Effect of Ti Addition on the Microstructure and Property of FeAlCuCrNiMo_{0.6} High-Entropy Alloy

Yunhai Su¹ · Xuewei Liang¹ · Yunqi Liu¹ · Zhiyong Dai¹

Received: 10 September 2019 / Revised: 1 December 2019 / Published online: 17 March 2020
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Abstract

FeAlCuCrNiMo_{0.6}Ti_x ($x = 0.2, 0.6, 0.8$) high-entropy alloys were fabricated with flux-cored wire and surfaced on low-carbon steel by gas metal arc welding. The effect of Ti addition on the microstructures and properties of FeAlCuCrNiMo_{0.6}Ti_x HEAs surfacing layer was systematically studied. The results show that FeAlCuCrNiMo_{0.6}Ti_x HEAs surfacing layer exhibits a majority of BCC solid solution based on Fe–Cr phase and a small amount of MC phase. Microhardness and wear tests results show that the addition of Ti can significantly improve the microhardness and wear resistance of the surfacing layer. The microhardness of surfacing layer can reach up to 642 HV_{0.5}, which is about 2.8 times that of the substrate. Meanwhile, the wear mass loss is minimum. Electrochemical corrosion test results show that the addition of Ti can improve the corrosion resistance of surfacing layer, which is better than 304 stainless steel substrate. With the increase in Ti, the corrosion resistance increases first and then decreases. When the Ti content is 0.6, the self-corrosion current density of polarization curve is the smallest, the capacitive impedance radius of electrochemical impedance spectroscopy is the largest, and the corrosion resistance effect is the best.

Keywords High-entropy alloy · Surfacing · Flux-cored wire · Microstructure · Microhardness · Wear resistance · Electrochemical corrosion

1 Introduction

In traditional alloys, one or two metal elements are used as the main elements (the content is more than 50%), and the remaining metal or non-metal elements are used as trace complementary elements to improve the properties of alloys. With the increasing element, intermetallic compounds including brittle phase will precipitate within the alloy, and the properties of the alloy will seriously decline [1]. The concept of the high-entropy alloys was firstly introduced in 1995, in a master's research paper [2]. High-entropy alloy (HEA) refers to a new type of alloy material without major components and with the same or similar atomic percentages of each component, which is characterized by large atomic chaos, multiple components and simple phase [3, 4].

Through reasonable alloy formulation design, alloy materials with high hardness, high strength, high wear resistance, corrosion resistance and other excellent characteristics can be obtained [5–8], which has been widely investigated by researchers. In the study of alloys, the types and contents of elements are adjusted so that they have a unique microstructure and a variety of properties. Zhou et al. [9] prepared AlCoCrFeNiTi_x HEAs with a single body-centered cubic (BCC) structure for the first time and carried out relevant researches on composition optimization. Sun et al. [10] prepared various alloys each of which had 5–8 elements, such as AlFeCuCoCr, AlFeCuNiCrV, AlFeCuCoNiCrTi and AlFeCuCoNiCrTiV, to study the impact of the element number on the microstructure and performance of HEAs. Results show that the alloy microstructure has a trend of transition from dendrites to equiaxed crystal. The hardness of alloy increases with the element number. Otto et al. [11] prepared CoCrFeMnNi alloy by arc melting and cold rolling and studied the change of mechanical properties with temperature at low temperature range.

Titanium alloy has high strength, low density, strong corrosion resistance, high temperature stability and other excellent

Available online at <http://link.springer.com/journal/40195>.

