

Carbides in High Chromium Cast Irons

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Abstract – High chromium cast irons are widely used as abrasion resistant materials. The mechanical properties and wear resistance of these irons depend on the type, hardness, morphology, distribution, volume fraction and orientation of the eutectic and precipitated carbides within their microstructures and on the nature of the matrix supporting these hard eutectic carbides. This paper is aimed to review the nature and crystallography of eutectic carbides and secondary carbides which are formed during solidification and precipitated during heat treatments of high chromium irons.

Keywords – High chromium cast iron, Eutectic carbide, Secondary carbide, Microstructure.

1. INTRODUCTION

High chromium irons can bridge the gap between the low toughness/good abrasion resistance Ni-Hard irons and the higher toughness/lower abrasion resistance high manganese steels. The abrasion resistance of high chromium irons is 20-25 times better than low carbon steels when the abrasive particles, such as quartz and garnet particles, are softer than the carbides [1]. Nearly all high chromium cast irons used for abrasion resistance are hypoeutectic alloys containing 10-30 wt%Cr and 2-3.5 wt%C. The alloys containing 12 wt%Cr are the cheapest, but 18-22 wt%Cr irons are the most popular range for general abrasion resistance such as rollers and tables in coal pulverizing or as liners in dry ball mills. The alloys containing 27-30 wt%Cr and 2.0-2.7 wt%C have been specially developed for combined abrasion and corrosion resistance in wet wear applications, for example slurry pumping in extraction processes [2-4]. Irons with 30-35 wt%Cr are used to resist oxidation and corrosion at high temperatures in applications such as furnace and burner parts. Microstructures of these alloys consist of ferrite and eutectic carbides [5-8].

The hypoeutectic high Cr irons solidify as primary austenite dendrites with a network of interdendritic eutectic carbides and during cooling some of the eutectic austenite around the eutectic carbides transforms to martensite [4, 5, 9-11]. The hardness levels of the austenitic irons are 500-520 HV. In some applications, as-cast austenitic irons can be used without heat treatment since the austenite can work harden at wear surfaces to provide a self-replacing wear resistant surface structure

[12, 13] in a similar way to the austenitic high Mn steels. For most applications, however, heat treatment is required. The primary purpose is to harden by destabilizing the austenite *via* precipitation of Cr-rich secondary carbides generally at 920-1060 °C for time of 1-6 hours depending on the alloy content and air quenching to room temperature during which the austenite with less Cr and C content can transform to martensite [14]. This gives a distribution of secondary carbides in a martensite matrix with small amounts of residual austenite and the hardness levels are increased to about 700-850 HV. This is so-called destabilization heat treatment. Tempering heat treatment after destabilization at 450-650 °C for normally 4 hours is to reduce the amount of the retained austenite in the matrix and the residual stress after quenching [3, 7, 15-18]. Karantzalis [19] found that tempering after destabilization led to the transformation of the martensite to pearlite structure, coarsening of previously secondary carbide precipitated and the formation of new finer carbide particles. Subcritical annealing heat treatment at 690-705 °C for several hours is also used to soften castings for machining by producing pearlitic structure and lowering hardness levels down to 400-450 HV. Full annealing involves re-austenitising in the range 900-1010 °C followed by slow cooling to 760 °C or below the A1 temperature, and then holding at this temperature for 10-50 hours [6]. The microstructure after the full annealing treatment consists of coarse spheroidal carbides in a ferrite matrix [20, 21]. Some very high temperature treatments at 1150-1180 °C, such as spheroidisation of eutectic carbides, are used to increase fracture toughness by producing a more globular eutectic carbide shape and controlled dispersion of secondary carbides in the matrix [3, 4]. Figure 1 shows an example of the microstructure in an as-cast and destabilized 30 wt%Cr-2.4 wt%C iron. The as-cast structure contains a network of eutectic carbide and the austenite (γ) matrix which partly transformed to lath martensite during cooling in the mould i.e. in the vicinity around the eutectic carbide. The destabilized structure consists of the network of eutectic carbide(s) and secondary carbide(s) precipitated in the prior austenite matrix which has been mostly transformed to martensite. The carbide types known to be present in the high chromium cast irons are M_7C_3 , M_3C , $M_{23}C_6$ and M_6C , with increasing M:C ratio [22, 23]. The type and morphology of the eutectic carbides in high chromium cast iron depend upon both chemical composition and solidification rate [3, 14, 24-26].

