

Article

Microstructural, Mechanical, and Tribological Properties of Selective Laser Melted Inconel 718 Alloy: The Influences of Heat Treatment

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Abstract: The present study investigates the microstructural, mechanical, and tribological properties of Inconel 718 alloy produced by selective laser melting (SLM) in relation to heat treatment. The SLM-processed samples received a two-step heat treatment: solutionizing at 1065 °C for 1 h, followed by double aging at 720 °C for 8 h and 620 °C for 6 h. The as-built sample exhibited a grain structure mostly characterized by fine Laves phases, while the hardening phases γ' ($\text{Ni}_3(\text{Al}, \text{Ti})$) and γ'' (Ni_3Nb) precipitated during the heat treatment. Following heat treatment, a transformation in crystallographic texture and dislocation density occurred, yielding a random texture and reduced dislocation density, particularly in the XZ direction, attributed to the formation of new grains via recrystallization in the microstructure. The grain size in the XY plane decreased following heat treatment, whereas the texture in the $\langle 001 \rangle$ direction remained unaffected. The heat-treated samples had significantly higher tensile strength (1330 MPa vs. 960 MPa) and hardness (530 HV vs. 340 HV) relative to the as-built samples. The wear resistance of heat-treated samples surpassed that of the as-built sample due to enhanced mechanical properties resulting from the fine and dispersed γ' and γ'' precipitates in the microstructure with heat treatment.

Keywords: 3D printing; selective laser melting; super alloy; heat treatment; precipitation strengthening phases; deformation mechanism



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1. Introduction

Additive manufacturing (AM), commonly referred to as “3D printing”, comprises a collection of advanced manufacturing methodologies capable of fabricating three-dimensional components using computer-aided design (CAD) software-generated virtual models [1,2]. This peculiar feature allows the production of complex shapes that are almost unfeasible to fabricate using conventional methods, such as casting and forging [3]. AM allows designers to create customized or complicated components without excessive material waste, complex shape problems, or specialized tools [4,5]. Among AM methods, selective laser melting

(SLM) is highly favored due to its ability to manufacture close to net shape and unique surface quality parts with complex shapes and outstanding mechanical properties [6,7].

Inconel 718 superalloy has been acknowledged as one of the most valuable materials in the aerospace [8], turbine blades [9], automotive [10], and petrochemical [11] industries for various applications due to high corrosion and oxidation resistance, and strength in both room and elevated temperatures [12,13]. Thus, Inconel 718 is a candidate for industrial applications of SLM-manufactured parts [14]. Even though AM has many benefits over conventional methods, it still has to deal with some challenges, such as the formation of residual stress, macro- and micro-segregation, poor ductility, and rough surfaces on the parts it builds [15–18]. So far, attempts have been made to solve these problems by optimizing process parameters like laser power, scan speed, and layer thickness [19,20]. However, optimizing the parameters is insufficient to solve these drawbacks [21]. Particularly, Inconel 718 alloys produced using the SLM method experience residual stresses as a result of the high melting and solidification processes [22,23]. The high concentration of refractory metals such as Nb and Mo in Inconel 718 alloy results in microstructural segregation, which negatively impacts the alloy's mechanical properties [24,25]. Thus, post-heat treatment is widely used to achieve specified mechanical and microstructure properties by relieving residual stresses and promoting the formation of strengthening precipitates [26,27].

Recently, Inconel alloys produced by AM methods have become very popular. However, as a result of high melting and solidification rates, defects such as porosity [28], cracks [29], and residual stresses are caused in the internal structure [30]. Post-heat treatments are widely used to eliminate these defects and increase the mechanical properties of Inconel alloys [31–34]. Heat treatments have been employed on SLM-produced Inconel alloys to tailor the microstructural features and mechanical properties. The strengthening mechanism of Inconel 718 alloy involves heat treatment processes that include homogenization, solid solution, and aging treatment [28,29]. The homogenization and solid solution treatments are performed at temperatures between 980 and 1200 °C to dissolve undesirable phases. The precipitation hardening treatment involves two consecutive steps to precipitate the strengthening phases (e.g., such as γ' and γ'') [30,31]. Liu et al. [35] utilized solution heat treatment, including direct double aging at 950 °C and 1050 °C at different hold times, and examined the dissolution of precipitates in SLM-produced Inconel 718. Huang et al. [36] have investigated different heat treatments, such as homogenization, aging, and precipitation aging, on the mechanical properties of SLM-produced Inconel 718. They reported that as the heat treatment temperature rises, the microstructure at each fabrication position shifts from anisotropic to isotropic, and the degree of recrystallization notably increases [36]. Li et al. [32] studied the effects of double-aging treatment on both the microstructural and mechanical properties of SLM-produced Inconel 718. They observed that the δ -phase precipitated on the grain boundaries, inhibiting grain growth [32]. Moreover, Maurya et al. [37] investigated the effects of homogenization, solution, and aging heat treatments on the microstructural and mechanical properties of SLM-produced Inconel 718. They found that γ' and γ'' strengthening phases formed on grain boundaries, and heat treatment improved Inconel 718's mechanical properties compared to the as-built sample [37]. Zhao et al. [24] showed that after solution treatment at 1020 °C, Laves phase and carbides were dissolved into the matrix in heat-treated SLM-produced Inconel 718 samples at different solutions, and γ' and γ strengthening phases were precipitated in the matrix after aging [24]. To sum up, the variation of microstructural characteristics and mechanical properties of Inconel 718 alloy has recently been studied in the literature. Nonetheless, further research is required to clarify the correlation between the microstructural characteristics, mechanical properties, and tribological properties of Inconel 718 alloy processed by SLM.

This study, for the first time, comprehensively examines the microstructural, mechanical, and tribological properties of Inconel 718 samples fabricated via SLM, specifically focusing on the impact of heat treatment on these properties through extensive microstructural characterization and comprehensive mechanical and tribological testing. Thus, the study aims to contribute to the restricted literature on the efficacy of heat treatment on improving the performance of SLM-produced Inconel 718 alloys that are essential for their use in industrial applications such as aerospace [8], turbine blades [9], and automotive [10]. Novel findings regarding the microstructural characteristics, mechanical properties, and tribological properties of both as-built and heat-treated SLM-manufactured Inconel 718 alloy are presented, alongside an extensive discussion.

2. Materials and Methods

2.1. Processing of Inconel 718 Samples

Inconel 718 alloy samples (In718 samples) were produced using In718 alloy powder with a particle size ranging from 15 μm to 45 μm , exhibiting a spherical morphology, by the SLM Envision 120 machine (Ermaksan, Turkey). The chemical composition and morphology of the powders are given in Table 1 and Figure 1, respectively. The samples are rectangular with dimensions of 10 mm $\times$ 10 mm $\times$ 30 mm. The samples were produced with a Z-axis orientation using 350 W laser power and a scanning speed of 745 mm/s. The final linear energy density (LED) was 0.47 J/mm, and the layer thickness of the samples was set at 30 μm . A scanning strategy was used where laser scanning patterns were rotated by 67° to achieve uniform heat distribution and prevent porosity and residual stress. High-purity argon gas was utilized to prevent the oxidation of the build parts during the process.

Table 1. Chemical composition of In718 powder (wt. %).

Element	Ni	Cr	Mo	Nb	Ti	Al
Wt. %	Bal.	17.64	3.18	21.2	0.88	0.44

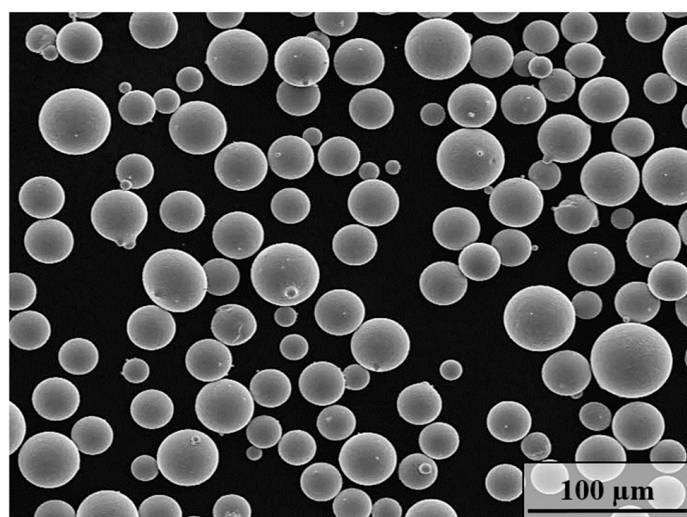


Figure 1. SEM image of In718 powders.