✉ Yunhai Su
su_yunhai@sut.edu.cn

¹ School of Materials Science and Engineering, Shenyang University of Technology, Shenyang 110870, China

properties, so the addition of Ti in HEAs can combine the excellent properties of HEA with those of titanium alloy. In addition, Ti is a passivated film-forming element with good corrosion resistance, so the properties of HEA containing Ti are widely investigated. The mechanical properties and corrosion resistance of HEAs containing Ti have been studied recently [12–15]. The preparation technology of HEA includes vacuum melting casting, powder metallurgies and surface modification. Block HEAs are mainly prepared with vacuum arc furnace melting and melt casting [16, 17], but the size is limited. Therefore, the preparation of large HEA composite materials by depositing HEA on low-carbon steel surface is necessary. It can reduce the cost of raw materials or avoid the replacement of parts, which is of great significance for practical applications. Consequently, the focus of this study is a new alloy design strategy, to prepare FeAlCuCrNiMo_{0.6}Ti_x HEAs surfacing layer by gas metal arc welding (GMAW) technology, and to discuss the phase composition, microstructure of HEAs surfacing layer and the relationship between microstructure and comprehensive properties.

2 Experimental

2.1 Preparation of the Surfacing Layer

Metal powders of Al, Cu, Cr, Ni, Mo and Ti with purity level greater than 99.9 wt% were selected as raw materials and were added into the steel strip according to different ratios to roll and draw for a flux-cored wire of 2.4 mm in diameter. The powder filling rate is 35%. Alloys with nominal composition of FeAlCuCrNiMo_{0.6}Ti_x were prepared (*x* refers to molar ratio, and *x*=0.2, 0.6 and 0.8, and denoted as Ti_{0.2}, Ti_{0.6} and Ti_{0.8}, respectively). The compositions of the steel strip and the substrate are given in Table 1. The powders need to be dried before preparation of flux-cored wire, because they are easy to absorb water and cause defects, such as pores and cracks in the weld [18].

Subsequently, the surface of substrate low-carbon steel was polished to remove oxides and then cleaned with pure ethanol. The powers were surfaced on the low-carbon steel by GMAW (argon: 99.99%) to obtain FeAlCuCrNiMo_{0.6}Ti_x HEAs. The welding process parameters [19] are given in Table 2. The surfacing layers were cut into a square sample with a size of 10 mm×10 mm×10 mm, grinded and polished and then etched by 4% nitric acid to observe microstructure.

Table 2 Welding process parameters

Welding current (A)	Arc voltage (V)	Welding speed (cm/min)	Gas flow (L/min)
160	24	8	12

2.2 Surfacing Layer Characterization

Phase structure of surfacing layer was analyzed by X-ray diffraction (XRD), and working parameters were: current of 30 mA, voltage of 40 kV, Cu target, scanning range from 20° to 90° and scanning speed of 4°/min. The grain structure and phase composition were characterized by a GeminiSEM 300 field emission scanning electron microscope equipped with an electron back-scatter diffraction (EBSD) and an energy-dispersive spectroscopy (EDS). Vickers hardness was determined under a load of 500 g for 15 s with microhardness tester (THVS-5). Wet sand rubber wheel wear tester was used to test the wear resistance of surfacing layers, and the parameters were: quartz sand of 250 μm, speed 240 r/min and a duration of 5 min. BL410F electronic balance (1 mg) was used to compare the mass of surfacing layer samples before and after wearing test and then calculate the wear mass loss. Electrochemical polarization curves and electrochemical impedance spectroscopy of the HEAs surfacing layer in 3.5% NaCl solution were measured by a electrochemical workstation (VSP-300), using the saturated calomel electrode as reference electrode and platinum electrode as the counter electrode at scanning speed of 10 mV/min.

3 Results and Discussion

3.1 Phase Structure

According to thermodynamics, Zhang et al. [20] applied Hume–Rothery law to the composition design of multi-component HEAs. Based on the parameters of mixing entropy (ΔS_{mix}), mixing enthalpy (ΔH_{mix}) and atomic radius difference (δ) of multi-component alloys, the rules of solid-solution phase formation of HEAs are proposed as:

$$\Delta S_{\text{mix}} = -R \sum_{i=1}^n C_i \ln C_i, \quad (1)$$

Table 1 Compositions of H08A steel strip and substrate (wt%)

