Characteristics Study on in-situ NbC Particles Reinforced Ni-based Alloy Composite Coating by Laser Cladding

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Abstract—The in-situ synthesized NbC particles reinforced Ni-based alloy composite coating was successfully fabricated by laser cladding a precursor mixture with a mass percentage of 80% Ni-based alloy powder (Nb10 powder) and 20% Cr3C2 powder on Q235 steel substrate. Another composite coating was also produced by laser cladding using 100% Ni-based alloy powder as cladding material for comparison. X-ray diffraction (XRD), scanning electron microscopy (SEM), energy spectrum analysis (EDS), microhardness measurement and wear resistance test were performed on the cladding coating. The results show that Ni-based alloy composite cladding coatings are homogeneous and free from cracks, the cladding coating and the substrate exhibit excellent metallurgical bonding. The main phases of the composite coating were (Fe,Ni) solid solution, NbC and Cr23C6 reinforced phases. Abundant NbC particles can be uniformly formed via in-situ reaction between niobium and Cr3C2 in the molten pool during the laser cladding process, and they are commonly precipitated in three kinds of morphologies: granular, quadrilateral and petaloid shape. However, the comparative coating was only composed of (Fe,Ni) solid solution, without the formation of any carbides. The microhardness and wear resistance of composite coating of NbC particle reinforced composite cladding coating were improved significantly, the results indicate that the existence of NbC particles plays a crucial role in enhancing the performance of the cladding coating.

1. Introduction

Metal matrix composites (MMCs) play a vital role in engineering field due to their enhanced specific properties and their multiple abilities to be tailored to suit many different applications. Laser cladding is a rapid non-equilibrium solidification technique that has been used recently to fabricate MMC coatings with fine microstructures and excellent properties[1,2]. Ceramic particles with high hardness, high melting point, high chemical stability, and excellent wear resistance were laser cladded on the surface of metal substrate that already have higher strength and higher plastic toughness to prepare ceramic particles reinforced metal matrix composites coatings, this method can effectively improve the surface hardness[3-5], wear resistance[6] and corrosion resistance[7]. Combining these two techniques can exert the favorable advantages of each, such as strong metallurgical combination, fine microstructure, low dilution rate, and modular performance[3,8].

According to the way in which the reinforced ceramic particles are added into the cladding coatings, the preparation method of the particle-reinforced composite cladding coating is divided into ex-situ method and in-situ method. The in-situ method is to add raw materials that can chemically generate reinforcements to the cladding material, then cladding materials are reacted in the high-temperature
molten pool under the laser spot and predetermined reinforcing particles are chemically generated. Compared to the ex-situ method, the in-situ method is a better preparation method, it has the following advantages: (a) The in-situ composite coatings are easily prepared, requiring fewer processing steps; (b) The in-situ formed reinforcements are thermodynamically stable at the cladding coatings, leading to less degradation in elevated temperature; (c) The interfaces between reinforcements and matrix are clean, resulting in a strong interfacial bonding; and (d) The in-situ formed reinforcing particles are finer in size and their distribution in the coating is more uniform than ex-situ method, yielding improved mechanical properties [9-14].

NbC is an extremely hard refractory ceramic material with outstanding characteristics: high melting point (up to 3873K)[4], high microhardness (up to HV2400) [13,15,16] that even harder than corundum, low thermal expansion coefficient (only 6.65×10^{-6}/K), and insoluble in cold and hot hydrochloric acid, sulfuric acid, nitric acid. In addition, the thermodynamic performance of NbC is also considerably stable. More importantly, its density (7.79g/cm3) is similar to nickel-based alloys[3,6,13,17,18], it is this characteristic that solves the problem of uneven distribution of ceramic particles[7]. Thus, NbC is an ideal ceramic reinforced phase among various carbides.

In the present study, NbC particles are designed to be synthesized by in-situ reaction of niobium and Cr_{3}C_{2} during laser process instead of NbC particles being directly added into laser molten pool. The feasibility of fabricating in-situ synthesized NbC particles reinforced Ni-based alloy composite coating using laser cladding technique is investigated. In addition, phases composition, microstructures, microhardness, wear resistance and formation mechanism of composite coatings are systematically studied. The results obtained can provide theoretical basis and technical support for the direct manufacturing.