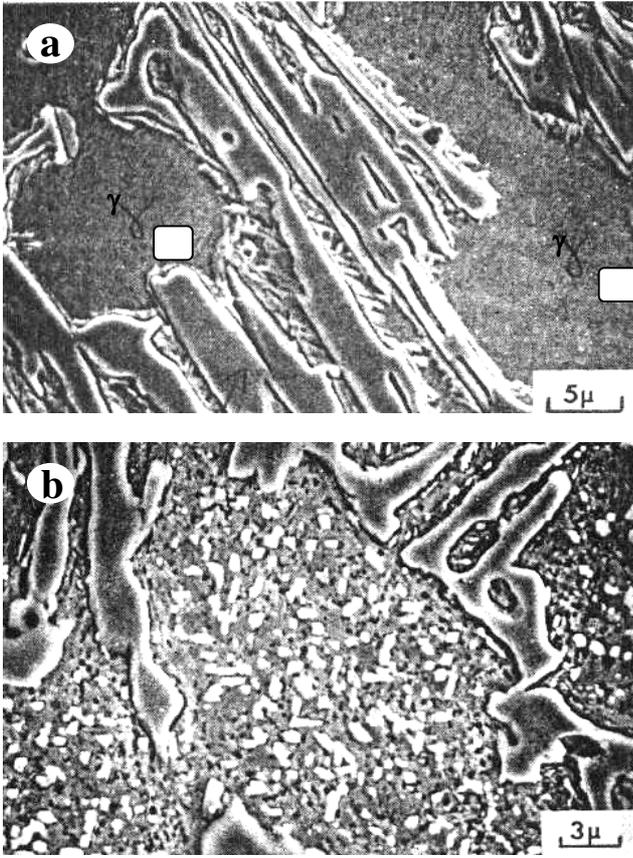


Figure 1 (a) micro-structure of as-cast 30 wt%Cr-2.4 wt%C with austenitic matrix, partially transformed to martensite (b) microstructure of destabilized 30 wt%Cr-2.4 wt%C containing secondary carbides with martensitic matrix [3].

2. EUTECTIC CARBIDE

2.1 M_3C Eutectic Carbide

M_3C carbide has orthorhombic structure with lattice parameters $a = 5.060 \text{ \AA}$, $b = 6.739 \text{ \AA}$ and $C = 4.499 \text{ \AA}$ [27]. In lower chromium irons with less than 12 wt%Cr, the eutectic carbide is M_3C with a hardness of around 1000 HV. Below about 6 wt%Cr, this carbide is in a continuous form which limits toughness. Figure 2 shows the eutectic carbide morphology in a 5 wt%Cr. When the amount of Cr is about 8-10 wt%Cr, the eutectic carbides are less continuous and can be duplex, consisting of an inner core of M_7C_3 and an outer shell of M_3C forming during solid state cooling in the mold [12]. Sinatora [28] found the morphologies of M_3C carbides in the 7 wt%Cr-(2.2-4.1) wt%C alloys as isolate plates and ledebulitic structure, resulting from cooperative growth with austenite. The amount of M_7C_3 was decreased as the carbon content was increased and the M_3C carbides were replaced the M_7C_3 carbides with carbon content exceeded 2.4 wt% at cooling rate of 550 °C/min. Reducing the cooling rate to 10 °C/min increased the eutectic reaction $L = \gamma + M_7C_3$ and delayed the reaction $L = \gamma + M_3C$. The M_3C carbides were observed at the end of the radial eutectic M_7C_3 carbides.

Inoue and Masumoto [29] found an orientation relationship between the M_3C and the M_7C_3 carbides in high carbon-chromium steels as;

$$(0\bar{1}1)_{M_3C} // (0001)_{M_7C_3}$$

$$(012)_{M_3C} // (1\bar{1}00)_{M_7C_3}$$

$$(\bar{1}00)_{M_3C} // (11\bar{2}0)_{M_7C_3}$$

Wang [30] found an orientation relationship between the M_3C and the ferrite matrix in 15 wt%Cr-1 wt%Mo-1.5 wt%V high chromium cast iron as;

$$(010)_{M_3C} // (100)_{ferrite}, [001]_{M_3C} // [011]_{ferrite}$$

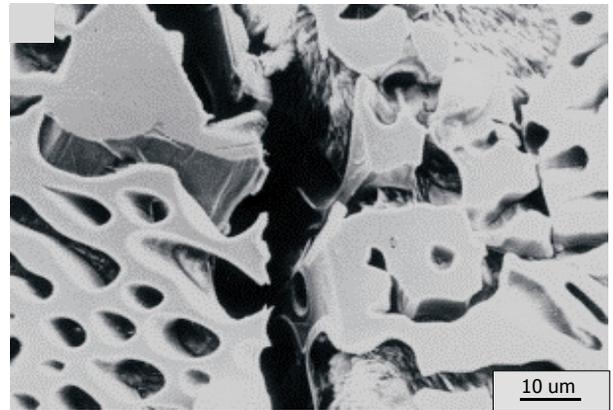


Figure 2 Deep etched microstructures of as-cast 5 wt%Cr-3.6 wt%C iron with continuous M_3C eutectic carbide [3].

