

Coupling effects of laser cladding parameters on the morphology and phase evolution of FeCoNiCrCu high-entropy alloy coatings

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Article Info	Abstract
Article History: Received 15 July 2026 Accepted 09 Sep 2026 Keywords: Laser cladding; FeCoNiCrCu high-entropy alloy; Coaxial powder feeding; Interfacial shear strength; Dilution rate; Microstructure; Phase precipitation; Annealing	Laser cladding of high-entropy alloys is an effective surface-strengthening technology, yet the coating formation mechanism still requires further optimization. In this work, FeCoNiCrCu high-entropy alloy powder and a 45-steel substrate were adopted to investigate laser-cladding behavior under four representative process states: under melting, baseline, high-energy over melting, and high powder-feeding rate. The powder was designed for equiatomic fraction; the nominal mass fraction of each element was 20 wt%, which deviates from strict equiatomic ratio. The results demonstrate that the under melting condition triggers incomplete powder melting and coating defects with a porosity of 4.23 %. Excessive heat input under the high-energy condition yields a high substrate dilution rate of 37.26 % and enlarges the heat-affected zone. A high powder-feeding rate induces elemental segregation. Under the baseline condition, the molten pool spreads uniformly, defects are suppressed, and elemental diffusion is sufficient, achieving the minimum porosity of 0.32 % and the optimal interfacial shear strength of 426.24 MPa. The high-energy over melting specimen (Condition 3) reaches an average hardness of 339.62 HV, though this process brings obvious substrate over-melting risk. The as-deposited coating consists of FCC and BCC solid-solution phases, whereas laves and σ phases precipitate after annealing. This study reveals the coupling effects of process parameters on coating morphology and phase evolution and provides guidance for process optimization of such quinary high-entropy-alloy laser-cladding systems.

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1. Introduction

Mechanical components frequently undergo prolonged exposure to severe friction and cyclic alternating loads, making them highly susceptible to surface degradation, wear, and subsequent delamination (peeling) failures. Laser cladding High-Entropy Alloy (HEA) coatings become a mainstream technology to improve the comprehensive performance of substrate surfaces at a low cost due to multi-principal element solid solution strengthening characteristics [1-2]. The quinary Iron-cobalt-nickel-chromium-copper (FeCoNiCrCu) HEA has a uniform and adjustable composition. It balances hardness and plasticity, showing outstanding application potential in the field of wear-resistant protection. However, the internal relationship among molten pool evolution, element diffusion, and phase precipitation under the coupling of multiple process parameters still lacks complete quantitative analysis. This knowledge gap restricts the standardized industrial adoption of the process [3-4]. To address these challenges, several research groups have investigated the cladding performance of various high-entropy alloy systems. Wen X et al. fabricated FeCoNiCrAl HEA composite coatings reinforced with varying contents of SiC ceramic particles via laser cladding and systematically analyzed their reinforcement mechanisms. They

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systematically analyzed the reinforcement mechanism of ceramic phases on coating microstructures. The hard phase can significantly improve the wear resistance of coatings. This work provides ideas for the design of high-entropy composite coatings [5]. Wu B et al. explored the interdiffusion behavior between multi-layer films of multi-element HEAs and aluminum substrates by means of atomistic simulation. They explained the causes of interfacial bonding quality from the perspective of microscopic atomic migration [6]. Li S et al. reviewed a large amount of existing research. They comprehensively summarized the corrosion evolution laws of HEA coatings prepared by laser cladding. The regulation logic of element segregation and phase structure on corrosion resistance was summarized [7]. Xiao D et al. adjusted the doping ratio of Niobium (Nb) to prepare Chromium-Iron-Cobalt-Nickel (CrFeCoNi) cladding coatings. They clarified that changes in alloy element content directly induce differential transitions in dendrite morphology and mechanical properties [8]. Kaviyarasan K et al. used gas tungsten arc welding to prepare HEA filled joints. They compared the effects of different entropy fillers on the joint microstructure and corrosion resistance [9]. Samoilova O V et al. explored the effects of silver doping on the micro-morphology, hardness, and surface oxidation behavior of AlCoCrCuFeNi HEA. They confirmed that trace alloy elements can change the phase precipitation tendency [10]. Xu K X et al. carried out electrochemical tests to compare the passivation and corrosion characteristics of CoCrFeNi HEA and 304 stainless steels in salt solution. The anticorrosion advantages of multi-element solid solution structures were quantified [11]. Fals H D C et al. improved cavitation and erosion resistance by adding Nb to Inconel coatings. They verified that alloying is an effective way to improve the service life of thermal spray coatings [12]. However, existing research mostly focuses on single element modification or single performance testing. Few studies simultaneously explore the comprehensive effects of multi-parameter coupling, such as laser power and powder feeding rate, on FeCoNiCrCu coating formation, thermophysical properties, mechanics, and phases. Therefore, this study sets four categories of gradient cladding conditions. Combined with characterization methods such as metallography, the changing rules of forming defects, element distribution, and mechanical indicators are quantified. The present work elucidates the synergistic regulation mechanisms the process coupling regulation mechanism to provide experimental support for process optimization of this type of quinary HEA laser cladding. The primary objective is to clarify the evolution of cladding cross-sectional morphology and molten pool spreading status under the synergistic action of laser power, powder feeding rate, and scanning speed. The differences in coating forming quality and mechanical properties under different process matchings are quantified. The innovation lies in coupling multi-indicator analysis. The regulation mechanism of multi-process synergy on the microstructure and defects of HEA cladding layers is fully revealed. This work provides systematic experimental support for the optimization of the laser cladding process of this type of quinary HEA.

2. Materials and Methods

2.1 Experimental Materials and Instruments

This study selects a common structural steel as the substrate and employs the equiatomic FeCoNiCrCu quinary HEA powder to perform laser cladding process coupling experiments. The detailed parameters of the raw materials, reagents, and shielding gas utilized in the experiment are listed in Table 1. The models, manufacturers, and specific applications of the primary laboratory instruments used in this work are listed in Table 2.

Table 1. Materials list

Material Name	Specification & Model	Manufacturer
45 steel (AISI 1045, Chinese GB standard)	100 mm×100 mm×10 mm	Ansteel Co., Ltd.
FeCoNiCrCu quinary high-entropy-alloy (HEA) powder	Particle size 150-300 mesh, sphericity ≥90%, equal-weight mixing ratio, each metallic element accounts for 20 wt%	Beijing General Research Institute of Nonferrous Metals
High-Purity Argon	Purity ≥99.999%	Hangzhou New Century Gas Co., Ltd.

Note: The powder was initially designed for equiatomic composition; the as-supplied powder adopts 20 wt% for each constituent element. Owing to differences in elemental atomic mass, 20 wt% does not yield equal atomic fractions. Calculated atomic fractions: Cr 22.1 at.%, Fe 20.4 at.%, Co 20.2 at.%, Ni 19.2 at.%, Cu 18.1 at.%.