2. Materials and experimental procedure

2.1 Raw materials

Q235 steel with the size of 100mm×100mm×10mm was used as the substrate, Nb10 alloy powders and Cr_{3}C_{2} powders were selected as cladding materials aiming at synthesizing NbC particles during laser process, which diameters were 70-110 μm. The cladding materials were homogeneously mixed together by mechanical stirring and dried in a 120 °C cabinet for 4 hours before laser cladding. The powder mixture was prepared from Nb10 alloy powders and Cr_{3}C_{2} powders in fixed ratio. The Nb10 alloy powders was 80wt% and Cr_{3}C_{2} powders was 20wt%. Table 1 and Table 2 showed the nominal composition of Nb10 powders and Q235 steel respectively.

| Elements | Cr  | Mo  | Nb   | Ni  |
|----------|-----|-----|------|-----|
| wt%      | 23±0.5 | 13±0.5 | 8~10 | Bal |

| Elements | C   | Si | Mn  | P    | S    | Fe   |
|----------|-----|----|-----|------|------|------|
| wt%      | <0.22 | <0.35 | <0.14 | ≤0.045 | ≤0.05 | Bal  |

Table 3. The chemical composition of cladding materials (wt%)

| Number | Nb10 alloy powder | Cr_{3}C_{2} |
|--------|-------------------|-------------|
| #1     | 100%              | 0           |
| #2     | 80%               | 20%         |

2.2 Laser cladding processing

Laser cladding coatings were produced by medium power fiber delivered direct diode laser system with a wave-length of 915nm (RFL-A2000D) and manufacturer is “Xi’an Zhongke Zhongmei Laser Technology Co, Ltd”. The parameters varied were the laser power, beam size, scanning speed, etc. The
experiments used the preset powder method and a single-pass cladding strategy. Before laser cladding, a 1mm layer of cladding powders were uniformly pre-placed on the surface of Q235 steel, and the focused beam is perpendicular to the steel surface during laser process. For convenience of comparison, a laser cladding composite coating which cladding material was 100wt% Nb10 alloy powders was prepared as well. Therefore, there were two different alloy powders selected as cladding materials which listed in Table 3.

The criteria for determining the optimum quality of coatings were based on a compromise of highest hardness, best homogeneity, and lowest occurrence of visible cracks and poles. The cladding procedure was repeated two times to get a two-layer coating with thickness of about 1.7 mm. The deposition direction and laser parameter of every layer are the same. The optimum parameters for fabricating coatings were established after a few trial runs. In the study, the best laser parameters were set as follows: laser power for 2000 W, laser scanning speed for 16.2mm/s, powder feeding rate for 15g/min, focal distance for 225m. An overlap of 57.1% between successive tracks was selected in order to ensure that the cladding coating surface morphology meets the standard.

2.3 Phase identification and microstructural characterization

The achieved cladded specimens were cut into pieces that were 10×10×11.6mm via wire electrical discharge machining. Then, the surfaces and transverse cross sections of the specimens were polished using standard mechanical polishing procedure. Specimens were finally ultrasonic cleaned using anhydrous ethanol and dried waiting for further examinations. The phases of specimens were identified by X-ray diffraction (XRD) with a D/max diffractometer. Cu Kα radiation (λ=1.5406Å) was applied at 40 kV and 100 mA; scans were performed in the continuous scanning mode at a speed of 8°/min; Microstructural characterization of coatings was analyzed using a S4800 scanning electron microscope (SEM) with an energy-dispersive spectroscopy(EDS).

2.4 Mechanical tests

The microhardness of coating cross-sections was measured using a FM300 micro Vickers hardness tester with a load of 500g at a loading time of 15s. From the top surface of the coating to the substrate, three sets of data were tested to calculate the average hardness. Each group was distributed at intervals of 80 μm.