The samples produced via the SLM underwent heat treatment: solid solution at 1065 °C for 1 h with air cooling (AC), followed by double aging at 720 °C with furnace cooling (FC) at 620 °C for 8 h and 6 h, respectively. Air cooling was conducted for 8 h following that. Figure 2 illustrates the solid solution + aging heat treatment cycle of the In718 samples.

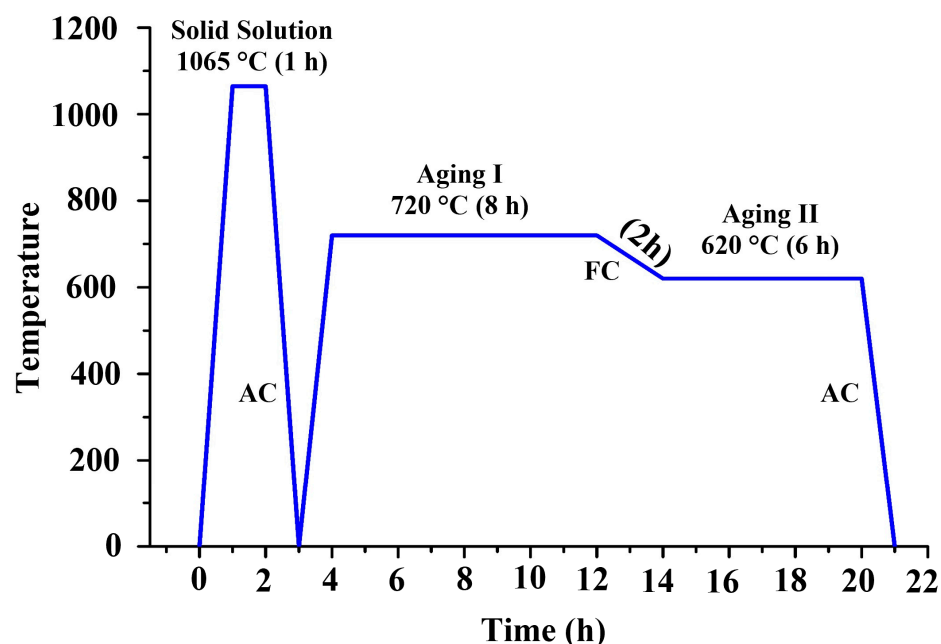


Figure 2. Heat-treatment cycle of SLM processed In718 samples.

2.2. Characterization of Microstructural Properties

The samples were embedded in resin and metallographically prepared for microstructural and phase analyses. The samples were ground using SiC papers (220–2500 mesh) and then polished with 3 μm diamond paste. Finally, the samples were etched with Kalling's reagent, which consists of 5 g copper chloride, 100 mL water, 100 mL hydrochloric acid, and 100 mL alcohol. The microstructure analyses of samples were performed with optical microscopy (Olympus BX41M-LED, Tokyo, Japan) and SEM (JEOL/JSM-6060 with energy-dispersive spectroscopy (EDS), Tokyo, Japan). An X-ray diffractometer (XRD-RigakuSA-HF3, XRD-RigakuSA) was utilized for phase characterization. The scan speed of the measurements was 1 deg/min, a step size of 0.001, and a scan range from 20° to 100° (2 θ).

Furthermore, EBSD analysis on the XY and XZ planes was performed to reveal the variation of microstructural features as a function of heat-treatment. Figure 3 displays the XY and XZ planes where EBSD analysis took place. The as-built and heat-treated specimens experienced metallographic processing procedures to prepare them for EBSD analysis. After the grinding and rough polishing stages, the polishing procedure concluded with 50 nm silica suspension. The EBSD images were obtained using a Zeiss-brand Gemini Sigma 300 scanning electron microscope equipped with an Oxford Instruments Aztec EBSD system. The EBSD study utilized a working distance of 15 mm, a mapping area of 100 $\times$ 100 μm , an accelerating voltage of 10 kV, and a step size of 0.18 μm . The obtained EBSD data was processed using the software TSL-OIM Analysis v7.3.1. Inverse pole figure (IPF), grain orientation spread (GOS), and recrystallization ratio (RC) maps were used to evaluate the effect of heat treatment on the microstructure of the SLM-produced In718 samples. Furthermore, Bunge's generalized spherical harmonic series expansion approach was utilized to perform texture evaluations [38].

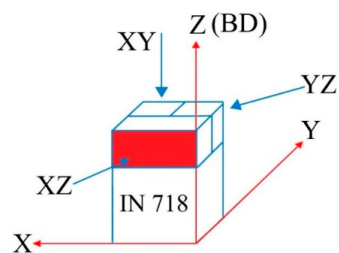


Figure 3. Schematic view of In718 that displays the XY and XZ planes for EBSD characterization (BD: build direction).

2.3. Characterization of Mechanical Properties

The hardness of as-built and HT samples was analyzed using a Zwick/Roell hardness tester with a load of 500 g and a dwell time of 10 s. For enhanced reliability and precision of the microhardness data, 10 indentations were performed on each sample.

Tensile tests were conducted at room temperature using an Instron 5985 Universal machine (Ulm, Germany) following the TS-EN-ISO-6892-1 standard method A. The crosshead speed was 5 mm/min. Three samples were repeated for each parameter. Figure 4 shows the dog-bone-shaped tensile test sample according to ASTM E8.

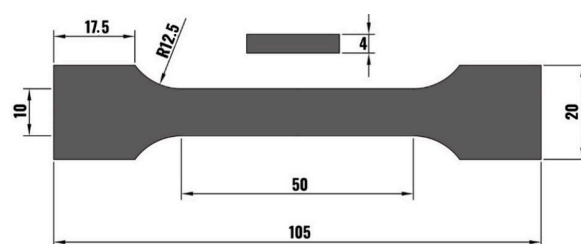


Figure 4. Schematic of dog-bone-shaped tensile test sample (dimensions in mm).

2.4. Characterization of Tribological Properties

The tribological characteristics of as-built and HT samples were assessed using a ball-on-disc test at room temperature with a Turkeyus POD/HT/WT tribometer (Bursa, Turkey) following the ASTM G133-05 standard. Silicon nitride balls with a diameter of 6 mm were used as the counter surface during wear tests. The wear tests were conducted with a normal load of 20 N over a sliding distance of 250 m.

3. Results

3.1. Microstructural Properties of As-Built and Heat-Treated Samples

The optical microscopy (OM) images of as-built and heat-treated (HT) samples are displayed in Figures 5 and 6. Figure 5 shows that the as-built sample has a cellular columnar structure on all planes, while the HT sample has equivalent grains following solid solution and age treatments.

As a result of rapid solidification, cracks and defects are typically encountered during the SLM process; however, neither the as-built nor the HT samples exhibited any cracks, indicating that the AM processing parameters are appropriate. Molten pools are distinguishable within the internal structure of the as-built sample, as illustrated in Figure 6a,b. Additionally, a greater number of equiaxed grains (as shown by white arrows in Figure 6d) are observed as a result of the solid solution temperature of 1065 °C surpassing that of the recrystallization temperature (approximately 1025 °C) [36]. New grains eventually nucleated and grew within the grain, replacing the coarse columnar crystals in the post-heat-treated sample shown in Figure 6c,d.