Table 2. Instrument list

Equipment	Model	Manufacturer	Country
3D Laser Profilometer	VR-6000	Keyence Corporation	Japan
Metallographic Specimen Preparation Equipment	Tegramin-20	Struers A/S	Denmark
X-ray diffractometer	D8 ADVANCE	Bruker Corporation	Germany
Vickers microhardness tester	HVS-1000Z	Laizhou Huayin Testing Instrument Co., Ltd.	China
Scanning Electron Microscope (SEM)	Sigma 300	Carl Zeiss AG	Germany
Optical Microscope	Axio Scope A1	Carl Zeiss AG	Germany
Vacuum heat-treatment furnace	SK-V-1200	Shanghai Jvjing Electric Furnace Co., Ltd.	China

2.2 Laser Cladding Experimental Equipment and Process Plan

Surface pretreatments are required for both the substrate and the alloy powder before the cladding experiments. The substrate steel plate is progressively ground with various grits of abrasive papers to eliminate surface oxide scales, rust, and oil contaminants, followed by ultrasonic cleaning in anhydrous ethanol and drying [13-14]. The FeCoNiCrCu HEA powder is dried in a vacuum drying oven at a constant temperature to thoroughly eliminate internal adsorbed moisture. After the pretreatment steps, the FeCoNiCrCu HEA coupling process experiments are conducted on a coaxial powder-feeding fiber laser cladding experimental platform.

The system configuration and working principle are shown in Figure 1.

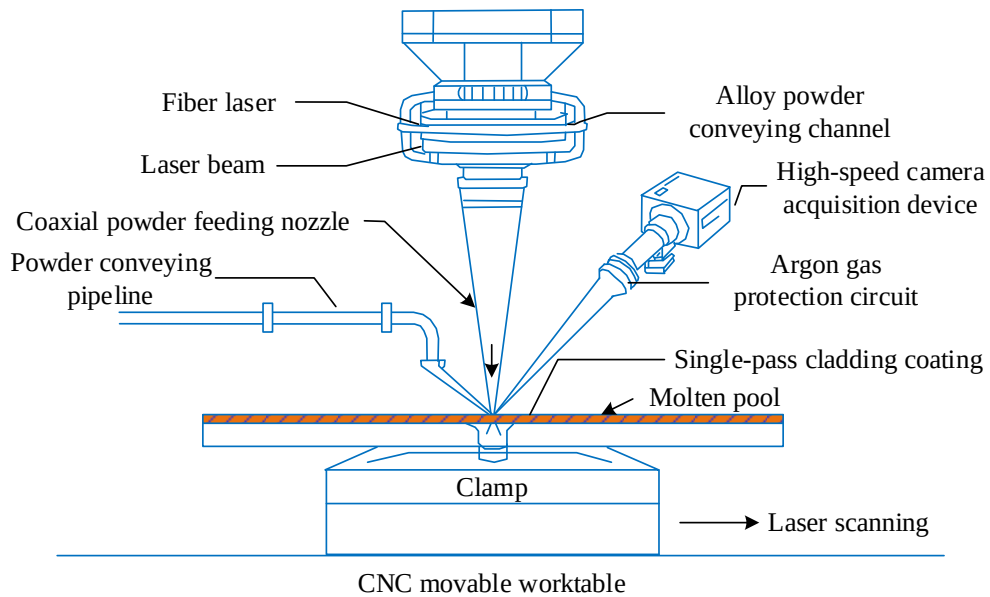


Fig. 1. Laser cladding experimental platform with coaxial powder feeding

In Figure 1, the laser beam emitted from the laser generator converges downward through a focusing nozzle, merging with the FeCoNiCrCu HEA powder delivered via the pipeline to create a molten pool on the substrate surface under continuous Argon shielding. The laser beam adopted a

Gaussian energy distribution with a fixed focal spot diameter of 2.0 mm on the substrate surface. During the process, the substrate workpiece is clamped onto a CNC moving stage. The laser scanning speed is regulated by adjusting the feed speed of the stage, completing the fabrication of cladding samples under diverse process parameters. All the steps in experiments are shown in Figure 2.

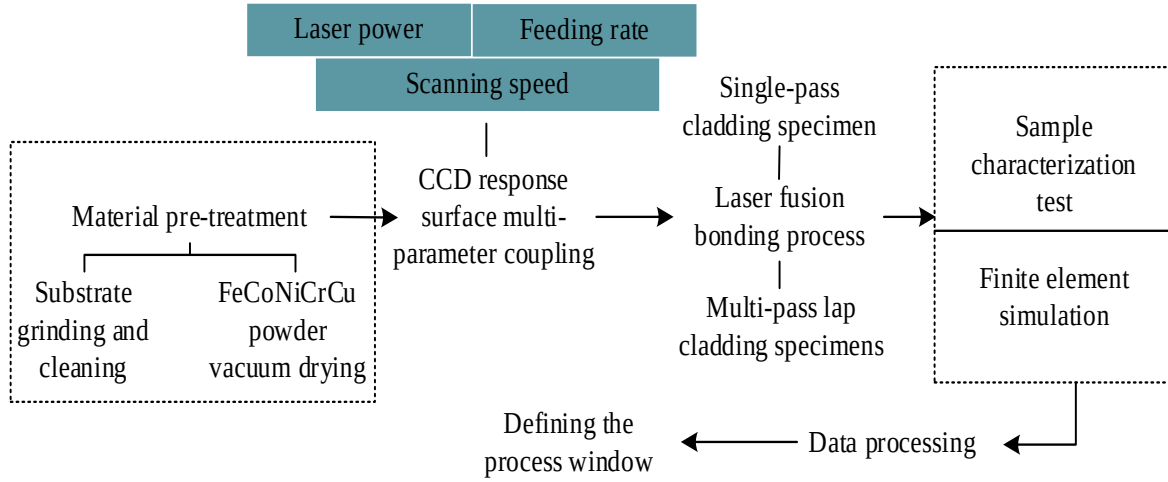


Fig. 2. Experimental technical route

As shown in Figure 2, the substrate undergoes grinding and decontamination in the early stage, while the FeCoNiCrCu HEA powder is dried in a vacuum environment to remove adsorbed moisture. The defocus distance was maintained as a constant value of 0 mm throughout all laser cladding tests. Single-track and multi-track overlapping specimens are fabricated on the laser cladding platform. In the experimental stage, the surface morphology, internal microstructure, and element distribution of the cladding layer are checked and analyzed via a three-dimensional profilometer, an optical microscope, and a scanning electron microscope equipped with energy dispersive spectroscopy. Three key parameters including laser power, scanning speed and powder feeding rate were selected. Four representative coupled process conditions were designed for comparative investigation. The detailed process parameters for the four representative conditions are presented below:

- Condition 1 (under melting): laser power = 1600 W, powder-feeding rate = 8 g/min, scanning speed = 8 mm/s
- Condition 2 (baseline): laser power = 2400 W, powder-feeding rate = 12 g/min, scanning speed = 6 mm/s
- Condition 3 (high-energy over melting): laser power = 3200 W, powder-feeding rate = 12 g/min, scanning speed = 6 mm/s
- Condition 4 (high powder-feeding rate): laser power = 2400 W, powder-feeding rate = 18 g/min, scanning speed = 6 mm/s

The defocus distance was kept constant at 0 mm throughout all laser cladding experiments.