2.2 M_7C_3 Eutectic Carbide

M_7C_3 carbide has a pseudo-hexagonal structure containing 56 iron atoms and 24 carbon atoms in a unit cell with lattice parameters $a = 13.9820 \text{ \AA}$ and $C = 4.5065 \text{ \AA}$, point group mmm and space group $Pmna$ [27, 31-33]. Powell [14] found that increasing the chromium and carbon contents in white iron changed the eutectic carbide from continuous M_3C to relatively discontinuous M_7C_3 , as shown in Figure 3, resulting in an increase in toughness. Above 12 wt%Cr, the eutectic carbide changes to the M_7C_3 type, which is often described as discontinuous or lamellar [5, 12].

The fibrous M_7C_3 has a higher hardness of around 1400-1600 HV, when compared to the continuous M_3C carbide with a hardness of about 1000 HV in lower chromium iron. This results in improved toughness and higher wear resistance compared to the lower chromium irons. In hypoeutectic irons containing between 20-35 vol% of eutectic carbides, the eutectic carbide is randomly nucleated from the liquid in the interdendritic areas between primary austenite dendrites, and then grows with a cellular interface [4,12,24]. The eutectic carbides consist of rod-like hexagonal pencil shaped crystals, which solidify as a colony structure, the rods join together to form blade-like

structures which are often described as lamellar. Cross sections of fine eutectic carbides are sometimes incorrectly described as globular. The rod-like M_7C_3 carbides become finer with increasing chromium content and with increasing rate of eutectic solidification [24-26]. The eutectic colony size is also decreased and the carbide spacing is also reduced with increasing the rate of eutectic solidification and with increase in chromium content. The length of eutectic cells becomes larger with an increase in the eutectic freezing range. The colony diameter in 15 wt%Cr and 30 wt%Cr irons was found to be 585 μm and 310 μm , respectively [24, 25].

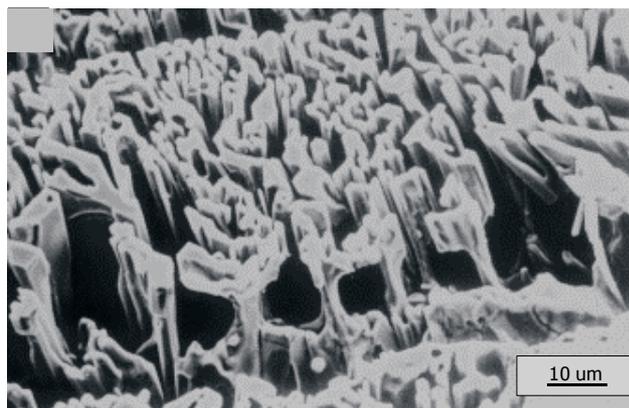


Figure 3 Deep etched microstructures of as-cast 30 wt%Cr-2.4 wt%C iron with finer and discontinuous M_7C_3 eutectic carbide [2].

Durman and Elwell [34] suggested that, in hypoeutectic alloys, the eutectic carbides form an interconnected network while, in hypereutectic material, the primary M_7C_3 exists as individual rods. Pearce [5] found that the eutectic M_7C_3 carbides grew as hexagonal rods or blades and that the overall diameter of the rods ranged from 1 to 4 μm with the smallest diameters of crystals within a cluster being in the order of 0.5 μm . There is only one growth direction, the $\langle 0001 \rangle$ [14]. Wiengmoon [35] found that the eutectic M_7C_3 carbide in as-cast semi-solid processed 27 wt%Cr cast iron consists of radiating clusters mixed with directional clusters. The size of these radiating eutectic colonies, measured from edge to edge was about 15 μm . In comparison, the thickness and the spacing of the eutectic carbide blades in as-cast conventionally cast 27 wt%Cr cast iron were about 0.6 μm and 0.25 μm , respectively, which were less than the values measured in the as-cast semi-solid processed 27 wt%Cr cast iron, i.e. about 0.9 μm and 0.4 μm , respectively. Eutectic M_7C_3 tends to grow as hollow rods with the diameter of the hollow core between 0.2 and 0.6 μm . This core is filled with matrix material of the same composition as that adjacent to the carbides. The cores can vary from circular to various other shapes and can move to the edge of the eutectic M_7C_3 during growth [5, 14]. Thin foil TEM studies [5, 35-38] of the eutectic carbides in 15-30 wt%Cr irons have confirmed that these carbides are M_7C_3 and that they are formed as rod- and/or blade-like structures. As mentioned above the eutectic M_7C_3 grows as hollow pencil

like crystals of hexagonal cross section, the blades being polycrystalline aggregates of rods as shown in Figure 4. M_7C_3 forms as eutectic carbide (or as primary carbides during solidification, or secondary carbides described later) with a distinguishable characteristic that it always appears to contain a high concentration of structural faulting. Stacking sequence faults likely occur on $\{10\bar{1}0\}$ and $\{11\bar{2}0\}$ planes giving rise to characteristic elongated reflections (streaking) in electron diffraction patterns and probably the presence of anti-phase domain boundaries in crystals [5, 36, 38-40]. Carpenter [36] reported that stacking faults within $(\text{Fe,Cr})_7\text{C}_3$ carbide occur by the omission of carbon atoms from the unit cell and as such do not destroy the hexagonal symmetry of the carbide.

