The Role of Reinforcing Particle Size in Tailoring Interfacial Microstructure and Wear Performance of Selective Laser Melting WC/Inconel 718 Composites

In this paper, both traditional Inconel 718 parts and WC/Inconel 718 composites were fabricated by selective laser melting (SLM). The size of WC particles was observed to play a crucial role in determining the microstructural evolution, distortion, and microcracks around the WC particles, which in turn also affected the effective mechanical properties of WC/Inconel 718 composites. The use of the 5.25 μm diameter WC particles resulted in fine dendrites at the interface between the WC particle and the Inconel 718 matrix. This was attributed to the formation of an annular heat flow and radially arranged temperature gradient directions around the WC particle that increased the contact area between the matrix and the particle, thereby also improving the interfacial bonding. A sound metallurgical bonding at the interface was achieved with negligible distortion and microcracks due to a relatively uniform temperature distribution and temperature gradient (4.7 × 10⁶ °C/mm) at the interface. This also explains the generation of dense and smooth interfacial bonding, which yielded a low average friction coefficient of 0.21. The wear properties were improved since grooves and spallation were reduced with the decrease of the WC size. [DOI: 10.1115/1.4040544]

Keywords: selective laser melting (SLM), particle-reinforced metal matrix composites (PRMMCs), microstructural evolution, distortion, microcracks, wear performance.
1 Introduction

Inconel 718, as a high-strength [1], corrosion-resistant [2,3] nickel chromium-based alloy, can be used at a high temperature up to ~700 °C. The ease and economy with which Inconel 718 can be processed have favored its use in a variety of areas for its excellent tensile, creep, fatigue, and rupture strength, including industrial gas turbine, aircraft engine, and turbine blade of nuclear reactor [4–6]. However, the accelerating development of materials science and engineering leads to the emergence of new materials with greater mechanical performance, thus resulting in the loss of competitive edge of traditional materials [7]. The conventional Inconel 718 alloy, for example, cannot meet the higher requirements (e.g., higher specific strength, higher specific stiffness, and higher thrust-weight ratio of modern aero engine) in fields like aerospace and automobile, since the service environment is becoming tougher in those areas for an improved output of the engine. The preparation of particle-reinforced metal matrix composites (PRMMCs) shows the great potential to address this concern through adding certain harder and stiffer ceramic reinforcements into the metal matrix. The PRMMCs have illustrated their superiority compared with the conventional metals for significantly increased hardness, wear resistance, and high-temperature mechanical performance [8–11]. The ceramics widely used in PRMMCs include TiC [12], WC [13], CrC [14], etc. Nevertheless, the insufficient densification and inhomogeneous microstructure caused by the addition of ceramic particles are more likely to occur in PRMMCs processed via conventional technologies (like stir casting, infiltration casting, squeeze casting), which are driving the appearance of novel processing technologies to fabricate Inconel 718 composites [15].

Selective laser melting (SLM) is a newly established branch of additive manufacturing technology [16–19]. Two-dimensional cross sections of a part are selectively radiated, fused, and then solidified in a layer-by-layer manner as to create a complete 3D object according to its computer model. It shows great capability of processing a wide range of metals, alloys, and PRMMCs. Furthermore, the layer-by-layer fabrication of selective laser melting allows for expanded design freedom with minimal feedstock waste, which enables the realization of structurally sound light weight structures, especially for the aerospace field [20,21]. Nevertheless, there is a significant challenge when PRMMCs are fabricated through SLM, which is the generation of interfacial microcracks or micropores between the ceramic particle and the matrix and even the local distortion and the fragmentation around the particle resulting from local thermal/stress concentration [22]. Those interfacial defects are likely to result in the premature failure of PRMMCs during mechanical loading due to the limited particle/matrix interfacial bonding ability, thus lowering the strengthened effect of ceramic particles on composites [23].

It is well known that the particle size plays an important role in determining the strength, stiffness, and even the failure mode of PRMMCs. On one hand, decreasing the size of ceramic particles is beneficial to improve the interfacial bonding ability between the WC particle and the matrix [24]; on the other hand, the agglomeration behavior of particles due to the large Van Der Waals attractive force among fine particles is also a technical challenge when the particle size is decreased to a certain level (like nanometer level), for obtaining homogeneous microstructure and enhanced mechanical properties of composites. Therefore, selecting appropriate ceramic particle size is a key issue for SLM-processed PRMMCs for excellent interfacial bonding ability. Thanks to the nonequilibrium, rapid melting/solidification process of SLM and controllable processing condition, the microstructure and thermal/stress distribution within the molten pool can be tailored, especially around the ceramic particles, which offers the technical foundation to improve the interfacial behavior by designing the size of the ceramic particle for uniform temperature distribution, acceptable cooling rates, appropriate temperature gradient, and controlled stress distribution around the particles [25]. Therefore, the intensive research efforts are needed to explore how the microstructure is developed around the interface during SLM process. The thermal behavior, control mechanisms of defects (microcracks, distortion, micropores, etc.), and corresponding mechanical properties around the particles should also be clarified based on different sizes of ceramic particles. In this case, the rapid development of numerical simulation technology offers us an efficient method to measure the thermal behavior at the condition that the rapid heating and cooling process is difficult to be monitored during SLM experiments.