	C	Si	Mn	S	P	Fe
H08A	0.03	0.01	0.30–0.55	0.013	0.011	Bal
Substrate	0.20	0.35	0.70	0.045	0.045	Bal

$$\Delta H_{\text{mix}} = \sum_{i=1, i \neq j}^n 4\Delta H_{ij}^{\text{mix}} C_i C_j, \quad (2)$$

$$\delta = 100 \sqrt{\sum_{i=1}^n C_i \left(1 - \frac{r_i}{\bar{r}}\right)^2} = \sum_{i=1}^n r_i C_i, \quad (3)$$

where r_i is the atomic radius of the i th element, C_i , C_j is the molar ratio, and $\Delta H_{ij}^{\text{mix}}$ is the enthalpy of mixing between the i th and j th elements. The parameter ΔH_{mix} indicates the chemical compatibility among components, reflecting the tendency of formation for intermetallic compounds. According to the available data from different HEAs systems, the criteria for forming a random solid solution in HEAs were determined as $1\% < \delta < 6.6\%$, $-15 \text{ kJ}\cdot\text{mol}^{-1} < \Delta H_{\text{mix}} < 5 \text{ kJ}\cdot\text{mol}^{-1}$ and $12 \text{ J}\cdot\text{mol}^{-1} \text{ K}^{-1} < \Delta S_{\text{mix}} < 17.5 \text{ J}\cdot\text{mol}^{-1} \text{ K}^{-1}$. In the case that the alloy system can form a solid solution, the valence electron concentration (VEC) is an important parameter that affects whether the alloy forms a FCC phase or a BCC phase. Guo et al. [21] proposed a VEC to predict the phase stability of HEAs with the following formula:

$$\text{VEC} = \sum_{i=1}^n C_i (\text{VEC})_i, \quad (4)$$

where C_i is the atomic percentage and $(\text{VEC})_i$ is the VEC for the i th element, respectively. According to the formula, the parameter calculation result is given in Table 3. In Guo's study, when $\text{VEC} > 8.0$, it is beneficial to the formation of FCC phase; when $\text{VEC} < 6.87$, it is beneficial to the formation of BCC phase. When $6.87 < \text{VEC} < 8.0$, two phases of BCC and FCC tend to coexist. Figure 1 shows the parameter values of FeAlCuCrNiMo_{0.6}Ti_x HEAs. It can be seen that our design system can be defined as a HEA, and it is reasonable that two phases of BCC and FCC coexist. When the Ti_{1.0} is calculated, the system also coexists in two phases, so this work only studies the first three alloys.

Figure 2 shows the X-ray diffraction patterns of FeAlCuCrNiMo_{0.6}Ti_x ($x=0.2, 0.6, 0.8$) HEAs. The HEA surfacing layer is composed of a BCC phase and a small amount of MC phase, and the main phase of alloy is the BCC phase. It can be seen from the data of PDF cards that the

Fig. 1 Parameter values of FeAlCuCrNiMo_{0.6}Ti_x HEAs

phase peak of this phase is consistent with Fe–Cr, indicating that the FeAlCuCrNiMo_{0.6}Ti_x HEA forms a solid solution based on Fe–Cr phase during surfacing process. The small peaks appearing near the diffraction peaks $2\theta = 35.906^\circ$ and $2\theta = 41.710^\circ$ are ascribed to the FCC phase, shown in Fig. 2b, which is a magnified view of the dotted line in Fig. 2a; small amount of FCC phase is obtained as MC eutectic. The peak of MC phase is mainly due to TiC phase (PDF: 32-1383) [22]. With increasing Ti content, the MC phase always exists, and the diffraction peak does not shift. It is consistent with the previously calculated parameters for prediction, such as ΔS_{mix} , ΔH_{mix} , δ and VEC.

