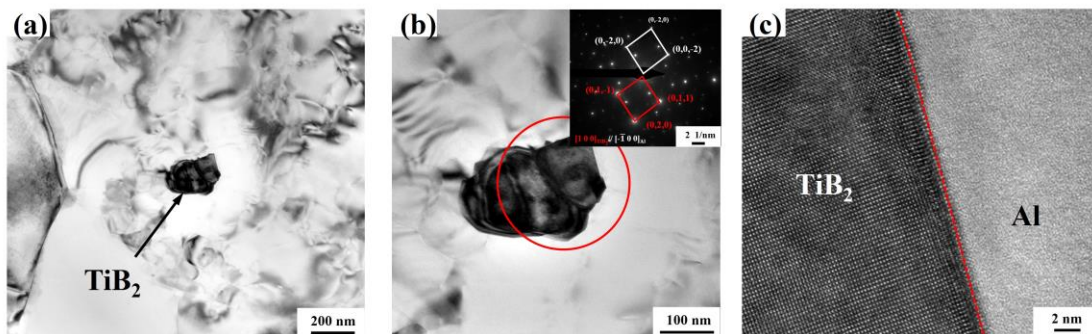


Fig. 9. (a) TEM bright-field image of as-built Al7075-M specimen; (b) SAED pattern; (c) HRTEM image of as-built Al7075-M specimen

The formation of Al-B phases was not found in present study and this is mainly because of the following factors. First, the Gibbs free energy calculations indicated that the reaction orders between Al, B and Ti are: $\text{Ti}+2\text{B}=\text{TiB}_2$, $\text{Al}+2\text{B}=\text{AlB}_2$ and then $3\text{Al}+\text{Ti}=\text{Al}_3\text{Ti}$, implying that the B element preferentially reacts with Ti to form TiB_2 (Fig. S3, the supplementary document). Second, our TEM analysis confirmed the significant enrichment of Al-Ti and Ti-B phases, with the identified phases of Al_3Ti and TiB_2 based on the SAED pattern analysis, while no apparent Al-B phase was detected, as shown in Fig. 8a and Fig. S4 (TEM analysis, the supplementary document).

3.3. Mechanical performance

Ambient temperature tensile tests were carried out to optimise the heat treatment; three specimens were tested with the same treatment to minimise possible errors (Fig. 10). Taking tensile strength as the optimization target, a range analysis was performed on the heat treatment process parameters. As shown in Table 3, the optimal solution temperature, solution time, and ageing time were 520 °C, 3 h and 36 h, respectively. Generally, post-processing is essential to guarantee the quality and consistency of final products within the overall manufacturing workflow. The optimized heat treat process for Al7075-M material in this study is slightly different from traditional T6 heat treatment process, so that the overall manufacturing workflow is not much affected but the mechanical performance of Al7075-M is improved via using the optimized heat treatment process.

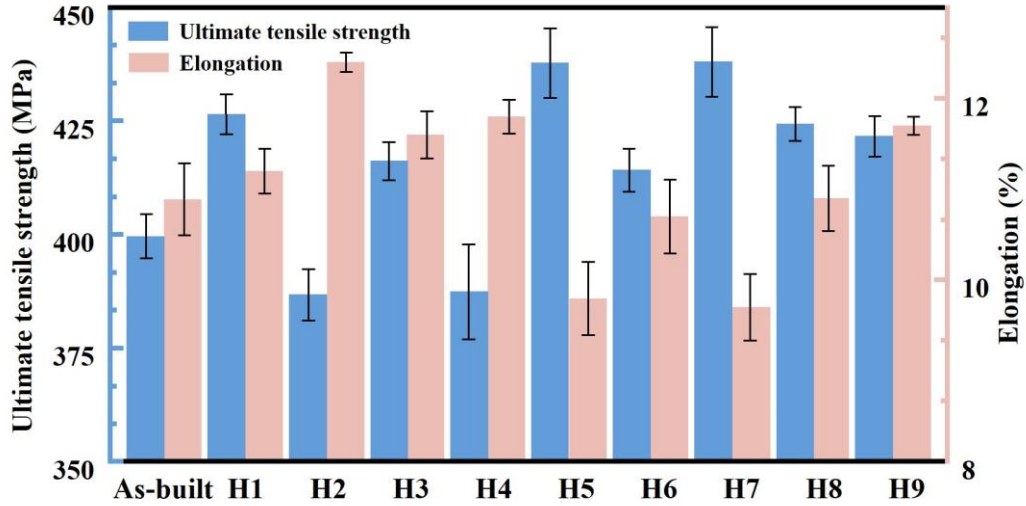


Fig. 10. Tensile properties of specimens under different heat treatments

Table 4.

