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High wear resistance hypoeutectic Fe–C–B alloy by hot rolling

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Abstract

A novel high wear resistance hypoeutectic Fe–C–B alloy by hot rolling with isolated hard lumps distributed in matrix is reported. The effect of deformation on microstructure and properties of the alloy during hot rolling process was systematically investigated. The results showed that with the increase of deformation, the continuous hard phase network in the alloy was crushed into isolated hard lumps, and the properties of the alloy were improved significantly. The alloy with 66.7% deformation displayed excellent mechanical properties with the impact value of 12.6 J while its hardness was still kept at 58.6 HRC. And the wear resistance of the alloy with 66.7% deformation was 1.21 times as that of the commercial Cr20Mo high chromium cast iron.

1. Introduction

Wear resistance materials are used in various fields, such as mining, metallurgy and mechanical industry [1, 2]. Various alloys such as Hadfield steel, high chromium white cast iron, etc have been applied for hundreds years. However, study on new wear resistance materials is still an attractive research field to improve the properties and lower the cost of them. Hypoeutectic Fe–C–B alloy is a new wear resistance alloy and gets more and more attentions due to their low cost and high wear resistance [3, 4]. Ternary Fe–C–B phase diagram shows that boron forms Fe2B, Fe3B2(C, C), Fe4B3(C, C)6, etc during solidification [5, 6], and enhances hardness and wear resistance of the alloy significantly [7]. However, these boron contained hard phases display continuous network distribution along with the grain boundaries in the alloy. It results low toughness value of the alloy. Consequently, it is not suitable for application under heavy impact conditions. Many works such as alloying, heat treatment and modification have been reported to improve the toughness of the alloy [8–11], but the results are still not good enough due to extraordinary stability of boron contained hard phases.

Thermomechanical treatment, integrated process of hot deformation and heat treatment, can improve carbide morphology and properties of high chromium white cast iron significantly [12–15]. It has got more and more application in the field for morphology and properties improvement of other alloys [16, 17]. Our previous work demonstrated forging process can effective to break the continuous boron contained hard phase and improve the properties of the hypoeutectic Fe–C–B alloy [18]. However, it was difficult to produce wear resistance plates by forging, and the effect of deformation that plays the important role in the process is still not clear.

In this paper, we reported a novel high wear resistance hypoeutectic Fe–C–B alloy by hot rolling. And the effect of deformation on microstructure and properties of the alloy during the hot rolling process was investigated. The results are helpful to produce low cost high wear resistance alloy, and propose a new way to improve morphology and properties of the alloys with continuous hard phases.

2. Experimental

The hypoeutectic Fe–C–B alloy was melted in an induction furnace with pig iron, B–Fe (contained 19.5 wt% B) and scrap steel. After being superheated to 1830 K and deoxidized by aluminum, the melt was poured into sand
mold at 1750 K to obtain samples of 30 mm $\times$ 30 mm $\times$ 120 mm. The composition of the alloy is listed in table 1.

The samples were heated to 1373 K for 10 min to be rolled under different deformation on a rolling mill. After that, the samples were heat treated with the process of austenized at 1323 K for 30 min following with water quenched.

Impact values were tested on a JB-30A impact tester according to non-notch Charpy impact method using the specimens of 10 mm $\times$ 10 mm $\times$ 55 mm machined from the samples. Hardness of the samples were measured by a HR-150 Rockwell hardness tester. Microstructures of the samples were analyzed by a Leica EZ4D optical microscope. Fracture and wear morphology was observed by scanning electron microscopy (SEM). The phase composition of the samples were analyzed by XRD on a D/MAX-2200 diffractometer scanned in the range of 5°–90° in a step-scan mode (0.02° per step) with an x-ray source of Cu Ka radiation at 40 kV and 200 mA.

Wear resistance tests of the samples were carried on a MDL-10 impact abrasive wear test machine with schematic drawing shown as figure 1 [18, 19]. The test used the specimens of $\Phi$ 10 mm $\times$ 30 mm machined from the samples, and quenched 40Cr steel ring of $\Phi$50 mm $\times$ $\Phi$30 mm $\times$ 20 mm with a hardness of 50 HRC was chosen as the coupled lower specimen. During the wear test, the upper sample impacted the lower coupled steel ring 82 times every minute with a impact energy of 1 J for 1 h. And quartz particles of 1.5 $\sim$ 3 mm in size and 950 HV of hardness were used as abrasive particles during the test. The average rate of the abrasive particles through the machine’s outlet was 2.0 kg min$^{-1}$. Prior to the 1 h test, each sample was carried a preliminary impact abrasive wear for 20 min to maximize its contact area with the lower specimen and reduce the test error. The inverse of the weight loss after 1 h impact abrasive test was used as the wear resistance. A Cr20Mo high chromium cast iron sample quenched under 1273 K with the hardness of 60 HRC and impact value of 6.4 J was tested for comparison.