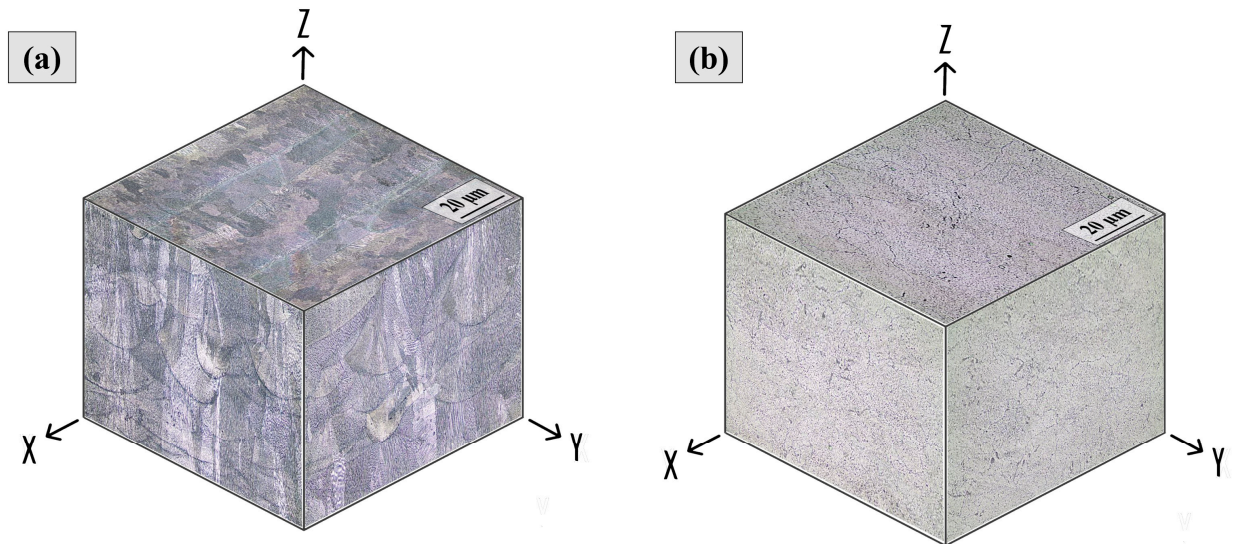


Figure 5. 3D OM images of (a) as-built and (b) HT samples.

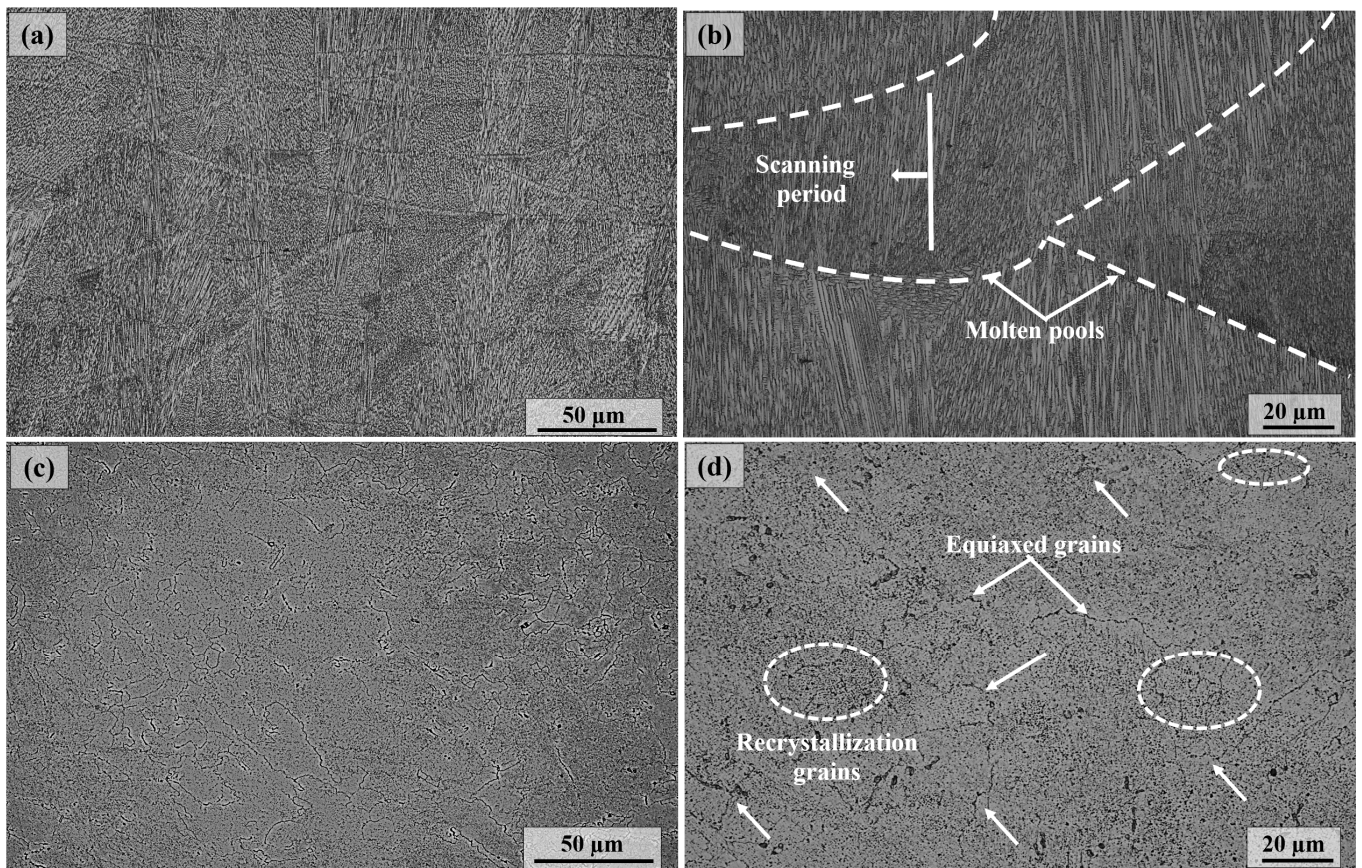


Figure 6. OM images of as-built (a,b) and HT (c,d) samples.

SEM images of as-built and HT samples are given in Figure 7 in order to further examine the influences of heat treatment on the microstructural properties. The as-built In718 sample exhibits a microstructure composed of fine cellular dendrites, and the molten pool boundary is clearly visible (Figure 7a,b). Evidently, molten pools are not perfectly symmetrical and are not identical to one another.

