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Improvement in microstructure and wear-resistance of high chromium cast iron/medium carbon steel bimetal with high vanadium

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Abstract

In the present study, a high chromium cast iron (HCCI) with high content of vanadium (15 wt.%) as the wear resistance layer was fabricated on the medium carbon steel (MCS) substrate by laser cladding process. The effect of high vanadium addition on microstructure, phase, hardness and wear properties of bimetal were investigated. The results show that the microstructure of HCCI layer was significantly improved and refined with addition of high vanadium. No defect such as microcracks and unbonded regions was observed on the interface of the bimetal. The x-ray diffraction revealed that the HCCI (15 wt.\% V) layer consists of \(\alpha\)-Fe matrix, \((\text{Cr, Fe})_7\text{C}_3\) and VC carbides. The SEM images shown that round-like VC particles uniformly distributed on the \(\alpha\)-Fe matrix and \((\text{Cr, Fe})_7\text{C}_3\) are of rod-like with distributed on grain boundary, such micromorphology makes it have higher hardness (883HV) and excellent wear resistance than HCCI layer without vanadium addition.

1. Introduction

High chromium cast iron (HCCI), containing high-volume \(\text{M}_7\text{C}_3\)-type (M: Fe and Cr) hard carbide formed at a certain Cr/C ratio, are widely used in machinery, metallurgy, mining fields for their good properties of hardness and wear resistance [1–4]. However, the severe wear caused by coarse primary carbide restricts their further utilization [5–7]. Moreover, \(\text{M}_7\text{C}_3\) carbide as the main wear-resistant phase shows the disadvantages of insufficient hardness and wear-resistant performance under long-term wear conditions. Therefore, improving the hardness and wear resistance of high chromium cast iron is of great significance to prolonging service life and saving material cost. Recently, a bimetal-materials consisting of high chromium cast iron as wear-resistance part and carbon steel as ductile part have been studied [8–11]. Yin et al [9] investigated the effect of Ti on the microstructure and properties of high chromium iron-based coating, the \((\text{Cr, Fe})_7\text{C}_3\) carbides have been refined significantly and the hardness was improved to 897HV. Li et al [10] reported that C atoms diffuse from low carbon steel side to high chromium cast iron side, which indicated uphill diffusion happened.

The wear resistance of high chromium cast iron is closely related to the microstructure of metal matrix and carbides, such as the type, morphology, distribution and volume fraction of carbides [12–14]. Therefore, these microstructure factors should be controlled and improved, some researchers have reported that addition of vanadium element into cast iron can refine carbides and improve morphology, distribution of carbides. Qi et al [15] researched that the effects of vanadium on microstructure and wear-resistance of high chromium cast iron hardfacing metal, the results shown that the size of primary carbides were significantly refined and the weight loss decreased. Wei et al [16] reported that the large amount of and uniformly distributed spherical vanadium carbides can be obtained with addition of 8.15–10.20 wt.\% vanadium. Wang et al [17] studied that vanadium could replace a part of chromium in \(\text{M}_7\text{C}_3\) carbides and their hardness and wear-resistance were obviously improved by increasing vanadium to 1.50 wt.\%.
It is well known that vanadium carbide has higher hardness than M$_7$C$_3$ carbide which making it suitable for the reinforced phase of wear-resistant materials, but there are few reports on adding high contents of vanadium into high chromium cast iron to form more vanadium carbides. In the present research, the bimetal materials that HCCI with high vanadium content (15 wt.%) as wear-resistance layer and carbon steel as substrate are prepared by laser cladding process. The effects of high vanadium on microstructure, phase, mechanical and wear properties of the bimetal materials were studied carefully by SEM, XRD, hardness tester and friction and wear tester.

### 2. Experimental procedures

The raw powder of HCCI with high vanadium content (15 wt.%) were prepared through atomization process by using high-carbon ferrochrome, ferrovanadium, molybdenum and gray cast iron as raw materials. The particle size of raw powders was measured by a laser particle size analyzer (HELOS-RODOS/M, Germany), it was revealed that their average size was about 97 μm. For the research, the atomized powders of HCCI and HCCI (15 wt.% V) were cladding on the medium carbon steel substrate by laser cladding, and its chemical compositions

![Figure 1](image1.png)  
**Figure 1.** Microstructure of aerosolized high chromium cast iron (15 wt.% V) powders: (a) SEM morphology of atomized powders, (b)–(c) microstructure of single atomized powders (backscattered electron image).

