Study of the wear resistance of Hypereutectic Fe–Cr–C Hardfacing Alloy reinforced with carbide particles

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Abstract
The microstructures and wear resistance of hypereutectic Fe–Cr–C hardfacing alloys reinforced with carbide particles was studied. The results showed that when the sizes of the M7C3 carbides were small, the wear resistance of the hardfacing alloy was considerably improved. During the wear process, the uniformly distributed small-sized carbides could also provide good support, and the uniform microstructure containing the fine M7C3 carbides could effectively reduce the force of the abrasive particles on the surface, evenly distribute the load, reduce the micro-cutting effect of the abrasive particles, and significantly improve the wear resistance of the hardfacing alloy. When the average size of the primary M7C3 carbides in the hardfacing alloy reached 0.5 μm, since the total volume fraction of the carbides of different sizes in the hardfacing alloy were basically the same, the reduction of the carbide size could significantly increase the phase interface of the M7C3 carbide and austenite. This could promote the occurrence of the M7C3 → M23C6 transformation, forming a two-phase composite structure with hard-core M23C6 carbide and soft-shell M23C6 carbide. This structure could reduce the interfacial stress between the M7C3 carbide and austenite and improve the spalling resistance of the carbides in the hardfacing alloy.

1. Introduction

Hypereutectic Fe–Cr–C (Fe-based) alloys have high hardness and wear resistance [1–3], and they can be clad on the surfaces of mechanical parts by arc hardfacing to prepare a wear-resistant surface [4]. These alloys have wide range of applications in mining, metallurgy, cement, and other industries.

The wear resistance of hypereutectic Fe–Cr–C alloys mainly depends on the high-hardness strengthening phase, M7C3 carbide, in its microstructure [5–8]. Studies have found that the use of optimized alloy components [9–12] can refine the primary M7C3 carbides and improve the wear resistance of hypereutectic Fe–Cr–C hardfacing alloys to a certain extent. Jiang et al [13] found that TiC can enhance the strength and wear resistance of Fe–Cr–C alloys. Under the action of an in situ-formed strengthening phase of TiC, the sizes of M7C3 carbides were greatly reduced, the wear resistance of the alloy was considerably improved, and its mechanical properties were significantly enhanced. Mourlas et al [14, 15] prepared Fe–Cr–C wear-resistant coatings on the surfaces of low-carbon steel specimens. After Ti and W alloy elements were added, the M7C3 strengthening phase that formed in the coating exhibited a better microstructure in subsequent dry friction tests, and the alloy showed higher wear resistance. However, the microscopic friction and wear mechanism of its strengthening phase remained to be further studied. Liu et al [16, 17] found that in Fe–Cr–C alloys, the microscopic morphologies of M23C6 carbides could be improved by forming in situ TiN and NbC. In addition, Nb, as a solid solution element,
could improve the properties of the M7C3 carbides. At the same time, TiN and NbC together with M7C3 carbides could be used as strengthening phases to enhance the wear resistance of Fe–Cr–C alloys. Zhou et al. [18] also found that when a hypereutectic Fe–Cr–C hardfacing flux–cored wire was prepared, the addition of Ti alloys could refine the primary M7C3 carbides in the hardfacing alloy and improve the uniformity of the carbide distribution. Imurai et al. [19, 20] found that the addition of different amounts of W to hypereutectic Fe–Cr–C alloys could form WC and improve the sizes and morphologies of the M7C3 carbides, which significantly improved the structure and wear resistance of the hardfacing alloy.