2.3 Sample Preparation Process

After the laser cladding experiments, the fabricated specimens undergo a series of preprocessing steps to facilitate subsequent macro-morphology examination and microstructure characterization. The complete process flow for metallographic sample preparation is shown in Figure 3. In Figure 3, standard-sized specimens are first sectioned from the formed workpieces using wire electrical discharge machining equipment. After hot mounting, the specimens are subjected to coarse and fine grinding using different grits of abrasive papers, followed by mechanical polishing to achieve a mirror-like surface. After grinding and polishing, the sample surface was cleaned with anhydrous ethanol. The metallographic surface was etched by nitric-acid-based corrosive agent for microstructure observation.

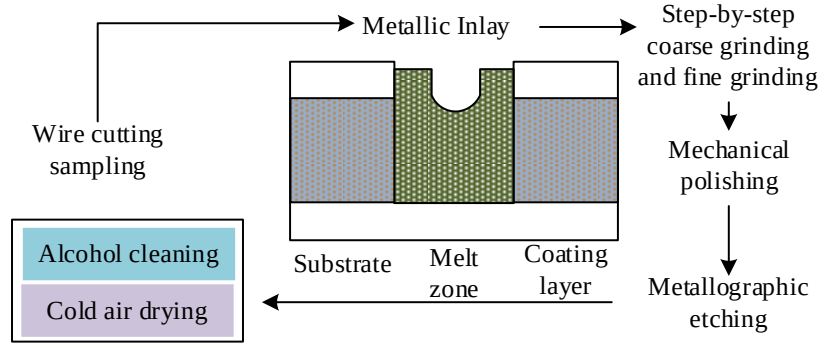


Fig. 3. Manufacturing process flow of metallographic specimens

Diamond lapping paste was adopted for fine polishing during specimen preparation. An appropriate etchant is selected to corrode the cross-section of the specimen to reveal the microstructure of the cladding layer. Finally, the specimens are rinsed with anhydrous ethanol and dried with cold air to complete the preparation. The calculation equation for the cladding Energy Utilization Rate (EUR) is shown in Equation (1) [15].

$$EUR = \frac{m_{\text{powder}} \cdot \Delta H_{\text{alloy}}}{P \cdot t} \times 100\% \quad (1)$$

In Equation (1), m_{powder} denotes the mass of the alloy powder consumed in a single track, m_{powder} is calculated as the product of powder feeding mass rate and single-track scanning duration. ΔH_{alloy} indicates the latent heat of fusion of the FeCoNiCrCu HEA, while P and t represent the actual output power of the laser and the total scanning time of a single track, respectively. The expression for the Specific Energy Consumption (SEC) is shown in Equation (2) [16].

$$SEC = \frac{P \cdot t_s}{m_{\text{powder}}} \quad (2)$$

In Equation (2), t_s is the total scanning time of single-track cladding (s). The calculation for the dilution rate of the cladding layer is shown in Equation (3) [17].

$$\eta = \frac{S_{\text{sub}}}{S_{\text{sub}} + S_{\text{coat}}} \times 100\% \quad (3)$$

In Equation (3), η is the dilution rate, while S_{sub} and S_{coat} represent the cross-sectional area of the melted substrate and the cross-sectional area of the cladded alloy layer within the molten pool, respectively. Part of the single-track laser-cladded specimens were subjected to vacuum annealing treatment. The annealing temperature was set to 800 °C with a holding time of 2 h. The heating rate was controlled at 5 °C/min. After heat preservation, the samples were cooled down to room temperature inside the furnace at a cooling rate of 3 °C/min. The whole heat-treatment process was carried out under a vacuum atmosphere of 5×10^{-3} Pa to prevent oxidation of the coating surface. Two-step chemical etching was adopted to reveal the microstructure of the substrate and the high-entropy alloy coating separately. The 45-steel substrate was etched with 4% nital solution (4% HNO_3 + 96% ethanol). The FeCoNiCrCu high-entropy alloy coating was etched with modified aqua regia solution ($\text{HCl}:\text{HNO}_3:\text{H}_2\text{O} = 3:1:2$) to clearly display dendritic boundaries and phase precipitation structures.

2.4 Specimen Testing and Characterization Methods

Four typical coupled process conditions were selected for comparison. Condition 1 is an under melting condition characterized by low laser energy and the minimum powder feeding rate. Condition 2 is a baseline matching condition featuring a balanced ratio of laser power to powder feeding rate. Condition 3 represents a high-energy over melting condition where the laser output power is increased while maintaining a moderate powder feeding rate. Condition 4 is a high

powder feeding condition where the powder feeding rate reaches the upper experimental limit and the laser power remains at a medium level. The comparative analysis is carried out on all these cladding specimens in a unified manner. The separate test methodologies are listed below:

Macro-morphology observation of single-track cross-sections: Cross-sectional specimens of the coating are cut via CNC wire electrical discharge machining. After mounting, grinding, and polishing, the cross-sectional profiles are observed using an Olympus vertical optical microscope. The cladding width, height, and depth dimensions are measured via ImageJ software according to standard metallographic preparation procedures [18-19].

Dendritic microstructure, surface micro-morphology, and elemental mapping of the coating were observed after a two-step etching procedure. The 45-steel substrate was etched with 4% nital solution, while the FeCoNiCrCu high-entropy alloy coating was etched with modified aqua regia solution (HCl: HNO₃: H₂O = 3:1:2). A Zeiss field emission scanning electron microscope was utilized to capture secondary electron morphologies of the coating surface. The use of an Oxford energy dispersive spectrometer which is attached is done to map the elemental distribution of Cr, Fe, Ni, and Si. Cr, Fe, Ni, Si, Co and Cu were selected for elemental distribution mapping to evaluate the chemical homogeneity and diffusion behavior across the cladding zone. Elemental detection and microstructure observations comply with the operational standards of scanning electron microscopy [20].

Thermophysical property testing of the substrate and HEA coatings: The laser flash method is used to collect thermophysical data at various temperatures, with the heating interval covering room temperature to 2000 K. The testing of thermophysical parameters refers to relevant standards for determining high-temperature thermophysical properties of metallic materials. **Process parameter testing of EUR, SEC, and aspect ratio of the track:** The measurement of the dimensional characteristics of the single-track cladding is done using a laser profilometer of Keyence, which is a three-dimensional laser profiler. The laser power and the mass flow of the powder are measured at the same time. The profile dimension measurement accuracy follows the operational specifications of three-dimensional morphology inspection equipment [21].