Range analysis of different heat treatment process parameters

Range analysis		Solution temperature	Solution time	Ageing time
Taking UTS as the optimization goal	K1	1229.54	1252.06	1265.16
	K2	1239.55	1248.98	1292.13
	K3	1284.11	1252.16	1195.91
	Range	31.5	22.8	73.25
	Optimal scheme	520°C	3h	36h

The optimal heat treatment process was then used to post-process the Al7075-M tensile specimens. Fig. 11a shows the strain-stress curves of as-fabricated Al7075, as-fabricated Al7075-M alloy and Al7075-M alloy under optimal heat treatment. The specimens were fabricated using the same parameters of 450 mm/s scanning speed and 190 W laser power and were tested vertically to the build direction. As shown in Fig. 11a, the Al7075 alloy exhibited very poor mechanical properties due to the presence of a large number of macro-cracks (Fig. 6a). With the incorporation of TC4 and B, the UTS value of the as-fabricated Al7075-M alloy reached 404 MPa, which was significantly improved compared with the Al7075 alloy. The optimised-T6 heat

treatment resulted in a further increase in the UTS value to 460 MPa, and the elongation was 13.0 %. The mechanical properties of the LPBF-fabricated Al7075 and Al7075-M specimens were compared to 7xxx aluminum alloys manufactured by LPBF (Fig. 11b). The Al7075-M specimen exhibited superior strength and ductility compared to other LPBF-fabricated 7xxx aluminum alloys.

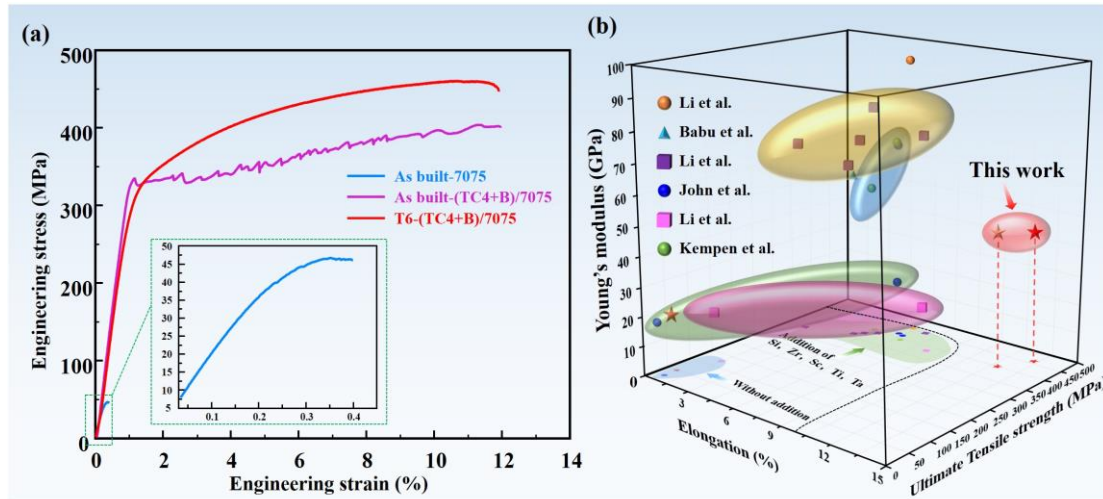


Fig. 11. (a) Tensile engineering stress-strain curves; (b) mechanical property comparison among the different Al-Zn-Mg alloys fabricated by LPBF [11,41,42]