The wear resistance of the surface coatings was examined at room temperature using wet sand abrasive wear tester with rubber wheel(MLS-225). To calculate the weight loss and wear rate in the abrasive wear experiment, the weight was measured before and after every 2000 revolutions of the rubber wheel for the specimens that had a dimension of 50×30×11.6mm using an electronic balance with an accuracy of 0.0001g. During the friction and wear, the weight loss was measured based on the equation (1):

$$\Delta M = M_1 - M_2$$  \hspace{1cm} (1)

where $M_1$ and $M_2$ are the weight (g) of the specimen before and after every 2000r wear, respectively, and $\Delta M$ is the weight loss(g) after every 2000r wear[5]. The wear rate was also obtained according to the following equation (2):

$$W_r = \frac{\Delta M \times 5}{\pi}$$  \hspace{1cm} (2)

Where $\Delta M$ is the weight loss after every 2000r wear, and $W_r$ is the wear rate. It should be noted that specimens need to be pre-wear (400r) to eliminate the experimental errors caused by the surface quality and uneven stress of coatings before the formal wear test. For the accuracy of the experiment, three parallel specimens were prepared for each kind of cladding coating and take the average.
3. Results and discussion

3.1 Phase composition and microstructure of cladding coatings

![XRD patterns of coating 1#](image1.png) ![XRD patterns of coating 2#](image2.png)

(a) XRD patterns of coating 1#  (b) XRD patterns of coating 2#

Figure 1. The analysis result of X-ray diffraction

![Microstructure of transverse section of the coating 1#](image3.png)

Figure 2. Microstructure of transverse section of the coating 1#: (a) the bottom (b) the middle and (c) the upper

XRD was utilized to show the phase constituents of two different coatings. Figure 1 presents the XRD patterns of the surface of the specimens prepared with different cladding material. The results showed that both of these coatings mainly composed of (Fe,Ni) solid solution that exhibited strong peaks corresponding to 43.494°, 50.673°, 74.539°, 90.459°, 95.818° of 2θ using ICDD. For coating 1#, due to the absence of carbon element in cladding material, so the phase composition of the coating 1# is relatively simple. The XRD pattern indicates that the coating 1# mainly consist of (Fe,Ni) solid solution, Fe2Nb, and FeNi3, no NbC or other carbides were observed from XRD result. In contrast, it should be noted that the diffraction peaks of NbC appeared in coating 2#, the 2θ had values of 34.730°, 40.316°, 58.337°, 69.718° and 73.306°, revealing that NbC phases were in-situ synthesized via reactions of C and Nb during laser cladding process. It can be also found that the diffraction peaks of NbC were relatively strong but slightly deviated in virtue of the changed interplanar spacing by Ni, Cr, Mo and Fe elements in the solid solution[19]. In addition, Cr23C6 phases were also detected in the coating 2# by XRD.

Figure 2 shows the microstructure of the coating #1 from bonding zone to near surface. It can be seen that the microstructure of the nickel-based laser cladding coating is a typical hypoeutectic structure, which is mainly composed of (Fe, Ni) solid solution and eutectic structure. The large area of grey
structure is (Fe, Ni) solid solution called primary crystal phase, which possesses a face-centered cubic structure, and the white network or flocculent structure is eutectic phase. The formation of this microscopic morphology is mainly related to the process characteristics of rapid heating and rapid cooling during laser cladding. The high-power energy input causes the temperature of the alloy powders to rise rapidly to form a molten pool, and crystallization occurs during the rapid cooling process. The Ni-rich primary crystal phases are formed. At the same time, high temperature also promotes the diffusion of other elements[20], which is also one of the reasons for the aggregation of Nb and Mo elements at the grain boundaries. As a result low melting point eutectic compounds are formed[6,13,20,21].

From SEM images, it can also be observed that there are some micro-pores inside the cladding coatings due to the gases generated and involved during the laser cladding process when cladding materials were melting and these gases are too late to escape during rapid solidification[20], these micro-pores may affect the mechanical properties of the coatings[22]. So many pores distributing in the clad makes the clad structure loosely that will reduce the layers compactability as well as the tensile strength. It can be deduced that once a micro-crack generates in the cladding coatings then it will propagate rapidly along pores especially the large size of pores[20,22].