| Material | Fe   | Cr   | C    | V    | Mo | Mn | Si |
|----------|------|------|------|------|----|----|----|
| HCCI (V) | Bal. | 16.6 | 4.2  | 15.3 | 1.1 | 0.8 | 0.9 |
| HCCI     | Bal. | 17.3 | 4.1  | —    | 0.7 | 0.7 | 1.0 |
| MCS      | Bal. | 0.1  | 0.5  | —    | —  | 0.6 | 0.2 |

![Figure 2](image2.png)  
**Figure 2.** XRD patterns of atomized powders of HCCI (15 wt.% V), carbon steel substrate, HCCI and HCCI (15 wt.% V) layer.
was shown in table 1. Prior to cladding, the substrate surface was cleaned with acetone and then was grit-blasted. The bimetal was processed into a cuboid \((10 \, \text{mm} \times 10 \, \text{mm} \times 15 \, \text{mm})\) by wire cutting machine. The cuboid samples were prepared by grinded from 240 to 1500 grit paper and metallographically polished with 0.5 \(\mu\text{m}\) diamond polishing paste.

The phase structures of bimetal were analyzed by x-ray diffraction \((\text{XRD, DX-2700, China})\) using Cu K\(\alpha\) radiation at a scanning rate of 0.06 \(^\circ\) s\(^{-1}\) in the range of \(2\theta = 30^\circ – 90^\circ\). The microstructures of the raw materials and bimetal were operated by scanning electron microscope \((\text{SEM JSM-6490LV, JEOL, Japan})\) equipped with an energy dispersive spectroscopy \((\text{EDS})\). The hardness of the polished cross-sections was measured by 452SVA Wolpert Wilson Instruments™ at a load of 49 N with a dwelling time of 15 s and each result was the average of more than five measurements. The wear-resistance was carried out by a reciprocating friction wear testing machine \((\text{MFT-5000, American})\) under a load of 10 N for 60 min, and the weight loss of wear was calculated every 10 min to evaluate the wear-resistance of the bimetal.

### 3. Results and discussions

#### 3.1. Microstructure of raw material powders

Figure 1 shows the microstructure of atomized raw powders of HCCI with high vanadium content, the powders were spherical as shown in figure 1(a) and the microstructure of single powder was revealed in figures 1(b)–(c) by SEM with backscattered electron modes. There were many lumps-like VC black phases \((\text{figure 1}(b))\) evenly distributed on the gray matrix phase, the volume fraction and average size of VC carbide was about 27%, 0.99 \(\mu\text{m}\) according to the analysis of Image pro plus software. Furthermore, figure 1(c) and the EDS results \((\text{see table 2})\) indicate that the eutectic \(\text{M}_7\text{C}_3\) distributed at the gray phase grain boundary.

#### 3.2. Phase and microstructure of bimetal material

The x-ray diffraction patterns of atomized powders of HCCI \((15 \, \text{wt.}\% \, \text{V})\), carbon steel substrate, HCCI and HCCI \((15 \, \text{wt.}\% \, \text{V})\) cladding layer are shown in figure 2. There was no obvious difference between the phases of HCCI \((15 \, \text{wt.}\% \, \text{V})\) layer and atomized powders, which were \(\alpha\)-Fe, VC and \((\text{Cr, Fe})_7\text{C}_3\). The main phase of the HCCI layer was composed of \((\text{Cr, Fe})_7\text{C}_3\) and \(\alpha\)-Fe. For the carbon steel substrate, only ferrite phase existed.

The microstructure of HCCI layer and HCCI \((15 \, \text{wt.}\% \, \text{V})\) \((\text{figure 3}(b))\) were composed of columnar gray phase and acicular black-gray phase, the EDS results \((\text{table 2})\) indicate that the eutectic \(\text{M}_7\text{C}_3\) distributed at the gray phase grain boundary.

### Table 2. EDS results of the point 1 shown in figure 1(c) (at.%).

| No. | Fe  | Cr  | C   | V   |
|-----|-----|-----|-----|-----|
| 1   | 54.6| 11.0| 27.4| 7.0 |

Figure 3. SEM micrograph (BSE) of the layer: (a) HCCI layer, (b) HCCI \((15 \, \text{wt.}\% \, \text{V})\) layer.
size was refined significantly. The reason is that VC precipitates firstly generated in the liquid phase, and the (Cr, Fe)\textsubscript{7}C\textsubscript{3} and Fe–Cr solid solutions nucleate with VC particles as core and grow rapidly. Furthermore, the (Cr, Fe)\textsubscript{7}C\textsubscript{3} carbides were distributed in the grain boundary in isolation (table 3), which weakens the cleavage effect on the matrix.