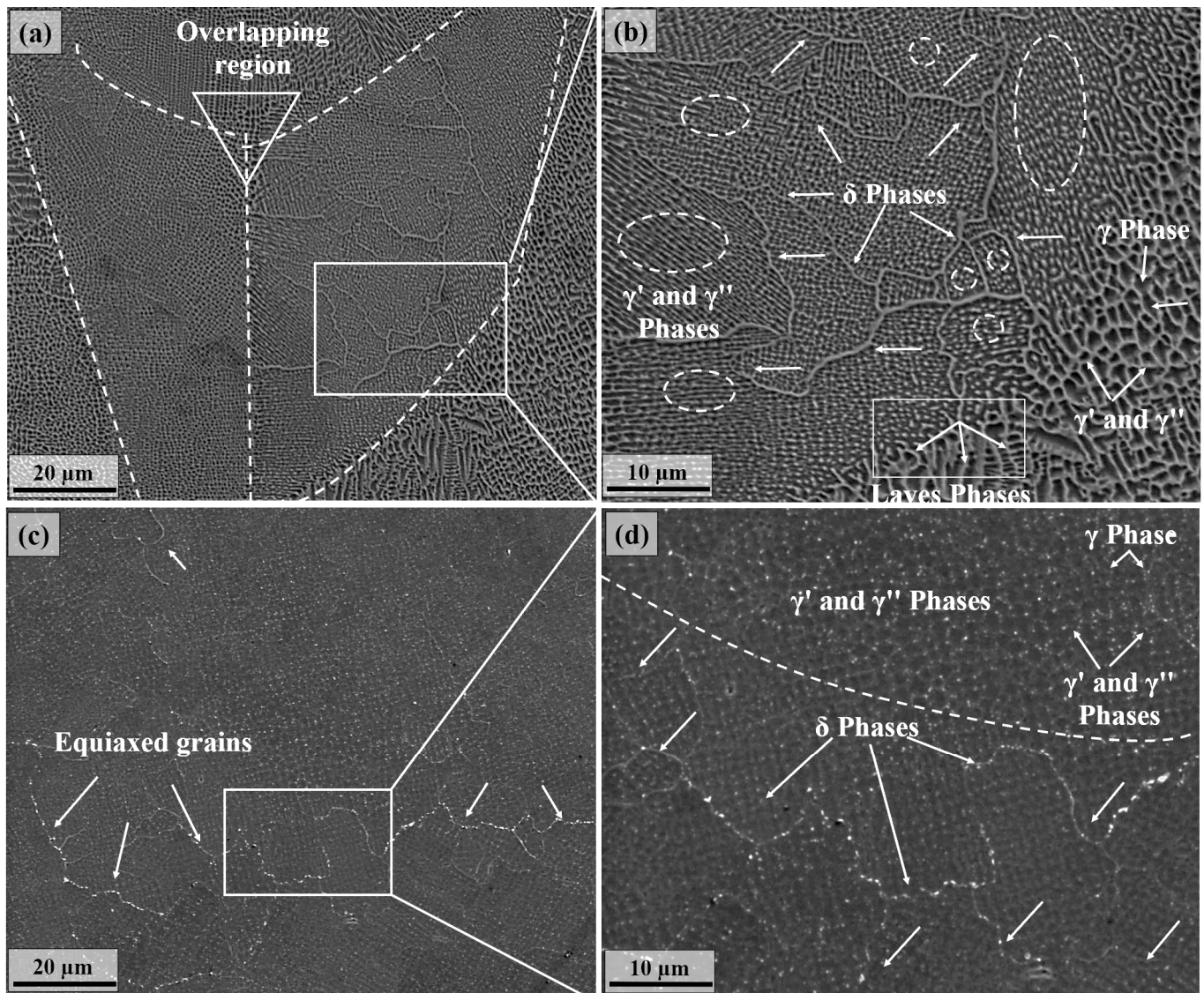


Figure 7. SEM micrographs of as-built (a,b) and HT (c,d) samples.

In Figure 8, the XRD patterns of In718 powder, as-built, and HT samples are illustrated to further clarify the evolution of phases within the microstructure through heat treatment. After heat treatment, strengthening γ' and γ'' phases, which typically have a face-centered cubic (FCC) crystal structure, are visible with minimal deviation and identical lattice structure [39]. XRD patterns of γ , γ' , and γ'' are visible, as evidenced by the diffraction peaks in the (111), (200), and (311-033) planes. NbC carbide was also observed on the (111) plane with angles ranging from 25° to 30°. This can be attributed to the presence of unsolidified Nb elements during the SLM process. Furthermore, the (200) peak exhibits greater intensity in the as-built and HT samples as a result of the precipitated strengthening phases (γ' and γ''). The γ and γ' phases at the (220) plane phases (70–80°) disappeared subsequent to HT due to dissolution.

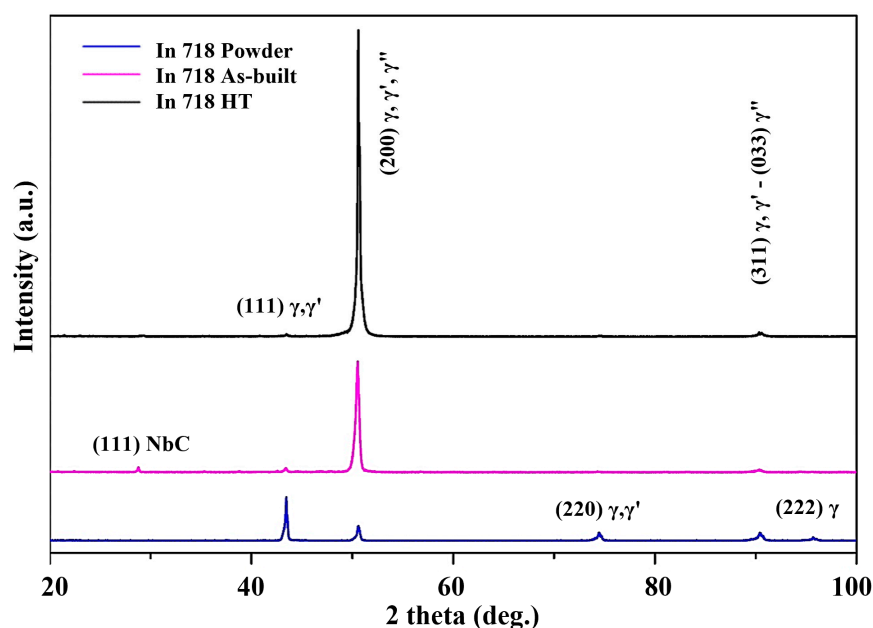


Figure 8. XRD spectrum of In718 powder, as-built, and HT samples.

After heat treatment, the Laves phases become encapsulated within the dendrite region of the inner cells, while the delta (δ) phases precipitate at the grain boundary (Figure 7b,c). The HT sample has equiaxed grains. Its cellular dendrite structure (as indicated by white arrows in Figure 7c,d) has disappeared completely because of solid solution application and aging. Furthermore, it is evident from Figure 7d that the Laves phases dissolve rapidly, and a considerable quantity of δ phases precipitate at the grain boundary. The Laves phase undergoes partial dissolution since the heat treatment temperature of 1065 °C exceeds the solvus temperature of the δ phases. As a result, the precipitation of the δ phases is not desired following HT [40].

In order to further understand the evolution of texture and microstructural features following heat treatment, Figure 9 shows the IPF map of the as-built and HT In718 samples in both the XZ and XY planes. The microstructure exhibits a columnar structure resulting from epitaxial growth (Figure 9a), while the columnar structure is oriented in the $\langle 101 \rangle$ direction within the XZ plane, revealing that the as-built sample possesses a strong texture in the XZ plane. Following heat treatment, the columnar structure remained unchanged in the XZ plane (Figure 9c), but the grain size decreased, as also evidenced by the discussed OM and SEM images (Figures 5–7). Recrystallization was caused by the heat treatment, leading to the formation of new grains in the as-built microstructure. Thus, the strong texture observed in the as-built condition transformed into a random texture following the heat treatment in the XZ plane. The reorganization of the grains led to a modification of the texture and facilitated a transition from an anisotropic to an isotropic structure. Upon analyzing the XY plane, it is observed that grains are oriented in the $\langle 001 \rangle$ direction in the as-built condition (Figure 9b). The grain size in the XY plane decreased following heat treatment (Figure 9d), as was observed in the XZ plane, whereas the strong texture in the $\langle 001 \rangle$ direction was not changed with heat treatment.

