

Microstructure of Martensite/Bainite Dual-phase Grey Cast Iron and Its Strengthening Mechanism

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By a process of combinations of microalloying with Ni, Mo and other elements, cast chilled by metal-mold and special heat-treatment, A martensite/bainite (M/B) dual-phase grey cast iron was fabricated. Quenching into oil after austenization at 900°C for 2 h, the hardness of the sample was above HRC 48. The microstructures of all samples were characterized by optical metallography or X-ray diffraction. The final microstructure was composed of M/B dual-phase and retained austenite. There were some “white-bright zone” composed of few high-carbon martensite plates and retained austenite, full of Mn and Mo elements, concentrating on the boundaries of eutectic cells. The nucleation site and mechanism of bainite were studied by short oil-quenching for 5 s after austenization at 860°C for 2 h. The result showed bainite first nucleated at interface between flake graphite and austenite, and austenite along the direction of flake graphite sharp corner was also tend to transform into bainite. The second site easy for bainite nucleation was the boundary of austenite, also the boundary of “white-bright zone”. Strengthening mechanism of M/B dual-phase to this grey cast iron was studied by using pressing-in method to press round head into the surface of corrosive M/B dual-phase grey cast iron. The initiation and extension of cracks were observed through SEM. The result showed that M/B dual-phase could effectively strengthen graphite–matrix interface and restrain cracks extension, tips of high-carbon martensites in “white-bright zone” were another site prone to initiate cracks under pressure. For lake of phase–phase interfaces, the cracks would extend straightly until to the whole interface of “white-bright zone” and matrix.

KEY WORDS: martensite/bainite; white-bright zone; nucleation; strengthening mechanism.

1. Introduction

Grey cast iron is a complicated, multiple and multiphase structure material which has great of development potential.^{1–3)} Because its structure and properties can change greatly with the way of solidifying, grey cast iron has been wildly used in manufacturing field. Take example of ball-flashing plate, in order to satisfying the purpose of cutting steel-balls quickly, it has two important factors to resolve: first is small and uniform flake graphite distribution, the second is high strength and wear-resistant matrix.^{4,5)} It has been well-known that flake graphite has some negative influences to grey cast iron.^{6–8)} For example, flake graphite hardly has any strength and its sharp edge would create stress concentration effect in the interface between graphite and matrix. Recently, theories of improving strength of grey cast iron have been studied by many researchers. It has been proved that flake graphite could not be looked as crack, and high-strength phase like martensite or bainite could strongly improve strength of grey cast iron by strengthening graphite–matrix (G–m) interface and restraining crack initiation.^{9,10)}

In 1969, Eduards had proposed that martensite/bainite (M/B) dual-phase structure had higher toughness than sin-

gle phase martensite or bainite.¹¹⁾ Ever since then, many researchers have investigated the properties of M/B structure. The results showed that M/B structure often has an excellent combination of strength and toughness, shows excellent performance under impact abrasion.^{12–14)} Until now, M/B dual-phase has been successfully used on wear resistant ductile iron and high-strength steel. If M/B dual-phase could be successful used on grey cast iron, it would highly improve its properties and widen its fields of application.

2. Experimental Details

2.1. Chemical Composition of the Grey Cast Iron

The chemical composition of the grey cast iron used in this work is shown in **Table 1**. To weakening the influence of flake graphite, the contents of C and Si in this grey cast iron were less than others and the number of Si/C was also adjusted. Too much of manganese and molybdenum would lead to chilling tendency or serious segregation boundaries of eutectic cells.

2.2. Casting and Heat-treatment

The cast iron was melted in a 150 kW medium frequency induction furnace. The tapping temperature was 1430–

Table 1. The chemical composition (wt%) of this grey cast iron.

