

Article

Surface Morphology, Relative Density, Microhardness and Microstructure of Tungsten Fabricated by Laser Powder Bed Fusion

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Abstract

This study investigates the effects of laser power and scanning rate on the surface morphology, relative density, microhardness and microstructure of pure tungsten fabricated by laser powder bed fusion (LPBF). Increasing the laser power or decreasing the scanning rate effectively suppresses spheroidisation and enhances densification, achieving a maximum relative density of ~98%. However, excessive laser power intensifies Marangoni convection, leading to surface protrusions that reduce density. Microstructural analysis reveals that the laser-scanned surface is dominated by fine columnar grains (390–480 HV), whereas the side surface comprises coarser columnar grains with lower hardness (~390 HV). Electron backscatter diffraction analysis confirms that the side surface contains a high proportion of grains exceeding 100 μm and reveals a significant peak (~41.8%) at ~3.5° for low-angle grain boundaries, indicating substantial internal stress and microstrain. Pole figures show a weak preferred orientation (maximum texture intensity of 3.161). Phase analysis shows no significant phase transformation after LPBF, while internal stress and microstrain increase notably.

Keywords: laser powder bed fusion; tungsten; surface morphology; relative density; microstructure

1. Introduction

Tungsten exhibits the highest melting point among all metals, along with high density, high hardness, excellent thermal conductivity and outstanding corrosion resistance [1]. Owing to these merits, tungsten has been extensively utilised in the military, chemical, medical, aerospace, aviation, nuclear and electronic industries. Moreover, tungsten effectively absorbs and shields high-energy radiation. Its high strength and low thermal expansion coefficient ensure the high dimensional accuracy and stability of tungsten-formed components during service. Accordingly, tungsten is regarded as an irreplaceable material in fabricating components for advanced medical imaging devices such as computed tomography (CT) and positron emission tomography (PET)–CT systems. A core component of CT detectors is the anti-scattering grid (ASG), which eliminates Compton-

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scattered X-rays through a series of crisscrossing thin walls of pure tungsten, thereby considerably enhancing the quality and spatial resolution of CT images [2,3]. Currently, the tungsten sheets of ASGs are unidirectionally arranged at different angles in CT systems, whereas in PET–CT systems they are interlaced to form a mesh structure with square straight pores. Both types of ASG are fabricated by rolling sintered tungsten sheets to a specific thickness, followed by cutting and assembly. As science and technology advance and the demand for medical techniques increases, ASGs in future medical CT systems are expected to contain complex and intricate mesh structures with angled and fine pores, where the thin tungsten walls are bidirectionally arranged at specific angles. To promote the future development of ASGs, research should focus on designs with smaller pore sizes and thinner wall thicknesses, which cannot be achieved using conventional forming processes [4,5]. ASGs with such complex and intricate structures can potentially be fabricated using metal additive manufacturing technologies, particularly the laser powder bed fusion (LPBF) technique [6].

The dimensional accuracy, powder adhesion, density and microhardness of pure tungsten thin-walled components fabricated by LPBF have already been demonstrated in previous studies, providing a theoretical foundation for the LPBF-based production of ASGs with intricate porous, thin-walled structures [7]. During the LPBF process of pure tungsten ASG, it is essential to achieve the fabrication of not only complex and intricate porous, thin-walled structures but also dense solid structures. Since the ASG is often subjected to high-speed rotation during service, the porous thin-walled and dense solid structures should possess high density and sufficient mechanical property to remain undeformed and maintain high precision during high-speed rotation. However, the relationship between the ability to manufacture thin-walled structures and the production of fully dense bulk parts is not straightforward, as these involve distinctly different melting and solidification mechanisms. Thin-wall fabrication relies on single-track or multi-track scanning with minimal overlap, where heat dissipation is predominantly lateral, whereas bulk part fabrication involves complex multi-layer, multi-track thermal cycles with significant re-melting and heat accumulation. Understanding this distinction is critical for optimising the LPBF process for ASG manufacturing, yet systematic studies addressing both aspects remain limited.

Numerous studies have been conducted on the microstructural characteristics, mechanical properties and densification of LPBF-fabricated pure tungsten components [8]. Zhang et al. [9] investigated the influence of laser processing parameters on the microstructural characteristics of LPBF-fabricated pure tungsten parts, showing that increasing the laser power and decreasing the scanning rate led to denser pure tungsten samples. Zi et al. [10] compared the LPBF processes of spherical and ordinary tungsten powders, revealing that under identical process conditions, samples fabricated using spherical tungsten powders—with superior fluidity and bulk density as raw materials—exhibited higher density, lower porosity and fewer cracks than those fabricated using ordinary tungsten powders. Wang et al. [11] explored the LPBF process of pure tungsten parts and obtained pure tungsten samples with a relative density of 96%, an average grain size of 5.12 μm and a microhardness of 3.79 GPa. Similarly, Sidambe et al. [12] studied LPBF-fabricated pure tungsten parts and reported that the formation of microcracks and defects was primarily caused by residual stress at macroscopic and microscopic scales within the samples. Microstructural analysis further demonstrated that grain growth with preferential orientation was governed by an epitaxial growth mechanism. The present study focuses on laser power and scanning rate, as these are the most accessible parameters and their combined effect ($LED = P/v$) directly controls the melt pool behaviour [13,14].

Despite these advances, the specific gap in the literature that the present study aims to address is the systematic investigation of how laser power and scanning rate—within

a carefully selected processing window—affect the surface quality, densification, and microstructural evolution of LPBF-fabricated pure tungsten bulk samples, with particular emphasis on the formation and suppression mechanisms of surface spheroidisation, protrusions and internal pores. The rationale for focusing on laser power and scanning rate is twofold: (1) these two parameters are the most accessible and adjustable on commercial LPBF systems, and their combined effect, expressed as linear energy density ($LED = P/v$), directly determines the melt pool temperature and solidification behaviour; and (2) the selected ranges (120–200 W for power and 200–400 mm/s for scanning rate) were chosen based on preliminary experiments to cover the transition from insufficient melting (resulting in porosity) to excessive energy input (resulting in keyhole-mode defects and spatter-ing).

The present study addresses the gap in the literature by systematically investigating how laser power and scanning rate—within a carefully selected processing window (120–200 W and 200–400 mm/s)—affect the surface quality, densification, and microstructural evolution of LPBF-fabricated pure tungsten bulk samples, with particular emphasis on the formation and suppression mechanisms of surface spheroidisation, protrusions and internal pores. The rationale for focusing exclusively on these two parameters is that they are the most accessible and adjustable on commercial LPBF systems, and they directly govern melt pool behaviour. The selected ranges were chosen based on preliminary experiments to cover the transition from insufficient melting to excessive energy input. Through this systematic investigation, the relationships among processing parameters, microstructural characteristics and mechanical properties are established to enhance the performance of LPBF-fabricated tungsten.

2. Materials and Methods

Experiments were performed on pure tungsten powders (purity 99.9%) supplied by Tekna Advanced Materials Inc. (Sherbrooke, QC, Canada). The samples were fabricated using an EOS M290 industrial metal additive manufacturing system (EOS GmbH, Krailling/Munich, Germany). The system is equipped with a 400 W Yb-fibre laser with a spot diameter of 100 μm . The powder morphology was observed using a Hitachi TM3030 desk-top tungsten-filament scanning electron microscope (SEM, Hitachi, Ltd., Tokyo, Japan) and the particle-size distribution was measured using an LS-POP (6) laser particle-size analyser (OMEC Instruments Co., Ltd., Zhuhai, China). As shown in Figure 1, the pure tungsten powders exhibit an obviously smooth, spherical morphology adhered with a limited number of fine satellite particles. The particle-size distribution of the powders ($D_{10} = 17.4 \mu\text{m}$, $D_{50} = 25.9 \mu\text{m}$ and $D_{90} = 32.8 \mu\text{m}$; average particle size $28.3 \mu\text{m}$) demonstrates that ~80% of powders are smaller than 30 μm , as shown the blue line in Figure 1b; very few particles are larger than 40 μm .