Studies have found that improving the structure of M7C3 carbide can also enhance the wear resistance of hypereutectic Fe–Cr–C hardfacing alloys [21, 22]. Cui et al. [23] found that when the C content in the hypereutectic Fe–Cr–C alloy was 5 wt.% and the Cr content was 20–30 wt.%, the M7C3 carbide that formed in the alloy had the highest hardness. The size and morphology distribution of the carbide were relatively uniform, and the alloy had good wear resistance. Frisk et al. [24] found that alloying element V could shift the eutectic point of the hypereutectic Fe–Cr–C alloys to the left and be solutionized in the M7C3 carbide to improve its stability. Wang et al. [25] found that the primary M7C3 carbide underwent varying degrees of deformation at 450 °C. The carbide underwent lattice distortion and shear–torsion deformation, and a large number of dislocations and deformation twins were generated inside the carbide. After deformation, the hardness of the M7C3 carbide increased to a certain extent, and the phenomenon of work hardening occurred. Based on the research results described above, we speculate as to whether improving the structure of the M7C3 carbides by refining the primary M7C3 carbides in the hypereutectic Fe–Cr–C hardfacing alloy can reduce the interfacial stress between the primary M7C3 carbide and the substrate, thereby improving the anti-spalling performance of the alloy while further improving its wear resistance.

M7C3 carbide is a metastable phase [26]. It has been an active research area to determine whether M7C3 carbide can be transformed under suitable conditions. Wang et al. [27] found that in high-performance microalloys, the M7C3 carbides were partially transformed to M23C6 carbides when the temperature was held at 1000 °C for different times. The formed M23C6 carbide and γ-Fe had a good crystallographic relationship. Wiengmoon et al. [28] found that the secondary carbides M23C6 were attached to the prior austenite dendrites of the eutectic substrate, and M23C6 had a good crystal coherent relationship with the austenite substrate. The precipitation of secondary carbides M23C6 improved the properties of hypoeutectic cast iron to a certain extent. The above research results showed that metastable M7C3 carbides could be transformed into more stable M23C6 carbides. The hardness of the M23C6 carbides, which formed between M7C3 carbide and austenite, was significantly lower than that of the M7C3 carbides.

Therefore, according to the previous research of the microstructure of hypereutectic Fe–25Cr–4C–0.5Ti–0.5Nb–0.2N–2LaAlO3 hardfacing alloy, the wear properties and mechanism of reinforced with ultrafine M7C3 carbide particles in this paper were investigated.

2. Experimental materials and methods

2.1. Experimental materials

In this paper, a coupled rolling-drawing process was used to prepare flux–cored welding wire. The outer sheath of the welding wire was an H08A steel strip, and the flux components were mainly composed of chromium–iron carbide, chromium–iron nitride, electrolytic manganese, silicon carbide, graphite, ferrotitanium, ferro niobium, sodium fluoride, and magnesium powder. The composition of the hardfacing alloy was selected based on that of the Fe–25Cr–4C hardfacing alloy. Ti, Nb, 0.2 wt.% N, and 2.0 wt.% LaAlO3 were added to optimize the composition of the Fe–25Cr–4C alloy, and the Fe–25Cr–4C–0.5Ti–0.5Nb–0.2N–2.0LaAlO3 hardfacing alloy was obtained. The detailed composition of the based and prepared hardfacing alloy is shown in table 1.

2.2. Experimental methods

After the welding wire was prepared, a Q235 substrate was used to prepare a hardfacing alloy sample. The chemical composition of the substrate is shown in table 2, and the size was 150 mm × 100 mm × 20 mm. The

| Table 1. Composition analysis of hardfacing alloys (wt.%). |
|---|---|---|---|---|---|---|---|---|
| Number | C | N | Cr | Mn | Si | Ti | Nb | Fe LaAlO3 |
| Base alloy | 4.0 | — | 25 | 2.0 | 1.0 | — | — | Bal. | — |
| No.1 | 4.16 | 0.05 | 25.32 | 2.06 | 1.01 | — | — | Bal. | — |
| No.2 | 3.95 | 0.20 | 25.59 | 2.21 | 1.06 | 0.59 | 0.61 | Bal. | 2.0 wt.% |

