

# The influence of vanadium on fracture toughness and abrasion resistance in high chromium white cast irons

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The influence of vanadium on wear resistance under low-stress conditions and on the dynamic fracture toughness of high chromium white cast iron was examined in both the as-cast condition and after heat treatment at 500 °C. A vanadium content varying from 0.12 to 4.73% was added to a basic Fe–C–Cr alloy containing 2.9 or 19% Cr. By increasing the content of vanadium in the alloy, the structure became finer, i.e. the spacing between austenite dendrite arms and the size of massive  $M_7C_3$  carbides was reduced. The distance between carbide particles was also reduced, while the volume fraction of eutectic  $M_7C_3$  and  $V_6C_5$  carbides increased. The morphology of eutectic colonies also changed. In addition, the amount of very fine  $M_{23}C_6$  carbide particles precipitated in austenite and the degree of martensitic transformation depended on the content of vanadium in the alloy. Because this strong carbide-forming element changed the microstructure characteristics of high chromium white iron, it was expected to influence wear resistance and fracture toughness. By adding 1.19% vanadium, toughness was expected to improve by approximately 20% and wear resistance by 10%. The higher fracture toughness was attributed to strain-induced strengthening during fracture, and thereby an additional increment of energy, since very fine secondary carbide particles were present in a mainly austenitic matrix. An Fe–C–Cr–V alloy containing 3.28% V showed the highest abrasion resistance, 27% higher than a basic Fe–C–Cr alloy. A higher carbide phase volume fraction, a finer and more uniform structure, a smaller distance between  $M_7C_3$  carbide particles and a change in the morphology of eutectic colonies were primarily responsible for improving wear resistance.

## 1. Introduction

High chromium white cast irons are an important class of wear-resistant materials. A volume of approximately 25–30% carbide resulted in a balanced combination of fracture toughness and abrasive wear resistance [1]. There is a general consensus [1–6] that optimum resistance to abrasion and resistance to spalling and fracture under conditions of severe repeated impact are obtainable in irons that have undergone heat treatment to produce a martensitic matrix. Some investigations [7, 8] have shown, however, that the greater the quantity of retained austenite in the microstructure, the higher the impact-abrasive wear resistance. In certain applications the presence of retained austenite in the matrix is even beneficial, since it imparts a greater degree of toughness to the casting, but with only a corresponding slight loss in abrasion resistance [1, 9]. The demand for longer service life has generated an interest in improving the quality of these wear-resistant alloys.

It has been reported that the addition of strong carbide-forming elements, such as vanadium, tung-

sten, niobium and titanium, improve the mechanical properties of high chromium white irons [10–16]. Vanadium improves abrasion resistance [15, 16], but does not substantially change fracture toughness [15]. Alloying high chromium iron with vanadium makes the structure finer [15]. Nevertheless, the volume fraction of eutectic  $M_7C_3$  carbides changes [17, 18].

Some studies [17] conclude that a higher content of vanadium increases the amount of  $M_7C_3$  carbides, while others [18] conclude a decrease in the amount of carbides. Furthermore, some investigations [15] show that vanadium changes the morphology of  $M_7C_3$  carbides, while others [18] indicate the absence of morphological changes. Where the content of vanadium exceeds 3% [15], i.e. in excess of 5% [18, 19], vanadium carbides of the VC type are formed.

Vanadium is soluble in eutectic  $M_7C_3$  carbides as well as in austenite [18] and influences the transformation of austenite in high chromium iron [16, 19]. In lower concentrations, vanadium effectively prevents pearlitic transformation [16]. In alloys containing 2–3% V, austenite largely transforms into bainite

[19]. Where the vanadium content is higher than 4%, precipitation of dispersive secondary carbides of the VC type [16, 19] in austenite are observed, which is favourable for martensitic transformation [16].

In alloys containing 25–30% Cr and over 5% V, the matrix microstructure is ferritic, which reduces hardness and wear resistance [16, 19, 20]. The tendency of oxide formation also increases [16].