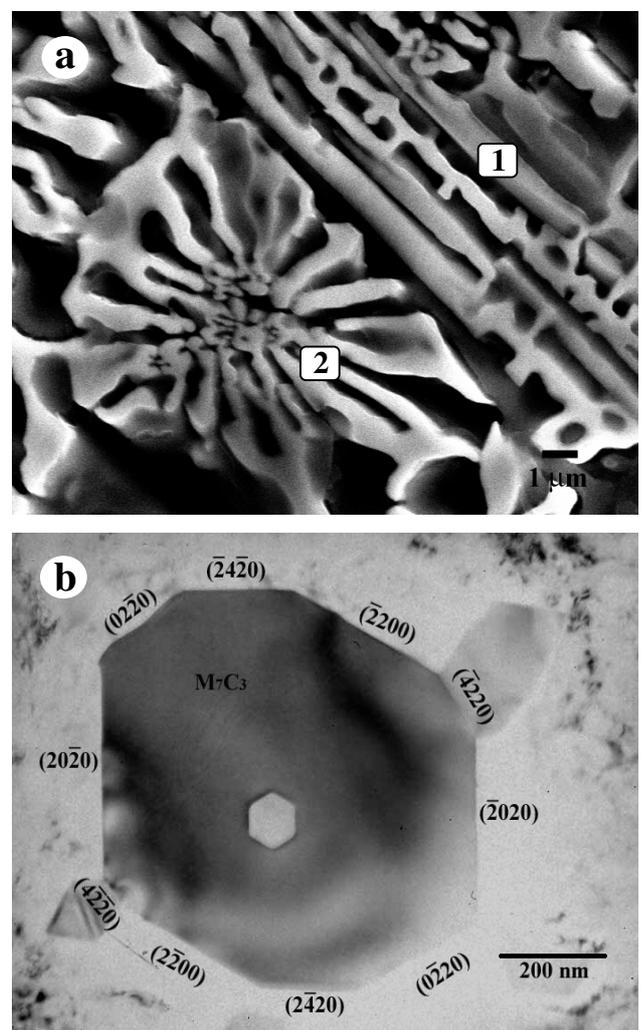


Figure 4 (a) Blade-like (1) and radiating clusters (2) morphologies of the eutectic carbide in the as-cast 27 wt%Cr cast iron, (b) BF-TEM micrograph shows the encapsulated core of the prior austenite within the M_7C_3 eutectic carbide in the as-cast semi-solid processed 27 wt%Cr cast iron [35].

In hypereutectic irons with volume fraction of carbides over 35 vol%, the eutectic hexagonal M_7C_3

carbides solidifies around the primary carbide as colonies [12, 24].

Shtansky [32] found the orientation relationships between the M_7C_3 carbide and the ferrite (bcc) as;

$$\begin{aligned} (\bar{1}13)_{ferrite} // (\bar{1}100)_{M_7C_3} \quad (\bar{1}13)_{ferrite} // (11\bar{2}0)_{M_7C_3} \\ (\bar{2}\bar{5}1)_{ferrite} // (11\bar{2}0)_{M_7C_3} \quad \text{or} \quad (\bar{2}\bar{5}1)_{ferrite} // (1\bar{1}00)_{M_7C_3} \\ (\bar{3}1\bar{1})_{ferrite} \approx // (0001)_{M_7C_3} \quad (\bar{3}1\bar{1})_{ferrite} \approx // (0001)_{M_7C_3} \end{aligned}$$