4. Discussion

4.1. Cracking elimination mechanism

As mentioned in section 3.2, the cracks in the LPBF-fabricated Al7075 alloy mainly appeared at the grain boundaries. Since the cracks appeared in the center of the melt pool or passed through multiple melt pools, the cracks in the Al7075 alloy fabricated by LPBF were known to be solidification cracks. Solidification cracking, which is a type of hot cracking, occurs in the final stage of solidification. The main reason for the occurrence of such cracking is solidification shrinkage. Because the Al7075 alloy has a wide solidification range, it has in a large interdendritic area, with the dendrites in a mushy state. In such region, columnar dendrites continue to grow longitudinally and transversely, colliding with each other and overlapping the dendrites. The

extremely high cooling rate leads to severe thermal shrinkage, resulting in solidification shrinkage once the liquid solidifies; the stress from shrinkage causes the regions between dendrites to start separating. A cavity forms at the bottom of the area, making backfilling difficult. Further thermal cycling and solidification shrinkage typically lead to the development of cracks parallel to the columnar crystals. Repeated thermal cycles lead to the accumulation of residual stress, resulting in crack propagation [43].

The solidification paths of Al7075 and Al-7075M alloy were calculated with Thermo-Calc software, using the Gulliver-Scheil model (Fig. 12) B [44]. The precipitation sequence of Al7075 alloy is as follows: $L \rightarrow \alpha\text{-Al} \rightarrow \text{Mg}_2\text{Si} \rightarrow \text{S} \rightarrow \text{T}$. $\alpha\text{-Al}$ directly solidifies from liquid when the temperature decreases from 635.55 °C to 518.36 °C in the Al7075 alloy, with the corresponding solid fraction f_s increasing from 0 to 0.9. During such a long solidification process ($f_s = 0 \rightarrow 0.9$), the solidified structure might develop into large columnar grains because of the prevailing solidification conditions that favour epitaxial growth; the rest of the liquid metal existing between the grains covers the grain boundaries in the form of liquid films, leading to cracks [45]. With the addition of 1.9 wt.% Ti and 0.7 wt.% B, the precipitation sequence changes to the following: $L \rightarrow \text{AlB}_2 \rightarrow \text{Al}_3\text{Ti} \rightarrow \alpha\text{-Al} \rightarrow \text{Mg}_2\text{Si} \rightarrow \text{S} \rightarrow \text{T}$. The solubility of Ti in cast Al alloys is typically 0.15 wt.%. In the LPBF-fabricated Al-Zn-Mg-Cu-Ti alloy, the Ti solute is expected to be less than 0.15 wt.% [23] because of the solute becoming trapped with reduced partitioning during rapid solidification in LPBF. In general, the Al_3Ti phase precipitates within the temperature range of 734.83–638.36°C. A large amount of Al_3Ti was observed in the as-fabricated Al-7075M via TEM. In general, the large amount of precipitated Al_3Ti phase in the Al melt has the potential to not only significantly improve the nucleation rate of $\alpha\text{-Al}$ and refine the grains, but also to accelerate the morphological transition from columnar crystals to equiaxed grains.

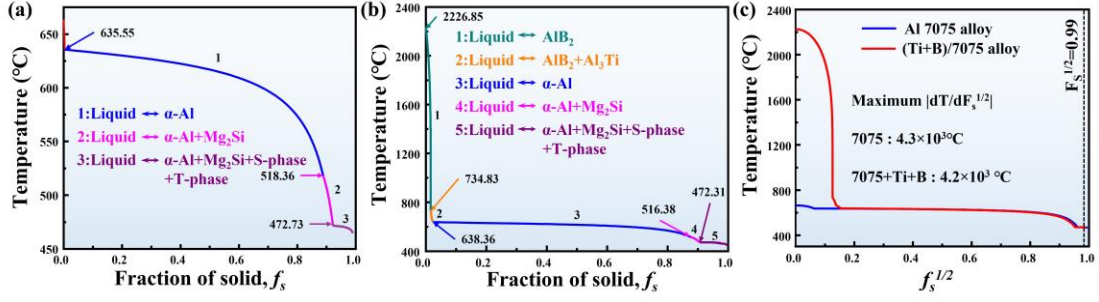


Fig. 12. Solidification curves (generated via Thermo-Calc) for (a) Al7075 alloy; (b) Al-7075M alloy; (c) comparison between Al7075 and Al-7075M alloy