In addition to the fact that data are often in disagreement and to a large measure based on suppositions, there is little information available on the influence of vanadium on fracture toughness or how much of this element should be combined with high chromium white iron to obtain optimum fracture toughness and abrasive wear resistance. In this paper, the influence of vanadium on the dynamic fracture toughness and abrasive wear resistance of high chromium white cast iron under low-stress conditions is examined.

## 2. Experimental procedure

Varying contents of vanadium, namely, 0.12, 0.49, 1.19, 2.02, 3.28 and 4.73%, were added to a basic Fe–C–Cr alloy containing 2.89% C, 19.03% Cr, 0.48% Mo, 0.99% Cu, 0.85% Si and 0.71% Mn. An induction furnace was used for melting. Samples (bars 200 mm long and 25 mm in diameter,  $10 \times 20 \times 400 \text{ mm}^3$  test blocks for abrasion tests and  $13 \times 13 \times 100 \text{ mm}^3$  test blocks for fracture toughness) were cast in sand moulds. Samples for structural analysis and hardness were selected from among the bars.

Samples were heat treated in an electric furnace with no protective atmosphere at  $500^\circ\text{C}$  for 4 h, followed by cooling to room temperature in still air.

The microstructure was examined using conventional optical microscopy. The size and volume fraction of phases were determined by semi-automatic image analysis using a MOP-Videoplan analyser. The morphology of eutectic  $\text{M}_7\text{C}_3$  carbides was examined by a scanning electron microscope, JEOL 733-FCXA, using an accelerating voltage of 25 kV. For this examination, the polished samples were deep-etched in a 10% HCl solution in methanol for 24 h then cleaned in an ultrasonic bath.

$\text{CuK}_\alpha$  radiation was used for measuring the amount of retained austenite by means of X-rays of flat specimens. At a scanning rate of  $1^\circ \text{min}^{-1}$ , the integrated intensities under the peaks  $(200)\alpha$ ,  $(220)\gamma$  and  $(311)\gamma$  were measured by a diffractometer.

Abrasive wear resistance was evaluated by measuring the mass loss in as per the procedure described in ASTM Standard Practice G-65, Procedure B (Rubber Wheel Abrasion Test) [21]. The volume loss,  $\Delta V$ , was then calculated by dividing mass loss by the alloy density ( $7.8 \text{ g cm}^{-3}$  [22]). The reciprocal value of the volume loss,  $\Delta V$ , due to wear is called wear resistance,  $\Delta^{-1}$ .

Dynamic fracture toughness,  $K_{Ic}$  was measured using an impact test machine equipped with an instrumented Charpy tub. The standard Charpy specimen was notched to electrical discharge machining (EDM) a 0.2 mm radius slot, 2 mm deep. The fracture load

was used to calculate  $K_{Ic}$  values using the equation for bend specimens provided in ASTM E399-90 [23].

## 3. Results

The microstructure of the alloys examined in as-cast and heat-treated conditions are shown in Figs 1–5. In addition to primary austenite dendrites, eutectic composed of  $\text{M}_7\text{C}_3$  carbides and austenite,  $\text{V}_6\text{C}_5$  carbides were also present in the structure of Fe–C–Cr–V alloys [24].

The volume fraction of primary austenite dendrites was reduced, whereas the amount of  $\text{M}_7\text{C}_3$  and  $\text{V}_6\text{C}_5$  carbides increased with a higher content of vanadium in the alloy (Fig. 6).

Vanadium also influenced the refinement of the structure of high chromium iron as indicated by measuring the spacing of austenite dendrite arms and the size of eutectic  $\text{M}_7\text{C}_3$  carbides (Figs 7–10).

The linear method was used to determine the dendrite arm spacing in the alloys examined. The average values obtained by measuring 500 dendrites (microscopic magnification of  $\times 500$ ) are shown in Fig. 7. The higher the vanadium content in an alloy, the lower average value of dendrite arm spacing. The dispersion of values also decreased, as indicated in the distribution histogram of the length of dendrite arms' segments by classes (Fig. 8).