C	Si	S	P	Mn	Mo	Ni
2.1-2.5	2.0-2.3	<0.03	<0.1	0.8-1.0	0.8-1.3	2.0-2.5

1450°C. After inoculation with BaSiFe, the melt was solidified in metal-mould. Cut a sample whose size was 15 mm×15 mm×10 mm and put it in medium temperature electrical resistance furnace preparing for heat treatment. By austenizing at 880–900°C subsequent with oil-cooling, the surface hardness of the sample can reach 48.3–52.8HRC. By austenizing at 860°C subsequent with oil-quenching for 5 s, the nucleation of bainite in this M/B dual-phase cast iron was observed.

2.3. Press-in Method

The initiation of cracks in Ni–Mo M/B dual-phase grey cast iron was studied by pressing round head into the corrosive surface. Cracks would initiate and extend in some site of M/B grey cast iron along with pressure's increasing.

3. Microstructure and Analyses

After polishing, all samples were observed by an optical microscope (OM). The phase analysis of quenching sample was separately performed on a Siemens D500 X-ray diffraction (XRD) autoanalyzer, to which an APPCE calculation program was attached.

3.1. Microstructure Analyses of Heat-treated Sample

Figure 1 displays the characters of microstructure as distance is different away from the graphite. It could be observed that there was a certain layer of closed bainite needles around the graphite, then is the admixture of needle-type lower bainite and martensite (the so-called M/B dual-phase) with few retained austenite between them. In the further distance, there are more retained austenite and definite martensite sheets which have an angle of 60 or 120 between each two sheets. In **Fig. 2**, the M/B dual-phase could be observed as matrix, and because of the segregation of manganese and molybdenum, a “white-bright zone” mainly existed at the boundary of two eutectic cells, which were mainly composed of high-carbon stabilized austenite and cryptocrystalline martensite according to the XRD results (**Fig. 3**), and the hardness could attain 870HV.

Because of higher content of normal-segregation alloys such as manganese and molybdenum, magnitude of which enriching at the boundary of eutectic cells is twice as the mean value, the non-atomic diffusion is hindered there. So the transition rate of bainite is intensely slow down, the retained austenite have little time to transform under rapid cooling. Further, silicon strongly prohibits the precipitation of carbide during bainitic transformation and enriches carbon in undecomposed austenite.

3.2. Bainite Nucleation Analyses

The bainite nucleation could be observed from the metallographic structure of Ni–Mo grey cast iron austenizing at 860°C subsequent with oil-quenching for 5 s (**Fig. 4**). The result showed the first nucleation site of bainite, which is

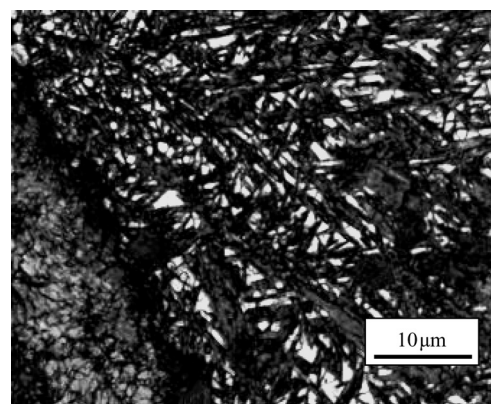


Fig. 1. The matrix of Ni–Mo grey cast iron after austenizing at 880°C for 2 h and quenching in oil.

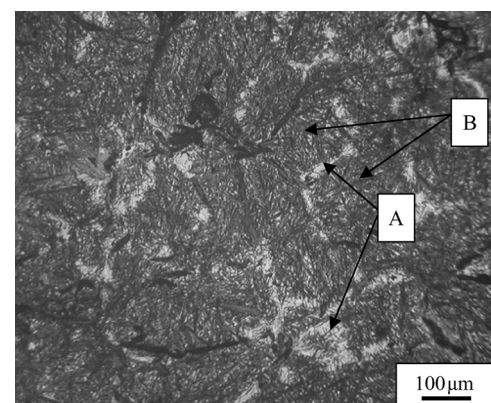


Fig. 2. The “white-bright zone” at the boundary of two eutectic cells. A zone is white-bright zone, B zone is eutectic cells.

