

ORIGINAL RESEARCH ARTICLE

Tailoring the microstructure through laser processing parameters for property optimization of a high-entropy alloy fabricated by laser powder bed fusion

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Abstract

Laser powder bed fusion (LPBF) offers a promising route to high-entropy alloys (HEAs) with refined microstructures. Here the individual roles of laser power and scanning speed in governing the densification, microstructure, and tensile behavior of FeNiCrCo HEA are systematically examined, with particular attention to effects that go beyond the volumetric energy density (VED) descriptor. Through orthogonal experiments (laser power: 80–160 W, scanning speed: 400–1200 mm/s), the optimal processing window was identified at 160 W and 400 mm/s, achieving a relative density exceeding 99.8%. The as-built alloy exhibits a single-phase face-centered cubic (FCC) structure with homogeneous elemental distribution, consisting of columnar grains with subgrain boundaries composed of dense dislocation networks. Notably, laser power and scanning speed exert differentiated effects that VED alone cannot fully capture: increasing laser power promotes the formation of low-angle grain boundaries and enhances texture intensity, while varying scanning speed produces distinct effects on grain morphology and dislocation density. The ultimate tensile strength ranges from 596 to 635 MPa, increasing with laser power but decreasing with scanning speed, whereas the total elongation remains near $33 \pm 4\%$ for all conditions. This study establishes that decoupling laser power and scanning speed effects is essential for precise microstructural control in LPBF-fabricated HEAs.

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1. Introduction

Since their conceptual introduction by Yeh *et al.* in 2004,¹ high-entropy alloys (HEAs) have reshaped the conventional philosophy of alloy design. Unlike conventional alloys that rely on one or two principal elements, HEAs contain multiple principal elements in near-equiatomic proportions, typically five or more, with each element ranging from 5 to 35 at.%.^{2,3} Such multi-principal-element compositions give rise to four characteristic effects, namely the high-entropy effect, severe lattice distortion, sluggish diffusion kinetics, and the cocktail effect.⁴ Together, these effects endow HEAs with exceptional properties such as outstanding cryogenic toughness, remarkable strength-ductility combinations, and

superior oxidation, corrosion, and radiation resistance.^{5–8} Among the various HEA systems, the FeNiCrCo alloy, characterized by its single-phase face-centered cubic (FCC) structure, has been extensively investigated owing to its exceptional formability and satisfactory ductility.^{9,10} These attributes position FeNiCrCo-based HEAs as promising candidates for demanding structural applications in aerospace, automotive, and energy sectors.¹¹

Despite these advantages, the practical application of FeNiCrCo HEAs fabricated by conventional methods such as casting faces a critical bottleneck: insufficient room-temperature strength.¹² The yield strength of cast FeNiCrCo HEAs typically falls below 300 MPa, which is inadequate for most structural applications.¹³ To address this limitation, researchers have explored various strengthening strategies, including grain refinement through severe plastic deformation, precipitation hardening via alloying additions, and second-phase reinforcement through the incorporation of ceramic nanoparticles.^{14–17} However, many of these approaches inevitably compromise ductility, leading to the well-known strength-ductility trade-off.¹⁸

Among additive manufacturing techniques, laser powder bed fusion (LPBF) has emerged as an attractive route for producing metallic components that combine complex geometries with superior mechanical performance.^{19–23} During LPBF, the material experiences extremely high heating/cooling rates (roughly 10^6 – 10^8 K/s), rapid solidification, and repeated thermal cycling.^{24,25} These unique thermal conditions enable the production of materials with refined microstructures, non-equilibrium phases, and enhanced mechanical performance that are unattainable through conventional manufacturing routes.^{26–28} The layer-by-layer deposition nature of LPBF also offers unparalleled design flexibility for creating components with customized geometries and functionally graded structures.^{29–31}

The volumetric energy density ($VED = P/(\nu \cdot h \cdot t)$, with P the laser power, ν the scanning speed, h the hatch spacing, and t the layer thickness) is widely adopted as a global indicator of the densification and microstructural state of LPBF-built parts.^{32–34} Nevertheless, accumulating evidence indicates that, at a fixed VED, the individual choices of laser power and scanning speed modify the thermal history, solidification kinetics, and defect populations in ways that a single VED value cannot capture.^{35,36} Understanding these nuanced effects is essential for achieving precise microstructural control and optimizing mechanical properties.

One of the most distinctive microstructural features of LPBF-processed metals is the presence of dislocation network structures at subgrain boundaries.^{37,38} These

networks, formed during rapid solidification to accommodate thermal contraction-induced strain, have been identified as key contributors to the exceptional strength-ductility synergy observed in additively manufactured materials.³⁹ The dislocation networks act as effective barriers to dislocation motion during deformation, thereby enhancing strength while preserving ductility.⁴⁰ However, the formation mechanisms and microstructural characteristics of dislocation networks in LPBF-fabricated FeNiCrCo HEAs, and their dependence on processing parameters, remain insufficiently elucidated.

Brif *et al.*⁹ first demonstrated the feasibility of producing high-strength and high-ductility FeNiCrCo HEA via LPBF in 2015. Subsequent studies have explored the printability, microstructure evolution, and mechanical properties of various HEA systems, including CoCrFeMnNi,^{41–43} AlCoCrFeNi,⁴⁴ and FeCoCrNi-based alloys.^{34,45} These investigations have revealed that LPBF-fabricated HEAs typically exhibit epitaxially grown columnar grains, cellular and columnar subgrain structures, and dense dislocation networks—characteristics that are distinctly different from conventionally processed counterparts.^{46,47}

Despite these advances, a systematic understanding of how laser power and scanning speed individually influence the microstructural evolution and mechanical properties of FeNiCrCo HEAs remains lacking. For this alloy system, however, how each parameter separately regulates the grain morphology, texture, dislocation network, and mechanical response remains unclear. To address these knowledge gaps, the present study systematically investigates how laser power and scanning speed, acting independently, govern the densification, microstructural and texture evolution, and tensile response of LPBF-fabricated FeNiCrCo HEAs (Figure 1a). Through orthogonal experimental design, we establish the optimal processing window and elucidate the processing–microstructure–property relationships. Particular emphasis is placed on investigating the differential effects of laser parameters on the microstructure and properties of HEAs. The findings of this study provide critical insights for tailoring processing windows and achieving desired material performance in LPBF-fabricated HEAs.

2. Materials and methods

2.1. Powder materials

Pre-alloyed FeNiCrCo HEA powder (99.99% purity, 15–53 μm) produced by gas atomization served as the feedstock. The powder morphology and particle size distribution are presented in Figure 1b. Overall, the powder exhibits good flowability. The powder particles are predominantly

spherical, with a small fraction of agglomerated particles appearing as satellites or elongated cylinders, which is typical for gas-atomized powders.

2.2. LPBF fabrication

An Ampro SP100 platform fitted with a Gaussian-beam fiber laser (maximum output 200 W) was used for the LPBF experiments. Throughout fabrication, the chamber was back-filled with high-purity argon so that the oxygen concentration remained below 200 ppm. A 316L stainless steel substrate was used as the building platform.

The LPBF process parameters are summarized in Table 1. The layer thickness was set to 24 μm , and the hatch spacing was 90 μm . A 67° rotation scanning strategy was applied between successive layers. Laser power was varied between 80 and 160 W and scanning speed between 400 and 1200 mm/s, and VED was evaluated as $P/(v \cdot h \cdot t)$. Figure 1c presents the small block specimens fabricated using the orthogonal experimental design.