The M_7C_3 contains 24-50 %Cr content and 8.6-8.9 %C content with the Gibbs free energy of formation - 10 to -20 kJ/mole [7]. Pearce [2] reported that the microprobe analysis of the eutectic carbide in the as-cast 30 wt%Cr-2.4 wt%C iron suggests a formula of $(Fe_{1.8}Cr_{5.4})C_3$ for eutectic M_7C_3 . This result is in agreement with Dyson and Andrews [40] who studied the carbides in alloy steel containing 3.8 wt%Cr-0.14 wt%C-0.3 wt%Mo by x-ray fluorescence analysis, finding that the carbide formula is $(Fe_{1.4}Cr_{5.6})C_3$, the average M : C ratio is 2.36 : 1 and the average percentage of carbon is 8.63 wt%. Dogan [41] found that the formula of eutectic carbide in the 25 wt%Cr-3 wt%C iron determined by wavelength dispersive spectroscopy was $(Fe_{2.2}Cr_{4.8})C_3$. Wiengmoon [38] reported that the formula of eutectic carbide in the as-cast 30 wt%Cr-2.3 wt%C iron by microprobe analysis and electron energy loss spectroscopy is $(Fe_{1.8}Cr_{5.9})C_3$ and $(Fe_{1.3}Cr_{6.3})C_3$, respectively. The chemical analysis of eutectic carbide in the TEM thin foil of the as-cast 26.6 wt%Cr-2.7 wt%C iron was studied by Carpenter [42]. It was found that the chemical formula at the centre part of the eutectic carbide is $Fe_2Cr_5C_3$, whilst that at the edge is $(Fe_{2.2}Cr_{4.8})C_3$. Pearce and Elwell [43] reported the Cr : Fe ratios of the 30 wt%Cr-2.44 wt%C iron by energy dispersive analysis which varied from 2.29-2.77 for the M_7C_3 .

2.3 $M_7C_3 - M_{23}C_6$ Duplex Eutectic Carbides

During conventional destabilization or annealing treatments, the eutectic M_7C_3 carbides in 10-25 wt%Cr irons do not appear to undergo any structural change, however, in the 30 wt%Cr irons, a transition from M_7C_3 to $M_{23}C_6$ has been observed in the eutectic carbides [11, 44]. This results in duplex eutectic carbide structures consisting of cores of M_7C_3 surrounded by shells of $M_{23}C_6$. TEM investigation is needed to reveal full details of this duplex structure and to study the mechanism by which the transformation takes place. Figure 5 reveals $M_{23}C_6$ shells surrounding the remaining cores of eutectic M_7C_3 with faulting contrast. The extent of the transformation depends on destabilization temperatures and times [37]. This transformation has only been observed within the very high chromium content irons, probably because $M_{23}C_6$ carbide is stable only at the high chromium content [45]. The $M_{23}C_6$ is believed to nucleate at the original interface between the eutectic M_7C_3 and the matrix; it then grows gaining metal atoms from the matrix and consuming the M_7C_3 [37]. The behavior of alloys containing 26 wt%Cr and 30wt%Cr have been compared by Pearce [45]. It was found that, as expected, the hardness of the irons were increased after hardening heat treatment. However, the

heat-treated 26 wt%Cr iron gave 36% increase in dry wear resistance, whereas the heat-treated 30 wt%Cr iron gave only 4% increase in dry wear resistance. In wet wear conditions, it has been reported [7] that the as-cast 30 wt%Cr iron was better than the as-cast or heat-treated 25 wt%Cr irons, however, the heat-treated 30 wt%Cr iron performed surprisingly worse. It was believed [45, 46] that the lack of improvement in both dry and wet wear resistance of the 30 wt%Cr irons after destabilization and hardening-tempering treatments was due to the eutectic $M_7C_3 \rightarrow M_{23}C_6$ transformation forming the duplex eutectic carbides. These duplex carbides appear to have a greater tendency to crack and become detached during wear than single M_7C_3 carbide [45]. The corrosive wear resistance of hardened 30 wt%Cr irons appears to be inferior to that of the as-cast material, while the opposite is true for 25 wt%Cr iron where there is no evidence of a $M_7C_3 \rightarrow M_{23}C_6$ transition in the eutectic carbides during heat treatment [47, 48]. Pearce and Elwell [43] reported the Cr : Fe ratios of the 30 wt%Cr-2.44 wt%C iron by energy dispersive analysis which varied from 1.61-1.68 for the $M_{23}C_6$ shell.

Inoue [29] found the orientation relationships between the M_7C_3 carbide and the $M_{23}C_6$ as;

$$\begin{aligned} (\bar{1}2\bar{1})_{M_{23}C_6} // (0001)_{M_7C_3} \\ (111)_{M_{23}C_6} // (1\bar{1}00)_{M_7C_3} \\ (10\bar{1})_{M_{23}C_6} // (11\bar{2}0)_{M_7C_3} \end{aligned}$$

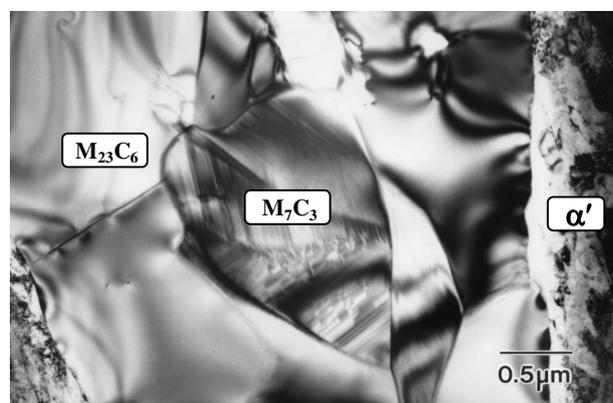


Figure 5 BF-TEM shows a duplex core-shell structure of the eutectic carbides after destabilization in 30 wt%Cr -2.3 wt%C [38].