3.2 Microstructures

Figure 3 shows the microstructures and morphologies of FeAlCuCrNiMo_{0.6}Ti_x ($x=0.2, 0.6, 0.8$) HEAs. It is known that the surfacing layer is mainly composed of dendritic structure. Typical gray dendritic (DR) and white interdendritic (ID) structures were clearly observed in Ti_{0.2}, Ti_{0.6} and Ti_{0.8} HEAs (Fig. 3a–c). Furthermore, the close-up view showed that some fine particles precipitated in FeAlCuCrNiMo_{0.6}Ti_x HEAs. The chemical compositions of different regions obtained by EDS are listed in Table 4. In FeAlCuCrNiMo_{0.6}Ti_x HEA surfacing layer, DR was rich in Fe, Cr, Ni and depleted in Cu, while ID was rich in Cu, Mo, Ti and Al is partially burned. The precipitate is TiC phase. The amount of precipitated phase TiC gradually increases with the increase in Ti content. The reason is that Ti is a strong carbon compound forming element and a high volume fraction. However, the Fe content is higher, because the high-energy arc makes the heat absorption of surfacing layer increase, the $t_{8/5}$ of surfacing layer and substrate will become long, more Fe in the substrate diffuses to the surfacing layer, so the dilution is higher [23,

Table 3 Theoretical parameter values

Element	ΔS_{mix} ($\text{J}\cdot\text{mol}^{-1} \text{ K}^{-1}$)	Δ (%)	ΔH_{mix} ($\text{kJ}\cdot\text{mol}^{-1}$)	VEC
Ti _{0.2}	15.516	5.74	−5.65	7.31
Ti _{0.6}	15.991	6.16	−9.12	7.10
Ti _{0.8}	16.063	6.32	−10.52	7.00
Ti _{1.0}	16.075	6.44	−11.73	6.91

Fig. 2 X-ray diffraction patterns of FeAlCuCrNiMo_{0.6}Ti_x HEAs: **a** overall diagram, **b** FCC phase diagram

Fig. 3 Microstructures and morphologies of the FeAlCuCrNiMo_{0.6}Ti_x HEAs: **a** Ti_{0.2}, **b** Ti_{0.6}, **c** Ti_{0.8}

Table 4 Chemical compositions of FeAlCuCrNiMo_{0.6}Ti_x HEAs surfacing layer at various regions in Fig. 3 (at.%)

Alloys	Region	Fe	Al	Cu	Cr	Ni	Mo	Ti	C
Ti _{0.2}	Precipitates	25.99	7.35	2.70	1.56	2.35	0.69	48.29	11.07
	DR	70.11	2.39	5.83	9.35	9.60	1.12	1.60	–
	ID	63.92	3.30	10.63	7.24	7.47	4.08	55.73	–
Ti _{0.6}	Precipitates	14.17	7.82	2.57	1.72	2.21	0.79	3.36	14.99
	DR	69.57	2.24	4.21	10.32	9.56	1.56	2.54	–
	ID	57.30	3.37	13.02	8.12	7.69	4.14	6.36	–
Ti _{0.8}	Precipitates	10.92	8.53	1.65	1.74	1.34	0.47	62.47	12.86
	DR	67.11	2.88	6.42	10.33	9.13	1.43	2.70	–
	ID	62.01	3.16	9.15	8.06	7.49	3.98	6.15	–

24]. Combined with XRD analysis, the microstructure is mainly Fe–Cr phase with BCC structure, and the fine particles of TiC are usually dispersed in alloy matrix [22].

Figure 4 shows the EDS maps of FeAlCuCrNiMo_{0.6}Ti_{0.8} HEA. It can be seen that Ti element rich in white bulk particles substantially coincides with C. From this, it was confirmed that the white particles were TiC phase. The distribution of TiC can only be judged by the content of Ti, because there is always a large error in the content of C obtained by EDS [23].