The thickness of the pre-laid powder layer was set to the minimum allowed by the equipment (20 μm). As most of the raw tungsten powder particles were larger than 20 μm in diameter, the first ten layers were remelted three times by the laser during the LPBF process. Remelting improved the surface flatness and reduced the height of the surface protrusions, effectively avoiding collisions between the scraper and surface of the as-fabricated sample during the pre-laid powder-spreading process. The LPBF-fabricated samples were sized 5 mm (length) $\times$ 5 mm (width) $\times$ 5 mm (height). The laser-scanning path, as shown by the arrows in Figure 2, followed the cross-hatching scanning strategy, rotating the laser-scanning direction by 67° between layers. During the LPBF process, the overlap width between two adjacent laser-scanning tracks was set to 0.05 mm. The effects of laser powers and laser-scanning rates on the surface morphologies, microstructure evolution and mechanical performance of the LPBF-fabricated pure tungsten bulk specimens were explored. Previously, it was found that specimen formability requires a linear laser-

energy density (ratio of laser power to laser-scanning rate) of 0.34 J/mm or higher [7]. Here, the laser parameters for forming the 20 μm thick pre-laid powder layer were divided into two distinct groups. In the first group, the laser power was varied (120, 140, 160, 180 and 200 W) at a constant laser-scanning rate (300 mm/s). In the second group, the laser-scanning rate was varied (200, 250, 300, 350 and 400 mm/s) under a constant laser power (160 W). The rationale for this experimental design is as follows: the constant scanning rate of 300 mm/s was chosen as the baseline from our previous work on thin-wall fabrication [7], while the constant laser power of 160 W was selected because it produced the best surface quality in the power-variation experiments. These choices allowed us to isolate the individual effects of each parameter while maintaining practical relevance to ASG manufacturing. Figure 3 is a macroscopic photograph of the LPBF-fabricated pure tungsten bulk samples.

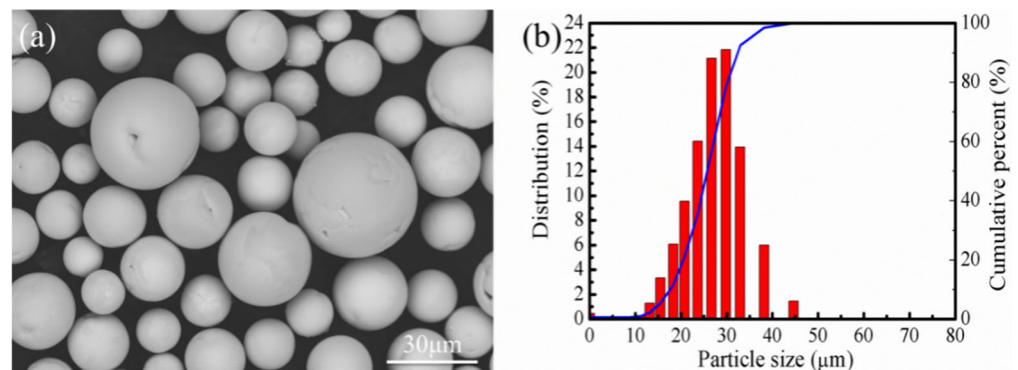


Figure 1. (a) SEM image of pure tungsten powders showing their spherical morphology; (b) particle-size distribution of the pure tungsten powders.

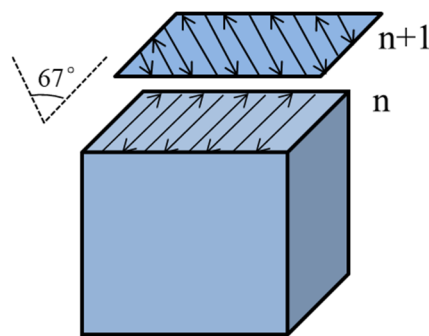


Figure 2. Schematic of the laser-scanning path during the LPBF process. The arrows represent the laser-scanning path of each layer.

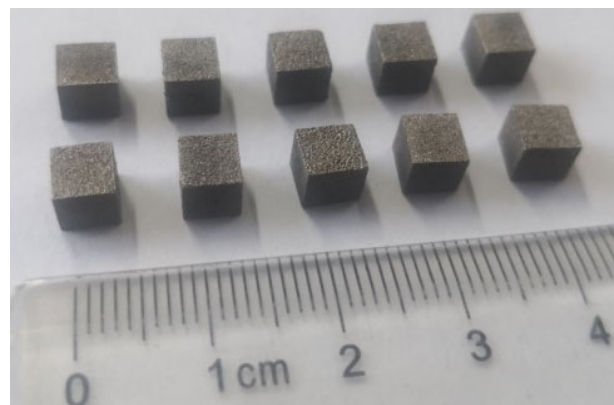


Figure 3. Macroscopic photograph of the LPBF-fabricated pure tungsten bulk samples.

The laser-scanned surface morphologies of the LPBF-fabricated pure tungsten bulk samples were observed using the Hitachi TM 3030 SEM. To prevent the effect of unmelted raw tungsten powders on the surface morphologies, the specimens were ultrasonically cleaned in alcohol for 30 min before SEM observation. The porosity fraction of the sample surface in the SEM image was statistically analysed using the ImageJ 1.54p software. The densities of the LPBF-fabricated bulk samples were measured using the Archimedes draining method. To improve the accuracy of the measurements, all samples were polished following the same procedure to remove surface-attached powder layers and open porosity before density measurement. The density of each sample was measured five times, and the measurements were averaged to give the final result. The pore distributions on the sample surfaces were examined on the Hitachi TM 3030 desktop SEM. Prior to pore observation, the specimen surfaces were sequentially ground with 200#, 400#, 800#, 1200#, 1600# and 2000# abrasive papers, followed by polishing with 1 μm and 0.5 μm polishing agents on nylon cloth. The grain morphologies of the laser-scanned and side surfaces of the specimens were observed under a Zeiss optical microscope (Zeiss, Oberkochen, Germany). Prior to observation, the polished sample surfaces were chemically etched with a pure tungsten etching solution ($\text{NaOH}:\text{K}_3[\text{Fe}(\text{CN})_6]:\text{H}_2\text{O}$; mass ratio 1:1:10) for 10 min, sequentially cleaned with deionised water and alcohol, and finally air-dried. The microhardness values of the laser-scanned and side surfaces of the specimens were obtained on a HXD-1000TMC/LCD digital microhardness tester (Shanghai Optical Instrument Factory, Shanghai, China) under an applied load of 1.961 N for a dwell time of 15 s. This load was chosen according to the ASTM E384-22 standard [15] to ensure that the indentation diagonal length was at least 20 μm , thereby minimising the influence of surface roughness and individual microstructural features on the measured hardness values. Each test surface was ground, polished, and subjected to five microhardness measurements at different sites. The phase compositions of the specimens were derived from X-ray diffraction (XRD) patterns acquired by an X'Pert Pro X-ray diffractometer (Malvern Panalytical B.V, Almelo, The Netherlands) with a Cu target, a tube voltage of 49 kV and a tube current of 40 mA. The angle was scanned from 30° to 90° at 6°/min. Electron backscatter diffraction (EBSD) analysis was performed on an EDAX-TSL EBSD detector produced by EDAX Inc. (EDAX Inc. Mahwah, NJ, USA). The polished specimens were further vibration-polished for 6 h with 0.05 μm colloidal silica suspension to eliminate surface scratches and residual stress. Tests were conducted at 20 kV accelerating voltage, 15 mm working distance and 1 μm step size, covering an 800 $\mu\text{m} \times 600 \mu\text{m}$ area on the side surfaces of the sample. Raw data were processed with OIM Analysis 7.2 software. Texture was calculated via harmonic series expansion ($L = 16$, Gaussian smoothing 5.0°), and grain boundaries with misorientation $> 2^\circ$ were identified, with $> 22^\circ$ defined as high-angle grain boundaries for statistics. XRD and EBSD analyses were performed on the sample fabricated at $P = 160 \text{ W}$ and $v = 300 \text{ mm/s}$, as this sample exhibited a relatively balanced combination of surface quality, density, and hardness, making it suitable for phase and crystallographic analyses.