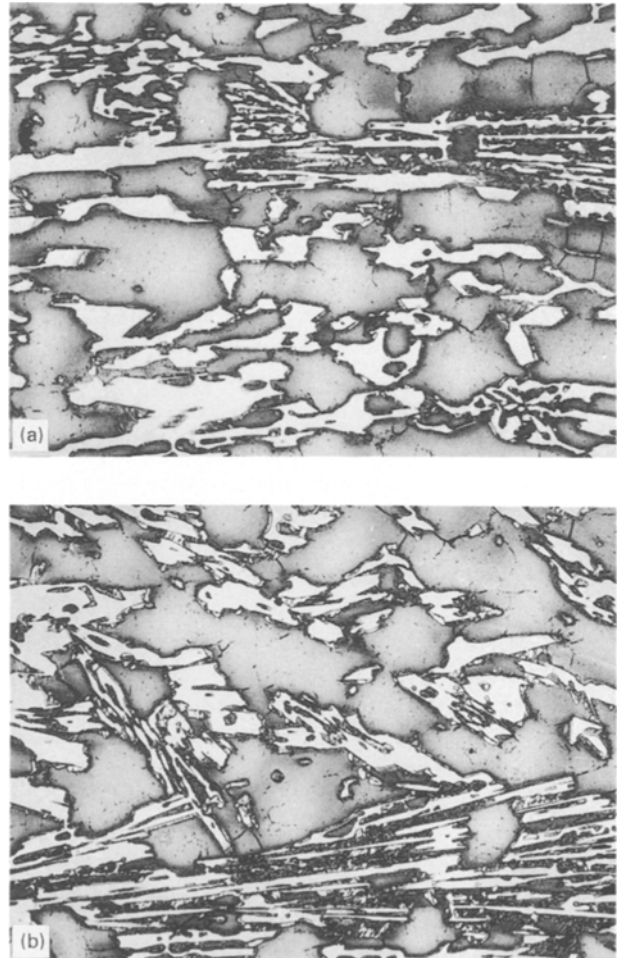


Figure 1 Microstructure of basic Fe–C–Cr alloy: (a) as-cast; (b) heat-treated  $500^\circ\text{C}/4 \text{ h}$  (magnification  $\times 390$ ).

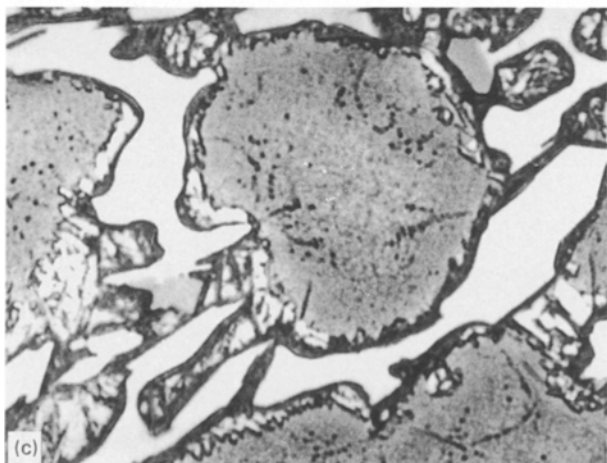
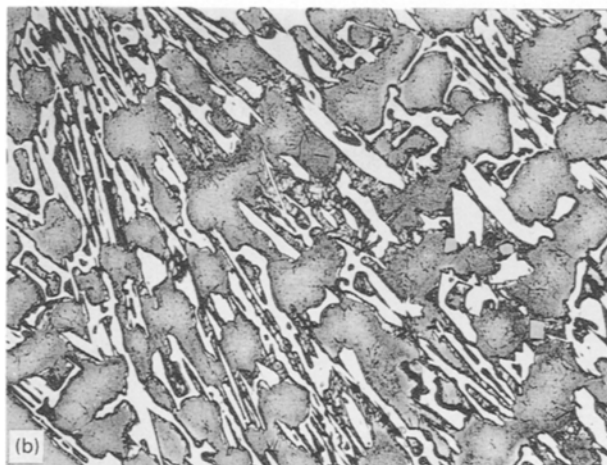
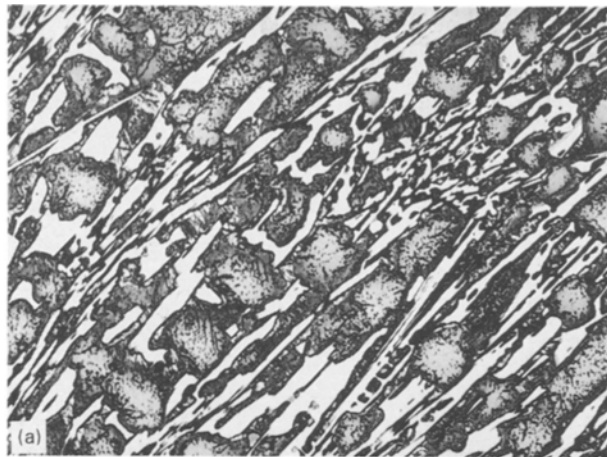