Figure 5 shows the EBSD texture maps and the grain size of FeAlCuCrNiMo_{0.6}Ti_x ($x = 0.2, 0.6, 0.8$) HEAs. These images exhibit the colors corresponding various grain orientations, as indicated by the unit triangles in Fig. 5a. It shows that the microstructure consists of many blocky, elongated and small amounts of equiaxed grains [23]. After calculating the average grain size, the value is 3.27 μm for Ti_{0.2}, 3.26 μm for Ti_{0.6} and 3.19 μm for Ti_{0.8}. With increasing amount of Ti, the grain size decreased, the texture orientation distribution was uneven, and the grain did not have preferred growth orientation. For Ti_{0.8}, the ratio of grains size (the area fraction) of less than 3.0 μm increased from 74.5% to 78%, as shown in Fig. 5b, d and f. In summary, the addition of Ti is advantageous for grains refinement.

Figure 6 shows the EBSD phase maps and histograms of the phase of FeAlCuCrNiMo_{0.6}Ti_x ($x = 0.2, 0.6, 0.8$) HEAs. There is a majority of BCC, few FCC phase and an unknown phase, the red region represents the BCC phase, and the blue and yellow-green regions represent FCC phases. As can be seen in Fig. 6a, c and e, FCC phases are mainly dispersed in BCC grains, and the proportion increases gradually. EBSD results are consistent with the XRD results, and the surfacing layers are mainly composed of solid solutions including FCC and BCC phases. The BCC phase occupies most of the area. In Fig. 6b, d and f, due to the addition of Ti, the FCC phase ratio increases, the FCC phase in Ti_{0.8} is 0.96%, and the ratio of BCC to FCC phase is 101.9.

3.3 Microhardness and Wear Resistance

Figure 7 shows the microhardness profiles along the cross section of the FeAlCuCrNiMo_{0.6}Ti_x ($x = 0.2, 0.6, 0.8$) HEAs. With increasing Ti content, the normal direction microhardness of HEAs surfacing layer increases gradually. The highest microhardness of HEAs surfacing layer can reach 642 HV_{0.5}, which was at least 2.8 times that of substrate, and the thickness of surfacing layer is about 3.0 mm. In addition, the microhardness of fusion zone is lower than that of upper surface. Figure 8 shows the cross-sectional microstructure of FeAlCuCrNiMo_{0.6}Ti_{0.8} HEA. The surfacing layer is mainly composed of solid solutions including FCC and BCC phases, where it contains many elements. These alloy elements enhance the microhardness of the surfacing layer by solid-solution strengthening and precipitation strengthening. The fusion zone is located in the transition zone between the surfacing layer and the substrate metal. The fusion zone experienced obvious element diffusion and migration due to the action of welding heat. The composition of fusion zone should be a mixture of substrate and surfacing layer, and the microhardness of fusion zone is also between them. Therefore, the microhardness of the surfacing layer is the highest, and that of the fusion zone is lower.

Figure 9 shows the comparison of average microhardness and wear mass loss of FeAlCuCrNiMo_{0.6}Ti_x ($x = 0.2, 0.6, 0.8$) HEAs. Intuitively, the microhardness curve of surfacing layer is increasing, and the average microhardness of Ti_{0.8} is the highest (626 HV_{0.5}); The wear mass loss curve showed a decreasing trend, and the amount of wear mass loss of Ti_{0.8} was the smallest (0.28 g). It is suggested that the smaller the crystal grains of material, the higher the microhardness, and the better the wear resistance of alloy. The average microhardness of substrate is 229 HV_{0.5}, which is lower than the microhardness of HEAs surfacing layer. Ti is a strong carbide forming element, the atomic radius is relatively large, lattice distortion occurs, and the resistance of dislocation motion is enhanced, so that the microhardness of alloy is much higher than that of substrate [25, 26].