3. Results and Discussion

3.1. Surface Morphology Analysis

Figure 4 shows the laser-scanned surface morphologies of LPBF-fabricated pure tungsten bulk samples produced under different laser powers at a laser-scanning rate of 300 mm/s. For the tungsten sample fabricated under the lowest laser power (120 W), the continuity of the melt tracks was poor and the bonding between adjacent laser tracks was insufficient, resulting in spheroidisation and the appearance of numerous surface pores (Figure 4a). The porosity fraction is approximately 11.915%. The microstructural evolution and properties of metallic materials during the LPBF process largely depend on their in-

trinsic physical and chemical properties, particularly the melting point, thermal conductivity, surface tension and dynamic viscosity in the molten state [16]. When the melting point of the metal powders is relatively high compared with the laser-energy input, the molten powders exhibit high viscosity and poor spreadability. Consequently, the molten metal flow cannot easily fill the gaps between the pre-laid powders, leading to pore formation. In contrast, when the metal powders possess high thermal conductivity, the heat within the molten pool is readily transferred to the surrounding metal powders, resulting in a relatively short solidification time. If the surface tension of the molten metal is large, the spreading process becomes slow and the time required for spreading exceeds the time for solidification. Consequently, the molten metal tends to spontaneously shrink into spheres rather than spread uniformly during solidification. For pure tungsten melt, which has a high density and large surface tension near the melting point, the slow spreading requires a prolonged time to achieve complete spreading. Accordingly, LPBF-fabricated pure tungsten samples are more prone to spheroidisation than other metallic materials. These observations are consistent with previous studies on single laser track formation in tungsten [17,18], which demonstrated that the balling phenomenon in tungsten is primarily caused by the high surface tension and the short solidification time relative to the spreading time. The comparison between our observations and those from single-track studies confirms that the mechanisms identified at the single-track level scale up to multi-track and multi-layer bulk samples.

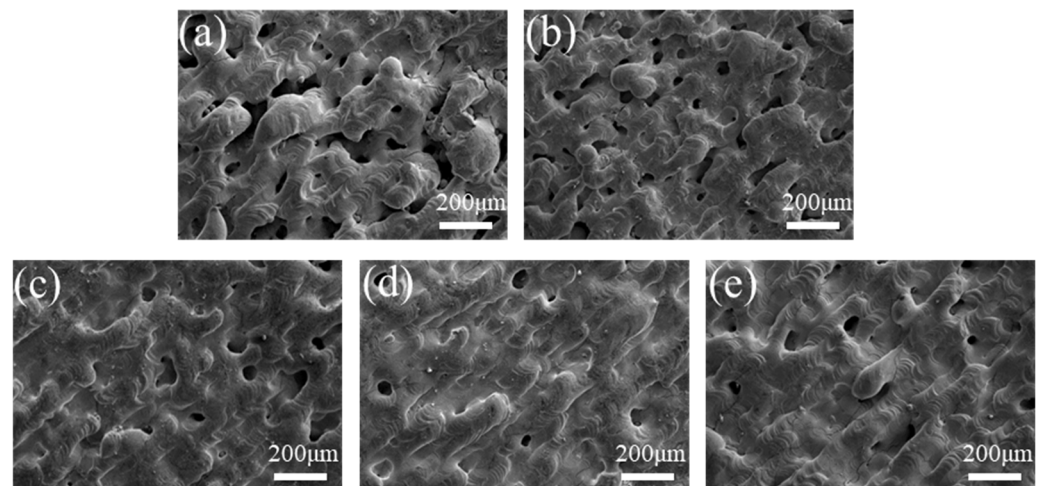


Figure 4. Laser-scanned surface morphologies of LPBF-fabricated pure tungsten bulk samples fabricated with a laser scanning rate of 300 mm/s and different laser powers: (a) 120 W, (b) 140 W, (c) 160 W, (d) 180 W and (e) 200 W.

Besides the intrinsic physical properties of metallic materials, the temperature of the molten droplets influences the spreading process. High temperatures lower the surface tension and dynamic viscosity of the molten droplets, enhancing their fluidity and spreading ability and mitigating the spheroidisation phenomenon to a certain extent. Young's equation relates the wetting angle θ of a molten droplet on a substrate surface to the solid–liquid surface tension (σ_{SL}), the solid–gas surface tension (σ_{SG}) and the liquid–gas surface tension (σ_{LG}) as follows:

$$\sigma_{SG} - \sigma_{SL} = \sigma_{LG} \cos \theta \quad (1)$$

Figure 5 illustrates the spreading behaviour of a pure tungsten droplet on a previously solidified layer surface. At wetting angles between 0° and 90° , the molten droplet exhibits good wetting and spreading behaviour and both its wetting and spreading abili-

ties improve with increasing θ (Figure 5a). A molten droplet with a wetting angle infinitesimally close to 0° is maximally wettable and can completely spread over the previously layered surface. In contrast, as presented in Figure 5b, a molten droplet with a wetting angle between 90° and 180° exhibits poor wetting and spreading behaviour and both its wettability and spreading abilities diminish with increasing θ (Figure 5b). A molten droplet with a wetting angle infinitesimally close to 180° is minimally wettable and cannot spread over the previously layered surface. The value of $\cos\theta$ in Equation (1) varies between 1 at $\theta = 0^\circ$ and -1 at $\theta = 180^\circ$. To ensure the equality given by Equation (1), σ_{LG} should be minimal when $\cos\theta$ is maximum, indicating that the surface tension of the molten droplet must be minimised to maintain good wettability and spreading behaviour.

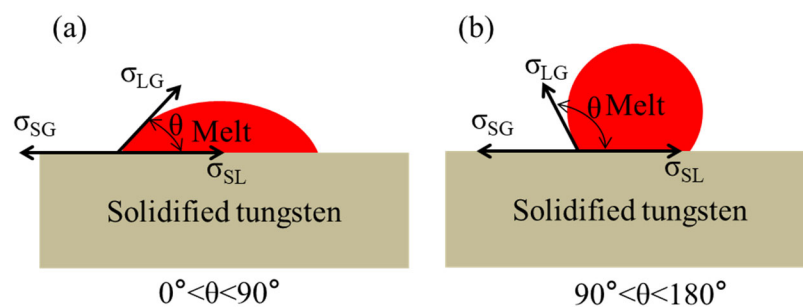


Figure 5. Schematic of spreading behaviour of a pure tungsten droplet on a previously solidified layer surface: (a) $0^\circ < \theta < 90^\circ$; (b) $90^\circ < \theta < 180^\circ$. σ_{LG} , σ_{SG} and σ_{SL} denote the liquid–gas, solid–gas and solid–liquid surface tensions, respectively.

Because the surface tension of molten pure tungsten is inversely proportional to the corresponding temperature, the molten droplet should remain sufficiently heated to lower its surface tension. During the LPBF process, the input laser energy can be enhanced by increasing the laser power or decreasing the laser-scanning rate. The melt temperature is then increased to reduce the melt surface tension σ , achieving superior wettability and spreading behaviour of the melt on the previously layered surface.

Figure 4b shows the laser-scanned surface morphology of the pure tungsten bulk sample fabricated under a laser power of 140 W. The porosity fraction decreases to approximately 8.640%. Moreover, the melt tracks become continuous with tight overlaps between adjacent tracks, while the spheroidisation phenomenon is remarkably reduced. This improvement can be attributed to the increased laser power during the LPBF process, which heats the pure tungsten powders to a temperature above their melting point, thereby reducing surface tension and viscosity and enhancing the wettability and spreading behaviour of the pure tungsten droplet on the substrate surface. In addition, high temperatures prolong the solidification time of the molten droplets, providing sufficient time for spreading, flowing and wetting. In the LPBF fabrication process of metallic materials, which involves layer-by-layer melting of pre-laid powders by laser scanning, interstitial voids are inevitably left between the pre-laid powders and must subsequently be filled by the flow of molten droplets. Under low laser power, when the molten droplets are insufficiently fluid (Figure 4a), numerous pores and spheroidisation appear on the laser-scanned surface. When the laser power is increased, the temperature of the molten droplets rises (Figure 4b), enhancing their fluidity and spreadability and allowing sufficient time for droplet flow and molten-pool spreading. Consequently, the gaps between the metal powders during the pre-laid powder process are effectively filled by the flowing metal melt, thereby decreasing the number and size of pores. Furthermore, the improved wettability and reduced surface tension enable the molten droplets to spread easily across the solidified pure tungsten surface, thus suppressing spheroidisation to a certain extent.