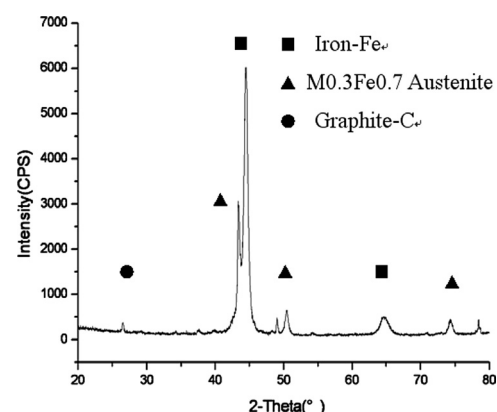


Fig. 3. The XRD result of the white-bright zone.

indicated as A zone, was the interface between flake graphite and austenite. Carbon poverty and silicon enrichment at the interface between flake graphite and austenite in addition with the symbiosis relationship between bainite ferrite(111) and superficial crystal of flack graphite(0001) make the bainite to first nucleate here. Another, analyzing from transforming thermodynamics, there are a lot of face

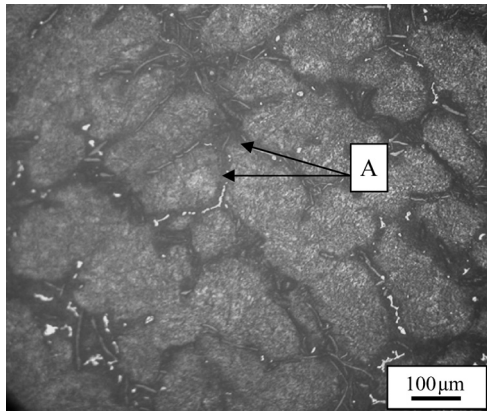


Fig. 4. Bainite nucleation of Ni-Mo grey cast iron after austenitizing at 860°C for 2 h and quenching in oil for 5–6 s. A zone is the first nucleation site of bainite.

defects around the graphite interface in favor of bainite nucleation because these face defects have higher energy. Observing from Fig. 4, austenite in the direction of flake graphite sharp corner also tended to transform into bainite. The reasons are the following aspects: On the one hand, in grey cast iron, when the sample was heated and insulation with austenizing temperature, because graphite is the proliferation of carbon source, carbon atoms, which are affected by graphite's chemical potential, have a tendency to move from austenite to graphite. As a result, the overall decline in free energy and austenite around the graphite has poor carbon. Because carbon-poor austenite is instability, it is very easy to transform into bainite. On the other hand, dislocation density of austenite around the graphite was very high, which can help bainite to nucleate as shear transformation.

The next site easy for bainite nucleation was the boundary of austenite, also the boundary of “white-bright zone”. The bainite transformation temperature here was lower than the first kind. The reason of this bainite deformation was because that austenite boundary was serious segregated zone of Mn and Mo elements, besides the co-segregation relationship between C–Mn and C–Mo induce carbon concentration here. So the narrow zone of austenite boundary which was also close to “white-bright zone” formed a carbon-poor zone. Because of Mn and Mo elements seriously stabilizing austenite, the lower-bainite nucleated here at a lower temperature.

3.3. Cracks Initiation and Extension

Figure 5 shows the first nucleation sites of cracks in this M/B dual-phase grey cast iron. It could be observed that cracks exist at the interface between graphite and matrix, especially at the sharp corner of flake graphite. Surface of graphite is also prone to initiate cracks under pressure. Increasing pressure, crack begin to increase inside of “white-bright zone”, which could be observed in **Fig. 6**. The accurate nucleation sites are cross corner of two martensite plates and interface between martensite corner and austenite.