The solidification crack index (SCI), proposed by Kou, was used to predict hot-cracking susceptibility in this study. Previous research has proved that SCI is valid in the LPBF processing of Al alloys. The SCI indicator was computed using the following equation [46]:

$$SCI = \left| \frac{dT}{d(f_s^{1/2})} \right|_{f_s=0.9-0.99} \quad (9)$$

where f_s is the molar fraction of a solid and T represents temperature. Because solidification cracking mainly occurs at the end stage of solidification, the calculation only took the final stage into consideration, where the solid fraction was 0.9-0.99. The higher value of SCI indicated a lower lateral growth rate of grains and a longer liquid feeding channel, which may have reduced the cracking resistance of the alloys during solidification. The calculation results in Fig. 12c show that the addition of Ti and B had a negligible impact on the terminal stage of the solidification path. In other words, the content of Ti and B had little effect on hot-cracking susceptibility according to Kou's theory. The Hunt criterion showed that a high density of nucleation sites could overcome the high G/R effect (the ratio of temperature gradient [G] and rate of growth [R]) and could play a more important role in the formation of equiaxed crystals [47]. Since the LPBF process parameters of all specimens were the same, the grain refinement effect was not controlled by the temperature gradient G and growth rate R, but rather was caused by heterogeneous nucleation.

In this study, the columnar grains were related to the directional epitaxial

growth of α -Al under the thermal gradient. If the overall undercooling (ΔT), consisting of constitutional supercooling (ΔT_{CS}) and thermal undercooling (ΔT_{therm}), at the front of the solid-liquid interface surpassed the critical nucleation undercooling (ΔT_n), then heterogeneous nucleation would occur [48,49]. The Ti element in TC4 offered good segregation, which could form a component supercooling zone at the front of the solid-liquid interface, thus activating the nucleating agent present at the front of the solid-liquid interface, and limiting the growth of dendrites. The limiting effect can be used to quantify the limiting factor. The growth restriction factor (GRF) is defined as [50]:

$$GRF = m(k - 1)C_0 \quad (10)$$

where m is the slope of the liquidus line, k represents the distribution coefficient, and C_0 represents the solute concentration. The $m(k-1)$ value of the Ti element is as high as 245.6 [50]. Ti has a high GRF value in Al alloys, which contributes to the rapid generation of constitutional supercooling zones, that is, increasing the ΔT_{CS} thus achieving sufficient ΔT to surpass the critical undercooling value for nucleation, ΔT_n [48].

Grain refinement can be attributed on the one hand to the formation of peritectic Al_3Ti particles. The equilibrium phase $D0_{22}-Al_3Ti$ and α -Al still have a coherent crystal plane with a lattice mismatch of less than 5%, such as (111) Al // (112) Al_3Ti (the blue plane in Fig. 15a and c). Due to the low cooling rate of conventional casting, the Al_3Ti that forms are often in an equilibrium phase with a tetragonal structure, as shown in Fig. 13a, while the extremely high cooling rate during the LPBF process allows for the easy formation of a meta-stable phase with a cubic $L1_2-Al_3Ti$ structure (Fig. 13b). The lattice constant of metastable $L1_2-Al_3Ti$ particles ($d = 3.967-4.05\text{\AA}$) is similar to α -Al ($d = 4.049\text{\AA}$) [51]. According to the lattice matching model, the lattice mismatch between the metastable phase $L1_2-Al_3Ti$ and α -Al is small, which can better promote the non-phase transformation and uniform nucleation of α -Al [52]. According to Liu et al. [53], the Al_3Ti phase matches the Al matrix almost perfectly, providing an ideal crystal nucleus for α -Al crystallization.