In this work, traditional Inconel 718 parts (as comparison) and WC particle-reinforced Inconel 718 composites (WC/Inconel 718) with different sizes of WC particles were prepared by SLM; a corresponding 3D finite element model was established to predict the thermal and physical behavior (e.g., temperature distribution, temperature gradient distribution, cooling rate, etc.) at the interface between the WC particle and the Inconel 718 matrix. It should be noted that the numerical simulation and SLM...
experiments were performed using same processing conditions in this work. The microstructural characteristics and interfacial defects around the particles were studied, and the mechanical performance in terms of wear resistance was assessed. The aim of this study is to explore the mechanism of microstructural evolution around the interface and the control mechanisms of interfacial defects as to fabricate WC/Inconel 718 composites with excellent interfacial bonding ability and desired comprehensive mechanical performance, which may be applicable and transferable to process other kinds of PRMMCs through laser additive manufacturing technologies.

2 Experimental/Numerical Simulation Procedures

2.1 Powder Preparation. The gas-atomized, spherical Inconel 718 powder with the particle size distribution of 15–45 μm was applied as starting material in this work. The chemical composition of Inconel 718 is 18.4 Cr, 17.7 Fe, 5.1 Nb, 4.2 Mo, 0.9 Ti, 0.3 Al, 0.08 C and Ni balance (wt %). The polygonal-shaped WC particles with the initial average size of 21 μm were selected as the reinforcement in the WC/Inconel 718 composites. The particle size distribution of those WC particles was 18–24 μm. In order to obtain the WC particles with various sizes, WC particles were milled in a Fritsch Pulverisette 6 planetary ball milling (Fritsch GmbH, Idar-Oberstein, Germany) using a ball-to-powder weight ratio of 5:1, a rotation speed of the main disk of 200 rpm, and the mixing time of 0 h, 4 h, and 8 h, respectively. As a result, three kinds of WC particles with the mean size of 21 μm (particle size distribution of 18–24 μm), 10.5 μm (particle size distribution of 7–14 μm), and 5.25 μm (particle size distribution of 3–8 μm) were obtained after the necessary screening process. Afterward, they were mixed into Inconel 718 powder homogeneously through ball milling as three experimental groups. The weight ratio of WC and Inconel 718 was 0.25:0.75. The same gas-atomized, spherical Inconel 718 powder (particle size distribution of 15–45 μm) was used in the comparison group.

2.2 Selective Laser Melting Process. Selective laser melting experiments were conducted based on a SLM-150 system, which is developed by Nanjing University of Aeronautics and Astronautics (NUAA). The system consists of an IPG Photonics Ytterbium YLR-500-WC fiber laser with a maximal output power of 500 W and a laser spot diameter of 70 μm, an automatic powder delivery apparatus, an inner argon gas protection system, and a computer system for process monitor and control. The powder layers were deposited with a thickness of 50 μm in a layer-by-layer manner and WC/Inconel 718 parts were created with dimensions of 5 mm × 5 mm × 8 mm. Based on the preliminary SLM experiments, an optimized combination of laser power (P, 125 W) and laser scan speed (v, 100 mm/s) was used in this work, focusing on the local microstructural behavior and metallurgical defects (holes, cracks, etc.) around the interface and wear properties of WC/Inconel 718 composites.

2.3 Microstructural Characterization and Tribological Tests. Samples of SLM-processed Inconel 718 and WC/Inconel 718 composites were first cut along the direction perpendicular to the laser scan direction. The cross section were then ground and polished for metallographic examinations according to the standard procedures and etched in the solution containing HCl (10 ml) and H2O2 (3 ml) with an etching time of 5 s. Microstructural features and metallurgical performance around the ceramic particles were characterized using a field emission scanning electron microscopy (FE-SEM; Hitachi S-4800, Tokyo, Japan).

Dry sliding wear tests were carried out using a HT-500 ball-on-disk tribometer under ambient temperature. A bearing steel GCr15 ball with a diameter of 3 mm and an average hardness of HRC60 was taken as the counterface material. Based on the wear tests of MMCs published in previous work [26,27], the test load was chosen to be 430 g, and the friction unit was rotated at 560 rpm for 15 min with the rotation radius of 2 mm. The coefficients of friction (COFs) were recorded in real time during the tests. After the wear tests, the worn surfaces were examined using the FE-SEM.