Wang et al. [18] and Zhou et al. [19, 20] found that the secondary carbides M23C6 formed by the transformation of M7C3 carbides can re
The polished sample was corroded using an etchant containing 5 g of FeCl₃, 10 ml of HNO₃, 3.0 ml of HCL, and 87 ml of ethyl alcohol. Field-emission scanning electron microscopy (FESEM, Hitachi S4800) was used to observe the microstructure and morphology of the hardfacing alloy. An EMAX energy spectrometer was used to analyze the chemical composition of each phase in the hardfacing alloy, and measurements were carried out simultaneously with the FESEM. After a 0.3 mm thin sample was cut off, it was mechanically ground to a thickness of less than 40 μm. The model 691 Precision Ion Polishing System (PIPS) was used to polish the sample and prepare a cylindrical transmission electron microscopy (TEM) sample (Ø3 mm). The JOEL 2010 transmission electron microscope was used to observe the morphology of the hardfacing alloy and perform electron diffraction on a selected area to determine the crystal structure based on the calibrated electron diffraction spots.

A self-made abrasive-belt friction and wear tester [21] was used to carry out the abrasive belt wear experiments. A 40-mesh SiC abrasive belt was selected as the abrasive material. The abrasive belt speed was 300 mm s⁻¹, the sample load was 10 N, and an electronic balance with an accuracy of 0.1 mg was used to measure the sample mass and calculate the mass loss. The self-developed impact abrasive-particle wear tester [21] was used to evaluate the impact wear resistance of the hardfacing alloy. SiC particles were selected as the impact abrasive particles with an average particle size of 350 μm. In the particle impact abrasion process, the abrasive particle injection speed was 80 m s⁻¹, the particle aggregation degree was 5200 g m⁻³, the blast nozzle diameter was 8 mm, angles of 30° and 90° between the incident direction of the abrasive particles and the specimen holder were selected for testing, and the wear time was 10 min. After the test was completed, each sample was ultrasonically cleaned in alcohol, and finally, an electronic balance was used to measure the mass of the sample after abrasion to determine the mass. To ensure the accuracy of the test, five sets of wear samples were prepared for each hardfacing alloy to conduct wear experiments, and finally, the average value of the wear-related mass loss was calculated according to the amount of wear. A ConScan confocal optical profilometer was used to collect the surface morphology data of the sample after the friction and wear test, and the corresponding Image Plus software was used to analyze the morphology data.

### 3. Experimental results

#### 3.1. Microstructure and morphology of hardfacing alloy

Figure 1 shows the microstructures of the two types of hardfacing alloys. Figure 1(a) displays the microstructure of the No. 1 hardfacing alloy. A typical bulky primary M₇C₃ carbide and a substrate composed of crystalline austenite and eutectic M₇C₃ carbide were observed. The bulky primary M₇C₃ carbides were in a hexagonal or irregular-hexagonal shape, and the eutectic carbides had flaky or fine rod-like structures. The average size of the primary carbides was 40 μm. Figure 1(b) displays the microstructure and morphology of the No. 2 hardfacing alloy. The composition of area A was analyzed by EDS, which shows that the composition is mainly composed of the alloying elements Fe, Cr, Mn, Si and C. Unlike the No. 1 alloy, there were no bulky primary carbides in the microstructure of the No. 2 alloy. Not only were the sizes of the carbides very small but also the fine carbides had a very high distribution density, and the distribution of carbides was very uniform. Additionally, the distribution density and uniformity of these fine carbides were higher than those of carbides in the eutectic substrate of the
Figure 1. Microstructure and morphology of hardfacing alloy: (a) No. 1 alloy, (b) No. 2 alloy, (c) No. 2 alloy under high magnification, and (d) morphology of the longitudinal section of No. 2 alloy.

No. 1 alloy. Figure 1(c) shows the enlarged image of the microstructure of the No. 2 alloy, and it was found that the carbide also had a hexagonal or irregular-hexagonal shape. Figure 1(d) shows the microstructure of the No. 2 alloy perpendicular to the surface of the hardfacing alloy. Based on figures 1(c) and (d), it was inferred that the carbides in the No. 2 hardfacing alloy all had fine rod-shaped structures, and the average size was about 0.5 μm.

