

Article

Microstructure and Mechanical Properties of Nickel-Based Coatings Fabricated through Laser Additive Manufacturing

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Abstract: In this study, single-layer and three-layer nickel-based coatings were fabricated on 316L SS by laser additive manufacturing. The phase characterization, microstructure observation, and microhardness analysis of the coatings were carried out by X-ray diffraction (XRD), scanning electron microscope (SEM), and microhardness tester. And the wear resistance of the coatings was analyzed through dry sliding friction and wear test. The results show that the cross-section microstructure of the three-layer nickel-based coating is different from that of the single-layer one under the influence of heat accumulation; the dendrite structure in the central region of the former is equiaxial dendrite, while that of the latter still remains large columnar dendrites. The existence of solid solution phase γ -(Fe, Ni) and hard phases of $\text{Ni}_{17}\text{Si}_3$, Cr_5B_3 , Ni_3B in the coating significantly improve the wear resistance of the coating, and the microhardness is nearly 2.5 times higher than that of the substrate. However, the average microhardness of multilayer cladding coating is about 48 $\text{HV}_{0.2}$ higher than that of the single-layer cladding coating. In addition, the fine surface structure of the three-layer nickel-based coating improves the wear resistance of the coating, making this coating with the best wear resistance.



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Keywords: laser additive manufacturing; 316L ss; nickel alloy; microstructure; tribological behavior

1. Introduction

Due to its excellent corrosion resistance and mechanical properties, 316L stainless steel is widely used in the chemical industry and aerospace field [1–3]. However, the low hardness and the poor wear resistance limit its applications to a certain extent. Therefore, surface modification is needed to further improve the properties of the material [4,5], including atmospheric plasma spraying (APS) [6], chemical vapor deposition (CVD) [7] and laser additive manufacturing (LAM) [8]. As a new manufacturing technology integrating laser, digitization, materials science, and other disciplines, LAM has obtained widespread attention in recent years since it can realize dimension reduction manufacturing, complex forming, and high material utilization [9–11]. According to material feed-in methods, LAM can be divided into powder spreading type selective laser melting and powder feeding type laser melting deposition [12,13]. Laser melting deposition technology is also called laser cladding (LC), which uses a high-energy laser beam to metallurgically bond the cladding material to the substrate surface, with a small heat-affected zone (HAZ), fast cooling speed, and many other advantages [14,15]. In addition, LC can inject powder with low dilution and specific quality into the substrate to improve material performance and repair material surface defects, which has been the focus of material researchers in recent years [16–18].

Ti-, Fe-, Co- and Ni-based alloy powders are widely used in laser additive manufacturing due to their excellent high-temperature resistance, high hardness, and excellent wear

resistance. Ertugrul et al. [5] studied the structure and hardness behavior of 15 vol% TiC reinforced 316L coating and found that the increase in material hardness was due to the refinement of TiC grains and the appearance of new carbides during solidification. Cheng et al. [19] found that the mechanical properties and wear resistance of the FeBSiNb coating prepared by broad-beam laser cladding (BLC) were significantly improved, because the Nb-B particles generated in situ were uniformly dispersed in each part of the coating. Yan et al. [14] applied laser cladding technology to prepare TaC/Stellite X-40 Co-based composite coating on a nickel aluminum bronze (NAB) substrate. The results showed that intermetallic reinforcement materials such as TaC, Cr_3C_2 , and Co_3Ta were distributed in the substrate uniformly, improving the wear resistance and the electrochemical corrosion performance of the alloy. When the TaC content is 20 wt%, the material reached the best wear resistance and corrosion resistance.

Among the cemented carbide powders mentioned above, Ni-based alloy powders have received extensive attention due to their excellent surface properties. Liu et al. [20] deposited a nickel-based WC composite powder on the surface of the tunnel boring machine (TBM) tool ring by LC and found that there was a good metallurgical bond between the coating and the substrate. The widely distributed WC, W_2C , and other particles in the coating effectively improved the wear resistance of the surface of the TBM tool ring. Liu et al. [21] adopted a multilayer laser cladding method to prepare a nickel-based alloy coating on the surface of graphite cast iron. The results showed that with the increase in the number of cladding layers, the volume fraction of dendrites and the distance between secondary dendrite arms increased significantly. When 6 layers of the coating were deposited, the material received the best corrosion resistance at room temperature. SouSa et al. [22] tested the tribological properties of the Ni-Cr-Bo-Si coating prepared by the laser cladding method, and the results showed that the coating had good surface adhesion. With the increase of the CrC, the microhardness of the coating was increased by 10%.

From the above research, it can be found that the cladding of nickel-based alloy powder on the substrate surface can effectively improve the wear resistance of the material. However, most studies have focused on the influence of different nickel-based alloy powder compositions on coating properties, and few studies on process parameters. During multilayer cladding, the existence of heat accumulation is likely to affect the structure and performance of the coating. Therefore, in this paper, the single-layer nickel-based coating and the three-layer nickel-based coating were prepared. Physical characterization, microstructure observation, and microhardness analysis of the coatings were performed using XRD, SEM, and microhardness tester. The wear resistance and wear mechanism of the coating were analyzed by a dry sliding friction wear test.