3. SECONDARY CARBIDE

As mentioned earlier, the austenite matrix of high chromium iron can be destabilized by heat treatment in which the C and Cr contents of this matrix become reduced as a result of the precipitation of secondary carbides. This raises the martensite-start temperature (M_s) of the austenite so that after destabilization and air hardening, the microstructure consists of secondary carbides in a matrix of martensite and some retained austenite. It has been reported that secondary carbides formed during the

destabilization of Cr irons do not nucleate and grow on the eutectic carbides, but precipitated within the dendritic matrix [4, 5, 11, 44]. However, an argument has also been raised [35] that the $M_{23}C_6$ can nucleate on the M_7C_3 eutectic carbide. The precipitation occurs on preferred planes in the regions of the austenite matrix remote from the eutectic carbides [49]. Normally, it is believed that secondary carbides precipitate on slip bands or in subgrain boundaries within the austenite regions [4, 49]. Slip bands or subgrain boundaries occur due to stress, which is produced by the difference in thermal expansion between the eutectic carbides and the matrix [34]. The type of secondary carbides depends on the alloy composition and destabilization temperature.

3.1 M_3C Secondary Carbide

M_3C type carbide can also form in lower Cr irons with a plate-like or acicular morphology [11, 21, 44, 46, 49]. From Rivlin's review [50] of the equilibrium ternary C-Cr-Fe system, partial isothermal section at 870 °C and 1000 °C are shown in Figure 6(a) and (b), respectively. Powell and Laird [11] used these isotherms to predict the resulting secondary carbides formed during isothermal soaking. It was reported that after destabilization of the alloy contain 8.8 wt%Cr-3.0 wt%C, at 870 °C and 1000 °C, the secondary carbides known to precipitate in the alloy were M_3C shown in Figure 7.

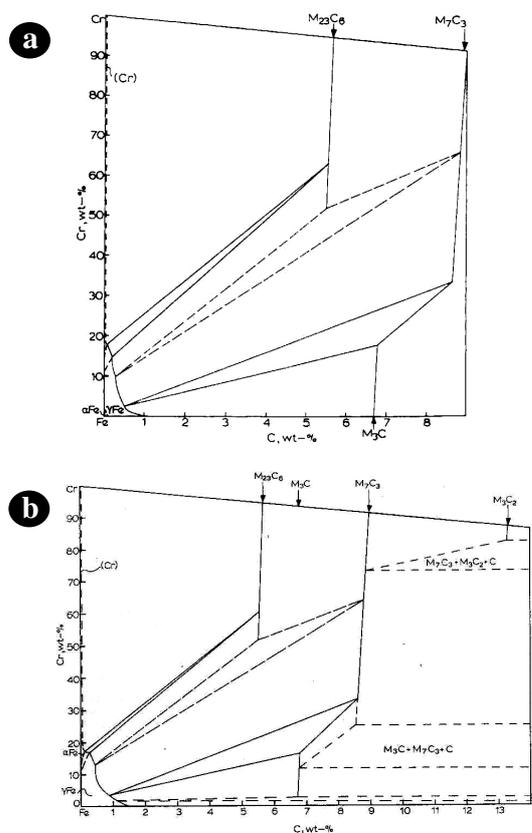


Figure 6 Solid-state isotherms of the C-Cr-Fe system at (a) 870 °C and (b) 1000 °C [50].

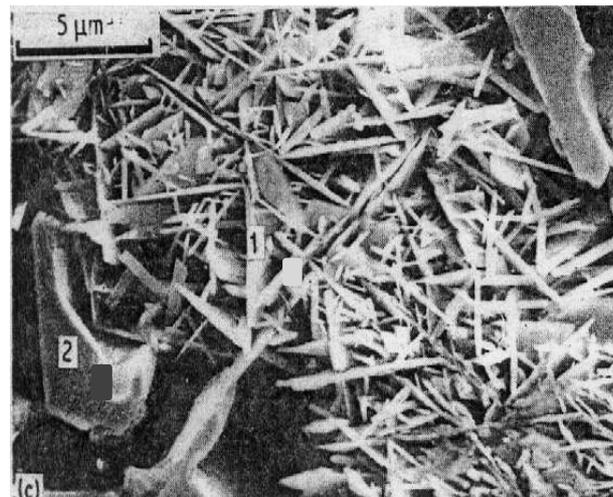


Figure 7 M_3C secondary carbides with plate-like shape in 8.8 wt%Cr-3.0 wt%C iron [11].