Figure 4 shows the SEM micrograph (BSE) of the HCCI/MCS and HCCI (15 wt.% V)/MCS joint interface. The left side indicates the MCS layer, and the right side does the HCCI layer, as shown in figures 4(a)–(b). In figure 4(a), the HCCI layer is composed of cellular crystals approaches to interface and columnar crystals growing perpendicular to the interface. The formation of cellular crystals was caused by the low solidification rate and large temperature gradient of the HCCI layer near the interface, which made it difficult to grains grow. The increase of HCCI layer temperature away from interface increased the solidification rate and reduced the temperature gradient, which causes the morphology of the HCCI layer to gradually change from cellular crystal to columnar crystal, and the columnar crystal was basically arranged along the direction of maximum heat dissipation. The HCCI layer was characterized by cellular crystals and acicular (Cr, Fe)\textsubscript{7}C\textsubscript{3} carbide clusters distributed on the grain boundaries as shown in figure 4(a1). The bonding interface of HCCI/MCS was almost a wavy line, and a diffusion zone about 8 \( \mu \text{m} \) existed between HCCI layer and interface, which indicates the bimetal achieved metallurgical bonding with a good condition of joint bonding [18]. The HCCI (15 wt.% V) layer consists of black VC particles, (Cr, Fe)\textsubscript{7}C\textsubscript{3} carbide and ferrite as shown in figures 4(b)–(b1). The spherical VC particles were uniform embedded in the ferrite matrix and (Cr, Fe)\textsubscript{7}C\textsubscript{3} carbide. It is obvious that the size of

Table 3. EDS results of the points shown in figure 3 (at.%).

| No. | Fe   | Cr   | C    | V    | Mn  |
|-----|------|------|------|------|-----|
| 1   | 71.1 | 17.8 | 11.1 | —    | —   |
| 2   | 34.2 | 29.1 | 36.7 | —    | —   |
| 3   | 75.5 | 13.5 | 7.7  | 2.5  | 0.8 |
| 4   | 24.2 | 25.0 | 37.5 | 12.4 | 0.9 |
| 5   | —    | —    | 49.6 | 50.4 | —   |
eutectic microstructure was evident refined and no coarse columnar crystal was found in the layer. In the solidification process, VC particles precipitated from the liquid phase could be used as heterogeneous nucleation points of carbide and ferrite [19], which greatly increased the nucleation rate and made the grain size significantly refined.

Figure 5 is the EDS map of the bonding interface of HCCI (15 wt.% V)/MCS, figures 5(c)–(f) shows the corresponding distribution of Fe, Cr, V and C element, respectively. The left side is the MCS layer, and the right side is the HCCI (15 wt.% V) layer. As shown in figures 5(c)–(d), on the HCCI (15 wt.% V) layer, Fe and Cr element concentrated in ferrite, (Cr, Fe), C₃ and diffusion zone. The distribution of V element is shown in figure 5(e), it revealed that the V element mainly concentrated in VC and diffusion zone. In figure 5(f), C element is detected on the MCS side, which indicates C atoms moves to the MCS side across the interface because of its small atomic radius, thus a carbide-free zone approach to interface on the HCCI (15 wt.% V) side formed in
According to the Kirkendall effect, atom moves from a region of high concentration to a region of low concentration during the diffusion process [10]. In current study, diffusion of V, Cr and C occurred apparently toward the MCS side because of their higher content in the HCCL (15 wt.% V) layer. The V and C elements moved to the diffusion zone to form VC particles. The low temperature and fast cooling speed close to the interface improved the nucleation rate of VC resulting in suppressing grain growth, and the acicular morphology of VC is formed due to the effect of dilution of MCS near the interface.