Figure 2 Microstructure of basic Fe-C-Cr-V alloy containing 0.49% V: (a) as-cast; (b) heat-treated 500°C/4 h (magnification  $\times 390$ ); (c) heat-treated 500°C/4 h (magnification  $\times 1200$ ).

Fig. 9 shows the size of massive  $M_7C_3$  carbides as a function of vanadium content in the alloy. Area of 2000 carbide particles (microscopic magnification of  $\times 500$ ) was measured using a planimetric method. The higher the vanadium content the lower average area value of carbide (Fig. 9). By decreasing the average carbide size value, in addition to an increase in the number of fine carbides in the structure, the dispersion of values was reduced (Fig. 10).

The shape of eutectic  $M_7C_3$  carbides present in the structure of Fe-C-Cr alloys and Fe-C-Cr-V alloys con-



Figure 3 Microstructure of basic Fe-C-Cr-V alloy containing 2.02% V: (a) as-cast; (b) heat-treated 500°C/4 h (magnification  $\times 390$ ).

taining 1.19 and 4.73% V in two- and three-dimensional space, is shown in Fig. 11. The scanning electron microphotographs of deeply edged samples revealed that single  $M_7C_3$  carbides in all the alloys examined were rod or blade shaped. Vanadium, however, changed the morphology of eutectic colonies (Fig. 11) whose shapes depended mainly on the amount and shape of austenitic dendrites.

A larger number of rods usually also formed sheafs in different directions. Sheafs of rods and blades interlaced among one another to form a continuous carbide network. In alloys containing 4.73% V, regions in which the border between carbide rod sheafs of different directions was clear, could also be observed (Fig. 11f).

The section plane form of eutectic  $M_7C_3$  carbides can be quantitatively defined by the form factor,  $\phi$ , which has been calculated using the following equation [25]:

$$\phi = 4\pi A/L_p^2$$

where  $A$  is the area of carbides and  $L_p$  is the perimeter of carbides.

Fig. 9 shows the average area value of carbide,  $\bar{A}$ , and average value of perimeter,  $\bar{L}_p$ . These factors decreased, although the average form factor value increased slightly with a higher vanadium content in