2.4 Numerical Simulation of Thermal Behavior During Selective Laser Melting. The numerical simulation of WC/Inconel 718 and Inconel 718 was carried out based on the ANSYS software for quantitative thermal behavior (temperature distribution, temperature gradient, cooling rate, etc.) around the interface between the WC particle and the Inconel 718 matrix during SLM. Based on the real SLM experiments [26], the 3D finite element model and laser scan strategy of SLM process used in this paper were designed, as shown in Fig. 1(a). The model consisted of a WC/Inconel 718 or Inconel 718 powder bed (595 μm × 315 μm × 50 μm) and a Inconel 718 substrate (715 μm × 415 μm × 80 μm). A tailored model of the polygonal-shaped WC particle was inserted into the powder bed, showing a distance of 14.5 μm below the top surface (point 1 is the center of the top surface of the powder bed, Fig. 1(b)). The complete morphology of WC particle is shown in the top-right of Fig. 1(b). According to the real sizes of WC particles used in this work, the model of the particle was designed into three dimensions (24 μm × 21 μm × 21 μm, 12 μm × 10.5 μm × 10.5 μm, and 6 μm × 5.25 μm × 5.25 μm, respectively). It should be pointed out that the distance of 14.5 μm was tailored in this work in order to have the largest WC particle located at the center of the powder bed and guarantee three kinds of WC particles in the simulation that can generate equal heat flux from the top surface of the powder bed for controlling experimental variables. Furthermore, three points on the path 1 were monitored to record the thermal behavior around the interface between the particle and the Inconel 718 matrix during the
SLM, and those three points located out of the particle, at the interface and in the particle, respectively, with a distance of 3 μm between each other (as shown in the bottom-right of Fig. 1(b)). It should be pointed out that path 1 is the central axis of the particle model and the direction of path 1 is parallel to the direction of Y-axis. As the comparison, same points for monitoring temperature in Inconel 718 were selected as those in the case with the 21 μm-sized WC particle model. In order to guarantee the computational precision and simulation efficiency both, the top of the powder bed was divided into ANSYS Solid70 hexahedron elements with dimensions of 17.5 μm × 17.5 μm × 12.5 μm; the tetrahedron mesh was adopted in the tailored particle, the bottom of the powder bed and Inconel 718 substrate; multiscale elements were designed at the interface for a good mesh convergence behavior in the simulation work (Fig. 1(b)). The model was meshed into 29,313 nodes and 21,635 elements in all. It should be noted that the laser scan strategy and other essential SLM process parameters (P = 125 W, v = 100 mm/s, layer thickness of 50 μm, hatch spacing of 50 μm, laser spot diameter of 70 μm, etc.) applied in experimental and numerical simulation process were same to guarantee the well-matched relationship between the experimental and numerical results.

During the SLM process, certain hexahedron elements on the top surface of the powder bed were irradiated according to the laser scan strategy. The interaction time (1.75 × 10^5 s) between those elements and the laser beam could be calculated according to the laser scan speed and the side length of the element, i.e., T_{time} = (Side length/Scan speed) = (17.5 μm/100 mm/s) = 1.75 × 10^5 s. The thermophysical parameters of Inconel 718 are shown in Table 1 [28]. The heat capacity (C_p), thermal conductivity (k), and density (ρ) of WC are 203 J/(kg°C), 63 W/(m°C), and 15,800 kg/m^3, respectively [29]. It should be noted that the thermophysical parameters of WC are influenced by the temperature. However, the fluctuation of thermophysical parameters of WC under different temperatures is weak due to the stable thermophysical characteristic of ceramic powders, so some thermophysical parameters of WC was thought to be unchanged during the SLM simulation, which was also helpful to get a good mesh convergence behavior in the simulation work. This operation was widely adopted in previous published work, and it had been certified that the influence from this operation was acceptable and negligible [29]. The specific simulation work (including the design of Gaussian heat source, the treatment of initial and boundary condition and the laser absorptivity of powder bed (0.72), etc.) was carried out on the basis of authors’ previous work which had been verified to be reliable and applicable in SLM process [28]. Specifically, the spatial and temporal distribution of the temperature field satisfies the heat conduction equation, which can be expressed as

$$ \rho C_p \frac{\partial T}{\partial t} = \frac{\partial}{\partial x} \left( k \frac{\partial T}{\partial x} \right) + \frac{\partial}{\partial y} \left( k \frac{\partial T}{\partial y} \right) + \frac{\partial}{\partial z} \left( k \frac{\partial T}{\partial z} \right) + \dot{Q} $$