2. Experimental Procedure

Commercially available Ni-based alloy (Ni-Cr-B-Si-Fe-C) powder was used as the cladding material. The SEM image of the powder is shown in Figure 1. It can be seen that its shape is almost equiaxed spherical the size is approximately 50–150 μm , and the d_{50} of powder is about 100 μm . 316L stainless steel cut into 80 mm $\times$ 50 mm $\times$ 10 mm (thickness) was applied as the substrate. Before cladding, sandblasting technique (Al_2O_3 ceramic particles) was used over the substrate to have a surface roughness (R_a) of 10 μm . Then, the samples were rinsed with alcohol and dried to keep the surface clean. The chemical compositions of 316L stainless steel and nickel-based alloy powder were tested by Tianjin Zhujin Technology Development Co., Ltd. (Tianjin, China) and shown in Table 1.

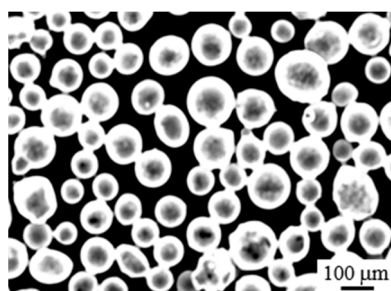


Figure 1. SEM image of Ni-based alloy powders.

Table 1. Chemical composition of powder and substrate metal.

Materials	Elements Composition (wt%)									
	C	P	Cr	S	Mn	Mo	Ni	Si	Fe	Co
316L steel	0.023	0.034	16.4	0.57	1.37	2.16	10.03	0.69	Bal.	-
Ni-based alloy	0.03	-	6	-	-	-	Bal.	1.5	0.38	3

RFL-C3300 (Raycus, Wuhan, China) high-power continuous fiber laser device was used in the laser cladding test. DPSF-2 powder feeder was used to feed the powder synchronously in an oblique direction, and argon was used as the power source to accurately send powder to the laser spot. During the cladding process, argon was used to protect the molten pool to avoid oxidation. In this experiment, single-layer (L1) and three-layer (L3) coatings were deposited respectively, as shown in Figure 2. The detailed parameters of laser cladding are listed in Table 2.

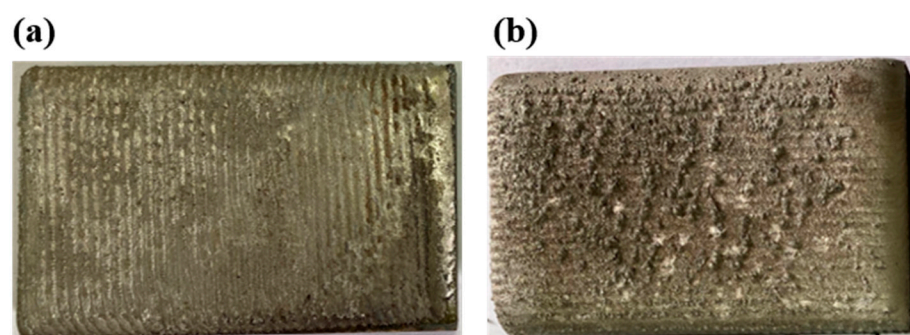


Figure 2. Macroscopic appearance of Ni-based laser-clad coating: (a) L1, (b) L3.

Table 2. Different parameters of laser cladding.

Laser Power (kW)	Powder Feeding Rate (g/min)	Scanning Velocity (mm/s)	Spot Diameter (mm)	Gas Flow Rate (L/min)	Overlap Rate
1.8	30	5	4	15	50%

When the laser cladding parameters are unchanged, the dilution rate of single-layer single-pass laser cladding is the same as single-layer multi-pass laser cladding and multi-layer multi-pass laser cladding. According to the laser cladding parameters in Table 2 (due to the single-layer single-pass cladding, the overlap rate parameter is not required here), the geometric dimensions of the single-layer single-pass laser cladding sample are shown in Figure 3. According to the literature, the dilution rate (λ) can be calculated by the area method [23], and the formula is simplified to:

$$\lambda = \frac{h}{H + h} \times 100\% \quad (1)$$

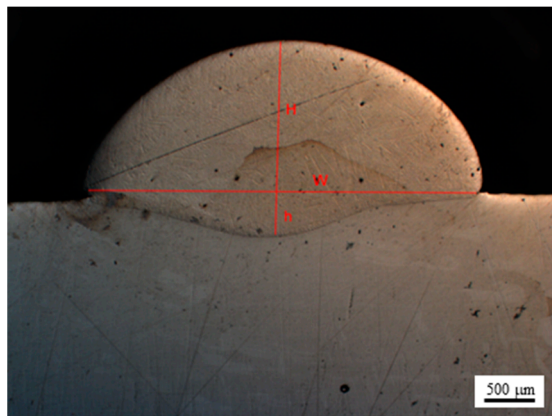


Figure 3. Feature size of single layer single pass laser cladding layer.

In the formula, h is the penetration depth of the base material, and H is the height of the cladding layer. According to the geometric dimensions shown in Figure 2, the dilution rate is 24.5%, indicating that the cladding layer and the substrate form a good bonding interface.