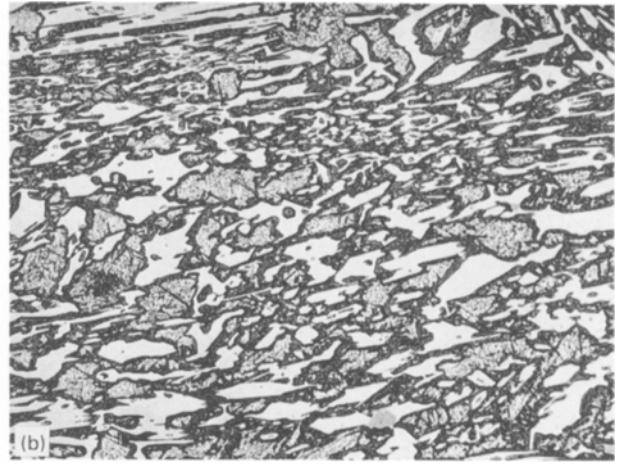
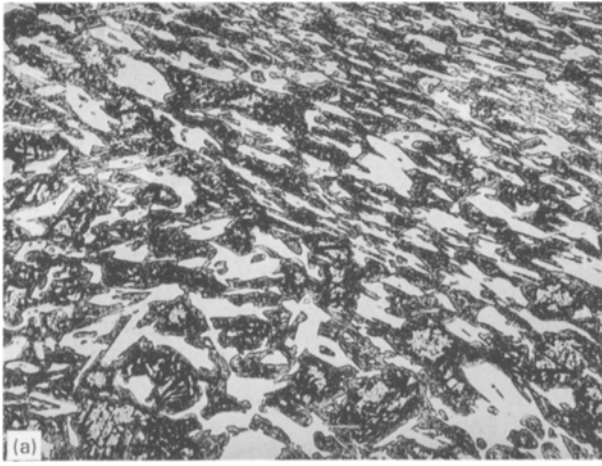


Figure 4 Microstructure of basic Fe-C-Cr-V alloy containing 3.28% V: (a) as-cast; (b) heat-treated 500°C/4 h (magnification  $\times 390$ ).

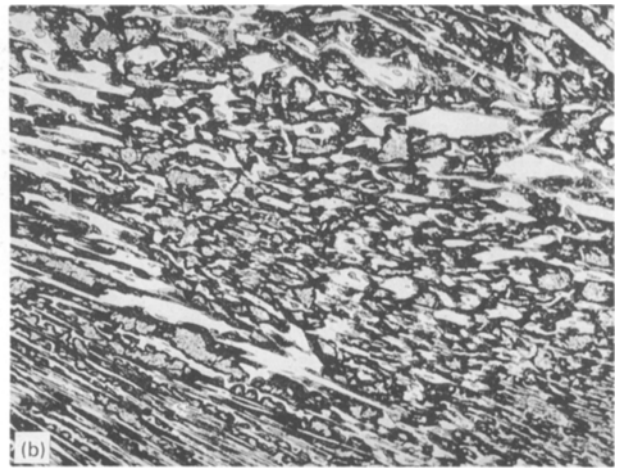
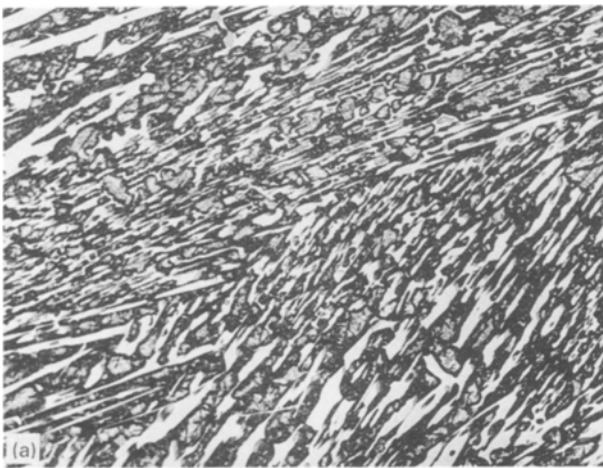


Figure 5 Microstructure of basic Fe-C-Cr-V alloy containing 4.73% V: (a) as-cast; (b) heat-treated 500°C/4 h (magnification  $\times 390$ ).