3.2 M_7C_3 Secondary Carbide

Pearce [5] found that, during the destabilization treatment of the 15 wt%Cr iron, the secondary M_7C_3 carbides formed with size ranging from 0.2-0.6 μm. Powell and Laird [11] reported that after destabilization of the alloy contain 17.7 wt%Cr-3.1 wt%C at 870 °C and 1000 °C, the secondary carbides were M_7C_3 as shown in Figure 8. In the conventionally cast 15 wt%Cr iron, the secondary carbides are M_7C_3 and show linear fringe contrast due to the presence of stacking faults as in eutectic M_7C_3 [5]. Dudzinski [39] reported that in the case of Cr_7C_3 , iron, molybdenum or vanadium can replace the chromium atoms in the lattice of this carbide to give $(Fe,Cr)_7C_3$ or M_7C_3 carbides.

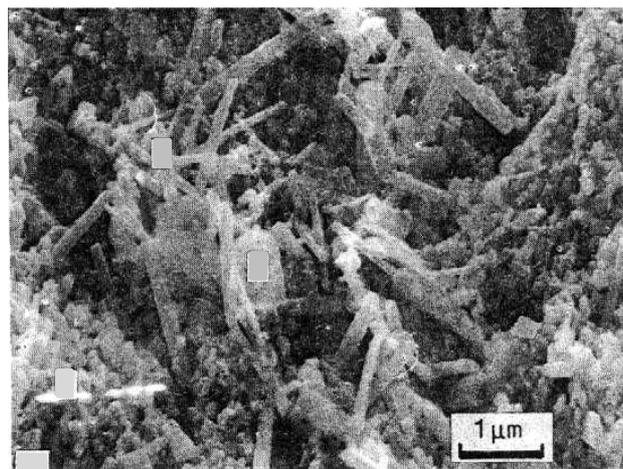


Figure 8 (a) M_7C_3 secondary carbides with rod-like shape in 17.7 wt%Cr-3.1 wt%C iron [11].

3.3 $M_{23}C_6$ Secondary Carbides

$M_{23}C_6$ carbide crystals are face-centered cubic, in which 92 metal atoms located at the 4(a), 8(c), 32(f) and 48(h) symmetry sites of space group $Fm\bar{3}m$ and with lattice parameter $a = 10.6214 \text{ \AA}$ [31, 51]. The $M_{23}C_6$ contains Cr contents with a lower limit of around 60 wt% and a C content of 5.3-5.7 wt%. The Gibbs free energy of the formation is -8 to -15 kJ/mole [7]. In the alloys with high chromium content above 25-30 wt%, $M_{23}C_6$ type carbides are formed as fine interconnected rods (fibrous shape). It was reported that after destabilization of the alloy contain 29.3 wt%Cr-2.5 wt%C at 870 °C and 1000 °C, the secondary carbides were $M_{23}C_6$ as shown in Figure 9. TEM results confirmed the type of secondary carbides in conventionally cast 30 wt%Cr iron as $M_{23}C_6$ [2, 37, 38], which has a cubic lattice and does not contain faults. In conventionally cast irons, SEM and TEM observations have shown that secondary carbides, either M_7C_3 or $M_{23}C_6$, do not nucleate on M_7C_3 eutectic carbide, but form separately within the matrix and grow at sub-boundaries in the matrix and/or along habit directions to develop fibrous network [5, 11, 38]. However, $M_{23}C_6$ secondary carbide in semi-solid cast 27 wt%Cr iron [35] was found to nucleate also on M_7C_3 eutectic carbide. In conventionally cast 18-20 wt%Cr irons, it has been shown that, as well as M_7C_3 secondary carbide, $M_{23}C_6$ carbides can also form and grow along $\{111\}$ planes in the original austenite matrix [21]. In conventionally cast 27 wt%Cr irons, both M_7C_3 and $M_{23}C_6$ cubic carbides can also found [8].

Various studies have revealed that the sizes of secondary carbides vary from 0.2-0.6 μm with both size and distribution being influenced by prior annealing and destabilization conditions, and by the segregation patterns in the original dendritic and eutectic austenite formed on solidification.