After cladding, the samples were cut, cleaned, and degreased with acetone alcohol, and then a standard metallographic sample preparation process (grinding, polishing, washing, and drying) was carried out to obtain mirror-surface metallographic and tribological samples. The phase identification of the cladding layer was performed by D8 Advance Multifunctional Powder Diffractometer (Bruker Corporation, Karlsruhe, Germany) with a Cu- $k\alpha$ radiation at a voltage of 40 kV. The micromorphology of the coating was observed by JSM-6480 tungsten filament scanning electron microscope (JEOL Corporation, Tokyo, Japan), and the element distribution in the micro area was measured by its equipped energy dispersive spectrometer (EDS). According to ASTM E92-2016 standard, KB-30S-FA Automatic Vickers Hardness Tester (KB Corporation, Assenheim, Germany) was used to test the hardness of the coating with the applied pressure of 0.2 kg and the holding time of 15 s. According to the ASTM G99-05 standard, the UMT-2 high temperature friction and wear tester (CETR Corporation, CA, USA) was used to study the dry sliding ball-on-plate wear of the nickel-based coating surface. For the dry sliding friction test, the sliding distance is 18 m, the relative speed between coating and counter-body is 10 mm/s, and the diameter of the counter-body ball is 10 mm. The surface roughness before wear test and the three-dimensional morphology of wear marks were observed by LEXTOLS4000 confocal laser scanning microscope (Leica Corporation, Wetzlar, Germany). Before the wear test, the surface roughness (S_q) of the substrate, L1, and L3 were 0.012, 0.044, and 0.046 μm , respectively. Finally, the micro wear morphology was observed through SEM.

3. Results and Discussion

3.1. Microstructure and Composition

XRD patterns of coatings are shown in Figure 4. According to Table 1, the content of Fe in nickel-based alloy powder is very low, but face centered cubic structural γ -(Fe, Ni) phase has been detected in XRD. This is because the molten pool is formed on the substrate surface under the irradiation of laser beam, and interdiffusion occurs between the substrate and the nickel-based coating [24], making a large amount of Fe enter the coating. The radius of Fe atoms is very close to that of Ni atoms, and the rapid rise and decrease of temperature are similar to solid solution treatment. Thus γ -(Fe, Ni) solid solution appears on the coating surface of the single-layer cladding sample. However, during multilayer cladding, Fe is difficult to diffuse to the coating surface. Therefore, although the Fe-containing main dominant peaks are detected at 44° for both L1 and L3 samples, the phase content in L3 has changed, which can also be proved from the EDS results in Table 3. In addition, hard phases Cr_5B_3 , $\text{Ni}_{17}\text{Si}_3$, and Ni_3B have also been detected on the coating surface, which is proved to be helpful to improve the hardness and wear resistance of the coating [25–29].

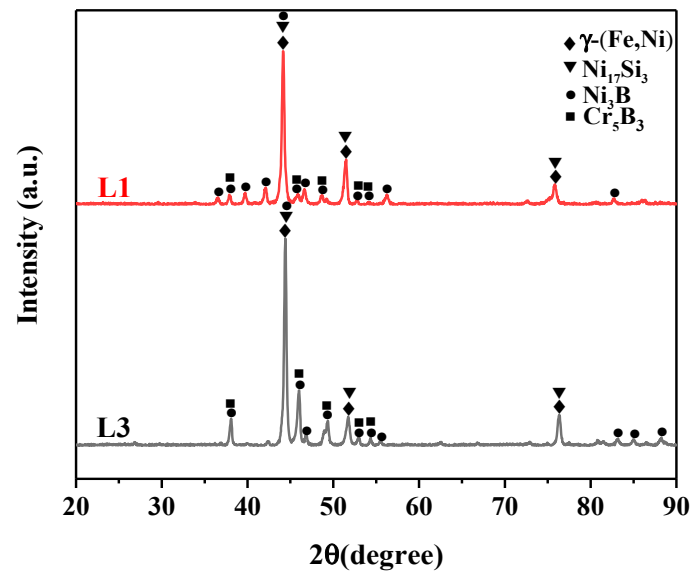


Figure 4. XRD patterns of the coatings.

Table 3. Chemical composition of phases in Figures 5 and 6.

Region	Element Composition (at.%)			
	Si	Cr	Fe	Ni
A	6.16	3.19	15.46	75.09
B ₁	6.77	3.08	13.75	76.40
B ₂	5.47	2.79	11.04	80.70
D ₁	15.77	-	-	84.23
D ₂	7.04	-	-	92.96
D ₃	5.59	-	0.87	93.54