Microhardness testing along the layer depth: A Vickers microhardness tester is utilized with a load of 0.1 kgf to perform step-by-step indentation testing from the coating surface toward the substrate. The hardness data collection interval is 0.05 mm, and the Vickers hardness testing of metallic materials strictly complies with the GB/T 4340.1-2018 standard. **Phase analysis of as-deposited and annealed coatings:** The coating surface is ground and polished before undergoing X-ray diffraction scanning, with a scanning angle range of 20° to 80°. The Face-Centered Cubic (FCC) and Body-Centered Cubic (BCC) solid solution phases, as well as intermetallic compounds, are identified according to general phase identification methods for metallic multi-component alloys.

Additional quantitative parameter test method to define quality: Calculation of wetting angle and porosity of the melt pool according to binary statistics of metallographic photographs. Surface roughness is measured using a white light interferometric three-dimensional profiler. The interfacial shear strength tensile test was performed with reference to GB/T 7124-2008. Adapted specimen geometry was adopted for metallurgically bonded coating-substrate samples rather than adhesive-bonded assemblies to measure coating-substrate interfacial shear performance. Thermal cracking susceptibility was evaluated by counting crack length and crack density based on metallographic cross-section micrographs [22].

3. Results and Discussion

3.1 Macro-Morphological Characteristics of Cladding Layers

Figure 4 presents the cross-sectional morphology of the single-track laser cladding coatings under four sets of process parameters. The regulatory effects of process parameters on the molten pool morphology are visual through three dimensional parameters: cladding width *W*, cladding height *H*, and cladding depth *D*.

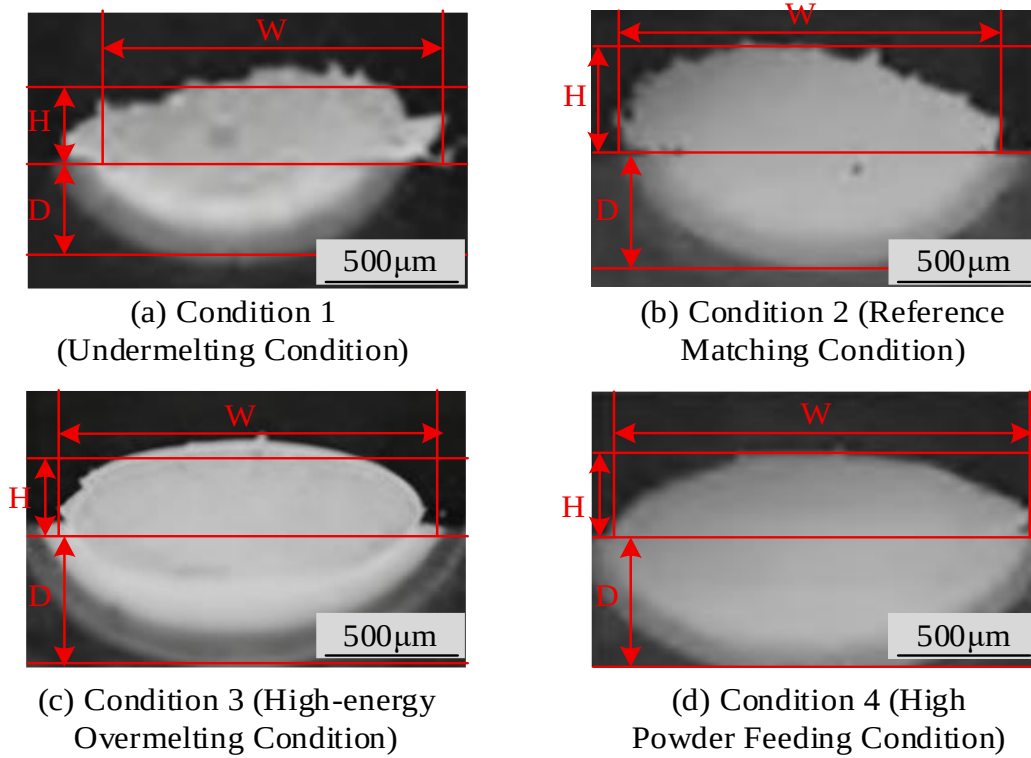


Fig. 4. Cross-sectional morphology of single-track laser cladding coatings

(Condition 1: 1600 W, 8 g/min, 8 mm/s; Condition 2: 2400 W, 12 g/min, 6 mm/s; Condition 3: 3200 W, 12 g/min, 6 mm/s; Condition 4: 2400 W, 18 g/min, 6 mm/s)

As shown in Figure 4(a), under the under melting condition, low laser heat input caused insufficient powder melting. The overall width and height of the cladding layer stayed relatively small, and the molten pool profile remained shallow. Slight irregular fluctuations occurred at the coating edges, which easily induced lack-of-fusion defects and led to poor metallurgical bonding quality. Figure 4(b) illustrates that the baseline condition offered a well-matched energy level. The molten pool exhibited a regular and symmetrical morphology with balanced width, height, and depth dimensions. The coating edges appeared flat and smooth, without obvious defects like pores or inclusions. The continuous and stable bonding interface between the substrate and the coating indicated the best forming quality. In Figure 4(c), excessive laser energy under the high-energy over melting condition significantly increased substrate melting, causing a pronounced enhancement in cladding depth. Although the powder melted fully and the coating formed uniformly, such excessive heat input under high-energy conditions is likely to induce substrate grain coarsening and deterioration of base-metal mechanical properties according to laser thermal-cycling characteristics. Direct microstructural verification of this phenomenon will be conducted in future work. In Figure 4(d), the high powder feeding rate required more laser energy per unit area to melt the powder. The cladding layer height and width increased further, whereas the cladding depth decreased slightly, yet the molten pool spread well. However, excessive powder feeding easily induced uneven powder deposition, leading to forming defects like surface agglomeration. Overall, the cladding layer achieved the highest forming integrity under the baseline condition, representing a relatively reasonable process parameter range.

3.2 Microstructure and Elemental Distribution Patterns of The Coating

Figure 5 displays the surface SEM morphology of the coatings and the EDS elemental mapping results of four major elements (Cr, Si, Fe, and Ni) under the four process conditions. Combining the observations from Figure 5, the coating surface under Condition 1 appeared relatively flat, and the overall element distribution was quite uniform. Si exhibited only scattered dot-like distribution without obvious element-rich zones. Low heat input restricted the degree of elemental diffusion, creating slight local composition segregation. In Condition 2, as the laser energy increased, the coating microstructure became finer and denser. The distribution uniformity of Cr, Fe, and Ni

improved, and Si began to aggregate in small areas, which indicated that elemental diffusion was enhanced. Condition 3 exhibited the most extensive elemental diffusion within the molten pool. The coating surface structure was dense and uniform, and the four elements dispersed globally. Although small Si enrichment zones existed, severe segregation did not occur. Sufficient element mixing was achieved in the molten pool. However, the excessive heat input of Condition 3 also induced undesirable substrate dilution and high cracking susceptibility.