3.2. Morphology and wear-related mass loss results of friction and wear experiments

3.2.1. Results of abrasive-belt friction and wear

To study the influence of the change in the hardfacing alloy structure on the abrasive wear resistance, an abrasive-belt friction and wear test was used to simulate the abrasive wear resistance of the hardfacing alloy. Figure 2 shows the curve of the wear-related mass loss of the hardfacing alloy with wear time. As the wear time increased, the wear-related mass losses of the two types of hardfacing alloys gradually increased, showing a basically linear growth trend. In each wear period, the wear-related mass loss of the No. 1 hardfacing alloy was significantly greater than that of the No. 2 hardfacing alloy. Therefore, the wear resistance of the No. 2 hardfacing alloy was higher than that of the No. 1 hardfacing alloy.

3.2.2. Results of impact abrasive-particle wear

Table 4 shows the results of the impact abrasive-particle wear test of the hardfacing alloy under impact angles of 30° and 90°. It was found that the wear rates of each hardfacing alloy at these two impact angles were different. The wear rates of the two hardfacing alloys under an impact angle of 90° were higher than those under an impact angle of 30°. However, the magnitudes of the increases in the wear rates of the two hardfacing alloys were different. When the impact angle was changed from 30° to 90°, the wear rate of the No. 1 hardfacing alloy changed significantly. When the impact time was 15 min, the mass loss of the alloy increased from 1.13 to 1.70 g, and the increase was 50.0%. However, the magnitude of the change in the impact abrasive-particle wear rate of the No. 2 hardfacing alloy was relatively low, the wear-related mass loss increased from 0.52 to 0.61 g, and the wear-related mass loss increased by 18.2%. If the wear resistance of No. 1 hardfacing alloy was normalized to 1, the wear resistance of No. 2 hardfacing alloy under impact angles of 30° and 90° are 2.17 and 2.78 respectively. Therefore, with the presence of ultrafine carbide particles, under 30° and 90° impact angle, the wear resistance of No. 2 hardfacing alloy increase by 117% and 178% respectively. It was concluded that for the hypereutectic Fe-25Cr-4C hardfacing alloy, the 90° impact angle was a more severe wear condition than the 30° impact angle.
Bitter confirmed [29] that the impact angle had a different impact on the wear rates for brittle materials compared to ductile materials. Materials with high ductility generally show a relatively high wear rate in the range of 30°–45°, while the wear rates of brittle materials gradually increase with increasing impact angle. The greater the brittleness of the material is, the more the wear rate increases with increasing angle. Based on the differences in the wear rates of the above-mentioned hardfacing alloys at the two impact angles of 30° and 90°, it was concluded that the two hardfacing alloys were brittle materials, but their brittleness characteristics were different. Under the same wear conditions, the wear rates of the two hardfacing alloys gradually decreased with decreasing carbide size, and the No. 2 hardfacing alloy exhibited a relatively high wear resistance.

### 3.3. Wear morphology analysis

#### 3.3.1. Morphology after abrasive belt wear

Figure 3 shows the scanning images of the two hardfacing alloys after the abrasive belt wear and the probability density distribution of the scanning data. As shown in figure 3(a), wear scars on the No. 1 hardfacing alloy surface were very evident. It was preliminarily speculated that this was caused by the large primary carbides prone to spalling. According to the surface data distribution of the wear scar in figure 3(b), the distribution range was 11.3–110.8 µm, the main distribution range was 37.6–92.5 µm, and the pattern of the distribution conformed to a normal distribution. As shown by the wear scar morphology of the No. 2 hardfacing alloy in figure 3(c), there were no rough wear scars on the surface. The worn surface showed shallow and evenly distributed wear scars, and the middle ‘platform’ did not appear in the pattern of the data distribution. The median value was 47.5 µm, the data also conformed to a normal distribution, and the distribution range was narrow at 21.5–73.4 µm.

