Effect of magnetic field strength on M$_6$C carbide precipitation behavior in W$_6$Mo$_5$Cr$_4$V$_3$ high speed steel during tempering

Y Wu$^1$, Z W Zhang$^2$, H H Li$^2$ and X Zhao$^{*2}$

$^1$Research Academy, Northeastern University, Shenyang 110819, Liaoning, China
$^2$ Key Laboratory for Anisotropy and Texture of Materials (Ministry of Education), Northeastern University, Shenyang 110819, Liaoning, China

*Corresponding author: zhaox@mail.neu.edu.cn

Abstract. Effect of high magnetic field strength on M$_6$C carbide precipitation morphology in W$_6$Mo$_5$Cr$_4$V$_3$ high speed steel was investigated. Results showed that at low and medium tempering temperatures, the high magnetic field significantly affects the precipitation morphology of M$_6$C carbides and shows strong spheroidization. This effect increases with the enhancement of the magnetic field strength. At high tempering temperature, the high magnetic field has no obvious effect on M$_6$C carbide precipitation behavior.

1. Introduction

In recent years, the influence of high magnetic field on phase transformations in steels has attracted much research interest, such as the study of magnetic field on martensite transformation [1-3], ferrite transformation [4-6], and pearlite transformations [7-9].

Carbides play an important role in the strengthening and toughening of microalloyed steels. Previous research has shown that the high magnetic field (10T or 12T or 14T) can affect the carbide growth behavior [10, 11] and change the carbides precipitation sequence [12-15]. These researches make great progress on the understanding of high magnetic field mechanism on carbide precipitation in steels. However, rare work has been involved in studying the effect of magnetic field strength on carbide precipitation behavior. In order to further investigate the mechanism of high magnetic field on carbide precipitation behavior during tempering, in this work, W$_6$Mo$_5$Cr$_4$V$_3$ high speed steel was selected and treated with different magnetic field strength. The effect of magnetic field strength on precipitation behavior of M$_6$C carbide was mainly focused.

2. Experimental

The alloy used in this study is an annealed HIP powder metallurgy W$_6$Mo$_5$Cr$_4$V$_3$ high speed steel. The chemical composition is (wt %): 1.24 C, 3.99 Cr, 5.25 Mo, 6.13 W, 3.01 V, 0.55 Si, 0.28 Mn, 0.18 Ni, 0.02S, 0.02 P, 0.14 Cu, 79.19 Fe. Specimens were first austenitized at 1180 °C for 5 min in a KSX-6-14 box-type furnace, and then immediately quenched in quenching oil at 20°C. The quenched specimens were then triple tempered at 200°C, 400°C and 560°C for 80min respectively, with different magnetic field strength (0T, 1T, 6T, 12T). In the case of the magnetic field tempering, the specimens were placed in the central (zero magnetic force) region of the furnace with their longitudinal direction parallel to the magnetic field direction (MD). The high magnetic field was applied during the whole heating, isothermal holding processes. The microstructure and composition
of the carbides is analyzed by scanning electron microscopy and Electron Probe Microanalysis (EPMA), respectively. In order to study the influence of high magnetic field on the precipitation morphology of M₆C carbides, 15 microstructure images (×2000) of different areas are randomly selected. The Image-pro Plus and Adobe Photoshop CS software are used to analyze statistically the spherical carbides (axis ratio < 1.2) proportion of M₆C carbides.

3. Results and discussion

Figure 1. Microstructure of specimens tempered at 200°C (a), 400°C (b) and 560°C (c) with different magnetic field strength.

Figure 1 shows the microstructures of W₆Mo₅Cr₃V₃ high speed steel tempered at 200°C, 400°C and 560°C with different magnetic field strength, respectively. In the figures, the magnetic field direction is perpendicular to the paper. On the basis of EPMA analysis (table 1), both the white angular-like and the ball-block-like carbides in figures are M₆C carbides. Here, M represents the metallic elements of Fe, Mo and W. From Figure 1(a) we can see that when tempered at 200°C with magnetic field of 0T, M₆C carbides precipitate as coarse angular-like along grain boundaries and ball-block-like in intracrystalline. When tempered with a magnetic field of 1T, 6T and 12T, the coarse angular-like M₆C carbides precipitating along grain boundaries disappear, and all the M₆C type carbides precipitate as ball-block-like and distribute dispersedly. From Figure 1(b) we can see that when tempered at 400°C with magnetic field of 0T, M₆C type carbides also precipitate as coarse angular-like along grain boundaries and ball-block-like in intracrystalline. With the increasing of magnetic field strength, the coarse angular-like M₆C carbides precipitating along grain boundaries decrease gradually; when tempered with a magnetic field of 12T, the coarse angular-like M₆C carbides disappear, and the M₆C type carbides precipitate as ball-block-like and distribute dispersedly. From Figure 1(c) we can see that when tempered at 560°C, the precipitation morphology of M₆C carbides is basically unchanged.
with the increasing of magnetic field strength, and all the M₆C carbides precipitate as ball-like. The above results indicate that at low and medium tempering temperatures, the high magnetic field significantly affects the precipitation morphology of M₆C carbides and leads to strong spheroidization, the stronger the magnetic field strength, the stronger the spheroidization effect. At high tempering temperature, the high magnetic field has no obvious effect on M₆C carbide precipitation morphology. The effect of high magnetic field on the precipitation morphology of M₆C carbides becomes weak with the increasing of the tempering temperature.