As shown in Figure 1d, cuboid samples ($10 \times 10 \times 5 \text{ mm}^3$) were built for density and microstructural analyses, whereas dog-bone tensile coupons (gauge length 35 mm, gauge width 3.16 mm) were produced for mechanical evaluation.

2.3. Microstructural characterization

Relative densities were determined by the Archimedes drainage method. Specimens were ground with diamond sandpaper (180# to 3000#) and polished with 2.5 μm and 0.04 μm SiO_2 polishing suspensions. Microstructures were examined with an MV3000 optical microscope (OM) and a MIRA 3 LMH field-emission SEM coupled to an Aztec Energy X-Max 20 EDS detector. Phases were identified on a Rigaku SmartLab SE diffractometer ($\text{Cu K}\alpha$, $\lambda = 1.54056 \text{ \AA}$) by scanning $2\theta = 10^\circ - 80^\circ$ in 0.02° steps. Electron backscatter diffraction (EBSD) analysis was performed using a TESCAN CLARA SEM equipped with a Symmetry 2 EBSD detector. Transmission electron microscopic (TEM) observations were carried out on an FEI Talos-F200S instrument at 200 kV, using foils prepared by mechanical grinding and subsequent Gatan PIPS ion milling.

2.4. Mechanical testing

Room-temperature tensile tests were performed on an Instron universal testing machine at a strain rate of $1 \times 10^{-2} \text{ s}^{-1}$, consistent with values adopted in previous tensile studies on LPBF-fabricated HEAs; given the low strain-rate sensitivity of FCC HEAs at room temperature, the measured strength and elongation represent quasi-static behavior. Each condition was repeated at least three times. The geometric dimensions of the specimen are shown in Figure 1e.

3. Results and discussion

3.1. Densification behavior

The relative density of LPBF-fabricated FeNiCrCo HEAs as a function of VED is presented in Figure 2a. The density values generally exceed 90% across all processing conditions and exhibit a non-linear relationship with VED. Within the VED range below 80 J/mm^3 , the relative density shows a clear positive correlation with VED, as the limited energy input causes incomplete melting, leaving unmelted particles and interlayer voids. In the VED range of $80 - 130 \text{ J/mm}^3$, the density exhibits some fluctuation, which can be attributed to the differentiated effects of laser power and scanning speed on the melting and solidification behavior.

The highest relative density, approaching 99.8%, was achieved at a VED of 185.2 J/mm^3 (laser power: 160 W, scanning speed: 400 mm/s). Optical microscopy observations (Figure 2b) reveal that the primary defects in the as-built specimens are gas pores and unmelted powder particles. Higher laser power progressively shrinks and eliminates the pores, whereas faster scanning promotes defect formation. Overall, the relative density of the fabricated parts increases with VED. However, the distinct effects of laser scanning speed and laser power are also evident. At an identical VED of 92.6 J/mm^3 , switching the parameter combination from 120 W/600 mm/s to 160 W/800 mm/s raises the relative density by nearly two percentage points (Figure 2a), and the corresponding cross-sections (Figure 2b) contain far fewer pores. Furthermore, it can be observed in this study that even at a VED of 81 J/mm^3 , due to the higher laser power applied, the internal pores in the sample are fewer compared to the specimen printed with parameters of 120 W and 600 mm/s, resulting in a higher relative density. A similar trend was also observed in the samples at VEDs of 138.9 J/mm^3 and 123.5 J/mm^3 . Laser power and scanning speed therefore act on the densification level in distinct ways, and VED alone cannot fully characterize their combined influence.

Table 1. LPBF processing parameters

Parameters	Values
Laser power (W)	80, 100, 120, 140, 160, 180
Laser scanning speed (mm/s)	400, 600, 800, 1000, 1200
Powder layer thickness (μm)	24
Hatch spacing (μm)	90

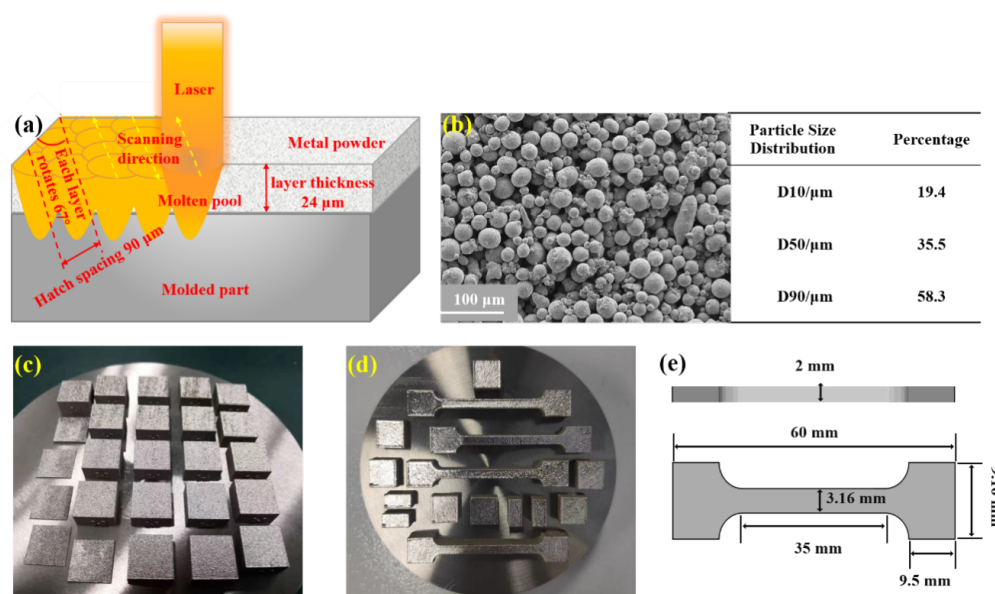


Figure 1. Schematic of the LPBF process, feedstock powder characteristics, and specimen geometries. (a) Schematic diagram of LPBF process and process parameters. (b) SEM images and particle size distribution of pre-alloyed FeNiCrCo high-entropy alloy powder. (c) Bulk formed parts with parameter optimization of LPBF high-entropy alloys. (d) Test characterization specimens. (e) Schematic diagram of tensile specimen dimensions. Abbreviations: LPBF: Laser powder bed fusion; SEM: Scanning electron microscopy.

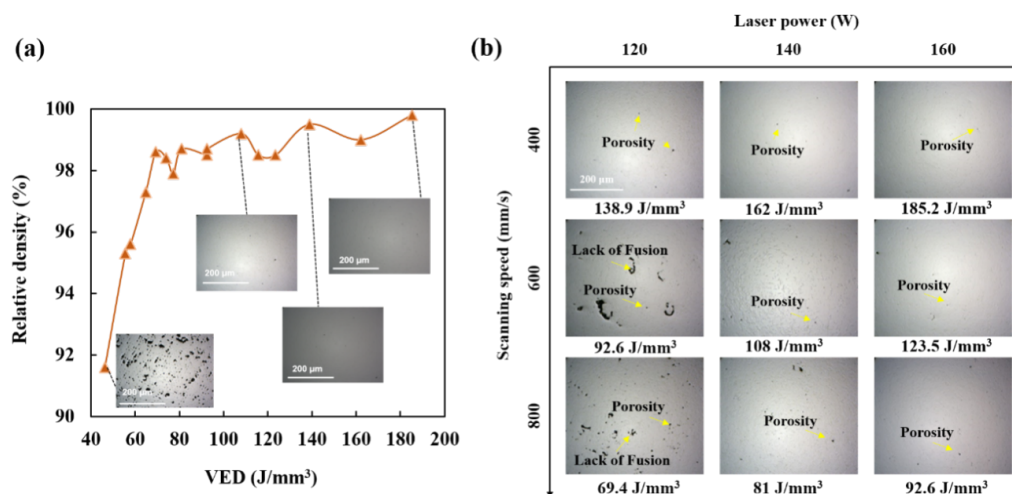


Figure 2. Relative density versus VED with optical micrographs of the defects. (a) Variation curve of the relative density of LPBF-processed FeNiCrCo HEAs as a function of VED. (b) Optical micrographs of LPBF-fabricated FeNiCrCo samples under different combinations of scanning speed and laser power.

