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Phase assemblage and properties of laser cladded Ti_{x}CrFeCoNiCu high-entropy alloy coating on aluminum

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Abstract

The Ti_{x}CrFeCoNiCu (x:molar ratio, x = 0, 0.2, 0.5, 0.8, or 1.0) coating was deposited on aluminum by laser cladding. The phase structure, microstructure, hardness, wear resistance and corrosion resistance were studied. The results show that with the increase of Ti content, the phase structure of the Ti_{x}CrFeCoNiCu coating changes from single FCC to FCC + B2, and FCC + Laves phase. When Ti is increased to 1.0, cracks appear in the coating. The hardness of the Ti_{x}CrFeCoNiCu coating is enhanced with the increase of Ti content, and ranges from 215HV_{0.2} to 585HV_{0.2}, which is about 3 to 7 times that of the substrate. The strengthening mechanism of Ti_{0.2}CrFeCoNiCu is solid solution strengthening, and when the Ti content is greater than 2, the strengthening mechanism of Ti_{x}CrFeCoNiCu coating is precipitation strengthening. The influence of Ti on the wear resistance exhibits the same trend as with hardness. When Ti increased from 0 to 0.8, the wear rate of the Ti_{x}CrFeCoNiCu coating changed from 2.26 × 10^{-4} mm^{3}N^{-1} to 9.92 × 10^{-7} mm^{3}N^{-1}; smaller than the substrate. The addition of Ti increases the current corrosion density of Ti_{x}CrFeCoNiCu coating, but both coatings still exhibits superior corrosion resistance relative to the substrate.

1. Introduction

Aluminum alloy has characteristics of low density and high specific strength and has been widely used in many fields [1–4]. However, due to the lower hardness and poor wear-resistance of aluminum alloy surfaces, its applications are limited in cutting-edge areas where reliability is more demanding. The popular aluminum alloy surface strengthening methods include chemical plating, electrodeposition techniques, and laser cladding (LC) [5–8]. Compared with other coating technologies, the laser cladding method has the advantages of producing lower pollution, higher efficiency, and thicker coating and a smaller heat-affected zone [9]. Traditional laser cladding materials include iron-based, nickel-based and ceramic materials. During laser cladding on aluminum, the increases of the laser heat input necessary to ensure the full melting of the those materials, but the dilution of Al in the substrate are increasing. Also, Al tends to react with traditional LC materials to form intermetallic compounds and cluster together in the cladding layer due to its high electronegativity and low density. The hardness and brittleness of intermetallic compounds are harmful characteristics as they can become the crack source in the cladding layer, resulting in the reduction of coating performance. Therefore, the preparation of coating on aluminum by LC has been a difficult problem in the field of laser surface engineering.

The design concept of high entropy alloy (HEA) is different from that of traditional alloy. The number of principal components in HEA is at least 5, and the mole fraction of each element is between 5% and 35% [10]. Existing research results show that HEAs do not tend to form a wide variety of hard and brittle intermetallic compounds after solidification [11–13]. Due to the high mixed entropy effect, it is easier to generate single-phase or dual-phase mixed structures with simple face-centered cubic, or body-centered cubic. Meanwhile, HEA has excellent properties such as high strength, high temperature oxidation resistance [14–16]. Fan et al prepared
as-cast AlxCu0.5(FeCrNiCo) (0.5 ≤ x ≤ 1.0) HEA, where the structure of an AlxCu0.5(FeCrNiCo) HEA transformed from a FCC into a BCC + FCC phase, and then to a BCC phase as increasing Al level. Moreover, the increase in Al content enhanced the yield strength and hardness of Alx(FeCrNiCo)Cu0.5 HEA. The hardness of the coating is 740 HV, about 10 times of that of the substrate, and the corrosion rate is much lower than that of the substrate [17]. Du et al have studied Ti addition effect on the formation of phase in Ti0FeCoCrNiCu0.5 HEAs, and both Ti0.8FeCoCrNiCu0.5 and Ti0.5FeCoCrNiCu0.5 HEA only consist of FCC phase, while x > 0.3, sigma and Laves phase were detected [18]. Siddiqui et al fabricated single-channel AlTiFeNiCu0.5 HEA coating on AA 1050 aluminum by LC. The coating consists of FCC1, FCC2 and BCC phases. Compared with the AA 1050 aluminum, the microhardness and corrosion resistance of the AlTiFeNiCu0.5 HEA coating were obvious enhance [19]. Our group used laser cladding to synthesized single-channel AlCrFeCoNiCu on 5083 aluminum and found that the coating showed better corrosion resistance in 1 mol L−1 H2SO4 solution, the corrosion current density was only 65.2% of the substrate [20].