(1)

where ρ is the density of the material, T is the temperature of the powder bed, C_p is the specific heat capacity, t is the interaction time between the laser beam, and the powder bed, (x, y, z) are the spatial co-ordinates, k is the thermal conductivity of powder bed and \( \dot{Q} \) is the heat generated per volume within the component. The initial condition of the model system at the time \( t = 0 \) can be defined as

$$ T(x, y, z, t=0) = T_{amb}(x, y, z) \in D \quad (2) $$

Table 1 Thermophysical parameters of Inconel 718

<table>
<thead>
<tr>
<th>T (°C)</th>
<th>20</th>
<th>100</th>
<th>200</th>
<th>400</th>
<th>600</th>
<th>800</th>
<th>1300</th>
</tr>
</thead>
<tbody>
<tr>
<td>k (W/(m°C))</td>
<td>10 12</td>
<td>14</td>
<td>17</td>
<td>20</td>
<td>26</td>
<td>31</td>
<td></td>
</tr>
<tr>
<td>c (J/(kg°C))</td>
<td>362 378</td>
<td>400</td>
<td>412</td>
<td>460</td>
<td>544</td>
<td>583</td>
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</tr>
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where \( T_{amb} \) is the ambient temperature (20°C). The thermal boundary condition for powder, liquid and solid is defined as

$$ k \frac{\partial T}{\partial n} - q + q_{con} + q_{rad} = 0 \quad (x, y, z) \in S \quad (3) $$

where S is the surfaces which are attached to input heat flux, radiation, and convection; n is the normal vector of surface S; the input heat flux q is presented in the following by Eq. (10). The heat loss \( q_{con} \) due to the natural convection of the fluid around the powder bed is defined as

$$ q_{con} = h(T - T_{amb}) \quad (4) $$

where h is the coefficient of heat transfer. The heat loss \( q_{rad} \) due to the radiation of the powder layer is

$$ q_{rad} = \sigma e (T^4 - T_{amb}^4) \quad (5) $$

where σ is the Stefan-Boltzmann constant (5.67 × 10^{-8} W/(m^2·K^4)), and e is the emissivity of powder bed which can be expressed as

$$ e = A_H e_s + (1 - A_H) e_h \quad (6) $$

where e_s is the emissivity of the solid. \( A_H \) and e_h are the area fraction of the surfaces which is occupied by the radiation-emitting holes and the emissivity of the hole, respectively.

$$ A_H = \frac{0.908 \varphi^2 - 908 \varphi^2 - 2 \varphi + 1}{908 \varphi^2 - 2 \varphi + 1} \quad (7) $$

and

$$ e_h = \frac{e_s \left[ 2 + 3.082 \left( 1 - \frac{1}{\varphi} \right)^2 \right]}{e_s \left[ 1 + 3.082 \left( 1 - \frac{1}{\varphi} \right)^2 \right] + 1} \quad (8) $$

where \( \varphi \) is the porosity of the powder bed which can be written as

$$ \varphi = \frac{\rho_s - \rho_p}{\rho_s} \quad (9) $$

where \( \rho_s \) and \( \rho_p \) are the density of the solid and powder material, respectively. The porosity is assumed to vary from \( \varphi = 0.4 \) for powder to \( \varphi = 0 \) for solid and it is taken as 0.4 in this paper.

During the SLM process, powder bed is irradiated by the laser beam and the distribution of the laser energy intensity follows nearly a Gaussian relationship, which is defined as

$$ \dot{Q} = \frac{2AP}{\pi R^2} \exp \left( -\frac{2r^2}{R^2} \right) \quad (10) $$

where A is the laser energy absorptivity of WC/Inconel 718 powder, P is the laser power, R is the radius of the Gaussian laser beam, and r is the distance from a point on the surface of the powder bed to the center of the laser beam.

3 Results and Discussion

3.1 Microstructural Evolution at the Interface. Figure 2 shows the microstructural features of SLM-processed WC/Inconel 718 composites and Inconel 718 parts (as comparison). Under the processing condition of 125 W (P) and 100 mm/s (v), in general, the sharp edges of the polygonal-shaped WC particles tended to be melted and the WC particles tended to become spheres after the interaction between the laser beam and the powder. That was because the sharp edges of the WC particles were easy to accumulate heat when they were heated [30–33], favoring significant

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smoothening and refinement of the initial WC particles. This reduced the cracking probability at the interface due to the less stress concentration. Another attractive phenomenon was that typical epitaxial columnar dendrites could be observed in cross sections, especially for Inconel 718 parts (Fig. 2(a)); while the microstructural features around the WC particles were different from the columnar dendrites, as shown in Figs. 2(b)–2(g). Fine dendritic grains around the interface between the particle and the Inconel 718 matrix could be found in SLM-fabricated WC/Inconel 718 composites, and those dendrites nucleated and grew at the interface and were almost perpendicular to the edge of the particle. This led to the formation of a gradient interface between the