As shown in Figs. 8 and 9, Ti in TC4 reacts with the α -Al matrix to generate a large number of fine $L1_2$ - Al_3Ti . According to the interface coherent correspondence theory, heterogeneous substrates with lower interface energy are better able to promote heterogeneous nucleation. The more matching interfaces between heterogeneous sites and the main phase and the less lattice mismatch, the stronger the heterogeneous nucleation ability. $L1_2$ - Al_3Ti is generally considered to be an effective nucleation site for aluminium because the atomic spacing along $[001]$ Al// $[001]$ $L1_2$ - Al_3Ti is small (0.24%), as is the interplanar spacing between matching planes (0.24%) (020) Al // (020) $L1_2$ - Al_3Ti [24] (the green surface in Fig. 13b-c). Measurement of the interplanar spacing between α -Al and Al_3Ti (Fig. 8f-g) revealed a minor difference between the two spacings, suggesting that Al_3Ti is more effective in promoting heterogeneous nucleation.

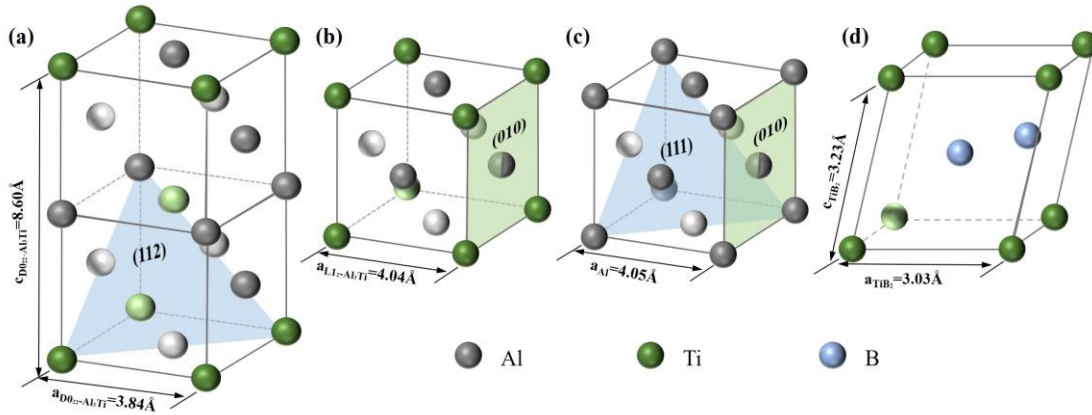


Fig. 13. Schematic illustration of unit cells of (a) $D0_{22}$ - Al_3Ti ; (b) $L1_2$ - Al_3Ti ; (c) Al [24]; (d) TiB_2

Some TiB_2 particles can serve as high-density non-uniform nucleation particles in the front of the solidification interface to reduce the critical nucleation undercooling (ΔT_n) required to form equiaxed grains and thereby refine the grains[50,54]. Other TiB_2 nanoparticles are pushed towards the grain boundaries, where they act as Zener pinning particles to prevent grain boundary expansion and further refine the grains [55]. A good lattice matching orientation relationship was observed between TiB_2 and Al [56]:

$$(200)_{\text{Al}} \parallel (10\bar{1}1)_{\text{TiB}_2}, [011]_{\text{Al}} \parallel [1\bar{2}10]_{\text{TiB}_2}$$

$$(220)_{\text{Al}} \parallel (11\bar{2}0)_{\text{TiB}_2}, [\bar{1}12]_{\text{Al}} \parallel [1\bar{1}00]_{\text{TiB}_2}$$

Fig. 14 shows a schematic illustration diagram of the microstructural changes and crack elimination in LPBF-fabricated Al7075 alloy with and without the addition of TC4 and B. Hot cracks occurred in Al7075 due to thermal shrinkage stress, as shown in Fig. 14a. With the introduction of the Ti element via the addition of TC4, Al_3Ti formed through the reaction between Ti and Al, achieving grain refinement and crack elimination, as shown in Fig. 14b. As Fig. 14c shows, the introduction of TC4 and B allowed the reaction between the Ti and B elements, forming TiB_2 ceramic particles that could act as nucleating agents, thus further refining the grains.