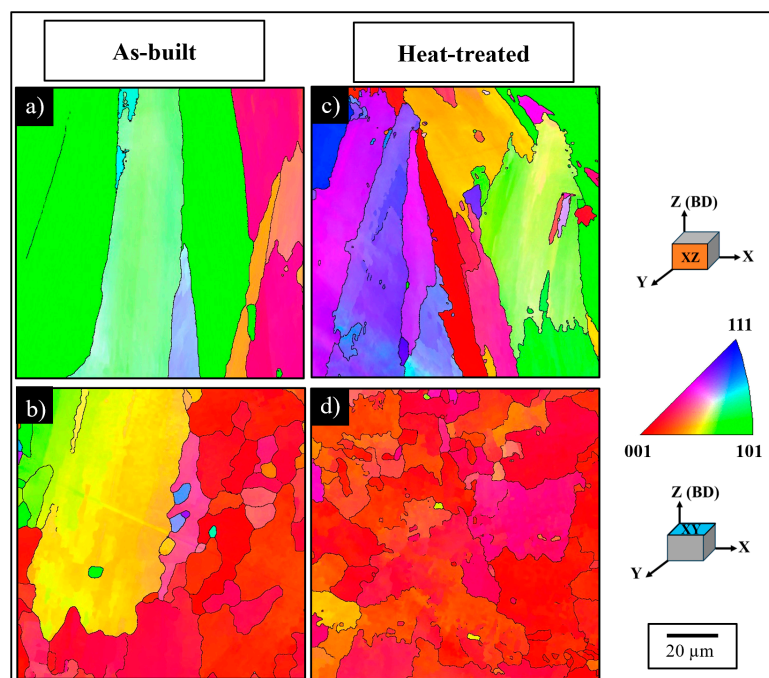


Figure 9. EBSD analysis showing IPF image maps for the microstructures of In718: As-built in (a) XZ and (b) XY orientations; HT in (c) XZ and (d) XY orientations.

3.2. Mechanical Properties of As-Built and HT SAMPLES

The stress–strain curves of the samples are given in Figure 10. The as-built and HT samples' yield strength (YS) and ultimate tensile strength (UTS) values are 650–960 MPa [41,42] and 1270–1330 MPa [24,43], respectively, agreeing well with the previous studies [24,41–43]. The elongation value of the as-built sample is roughly 58% higher than that of the HT sample. On the other hand, the as-built and HT samples' hardness values are 340 HV and 540 HV, respectively. The results clearly demonstrate that heat treatments significantly enhance the YS and UTS in alignment with the hardness values, while substantially reducing the elongation at failure. The low tensile strength in the as-built sample is due to the absence of strengthening phases γ' and γ'' that typically develop during the solid solution and aging treatment. In parallel with these results, the as-built materials exhibit lesser hardness compared to HT samples due to the lack of strengthening phases (γ' and γ''), which aligns with existing literature [35]. Furthermore, heat treatment helps separate Nb from the Laves phase and allows γ'' phases to precipitate effectively, leading to maximum strength.

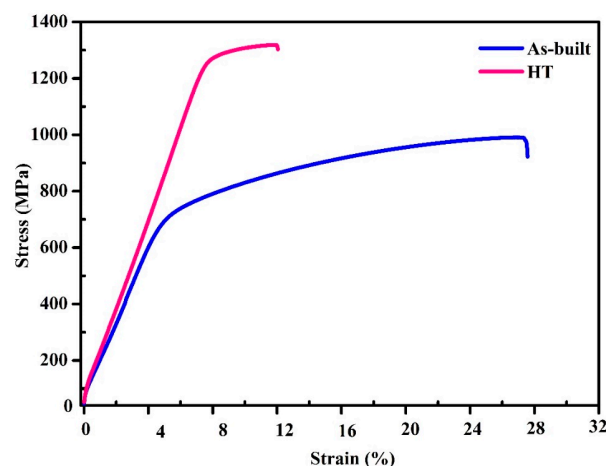


Figure 10. Stress–strain curves of as-built and HT samples.

The fracture surface morphologies of as-built and HT samples are given in Figure 11 to further elaborate on the tensile characteristics of the samples as a function of their microstructural features. Both the fracture surfaces of the as-built and HT samples display large and small dimples. As reported previously, the HT specimens having precipitated hard and brittle phases (δ and Laves) lead to a higher strength with a significant compromise in ductility. However, the fracture surface of HT samples still shows dimple formation, which indicates ductile failure [44,45].

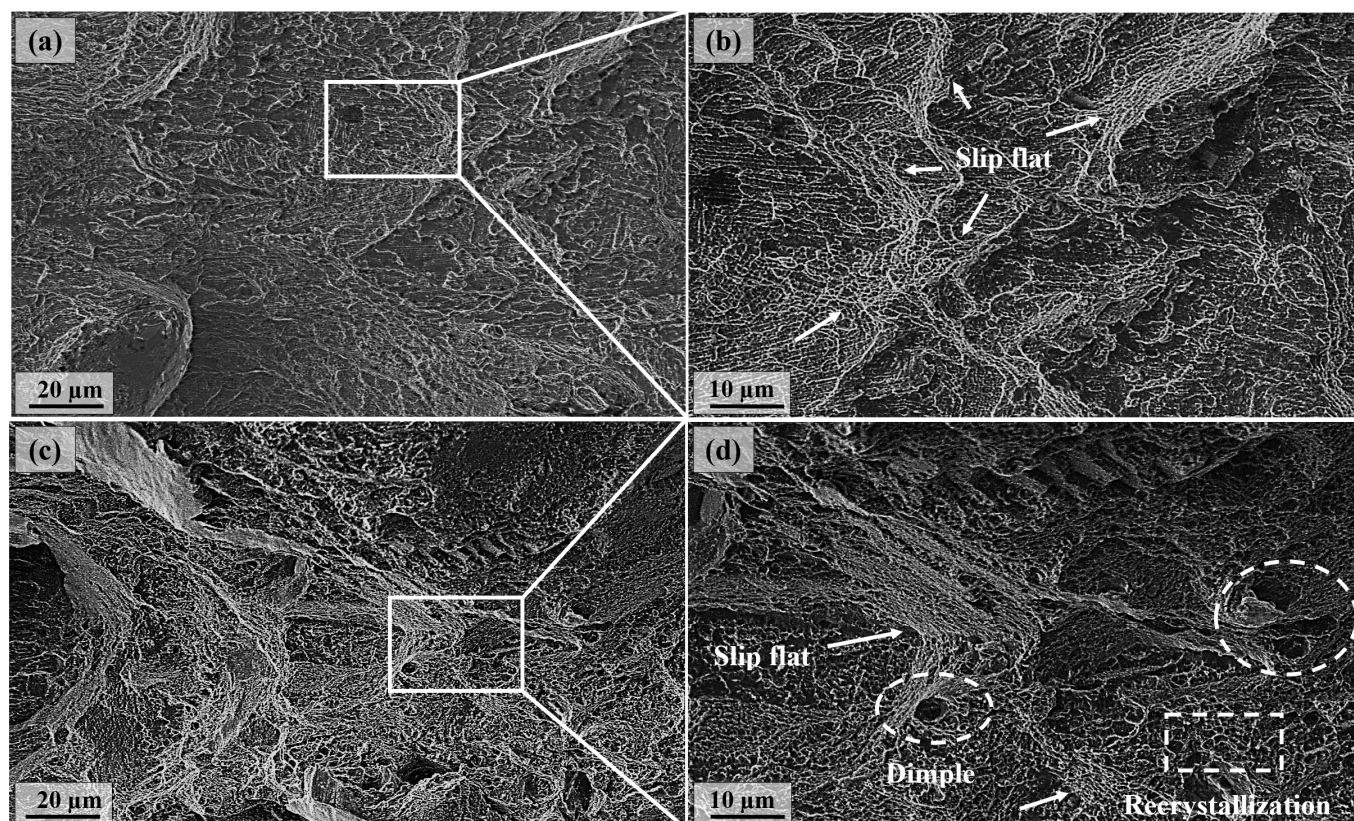


Figure 11. Fracture surface morphology of as-built (a,b) and HT (c,d) samples.

3.3. Tribological Properties of As-Built and HT Samples

Figure 12 displays the coefficient of friction (CoF) curves for as-built and HT surfaces, illustrating two distinct stages: a running-in stage and a steady-state stage. Both the as-built and HT surfaces showed signs of the running-in stage within the initial 15 m of sliding distance. During the initial stage, the CoF rapidly increased in both samples due to the limited contact area [46]. Once the steady-state stage was established, the CoF remained consistently stable throughout the duration of the tests. The average coefficient of friction values for the as-built and HT samples at a specific sliding distance were 0.452 ± 0.03 and 0.632 ± 0.04 , respectively. After experiencing solid solution and aging treatment, the CoF value decreased by approximately 40%. The wear rates of as-built and HT samples are $195 (\pm 30) \times 10^{-6} \text{ mm}^3/\text{Nm}$ and $97 (\pm 20) \times 10^{-6} \text{ mm}^3/\text{Nm}$, respectively, agreeing well with the measured CoF values.