3. Results and discussion

3.1. Microstructure and properties of the unrolled alloy

Microstructure of unrolled as-cast hypoeutectic Fe–C–B Alloy and unrolled sample quenched in water after austenized at 1323 K for 30 min were shown in figures 2(a) and (b). It indicated that the as-cast microstructure of
the alloy was made up of pearlite and ferrite, product of primary austenite, and continuous ledeburite network with eutectic hard phases. And after quenched at 1323 K, the matrix of the alloy transformed into martensite while the hard phase morphology kept unchanged as continuous network along with the grain boundaries. The x-ray diffraction results showed that the hard phases in the as cast and heat treated unrolled samples were consist of Fe₃(B,C) and Fe₂B, as shown in figure 2(c). Since there was no improvement of hard phase’s morphology, the impact value of the alloy was only enhanced from 3.1 J to 4.5 J although the hardness of the quenched alloy is enhanced from 34 HRC to 56 HRC due to the martensite transformation of matrix.

3.2. Effect of deformation on microstructure of the alloy
The microstructures of samples rolled under different deformation are shown in figure 3, which are 16.7% (a), 33.3% (b), 50% (c) and 66.7% (d), respectively.

Compared with the microstructure of unrolled alloy in figures 2(b), 3(a) showed that when the deformation is 16.7% the hard phase morphology of the alloy changed little, which still presented continuous network although the matensite particles transformed from proeutectic austenite grain had been deformed. Figure 3(b) showed that when the deformation was increased to 33.3%, the network hard phases in ledeburite of the alloy began to be break and the matensite particles were deformed increasingly. And figures 3(c) and (d) showed that when the deformation was increased up to 50% and 66.7%, the network hard phases in ledeburite were broken to isolated hard phase lumps. It also could be found that little bit of secondary hard phases precipitated from the matensite matrix, which were shown in figures 3(c) and (d).

The XRD diffraction patterns of the samples with 16.7% deformation and 66.7% deformation were shown in figure 3(e). It indicated that the hard phases in the deformed samples were still Fe₂B and Fe₃(B,C), which were not changed compared with the unrolled sample listed in figure 2(c).

The improvement of hard phase morphology was the result of hot rolling. During hot rolling process, when the alloy was heated to 1373 K, the matrix was transformed to plastic austenite while the hard phases were still hard and brittle. The unchanged hard phases would resist deformation on both vertical and lateral directions.
during hot rolling process. However, since it should keep the same deformation with plastic austenite during hot rolling, shearing stress perpendicular to rolling direction and tension stress parallel to rolling direction would accumulate in the hard phases. It resulted in fracture of hard phases, as shown in figures 4(a) and 4(b) respectively. With the increase of the deformation, the thickness of the sample reduced gradually, and more and more hard phases were crushed and fractured correspondingly.

**Figure 3.** Microstructure and XRD results of hypoeutectic Fe–C–B alloy rolled under different deformation. (a) 16.7% deformation, (b) 33.3% deformation, (c) 50% deformation, (d) 66.7% deformation, and (e) XRD results of the rolled alloys with 16.7% deformation and 66.7% deformation.
After the hard phase broken, the plastic austenite matrix would pile into the cracks rapidly under the roll force. It isolated the broken hard particles and forced them to move with the austenite during the rolling process. Apart from breaking continuous eutectic hard phase into hard particles, the rolling process would also lead the austenite in ledeburite to be desorbed and merged into matrix. It resulted in improvement of matrix continuity and the decrease of ledeburite in the alloy with the increase of deformation, just as shown in figure 3. At the same time, the equilibrium of interfacial energy would be broken due to the produce of new interface that could provide a driving force for carbon and boron diffused and separated out to maintain equilibrium. The process would make isolated hard particles to be spheroidized.

The hot rolling process also resulted in precipitation of secondary phase. Since the solubility of carbon and boron in $\gamma$-Fe were greater than in $\alpha$-Fe, the matrix would dissolve more C and B atoms when the alloy heated to 1373 K. During deformation, numerous dislocations would be produced in austenite due to stress concentration, and the boron and carbon dissolved in the austenite would accumulate around dislocations. It resulted nucleation, growth and precipitation of secondary hard phase, as shown in figures 3(c) and (d). Figure 3 also showed that the precipitation amount of secondary hard phases raised with the increase of deformation.