There is no obvious effect on the precipitation sites of M₆C carbides, and the M₆C carbides can precipitate at original austenite boundaries and in grain interior under different magnetic field strength.

| Element | Atomic% |
|---------|---------|
| C       | 13.44   |
| V       | 4.95    |
| Cr      | 4.61    |
| Fe      | 44.46   |
| Mo      | 17.01   |
| W       | 15.53   |

Table 1. The EPMA analysis of M₆C carbides in tempered specimens.

Figure 2 shows the spherical carbide proportion of M₆C carbides in specimens tempered with different magnetic field strength. From Figure 2 we can see that, the proportion of spherical M₆C type carbides increases remarkably with the increasing of magnetic field strength when tempered at 200°C and 400°C. When tempered at 560°C, the proportion of spherical M₆C type carbides is basically unchanged with the increasing of magnetic field strength.

Without magnetic fields, M₆C on grain boundaries shows a strong tendency to grow along boundaries, and coarse angular-like M₆C are eventually formed. This growth is energetically favorable as it can make use of the existing interfacial energy to lower the additional energy required for the formation of new carbide/matrix interfaces. But this growth tendency is not favored by the magnetic field. When M₆C carbides grow, the interfacial area becomes larger. If there is no magnetic field, the interfacial energy of M₆C /matrix is only determined by the crystal structure and composition differences between the two phases; conversely, if there is a magnetic field, the interfacial energy resulting from the difference in magnetization between the two phases should also be taken into account. For a solid phase, when magnetized by the applied magnetic field, the Gibbs free energy can be decreased. Therefore, the Gibbs free energy values for M₆C and matrix on both sides of the M₆C /matrix interface can be reduced by an amount corresponding to their magnetization degree. Conversely, the energy level of the interface remains unchanged, as the magnetization of interface is very low due to the disorderly atom arrangement and the large number of crystal defects. Thus, the interface energy increases relative to the grain interior, as suggested in Ref. [10]. Since the magnetic field can obviously raise the M₆C/matrix interfacial energy, the shape of M₆C that has minimum interface area is advantageous to minimize the final total interfacial energy. Therefore, the sphere-or particle-like M₆C is most favorable. In addition, the magnetostriction of M₆C
is different from that of the matrix. The growth of $M_6C$ will result in a difference in volume and therefore an increase in strain energy. According to the phase transformation theory, the strain energy resulting from this difference in volume depends on the shape of precipitates. When the hardness of the precipitates is much higher than that of the matrix, spherical precipitates generate the lowest strain energy [16]. Under the two effects of the magnetic field, the shape of particle-like $M_6C$ that has the minimum total interfacial area and minimum magnetostrictive strain energy is most favorable. Therefore, the magnetic field promotes the spheroidization of $M_6C$ type carbides, and this effect increases with the enhancement of the magnetic field strength. With the increasing of tempering temperature, the influence of temperature on Gibbs free energy increases, resulting in the weakening of the influence of magnetic field. So at high tempering temperature (560°C), the high magnetic field almost has no obvious effect on $M_6C$ carbide precipitation morphology.

4. Conclusions
Effect of high magnetic field strength on $M_6C$ carbide precipitation behavior in $W_6Mo_5Cr_4V_3$ high speed steel has been investigated, which leads to the following results: (1) At low and medium tempering temperatures, the high magnetic field obviously affects the precipitation morphology of $M_6C$ carbides and shows strong spheroidization, the stronger the magnetic field strength, the stronger the spheroidization effect. (2) The effect of high magnetic field on the precipitation morphology of $M_6C$ carbides becomes weak with the increasing of the tempering temperature.

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