![Fig. 2 SEM showing microstructural features around the interface between the particle and the matrix: (a) NON-particle, (b) particle size of 21 μm, (c) high-magnification SEM image in the rectangle of (b), (d) particle size of 10.5 μm, (e) high-magnification SEM image in the rectangle of (d), (f) particle size of 5.25 μm, and (g) high-magnification SEM image in the rectangle of (f)](https://manufacturingscience.asmedigitalcollection.asme.org)
matrix and the particle due to the complex reaction among chemical elements in composites as discussed in authors’ previous work and thus contributed to the good interfacial bonding [13]. When large-sized WC particles were used in WC/Inconel 718, well-developed dendrites formed with the trunk length of 2.85–5.42 μm. It was apparent that the dimensions of dendrites tended to be larger at sharp edges of the particle (5.42 μm) compared with those at smooth edges (2.85 μm) (Figs. 2(b) and 2(c)). With the WC size decreased to 10.5 μm, more intensive dendrites around the interface were obtained and guaranteed a hierarchical interface for better bonding ability at the interface (Figs. 2(d) and 2(e)). When the WC size decreased to 5.25 μm, the polygonal-shaped WC particle became an approximate sphere with typical dendrites arranging around the interface between the particle and the matrix. The trunk length of dendrites fluctuated around 0.89–2.13 μm with obvious secondary dendrites which were beneficial to further increase the effective contact area between the dendrites and the columnar grains in the matrix, and resultantly enhanced the bonding strength between the particle and the matrix (Figs. 2(f) and 2(g)).

Figure 3 depicts the heat flow patterns within the molten pool of SLM-processed WC/Inconel 718 composites and Inconel 718 parts. During the SLM process, the laser energy is rapidly absorbed by the powder bed through bulk-coupling and powder-coupling mechanisms when the laser beam irradiates the powder bed [34]. The black dashed line circles in temperature contour plots represent the melting temperature of Inconel 718 (1300 °C).

The temperature within the dashed line circle was higher than melting point of Inconel 718 with a small-scale molten pool forming. The width and depth of the molten pool were approximately 110 μm and 66 μm, respectively, under this process condition (P = 125 W, v = 100 mm/s). In general, the majority of heat in the powder bed was dissipated through the substrate and previous processed powder layers, leading to the temperature gradient along the building direction (Fig. 3(a)). This contributed to the nucleation and epitaxial growth of typical columnar dendrites along the opposite heat transmitting direction especially in Inconel 718 parts without the influence of WC (Fig. 2(a)). However, the WC particles could not be melted completely in the molten pool due to its higher melting temperature (1300 °C of Inconel 718 versus 2593 °C of WC); the capacity of heat transmission of WC is relatively weaker than that of Inconel 718 in high-temperature areas. Consequently, the heat flow was blocked significantly by the WC particle when heat was dissipated from the top surface, generating an annular heat flow around the particle. As a result, the directions of the temperature gradient around the WC were arranged in a radial pattern from the center of WC to the Inconel 718 matrix, as shown in Figs. 3(b)–3(d). This accounted for the establishment of dendrites which were almost perpendicular to the edge of the particle (Figs. 2(b)–2(g)). As the small-sized WC particle was used (5.25 μm), the isothermal curves through the WC particle were much more intensive than those of the matrix, compared with the curves in WC/Inconel 718 composites with large-sized particles (21 μm and 10.5 μm). This illustrated a more drastic change of the temperature between Inconel 718 matrix and the WC particle, which offered stronger thermodynamic potential for the growth of fine dendrites with distinct secondary dendrites (Fig. 3(d)).

![Fig. 3 Schematic showing the heat flow directions around the WC particle during SLM process: (a) NON-particle, (b) particle size of 21 μm, (c) particle size of 10.5 μm, and (d) particle size of 5.25 μm.](https://manufacturingscience.asmedigitalcollection.asme.org/doi/10.1115/1.4039225)

### 3.2 Distortion and Fragmentation at the Interface

Figure 4 shows the distortion and fragmentation of the WC particle within the molten pool of WC/Inconel 718 composites. Basically, the WC particle tended to be integrated without obvious distortion and microcracks when smaller-sized WC particles were used. At a large-sized WC of 21 μm, obvious bulges could be found along the interface between the particle and the Inconel 718 matrix with cracks forming at the edge between two adjacent bulges. The maximal width of those cracks could reach 0.48 μm (Figs. 4(a) and 4(b)). As the size of WC decreased to 10.5 μm, the bulges disappeared while the fragmentation developed dramatically at the edge of the previous bulges with short and intensive microcracks, which could be regarded as the expression of much more serious distortion instead of simple bulges within a WC particle (Fig. 4(c)). On further decreasing the particle size to 5.25 μm, an almost integrated particle with weak bulges was obtained, indicating a homogeneous and appropriate stress distribution around the WC particle (Fig. 4(d)).