When the laser power further increased to 160 W, the laser-scanned surface of the LPBF-fabricated pure tungsten bulk sample becomes smooth with good overlap between the adjacent melt tracks. Spheroidisation is essentially non-existent and few pores are observed (Figure 4c). The porosity fraction decreases further to approximately 7.130%. When the laser power reaches 180 W, the surface becomes even smoother with the lowest porosity among all samples (Figure 4d). The porosity fraction drops significantly to approximately 1.817%. This improvement is attributable to high energy absorption by the pure tungsten powder under high laser power. As the powder reaches a higher temperature, it yields a pure tungsten melt with low surface tension and viscosity and superior wettability and spreading ability, facilitating the fusion of individual pure tungsten droplets into a continuous molten pool that fills the gaps between the pre-laid powders through flow and wetting processes, thereby decreasing the internal porosity of the LPBF-fabricated pure tungsten sample. Meanwhile, the prolonged solidification time allows complete spreading of the molten droplets. When the molten droplets are nearly fully spread and solidified, the spheroidisation phenomenon is effectively reduced.

When the laser power is further increased to 200 W, the laser-scanned surface morphology of the LPBF-fabricated pure tungsten bulk sample exhibits protrusions and an increased number of pores (Figure 4e). The porosity fraction increases to approximately 4.603%. As discussed in the following density analysis (Section 3.2), the sample fabricated at 200 W is less dense than the one fabricated at 180 W. The schematic in Figure 6 illustrates the formation of protrusions on the surface of an LPBF-fabricated sample scanned at an excessively high laser power. The laser beam irradiates the pre-laid pure tungsten powders with a Gaussian energy distribution. Moreover, the thermal conductivity differs between the molten pool and its surroundings. Specifically, the powders have lower thermal conductivity than the solidified material, resulting in a lower molten-pool temperature near the solidified region and a higher molten-pool temperature near the powder region. Consequently, a temperature gradient forms within the molten pool. An increase in temperature lowers the surface tension and dynamic viscosity of the pure tungsten melt. Accordingly, the surface tension and viscosity of the molten pool are high near the solidified region and low near the pure tungsten powder region. Under these conditions, the melt at the centre of the molten pool tends to flow towards the solidified region, forming Marangoni convection [19]. Surface protrusions are then generated after the molten pool solidifies. An excessively high laser power produces a large temperature gradient in the molten pool, intensifying the Marangoni convection. When the protrusions formed by such convection cannot be eliminated by the molten-pool flow, they remain on the surface of the solidified sample, reducing the density and uniformity of the subsequent pre-laid powder layer.

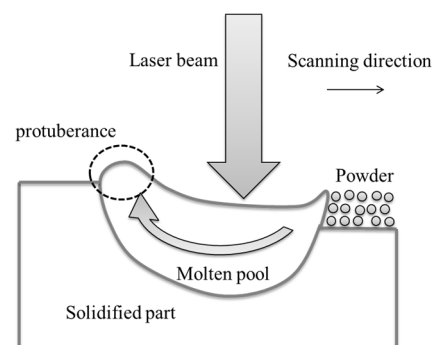


Figure 6. Schematic showing the formation of protrusions during LPBF.

Figure 7 shows the laser-scanned surface morphologies of the pure tungsten bulk samples fabricated at different LPBF scanning rates. Under a constant laser power (160 W

in this case), the laser-scanning rate can affect the surface morphologies of the samples in two main ways. First, it can determine whether the laser irradiation time is sufficient to allow surface-to-centre heat transfer among the individual tungsten powder particles, ensuring complete melting of the tungsten powders. Second, it can determine whether the laser-energy density can adequately heat the metal powders to a sufficiently high temperature, ensuring low surface tension and dynamic viscosity of the molten pool and allowing sufficient time for the flowing and filling of gaps between adjacent melt tracks.

During the LPBF process, the heat generated by laser irradiation on the metal-powder surfaces is transferred to the metal-powder interior. The time ΔT required for the entire powder mass to reach the same temperature depends on the thermal diffusion coefficient α of the powders and the powder particle size D [20]:

$$\Delta T \approx D^2/4\alpha \quad (2)$$

Meanwhile, the thermal diffusion coefficient α is associated with the thermal conductivity λ (W/m·K), density ρ (kg/m³) and the specific heat capacity c (J/(kg·K)) of the material as follows:

$$\alpha = \lambda/(\rho c) \quad (3)$$

The thermal conductivity, specific heat capacity, and density of tungsten are ~160 W/m·K, 132 J/(kg·K), and 19,350 kg/m³, respectively, giving a thermal diffusion coefficient (α) of 6.28×10^{-5} m²/s for pure tungsten. As the heat generated by laser irradiation on the metal-powder surface transfers to the interior of the metal powders, all parts eventually reach the same temperature at time ΔT , estimated as ~3.12 μ s.

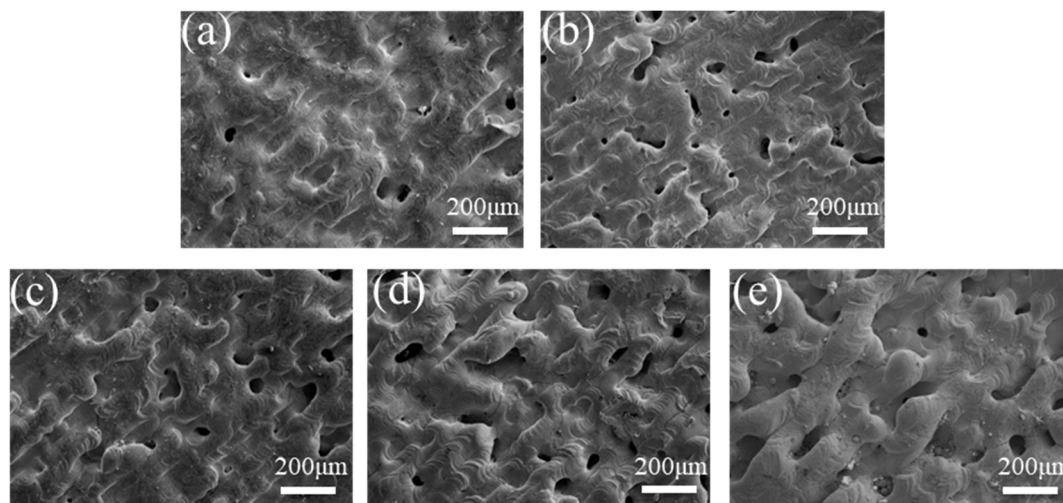


Figure 7. Laser-scanned surface morphologies of LPBF-fabricated pure tungsten bulk samples fabricated with laser power of 160 W and different laser-scanning rates: (a) 200, (b) 250, (c) 300, (d) 350 and (e) 400 mm/s.

Given the laser-scanning rate, one can compute the time required for complete laser irradiation of the scanned pure tungsten powders (Figure 8). This time t is a function of the laser spot diameter d_b (100 μ m), the diameter D of a spherical tungsten powder particle (~28 μ m) and the laser-scanning rate v (200–400 mm/s). The time t can also be calculated from the distance L moved by the laser beam during complete irradiation on the spherical tungsten particle (positions 1 to 2 in Figure 8). Hence, $L = d_b - D$ and t is calculated as

$$t = L/v = (d_b - D)/v \quad (4)$$

Accordingly, the time t required for complete laser irradiation of a single spherical pure tungsten powder particle is 180–360 μs , much longer than the 3.12 μs required for heat transfer from the irradiated surface to the interior of the metal-powder mass. Therefore, at the laser-scanning rate of this experiment, the heat generated by laser irradiation on the tungsten powder surface can reach the entirety of the powder. The laser-scanning rate influences the laser-scanned surface morphologies of the samples primarily by affecting the fluidity and wettability of the molten pool and the solidification rate of the molten pool.