No. 2 hardfacing alloy: (c) morphology of worn surface and (d) probability density distribution of scanning data.

| Hardfacing alloy | Wear-related mass loss under different impact angles (g) | Wear resistance (normalized) |
|------------------|--------------------------------------------------------|-------------------------------|
|                  | 30°  | 90°  | 30° | 90° |
| No.1             | 1.13 | 1.70 | 1   | 1   |
| No.2             | 0.52 | 0.61 | 2.17| 2.78|
Figure 3. No. 1 hardfacing alloy: (a) morphology of worn surface and (b) probability density distribution of scanning data.

Figure 4. Three-dimensional wear morphology and probability density distribution of scanning data of hardfacing alloy under different impact angles: (a) and (b) No.1 of 90°; (c) and (d) No.2 of 90°.
3.3.2. Morphology after impact abrasive-particle wear

The three-dimensional wear morphologies in figures 4(a)–(b) showed that after the impact abrasion of the hardfacing alloy, the worn surface was very rough, with many different sizes of dents distributed on the surface. A Gaussian distribution of the probability density of the surface data was found. At an impact angle of 90°, the data distribution range was 44.7–91.5 μm, with an intermediate value of 72.5 μm.

Figures 4(c)–(d) show the 3D wear morphology of the No. 2 hardfacing alloy and the probability density distribution of the surface morphology data. The dents of the wear morphology in the figure were relatively small, and the sizes of the dents were relatively uniform. The probability density distribution curve of the data showed that under an impact angle of 90°, the distribution range became 46.6–76.0 μm, with an intermediate value of 64.3 μm. Compared with the data distribution curve of the No. 1 hardfacing alloy, the curve distribution of the No. 2 hardfacing alloy was ‘tall and thin,’ and the data distribution range was more concentrated.

4. Discussion

4.1. Wear mechanism of hardfacing alloy

The results of the friction and wear experiments showed that the microstructures of the two hardfacing alloys were different, and their wear resistances were also considerably different. The No. 2 hardfacing alloy exhibited relatively high wear resistance under the two friction and wear experiments. To better analyze the microscopic wear mechanism of the two hardfacing alloys, we corroded the worn surfaces of the two hardfacing alloys.

Figure 5 shows the FESEM images of the worn surface microstructure that was obtained after the sample surface of the hardfacing alloy subjected to abrasive belt wear was corroded. When there were large pieces of primary M₇C₃ carbides in the No. 1 hardfacing alloy, because of the high hardness and brittleness of the carbides, a large number of cracks formed on the carbides, causing cracks. During the formation and propagation of the carbide cracks, many secondary, tertiary, or even multiple cracks were also produced. The sizes of the primary M₇C₃ carbides were large, and the distribution was uneven. The morphology image after corrosion showed that when the carbides were not cracked, the wear scars were narrow and shallow, there were no large blocks, and the carbide area wear scars were wide and deep. Since a large number of cracks formed on the carbide, it was speculated that the entire carbide did not spall off in one piece during the wear process but gradually cracked, forming small fragments as the cracks occurred, and then spalled off. The spalled carbides also had very high...
hardness and could form many sharp corners, which could be used as new wear particles to further cause more severe wear.

Figure 5(b) shows that the wear scar of the No. 2 hardfacing alloy was relatively uniform, and no deep furrows were evident. The corrosion morphology of the worn surface of the No. 2 hardfacing alloy was observed after magnification. Some carbides were bent and deformed, and the deformed carbides could also provide 'skeletal support' for the material during the wear process. There were no large pieces of primary M7C3 carbides in the No. 2 hardfacing alloy, so it had no carbides for cracking and spalling. Although the fine carbides were partially broken, the small fragments of the fractured carbides did not spall off. They were still embedded around the austenite substrate and could still be gradually worn away after a period of support during the wear process. Therefore, the No. 2 hardfacing alloy exhibited higher wear resistance.