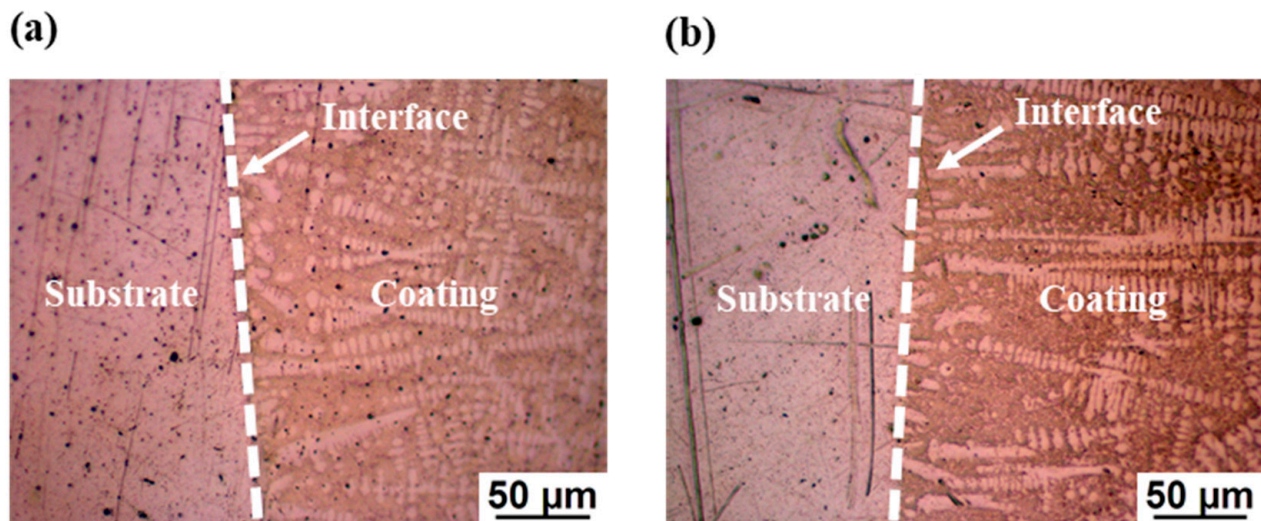


Figure 5. Cross-section OM images of (a) L1 and (b) L3.

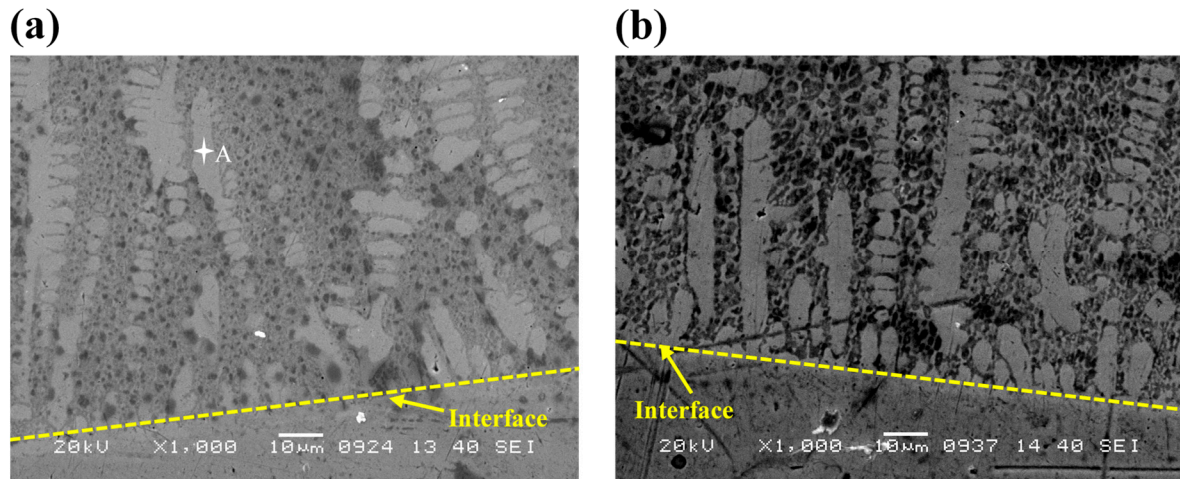


Figure 6. Cross-section morphology (SEM) near the coating-substrate interface (a) L1, (b) L3.

Figure 5 displays the metallographic diagram of the cross section of the nickel-based coating. It can be seen that the coating is well bonded with the substrate, and there are no obvious cracks. A lot of gas pores are observed in the coating of sample L1 (Figure 5a). During laser cladding, the temperature rises and decreases so rapidly that the gases have little time to escape, leaving them to remain in the coating and form pores. Moreover, these pores are randomly distributed and have nothing to do with the trajectory of the boundary, and the existence of pores will not affect the bonding quality between the substrate and the coating [30]. However, not many gas pores can be seen in the coating of sample L3 (Figure 5b), which is because the multilayer cladding will reheat the cladding area, greatly reducing the pores.

The SEM images of the cross section near the interface between coating and substrate are shown in Figure 6. According to Kou's study [31], the ratio of G to R can be used to describe the change of microstructure, where G is the temperature gradient and R is the growth rate. It is well known that the ratio of G to R is high due to the good heat dissipation of the substrate and the short solidification time of the solution near the interface during the solidification of the coating. Thus, some cellular crystals appear along the boundary line. As shown in Figure 6a,b, with the decrease of the temperature gradient, an obvious columnar dendrite structure can be observed near the boundary line, and the columnar dendrite grows along the temperature gradient direction. According to the EDS data in Table 3, the main component of point a in Figure 6a is Ni, with a small amount of Si, Cr, and Fe, indicating that the composition of dendrite is γ -Ni phase and the binary eutectic γ -Ni formed by γ -Ni phase and other metal elements (Si, Cr, and Fe). Due to the high cooling rate, a dense eutectic phase was both observed between the γ -Ni dendrites in Figure 6a,b, which shows a network structure in sample L1 while a honeycomb texture in sample L3 due to multiple heating. Many dark regions can be seen between the eutectic phases. Paul believed that the main component of this region is borides [32], and the XRD results also verify the existence of borides. Therefore, the dendrite structure of eutectic is boride and binary eutectic γ -Ni. The microstructure near the coating-substrate interface is a uniform combination of columnar dendrites and interdendritic crystals.