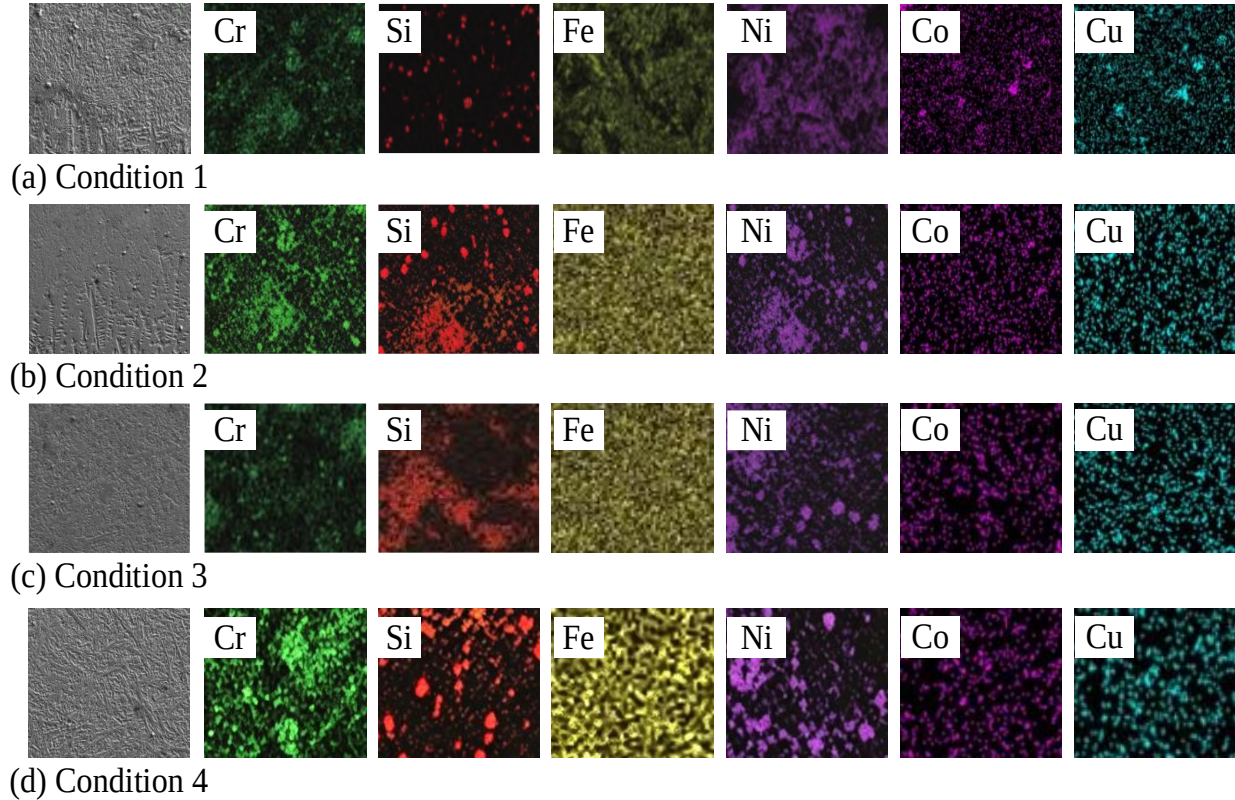


Fig. 5. SEM morphology and representative elemental EDS-mapping results of coatings under four different process conditions

Under Condition 4, the high powder feeding rate caused a notable increase in Cr and Ni enrichment regions and aggravated Si agglomeration, which raised local composition segregation. There was a considerable quantity of powder that melted within a relatively short period of time, but some materials had quickly solidified before there was adequate diffusion. In conclusion, Condition 3 gave the most sufficient diffusion of elements, while Condition 2 delivered balanced comprehensive coating performance with low dilution and cracking risk. Silicon is not an initial constituent element of FeCoNiCrCu powder. The observed Si enrichment originates from trace-element diffusion from the 45-steel substrate into the molten pool during laser melting. Cu element presents scattered local enrichment regions across the coating. This segregation behavior originates from the positive mixing enthalpy between Cu and other constituent metallic elements, which tends to drive Cu atoms to accumulate within interdendritic areas during rapid laser-cladding solidification. No severe segregation was observed for Co element within the detection range of this test.

3.3 Thermophysical Property Differences

Figure 6 demonstrates reference temperature-dependent variations in specific heat capacity and thermal conductivity for both the FeCoNiCrCu HEA coating and 45-steel substrate. Within this temperature range, the variations in specific heat capacity and thermal conductivity between the two materials directly determined the cooling efficiency and high-temperature duration of the molten pool, serving as the core physical foundation for the differential evolution of cladding morphology under diverse processes.

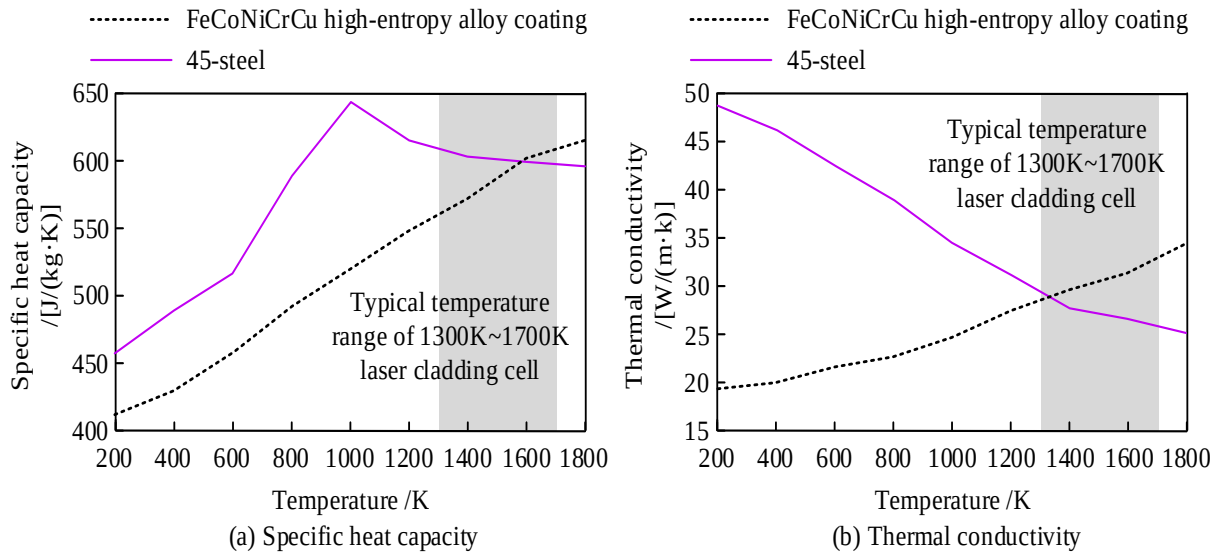


Fig. 6. Reference thermophysical property-temperature curves of FeCoNiCrCu HEA coating and 45-steel substrate