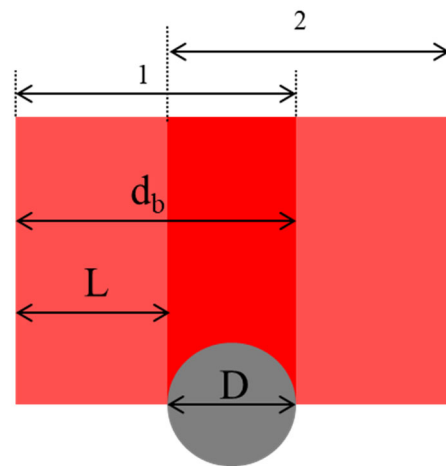


Figure 8. Schematic of laser irradiation on a pure tungsten powder particle of diameter D . d_b is the laser spot diameter and L is the distance over which the laser beam completely irradiates the particle (between positions 1 and 2).

As shown in Figure 7a, a low laser-scanning rate generates few pores and good overlap between the adjacent melt tracks on the LPBF-fabricated sample; consequently, the surface is smooth. The porosity fraction is approximately 1.915%. This outcome is attributable to the prolonged time of laser irradiation on the individual powder particles at a low laser-scanning rate. The extended laser irradiation raises the resulting molten-pool temperature, improving the fluidity and wettability and prolonging the solidification time of the molten pool. The molten pool is then granted sufficient time to flow and fill the gaps between the pre-laid powders and stabilise the fluctuations induced by Marangoni convection. As the laser-scanning rate increases, more pores appear on the laser-scanned surface of the LPBF-fabricated sample and the surface protrusions and spheroidisation phenomenon become more prominent (Figure 7b–e). The porosity fractions are approximately 3.720%, 7.130%, 9.217%, and 9.803% respectively. The laser-scanning rate reaches 400 mm/s; the LPBF-fabricated sample exhibits discontinuous melt tracks and numerous pores on its surface. It can be verified that increasing the laser-scanning rate shortens the time of laser irradiation on a single pure tungsten particle, lowering the molten-pool temperature, reducing the fluidity and wettability and shortening the solidification time. Consequently, the melt tracks are discontinuous, the adjacent melt tracks poorly overlap, and numerous pores and spheroidisation effects are generated.

Based on the aforementioned analysis, an LPBF-fabricated pure tungsten bulk sample with a smooth surface morphology can be obtained after appropriately increasing the laser power or reducing the laser-scanning rate. It should be noted that the hatch spacing (0.05 mm) was kept constant in this study, which was selected based on preliminary experiments to ensure sufficient overlap between adjacent tracks. The effect of hatch spacing on surface morphology and densification has been systematically investigated in previous

studies [21,22], where it was shown that optimising hatch spacing is as critical as laser power and scanning rate for achieving full densification.

3.2. Density Analysis

Figure 9 shows the effects of laser power and scanning rate on the relative densities of the LPBF-fabricated pure tungsten bulk samples. As the laser power gradually increases from 120 to 200 W at a constant scanning rate of 300 mm/s, the sample densities initially increase and then decrease (Figure 9a). At a laser power of 120 W, the sample density is ~91%. Increasing the laser power to 180 W raises the density to ~96%. Generally, the temperature of the molten pool is inversely proportional to its dynamic viscosity; thus, a higher temperature corresponds to a lower viscosity. A lower dynamic viscosity promotes the spreading and flowability of the molten pool, allowing it to more effectively fill the interparticle voids between the raw pure tungsten powders, thereby increasing the sample density [23–25]. When the laser power is further increased to 200 W, the sample density slightly decreases to ~95%. As discussed in Section 3.1 on the surface morphologies of LPBF-fabricated pure tungsten bulk samples, excessive laser power induces intense Marangoni convection, which generates protrusions on the sample surface. These protrusions reduce the density and uniformity of the subsequent pre-powder layer, resulting in numerous and large gaps that are difficult to completely fill through the flow and spreading of the molten pool, thus reducing the sample density.

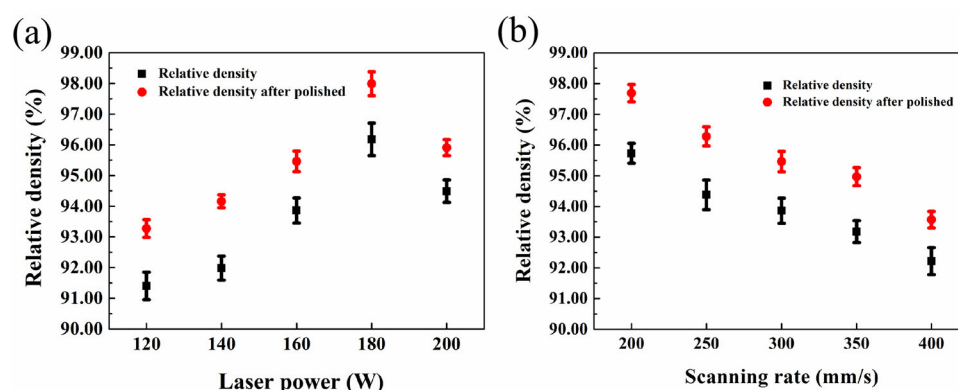


Figure 9. Relative densities of LPBF-fabricated pure tungsten bulk samples at (a) different laser powers at a laser-scanning rate of 300 mm/s and (b) different laser-scanning rates at a laser power of 160 W.

Figure 9b shows the changes in relative density of the pure tungsten bulk samples fabricated at a laser power of 160 W as the laser-scanning rate increases from 200 to 400 mm/s. At a scanning rate of 200 mm/s, the sample density reaches ~95%, but it gradually decreases with increasing scanning rate. This reduction is attributed to the generation of surface spheroidisation and pores at increased scanning rates, as shown in Figure 7.

Additionally, the four lateral surfaces of all the samples were ground and polished to remove the adhered powder layers, after which the corresponding densities were also measured. The densities of the polished samples are considerably higher than those of the unpolished samples. The adhered powder layers on the lateral surfaces correspond to loosely packed regions containing numerous open pores, which lower the density values measured by the Archimedes draining method.

Figure 10a–c show the surface morphologies of the polished LPBF-fabricated pure tungsten bulk samples at different laser-scanning rates, with the laser power maintained at 160 W. At a scanning rate of 200 mm/s, the surface is essentially free of pores, indicating high density. Indeed, the density of this sample reaches ~98% (Figure 9b). When the laser-

scanning rate is increased to 300 mm/s, numerous pores appear on the laser-scanned surface of the as-polished sample. Further increasing the scanning rate to 400 mm/s increases the number and size of pores on the polished surface. The pores on the laser-scanned surface of the as-polished sample are regularly arranged in columns with an intercolumn spacing of $\sim 50\ \mu\text{m}$, consistent with the overlap distance between two adjacent laser tracks. Therefore, these pores probably originate from the regions at the bottom of overlapping molten pools that are insufficiently fused. In such regions, either the residual unmelted tungsten powders or the molten tungsten powders cannot fully fill the gaps between adjacent laser tracks (Figure 10d). As the laser-scanning rate increases, the temperature of the molten pool decreases, the solidification time shortens and the melt flowability deteriorates. Consequently, the molten pool volume is reduced, the unfused regions at the molten-pool bottom are enlarged and the density of the LPBF-fabricated sample decreases. Based on the analysis of pore formation on the laser-scanned surface of the as-polished sample, it was inferred that at increased laser-scanning rates, a dense pure tungsten sample can be achieved by suppressing internal pore formation and reducing the overlap distance between adjacent laser tracks, or by appropriately increasing the laser power [26,27].