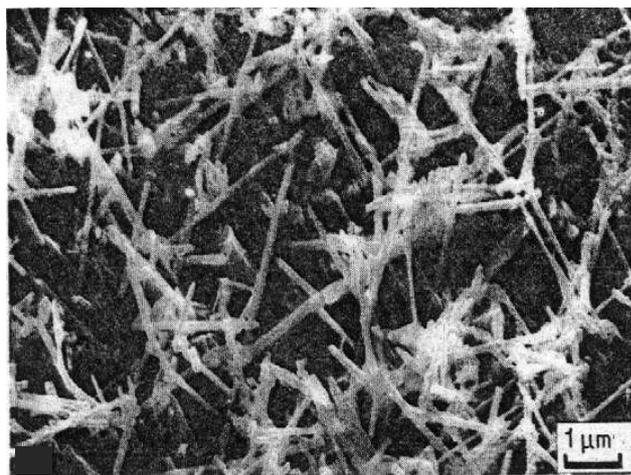


Figure 9 $M_{23}C_6$ secondary carbides with fibrous shape in 29.3 wt%Cr-2.5 wt%C iron [11].

3.4 M_6C Secondary Carbide

In conventionally cast 18-20 wt%Cr irons, M_6C secondary carbides can form together with $M_{23}C_6$ and M_7C_3 [21]. M_6C carbide crystal is FCC with lattice parameters $a = 11.0823 \text{ \AA}$, point group $m\bar{3}m$ and space group $Fd\bar{3}m$ [31, 33]. Kibble [52] found that both M_6C and M_7C_3 secondary carbides were precipitated in 19 wt%Cr iron after annealing at 800 °C and followed by destabilization at 1000 °C. Inoue [29] found the transformation of $M_{23}C_6$ carbide to M_6C carbide in 4 wt%C-18 wt%Cr with 8 wt%W or 8 wt%Mo steels during tempering at 700 °C for longer time than about 1 hour. The orientation relationships among $M_{23}C_6$ to M_6C were;

$$\begin{aligned} (\bar{1}10)_{M_{23}C_6} // (1\bar{1}0)_{M_6C} \\ (\bar{1}\bar{1}1)_{M_{23}C_6} // (11\bar{1})_{M_6C} \\ (112)_{M_{23}C_6} // (112)_{M_6C} \end{aligned}$$

4. EFFECTS OF ADDITIONAL ALLOYING ELEMENTS

Irons which are required to be predominantly austenitic but which have insufficient as-cast hardenability to avoid pearlite formation are described as under-alloyed. To prevent this, the correct level of austenite-stabilising elements must be added [12]. For heat treated irons, since chromium is present in both the eutectic and secondary carbides, only part of the total Cr content of the iron is retained in the matrix to increase hardenability and supplementary addition of Mo, Ni, Cu and Mn are needed if heavier sections are to be fully hardened [2, 4, 7].

The effects of additional alloying elements in Cr irons have been extensively studied [53-57]. Normally, alloying additions such as molybdenum, manganese, nickel and copper are used to increase hardenability and to prevent pearlite formation in the thicker sections of castings. Irons alloyed with carbide-forming elements such as molybdenum, vanadium and tungsten have been developed for special applications such as hot working mill rolls in the steel industry [4, 13]. Overalloying must be avoided since too high content of alloying element increases both the time for destabilization process and the levels of retained austenite in the structure, the latter resulting in lower abrasion resistance. Under-alloying is equally as damaging since too low a content of alloying element encourages pearlite formation in both as cast irons and after hardening [12].

5. CONCLUSIONS

Wear resistance and mechanical properties of high chromium cast irons depend on the eutectic and precipitated carbides within their microstructures and on the nature of the matrix supporting the hard eutectic carbides. Table 2 gives a conclusion of the types, crystallography, morphologies, and hardness of carbides in high chromium cast irons, and comparison of the hardness of some abrasive materials in Table 3 [4, 7].

Table 2 Types, crystallography, morphologies, and hardness of carbides in high chromium cast irons.

Carbide, abrasive materials	Crystal structure	Morphology	Hardness (HV)
M ₃ C	Orthorhombic Lattice parameters $a = 5.060 \text{ \AA}$, $b = 6.739 \text{ \AA}$, $c = 4.499 \text{ \AA}$	Continuous plate	800-1100
M ₆ C	FCC Lattice parameters $a = 11.0823 \text{ \AA}$ Point group $m\bar{3}m$ Space group $Fd\bar{3}m$	Facet and single rod	1200-1800
M ₇ C ₃	Hexagonal Lattice parameters $a = 13.982 \text{ \AA}$, $c = 4.506 \text{ \AA}$ Point group mmm Space group $Pm\bar{3}n$	Hexagonal rod and blade-like	1000-1800
M ₂₃ C ₆	FCC Lattice parameter $a = 10.6214 \text{ \AA}$ Point group $m\bar{3}m$ Space group $Fm\bar{3}m$	Irregular shapes, rod, needle, plates, cubic	1000

Table 3 Hardness of some abrasive materials.

Abrasive materials	Hardness (HV)
SiC	2600
VC	2000-3000
Al ₂ O ₃	1800
silica	1430
high carbon martensite	500-1010

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