The SEM cross-section morphologies of the middle and surface areas of the coatings are shown in Figure 7. Due to the change of composition and the decrease of temperature gradient, the growth of columnar dendrites is limited. Thus, columnar dendrites are not observed in the middle of the coating, while uniformly arranged equiaxed dendrites appear instead. Meanwhile, the network eutectic structure between equiaxed dendrites becomes uniform due to the decrease of temperature gradient. Figure 7b is the SEM picture of the single-layer cladding coating surface and both columnar dendrites and steering dendrites can be observed. Due to the change of temperature gradient caused by the co-effect of the contact between the coating surface and air and the movement of laser beam [33], not only

the columnar dendrites growth along the temperature gradient appear, but also the steering dendrites grow along the scanning direction on the coating surface occur. When the coating surface contacts the air, the cooling rate will be faster, and the grain growth will be limited. Thus it can be seen that the grain size in Figure 7b is significantly smaller than that in Figure 7a. In addition, more Fe elements have been detected in B₁ and B₂ in Figure 7b, indicating that single-layer cladding is more conducive to the diffusion of Fe elements to the coating surface so as to form γ -(Fe, Ni) lamellar eutectic with Ni [4]. Figure 7c is the SEM picture of the middle area of the multilayer cladding. When the second cladding layer solidifies, although the first cladding layer has solidified, it still has a certain amount of heat, resulting in heat accumulation. Due to the accumulation of heat, the value of G/R is not high enough to form planar crystal and cellular crystal, but columnar dendrite appears directly in Figure 7c. At the same time, the effect of heat accumulation increases the reheating area, and the solidification rate of grains decreases, resulting in larger columnar dendrite size [34]. On the other hand, heat accumulation reduces the effect of temperature gradient on the grain growth direction, making the columnar dendrites present different directions. From Figure 7d, it's hard to find obvious dendrites. This is because when it is near the top of the molten pool, the solidification rate of the molten pool increases rapidly due to the contact with air, so that the dendrites cannot grow sufficiently. Moreover, a large number of fine particles appear in Figure 7d and Mohamad et al. [35] believed that the particles not completely melted on the surface of the cladding layer would play the role of heterogeneous nucleation points. As the EDS data in Table 3 show, the main components of the points D₁, D₂, and D₃ are Ni₁₇Si₃, indicating that the large number of Ni₁₇Si₃ distributed on the coating surface during coating solidification help with the formation of a finer surface structure of the coating.

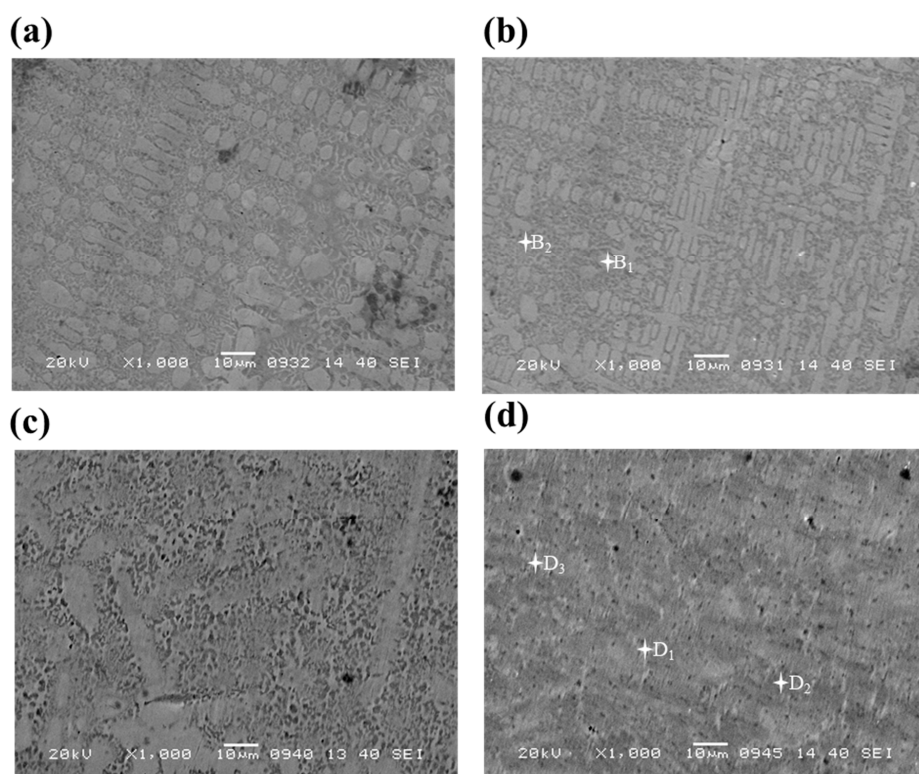


Figure 7. Cross-section SEM morphologies of the middle and surface of the coatings: (a) Middle region of L1, (b) Surface region of L1, (c) Middle region of L3, (d) Surface region of L3.

3.2. Wear Analysis

Figure 8 shows the cross-sectional microhardness distribution of different samples. It can be seen that the microhardness of both coatings is not less than 600 HV_{0.2}, which

is nearly 2.5 times higher than that of the 316L SS substrate (about 250 HV_{0.2}). Due to the bonding of the coating and the substrate, the substrate is inevitably diluted [36]. Comparing Figure 8b to Figure 8a, the microhardness of the substrate close to the HAZ decreases slightly because of Fe dilution [4]. The HAZ of laser cladding is very small, Landowski [37] found that the HAZ of laser welded 316L SS joint is about 20 µm. Therefore, the presence of HAZ has a very limited effect on the microhardness of the substrate. During cladding, the microhardness distribution of the cross-section changes owing to the change of temperature. For the multilayer cladded L3 sample, the temperature changes much greater than the single-layer cladded L1 sample during laser cladding, resulting in a greater fluctuation of cross-sectional microhardness than L1. The significant improvement of microhardness of the coating compared with 316L SS substrate can be attributed to the addition of alloying elements and the γ -(Fe, Ni) solid solution phase and hard phases Ni₁₇Si₃, Cr₅B₃, and Ni₃B formed during the cladding process. In addition, the rapid rise and fall of temperature during laser cladding will produce residual stress in the coating, which also helps improve the microhardness of the coating [38]. The average microhardness of L1 coating is 593 (± 6) HV_{0.2}, while that of L3 coating is 640 (± 20). According to the analysis above, the improvement of microhardness is related to the fine surface structure of the coating.