The worn surface morphology of as-built and HT samples is shown in Figure 13. The worn surface of the as-built sample has spalling and fragmentation, and shear bands enlarged into the cracks, while the worn surface of HT samples has limited micro-cutting. The plastic deformation produces adhesive layers for both samples [47]. As-built samples have experienced delamination as a result of severe plastic deformation during the wear process, indicating adhesive wear is most predominant. HT samples exhibit lower delamination

and mostly shallow micro-scratches, indicating the dominant wear mechanism is abrasion. Thus, heat treatment seems to decrease the adhesive wear on the surface, while a limited amount of abrasive wear is still visible. This agrees well with the reduced CoF and wear rate for the HT samples compared to the as-built samples.

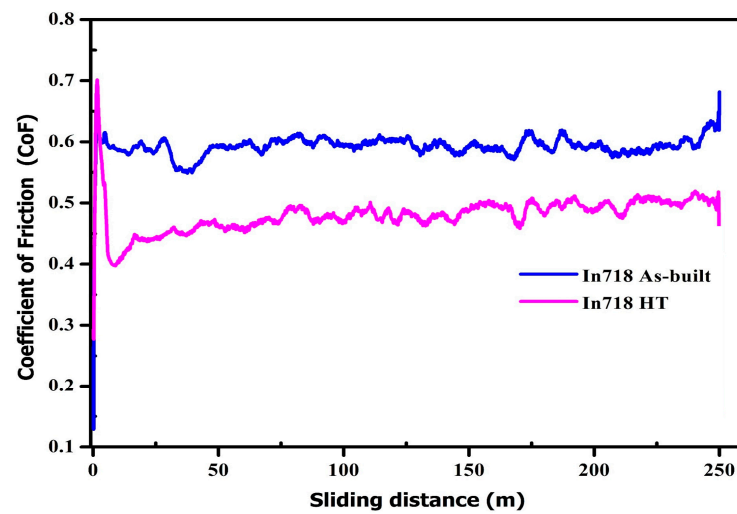


Figure 12. CoF vs. sliding distance for as-built and HT samples.

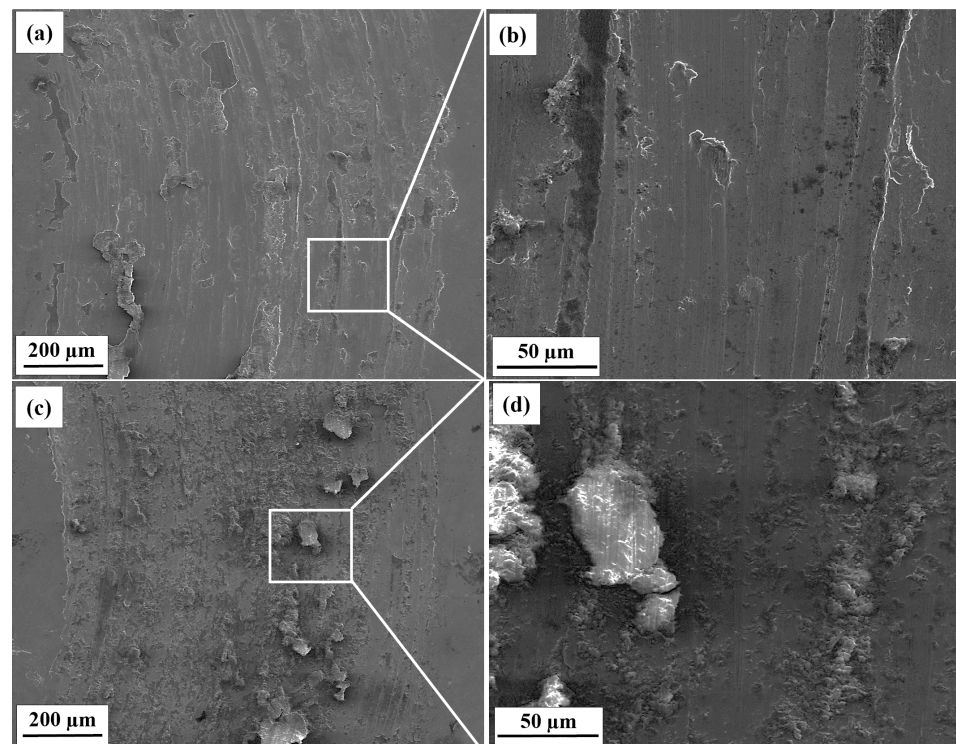


Figure 13. Worn surface morphology as-built (a,b) and HT (c,d) samples.

4. Discussion

4.1. Modification of Microstructural Characteristics with Heat Treatment

It is clear that the microstructural characteristics of as-built In718 alloy significantly change with heat treatment. The as-built In718 sample showed grains that were surrounded by sub-grain boundaries in Figure 7a,b. The as-built sample possesses both cellular and columnar microstructural features, and average dendrite arm spacing is less than about 2–4 μm . Small pores with sizes ranging between 1 μm and 4 μm are observed in both

the as-built and HT samples due to the rapid cooling. Teng et al. [43] observed cellular and columnar structures for In718 samples processed via SLM. Xie et al. [48] also reported cellular microstructures for SLM-processed In718 alloy. The IPF map of In718 samples (Figure 9) displays a columnar structure resulting from epitaxial growth in the XZ plane. The distinctive process of SLM results in a predominant alignment of heat flow parallel to the structural direction during solidification, leading to the formation of a robust $\langle 001 \rangle$ texture (Figure 9b). The resultant textures are mostly related to the SLM processing factors, such as scanning speed and laser power [49,50]. Consequently, it is justifiable to claim that the texture structure acquired in this study corresponds with the results of previous studies in the literature. For example, Wang et al. [49] showed that several layers along the build direction (Z) consisted of typical columnar grains in SLM-processed In718 alloy, with the resulting texture being affected by the scanning speed.

In the context of rapid solidification during SLM processes, reduced phase precipitation and elevated solute supersaturation may occur, which facilitates solid phase transformation during subsequent heat treatment. The δ phase typically exhibits an orthorhombic crystal structure, with a Nb atomic content exceeding 8% and a precipitation temperature between 860 and 995 °C. During the heat treatment of In718 alloy, several Nb-rich Laves phases dissolved into the γ matrix. Since Nb had a large driving force of atom diffusion, Nb distributed in the matrix more homogeneously under solid solution treatment (1093 °C) [43]. As noted above, Nb was the main constituent element of δ phases (Figures 7 and 8), and the uniform distribution contributed to the random precipitation of δ phases at both inside grains and grain boundaries. After heat treatment (Figure 7c,d), it can be found that the cellular structure in as-built In718 has largely disappeared and acute grain boundaries have appeared, indicating that recrystallization has occurred in the high-temperature heat-treated specimen. At the same time, a large number of microscopic precipitates were observed within the grains and at the grain boundaries, containing granular and precipitated phases; a small number of irregular precipitates of large size were also present.

The recrystallization rate of the as-built and HT samples is shown in Figure 14. The yellow areas in the images depict the substructured grains, while the blue regions indicate the grains that have undergone recrystallization. Figure 14 also provides statistical data on the recrystallization percentages of the XZ and XY planes. The recrystallization of the XZ plane in the as-built state increased from 2.42% to 14.1% after heat treatment. Similarly, the recrystallization rate in the XY plane increased from 7.2% to 16.2% after heat treatment. Consequently, the application of heat treatment resulted in the formation of newly formed grains within the structure, leading to a reduction in grain size and a decrease in dislocation density, as also shown by IPF maps and SEM studies, respectively (Figures 7 and 9).