It can be seen that strong diffraction peaks of NbC and Cr$_{23}$C$_6$ were detected in the coating 2# except the main phase (Fe, Ni) solid solution. Figure. 6 is the full scale image of the lower, middle, upper section of coating #2 at 1000 times magnification. It can be clearly seen that there are abundant white fine particles which evenly distributed in the cladding coating, and these large amounts of white particles are NbC, which is consistent with the previous research results[20,23,24]. The reason for the formation of NbC is that synthesis reaction of carbon with niobium occurs in the molten pool under the high temperature of laser, which is in the Ni-Fe-Cr-Mo-Nb-C reaction system. Before the laser beam reached the substrate, it passed through the pre-placed powder on the

![Figure 3. Microstructure of transverse section of the coating 2#: (a)the lower(b)the middle(c)the upper substrate which could absorb the laser energy, the residual laser energy triggered synthesis reaction and subsequent laser alloying of the synthesis reaction products onto Q235 steel substrate. Although the Nb element has a high melting point, the high temperature effect of the laser can still melt it into Nb atoms. Similarly, Cr$_3$C$_2$ will also melt and decompose in the high temperature molten pool, resulting in the dissolution of Cr atoms and C atoms in the Fe, Ni liquid phase. There may be multiple reactions in the reaction system of this molten pool. It has been concluded that possible reactions between Nb-C, Cr-C, Mo-C, Ni-C and Fe-C should be considered in this system. The formation of possible products can be predicted in view of the Gibbs free energy for the Ni-Fe-Cr-Mo-Nb-C system during laser cladding. The reaction equations accompanied by their respective free energy changes are as following equations (3)-(11)[25-27]:

\[
\begin{align*}
\text{(3)} & \quad \text{Reactions in the reaction system:} \\
\text{(4)} & \quad \text{Possible reactions in the Ni-Fe-Cr-Mo-Nb-C system:} \\
\text{(5)} & \quad \text{Predicted reactions:}
\end{align*}
\]
Figure 4. The Gibbs free energy of the synthesis reactions versus temperature.

\begin{align*}
7\text{Cr} + 3\text{C} &\rightarrow \text{Cr}_7\text{C}_3 \\
\Delta G^0(\text{KJ}\cdot\text{mol}^{-1}) &= -174.401 - 0.0259T (298\text{K} < T < 2171\text{K}) \\
23\text{Cr} + 6\text{C} &\rightarrow \text{Cr}_{23}\text{C}_6 \\
\Delta G^0(\text{KJ}\cdot\text{mol}^{-1}) &= -309.616 - 0.0774T (298\text{K} < T < 1773\text{K}) \\
3\text{Cr} + 2\text{C} &\rightarrow \text{Cr}_3\text{C}_2 \\
\Delta G^0(\text{KJ}\cdot\text{mol}^{-1}) &= -79.078 - 0.01766T (298\text{K} < T < 2130\text{K}) \\
3\text{Fe} + \text{C} &\rightarrow \text{Fe}_3\text{C} \\
\Delta G^0(\text{KJ}\cdot\text{mol}^{-1}) &= 25.92 - 0.023T (298\text{K} < T < 463\text{K}) \\
&\hspace{1cm}+ 0.025T (463\text{K} < T < 1115\text{K}) \\
&\hspace{1cm}+ 0.010T (1115\text{K} < T < 1808\text{K}) \\
\text{Mo} + \text{C} &\rightarrow \text{MoC} \\
\Delta G^0(\text{KJ}\cdot\text{mol}^{-1}) &= -75.3 - 0.00544T (298\text{K} < T < 973\text{K}) \\
2\text{Mo} + \text{C} &\rightarrow \text{Mo}_2\text{C} \\
\Delta G^0(\text{KJ}\cdot\text{mol}^{-1}) &= -45.6 - 0.00418T (298\text{K} < T < 1373\text{K}) \\
3\text{Ni} + \text{C} &\rightarrow \text{Ni}_3\text{C} \\
\Delta G^0(\text{KJ}\cdot\text{mol}^{-1}) &= 33.9472 - 0.007116T (298\text{K} < T < 1000\text{K}) \\
2\text{Nb} + \text{C} &\rightarrow \text{Nb}_2\text{C} \\
\Delta G^0(\text{KJ}\cdot\text{mol}^{-1}) &= -192.549 + 0.00419T (1180\text{K} < T < 1370\text{K}) \\
\text{Nb} + \text{C} &\rightarrow \text{NbC} \\
\Delta G^0(\text{KJ}\cdot\text{mol}^{-1}) &= -130.1800 + 0.00167T (1180\text{K} < T < 1370\text{K})
\end{align*}