In order to further investigate the formation mechanism of distortion and fragmentation within the WC particle, the temperature distribution and temperature gradient distribution along path 1 during SLM process of Inconel 718 and WC/Inconel 718 are illustrated in Fig. 5. The most important thing was that a sudden change of temperature distribution could be observed at the interface where a significantly high temperature gradient existed. Specifically, the temperature of Inconel 718 changed smoothly from the periphery with a maximal temperature gradient of 2.7 × 10³ °C/mm at the edge of path 1 (Fig. 5(a)). While at a large-sized WC particle of 21 μm, the temperature around the particle reduced from 1427 °C at the center of the particle to 1392 °C out of the particle. Two peaks of the temperature gradient (6.0 × 10³ °C/mm and 5.3 × 10³ °C/mm, respectively) generated at the interface between the particle and the matrix (Fig. 5(b)). It is believed that a high temperature gradient is likely to cause the stress accumulation during solidification, resulting in the bulges or cracks within the particle (Fig. 4(a)) [28]. Reducing the WC size to 10.5 μm, the temperature changed...
Fig. 4 SEM showing distortion and fragmentation of the WC particle during SLM process: (a) particle size of 21 μm, (b) high-magnification SEM image in the rectangle of (a), (c) particle size of 10.5 μm, and (d) particle size of 5.25 μm

Fig. 5 Temperature distribution and temperature gradient distribution along path 1 during SLM process: (a) NON-particle, (b) particle size of 21 μm, (c) particle size of 10.5 μm, and (d) particle size of 5.25 μm
from 1450°C to 1466°C with a decreased maximal temperature gradient of $5.0 \times 10^3$ C/mm (Fig. 5(c)). The lower temperature gradient was expected to alleviate stress accumulation and relieved the distortion at the interface; while fragmentation, the expression of severe distortion, happened in this case. At a smaller size of WC (5.25 µm), the temperature fluctuated around 1484–1490°C near the WC particle and the maximal temperature gradient reduced to a lower figure for $4.7 \times 10^3$ C/mm, which contributed to a homogeneous stress distribution around the WC particle and the resultant integrated WC particle without cracks or obvious distortion (Fig. 5(d)).

As per the force analysis around the WC particle shown in Fig. 6, the matrix experienced a relatively more obvious contraction during the solidification due to the higher coefficient of thermal expansion of Inconel 718. As a result, the WC particle had to bear compressive stress from Inconel 718 matrix, and there was tensile stress in the matrix around the WC particle. It was believed that the compressive stress from the matrix was also responsible for the distortion of the particle, such as bulges and fragmentation (Figs. 4(a)–4(c)). In addition, the poor bonding ability between the matrix and the WC particle (like pores and cracks at the interface) would increase the possibility of distortion of the particle since those defects provide space for releasing stress. Furthermore, the heterogeneous contraction due to the local high temperature gradient at the interface further aggravated the generation of distortion at the edge of the particle. The large WC particle of 21 µm owned enough volume to undertake larger compressive stress, so only some bulges and cracks formed within the particle due to the reciprocal extrusion of bulges (Figs. 4(a) and 4(b)). We termed this kind of crack “compressive stress induced microcrack.” The smaller WC particle of 10.5 µm was not large enough to bear large compressive stress, resulting in the fragmentation of the particle (Fig. 4(c)). At the smallest size of WC (5.25 µm), there were only some weak bulges which benefited from the low temperature gradient and relatively uniform thermal distribution around the interface (Figs. 4(d) and 5(d)).

3.3 Microcracks at the Interface. Figure 7 shows the micropores/voids and microcracks at the interface between the particle and the matrix. Compared with the intensive, short, and fine cracks within the WC particle in Fig. 4, microcracks in Fig. 7 were prone to single, long, and coarse cracks which always originated from the poor metallurgical bonding area at the interface and further propagated into the particle. When large-sized WC particles (21 µm) were applied in WC/Inconel 718 composites, a porous interface with poor metallurgical bonding formed, which mainly attributed to the thermocapillary instability resulting from the difference of thermal conductive ability between Inconel 718 and WC and resultant insufficient wettablity of metal liquid between the matrix and ceramic particles. Furthermore, a long crack (~10.7 µm) could be found at the interface and further propagated into the center of the particle (Fig. 7(a)). On decreasing the size of the particle to 10.5 µm, the interfacial bonding ability was improved with the number of micropores decreased. The length of cracks reduced to about 9.7 µm (Fig. 7(b)). When the particle size further decreased to 5.25 µm, the dense morphology between the WC particle and the Inconel 718 matrix was obtained. With the improvement of interfacial behavior, cracks which developed from the interface and propagated into the particle disappeared and the length of the crack reduced to ~7.23 µm (Fig. 7(c)).