Abbreviations: HEA: High-entropy alloy; LPBF: Laser powder bed fusion; VED: Volumetric energy density.

3.2. Microstructural characteristics

3.2.1. Grain morphology and crystallographic texture

Figure 3 presents the SEM micrograph, elemental maps, and X-ray diffraction (XRD) pattern of the alloy built at 185.2 J/mm^3 . The XRD pattern of the specimen reveals a

single-phase FCC structure, with no evidence of secondary phases. Energy-dispersive X-ray spectroscopy (EDS) mapping (Figure 3a) confirms the homogeneous distribution of Fe, Ni, Cr, and Co elements within the matrix, consistent with the rapid solidification characteristics of the LPBF process that suppress elemental segregation.

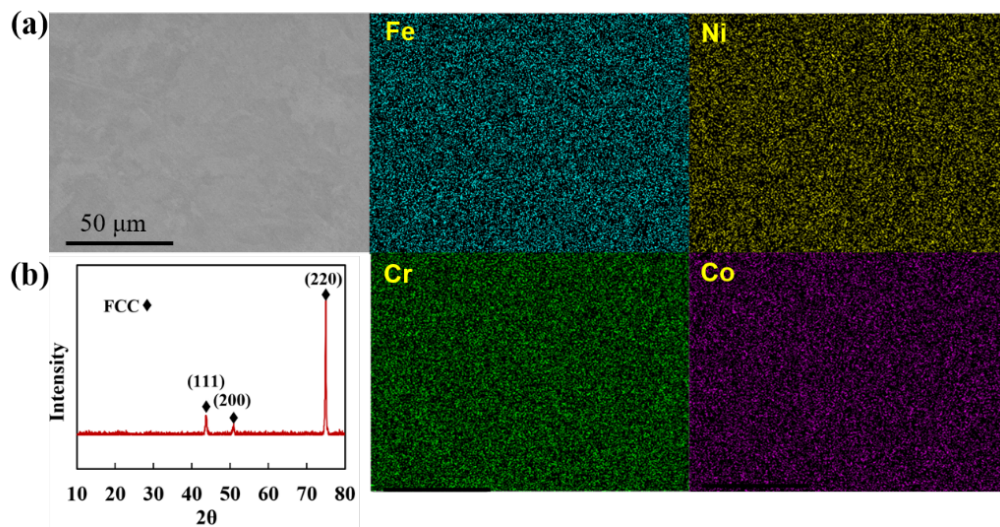


Figure 3. Phase characterization of the as-built alloy. (a) SEM image and corresponding elemental distribution map of the LPBF-fabricated FeNiCrCo sample at a VED of 185.2 J/mm³, and (b) XRD pattern.
Abbreviations: SEM: Scanning electron microscopy; VED: Volumetric energy density; XRD: X-ray diffraction.

Only specimens whose relative density exceeded 98.5% were selected for the crystallographic analyses. The primary objective of these tests is to thoroughly investigate the effects of laser power and scanning speed on the crystal structure of the material. Specifically, two series were characterized by EBSD: one at a fixed power of 160 W with varied scanning speeds, and one at a fixed speed of 400 mm/s with varied powers.

Figure 4 presents the inverse pole figures (IPFs) on the XY plane for the specimens fabricated under different processing parameters. The figures clearly reveal the heterogeneous crystal structure of the LPBF-printed FeNiCrCo. On the XY plane, track lines of approximately 90 μm can be clearly observed. Given their dimensions are comparable to the hatch spacing, they can be identified as the boundaries of laser scanning tracks. Elongated fine grains surround larger grains, and continuous color gradients within certain grains indicate local low-angle misorientations. The grains at the scanning boundaries are significantly finer than those within the scanning tracks. This is because the boundaries of the melt pool are in direct contact with the previously solidified matrix, resulting in a significantly higher solidification rate compared to the interior of the melt pool. Consequently, the grains solidify before they have sufficient time to grow. Additionally, the crystallographic orientation of the grains at the boundaries differs from that of the interior. The fine grains are predominantly red, whereas the coarse grains are primarily yellow-green. This phenomenon occurs because grain

growth follows the heat dissipation direction of the melt pool, which has a distinct angular relationship, thereby leading to the formation of grains with varying orientations. Raising the laser power or lowering the scanning speed markedly enlarges the green (<110>-oriented) regions in the IPFs. Meanwhile, gas pores are also observed in certain regions, which aligns well with the optical micrographs.

Figure 5 shows the pole figures (PFs) on the XY plane for the specimens under different processing parameters. The overall texture strength rises progressively as the laser power increases or the scanning speed decreases. The 120 W/400 mm/s condition shows essentially random grain orientations. However, under other parameter conditions, the specimens exhibit relatively similar orientation characteristics. As can be seen from the (100) pole figures in Figure 5b–5e, the <001> oriented grains undergo a fairly uniform angular rotation along the build direction (BD), forming a ring-like pattern along the outermost circumference. This is attributed to the 67° multi-layer rotation scanning strategy.³⁵ Similarly, ring-like patterns are also observed in the (110) and (111) pole figures, with <110> aligning well with the BD, confirming the presence of a preferred texture. Overall, the specimens do not possess a dominant crystallographic orientation; rather, the <001>, <110>, and <111> directions all serve as preferred growth directions during this fabrication process.

The grain boundary characteristics are shown in Figure 6. At a fixed scanning speed, raising the laser power increases the fraction of low-angle grain boundaries (LAGBs, 2°–15°),

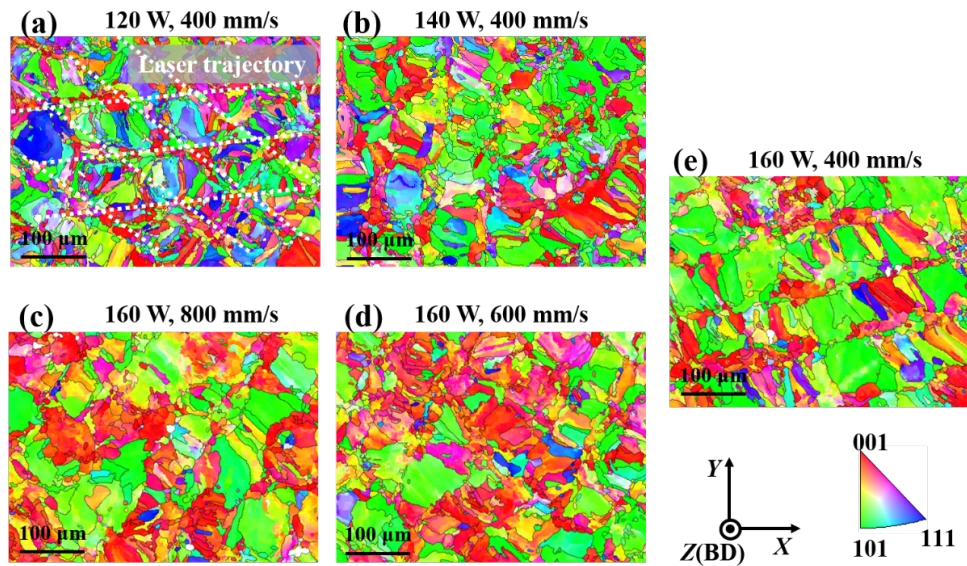


Figure 4. Inverse pole figures (IPFs) with grain boundaries on the XY plane of the samples for the following laser parameters: (a) 120 W–400 mm/s; (b) 140 W–400 mm/s; (c) 160 W–800 mm/s; (d) 160 W–600 mm/s; (e) 160 W–400 mm/s