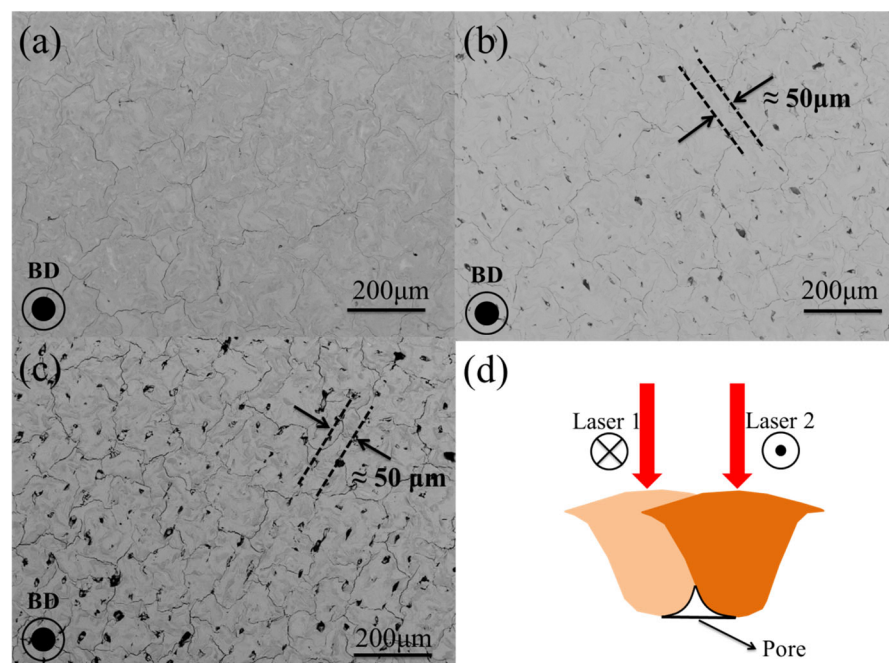


Figure 10. Laser-scanned surface morphologies of polished LPBF-fabricated pure tungsten samples at different laser-scanning rates: (a) 200, (b) 300 and (c) 400 mm/s. The laser power is 160 W. (d) Schematic of pore formation within the samples. (The arrows indicate the width of the two dashed lines, which is the spacing between each column of pores.)

It is important to note that in addition to the pores discussed above, microcracks are clearly visible in Figure 10a–c. These cracks are a significant limitation of the current study and are discussed in detail below. The formation of these cracks is attributed to the high residual thermal stresses generated during rapid solidification, combined with the intrinsic brittleness of tungsten at temperatures below its ductile-to-brittle transition temperature (DBTT). Similar crack networks have been extensively reported in LPBF-fabricated pure tungsten [28–30]. The cracks propagate primarily along high-angle grain boundaries, as confirmed by our EBSD analysis (Section 3.5), and their density and morphology are influenced by the laser scanning strategy [31]. The presence of these cracks, despite the high relative density achieved, represents a key challenge for the structural application of LPBF tungsten.

3.3. Phase Analysis

Figure 11 shows the XRD patterns of the raw tungsten powders and the LPBF-fabricated sample ($P = 160$ W, $v = 300$ mm/s). Both patterns exhibit four strong characteristic diffraction peaks corresponding to the (110), (200), (211) and (220) planes of body-centred cubic tungsten, indicating that the LPBF process does not introduce any significant phase transformation. No tungsten oxide peaks are detected in the XRD pattern of the LPBF-fabricated sample. However, this does not conclusively prove that the samples are entirely oxide-free, as the detection limit of conventional XRD is approximately 5% [13]. Indeed, other studies [14,24] have revealed the presence of oxides in LPBF-fabricated tungsten using XPS and fractography analyses, even when XRD did not detect them. Given the high purity of the starting powder (99.9%) and the controlled argon atmosphere ($O_2 < 100$ ppm) maintained during processing, any oxidation, if present, is likely below the XRD detection limit.

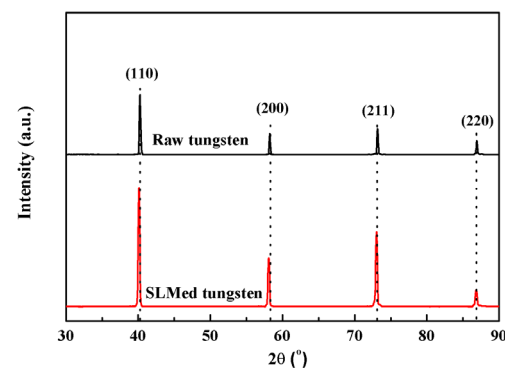


Figure 11. XRD patterns of the raw tungsten powders (**top**) and LPBF-fabricated pure tungsten samples (**bottom**).

The XRD patterns were fitted using Jade 6.5 software, and the positions and full-width-at-half-maximum (FWHM) values of the four characteristic diffraction peaks for each sample are listed in Table 1. The peak positions of the raw powders are close to those in the standard PDF#04-0806 pattern [32], whereas those of the LPBF-fabricated sample show small shifts of approximately $0.2\text{--}0.4^\circ$. According to Bragg's law ($2d\sin\theta = n\lambda$), variations in θ reflect changes in the interplanar spacing (d) caused by internal stress generated during rapid solidification. However, the magnitude of these shifts is relatively small and does not constitute a significant change.

Table 1. Fitting data of the XRD patterns.

Samples	2 θ (°)				FWHM (B)
	(110)	(200)	(211)	(220)	
PDF#04-0806	40.264	58.274	73.195	87.021	
Raw tungsten powders	40.242	58.227	73.159	86.911	0.196
LPBF-fabricated tungsten	40.116	58.101	73.012	86.827	0.237

More notably, the FWHM of the main peak increases from 0.196 in the raw powders to 0.237 in the LPBF-fabricated sample, and similar broadening is observed for the other peaks. Since the Scherrer equation applies only to grains smaller than 100 nm, this broadening cannot be solely attributed to grain refinement. Instead, the increased FWHM values primarily arise from microstrains induced by internal stresses formed during the rapid solidification of the molten pool [18]. The FWHM broadening is therefore a more reliable indicator of residual stresses and microstrains than the small peak shifts. Given the intrinsic brittleness and narrow solidification-temperature range of pure tungsten,

these microstrains tend to manifest as macroscopic cracks [19,20], consistent with the crack networks observed in Figure 10a–c.

3.4. EBSD Analysis

To further quantitatively analyse the crystallographic orientation, grain size distribution, and micro-texture evolution, EBSD analysis was performed on the side surface of the LPBF-fabricated pure tungsten sample (laser power 160 W, scanning rate 300 mm/s). Figure 12 presents the EBSD results.

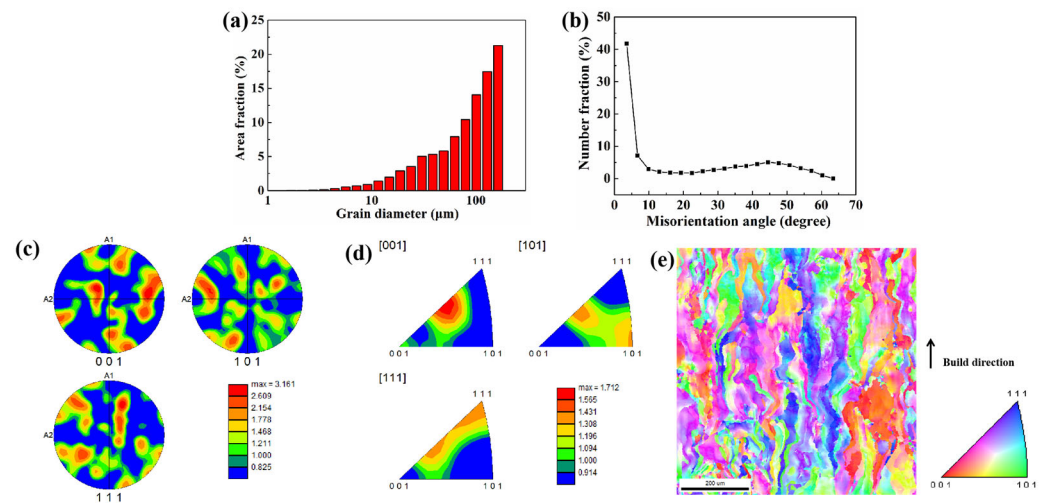


Figure 12. EBSD analysis of the LPBF-fabricated pure tungsten bulk sample: (a) grain size distribution, (b) misorientation angle distribution, (c) pole figures (PF), (d) inverse pole figures (IPF) and (e) orientation map.