Figure 6(a) depicts the specific heat capacity curves of the two materials. In the low-temperature stage, the 45-steel substrate possessed a significantly higher specific heat capacity than the HEA. As temperature rose, the substrate specific heat capacity increased rapidly first, reaching a peak around 1100 K before dropping slowly. The specific heat capacity of the HEA steadily enlarged with increasing temperature. Within the working temperature range of the molten pool, its specific heat capacity gradually surpassed that of the substrate material. This phenomenon meant that the alloy coating could store more heat during heating, which helped stabilize the temperature field of the molten pool. Figure 6(b) shows the trend of thermal conductivity variation. While the thermal conductivity of the 45-steel substrate kept decreasing continuously with the increase in temperature, that of the HEA kept increasing with the increase in temperature. The point of intersection of the thermal conductivities was about 1300K. In the temperature range of the cladding, the thermal conductivity of the HEA was higher than that of the substrate. This feature accelerated the outward transmission of excess heat from the molten pool, reduced the internal temperature gradient of the coating, relieved thermal stress concentration, and suppressed defects like cracks and deformation, thereby enhancing coating forming stability.

3.4 Influence of Process Parameters on Cladding Energy Consumption and Track Geometry

Figure 7 reveals the influence of three process parameters (laser power, powder feeding speed, and scanning speed) across four levels on three evaluation indicators: EUR, SEC, and track aspect ratio. The factor levels correspond to actual experimental values as follows: Laser power: Level 1 = 1600 W, Level 2 = 2400 W, Level 3 = 2800 W, Level 4 = 3200 W. Powder-feeding rate: Level 1 = 8 g/min, Level 2 = 12 g/min, Level 3 = 15 g/min, Level 4 = 18 g/min. Scanning speed: Level 1 = 4 mm/s, Level 2 = 6 mm/s, Level 3 = 7 mm/s, Level 4 = 8 mm/s. The symbol α in the figure represents the orbital cross-sectional ratio, which is defined as the ratio of the cladding width to the cladding height ($\alpha = W/H$).

In Figure 7(a), the EUR increased continuously with the increase of laser power and powder feeding speed. When the laser power rose from Level 1 to Level 4, the EUR increased from 10.65% to 13.28%. The powder feeding speed brought an increase in the utilization rate from 10.23% to 14.17%. Increasing the scanning speed only changed the utilization rate slightly from 12.13% to 12.32%. Higher power and powder volume allowed the laser energy to melt the powder more efficiently, whereas excessive scanning speed shortened the thermal interaction time and caused energy loss. As shown in Figure 7(b), the SEC decreased significantly with the increase of powder feeding speed, dropping from 31200 J/g at Level 1 to 15300 J/g at Level 4.

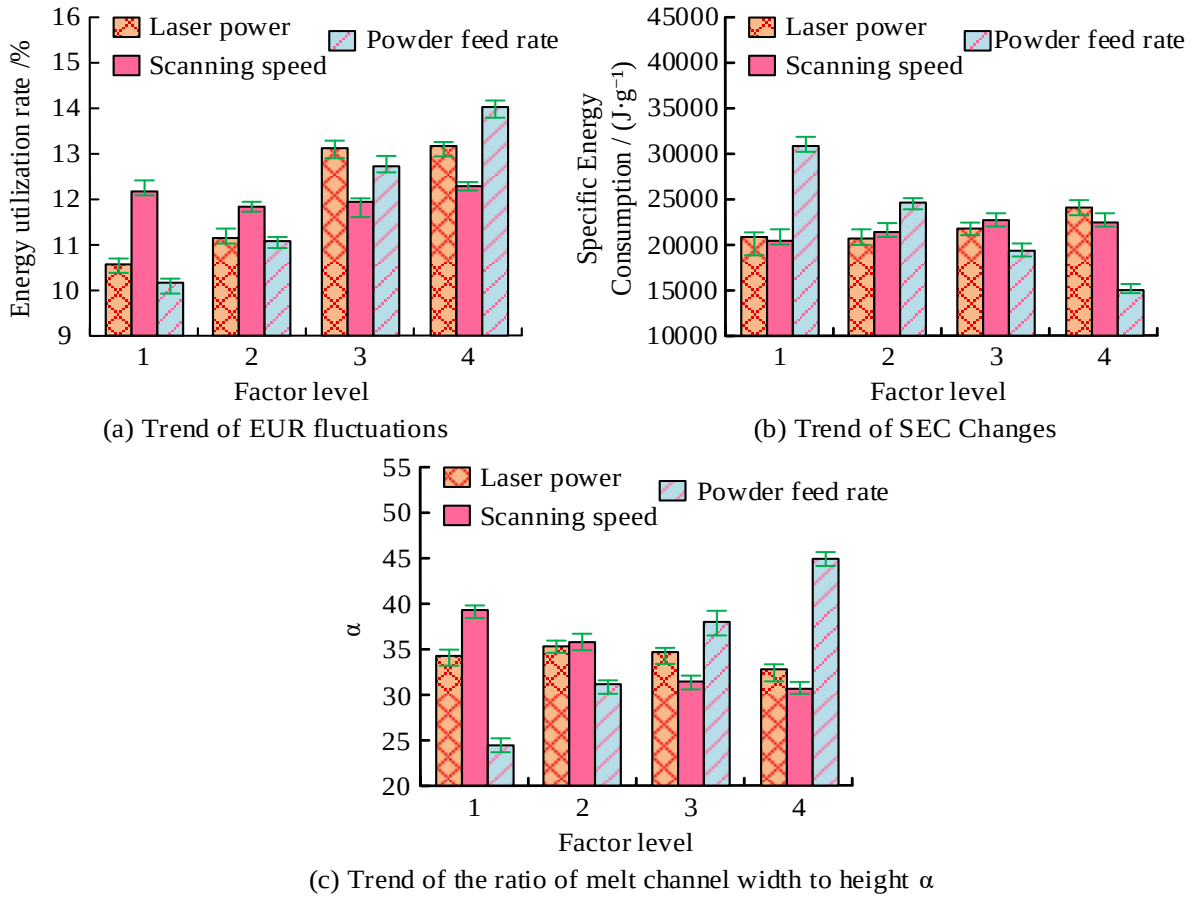


Fig. 7. Influence trends of process factors on cladding evaluation indicators

Increasing the laser power slightly raised the SEC, while the scanning speed caused minor fluctuations in energy consumption. The increase in powder delivery enhanced material utilization efficiency, which fully utilized the laser energy consumed per unit of powder. The results in Figure 7(c) demonstrate that the powder feeding rate had the greatest influence on the track aspect ratio. When the powder feeding rate was increased from Level 1 to Level 4, the aspect ratio increased from 24.83 to 44.98. The laser power increased the aspect ratio, whereas the scanning speed decreased the aspect ratio. Higher powder feeding increased the cladding accumulation height, whereas accelerating the scan promoted lateral spreading of the molten pool, which ultimately reduced the track aspect ratio.