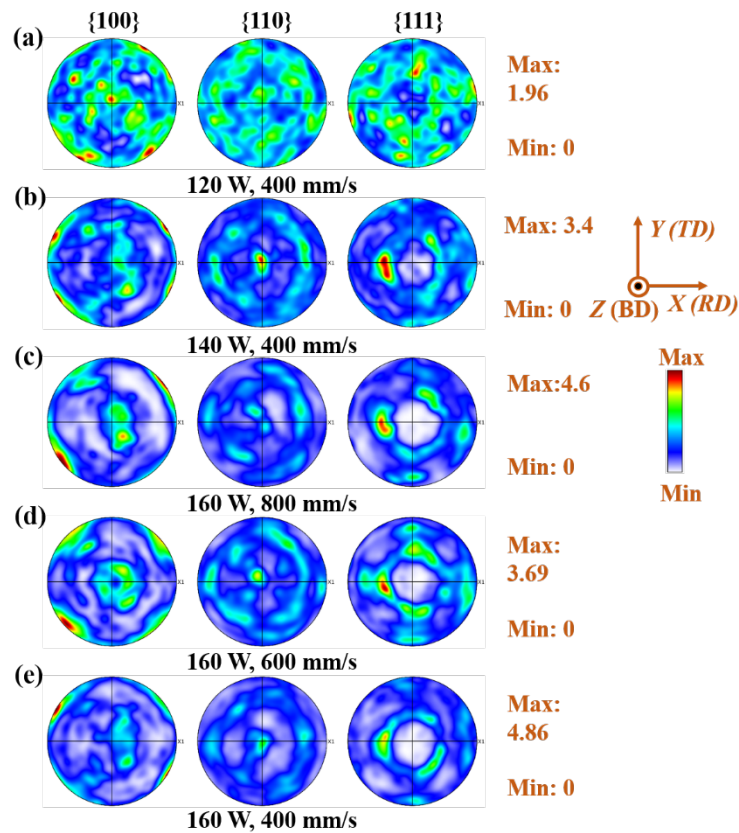


Figure 5. Pole figures (PFs) on the XY plane of the samples for the following laser parameters: (a) 120 W–400 mm/s; (b) 140 W–400 mm/s; (c) 160 W–800 mm/s; (d) 160 W–600 mm/s; (e) 160 W–400 mm/s

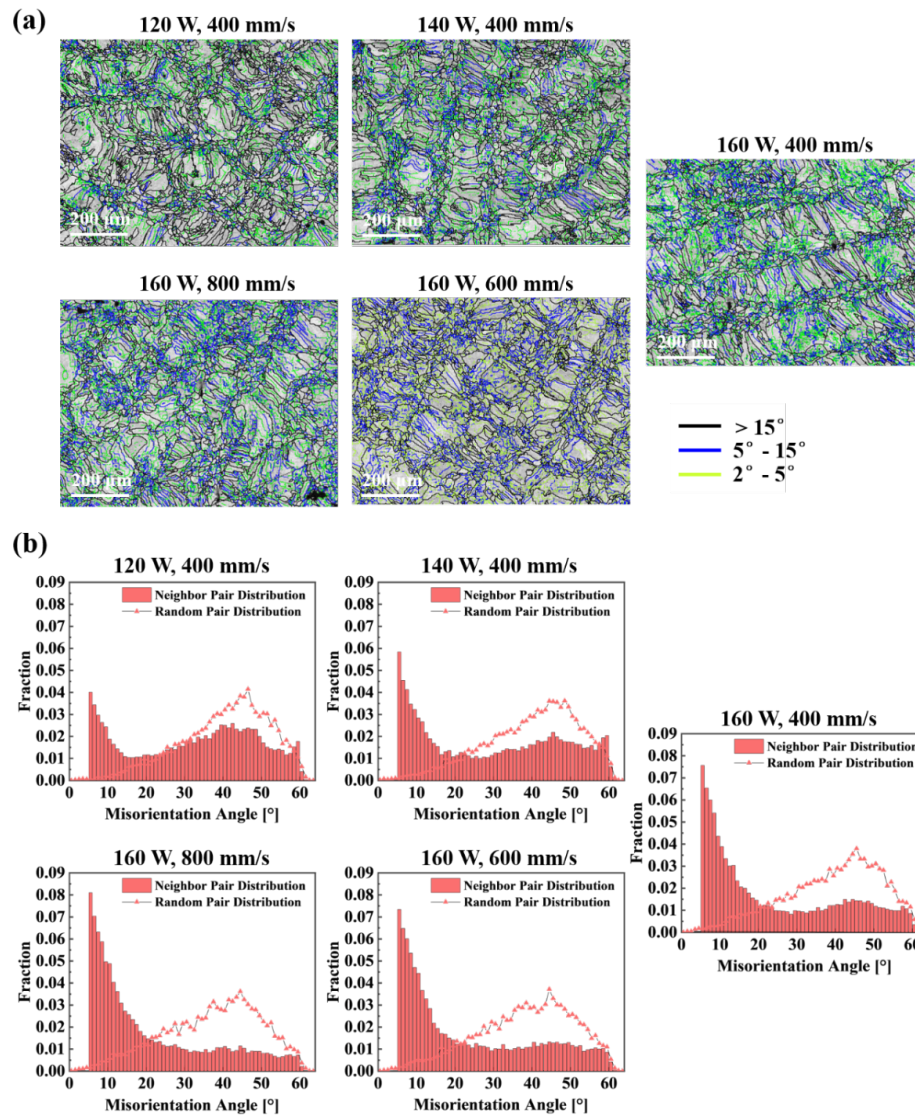


Figure 6. Distribution and statistics of grain boundary characteristics versus laser power and scanning speed. (a) Grain boundary distribution maps on the XY plane of the samples under different laser parameters: 120 W-400 mm/s; 140 W-400 mm/s; 160 W-800 mm/s; 160 W-600 mm/s; 160 W-400 mm/s. (b) Statistical distribution charts of grain boundary misorientations on the XY plane of the samples under different laser parameters: 120 W-400 mm/s; 140 W-400 mm/s; 160 W-800 mm/s; 160 W-600 mm/s; 160 W-400 mm/s.

whereas changing the scanning speed perturbs the overall misorientation distribution only weakly; in the 2°–5° range, for instance, the boundary fraction rises from ~0.04 to ~0.075 as the power increases from 120 to 160 W at 400 mm/s (Figure 6b). However, the fraction of LAGBs in the 2°–5° range shows a slight increase with increasing scanning speed. LAGBs are known to be effective barriers to dislocation motion, contributing to enhanced strength through dislocation–dislocation interactions.⁴⁸

Kernel average misorientation (KAM) maps of the XY

plane are presented in Figure 7a for the different printing conditions. As the laser power increases or the scanning speed decreases, the KAM distribution gradually becomes denser. The KAM signal is concentrated mainly along the grain boundaries. Because the grains at the boundaries of the laser scanning tracks are finer and more densely packed, the KAM values also tend to aggregate at these boundaries, resulting in a gradient distribution of KAM.