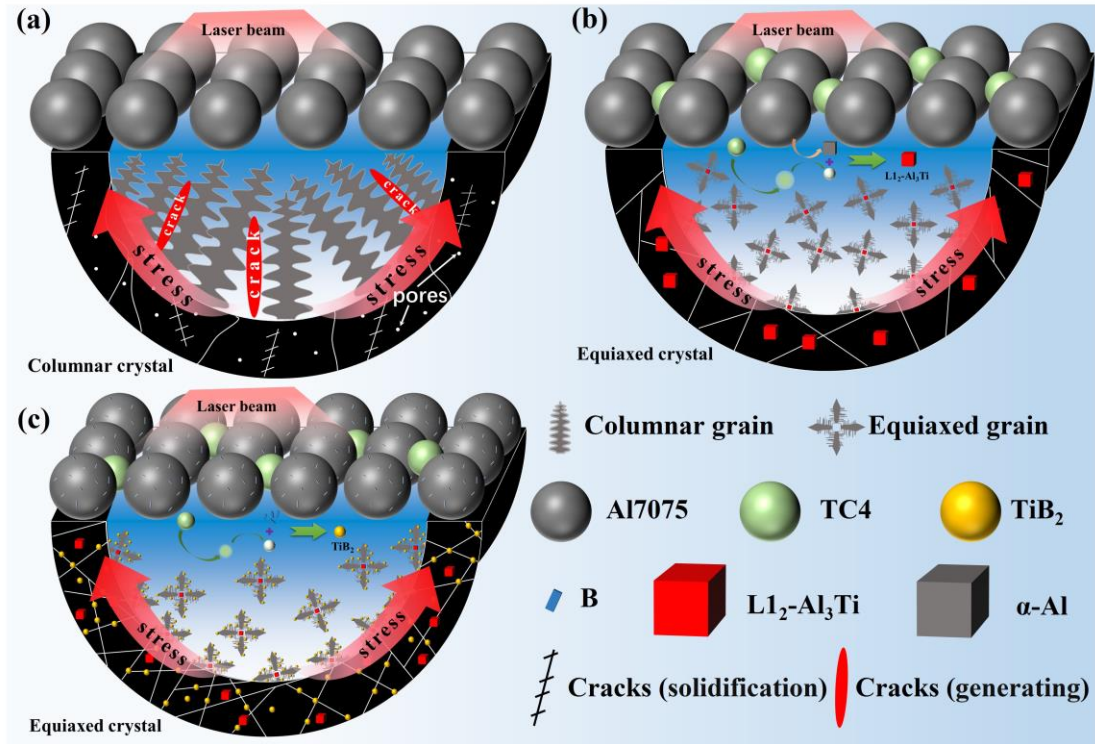


Fig. 14. Diagram of (a) typical columnar grain structure of LPBF-fabricated Al7075 alloys without any modification; (b) grain refinement induced by TC4 powder; (c) TC4 powder and B powder synergistically induced grain refinement

The in-situ reactions that occurred in the Al7075-M alloy are as follows [57]:



In general, within the Al7075-M system, the probability of combustion chemical reaction and priority is related to the Gibbs free energy (ΔG) of each reaction. Based on the known thermodynamic data, the Gibbs free energy change per mole of reactants in Equations 11 and 12 was calculated. The Gibbs free energy was calculated using the following equation:

$$\Delta G = \Delta H - T\Delta S \quad (13)$$

where ΔH is the standard enthalpy change, ΔS is the standard entropy change and T represents the thermodynamic temperature. As shown in Fig. 4c, the maximum temperature in the melt pool was about 2400 K under the optimal forming parameters. The ΔG of reaction (Eq. 11) is -26.04 kJ/mol, and the ΔG of reaction (Eq. 12) is -224.50 kJ/mol. The calculated Gibbs free energy values of reactions (Equation 11 and 12) were both negative. Thermodynamic analysis showed that the reactions mentioned above could occur spontaneously, and therefore, $L1_2\text{-Al}_3\text{Ti}$ and TiB_2 may be in-situ formed before solidification. These findings were also proven through TEM analysis, as shown in Figs. 8 and 9.