Fig. 4 Qualitative element distribution EDS maps of FeAlCuCrNiMo_{0.6}Ti_{0.8} HEAs

The Fe–Cr solid solution of BCC structures in alloy inflicts solid-solution strengthening, and the fine particle TiC precipitation strengthening effect of FCC phase improves the microhardness of alloy [27–29]. Comparing the atomic radius of each element with Fe–Cr lattice gap, the following relationship exists:

$$R(\text{Ti}) > R(\text{Al}) > R(\text{Cu}) > R(\text{Ni}) > R(\text{Cr}) > R(\text{Fe}) > \text{Lattice gap}$$

Since the Fe and Cr atomic radius difference and the electronegativity are smaller than those of other atoms, it is impossible for metal atom to exist in the unit cell as an interstitial solid solution, and it is most likely that it exists in the surfacing layer in a form of substitutional solid solution. Compared with other elements, the atomic radius of Ti element is relatively large. The addition of

Ti element may produce a substitutional solid solution. As the Ti content increases, the entropy value increases, the Gibbs free energy decreases, the degree of alloy chaos increases, and element segregation is reduced, the phase is more stable [30]. According to the above analysis, there are two strengthening mechanisms in the test alloy system, namely solid-solution strengthening and precipitation strengthening, and the two strengthening effects enhance the mechanical properties of alloy as well.

3.4 Corrosion Resistance

Figure 10 shows the electrochemical polarization curves in 3.5 wt% NaCl solution for 304 stainless steel and FeAlCuCrNiMo_{0.6}Ti_x ($x = 0.2, 0.6, 0.8$) HEAs. The self-corrosion potential and self-corrosion current density of

Fig. 5 EBSD texture maps and grain sizes of FeAlCuCrNiMo_{0.6}Ti_x HEAs: **a, b** Ti_{0.2}, **c, d** Ti_{0.6}, **e, f** Ti_{0.8}

FeAlCuCrNiMo_{0.6}Ti_x HEAs surfacing layer with different Ti contents are not consistent with each other. With increasing Ti content, the self-corrosion potential firstly increases and then decreases, and the self-corrosion current density decreases at first and then increases. The self-corrosion potential is an important parameter of measuring the thermodynamics of corrosion reaction. The higher the potential, the more difficult the corrosion reaction begins. Self-corrosion current density is a physical parameter that

reflects the kinetics of reaction. The smaller the current density, the slower the corrosion rate [31, 32]. Since the corrosion rate is positively correlated with the self-corrosion current, the corrosion rate of HEAs also decreases at first and then increases. The data of self-corrosion potential and self-corrosion current density are given in Table 5. It can be concluded that FeAlCuCrNiMo_{0.6}Ti_{0.6} has best corrosion resistance, which possesses the highest self-corrosion potential, about -1.03409 V, and the lowest self-corrosion current

Fig. 6 EBSD phase maps and histograms of phases within $\text{FeAlCuCrNiMo}_{0.6}\text{Ti}_x$ HEAs: **a, b** $\text{Ti}_{0.2}$, **c, d** $\text{Ti}_{0.6}$, **e, f** $\text{Ti}_{0.8}$

density, about $2.1276 \times 10^{-4} \text{ mA/cm}^2$. Compared with 304 stainless steel, HEAs surfacing layer has better corrosion resistance, smaller self-corrosion current density and higher corrosion potential.

Figure 11 shows Nyquist plots of 304 stainless steel and $\text{FeAlCuCrNiMo}_{0.6}\text{Ti}_x$ ($x = 0.2, 0.6, 0.8$) HEAs in 3.5 wt% NaCl solution. The impedance spectra are expressed in the high-frequency region, when the capacitive reactance

arcs form parallel connections at the charge transfer resistance. During the corrosion process, the passivation film is continuously damaged by the Cl^- -induced corrosion in solution, the impedance gradually reduces, and the arc gradually shrinks. In addition, in this process, an inductive arc appears in the low-frequency region, which is related to the passivation film and Cl^- -induced corrosion [33, 34]. The smoother the amplitude–frequency curve,

Fig. 7 Microhardness distributions of FeAlCuCrNiMo_{0.6}Ti_x HEAs

Fig. 10 Electrochemical polarization curves of FeAlCuCrNiMo_{0.6}Ti_x HEAs

Fig. 8 Cross-sectional microstructure of FeAlCuCrNiMo_{0.6}Ti_{0.8} HEA

Fig. 9 Comparison of microhardness and wear mass loss of FeAlCuCrNiMo_{0.6}Ti_x HEAs