During the LPBF process, continuous melting and solidification occur, accompanied by material shrinkage

upon solidification. To maintain material continuity and prevent the formation of thermal cracks, geometrically necessary dislocations (GNDs) are generated at the interfaces with strain gradients to accommodate these strains.^{10,39} As a result, as-deposited samples retain a high GND density even without external loading. Therefore, a high concentration of GNDs is commonly observed in LPBF-printed specimens.^{39,49} Because the KAM value reflects the GND distribution, the resulting heterogeneous dislocation arrangement contributes to strengthening. It should be emphasized, however, that the KAM-derived GND density primarily probes geometrically necessary dislocations associated with local lattice curvature at the EBSD sampling resolution; it therefore provides a lower-bound estimate of the total dislocation density, which additionally comprises statistically stored dislocations such as those imaged by TEM (Figure 8).^{48,50}

Figure 7b presents the statistical charts of KAM on the XY plane of the samples fabricated under different printing parameters. As the laser power increases or the scanning speed decreases, the average KAM value gradually increases, and the GND density exhibits a corresponding upward trend. When the laser parameters are set to 160 W and 400 mm/s, the average KAM value of the specimen is $\sim 1.26^\circ$, and the calculated GND density is $1.81 \times 10^{14} \text{ m}^{-2}$, consistent with Figure 7b. Quantitatively, at a fixed scanning speed of 400 mm/s, increasing the laser power from 120 to 160 W raises the average KAM from 0.86° to 1.26° and the GND density from 1.46×10^{14} to $1.81 \times 10^{14} \text{ m}^{-2}$, whereas at a fixed power of 160 W the KAM increase following a reduction of the scanning speed from 800 to 400 mm/s is smaller (average KAM from 1.12° to 1.26°), indicating that laser power is the dominant factor governing dislocation accumulation. The GND density can be calculated using the strain gradient theory⁵¹:

$$\rho_{\text{GND}} = 2\theta / \mu b \quad (1)$$

where θ is the local misorientation angle, μ is the EBSD step size (the unit length over which the orientation gradient is evaluated), and b is the magnitude of the Burgers vector.

3.2.2. Grain and subgrain

The densest specimen (160 W, 400 mm/s, VED = 185.2 J/mm^3) was then examined in detail by optical microscopy, SEM, EBSD, and TEM.

From the metallographic micrograph on the XY plane (Figure 9a), the laser melt tracks between different layers can be clearly observed. The angle between adjacent melt tracks is approximately 67° , which is consistent with the preset laser scanning strategy. Additionally, Figure 9b presents the metallographic micrograph on the XZ plane,

distinctly revealing the layer-by-layer printing feature of the LPBF process. The cross-section of the melt pool exhibits a typical U-shape, being relatively wide but shallow, which is primarily attributed to the Gaussian distribution of the laser beam energy. Due to the superposition of multiple melt pools and layers, the build direction (BD) plane displays an overall wavy morphology. The interlayer bonding is highly dense, with no obvious cracks or other defects present.

Figure 9c illustrates the microstructural morphology of the specimen on the BD plane, which mainly consists of numerous columnar grains with highly diverse growth directions. This morphological characteristic is primarily attributed to the extremely high cooling rate and the complex variation in the temperature gradient direction during the LPBF process. According to Zhang *et al.*,⁵² in LPBF fabrication without interlayer rotation, the grain growth modes exhibit significant differences across cross-sections of varying orientations. Specifically, on the cross-section perpendicular to the scanning direction, the melt pool presents a complete semi-circular morphology, and the columnar grains grow at an inclined angle toward the center of the melt pool; whereas on the cross-section parallel to the scanning direction, the columnar grains grow almost vertically upwards, forming an ordered layered structure. In this study, the columnar grains grow across the fusion line, exhibiting typical epitaxial growth characteristics, which is consistent with previous studies.^{27,46,53} Grains always grow along the direction of the maximum temperature gradient, causing the subgrains to tend to grow perpendicular to the fusion line.³⁵ Because the 67° interlayer rotation reorients the heat-flux direction layer by layer, the columnar grains on the BD-parallel section grow with highly irregular morphology and direction.

A large number of columnar and cellular subgrain morphologies exist within the columnar grains (Figure 9d and 9e). However, these two structures are merely morphological manifestations on cross-sections of different orientations, and cannot comprehensively and effectively reflect the true three-dimensional morphology and structural characteristics of the substructures in space. To further investigate the true 3D structure of the cellular and columnar substructures, an over-etching technique was employed to deeply etch the specimens. Through lateral observation of the cellular substructures, it was found that they exhibit a columnar morphology in the thickness direction (Figure 9f). This indicates that the substructures are essentially all columnar structures, and the two-dimensional cellular and columnar morphologies are merely the results of observing different cross-sections, with the specific principle illustrated in Figure 9g.