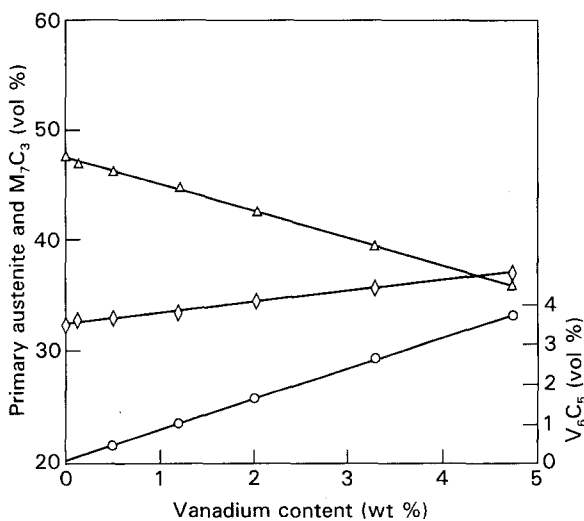


Figure 6 Volume fraction of phases present in the structure of examined alloys.  $\diamond-\diamond$ ,  $V_6C_5$ ;  $\triangle-\triangle$ , primary austenite;  $\circ-\circ$ ,  $M_7C_3$ .

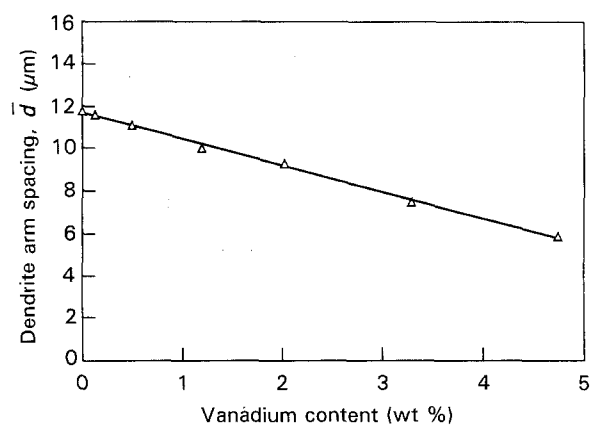


Figure 7 Austenite dendrite arm spacing,  $d$ , as a function of vanadium content in the alloy.

the alloy. It must be specified that as the size of carbides decreased, the perimeter value showed a larger influence on the form factor and contributed to its greater value. In any case, on the basis of the average form factor values obtained, it can be said that the

shape of eutectic carbides were, for the most part, elongated.

Distribution histograms of the form factor of eutectic  $M_7C_3$  carbides by class are presented in Fig. 12. In all the alloys examined there was a wide range of form factor values. The wide dispersion of carbide cross-section shapes indicated different directions of rod sheafs in the space.



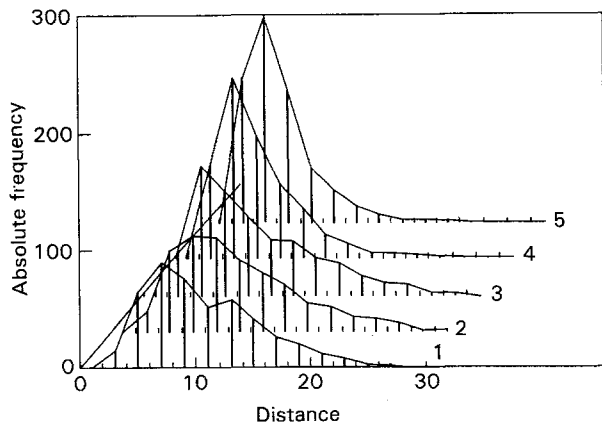


Figure 8 Distribution histograms of austenite dendrite arm spacing (1, Fe-C-Cr; 2, Fe-C-Cr-V with 0.12% V; 3, Fe-C-Cr-V with 1.19% V; 4, Fe-C-Cr-V with 3.28% V; 5, Fe-C-Cr-V with 4.73% V).