As we all know that graphite is the most soft phase in the matrix, initiation of crack at G–m interface has relationship with the impediment effect of graphite to dislocation movement. Its performance for the relationship is reducing the

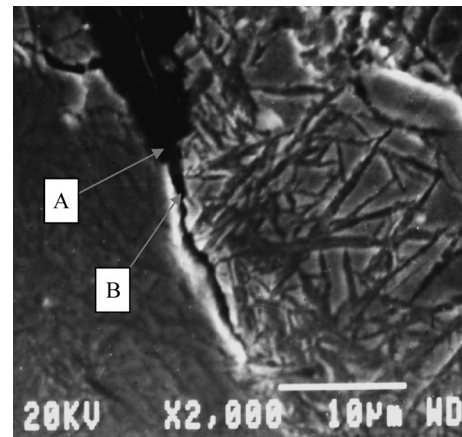


Fig. 5. The crack initiation around the flake graphite. A zone is the sharp corner of flake graphite; B zone is the crack initiation.

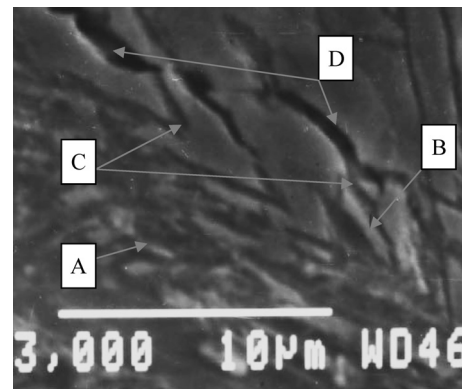


Fig. 6. The crack initiation in the “white-bright zone”. A zone is M/B dual-phase; B zone is “white-bright zone”; C zone is flake martensite; D zone is the crack.

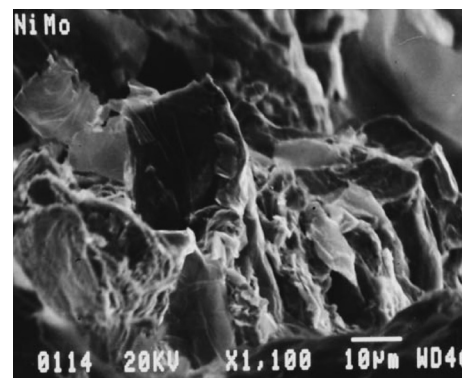


Fig. 7. The impact-fracture pattern of M/B dual-phase grey cast iron.

elastic energy of dislocation. Because of lower shear elasticity, soft graphite is nearly looked as “cavity”, and its effect to dislocation is attractive force rather than resistance force, and it could not create dislocation block or stress concentration. So, the fact is that crack will initiate at G–m interface when the applied stress was big enough. This viewpoint could also be approved by the impact fracture of this M/B dual-phase grey cast iron (**Fig. 7**), from which the completely correspondence of cleavage facet and former austenite crystal and the cracks extension mainly along with G–m interfaces could be clearly observed.

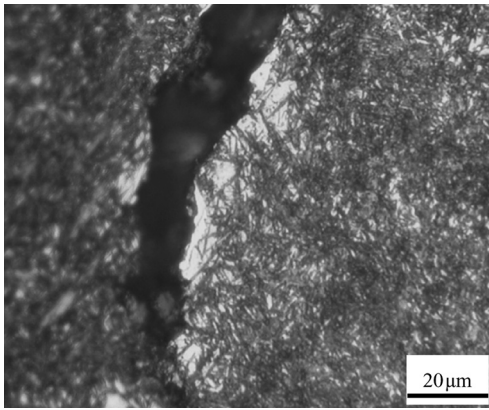


Fig. 8. The crack extended along with interface of “white-bright zone” and matrix.