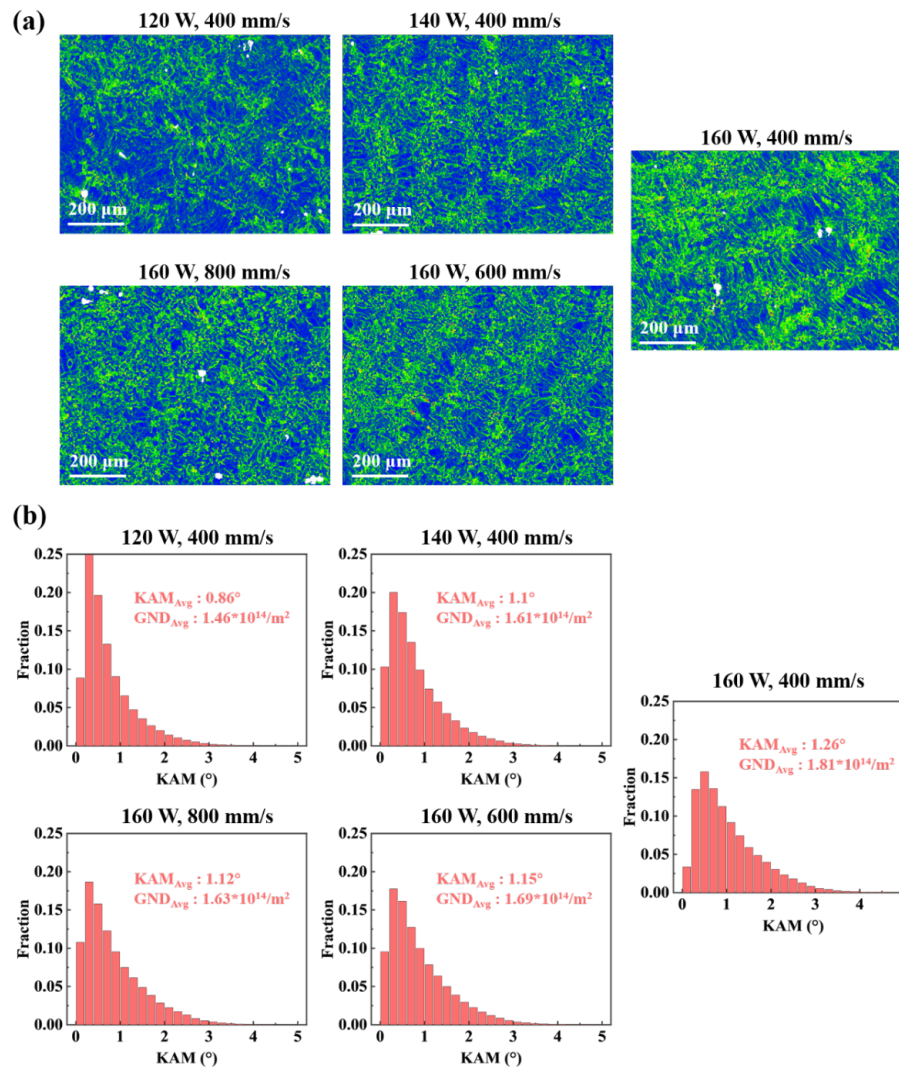


Figure 7. Distribution and statistics of KAM characteristics versus laser power and scanning speed. (a) KAM distribution maps on the XY plane of the samples under different laser parameters: 120 W–400 mm/s; 140 W–400 mm/s; 160 W–800 mm/s; 160 W–600 mm/s; 160 W–400 mm/s. (b) Statistical charts of KAM on the XY plane of the samples under different laser parameters: 120 W–400 mm/s; 140 W–400 mm/s; 160 W–800 mm/s; 160 W–600 mm/s; 160 W–400 mm/s.

Abbreviations: GND: Geometrically necessary dislocation; KAM: Kernel average misorientation.

In the actual LPBF process, the laser scanning of each layer remelts the previous layer. This process causes the microstructure in the original melt pool zone to form diverse structures during the overlapping and remelting processes. This effect leads to chaotic growth directions of subgrains within different grains, with only a very few grains at the bottom of the melt pool retaining their initial orientation. Substructures with various orientations are therefore observed; this directional disorder underlies the simultaneous presence of cellular and columnar substructures.

Figure 10 presents the EBSD characterization of the

sample on the XZ plane. In the microstructure observed on the plane parallel to the build direction (BD), the epitaxial columnar grains are collectively inclined with respect to the BD and display large aspect ratios (Figure 10a). Due to the epitaxial growth mechanism, LPBF-processed materials typically exhibit strong anisotropic microstructures and texture intensities along the BD. However, the 67° interlayer rotation prevents the grain orientations from locking to the BD, producing a texture spread around the Z-axis (Figure 10b). The grain size, evaluated as the number-weighted mean over the EBSD map, is $40.5 \pm 57.8 \mu\text{m}$; the large standard deviation reflects a strongly right-skewed distribution (median $\approx 30 \mu\text{m}$,

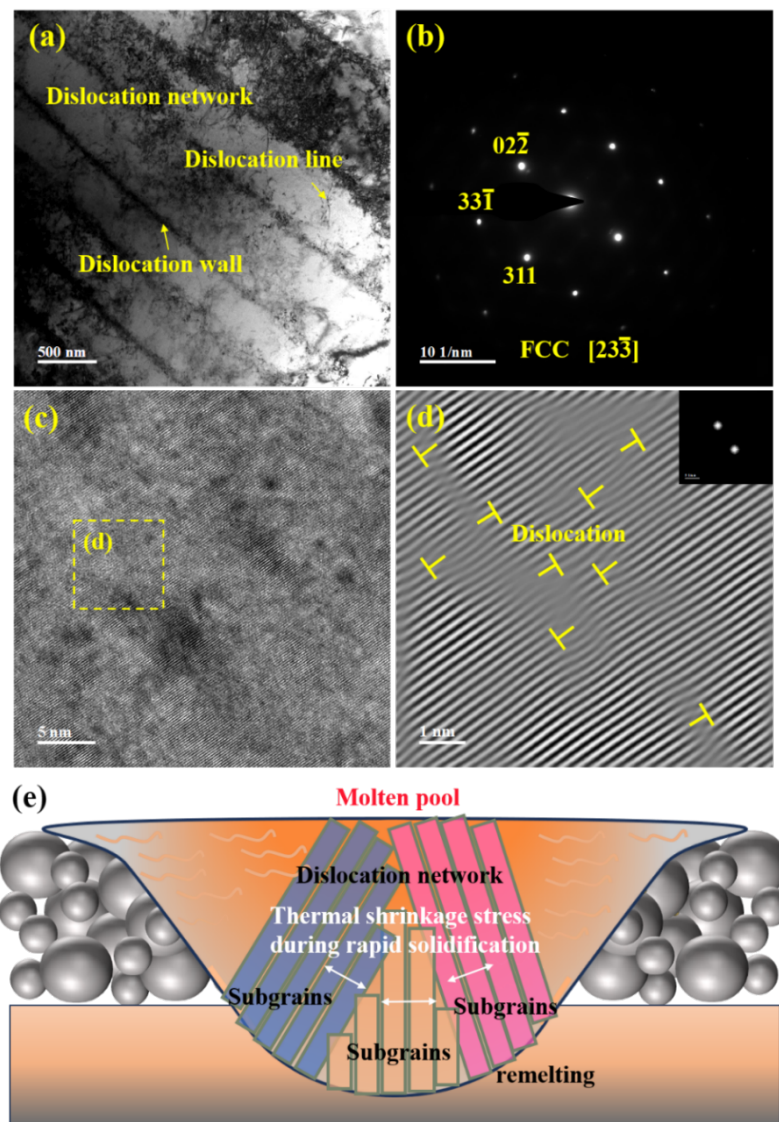


Figure 8. Transmission electron microscopy (TEM) images of the LPBF-fabricated FeNiCrCo high-entropy alloy: (a) dislocation network; (b) indexed SAED pattern; (c) HRTEM image; (d) IFFT image of the region boxed in (c), with dislocations marked; (e) schematic of the dislocation-network formation mechanism

Abbreviations: IFFT: Inverse fast Fourier transform; HRTEM: High-resolution transmission electron microscopy; LPBF: Laser powder bed fusion; SAED: Selected-area electron diffraction.

Figure 10c) produced by a small population of epitaxial columnar grains exceeding 300 μm that traverse several melt pools, confirming epitaxial growth across melt-pool boundaries.