After receiving heat treatment, the sample exhibited a transformation in both its crystallographic texture and dislocation density, resulting in a random texture and reduced dislocation density, particularly in the XZ direction. However, in the XY plane, the grains were aligned along the $\langle 001 \rangle$ direction, and their size decreased because of heat treatment. The robust texture in the $\langle 001 \rangle$ orientation remained unaltered. Li et al. [50] reported similar results for heat-treated SLM-processed In718 alloy. In their study, coaxial grains with $\langle 001 \rangle$ texture were predominant in the XY plane, and the average grain size was determined to be 20 μm . Thus, in parallel to recent literature studies [51,52], the direction of the thermal gradient and the competitive growth of the grains influence the formation of the particular crystallographic texture in the present study.

The GOS maps of the samples display the regions with the lowest and highest dislocation density, as indicated by the colors blue and red, respectively (Figure 15). The GOS values of the samples slightly decreased after heat treatment in both the XZ and XY

planes. The formation of new grains through recrystallization in the microstructure caused a reduction in the deformation density of the structure [53]. The results obtained regarding the impact of heat treatment on the dislocation density of the material produced by SLM are consistent with the published data [54–56].

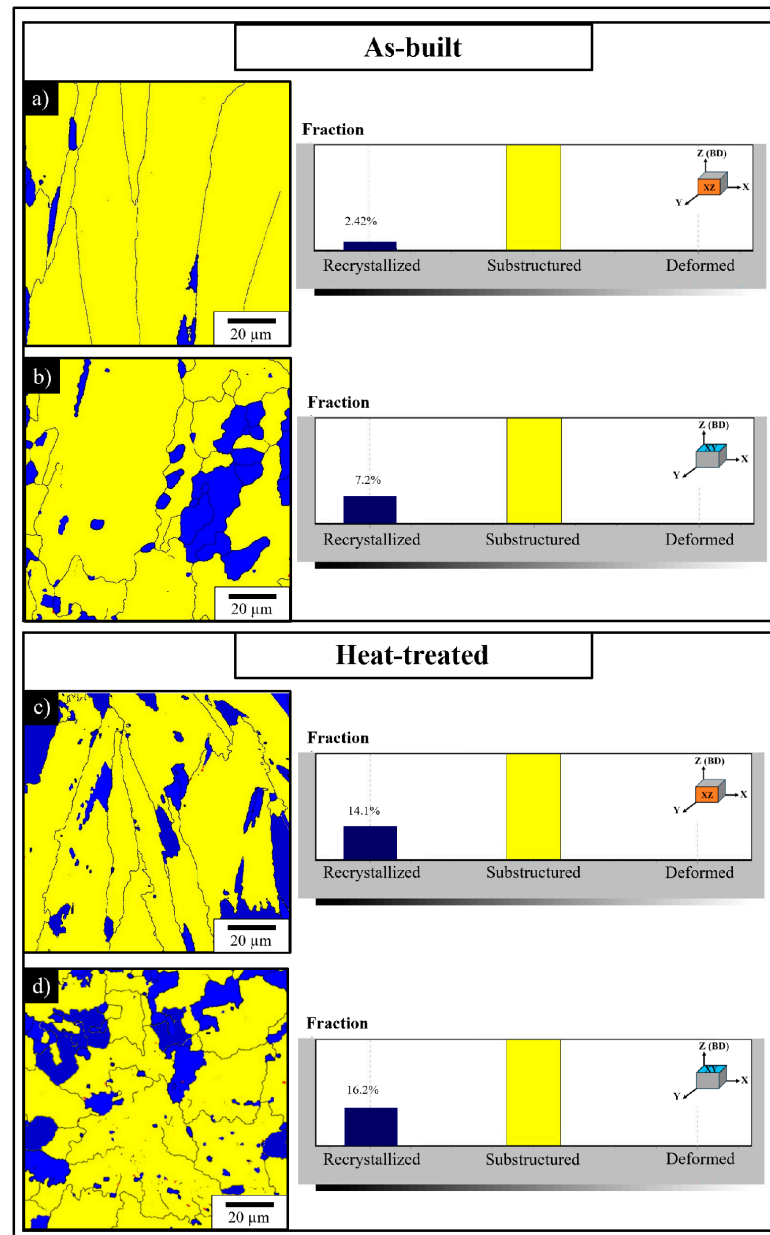


Figure 14. The RC maps of In718: As-built in (a) XZ and (b) XY orientations; HT in (c) XZ and (d) XY orientations.

4.2. Variation of Mechanical Properties with Heat Treatment

Figure 16 compares the yield strength (YS), tensile strength (UTS), elongation, and hardness results with those in the literature [24,31,36,57–60]. YS, UTS, and elongation values are strongly dependent on those of the corresponding as-built and HT In718 samples, which range between 650 and 1100 MPa, 900 and 1400 MPa, and 11 and 35%, respectively. The heat treatment has tended to increase the YS, UTS, and hardness, but decrease the elongation [24,31,36,58–62].

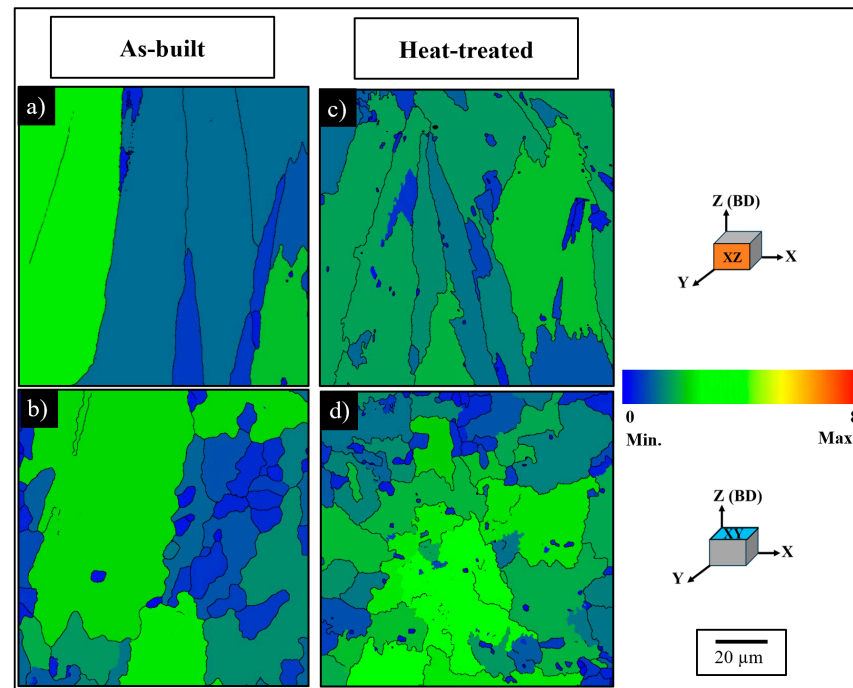


Figure 15. The GOS maps of In718: As-built in (a) XZ and (b) XY orientations; HT in (c) XZ and (d) XY orientations.