The above survey shows that HEAs have good mechanical properties and corrosion resistance, which enables them as coatings. Al exhibits high solid solubility in the HEA system comprising of Fe, Co, Ni, Cr, and Cu. This means that the Al in the substrate and the added powder will tend to form solid solutions after reaction, thereby reducing the cracks caused by the intermetallic compound clusters. But this theory is still required to be verified by experiments. There is only limited research available on the improvement of surface wear resistance of the aluminum alloy with HEA coating. In addition, Ti has been proved to improve the mechanical properties of HEA due to its large atomic radius and the ability to promote the formation of reinforced phase. It would be interesting to study the influence of Ti on the properties of as-cladded HEA coating containing Fe, Co, Ni, Cr, and Cu elements. In this study, TiCrFeCoNiCu coating with different Ti content was prepared on aluminum, and its microstructure and properties were analyzed in detail.

2. Materials and methods

5083 aluminum alloy was selected as the substrate due to its good superplasticity and weldability. Its composition as follows: 4.0% ≤ Mg ≤ 4.9%, Si ≤ 0.4%, Cu ≤ 0.1%, Fe ≤ 0.4%, 0.5% ≤ Cr ≤ 0.25%, Zn ≤ 0.25%, Ti ≤ 0.15%, and the rest is Al. A sample size of 50 × 30 × 10 mm was used. As the oxide film on substrate can affect the bonding strength, the surface of the substrate was polished with 800–325 meshes. The mixed powder was ball milled in an argon atmosphere for 2 h and was then placed in a vacuum oven to dry for 2 h. It was then finally sealed and stored. The powders ratio and abbreviations of all samples are shown in table 1.

The DC050 slat-type CO2 laser equipment by Rofin Company was used. The motion device was an CNC machine tool. The powder feeding device was used, and the shielding gas was argon. The schematic diagram of laser cladding is shown in figure 1. Based on previous experiments [20–22], the process parameters are as follows: spot diameter 1.2 mm; laser power 1850 W; powder feed rate 5.6 g min−1, scanning speed 120 mm min−1, duty cycle 70%, pulse frequency 50 Hz, overlap rate 30%.

The X-ray diffraction (XRD, Empyrean), scanning electron microscope (SEM, 6510F), transmission electron microscope (TEM,2100F) were used to analyze phase structure and microstructure of the samples. The wear test was performed at room temperature. The wear track used a GCr15 steel ring with a hardness of 61 HRC, wear sample size of 7 × 14 × 10 mm. The test load was 98 N, the speed was 400 r min−1, and time allowed was 30 min. The wear rate of the material was evaluated by calculating the wear volume. The formula is as follows:

$$\omega = \frac{V_{loss}}{L \times N}$$  \hspace{1cm} (1)

$$L = \frac{2\pi Rvt}{2}$$  \hspace{1cm} (2)

| Coating Abbreviation | Ti | Fe | Co | Ni | Cr | Cu |
|----------------------|----|----|----|----|----|----|
| FeCoNiCrCu           | Ti0| 0  | 20 | 20 | 20 | 20 |
| Ti0.2CrFeCoNiCu      | Ti0.2| 3.84| 19.23| 19.23| 19.23| 19.23|
| Ti0.5CrFeCoNiCu      | Ti0.5| 9.09| 18.18| 18.81| 18.81| 18.81|
| Ti0.8CrCoNiCu        | Ti0.8| 13.79| 17.24| 17.24| 17.24| 17.24|
| TiCrFeCoNiCu         | Ti1.0| 16.60| 16.66| 16.66| 16.66| 16.66|