Generally, the growth characteristics of the microstructure during the LPBF process are profoundly influenced by the thermal cycling, temperature gradient, and cooling rate during fabrication. During the manufacturing process, when a new layer is deposited, the melt-pool bottom, being in direct contact with the

underlying solid, solidifies first. Due to the extremely high solidification rate in the LPBF process (10^6 – 10^8 K/s),²⁴ a significant temperature gradient difference exists from the center to the bottom of the melt pool. Because heat is conducted from hot to cold regions, the solidifying microstructure grows antiparallel to the heat flux. Consequently, the microstructure exhibits a tendency to grow perpendicularly to the melt pool boundary towards the center of the melt pool. Furthermore, due to the remelting effect during the LPBF process, when the laser scans the topmost layer, only the grains located at

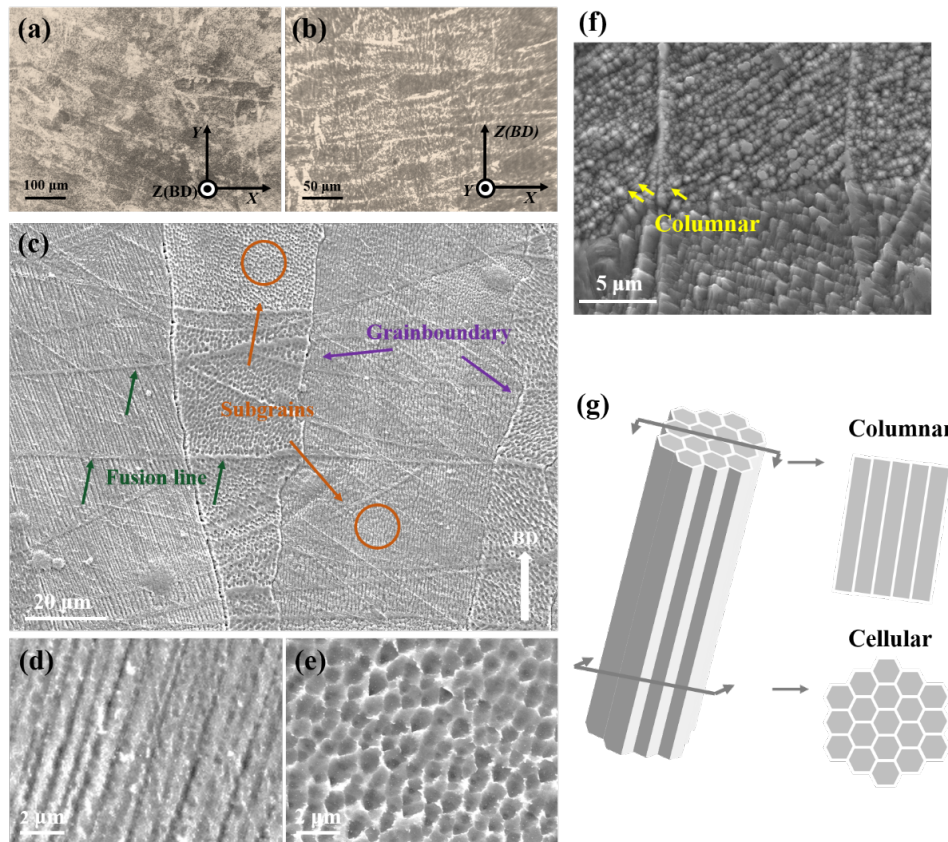


Figure 9. Microstructural characterization of the etched sample at a VED of 185.2 J/mm^3 : (a) Optical microscopic image showing the laser melt track morphology on the XY plane. (b) Optical microscopic image showing the melt pool morphology on the XZ plane. (c) SEM image of the microstructure, along with the corresponding columnar substructure (d) and cellular substructure (e). (f) SEM images of the actual three-dimensional structure of the columnar subgrain morphology after over-etching. (g) Schematic illustration of the columnar subgrain morphology on different cross-sections. Abbreviation: SEM: Scanning electron microscopy.

the bottom of the preceding melt pool survive remelting, and the newly solidified grains inherit their orientation, yielding high-aspect-ratio epitaxial columnar grains.

TEM characterization of the specimens also revealed a columnar substructure, with the substructure boundaries composed of dislocations (Figure 8a). The dense boundaries constitute a dislocation network, which exhibits a parallel columnar structure in the image, showing high consistency with Figure 9. This actually represents the morphological manifestation of the subgrains on the columnar plane, as illustrated in Figure 9g. Such regular dislocation networks are regarded as a signature structural feature of LPBF-processed materials.¹³ A large number of dislocation lines are also present within the columnar subgrains. The indexed selected-area electron diffraction (SAED) pattern confirms a single FCC phase free of compositional segregation or secondary phases (Figure 8b). This is fully consistent with the XRD/EDS results in Figure 3.

Figure 8c displays the high-resolution TEM (HRTEM)

image of the subgrain interior, revealing a massive amount of atomic vacancies and lattice distortions. Through inverse fast Fourier transform (IFFT) analysis (Figure 8d), the dislocations within the yellow box in Figure 8c have been indexed. A very high dislocation density is confined within an extremely small region. These dense dislocations and the dislocation network play a crucial role in strengthening the alloy.

These dislocation networks are generally considered to be induced by the thermal contraction linear strain that occurs during the rapid cooling process of the LPBF. As shown in Figure 8e, during the non-equilibrium solidification of the melt pool, only the FCC phase nucleates in the molten metal. During the layer-by-layer melting and solidification process of LPBF, the molten upper-layer material shrinks upon cooling, but this shrinkage is constrained by the previously solidified lower-layer material.⁵⁴ Consequently, the tensile effect exerted by the upper layer on the lower layer generates linear strain⁵⁵,

leading to significant tensile residual stresses within the solidified material. To accommodate the lattice interfaces and maintain material continuity, dislocation networks are formed.

3.3. Tensile properties

The representative engineering stress-strain curves of LPBF-fabricated FeNiCrCo HEAs under various processing conditions are shown in Figure 11. The ultimate

tensile strength (UTS) ranges from 596 to 635 MPa, with a total elongation of $33 \pm 4\%$. At constant laser power (160 W), the UTS decreases with increasing scanning speed. At constant scanning speed (400 mm/s), the UTS increases with increasing laser power; the corresponding UTS values are approximately 596, 607, 607, 617, and 635 MPa for 120 W–400 mm/s, 140 W–400 mm/s, 160 W–800 mm/s, 160 W–600 mm/s, and 160 W–400 mm/s, respectively (Figure 11). The highest UTS, 635 ± 15 MPa, is obtained for the 160 W–400 mm/s specimen.

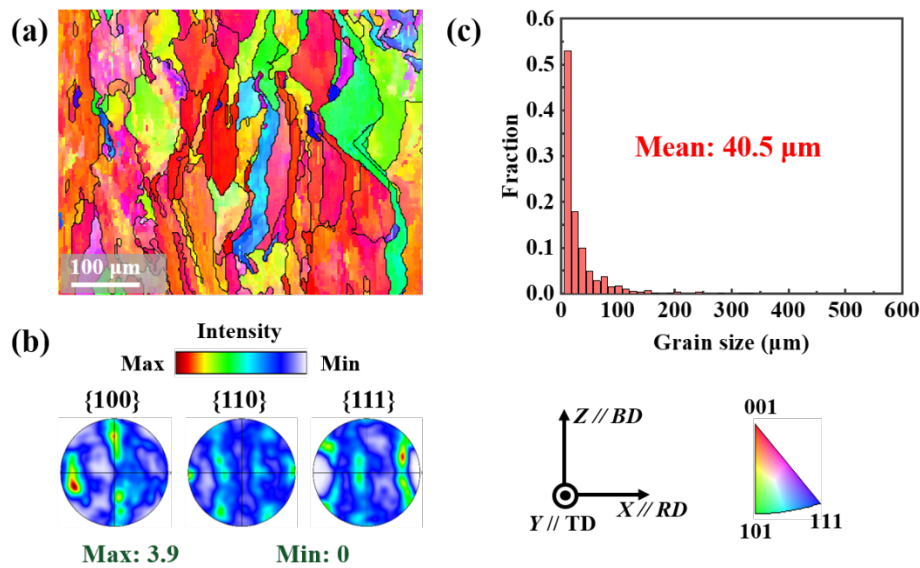


Figure 10. Electron backscatter diffraction (EBSD) characterization of the sample on the XZ plane: (a) inverse pole figure (IPF); (b) pole figure (PF); (c) grain size statistical chart