Table 5 Self-corrosion potential and current density of FeAlCuCrNiMo_{0.6}Ti_x HEAs surfacing layer

HEA surfacing layer	Self-corrosion potential (V)	Self-corrosion current density (mA/cm ²)
304 Stainless steel	-0.9008	2.3549×10^{-3}
Ti _{0.2}	-0.9437	1.9474×10^{-3}
Ti _{0.6}	-1.03409	2.1276×10^{-4}
Ti _{0.8}	-1.03038	3.9298×10^{-4}

Fig. 11 Nyquist plots of FeAlCuCrNiMo_{0.6}Ti_x HEAs

Fig. 12 Corrosion morphologies of FeAlCuCrNiMo_{0.6}Ti_x HEAs: **a** FeAlCuCrNiMo_{0.6}Ti_{0.6}, **b** 304 stainless steel

the smaller the corrosion variation, indicating that the corrosion resistance is worse; the larger the radius of the capacitive arc, the better the corrosion resistance [35]. It can be seen from Fig. 11 that the corrosion rate decreases at first and then increases, and the Ti_{0.6} has the largest radius of capacitance and the best corrosion resistance. This is because the BCC structure is dominant in Ti_{0.6}, and the BCC structure is mainly related to Fe–Cr phase, which contains a large amount of Cr element, and the passivation film can be more easily formed [36]. On the other hand, Ti_{0.6} has the lowest self-corrosion current density and the highest charge transfer resistance. Therefore, the capacitance radius of Ti_{0.6} is much larger than other components, and the corrosion resistance is the best.

Figure 12 shows the corrosion morphologies of FeAlCuCrNiMo_{0.6}Ti_x ($x = 0.2, 0.6, 0.8$) HEAs. The difference in morphologies is obvious by comparing Fig. 12a and b; the corrosion resistance of FeAlCuCrNiMo_{0.6}Ti_x HEAs is stronger than that of 304 stainless steel, which is mainly due to the single solid-solution phase, and the improved corrosion resistance of alloy. The microstructure corrosion of FeAlCuCrNiMo_{0.6}Ti_x HEAs occurs mostly within the interdendritic region. From the comparison of EDS components of DR and ID in Table 4, it is known that the dendritic region is less Cu and the interdendritic region is rich in Cu forming a galvanic couple, resulting in the intergranular zone to be firstly corroded [37]. The less phase of corroded material during electrochemical corrosion process guarantees more uniform structure and provides more favorable corrosion resistance of alloy [38]. Cr and Ni elements are easily passivated in oxidative medium, which is beneficial to the formation of passivation film on the surface of alloy, thereby preventing the corrosion from proceeding further. On the other hand, the single solid-solution phase improves the corrosion resistance of alloy [39].

4 Conclusions

1. FeAlCuCrNiMo_{0.6}Ti_x HEAs surfacing layer was prepared on the surface of low-carbon steel by gas metal arc welding technology. The microstructure of alloy system is dominated by a body-centered cubic Fe–Cr solid-solution phase with a small amount of carbide (TiC).
2. The microstructure of surfacing alloy is a typically dendritic structure. With the addition of titanium, the grain size becomes smaller and the grain boundaries increase. The alloy system shows a combination of solid-solution strengthening and precipitation strengthening effects.
3. The addition of titanium can significantly improve the microhardness and wear resistance of surfacing layer. FeAlCuCrNiMo_{0.6}Ti_{0.8} has a microhardness of up to 642 HV_{0.5} and a wear mass loss of 0.28 g. The addition of titanium brings about more significant corrosion resistance than 304 stainless steel, and the appropriate amount of Ti content for corrosion resistance can be optimized, and FeAlCuCrNiMo_{0.6}Ti_{0.6} has the best electrochemical corrosion resistance.

Acknowledgements This work was financially supported by the National Key R&D Program of China (No. 2017YFB1103603) and the Natural Science Foundation Program Key Projects of Liaoning Province (No. 20180510030).

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