Table 1. Nominal components and abbreviations of TixCrFeCoNiCu coatings(at%).
where \( \omega \) is the wear rate, \( V_{\text{loss}} \) is the wear volume (mm\(^3\)), \( B \) is the wear track length (mm), \( b \) is the wear track width (mm), \( \nu \) is the rotating speed of the wear ring (r min\(^{-1}\)), and \( L \) is the total wear ring sliding length (mm).

Figure 2 shows the schematic diagram of the wear test.

The polarization test uses an electrochemical workstation (Zahner X). A three-electrode working system was used: the working electrode was the experimental sample (WE), the counter electrode (CE) was platinum, and the reference electrode (RE) was saturated KCl solution. Select 3.5% NaCl solution for corrosion solution. The polarization test of the HEA coatings were measured with scanning rate of 0.3 mV s\(^{-1}\), 0.5 mV s\(^{-1}\) and 5 mV s\(^{-1}\).

3. Results and discussion

3.1. Phase and microstructure

Figure 3 shows the XRD pattern of TixCrFeCoNiCu HEA coating, all samples show FCC phase diffraction peaks. The enlarged view of near \( 2\theta = 43.5^\circ \) shows that the main diffraction peak shifts to the low angle with the increase of Ti, indicating that addition of Ti increases the lattice parameter of the FCC phase, of which from 3.918 Å to 3.954 Å. This is due to Ti has the largest radius among added powders, addition of Ti increases the degree of lattice distortion of the solid solution [23]. When Ti increases to 0.5, the diffraction peak at near \( 2\theta = 63.5^\circ \) appeared, which is identified as the B2 phase, with lattice parameter of 2.972 Å, should be ranged as the Ni-Ti type. For Ti0.8 and Ti1.0 coatings, the B2 diffraction peak disappeared, and the diffraction peak of Laves phase is detected. The lattice parameters of Laves phase is \( a = 4.767 \) Å and \( c = 7.723 \) Å, corresponding to an Fe\(_2\)Ti type.

Figure 4 shows the surface macro-photograph and EDS line scanning along the cross section of FeCoNiCrCu coating. It can be observed that cracks appeared on the top of the coating and pores appeared at the interface near...
the substrate. The main reasons for the formation of cracks is a high temperature gradient in the coating during processing. When the beam leaves the coating, the cooling speed at the top of the coating is rapid, resulting in the local area on the top of the coating being subjected to larger shrinkage stress during the crystallization process and eventual crack formation. The formation of pores is speculated as follows: compared with coating, the melting point and boiling point of the aluminum alloy are lower. On the one hand, under the action of high-energy laser, the vaporization of the substrate provides a source for the pores [24]. On the other hand, the lower melting point of the substrate causes the final solidification during LC. While LC is a process of rapid cooling that results in the gas not having enough time to overflow, it remains in the substrate. It can be seen from the result of EDS line scanning along the coating cross section that there is element diffusion between the coating and the substrate, indicating that the coating and the substrate form a metallurgical bonding. It is found that the distribution of Al elements is a negative gradient along the thickness direction. The Cu content is slightly lower than that of the other four added powders, which indicates that the vaporization of Cu occurs during LC.