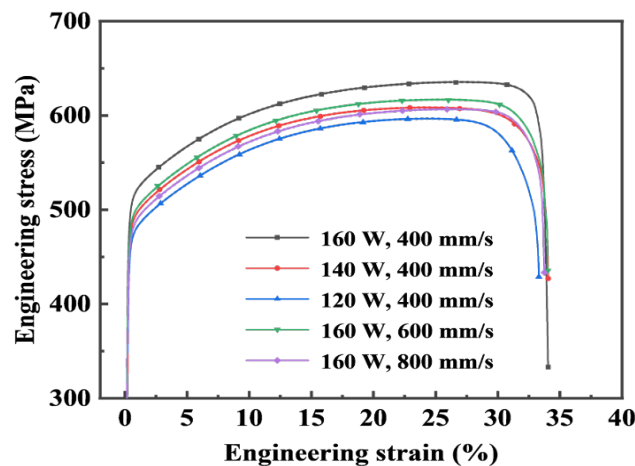


Figure 11. Representative engineering stress-strain curves of the specimens printed with different parameters

Representative fracture surfaces of the tensile specimens are compared in Figure 12 for the different laser parameters. As can be seen from Figure 12a₁–12e₁), all specimens exhibit a certain degree of necking, indicating the characteristics of ductile fracture. Meanwhile, numerous large-sized defects are observed in the specimens. In contrast, the specimens fabricated with optimized parameters have relatively fewer pores. Compared to the undeformed specimens observed under optical microscopy, the pore sizes in the fractured specimens have increased manifold, suggesting that pores in the LPBF specimens are the primary cause of premature fracture.

The fracture surfaces of all specimens are relatively flat, indicating that the material primarily undergoes transgranular fracture, which is another characteristic of its high ductility. Additionally, a small amount of quasi-cleavage fracture features is observed in the fractures

(Figure 12a₂–12e₂). This morphology is quite common in LPBF-processed samples, which is attributed to the reduced plasticity caused by the high defect density.⁵⁶

High-magnification SEM observations of the fracture surfaces reveal that dimples of varying sizes are formed in all specimens, serving as an important feature of ductile fracture (Figure 12a₃–12e₃). Large and small dimples are distributed alternately in a strip-like pattern. Based on the micro-morphology of the specimens, it is speculated that this phenomenon may be induced by the regular dislocation networks. The plastic deformation is inconsistent between the subgrain interiors and the dislocation networks. During the deformation process, the dislocation networks hinder dislocation movement but still maintain the continuous flow of plastic deformation. Consequently, relatively smaller dimples are generated at the dislocation networks.

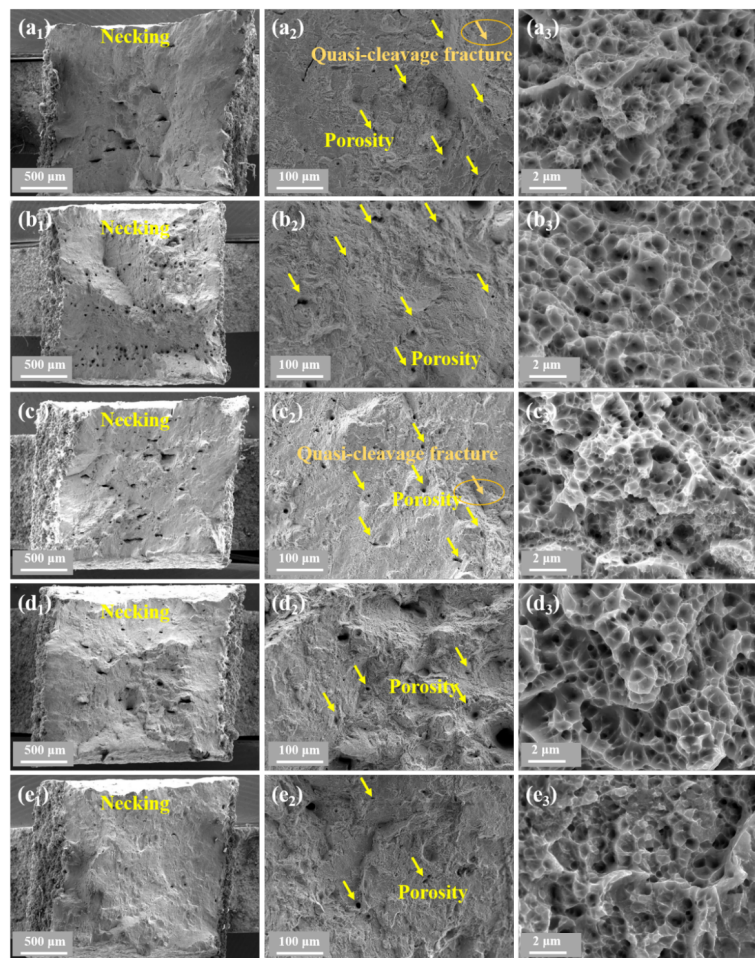


Figure 12. Tensile fracture morphologies of the LPBF-printed specimens under different laser processing parameters: (a₁–a₃) 120 W, 400 mm/s; (b₁–b₃) 140 W, 400 mm/s; (c₁–c₃) 160 W, 800 mm/s; (d₁–d₃) 160 W, 600 mm/s; (e₁–e₃) 160 W, 400 mm/s

4. Conclusion

In this study, the respective influences of laser power and scanning speed on the densification, microstructural evolution, and mechanical response of LPBF-fabricated FeNiCrCo HEAs were systematically investigated. The following conclusions can be drawn:

- (i) *Processing optimization and parameter decoupling:* The optimal processing window was identified at 160 W and 400 mm/s, achieving near-full densification (>99.8%). Critically, laser power and scanning speed affect the densification, grain structure, and texture evolution in ways that VED alone cannot capture, demonstrating that the two parameters must be decoupled for precise process control.
- (ii) *Microstructural evolution and dislocation network formation:* The as-built alloy exhibits a single-phase FCC structure with columnar grains and epitaxial growth characteristics. Dense dislocation networks form at subgrain boundaries during rapid solidification to accommodate thermal contraction-induced strain. Increasing laser power promotes LAGB formation and intensifies crystallographic texture, while GND density increases with higher laser power or lower scanning speed.
- (iii) *Mechanical performance and strengthening mechanism:* The UTS ranges from 596 to 635 MPa, increasing with laser power and decreasing with scanning speed, whereas the elongation remains at approximately $33 \pm 4\%$ irrespective of the parameters. The excellent strength-ductility synergy arises from the dislocation network structure, which effectively impedes dislocation motion while preserving ductility.

Overall, this work establishes a processing-microstructure-property framework showing that independently controlling laser power and scanning speed enables rational microstructural engineering and property optimization in LPBF-fabricated HEAs.

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Conflict of interest

The authors declare they have no competing interests.

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Conceptualization: Tan Shu, Chee Kai Chua

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Investigation: Tan Shu

Methodology: Tan Shu

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Ethics approval and consent to participate

Not applicable.

Consent for publication

Not applicable.

Availability of data

The data presented in this study are available upon request from the corresponding author due to ethical or legal restrictions on public sharing.

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