4.2. Strengthening mechanism

The strengthening mechanisms of the as-fabricated Al7075-M alloy generally include the grain boundary strengthening ($\Delta\sigma_{H-p}$), dislocation strengthening ($\Delta\sigma_d$), solid-solution strengthening ($\Delta\sigma_{ss}$), and Orowan strengthening ($\Delta\sigma_{Or}$) and they may be calculated by the following equation:

$$\Delta\sigma = \Delta\sigma_{H-p} + \Delta\sigma_d + \Delta\sigma_{ss} + \Delta\sigma_{Or} \quad (14)$$

Equiaxed grains without preferred orientation, rather than columnar grains with a strong texture, were found in the Al7075-M specimen. According to the Hall–Petch formula [58], the yield strength (YS) of an alloy is strongly related to the grain size. The strengthening may be described by the following equation:

$$\Delta\sigma_{H-p} = Kd_g^{-1/2} \quad (15)$$

Dislocation strengthening ($\Delta\sigma_d$) may be calculated using the Bailey–Hirsch equation:

$$\Delta\sigma_d = M\alpha Gb\rho^{1/2} \quad (16)$$

Solid-solution strengthening occurs when other elements are alloyed with a metal matrix as solute atoms that differ from the matrix atoms in size and/or shear modulus, which can cause a variation of strain fields. Local strain fields are created that then interact with dislocations and impede their motion, leading to an increase in the YS of the material. Researchers generally accept that solid-solution strengthening is governed by the Fleischer equation [58]:

$$\Delta\sigma_{ss} = MGb\varepsilon_{ss}^{\frac{3}{2}}\sqrt{c} \quad (17)$$

When Al_3Ti and TiB_2 are uniformly distributed in the matrix phase, the YS increases because the particles act as obstacles to dislocation movement. The increase in YS can be estimated by the Orowan mechanism [59]:

$$\Delta\sigma_{or} = M \frac{0.4Gb}{\pi\sqrt{1-\nu}} \frac{\ln\left(\sqrt{\frac{2}{3}}d_p/b\right)}{\sqrt{\frac{2}{3}}d_p\left(\sqrt{\frac{\pi}{4f}}-1\right)} \quad (18)$$

where K is the strengthening effect coefficient, which is a constant of $0.17 \text{ MPa}\cdot\text{m}^{1/2}$, and d_g is the average grain size. The average sizes of Al7075 and Al7075-M measured using EBSD were 20.25 and 1.95 μm , respectively. M is the mean orientation factor (3.06 for face-centered cubic [fcc] metals), α is the constant (0.2 for fcc metals), G is the shear modulus (26.9 GPa for Al7075 alloy), b is the magnitude of the Burgers vector (0.286 nm for fcc Al) and ρ is the dislocation density, consisting of geometrically necessary dislocation (ρ_{GND}) and statistically stored dislocation (ρ_{SSD}) [58]. ε_{ss} is lattice strain [60] and c is the concentration of solute atoms, which in this study mainly consists of Zn, Mg and Cu atoms.

The contents of Zn, Mg and Cu elements in the matrix were determined by ICP-AES as 4.6, 2.75 and 1.53 wt.%, respectively. ν is Poisson's ratio (0.33 for Al7075 alloy), d_p is the radius of the precipitate and f is the volume fraction of the precipitate. The size of the Al_3Ti particles was found to be about 80 nm (Fig. 9). The volume fraction of Al_3Ti may be calculated based on the added amount.

The volume fraction of Al_3Ti was the highest, at 4.61%. The size of TiB_2 particles was about 200 nm, and the highest volume fraction was 1.81 at. %. The calculated strength increments are listed in Table 5.

Table 5.

The contribution of Hall–Petch, dislocation strengthening, solid-solution strengthening and Orowan strengthening in the YS

$\Delta\sigma_{\text{H-p}}$	$\Delta\sigma_d$	$\Delta\sigma_{ss}$	$\Delta\sigma_{\text{Or}}$
84 MPa	33.6 MPa	85.6 MPa	130 MPa

The total strengthening contribution of the printed specimens to YS was approximately 334 MPa. Obviously, the strengthening effect produced by the second phase played a fundamental role in the high YS of the specimen. It should be mentioned that, as shown in Fig. 15, the theoretical YS value agreed well with the experimental value. Elucidating the strengthening mechanism of the as-built Al7075-M alloy improved the mechanical properties and promoted the application of precipitation-hardened aluminum alloys during additive manufacturing.