where $T$ (K) is the absolute temperature. The values of Gibbs free energy $\Delta G^0$ for all reactions are found to be negative in the range of 1000-2000°C which indicates that they are thermodynamically possible to take place. During the fast solidification process, the Cr$_{23}$C$_6$ and Cr$_7$C$_3$ phases would be first synthesized in the molten pool according to the values of $\Delta G^0$ for all reactions, as shown in Figure 4. However, microstructure observation in Figure 3 and phases identification in Figure 1 show clearly that the NbC phases first precipitate in the molten pool. The first reason is that Nb is strong carbide formation element, capable of easy reaction with C to form NbC which has the highest melting point among above carbides[28]; the second reason is that the free energy of formation of NbC phase does not change significantly with temperature; the third reason is the rapid heating and rapid cooling rate by laser cladding technique causing the non-equilibrium solidification in the molten pool. The above three reasons are considered to be responsible for the phenomenon in this case[26].

Hence, it is safely assumed that NbC will be the first phase synthesizing by laser cladding and precipitating during solidification due to its high melting point. Once NbC particles were formed, they naturally became the heterogeneous nucleus of chromium carbides during the solidification process in
the molten pool. From XRD patterns results, Cr$_2$C$_6$ phases left in the final structure instead of Cr$_7$C$_3$. This result is attributed to two possible reasons, one is that Cr$_2$C$_6$ is thermodynamically the most stable among these Cr-containing carbides\cite{29}, Cr$_7$C$_3$→Cr$_2$C$_6$ transformation may occur during solidification of molten pool; another reason is that Niobium (Nb) is a strong carbide formation element, which can easily react with C to form a NbC, the formation of NbC particles results in a lack of C element in the molten pool. Insufficient content of carbon effectively suppresses the formation of Cr$_7$C$_3$ phases\cite{28}.

It is clear that Nb$_2$C has a lower Gibbs free energy than NbC. Therefore, Nb$_2$C was expected to be the first carbide phase to emerge. S. Barzilai et al.\cite{30} reported that the Nb$_2$C and NbC phases are formed on a graphite substrate by heat treatment in the temperature range of 800–1500 °C. In addition, the formation of Nb$_2$C in the Nb-Zr-C alloy and its phase transformation were studied by B. Vishwanadh et al\cite{31}. Nevertheless, the NbC phase was detected in the coating 2# from XRD patterns instead of Nb$_2$C phase. In an investigation by Cai, et al\cite{5}, it was reported that diffusion kinetics are dominant for its formation rather than the thermodynamics. The C atom radii (0.091 nm) is smaller than Nb atom (0.148 nm), leading to different diffusion mechanisms. C atoms move via interstitial diffusion and Nb atoms move via void diffusion\cite{32}, which results in the diffusion rate of the C atom being significantly faster than that of the Nb atom in the molten pool. The growth of the NbC coating can be largely attributed to the C atoms via diffusion for reaction with Nb atoms. There is another possible reason that explain the absence of Nb$_2$C phases, that is, Nb$_2$C phases form as a first carbide phase only when the supply of carbon is limited\cite{30,33}. In this research, however, the chemical composition of the cladding materials includes an abundant amount of C with a value of 2.67wt% (calculated from the Cr$_3$C$_2$ content in Table 3), which can continuously offer a C source for the in situ reaction\cite{5}.

It can be clearly seen from the SEM image that NbC particles are evenly distributed on the upper, middle and lower sections of the coating 2#, and their size is about 2 to 3μm as the result of the characteristics of rapid heating and rapid cooling of the laser cladding process which does not give synthesized NbC crystal nucleus enough time to grow after nucleation\cite{34}. From Figure. 3, although there is a good metallurgical bond between substrate and cladding coating, it can be found that there are a small number of micro-pores in the cladding layer, but the number of micro-pores is far less than the coating 1#.