Containing of retained austenite and few high-carbon martensite, “white-bright zone” is the other site prone to initiate cracks. The crystal structure of martensite is supersaturated solid solution of carbon in α -ferrite. The supersaturated carbon prone to lead distortion of α -ferrite cubic lattice, c -axis is elongated and a -axis is shortened, which creates large stress concentration. So this kind of martensite hardly has any plasticity and crack-sensitivity is much big. When the immediate-formation of martensite plates impact with each other and grain boundary of retained austenite, cracks would be easy to initiate at the cross corner of two martensite plates and interface between martensite corner and austenite. Because of low plasticity of “white-bright zone”, the formative micro-cracks can easily extend and stretch along with the boundary of “white-bright zone” when the applied stress was much big (Fig. 8).

3.4. Strengthening Mechanism of M/B Dual-phase Cast Iron

Along with the applied stress increasing, cracks initiated at G–m interface would prefer to extend along with the interface (Fig. 9). For those cracks, their extension at G–m interface needed the M/B dual-phase in a short range to do some backward deformation. To achieve this, the stress located at G–m interface must be greater than the yield strength of M/B dual-phase to destroy G–m interface. This is one of the reasons why M/B dual-phase could strengthen grey cast iron.

Cracks initiated G–m interface would go into matrix finally after the G–m interface was destroyed completely, especially those initiate at sharp corners of flake graphites would extend straightly. For the others, the extension would first meet low-bainite which has excellent plasticity to produce delaying effect, so the cracks would be ended or turn to extend along with grain boundary. Increasing pressure, cracks would go through low-bainite to form transcrystalline extension. Also the quantity of martensite is generally increasing here and there is some retained-austenite between acicular structures. M/B dual-phase has much more phase interfaces than single phase, which could further absorb the energy needed in the course of crack extension. The reasons are the following aspects: Firstly, the glide band was hard to cross large-angle grain boundary because the M/B dual phase has more phase interfaces, and the

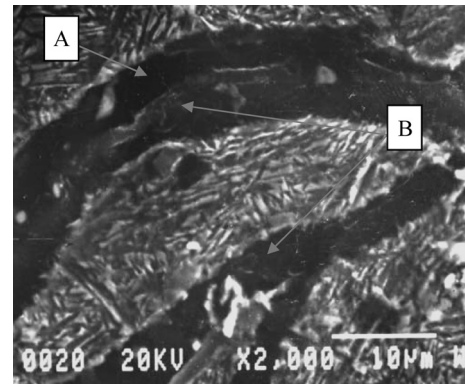


Fig. 9. The crack extended along with interface of graphite and matrix. A zone is the crack; B zone is flake graphite.

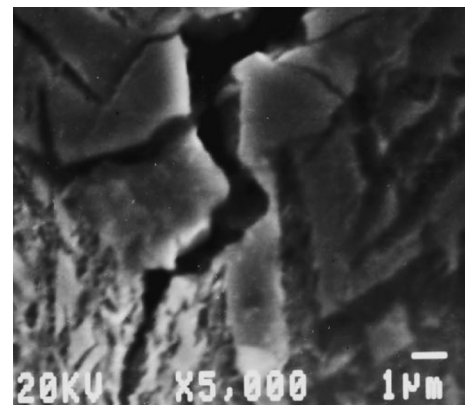


Fig. 10. The crack extension in the M/B dual-phase matrix.

stress concentration resisted by phase-interface would reduce largely in the plastic deformation which could delay the crack extension. Secondly, there is adequate low-bainite, which has excellent plasticity and toughness, located in the hard martensite matrix. The excellent plasticity and toughness of low-bainite has relationship with its metastructure that was composed by dislocation. Furthermore, in M/B dual phase, most carbon in low-bainite precipitated from ferrite in the form of uniform distributing fine carbide which not only effectively reduce its adverse effect to plasticity but also keep its favorable effect to the strength. All of these could relax the tip three-dimensional stress concentration of cracks and slow their extending speed and could further absorb the energy needed in the course of crack extension. Thirdly, the retained austenite between needle-like bainite and martensite could also help to resist crack's extension in the M/B dual-phase grey cast iron. Because of rich carbon content and restrained by surrounding low-bainite and martensite, the stability of this retained austenite was correspondingly improved. When subjected by applied-stress, retained austenite would passivate the tip of crack and relax its front-edge stress. So the route represented in a form of “Z” when crack coming across this M/B dual-phase and retain austenite (Fig. 10).