Figure 6 shows the FESEM images of the worn surfaces of the two hardfacing alloys after the impact abrasive-particle wear test. Under an impact angle of 30°, multiple cracks formed inside the primary carbides in the No. 1 hardfacing alloy, as shown in figure 6(a). This was because when the abrasive particles impacted the carbides, the carbides with high hardness could not deform and resist the impact of the abrasive particles, so cracks were generated and spread radially. This high-speed abrasive particle impact was the main reason for the formation of cracks in the carbide. When the abrasive particles continued to impact the carbide, this radial crack could cause the carbide to quickly produce cracks and spall off. Figure 6(b) shows the carbides that had been broken and were about to spall off, indicating that as the impact angle became larger, the large primary carbides were highly prone to cracking and spalling off under the action of a higher impact force of the abrasive particles. As shown in figure 6(c), under the action of small-angle erosion of the abrasive particles in the No. 2 hardfacing alloy, since the carbides were very small, no cracks inside the carbides were observed. Because the sizes of the carbides were very small, if cracks formed, they could be separated from the original carbide. However, due to the very small sizes of the carbides, the contact area between a carbide and the austenite substrate was relatively large, and the carbide did not spall off immediately under the protection of the austenite. The spalling off and transfer of materials occurred during the wear process.

As shown in figure 6(d), when the impact angle was 90°, the surrounding materials had deformed, and the austenite substrate wrapped around a large number of small carbides that underwent certain plastic deformation, and the original straight rod-shaped carbides showed a certain degree of bending. Based on this phenomenon, it was inferred that the ductility and plasticity of the No. 2 hardfacing alloy had been greatly improved.

The behavior of fine carbides in the wear process was observed in more detail. Figure 7 shows the enlarged image of the surface morphology of the No. 2 hardfacing alloy after wear. It was found that one carbide seemed to split and form two carbides according to the morphological characteristics of the carbide (point A indicated by the arrow). A certain degree of bending occurred, and no cracks formed on the carbides. Combined with the
TEM image of the No. 2 hardfacing alloy in figure 8, it was found that there were two growing carbides that were attached to each other. It was inferred that the splitting behavior of the carbides was caused by the separation of two carbides that were originally attached to each other during the wear process. The observation of the carbide at point B showed that the right end of the carbide suddenly became smaller, which was presumably caused by the spalling of a layer of material in this area. That is, this small carbide exhibited a phenomenon of layer-by-layer spalling off during the wear process.

In the above-mentioned impact abrasive-particle wear test, small solid particles as abrasive particles impacted each hardfacing alloy at a relatively high speed. Therefore, such small particles carried high kinetic energy, and part of the kinetic energy could be converted into energy for cracking or deformation of the carbides in the alloy through energy transfer during the impact process. When the impact point of the abrasive particles was exactly on a large carbide, the carbide could not absorb the impact energy of the abrasive particles through deformation. Therefore, with the impact point as the source of the cracks, radial cracks formed, and brittle spalling of the material occurred under the continuous impact. When the impact angle became larger, according to the momentum theorem, a greater impact force would be vertically loaded on the surface of the hardfacing alloy, forming more cracks, and the phenomenon of carbide spalling would become more severe.

When there were no large carbides in the hardfacing alloy, in the above impact wear test, the surface of the material underwent plastic deformation under the action of the impact of small solid particles, and the carbides absorbed the energy in the impact process by bending or layer-by-layer spalling. Under the action of continuous impact indentation, the surface of the material eventually ruptured and spalled off, resulting in abrasion. Compared with the spalling of large carbides, the speed of the material spalling caused by multiple indentation deformation was smaller, the wear rate of the hardfacing alloy was reduced, and the wear resistance was improved. Therefore, when the impact angle became larger, the higher the ductility of the material and the higher the wear resistance became.