3.3. Hardness and wear properties of bimetal material

The Vickers hardness taken across the interface of the HCCI/MCS and HCCI (15 wt.% V)/MCS bimetal as a function of the distance are plotted in figure 6. Each hardness value was the average of five measurements. The Vickers hardness changes in three stages with the distance from the interface, corresponding to structure of matrix, interface and cladding layer from low to high, respectively. On the substrate side, the hardness of HCCI/
MCS and HCCI (15 wt.% V)/MCS bimetal both exhibited a lower value (~260 HV). On the HCCI layer, owing to the existence of rod-like (Cr, Fe)$_7$C$_3$ hard phases, the values of Vickers hardness significantly increased to ~657HV. Moreover, the HCCI (15 wt.% V) layer showed the highest hardness (883HV), which could be attributed to the round-like VC carbides with higher hardness were embedded on the ferrite matrix.

The evolution of the friction coefficient for the HCCI and HCCI (15 wt.% V) layer presented in figure 7 consists of running-in stage and stable stage. The friction coefficient of HCCI and HCCI (15 wt.% V) layer rapidly increased during a running-in stage of about 100 s, which was mainly due to the increase of contact area.
between the sample and the friction pair. In the stable stage, with the contact area tended to stable, the friction coefficient fluctuated within the range of about 0.08 and 0.14, respectively. HCCI (15% wt.% V) layer showed a higher friction coefficient and fluctuation than HCCI layer, the average friction coefficient (calculated from 100 s during the stable wear stage) of HCCI and HCCI (15% wt.% V) layer was about 0.703 and 0.781, which indicated that the resistance of the cladding layer to the movement of the friction pair increased with the addition of V element.

The three-dimensional worn surface images of HCCI layer and HCCI (15 wt.% V) layer are presented in figure 8. The surface of HCCI layer (figure 8(a)) became rough obviously with a maximum wear depth of 1.8 μm, while the HCCI (15 wt.% V) layer (figure 8(b)) was rather flat with a maximum wear depth of 1.1 μm, indicating much shallower wear scar than HCCI. There are some furrows of HCCI layer parallel to the wear direction in the wear surface and significant plastic deformation area was generated at the edge of wear scar. The ploughing effect of the friction pair on the HCCI (15 wt.% V) surface was significantly weakened with the addition of vanadium, the deep furrow and plastic deformation area on the worn surface disappear, which is attributed to the strengthening effect of VC on the matrix.

Figure 9 gives the specific weight loss as a function of wearing time in the wear test of the HCCI and HCCI (15 wt.% V) coating. The weight loss of HCCI layer was approximately linear with wearing time, and HCCI (15 wt.% V) layer showed flatbeds. Compared with the cladding layer of HCCI, the loss of weight of the cladding layer of HCCI (15 wt.% V) was reduced by about 10 times, which indicated that the wear-resistance of HCCI was improved significantly with the addition high vanadium element. The (Cr, Fe)7C3, as a wear-resistant phase of HCCI layer, was mainly distributed at the grain boundary of α-Fe, while in HCCI (15 wt.% V) layer, the wear-resistant phases were (Cr, Fe)7C3 and higher hardness of VC particles, moreover, VC particles were embedded in the α-Fe matrix could protect the soft matrix from peeling off, thus greatly reducing the weight loss of HCCI (15 wt.% V) layer.

The wear surfaces of HCCI and HCCI (15 wt.% V) layer were carefully examined under SEM are shown in figure 10. The wear tracks of HCCI and HCCI (15 wt.% V) layer were shown in the pictures having grooves parallel to the sliding direction, which indicated that the dominant wear mechanism was abrasive wear. In accordance with the result of the weight loss measured previously, it can be seen clearly from the figures that the wear-resistance of HCCI (15 wt.% V) layer was better than HCCI layer. The grooves caused by hard carbides plowed into soft ferrite matrix were shallower in HCCI (15 wt.% V) layer than HCCI layer, which was mainly due to the VC hard particles embedded in ferrite played the role of wear-resistant skeleton to prevent the soft ferrite matrix from being worn during the wear process.

4. Conclusions

The high vanadium (15 wt.%) HCCI/MCS bimetal having desirable microstructure and excellent mechanical properties was fabricated by laser cladding. Compared with HCCI/MCS bimetal, the spherical VC hard particles with higher hardness replaced the rod-like (Cr, Fe)7C3 carbides and distributed uniformly on the ferrite matrix, which significantly improved the hardness of bimetal (~200HV). Furthermore, with lower weight loss and shallower grooves, the high vanadium (15.3 wt.%) HCCI/MCS bimetal exhibited better wear resistance than HCCI/MCS bimetal.
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Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

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