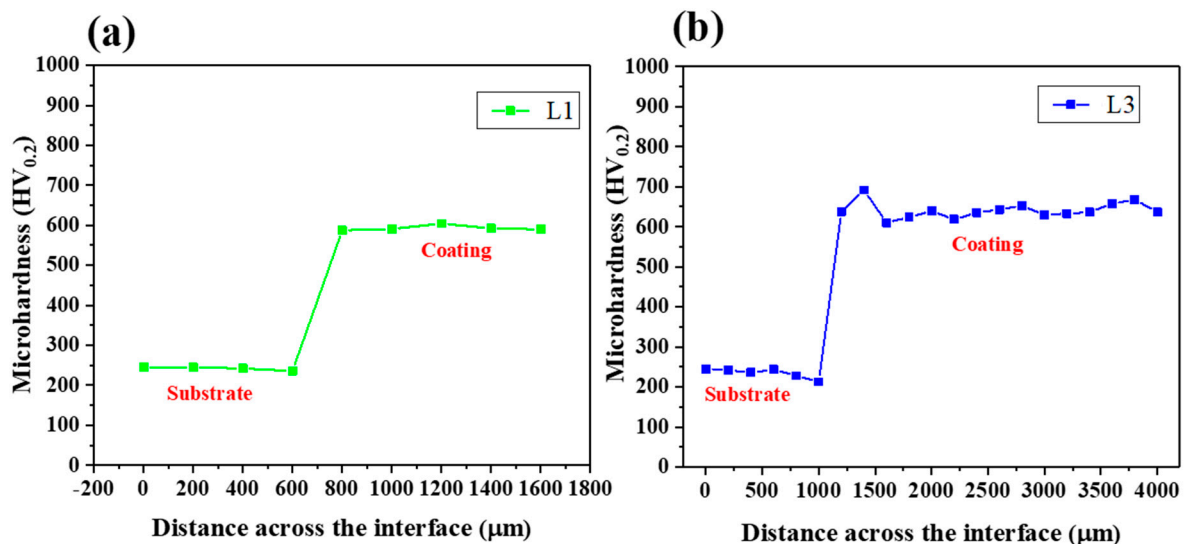


Figure 8. Microhardness profile of laser cladded cross section: (a) L1, (b) L3.

The dry sliding friction and wear test with a sliding friction load of 10 N and a wear time of 1800 s was conducted to evaluate the friction performance of the substrate and the coatings. The obtained friction coefficient curve of the substrate and the coatings and the relevant histogram of the average friction coefficient calculated from the friction coefficient curve are shown in Figure 9a,b, respectively. As shown in Figure 9a, all samples reach a high limited value of the friction coefficient before it stabilizes. This is because in the early stage of sliding friction, the material is continuously sheared, and the deformation layer is continuously accumulated. And the accumulated deformation layer is continuously washed to ensure the smooth progress of the experiment, so the friction coefficient is increased [39]. In addition, since the hardness of the substrate is relatively lower, the deformation layer accumulated in the sliding friction experiment is easier to be washed, so the friction coefficient of the substrate tends to be stable earlier than that of the coating. As the sliding friction experiment continues, the friction coefficient of the substrate shows a slight decrease. Kumar believes that this is because the transfer of materials during the sliding friction increases the carbon concentration on the surface of 316L SS, thus promoting the formation of the friction film [40]. It can be concluded from Figure 9b that the average friction coefficient of different samples meets the substrate (0.53 ± 0.02).

> sample L1 (0.47 ± 0.05) > sample L3 (0.44 ± 0.03), indicating that nickel-based alloy coating can effectively improve the wear resistance of 316L SS, and the improving effect of multilayer cladding is better than that of single-layer cladding.