3.5. The Relation between the Suppression of Crack Extension and the Condition of Manufacturing Process

We speculate that there are some relation between the suppression of crack extension and the condition of manu-

facturing process. As mentioned above this paper, Ni–Mo alloying and heat treatment process could promote the formation of M/B dual-phase, after cracks entrance to matrix along with the applied stress increasing, cracks extension will be restrained by matrix, because M/B dual-phase has an enhancement on matrix. So Ni–Mo alloying and the heat treatment have an indirect enhancement on the suppression of crack extension.

4. Conclusions

(1) By a combination of Ni–Mo alloying and special heat-treatment, M/B dual-phase grey cast iron could be got in this article. Also, there were some “white-bright zones” exist at the boundary of two eutectic cells, which were mainly composed of high-carbon stabilized austenite and cryptocrystalline martensite, and the hardness could attain 870HV.

(2) The result showed the first nucleation site of bainite was the interface between flake graphite and austenite. Austenite in the direction of flake graphite sharp corner also tended to transform into bainite because of the high dislocation density here.

(3) The next site easy for bainite nucleation was the boundary of austenite, also the boundary of “white-bright zone”. The bainite transformation temperature here was lower than the first kind because of Mn and Mo elements seriously stabilizing austenite.

(4) Under pressure, cracks prefer to initiate at the interface between graphite and matrix (G–m), especially at the sharp corner of flake graphite. The “white-bright zone” was

the other site prone to initiate cracks, which initiated at the cross corner of two martensite plates or interface between martensite corner and austenite, and could easily extend along with the boundary of “white-bright zone” with the applied stress increasing.

(5) M/B dual phase can strengthen the G–m interface. The crack nucleated in the G–m interface would prior to extend along G–m interface. Increasing pressure, the crack will go into the matrix and extend as the shape of “Z”. It is the result of three factors: better strength and plasticity of bainite, a lot of interfaces between M/B dual phases, and good plasticity of retained austenite, which can restrain the extension of crack.

REFERENCES

- 1) W. Xu, M. Ferry and Y. Wang: *Mater. Sci. Eng. A*, **A390** (2005), No. 15, 326.
- 2) Y. Ikawa and T. Itami: *ISIJ Int.*, **30** (1990), No. 9, 756.
- 3) J. S. Zhang and P. L. Zhao: *Res. Iron Steel*, **7** (1995), No. 4, 50.
- 4) Q. C. Wang: *Foundry*, **12** (1992), 10.
- 5) L. F. Nie and X. J. Zhao: *Bearing*, **1** (2001), 16.
- 6) E. Hornbogen and J. M. Motz: *AFS Int. Cast Met. J.*, **2** (1977), No. 4, 31.
- 7) W. Patterson: *AFS Trans.*, (1962), 176.
- 8) W. Hiller: *Foundry*, **12** (1962), 54.
- 9) Z. R. He, G. X. Lin and S. Ji: *Adv. Fract. Resist. Mater.*, **3** (1996), 265.
- 10) Z. R. He and G. X. Lin: *Acta Mech. Solida Sin.*, **8** (1994), No. 5, 540.
- 11) Y. P. Zhang: *Acta Metal. Sin.*, **11** (1998), No. 1, 25.
- 12) H. Vettors: *Steel Res.*, **67** (1996), No. 10, 408.
- 13) D. P. Eduards: *JISI*, **207** (1969), 1494.
- 14) M. Delia: *Damage Mech.*, **7** (1998), No. 2, 103.