Based on the behaviors of the carbides of different sizes during the wear process in the abrasive belt and impact abrasive-particle wear experiments, we inferred that when the abrasive particles were pressed into the carbide surface under the action of a normal load, the friction force caused the shearing, ploughing, and cutting

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Figure 7. Behavior of carbides on the worn surface of No. 2 hardfacing alloy: (a) splitting characteristics of the carbide and (b) layer-by-layer spalling of the carbide.
of the surface through the ploughing effect of the abrasive material when it slid, resulting in groove-like wear scars. When the abrasive particles were pressed into the friction surface under the action of squeezing to produce an indentation, a certain amount of plastic deformation could occur in the plastic area of the friction surface, and cracks that formed in the brittle area and the material spalled off. Finally, under the action of the cyclic contact stress generated by the abrasive particles on the friction surface, the surface material spalled off due to fatigue. Therefore, reducing the carbide sizes could reduce the cracking and spalling tendency of the carbides. The average size of the M7C3 carbides in the No. 2 hardfacing alloy reached 0.5 \( \mu \text{m} \), which significantly reduced the cracking and spalling of the carbide.

In the hypereutectic Fe-25Cr-4C hardfacing alloy with ultrafine carbides, there were no large bulk primary carbides. All carbides were very small in size, and the volume fraction of the carbides reached more than 40%, which had a good support effect. Meanwhile, the carbide distribution was uniform, and the carbide distribution density was very high. The microstructure could reduce the force of the abrasive particles on the surface and evenly distribute the load, which had a significant effect on reducing the micro-cutting effect of the abrasive particles and could significantly improve the wear resistance of the hardfacing alloy.

### 4.2. Mechanism of formation of shell carbides

As a metastable phase, M7C3 carbide can undergo the \( \text{M}_7\text{C}_3 \rightarrow \text{M}_{23}\text{C}_6 \) transformation under certain conditions. However, there are no reports of \( \text{M}_7\text{C}_3 \) carbide transforming to \( \text{M}_{23}\text{C}_6 \) carbide under the condition of fast cooling during hardfacing welding. Previous studies have found that in the Fe-Cr-C alloy, when the transformation of \( \text{M}_7\text{C}_3 \) carbide \( \rightarrow \text{M}_{23}\text{C}_6 \) carbide occurs, the Cr in the austenite substrate exhibits uphill...
diffusion into the primary carbide. When the Cr content reaches a critical value, the external layer first undergoes a transformation to form M\textsubscript{23}C\textsubscript{6} carbide.

According to previous research results [22], the volume fractions of M\textsubscript{1}C\textsubscript{3} carbides in the two hardfacing alloys were basically the same. In the No. 1 hardfacing alloy, the average size of bulky carbides was 40 \( \mu \)m, and the volume of carbides with lengths of 1 \( \mu \)m was 400\( \pi \)\( \mu \)m\textsuperscript{3}. When the carbide was refined to an average diameter of 0.5 \( \mu \)m, if the length of the carbide was also 1 \( \mu \)m, then 6400 such carbides are needed to achieve a total volume of 400\( \pi \)\( \mu \)m\textsuperscript{3}. For a carbide with a size of 0.5 \( \mu \)m, the two-phase contact area between the primary M\textsubscript{23}C\textsubscript{6} carbides and the austenite substrate was 3200\( \pi \)\( \mu \)m\textsuperscript{2}, and the two-phase contact area between the primary M\textsubscript{23}C\textsubscript{6} carbides with an average size of 40 \( \mu \)m and the austenite substrate was 40\( \pi \)\( \mu \)m\textsuperscript{2}. The phase interfaces of the two different sizes of primary carbides were 80 times different. Therefore, the contact area of the primary carbides with diameters of 0.5 \( \mu \)m and the austenite substrate was considerably larger than that of primary carbides with an average size of 40 \( \mu \)m. As a result, the carbides with smaller sizes were more likely to undergo the M\textsubscript{1}C\textsubscript{3} \textrightharpoonup M\textsubscript{23}C\textsubscript{6} transformation. Although the weld pool cooled quickly under welding conditions, the contact area between the two phases was very large, so some fine primary carbides underwent the M\textsubscript{1}C\textsubscript{3} \textrightharpoonup M\textsubscript{23}C\textsubscript{6} carbide phase transformation, forming a composite phase structure with a hard core and a soft shell layer.