The grain size distribution (Figure 12a), inverse pole figure (IPF) (Figure 12d) and orientation map (Figure 12e) collectively reveal that the side surface consists of coarse columnar grains extending parallel to the building direction (BD), with a high proportion (>20% area fraction) of grains exceeding 100 μm . The orientation map clearly visualises the continuous epitaxial growth of columnar grains across multiple deposited layers. The formation of such elongated columnar grains is attributed to the epitaxial growth mechanism, where solidification continuously inherits crystal orientations from the pre-existing grains at melt pool fusion boundaries, as widely reported in previous LPBF pure tungsten research [12,33]. Unlike the literature that obtained ultra-fine grains (<10 μm) under ultra-high scanning speeds [21], the relatively low scanning rate (300 mm/s) and high linear energy density adopted in this study prolong the melt pool solidification time, providing sufficient driving force for grain coarsening and generating the large columnar grains observed in the orientation map.

Figure 12b shows the misorientation angle distribution. The average misorientation angle is 20.62°. A prominent peak at ~3.5° is observed, with a number fraction of ~41.8% for low-angle grain boundaries (LAGBs, <15°). This high density of LAGBs is a direct indication of substantial residual stress and microstrain, originating from the rapid heating and cooling cycles inherent to the LPBF process. This finding aligns with previous studies on LPBF tungsten [27,28], which also reported a significant fraction of LAGBs in as-built samples that can be relieved through post-annealing. The formation of LAGBs is closely related to the energy input level: higher linear energy density tends to generate more pronounced thermal gradients and faster cooling rates, thereby promoting the development of dislocation substructures and LAGBs within the grains [28].

The pole figures (PFs) and the corresponding maximum texture intensity (Figure 12c, $\max = 3.161$) indicate a relatively weak preferred crystallographic orientation on the side surface. The weak texture intensity can be primarily attributed to the employed 67° rotation scanning strategy (Figure 2), which disrupts the strong preferred orientation by generating overlapping multi-directional molten pools [34]. This rotation strategy has been shown to randomise grain orientation and reduce texture intensity in other refractory metals fabricated by LPBF [31]. Furthermore, the texture evolution in BCC tungsten is less straightforward than in FCC metals. For tungsten, a high thermal conductivity and a deep bowl-shaped melt pool can lead to a growth direction that deviates from the typical $\langle 001 \rangle$ texture, potentially resulting in weak $\langle 110 \rangle$ or $\langle 111 \rangle$ textures [12]. This complex interplay between melt pool geometry and crystallographic growth, combined with the rotation scanning strategy, explains the absence of a strong $\langle 001 \rangle$ fibre texture in this study. The combination of coarse columnar grains, a high density of LAGBs, and weak texture collectively contributes to the anisotropic mechanical properties observed between the top and side surfaces of the LPBF-fabricated tungsten.

It is important to note that the EBSD analysis presented here is limited to the side surface of a single sample ($P = 160$ W, $v = 300$ mm/s). A comprehensive EBSD study of all samples and both surfaces would provide a more complete picture of the microstructural evolution under different processing conditions. However, the current analysis serves as a representative case study to illustrate the key crystallographic features of LPBF-fabricated tungsten, while the XRD analysis provided phase information for all samples.

3.5. Microhardness Analysis and Microstructure Evolution

Figure 13 shows the microhardness values of the laser-scanned surfaces for all 10 samples. As presented in Figure 13a, with a fixed scanning rate of 300 mm/s, the microhardness increases from approximately 430 HV at 120 W to a peak of approximately 480 HV at 180 W, and then decreases to approximately 440 HV at 200 W. Figure 13b shows that with a fixed laser power of 160 W, the microhardness increases from approximately 390 HV at 200 mm/s to a peak of approximately 470 HV at 300 mm/s, and then decreases to approximately 450 HV at 400 mm/s. The microhardness values obtained in this study (390–480 HV) are comparable to or slightly higher than those reported in the literature for LPBF-fabricated pure tungsten, e.g., 428 HV₃ by Wen et al. [22] and 474 HV by Guo et al. [21]. In general, the microhardness of a metal sample is influenced by the synergetic effects of density, grain size, and internal stress. However, the relationship between hardness and density is not strictly linear: the highest density ($\sim 98\%$) at 200 mm/s corresponds to only ~ 390 HV due to the coarse grain structure (Figure 14a,d), whereas the peak hardness (~ 470 – 480 HV) at 300 mm/s and 180 W results from significant grain refinement (Figure 14b,e) despite moderate porosity. Large internal stress can also distort the lattice and increase hardness, as evidenced by XRD peak broadening (Table 1) and the high fraction of low-angle grain boundaries observed in EBSD analysis (Figure 12b). Pores within the metal sample, on the other hand, provide space for atomic movement and reduce hardness. In addition, the grain-boundary density increases with decreasing grain size, effectively impeding dislocation motion [35] and enhancing hardness [22].

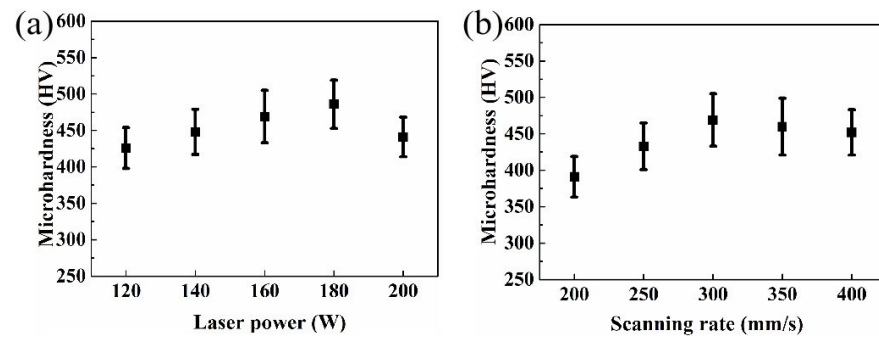


Figure 13. Microhardness of laser-scanned surfaces of pure tungsten bulk samples at (a) different laser powers at a laser-scanning rate of 300 mm/s and (b) different laser-scanning rates at a laser power of 160 W.

Figure 14 shows the microstructural characteristics of the laser-scanned surfaces of pure tungsten bulk samples fabricated at different laser-scanning rates under a laser power of 160 W. It should be noted that the microstructures shown in Figure 14 are actually the solidification morphology of the melt pools, which does not reveal individual grain boundaries. The observed features are the result of the scanning strategy (67° rotation) and melt pool geometry. As established in the literature [12], columnar grains in LPBF-fabricated samples are primarily oriented along the build direction (visible on side surfaces) due to the thermal gradients involved during solidification. The laser-scanned surface (top surface) typically shows melt pool boundaries and solidification substructures rather than clearly defined columnar grains. The samples fabricated at 200 mm/s comprise fine columnar grains and some relatively coarse grains (Figure 14a,d). The formation of columnar grains can be attributed to heat conduction between the surrounding powders and the solidified material within the molten pool. Nucleation preferentially occurs at the edge of the molten pool and solidification proceeds towards the centre, leading to the formation of columnar grains. At a lower laser-scanning rate, the solidification rate of the molten pool decreases, resulting in more uniform growth rates of crystal nuclei in different directions. This provides sufficient time for the nuclei to grow in multiple directions, forming relatively coarse grains. A coarse-grained microstructure has a low density of grain boundaries due to the increased grain size, which reduces microhardness. Consequently, the resistance to dislocation slip within the crystal decreases, allowing atoms to move more easily under an external force. This promotes plastic deformation, manifesting as reduced microhardness.