3.5 Microhardness Distribution of Coatings Along the Layer Depth Under Different Conditions

At each depth interval, five independent Vickers indentations were performed, and the average hardness value was calculated ($n=5$). Small error bars in the bar charts indicate good repeatability of hardness tests and high process stability. Figure 8 presents the microhardness test results of the HEA cladding coatings under four sets of processes. The influence of different processes on the mechanical properties of the coatings was analyzed through the hardness distribution curves along the depth and the bar charts of average hardness.

As shown in Figure 8(a), within the coating region of 0 to 0.6 mm, the microhardness of the four specimens remained stable overall. The values followed the order of Condition 3 > Condition 2 > Condition 4 > Condition 1. The small hardness fluctuation inside the coatings indicated good microstructure uniformity across all groups. Upon entering the heat-affected zone of 0.6 to 1.0 mm, the hardness of all four specimens dropped noticeably, which was mainly because heating induced tempering softening of the substrate. When the depth exceeded 1.0 mm and entered the substrate region, the four curves converged gradually, and the hardness stabilized near 150 HV0.1, matching the hardness of the base metal.

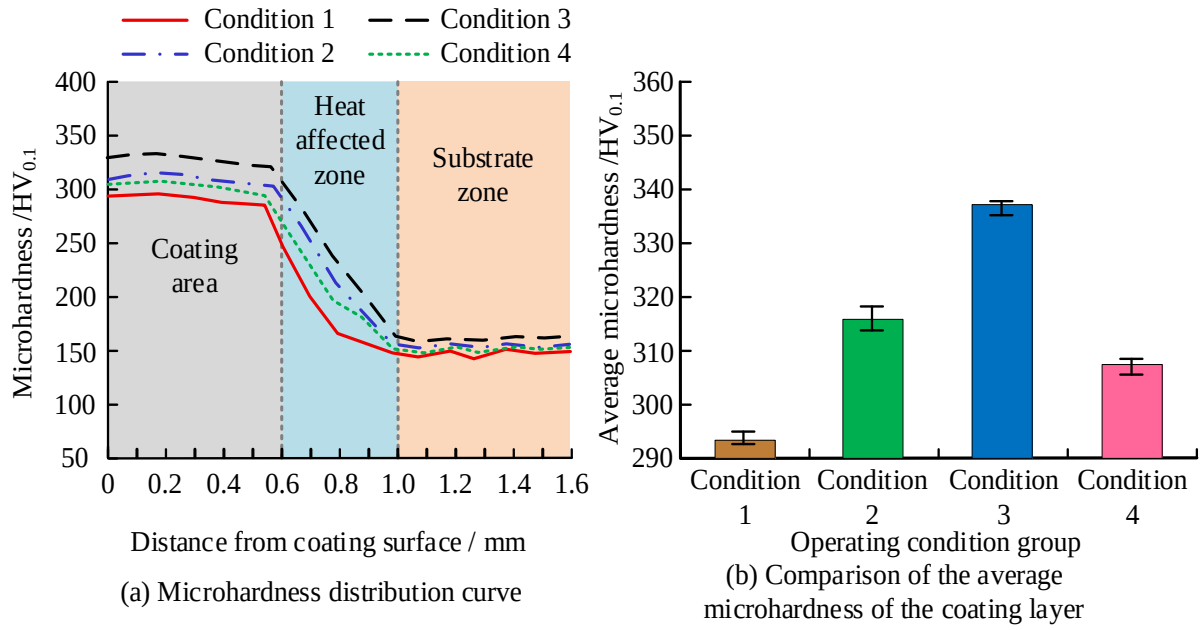


Fig. 8. Microhardness test results

Considering the average hardness statistics in Figure 8(b), Condition 3 yielded the highest average coating hardness of 339.62 HV_{0.1}. Here, fine-grain strengthening and solid-solution strengthening showed the most prominent effects. Condition 1 lacked sufficient heat input, creating unmelted defects inside the coating, so the hardness was the lowest. The small error bars in the bar charts indicated good repeatability of the hardness tests and high process stability. Comparative analysis demonstrated that the parameter combination of Condition 3 fabricated cladding coatings with optimal hardness performance.

3.6 Phase Evolution Analysis of As-Deposited and Annealed Coatings

Figure 9 shows the X-Ray Diffraction (XRD) phase patterns of the FeCoNiCrCu HEA coating in both the as-deposited and annealed heat-treated states. This analysis was utilized to compare the evolution of crystal structures and phase constitutions before and after heat treatment.

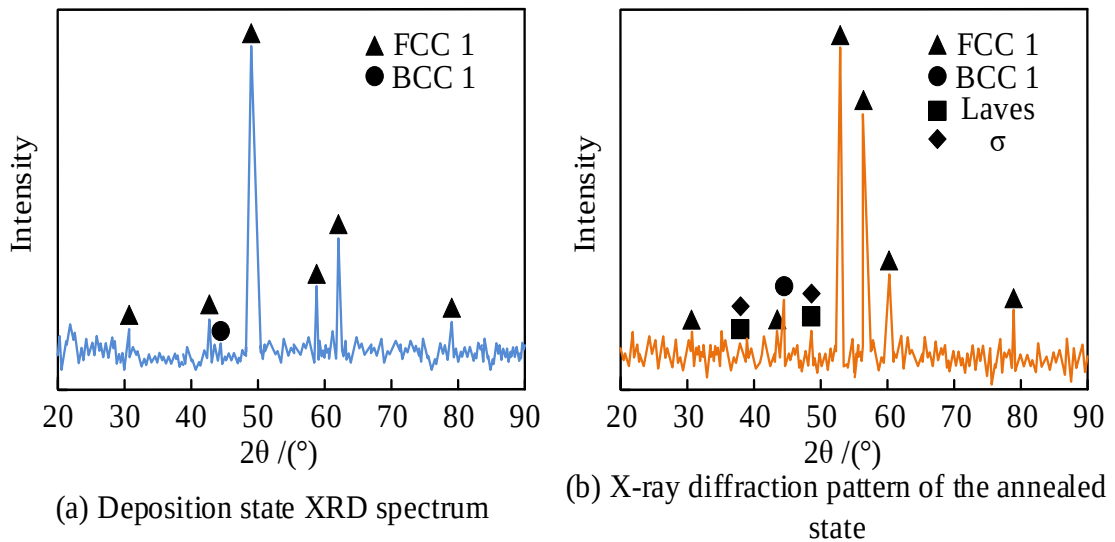


Fig. 9. XRD patterns of laser cladded FeCoNiCrCu HEA coating

The diffraction pattern of the as-coated layer in Figure 9(a) showed the presence of just two solid solutions, that is, FCC 1 and BCC 1. Diffraction peaks corresponding to intermetallic were absent. This was attributed to the very fast cooling process in the molten pool during laser cladding. Consequently, different alloying elements did not get sufficient time for long-range diffusion and segregation. Meanwhile, the broad diffraction peaks confirmed the characteristics of grain