It is generally believed that the nonsynchronous rapid cooling process of SLM was responsible for the reciprocal extrusion or stretch among the areas within the powder bed, thus inducing the formation of bulges or cracks in the SLM-processed parts (Figs. 4 and 7) [35]. Figure 8 illustrates the cooling rates at the interface...
between the matrix and the particle, out of the particle, and in the particle, respectively, along path 1 in order to investigate the quantitative cooling process around the interface (Fig. 1(b)). In order to provide a clearer discussion, the cooling rate at the interface was simply defined as “CRAT,” the cooling rate out of the particle was defined as “CROUT,” and the cooling rate in the particle was defined as “CRIN” in this work. It should be noted that analyzing the increments of the cooling rate from “CROUT” to “CRAT” and from “CRAT” to “CRIN” was much more meaningful than the analysis of the cooling rates themselves for a thorough understanding of the nonsynchronous cooling process at the interface. Basically, the increment of the cooling rate from “CROUT” to “CRAT” (9.53%) was lower than that from “CRAT” to “CRIN” (11.03%) in Inconel 718; while the increments of the cooling rate from “CROUT” to “CRAT” were obviously larger than those from “CRAT” to “CRIN” in WC/Inconel 718 composites due to the relatively weaker capability of heat transmission of WC under high-temperature environment. Specifically, when the particle size reduced from 21 μm to 5.25 μm, the increments of the cooling rate from “CROUT” to “CRAT” decreased from 9.31% to 1.50%, and the increments of the cooling rate from “CRAT” to “CRIN” reduced from 7.50% to 1.33%. The more rapid change of the cooling rate between “CROUT” and “CRAT” indicated a much more significant nonsynchronous cooling process out of the particle, indicating higher cracking possibility there. As analyzed in Sec. 3.2, tensile stress generally formed around the particle. The poor interface with metallurgical defects (like micropores, voids, and so forth) had higher cracking possibility under the impact from tensile stress due to the stress concentration within those defects, which accounted for the reason why the position of crack sources always located at the matrix near the interface in Figs. 7(a) and 7(b). The preferable ductility of Inconel 718 helped to prevent the crack growth into the matrix while fragile ceramic particles were easy to crack. Therefore, under the synergistic impact from tensile stress in matrix and the fragility of ceramics, microcracks were more likely to propagate into the particle in this case. We termed this kind of crack “tensile stress-induced microcrack.” When the small WC particles were used in WC/Inconel 718 composites, the relatively slower change of cooling rates obviously alleviated the cracking possibility and only fine cracks around the edge of the particle could be found, indicating the formation of interface with relatively sounder bonding between the WC particle and Inconel 718 matrix (Fig. 7(c)).

3.4 Wear Properties at the Interface. Figure 9 depicts the COFs of SLM-produced Inconel 718 parts and WC/Inconel 718 composites with different-sized WC particles. In general, the COFs all showed severe fluctuation at the start of the wear tests. This was attributed to the relatively rougher surface which reduced the contact area between the top surface of the parts and the wear counterparts. As the sliding friction process continued, the fluctuation behavior of the COFs was alleviated and became leveled off due to the reduced roughness of the surface. In addition, it could be observed that the duration of the fluctuation of COFs became shortened obviously when the smallest WC particles were used. This could be attributed to the decrease of the surface roughness due to the improved densification behavior at the interface between the Inconel 718 and the WC particle (fewer pores, fewer microcracks, and less distortion). As the size of the particle decreased from 21 μm to 5.25 μm, the average COF decreased from 0.40 to 0.21. By contrast, we could find that the average value of Inconel 718 (0.43) was higher than those of WC/Inconel 718 composites, showing the improvement of wear properties induced by WC particles.

In the interest of having an underlying understanding of the tribological behavior, the detailed morphologies of the worn surfaces of Inconel 718 parts and WC/Inconel 718 composites are performed in Fig. 10. Higher magnifications on specific locations in structures were also investigated and showed in Fig. 11 in order to study the inherent wear mechanisms. It was clear that the worn morphologies and wear mechanisms varied greatly from the applied size of the WC particle. As the comparison, large area of spallation and enormous fine debris could be found on the worn surface of Inconel 718 parts, showing the combination of adhesive wear and abrasive wear (Fig. 10(a)). High magnification showed obvious debris and cracks on the worn surface which depicted the fracture on the surface (Fig. 11(a)). When large-sized WC particles (21 μm) were used, severe material delamination and spallation with massive cracks could be observed on the worn surface, increasing the surface roughness and thus leading to the severe fluctuation of COFs. This also indicated the poor wear resistance induced by the fracture of the strain-hardened tribolayer (Fig. 10(b)). The high magnification showed massive and intensive pits after the material spallation in wear tests, which were the typical characteristics of adhesive wear (Fig. 11(b)). When the size of the particle reduced to 10.5 μm, a lot of fine and small wear debris could be found, and the WC particles were covered by the debris (Fig. 10(c)). Closer investigation exhibited that the size and shape of the debris was extremely diverse and some of them clumped under pressure, reflecting the characteristics of adhesive wear in some areas (Fig. 11(c)). This was believed to be responsible for the fluctuation of the COFs. On further decreasing the particle size to 5.25 μm, the worn surface of WC/Inconel 718 composites became smooth and, a continuous, adherent, strain-hardened tribolayer could be observed without any obvious
Fig. 10 SEM images showing worn surface morphologies on Inconel 718 parts and WC/Inconel 718 composites: (a) NON-particle, (b) particle size of 21 μm, (c) particle size of 10.5 μm, and (d) particle size of 5.25 μm