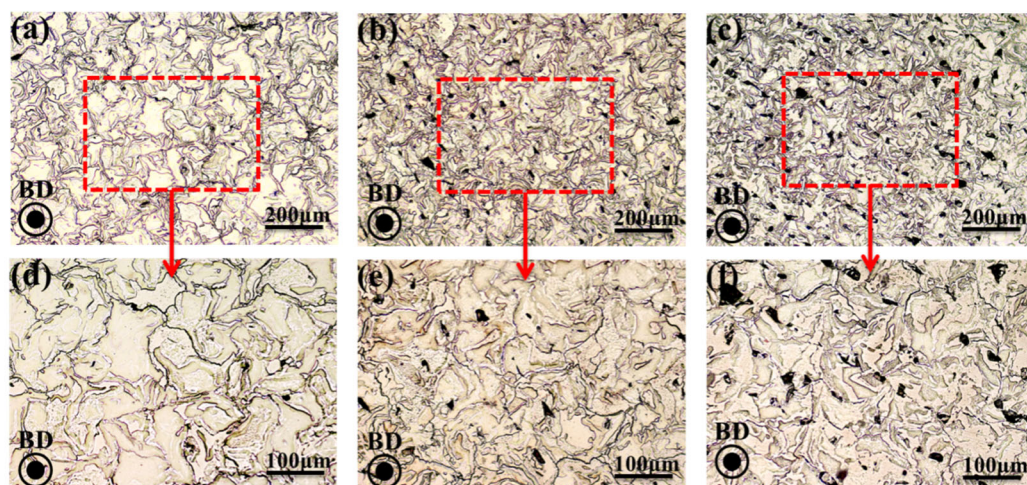


Figure 14. Microstructural characteristics of laser-scanned surfaces of pure tungsten bulk samples fabricated at different laser-scanning rates: (a,d) 200, (b,e) 300 and (c,f) 400 mm/s. The laser power is 160 W. (The arrows indicate the enlarged view of the dashed boxes is shown below.)

When the laser-scanning rate is increased to 300 mm/s, the microhardness of the laser-scanned surface of the LPBF-fabricated sample increases. The grain size on the laser-scanned surface decreases considerably compared with that of the sample fabricated at 200 mm/s (cf. Figure 14b,e with Figure 14a,d), and a few pores are visible. This improvement is attributable to the shortened solidification time of the molten pool at an increased laser-scanning rate, which promotes nucleation at the edge of the molten pool followed by rapid growth towards its centre and limited growth along the laser-scanning direction, resulting in elongated columnar grains. The reduced grain size increases the number of grain boundaries, thereby enhancing microhardness [34,36]. Meanwhile, as discussed in the previous subsection, increasing the laser-scanning rate promotes pore formation on the surface. In general, the presence of pores decreases the density and microhardness of metal samples. Considering the combined effects of grain refinement and pore formation on the microhardness of the laser-scanned surface, the increased laser-scanning rate likely achieves notable grain refinement despite a slight rise in pore number, ultimately enhancing the microhardness of the LPBF-fabricated sample.

When the laser-scanning rate is further increased to 400 mm/s, the microhardness of the laser-scanned surface does not increase appreciably but instead shows a slight reduction. The corresponding microstructural characteristics are presented in Figure 14c,f. At this laser-scanning rate, pronounced grain refinement occurs on the laser-scanned surface, and the number of columnar grains increases. However, the increased laser-scanning rate reduces the temperature of molten pure tungsten, thereby decreasing its fluidity and spreading ability and increasing pore density. Although the small grain size increases the number of grain boundaries, the numerous pores provide space for plastic deformation of the laser-scanned surface of the LPBF-fabricated sample under an external force, ultimately reducing rather than markedly increasing the microhardness.

Figure 15a shows the microhardness values of the laser-scanned and side surfaces of the sample fabricated under a laser power of 160 W and a laser-scanning rate of 300 mm/s. Overall, the grains within the LPBF-fabricated sample exhibit uneven lamellar morphologies. The microhardness reaches ~480 HV on the laser-scanned surface and ~390 HV on the side surface. Figure 15b,c show the corresponding microstructural characteristics of the laser-scanned surface and side surface, respectively. The clearly smaller grain size on the laser-scanned surface than on the side surface explains the higher microhardness of the laser-scanned surface. On the laser-scanned surface (Figure 15b), the columnar grains grow at a certain angle relative to the laser-scanning direction and extend along the laser-scanning direction. On the side surface (Figure 15c), most grains are columnar or S-shaped and commonly grow parallel to the layer-wise building direction of the sample. The lengths of most grains on the side surface (>100 μm) largely exceed the thickness of a single powder layer during the LPBF fabrication. This length extension is possibly caused by remelting of the solidified part during the laser-scanning process.

The above analysis demonstrates a preferred orientation of grain growth during the LPBF process of pure tungsten powders. Moreover, the sizes and morphologies of the grains can be regulated to some extent by appropriately adjusting the laser processing parameters or the laser-scanning strategy.

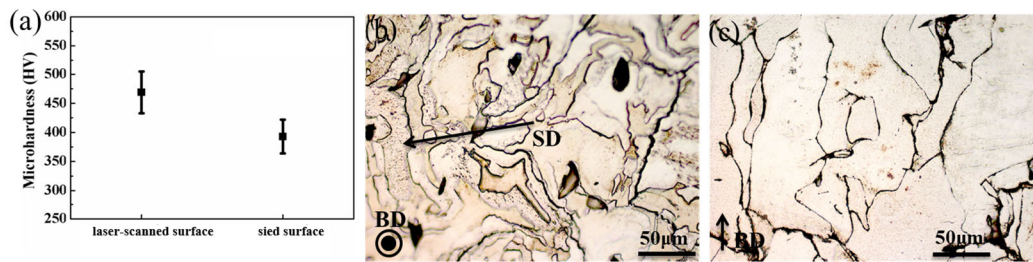


Figure 15. Microhardness values and microstructural characteristics of the LPBF-fabricated pure tungsten bulk sample under a laser power of 160 W and a laser-scanning rate of 300 mm/s: (a) microhardness values, (b) microstructural characteristics of laser-scanned surface and (c) microstructural characteristics of side surface.

4. Conclusions

This study systematically investigated the effects of laser power and laser-scanning rate on the surface morphologies, densities, microstructural characteristics and microhardness of LPBF-fabricated bulk samples. It also analysed the formation and suppression mechanisms of surface spheroidisation, protrusions and internal pores, and elucidated the phase constitution of the LPBF-fabricated samples. The main conclusions are summarised below.

- (1) Increasing laser power or reducing scanning rate promotes continuous melt tracks and suppresses spheroidisation. However, excessive laser power (200 W) intensifies Marangoni convection, creating surface protrusions that reduce density.
- (2) Pores form due to insufficient overlap at the bottom of adjacent molten tracks. Reducing the laser-scanning rate effectively suppresses pore formation, achieving a maximum relative density of ~98%.
- (3) The laser-scanned surface exhibits a transition from coarser to finer solidification substructures as the scanning rate increases, while the side surface consists of coarse columnar grains extending along the building direction. The EBSD analysis also reveals a high fraction (~41.8%) of low-angle grain boundaries at ~3.5°, indicating significant residual stress and microstrain.
- (4) A low scanning rate (200 mm/s) results in coarse substructures and lower hardness (~390 HV), while a moderate scanning rate (300 mm/s) refines the substructure and improves hardness to ~470–480 HV. An excessively high scanning rate (400 mm/s) generates numerous pores that offset the strengthening effect of substructure refinement, leading to a slight hardness reduction to ~450 HV. The side surface, with its coarser columnar grains, exhibits a lower hardness (~390 HV) compared to the laser-scanned surface (~480 HV) at the same processing conditions.
- (5) Phase analysis reveals no significant phase transformation after LPBF. However, internal stress and microstrain increase notably, as evidenced by the increased FWHM values in the XRD patterns and the high density of low-angle grain boundaries observed by EBSD.
- (6) EBSD pole figures show a weak preferred orientation (max texture intensity 3.161) on the side surface, primarily due to the 67° rotation scanning strategy. This weak texture, combined with coarse grains and LAGBs, causes anisotropic mechanical properties between different surfaces.

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visualisation, Z.J.; supervision, D.Y.; project administration, F.W.; funding acquisition, F.W. All authors have read and agreed to the published version of the manuscript.

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