Figure 5 shows typical microstructure of TiCrFeCoNiCu HEA coatings. The dendritic structure of the Ti0 exhibit fine cellular morphology characteristics, which is a typical rapid solidification structure (figures 5(a) and 4(b)) [25]. Table 2 shows the EDS result of each region in figure 5. The results of EDS show that dendrite region are enrich with Fe, Co, Ni, and Cr, while Cu segregation occurs in the interdendritic region. Furthermore, the result of XRD reveal that Ti0 is a single FCC phase, indicating that the dendrites and the interdendrites are two different components but having the same crystal structure. The dendrites are (Fe, Co, Ni, Cr)-rich FCC solid solutions, and the interdendrites are Cu-rich phases. Cu has a positive mixing enthalpy and a lower melting point, so it precipitates at the grain boundary [26]. The partial dendrites of the Ti0.2 exhibit a petal-like shape...
The proportion of elements in each region is close to that of Ti0. However, when Ti reaches 0.5, micron precipitation (marked as A) can be observed in figures 5(d) and 4(f). EDS results show that Ni(24.38 at-%) and Ti(20.45 at-%) are enriched in the precipitation, corresponding to B2 phase in XRD patterns. The more negative binary enthalpy of mixing between Ti and Ni promotes the formation of B2 phase \[18\]. When Ti is increased to 0.8, the net distribution of white precipitates along the grain boundary can be observed in figures 5(g) and (h). EDS results show that the Ti and Fe content in the white precipitates is higher than the nominal composition, and the ratio is close to 1:2, corresponding to a Laves phase. As shown in figure 5(i), crystalline cracks occur on Ti1.0, and the second-phase particles were mixed up with the fracture.

Table 2. EDS results of areas in figure 5 (at%).

|       | Al | Ti | Cr | Fe | Co | Ni | Cu |
|-------|----|----|----|----|----|----|----|
| Ti0   | Nominal | 0 | 0 | 20 | 20 | 20 | 20 |
|       | DR1 | 3.45 | 0 | 24.31 | 24.02 | 22.43 | 20.42 | 5.37 |
|       | ID1 | 4.39 | 0 | 7.21 | 6.39 | 8.42 | 13.34 | 60.25 |
| Ti0.2 | Nominal | 0 | 3.84 | 19.23 | 19.23 | 19.23 | 19.23 | 11.34 |
|       | DR1 | 2.43 | 1.15 | 21.32 | 24.5 | 20.45 | 19.86 | 11.34 |
|       | ID1 | 3.45 | 5.41 | 10.45 | 9.45 | 15.32 | 12.54 | 43.25 |
| Ti0.5 | Nominal | 0 | 9.09 | 18.18 | 18.18 | 18.18 | 18.18 | 18.18 |
|       | DR2 | 3.21 | 6.56 | 20.45 | 20.22 | 19.02 | 19.2 | 11.34 |
|       | ID2 | 2.47 | 14.25 | 12.58 | 13.45 | 15.32 | 14.24 | 27.69 |
|       | A   | 1.45 | 20.45 | 14.32 | 12.07 | 14.11 | 24.38 | 13.22 |
| Ti0.8 | Nominal | 0 | 13.79 | 17.24 | 17.24 | 17.24 | 17.24 | 11.34 |
|       | DR3 | 3.13 | 9.32 | 19.51 | 16.91 | 20.64 | 20.04 | 10.45 |
|       | ID3 | 2.76 | 18.24 | 10.32 | 36.12 | 8.14 | 5.17 | 19.25 |
|       | B   | 0.42 | 30.68 | 3.67 | 61.56 | 1.56 | 1.94 | <0.1 |
| Ti1.0 | Nominal | 0 | 16.66 | 16.66 | 16.66 | 16.66 | 16.66 | 16.66 |
|       | C   | 1.47 | 29.64 | 2.86 | 62.84 | 0.89 | 0.67 | 1.63 |

(figures 5(c) and 4(d)).
proportion of each element in the second-phase particles is close to the white precipitates of Ti0.8, so the second-phase particles is identified as a Laves phase. It is worth mentioning that although Al is detected in both samples by EDS analysis, the intermetallic compounds with Al-rich have not been observed, which indicates the coating defects caused by the dilution behavior of the substrate is controlled to some degree by process parameters and power ratio in this experiment.