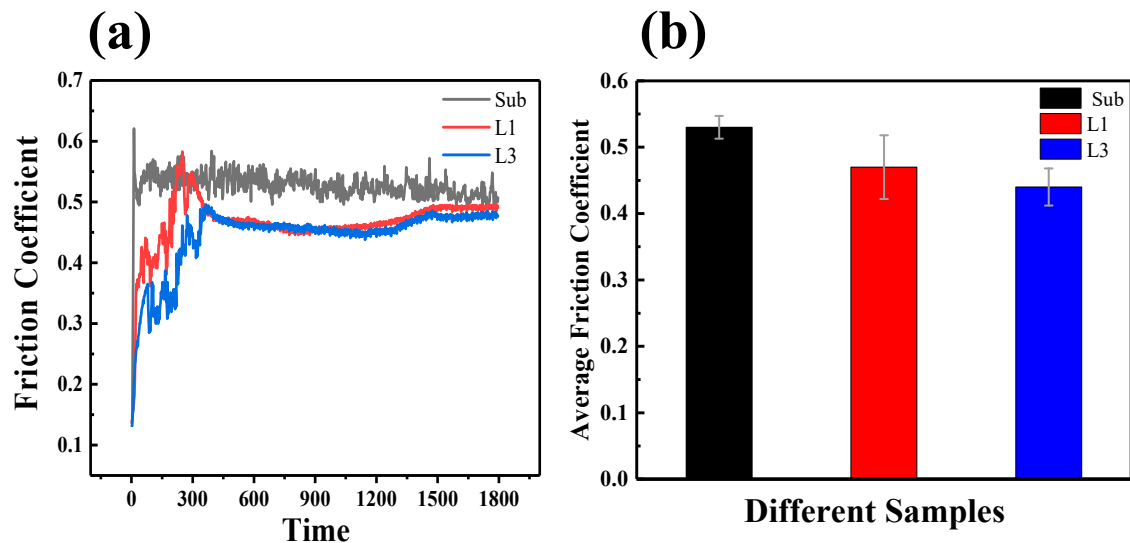


Figure 9. (a) Friction coefficient curves of the substrate (Sub), L1 and L3, (b) Histogram of average friction coefficient.

Figure 10 is the wear scar surface and the corresponding wear scar profile data obtained from the dry sliding friction and wear test. It is obvious that the wear resistance of 316L SS can be effectively improved by Ni-based alloy coating whether by single-layer cladding or multilayer cladding. Comparing Figure 10a–c, it can be found that the wear profile of the substrate surface shows more deflection. The difference in deflection depends on the difference in the contact area. Under the same load condition, the microhardness of the substrate surface is lower, which makes the contact area between the Al_2O_3 ceramic ball and the material surface larger. The higher the contact area, the higher the length of wear track and peak/valley heights, which in turn increases the wear rate of the material. At the same time, due to the lower microhardness of the substrate, greater deformation occurs during sliding friction, and obvious protrusions appear on both sides of the wear scar. In Figure 10b,c, samples L1 and L3 have similar 3D morphologies and the profiles of wear marks, indicating that the wear extent of L1 and L3 in dry sliding friction and wear test is similar. In addition, the relatively flat wear surface of samples L1 and L3 demonstrating that the higher microhardness of the coating surface makes the load applied to the coating be effectively distributed to the whole wear surface.

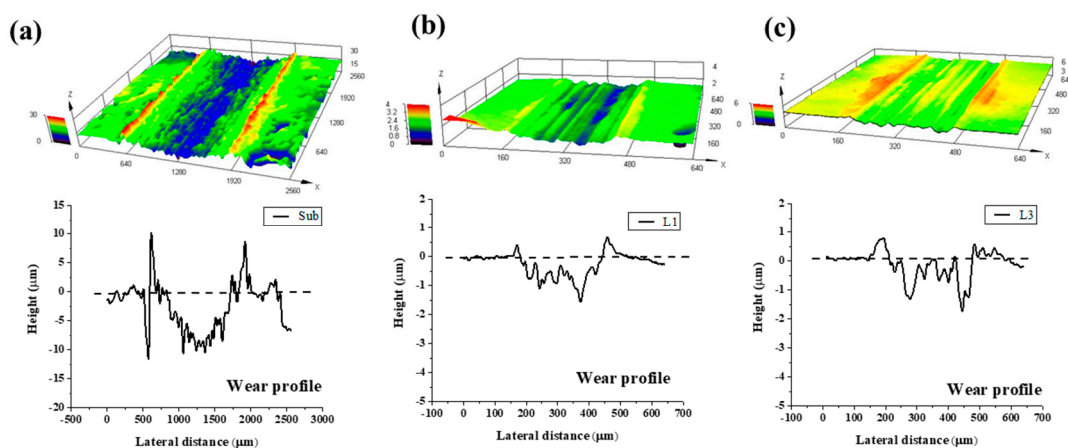


Figure 10. Surface wear of the substrate and the coating under the dry sliding friction and wear test conditions with a sliding friction load of 10N and a wear time of 1800 s: (a) Sub, (b) L1, (c) L3.

Figure 11 shows the SEM image of the morphologies of the substrate and the coatings after the dry sliding friction and wear test. Figure 11a,b are the wear SEM morphologies of the substrate. It can be seen that the wear surface is rough and uneven, and the particles peel off obviously, showing typical abrasive wear characteristics. As shown in Figure 11a, the substrate surface shows a high wear extent with some pores generated during sample preparation exposed. The surface in Figure 11b is even seriously worn, with large wear debris particles attached to it as well as many deep grooves in the area with severe particle spalling. Under a load of 10 N, particles frequently fall off from the surface, causing delamination on the wear surface, so as to greatly increase the wear rate. As seen in Figure 11c,d, the wear surfaces of the coatings are in good condition with few particles peeling off or delamination, and the coatings show excellent wear resistance. After an 1800 s sliding friction, some peeling-off appears on the surface of sample L1. At the same time, a small number of black patches appeared in Figure 11c, indicating that the wear mechanism of the L1 sample is a combination of abrasive wear and adhesive wear. In the figure, there are even obvious micro-cracks in the area where the spalling phenomenon is more serious. If the load or wear time continues to increase, the micro-cracks will expand and delamination will occur [41]. A large area of black patches can be seen in Figure 11d. During sliding friction, the particles will be transferred from the contact material. Because of the high hardness of the coating, the particles eventually appear in the form of patches, resulting in adhesive wear. Moreover, a large number of small pits in the middle of the patches indicate slight abrasive wear. Sometimes, the particles adhering to the contact surface will fill the gap (as shown in Figure 11d) during sliding, thus reducing the wear extent of the coating, but it won't affect the wear mechanism [42]. In the process of sliding friction, the positive pressure between the load and the tangential motion of Al_2O_3 ceramic ball causes plastic deformation of the material, resulting in shallow grooves along the sliding direction. In addition, the fine surface structure of the L3 coating can effectively improve the wear resistance of the coating, and there is no obvious particle spalling trace except slight wear marks. In general, no matter by single-layer cladding or by multi-layer cladding, nickel-based coatings change the wear mechanism of materials in the process of friction and wear, effectively reduce the abrasive wear extent and improve the wear resistance of materials; the three-layer nickel-based coating displays the best wear resistance.