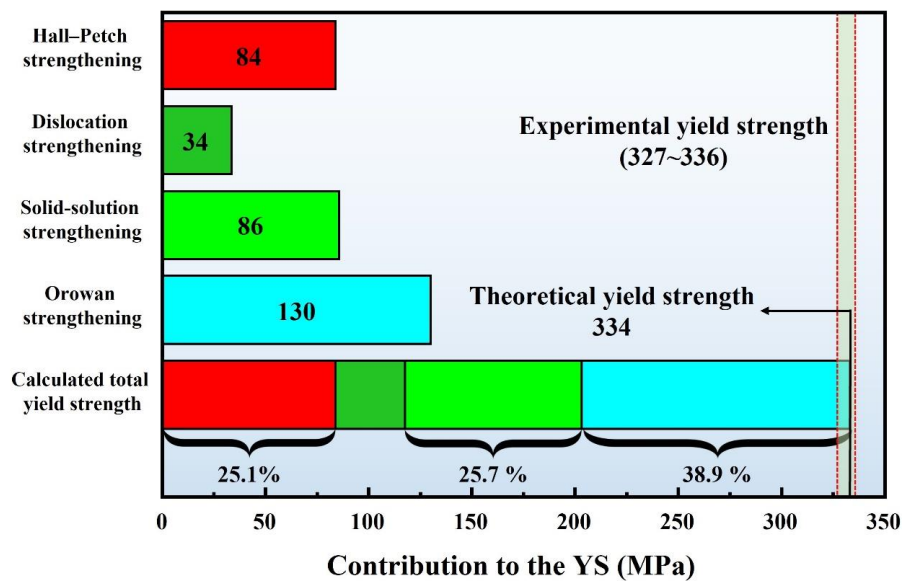


Fig. 15. Contribution of Hall–Petch strengthening, dislocation strengthening, solid-solution strengthening and Orowan strengthening to the YS

5. Conclusions

In this study, a novel approach was developed for the laser powder bed fusion (LPBF) of cracking prone Al7075 alloy. The aim of this approach was to co-incorporate submicron-sized B and micron-grade TC4 to provide an additional liquid phase and to improve fracturing toughness to resist hot cracking. This method has been proven to eliminate hot cracks and to effectively transform coarse columnar crystals into fine equiaxed crystals, thus improving the materials' mechanical properties. The main conclusions are as follows:

1. Macroscopic cracks formed in the as-fabricated Al7075 alloy, resulting in the poor ultimate tensile strength (UTS) and elongation of 47 MPa and 0.4%, respectively. The addition of 0.9 wt.% submicron B and 2.3 wt.% micron-grade TC4 particles into the Al7075 alloy through high-speed mixing eliminated the hot cracks within Al7075-M alloy. The optimal LPBF parameters for high-density specimens were found to be a hatch spacing of 120 μm and a scanning speed of 450 mm/s, with a layer thickness of 30 μm .
2. Large amounts of $\text{L1}_2\text{-Al}_3\text{Ti}$ and TiB_2 were found to have formed both inside the grains and at the grain boundaries in the Al7075-M specimen, indicating that in-situ reactions had occurred between TC4 and Al7075 and B during the LPBF process, respectively. Coherent crystal planes with small lattice mismatches formed due to the interaction between $\text{L1}_2\text{-Al}_3\text{Ti}$, TiB_2 and $\alpha\text{-Al}$, which resulted in significant grain refinement.
3. The addition of TC4 and B particles improved the mechanical properties of the Al7075 alloy. The improvement was mainly due to the combined effects of crack elimination, grain refinement and precipitation strengthening of Al_3Ti and TiB_2 . After an optimised T6 heat treatment, the UTS and YS values of the Al7075-M material were determined to be 460 MPa and 359

MPa with an high elongation of 13.0%.

Acknowledgements

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CRedit authorship contribution statement

Yanzhen Liang: Conceptualization, Methodology, Investigation, Writing-original draft. **Quanquan Han:** Conceptualization, Investigation, Writing - Original Draft, Funding acquisition. **Zhongyang Sui:** Validation, Visualization, Software. **Zhenhua Zhang:** Investigation, Validation. **Han Zhang:** Resources, Visualization. **Heng Gu:** Software. **Defan Wu:** Investigation. **Liqiao Wang:** Review & editing. **Hanlian Liu:** Conceptualization. **Rossitza Setchi:** Methodology.

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