3.2. Hardness
The average hardness of the substrate and TixCrFeCoNiCu HEA coating is shown in figure 6. As the Ti content increases, the hardness of the TixCrFeCoNiCu HEA coating gradually enhanced, and ranges from 215HV0.2 to 585HV0.2, which is about 3 to 7 times that of the substrate. Ti0 and Ti0.2 consist of only the FCC phase, so the hardness is obviously lower than that of T0.5 and Ti0.8. Ti0.5 and Ti0.8 showed higher hardness mainly due to the precipitation of harder B2 phase or Laves phases particles. In summary, when the Ti content is increased from 0 to 0.2, hardness enhancement mechanism of the coating is solid solution strengthening. To further increase the Ti content, the hardness enhancement mechanism of coatings are precipitation strengthening.

3.3. Wear test
Figure 7 shows the hardness and wear rate curves of TixCrFeCoNiCu coatings and substrates. Table 3 shows the calculation results of the wear rate. The wear rate is calculated according to formula (1)~(3). It can be observed that the wear rate of TixCrFeCoNiCu coating is significantly lower than that of substrate, and with the increase of Ti content, the wear rate of TixCrFeCoNiCu HEA coating decreases, which is opposite to the change trend of hardness. This result is in line with the Archard equation [27]. The wear rates of Ti0, Ti0.2, Ti0.5, and Ti0.8 are 2.26 × 10^{-4} mm^3 Nm^{-1}, 1.39 × 10^{-4} mm^3 Nm^{-1}, 8.16 × 10^{-5} mm^3 Nm^{-1}, and 9.27 × 10^{-4} mm^3 Nm^{-1}, respectively, which are 64.6%, 39.7%, 23.3%, 0.28% of the substrate. These results suggest that TixCrFeCoNiCu HEA coatings have wear resistance against substrate. From figure 5(i), it can be seen that the cracks occur in the Ti1.0, resulting in the coating have outlived application value. Therefore, Ti1.0 was not performed wear test.
Figure 8 shows the wear surface of \( \text{TixCrFeCoNiCu} \) and substrate (a) Substrate (b) \( x = 0 \) (c) \( x = 0.2 \) (d) \( x = 0.5 \) (d) partial enlargement of figure (d) (e) \( x = 0.8 \).

### Table 3. The measured results of wear scar. and wear coefficient.

| Alloy            | Length (mm) | Width (mm) | Wear coefficient (\( \text{mm}^3 \text{Nm}^{-1} \)) |
|------------------|-------------|------------|-------------------------------------------------|
| Substrate        | 7           | 14         | \( 3.50 \times 10^{-4} \)                        |
| \( \text{FeCoNiCrCu} \) | 7           | 12.16      | \( 2.26 \times 10^{-4} \)                        |
| \( \text{TixCrFeCoNiCu} \) \( 0 \) | 7           | 10.39      | \( 1.39 \times 10^{-4} \)                        |
| \( \text{TixCrFeCoNiCu} \) \( 0.2 \) | 7           | 8.73       | \( 8.16 \times 10^{-5} \)                        |
| \( \text{TixCrFeCoNiCu} \) \( 0.5 \) | 6.12        | 2.53       | \( 9.92 \times 10^{-7} \)                        |

Figure 8 shows the wear surface of the substrate and \( \text{TixCrFeCoNiCu} \) HEA coatings. As shown in figure 8(a), wear morphology of the substrate exhibit a stepped shape. Along the wear direction, a serious plastic deformation and elongated dimple appear, indicating that delamination fracture and serious adhesion wear occurred during the wear process. This wear characteristic is attributed to the aluminum alloy having stronger ductility and plasticity. At wear initialization, the materials near the initial crack tend to elongate, forming thin areas, rather than propagating along the track of the crack and fall off. As the wear continues, more thin areas are formed. Under cyclic shear stress, cracks with adjacent thin areas converge, causing the entire elongated piece to break altogether, at the same time, the broken surface produces plastic deformation and dimples. After that, the newly generated surface repeats the above process, and the wear surface of the substrate forms a stepped shape.