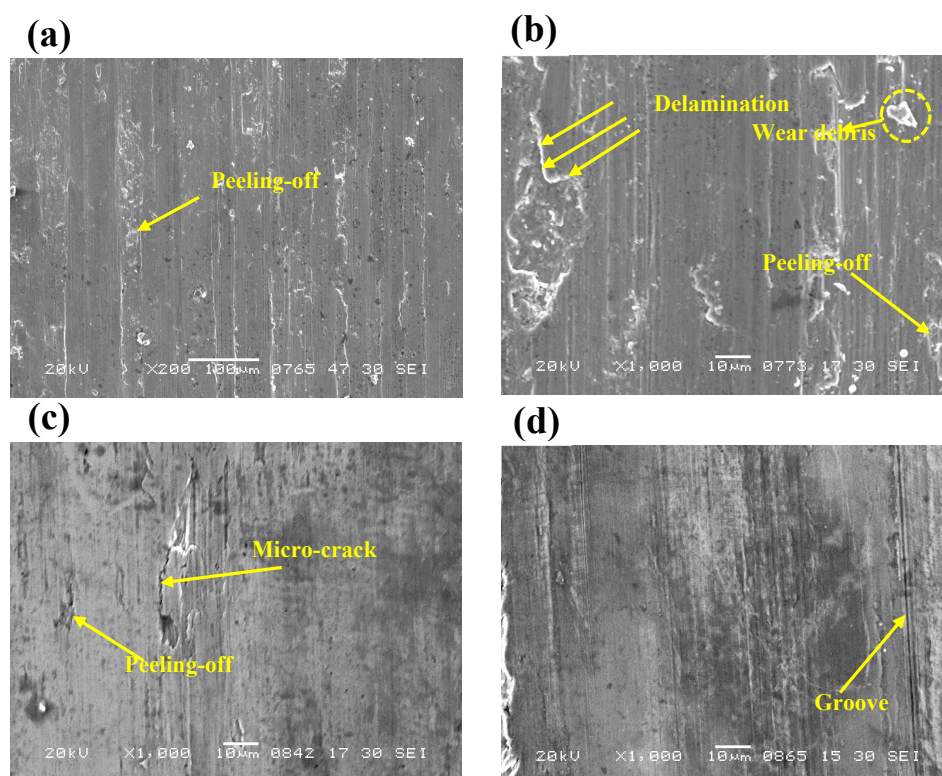


Figure 11. SEM morphologies of wear surfaces for (a,b) the substrate, (c) L1 and (d) L3.

4. Conclusions

The single-layer and three-layer nickel-based coatings have been fabricated on 316L SS substrate by laser additive manufacturing and are metallurgically bonded well with the substrate without obvious defects. The microstructure, surface precipitates and wear resistance of the coatings have been systematically studied and the following conclusions have been drawn:

- (1) The XRD patterns of single-layer cladding and multilayer cladding coatings are similar, both including γ -(Fe, Ni) solid solution phase, Cr_5B_3 , $\text{Ni}_{17}\text{Si}_3$, and Ni_3B phases. Due to the diffusion of Fe, the content of γ -(Fe, Ni) solid solution in the surface of single-layer coating is higher than that of the multilayer coating.
- (2) Cellular crystals are observed in the coating-substrate interface of both single-layer cladding and multilayer cladding samples. With the change of G/R ratio, equiaxed dendrites appear in the middle of the single-layer coating, and columnar dendrites and steering crystals appear on the surface. However, due to heat accumulation, columnar dendrites with large size and different directions appear in the middle region of the multilayer clad Ni-based coating. No obvious dendrite morphologies are observed in the surface area of the multilayer clad coating, and the massively distributed $\text{Ni}_{17}\text{Si}_3$ phases help to obtain a finer surface structure.
- (3) The surface microhardness of 316L SS has been significantly improved by Ni-based coating by about 2.5 times. The average microhardness of multilayer cladding coating is about 48 $\text{HV}_{0.2}$ higher than that of the single-layer cladding coating, and the average friction coefficient of three-layer coating is the lowest (0.44 ± 0.03) among the substrate and the single-layer coating. In addition, the fine surface structure of the three-layer nickel-based coating improves the wear resistance of the coating, making this coating with the best wear resistance.
- (4) The Ni-based coating shows excellent wear resistance. After dry sliding friction and wear test, the substrate has been severely worn down, and the wear mechanism is mainly abrasive wear. Some peeling-off appear on the surface of single-layer nickel-based coating, and the main wear mechanism is a combination of abrasive wear and adhesive wear, while the three-layer coating is in good condition without serious abrasion due to its fine surface structure and the main wear mechanism is adhesive wear.

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