![Figure 5. SEM image of the middle transverse section of the coating 2#](image)

| Region | Nb   | Cr   | Fe   | Ni   | Mo   | C    | S    | O   |
|--------|------|------|------|------|------|------|------|-----|
| A      | 57.73| 8.55 | 1.58 | 6.02 | 9.31 | 17.50| 1.30 | -   |
| B      | 2.41 | 45.99| 7.33 | 27.90| 7.80 | 8.38 | -    | -   |
| C      | 63.72| 1.00 | 0.44 | 1.34 | 7.72 | 22.91| -    | 2.87|
| D      | 4.14 | 30.03| 7.35 | 39.89| 12.74| 5.45 | -    | 0.41|

Figure 5 shows the SEM image from transverse section of middle section of coating 2#. It can be been that the dense white particles are uniformly precipitated in the cladding coating. The composition of this area is analyzed by EDS, and the results are shown in Table 6. SEM images and EDS results show that white particles (A,C) are rich in Nb and C, while other elements such as Cr, Mo, Ni, and Fe...
are much lower than those in non-white structure (B, D). Therefore, by combining XRD, SEM and EDS results, it was determined that the white structure was NbC particles, which were the enhancement phases precipitated firstly during non-equilibrium solidification [23]. There is a large amount of dark gray (B) structure densely distributed in the middle section of coating 2#, which is rich in Cr and Ni, and contains a certain amount of Fe and C. Combined with XRD analysis results, it can be judged that the dark gray structure is Cr$_{23}$C$_6$ phase, in which Cr atoms can be partially replaced by Ni or Fe atoms[35]. Also, a large amount of light gray structure mixed with a dark gray structure (D) exists around NbC particles and bulk Cr$_{23}$C$_6$ phases, which is rich in Ni and Cr, and contains a certain amount of Fe and Mo. This is a mixture of (Fe,Ni) solid solution and Cr$_{23}$C$_6$, where Cr atoms may be solute in (Fe,Ni) solid solution.

As mentioned above, the white particles in figure. 5 are NbC enhanced phases, it can be found that their shapes are close to granular, quadrilateral and petaloid. This morphology of NbC is consistent with that described in numerous literatures [5,6,10,15,19]. Ref[10] reported that the growth morphology of NbC is closely related to different carbon contents correspond to different enthalpies of reaction, which changes the dynamic temperature of the molten pool and the growing environment of crystals. Under constant nominal experimental parameters (laser power), the exothermic behavior of MC carbides generated in-situ affect the temperature dynamics of the molten pool, which in turn affects the solidification rate. Therefore, the growth morphology of the internal structure of the coating will also correspondingly change, especially the MC carbide particles that preferentially precipitate. As the carbide mass fraction in the cladding powder increases, the dynamic reaction heat of the molten pool will also increase, and the temperature gradient of the molten pool will decrease, which provides a longer growth time for carbide. In this paper, the carbon content of coating 2# is about 2.67 wt%, the growth orientation behavior and microstructure of NbC phase are similar to octahedron, which results in the morphology of NbC phases looks like petal and quadrangle in BSE image at a particular angle.

3.2 Microhardness of cladding coatings

Microhardness profile along the cross-section of the laser cladding coatings (1# and 2#) is shown in Figure. 6. The microhardness of both of coatings are increased obviously compared to substrate.