The wear mechanism of Ti0 is delamination fracture (figure 8(b)). Unlike the substrate, dimples are not found on the wear surface, mainly because the plasticity of Ti0 is lower than that of the aluminum alloy. The wear surface of Ti0.2 has large area spalling and transverse cracks perpendicular to the wear direction (figure 8(c)). This evidence suggests that adhesive wear and delamination fracture took place during the wear test. During testing, one part of the material in Ti0.2 surface adheres to the GCr15 grinding block under the action of pressure, and the adhesive position is destroyed in the subsequent relative movement. After the initial crack formation, it spreads easily along the surface under the action of the friction block due to the low ductility of the material, and finally the surface falls off, and spalling forms. The results of XRD show that Ti0 and Ti1.0 are only composed of FCC phase, its low hardness having little resistance to the micro-pitching in and micro-cutting from the asperities of the GCr15 steel ring, which leads to a greater degree of wear.

The scratches and a small amount of wear debris appear on the wear surface of Ti0.5 (figure 8(d)). Some of the material surfaces are stretched along the shear direction, showing ductile fracture characteristics similar to shear fracture. Figure 8(e) is a partial enlarged view of figure 8(d), cracks perpendicular to the wear direction can be observed, but there is no obvious spalling around the cracks. The microstructure of Ti0.5 is composed mainly of FCC matrix and dispersed B2 phase. As stated above, the hard phase can better hinder the micro-pitching in and micro-cutting from the asperities of the wear ring, so the wear surface of Ti0.5 is smoother than that of the Ti0.2. In addition, some of the hard phases of the coating will fall off during the wear process, it will become abrasive particles, thus scratches can be observed on the wear surface. Further more, there is no obvious delamination or fractures around the crack. It can be inferred that the crack was formed in the process of adhesive wear.
As shown in Figure 8(f), compared with the wear morphology of Ti0.5, the wear surface of Ti0.8 is smoother with less deformation and no transverse cracks, but there are furrows in different directions, and fine abrasive particles, indicating that the wear mechanism of Ti0.8 is abrasive wear. There are a lot of Laves phases in the Ti0.8, and the appearance of a hard phase improves the overall wear resistance. The reasons for the formation of furrows with different directions is as follows: the Laves phase falls off under the continuous shear stress, and its brittleness is more easily crushed into fine abrasive particles by the GCr15 friction block, while in the wear experiment, the wear tester will vibrate noticeably. The fine abrasive particles are more likely to deviate from the direction of the shear force wear ring during vibration, causing the direction of the furrows to be inconsistent with the wear direction.

3.4. Corrosion resistance
Figure 9 shows the polarization curve of Ti_xCrFeCoNiCu coating under different scanning rates. Table 4 shows the electrochemical parameters of samples after the potentiostatic polarization test under different scanning rates. It can be observed from figures 9(a) - (d) that the passivation interval appears in both samples, indicating that a stable passivation film will be formed on the working electrode (WE) surface during the electrochemical corrosion process. From the classical electrochemical theory, the free corrosion potential ($E_{corr}$) only represents the
thermodynamic trend of the WE during the corrosion process, while corrosion current density ($I_{corr}$) is caused by the dissolution of the WE. $I_{corr}$ is determined in the linear part of the polarization curve by deducting Tafel extrapolation. Therefore, $I_{corr}$ is the most important index for evaluating the corrosion resistance of WE [28, 29]. The $I_{corr}$ of Ti$_x$CrFeCoNiCu coatings increases with the increase of Ti, indicating that Ti reduces the corrosion resistance of Ti$_x$CrFeCoNiCu coatings [30]. However, the $I_{corr}$ of Ti0.8 is still lower than that of the substrate, indicating that the Ti$_x$CrFeCoNiCu HEA coating has superior corrosion resistance. In figure 9(c), the $I_{corr}$ obtained at 0.3 and 0.5 mV s$^{-1}$ have little difference, but are significantly lower than that obtained at 5 mV s$^{-1}$. This is because the electrode reaction did not reach steady-state during the potentiodynamic scanning process at 5 mV s$^{-1}$ [31]. While the scan rate decreased to 0.5 mV s$^{-1}$, the electrode reaction tended to steady-state [32].