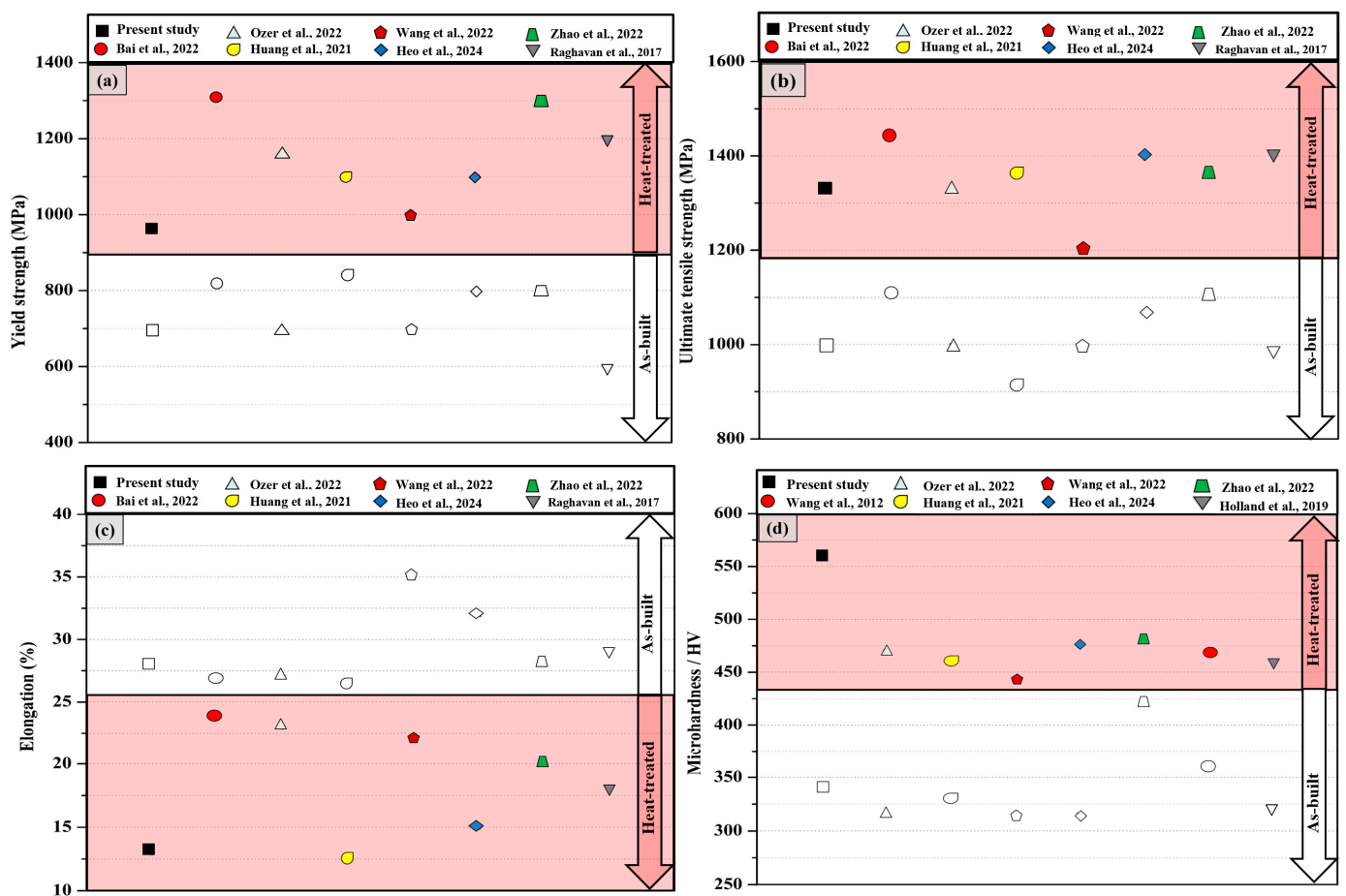


Figure 16. Comparison of the YS (a), UTS (b), elongation (c), and hardness values (d) obtained as a function of In718 loading with literature studies [24,31,36,57–62].

The results can be elucidated by the phase development subsequent to heat treatment conditions. The development of strengthening phases enhanced tensile strength; concurrently, the mobility of dislocations was impeded by these phases, resulting in a reduction in ductility. Furthermore, the emergence of δ phases diminished ductility. Following heat treatments, the columnar grains and small cellular structures were altered by recrystallized grains, resulting in a decreased density of defects at the melt pool boundaries. Consequently, the anisotropy was suppressed. Teng et al. [43] studied the effect of precipitated strength phases on mechanical characteristics and determined that the precipitated phases were brittle (δ and Laves phase) and allowed limited plastic deformation. Upon encountering the δ and Laves phases, dislocations would become crack sources and promote crack propagation. The enhancement in tensile strengths may be ascribed to the precipitated γ'' phases. Nonetheless, the unevenly distributed phases affect the motion of dislocations. The dislocation theory posits that precipitated phases may impede dislocation movement, resulting in an elevated dislocation density. In this study, the YS, UTS, and elongation values of both as-built and HT In718 samples align with existing literature.

4.3. Tribological Behavior

Generally, the dry contact sliding of nickel alloys, such as the Inconel series, involves multiple wear mechanisms that occur simultaneously. The worn surface morphologies of as-built and HT In718 samples are presented in Figure 13. As-built samples have spalling and fragmentation, and shear bands enlarged into cracks. Due to severe plastic deformation during wear, as-built samples have delaminated, indicating adhesive wear [63]. This phenomenon may arise from the temperature generated at their interface during the wear test [64] and also from augmented plastic strain at the contacting asperities. The HT samples exhibit micro-cutting and abrasion [47], showing lower delamination and shallow micro-scratches, indicating abrasion dominates [65]. After the heat treatment, wear debris particles of distinct sizes were produced on the worn surface (Figure 13c,d), indicating that abrasive wear is dominant. Nevertheless, it has been demonstrated that the wear of the sample after HT was minimal, characterized by fewer fine grooves and wear debris, indicating that the surface of the HT In718 alloy possesses high mechanical stability and compactness.

In the present study, heat treatment reduced the coefficient of friction (CoF) by approximately 40% (Figure 12). Similar to our study, Karabulut et al. [14] reported that the CoF of SLM In718 alloy decreased by approximately 30–40% after heat treatment. In another study, Zhao et al. [47] concluded that the CoF of SLM-processed In718 samples decreased after heat treatment. The reduction of the CoF can be attributed to the transformation of the melt pools into equiaxed grains and recrystallization after heat treatment. Also, the increased hardness with heat treatment also decreased the CoF [14]. The superior friction and wear characteristics of the alloy surface can be ascribed to the finely dispersed strengthening phases (γ' and γ'') precipitated within the microstructure after HT, supported by the enhanced surface hardness [24]. Similar results were observed by Rong et al. [66] for the SLM-processed In718, reporting that the correlation between microhardness and microstructure significantly influences the CoF and wear behavior. Furthermore, the wear rate of the as-built samples was approximately twice that of the heat-treated samples, which correlates with the diminished CoF values of In718 samples after heat treatment ($195 (\pm 30) \times 10^{-6} \text{ mm}^3/\text{Nm}$ vs. $97 (\pm 20) \times 10^{-6} \text{ mm}^3/\text{Nm}$).

5. Conclusions

This study focuses on revealing the changes in the microstructural features, mechanical properties, and tribological behavior of SLM-produced In718 after heat treatment:

solutionizing at 1065 °C for 1 h, followed by double aging at 720 °C for 8 h and 620 °C for 6 h. The study's critical findings are outlined below:

The as-built sample displayed a grain structure predominantly defined by fine Laves phases, while the hardening phases γ' (Ni₃(Al, Ti)) and γ'' (Ni₃Nb) precipitated during heat treatment. After heat treatment, the melt pools formed during the SLM process disappeared, and new grains were formed through recrystallization. We observed a change in crystallographic texture and dislocation density, leading to a random texture and diminished dislocation density, particularly in the XZ direction. The heat treatment reduced grain size in the XY plane but did not affect texture in the <001> direction.

The elongation value of the as-built sample is approximately 58% greater than that of the HT sample. The results clearly indicate that heat treatments substantially improve the YS and UTS in accordance with the hardness values while significantly decreasing the elongation at failure. This phenomenon is attributed to the precipitation of the strengthening γ'' and γ' phases within and the intergranular dissolution of the Laves and δ phases within the microstructure. The wear resistance of heat-treated samples surpassed that of the as-built sample due to enhanced mechanical properties resulting from the fine and dispersed γ' and γ'' precipitates in the microstructure with heat treatment.

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