![Figure 6. Microhardness distribution of coating 1# and coating 2#](image)

The microhardness of coating 2# was significantly increased, giving a high average hardeness of HV695, which indicated the microhardness was higher 5.26 times than that of Q235 steel (HV132) and 1.92 times than that of coating 1# (HV361). The significant increase in hardness of coating 2# can be summarized as the following two reasons: First of all, NbC particles in the cladding coating 2# exert important influence on enhancing the mechanical properties. The microhardness of NbC is up to
2400HV[33,36]. Such high microhardness of NbC particles were not only in-situ synthesized, but also distributed evenly in the cladding coating. They can effectively pin the grain boundaries, prevent the coarsening of the grains, control the size of the grains, and finally play the role of precipitation dispersion strengthening and fine grain strengthening. Secondly, the effects of alloying elements cannot be ignored. The main structure of the cladding coating is (Fe,Ni) solid solution, Cr atoms may replace the position of Ni atoms, and due to the difference in atomic size of the two atoms, lattice distortion will occur, strengthening the solid solution primary phases. However, Cr, Mo, Fe, and Nb atoms may also be dissolved in the matrix and play a role in solid solution strengthening[37].

3.3 Wear resistance of coatings
Figure 7 shows the wear results of cladding coatings. As seen in Figure 7, mass loss by abrasive wear test increased with the number of rubber wheel rotations. However, it could be clearly seen that the weight loss of coating 2# was lower than coating 1# after 10000 rotations of abrasive wear. Therefore, it could be judged that the wear resistance of cladding coating was significantly improved after the NbC particles in-situ synthesized[21]. Combined with the hardness results, the wear resistance was directly proportional to the microhardness of the cladding coating. The higher microhardness was, the better wear resistance coating exhibited[31,34].

![Figure 7. Wear weight loss of the cladding coatings](image)

The above analyses reveals that the key reason for the significant improvement in wear resistance of coating 2# is that numbers of NbC particles and chromium carbides(Cr23C6) are in-situ synthesized and evenly distributed in the cladding coating. The average distances among these particles were reduced, which protruded gradually from the surface to resist wear. During the process of abrasive wear, the binder phase (Fe, Ni) with low hardness in the cladding coating undergoes metal-ceramic contact with the rubber wheel at high sliding speed, indicating adhesion can take place between the tribo-surfaces, which also results in the subsequent material transfer. This in turn leads to the exposure of NbC particles to the steel counter surface[26]. However, the high hardness of NbC had pinning effect on the metal matrix during the initial stage of the wear, which effectively carried the load, resisted the plastic deformation, prevented abrasive from pressing in the matrix of coating, kept the surface from further abrasion and reduced the contact area between the matrix and the friction pair, it remarkably relieved the adhesive wear and decreased the wear debris[38]. Since the NbC particles were formed in-situ, the interfacial bonding between the (Fe,Ni) matrix and NbC reinforcing phase tended to be stronger, resulting in the absence of pullout of the NbC particles. Hence, the NbC particles were worn via micro-polishing mechanism. Taken together, the dominating wear mechanism for the cladding coating was a mixed adhesive and micro-polishing mechanism.
4. Conclusions
Two kinds of nickel-based alloy composite coatings are successfully prepared on Q235 steel by laser cladding. Their cladding materials are different, coating 1# (comparative sample) used 100wt% Nb10 powder as the cladding material, while coating 2# used a mixed powder of 20wt%Cr3C2 + 80wt%Nb10 alloy powder as the cladding material. XRD, SEM, EDS, microhardness and wear resistance test are carried out on the laser cladding coatings. The following conclusions are obtained by means of the above test results.

(1) Both coating 1# and coating 2# are continuously formed and smooth on their surface, without cracks, but with a small amount of pores. An excellent metallurgical bonding at the interface between coating and substrate can be obtained by laser cladding.

(2) The main phases of the coating 2# are (Fe, Ni) solid solution and Cr23C6. Besides, a large amount of NbC reinforced particles are produced by in situ reacting Cr3C2 and niobium during laser cladding process, they are uniformly dispersed in the coating 2# in three different morphologies: petaloid, quadrangular, and granular. However, the coating 1# is only composed of (Fe,Ni) solid solution, without the formation of any carbides or other reinforcing phases due to the absence of Cr3C2 in cladding material.

(3) The microhardness of coating 2# is higher 5.1 times than Q235 steel and 2 times than coating 1#, it also exhibits better tribological property compared to coating 1#. Abundant in-situ synthesized NbC particles which are evenly distributed in the coating 2# play an important role in fine grain strengthening and dispersion strengthening. The solid solution strengthening of alloying elements reinforces the coating as well.

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