refinement under rapid cooling conditions. At this stage, solid solution strengthening served as the primary mechanism to enhance the mechanical properties of the coating. Figure 9(b) indicates that after annealing heat treatment, the coating still retained the two matrix solid solution phases, FCC 1 and BCC 1. The main peak positions did not shift obviously, but the diffraction peaks became sharper, which suggested that the grains grew to some extent during high-temperature holding. Furthermore, a large number of new diffraction peaks appeared in the range of 35° to 48°, which were indexed as two types of intermetallic compounds: Laves phase and σ phase. The FCC solid-solution phase was identified according to ICDD-PDF #04-0850, the BCC solid-solution phase was matched with ICDD-PDF #06-0696. The Laves phase and σ phase were indexed with ICDD-PDF #15-0687 and ICDD-PDF #47-1423, respectively. Annealing provided sufficient energy for atomic diffusion. Elements segregated locally to precipitate secondary phases, which can further improve coating performance such as hardness and wear resistance through secondary phase strengthening. The Laves phase and the σ phase both belong to typical brittle intermetallic compounds. Although both can enhance the microhardness, their precipitation may increase the brittleness of the coating and the risk of crack initiation. This study only conducted a crack sensitivity assessment on the deposited-state coating and did not conduct crack behavior characterization on the annealed samples containing brittle precipitated phases. Therefore, before the practical application of this annealing process, further experimental evaluations are needed to assess the embrittlement risk brought by these intermetallic phases.

3.7 Quantitative Comparison of Cladding Forming Quality Under Multiple Conditions

To further quantify the comprehensive influences of different process parameters on cladding layer forming quality, interfacial bonding, and defect susceptibility, supplementary tests including molten pool wetting state, surface roughness, and dilution rate were performed. These tests quantitatively characterized the forming morphology variations and internal defect levels under the four conditions, and the results are listed in Table 3.

Table 3 Supplementary testing results of forming morphology and quality

Test indicators	Condition 1 (Under melting Condition)	Condition 2 (Baseline Condition)	Condition 3 (High- energy Overmelting Condition)	Condition 4 (High Powder Feeding Rate Condition)
Molten pool wetting angle (°)	128.67±3.24	76.32±1.47	58.75±2.08	109.43±2.69
Coating surface roughness (μm)	12.73±1.78	3.21±0.39	4.54±0.57	8.62±1.09
Dilution rate of the cladding layer (%)	8.34±0.87	21.52±1.18	37.26±2.47	12.63±0.98
Coating porosity (%)	4.23±0.68	0.32±0.09	0.84±0.17	2.13±0.47
Crack sensitivity coefficient	0.78±0.06	0.12±0.03	0.65±0.05	0.36±0.04
Interfacial shear strength (MPa)	218.37±11.86	426.24±17.93	352.18±14.87	287.42±13.79
Coefficient of variation of cladding track dimensions (%)	11.34±1.17	2.62±0.38	3.15±0.47	6.83±0.88
Width of the heat affected zone (μm)	126.35±7.89	245.27±11.92	487.46±17.83	189.32±9.78

Note: Condition 1: Laser power=1200 W, Scanning speed=8 mm/s, Powder-feeding rate=12 g/min; Condition 2: Laser power=1400 W, Scanning speed=10 mm/s, Powder-feeding rate=15 g/min; Condition 3: Laser power=1600 W, Scanning speed=10 mm/s, Powder-feeding rate=15 g/min; Condition 4: Laser power=1400 W, Scanning speed=10 mm/s, Powder-feeding rate=20 g/min.

The data in Table 3 indicate that Condition 1 lacked sufficient heat input, yielding a molten pool wetting angle of 128.67°. Incomplete powder melting led to a porosity of 4.23%, and the crack sensitivity coefficient was relatively high at 0.78, while the interfacial shear strength reached only 218.37 MPa. This result meant that forming defects were numerous and bonding strength was poor. Condition 2 provided the best matching baseline parameters, where the wetting angle decreased to 76.32°, the surface roughness was only 3.21 μm , and the porosity dropped to 0.32%. The crack tendency was weak, and the interfacial shear strength reached 426.24 MPa. The variation coefficient of track size was only 2.62%, indicating uniform and stable forming. In condition 3, the amount of heat input was too high, resulting in a very high dilution rate of 37.26%, and the width of the heat affected zone increased to 487.46 μm . This resulted in excessive thermal softening of the substrate, and the susceptibility to cracking increased again. Condition 4 involved an excessive powder feeding rate, which degraded molten pool spreading, causing the wetting angle to rebound to 109.43°. Its porosity and dimensional fluctuations both exceeded those of the baseline group, showing reduced forming stability. Considering all quantitative data, the baseline condition effectively balanced molten pool wetting, substrate dilution, and internal defects, making it the optimal process window that satisfied both forming appearance and interfacial mechanical properties.

4. Conclusions and Perspectives

Based on the experimental characterization results of laser-cladded FeCoNiCrCu high-entropy alloy coatings under four representative process conditions, the main findings can be summarized as follows. Under the under melting condition (Condition 1), insufficient laser heat input limited elemental diffusion. The cladding width and depth remained relatively small. Low energy input suppressed metallurgical bonding at the coating-substrate interface. Excessively high laser energy adopted in Condition 3 drove the 45-steel substrate dilution rate to 37.26% and substantially widened the heat-affected zone. According to laser thermal-cycling characteristics, such high heat input is likely to trigger undesirable substrate grain coarsening, which needs further microstructural verification in future studies. While the high-energy over melting condition (Condition 3) yielded the highest average microhardness of 339.62 HV0.1, the baseline process (Condition 2) represented the most balanced and optimal process window, effectively optimizing molten pool wetting, minimizing substrate dilution, and maximizing interfacial shear strength. Combined with the XRD results, FCC and BCC solid-solution phases formed in the as-deposited coating. Additional Laves phase and σ -phase precipitates were generated after annealing treatment.

Combining all quantitative indicators, the baseline process effectively balanced the melting state, 45-steel substrate dilution, and interfacial bonding. Notably, obvious Cu-rich segregation can be observed within the cladding coatings from EDS elemental mapping results, which is consistent with the metallurgical characteristics of FeCoNiCrCu high-entropy alloy systems. Although post-annealing treatment promotes the precipitation of hard Laves and σ -phases to improve coating hardness, such brittle intermetallic precipitates may also bring potential embrittlement risks to the coating. However, the present study is limited to a narrow window of laser processing parameters. Furthermore, in-service performance characteristics—such as wear and corrosion resistance—and multi-physics coupled numerical simulations were not evaluated. Further research will focus on expanding the experimental design matrix to investigate multi-track overlapping effects, conducting high-temperature wear and electrochemical corrosion testing to assess environmental durability, and refining the finite element models to accurately predict residual stresses and phase precipitation during post-cladding annealing.

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