Figure 8(a) shows two adjacent growing M\textsubscript{1}C\textsubscript{3} carbides, and high-density dislocations also formed around the carbides. The carbides that grew in close contact could not only provide sufficient support during wear but could also prevent crack propagation. The M\textsubscript{23}C\textsubscript{6} carbide formed at the upper J position in figure 8(b) and the K position at the lower right corner of the M\textsubscript{1}C\textsubscript{3} carbide. Electron diffraction was performed on the J and K regions, and the diffraction spots are shown in figures 8(c) and (d), respectively. The M\textsubscript{1}C\textsubscript{3} carbide possessed an orthorhombic crystal structure, with lattice constants of \( a = 7.01 \) nm, \( b = 12.2 \) nm, and \( c = 4.53 \) nm. The distances between the (200), (041), and (241) planes in figure 8(c) were measured to be 0.35, 0.24, and 0.20 nm, respectively. The measurement results were consistent with the characteristics of the crystal structure of M\textsubscript{1}C\textsubscript{3}. The electron diffraction spots in figure 8(d) were calibrated, and they were found to be face-centered cubic lattices. The (2 \( \bar{4} \) 2) and (6 \( \bar{2} \) 0) crystal plane spacings in the figure were measured to be 0.22 and 0.17 nm, respectively. The calculated lattice constant was \( a = 1.08 \) nm, and it was determined that M\textsubscript{23}C\textsubscript{6} carbide formed in this area. This was also consistent with the fact that the spalling of the outer layer material of the small-sized carbides in figure 7 occurred first.

5. Conclusions

(1) In the hypereutectic Fe-25Cr-4C hardfacing alloy, when the average size of the primary M\textsubscript{1}C\textsubscript{3} carbides reached 0.5 \( \mu \)m, the contact area between the small carbides and the austenite substrate greatly increased. The increase in the surface area of the carbides was favorable for the occurrence of the M\textsubscript{1}C\textsubscript{3} \textrightharpoonup M\textsubscript{23}C\textsubscript{6} transformation and promoted the formation of M\textsubscript{23}C\textsubscript{6} carbides. A layer of M\textsubscript{23}C\textsubscript{6} carbides was first formed on the outer layer of the M\textsubscript{1}C\textsubscript{3} carbides.

(2) When a relatively low-hardness M\textsubscript{23}C\textsubscript{6} carbide shell was formed around the M\textsubscript{1}C\textsubscript{3} carbide with high hardness, this hard core and soft shell dual-phase composite structure could improve the impact-abrasive-wear resistance of the hardfacing alloy during the wear process and improve its wear resistance.

(3) A more uniform microstructure could allow the wear process to be more uniform, reduce the force of the abrasive particles on the surface, and evenly distribute the load, without large-scale spalling, which had a significant effect on reducing the micro-cutting of the abrasive particles. Additionally, a uniform microstructure was conducive to reducing the high interfacial stress between the phase interfaces when the bulky primary carbides formed, reducing the spalling behavior of the carbides and improving the wear resistance of the hardfacing alloy.

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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).
CRediT authorship contribution statement

Jibo Wang: Conceptualization, Writing-Original draft preparation. Tiantian Liu: Writing- Reviewing and Editing. Xiaolei Xing: Investigation, Data Curation. Jian Li: Data Curation. Xiaowen Qi: Software, Formal analysis. Qingxiang Yang: Methodology, Supervision, Project administration.

Declaration of competing interest

The authors declare that they have no known competing financial interests.

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