Figure 10 shows the corrosion morphology of samples. Table 5 shows the results of EDS in each area of figure 10. The surface of the substrate has been completely corroded. The surface corrosion morphology of Ti$_x$CrFeCoNiCu HEA coatings is relatively complete. As shown in figures 10(c) and (d), the corroded surface of Ti0 is uneven, where corrosion products and grain boundaries can be observed. As shown in figures 10(e) and (f), the surface morphology of Ti0.2 is similar to that of Ti0, showing intergranular corrosion morphology. The EDS results show that the corrosion products of Ti0 and Ti0.2 are enriched with Cu, indicating that intergranular corrosion is caused by copper segregation. As shown in figures 10(f) and (g), the grain profile of Ti0.5 can be clearly observed. The corrosion products are enriched with Ti and Ni, proving that B2 phase is easier to corrode, which is consistent with the research results of Han et al. [33]. In figures 10(h) and (i), it can be noted that the intergranular corrosion characteristics of Ti0.8 are still significant, and a lot of diamond-shaped pits are

![Image](image-url)
observed. The corrosion products are enriched with Ti and Fe, making the Laves phase as the main reason for pitting corrosion. Although the corrosion time of coatings obtained at 0.5 mV s\(^{-1}\) is higher than that obtained at 5 mV s\(^{-1}\), the corrosion degree for each sample is still similar. This also confirms that the HEA coating has better corrosion resistance in 3.5% NaCl solution.

4. Conclusion

1. Under the current laser deposition process, cracks will form at the top of the coating, and pores will form near the substrate at the interface. Both TixCrFeCoNiCu coatings have FCC phase. Ti0 and Ti0.2 coatings only consist of FCC phase. When Ti content increased to 0.5, the discrete distribution B2 phase particles appeared in the coating. With the further increase of Ti content, the chain-like Laves phase appeared in the grain boundary of the coating. Ti1.0

2. The hardness of the TixCrFeCoNiCu HEA coating increases with the increment of Ti, ranging from 215 HV0.2 to 585 HV0.2, about 3 to 7 times of the substrate. The hardness strengthening mechanism of Ti0.2 coating is solution strengthening. When x > 0.2, the main mechanism of hardness strengthening of TixCrFeCoNiCu coatings is precipitation strengthening.

3. When Ti content increased from 0 to 0.8, the wear rate of the TixCrFeCoNiCu coating decreased from \(2.26 \times 10^{-6} \text{ mm}^2 \text{ N m}^{-1}\) to 9.92 \(\times 10^{-4} \text{ mm}^2 \text{ N m}^{-1}\). The wear mechanisms of Ti0 and Ti0.2 are delamination fracture and adhesive wear, Ti0.5 is adhesive wear and abrasive wear, and the Ti0.8 is only abrasive wear. The wear resistance of both samples are better than that of the substrate.

4. The corrosion resistance of the coating decreases with the increasing of Ti. The corrosion behavior of substrate is total corrosion, the corrosion behavior of Ti0, Ti0.2 and Ti0.5 is intergranular corrosion, while the corrosion behavior of Ti0.8 is pitting corrosion and intergranular corrosion. Compared with the substrate, all coatings show better corrosion resistance. The \(I_{corr}\) of HEA coatings obtained at 0.3 and 0.5 mV s\(^{-1}\) have little difference, but are significantly lower than that obtained at 5 mV s\(^{-1}\).

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Disclosure statement

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