### 3.3. Effect of deformation on mechanical properties of the alloy

Results showed that rolling deformation presents significant effect on the mechanical properties of the hypoeutectic Fe–C–B Alloy. Figure 5 showed that with the increase of deformation, the impact value of the alloy was increases from 6.5 J to 12.6 J while the hardness of the alloy kept steady, which were between 57.9 HRC and 58.6 HRC. It indicated that the properties of the alloy had been significant improved compare with the unrolled alloy whose impact value was 4.5 J and hardness was 56 HRC.

Figure 6 showed the impact fracture morphology of the unrolled alloy and the alloy with 16.7% deformation and 66.7% deformation. Figure 6(a) indicated hummocks and roundness pot holes with plane surface parts and few short ridge lines could be found in the unrolled alloy, which suggested its intercrystalline fracture mechanism. Figure 6(b) presented some ridge and a few river-like lines on the fracture morphology of the alloy with 6.7% deformation, which indicated the cleavage fracture mechanism of it. Figure 6(c) presented tear ridge morphology of the alloy with 66.7% deformation, which showed the quasi-cleavage fracture mechanism of the alloy with higher impact value.

The impact fracture morphologies of different processed alloys were corresponding with their properties and microstructures. For the unrolled alloy, the network hard phases were distributed along with the martensite crystal grain, which resulted in the fracture nucleating and propagating along the hard phase network rapidly. Since the matrix could impede fracture propagation, the fracture would extend along with the interface of hard phase and matrix, which led to intercrystalline fracture. After rolled to 16.7% deformation, a part of network hard phases in the alloy was broken, and the propagation of the fracture would be delayed. It resulted that the fracture mechanism of the alloy changed from intercrystalline to cleavage fracture, which was still belong to nonplastic fracture, and the impact toughness of the alloy was still not good. While when the alloy rolled to 66.7% deformation, the hard phase network was broken seriously, and the continuity of matrix was reinforced.
It resulted that the fracture propagation was prevented by high plastic matrix, which absorbed the crack propagation promptly and formed tear ridge morphology.

3.4. Effect of deformation on wear performance of the alloy

The wear data of the unrolled and the rolled alloy with different deformation were shown in figure 7. The data were also compared with that of commercial Cr20Mo high chromium cast iron. The composition of the
Cr$_{20}$Mo is listed in Table 2. It was quenched in the water after austenized at 1273 K for 30 min and reached 6.5 J of impact value and 61 HRC of hardness. It showed that the wear resistance of unrolled alloy was worst. And with the deformation increased gradually, the wear resistance ability of rolled alloys was enhanced. The sample with 66.7% deformation showed the best wear resistance among all samples, which was much better than that of Cr$_{20}$Mo high chromium cast iron. It reached to 1.21 times as that of Cr$_{20}$Mo.

Figure 8 showed the abrasive morphology of the unrolled alloy and the alloys with 16.7% deformation and 66.7% deformation. Figure 8(a) indicated that there were long deep narrow discontinuous microcut grooves and many fatigue spalling cracks on unrolled alloy, which implied that the wear of it was caused mainly by ploughing microcut and brittle fracture. Figure 8(b) showed that the abrasive morphology of the alloy with 16.7% deformation was different from figure 8(a). Although there were still many spalling cracks and microcut grooves, the gouges turned to short and wide.

Figure 8(c) showed that for the alloy with 66.7% deformation, most region of the worn surface was long wide shallow continuous microcut grooves. And wear debris could be found at the edge of grooves, which indicated that the wear mechanism had been changed to plastic deformation due to the improvement of hard phase morphology and significant enhancement of the properties. The crushed hard phases and isolated secondary

| Table 2. Chemical composition of the Cr$_{20}$Mo high chromium cast iron (mass fraction, %). |
|---|
| C | Cr | Mo | Mn | Si | P | S | Fe |
| 2.7 | 22.0 | 1.2 | 1.41 | 0.62 | 0.03 | 0.04 | balance |

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phases participated during rolling process not only reinforced the matrix but also prevented microcut of the quartz particles. It made the grooves were shallow.

4. Conclusions

(1) High wear resistance hypoeutectic Fe–C–B alloy was produced by hot rolling. And deformation had significant effect on the hard phase morphology and properties of the alloy.

(2) With the increase of deformation during hot rolling, the properties of the alloy are improved gradually. When the deformation of the alloy was increased from 16.7% to 66.7%, the impact value of the alloy was enhanced from 6.5 J to 12.6 J while the hardness of the alloy was kept between 57.9 HRC and 58.6 HRC.

(3) The abrasion performance of the alloy was enhanced significantly by hot rolling. The wear resistance of the alloy with 66.7% deformation was 1.21 times as that of Cr20Mo high chromium cast iron.

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