Fig. 11 High-magnification SEM images showing characteristic worn surface morphologies on Inconel 718 parts and WC/Inconel 718 composites in (a) Fig. 10(a), (b) Fig. 10(b), (c) Fig. 10(c), and (d) Fig. 10(d)
fracture or local plowing except some parallel fine-grained grooves (Fig. 10(d)). This kind of morphology explained the weak fluctuation of COFs in the wear tests and showed that the dominant wear mechanism changed to be partially abrasion wear. At a magnified state, the worn surface was smooth and continuous without apparent fracture and apparent material spallation or debris (Fig. 11(d)).

In order to further investigate the effects of WC particles on wear properties, the morphologies of the WC particle after the wear tests were obtained in Fig. 12. It was obvious that the worn morphologies varied obviously from the applied WC particles. Elongated particles could be observed on the worn surface as large WC particles (21 μm) were applied due to the impact from bearing steel ball in wear tests and, the occurrence of some microcracks in the center of the particle showed the potential of brittle fracture of the particle. These reflected the poor strengthened effect of large WC particles on WC/Inconel 718 composites and resultant inferior wear behavior and a high average COF (Fig. 12(a)). When relatively smaller WC particles (10.5 μm) were used, the obvious distortion of particles disappeared except some microcracks at the interface between the Inconel 718 matrix and the WC particle, leading to the improvement of interfacial bonding ability (Fig. 12(b)). It was also responsible for the improved wear properties and the lower average value of COFs. On further decreasing the particle size to 5.25 μm, fully dense interface with sound bonding morphology was achieved in SLM-fabricated samples, contributing to the superior wear properties and a lowest mean COF value in WC/Inconel 718 composites (Fig. 12(c)).

On the other hand, it should be pointed out that there was a close relationship between the wear properties and interfacial bonding based on the comparisons between the morphologies of the WC particle before and after the wear tests. Enormous bulges and cracks within the WC particle were partially responsible for the formation of debris and spallation as well as poor wear properties since they cannot undertake and deliver the loading effectively during the wear tests. This also explained the high average value of COFs when large-sized WC particles were used in the WC/Inconel 718 composites. In contrast, as the particle size decreased to a small value, the interface tended to be free of cracks, distortion, and pores. This was beneficial for particles to undertake, deliver, and disperse the loading from the counterparts between the matrix and the particle during the tests. In addition, the intensive and homogeneous distribution of fine WC particles within the matrix provided effective support to undertake the loading. Obviously, above descriptions were the main contributions of improved wear properties of 5.25 μm WC particle-reinforced Inconel 718 composites. Furthermore, the stress accumulation at interface induced by the drastic change of temperature gradient further aggravated the formation of cracks, distortion, and resultant poor bonding ability at the interface. Therefore, it could be concluded that the addition of 5.25 μm-sized WC particles in Inconel 718 was an effective way to enhance the wear properties, showing the desired strengthened effect of WC particles on WC/Inconel 718 composites.

4 Summary and Conclusions

(1) The microstructure of SLM-processed WC/Inconel 718 composites could be influenced by the addition of WC particles. When small WC particles were used in the composites, fine dendrites which were almost perpendicular to the edge of the particle formed under the impact from the annular heat flow and resultant radial-arranged temperature gradient directions around the WC. This could increase the contact area between the matrix and the particle and thus improve the interfacial bonding.

(2) Two kinds of microcracks formed within the WC particle. First, due to the more obvius contraction of Inconel 718 during the cooling process, the particle undertook the compressive stress from the matrix. This caused the formation of bulges and short and intensive microcracks within the WC particle due to the reciprocal extrusion of bulges. We termed this kind of crack “compressive stress induced microcrack.” Another kind of microcrack originated from the irregular pores at the interface and propagated into the particle, mainly because the preferable ductility of Inconel 718 helped to prevent the crack growth while fragile ceramic particles were likely to crack. We termed this kind of crack “tensile stress induced microcrack.”

(3) The appropriate thermal behavior and resultant uniform stress distribution promoted the formation of a sound metallurgical bonding between the WC particle and the Inconel 718 matrix when the small WC particle (5.25 μm) was used in the composites. This improved the mechanical properties of WC/Inconel 718 composites, showing a uniform
distribution of the friction coefficients with a considerably lower average value of 0.21 in wear properties.

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