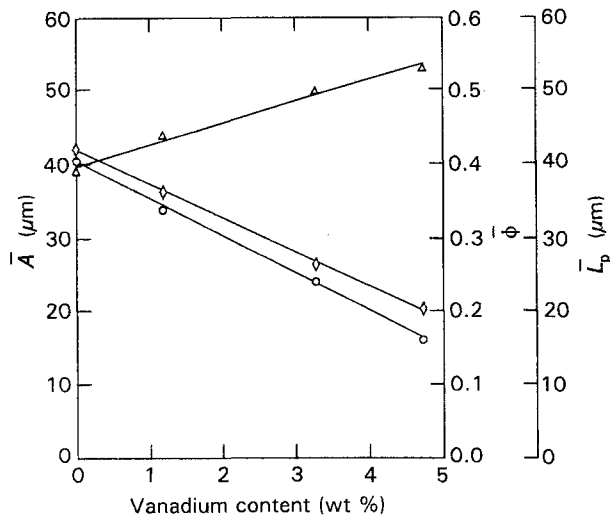


Figure 9 Mean area ( $A$ ), perimeter ( $L_p$ ) and form factor ( $\phi$ ) of eutectic carbides as a function of vanadium content in the alloy.  $\circ$ — $\circ$ ,  $A$ ;  $\diamond$ — $\diamond$ ,  $L_p$ ;  $\triangle$ — $\triangle$ ,  $\phi$ .

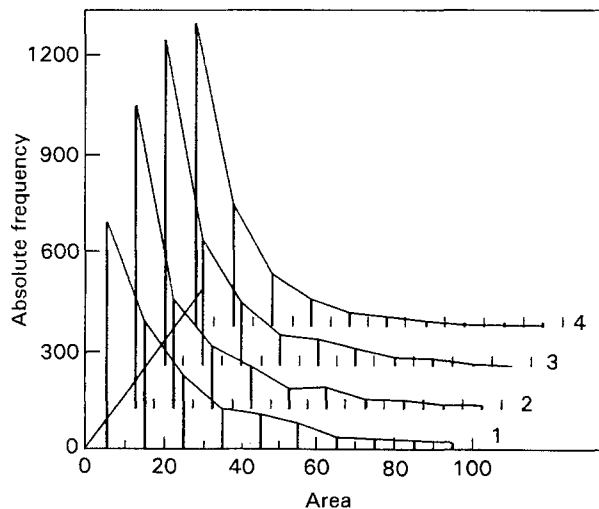


Figure 10 Distribution histograms of area of eutectic carbides (1, Fe-C-Cr; 2, Fe-C-Cr-V with 0.12% V; 3, Fe-C-Cr-V with 1.19% V; 4, Fe-C-Cr-V with 3.28% V; 5, Fe-C-Cr-V with 4.73% V).

The mean free path,  $\lambda$ , which described the distribution of eutectic  $M_7C_3$  carbides in the space, could be calculated using the following equation [26]:

$$\lambda = L_3(1 - V_v)/V_v$$

where  $L_3$  is the size of carbides and  $V_v$  is the volume fraction of carbides.

An increase in the volume fraction of eutectic carbides and a decrease in size, due to a higher content of vanadium in the alloy, would reduce the mean free path, i. e. the average distance between carbides (Fig. 13).

The matrix microstructure of a basic Fe-C-Cr alloy is mainly austenitic. Only eutectic austenite was partially transformed into martensite (Fig. 1a). As micrographs (Figs 2a, 3a, 4a and 5a) reveal, large numbers of

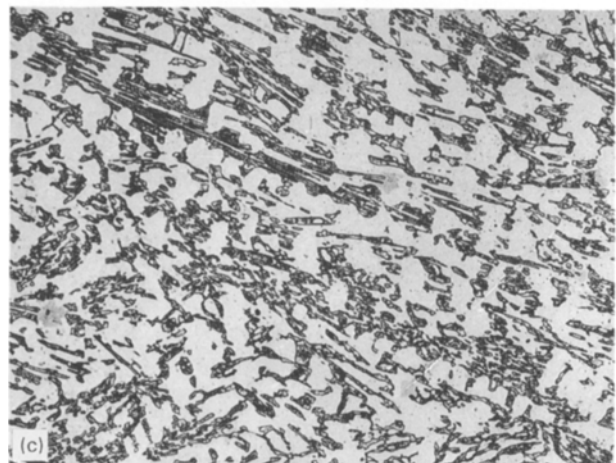
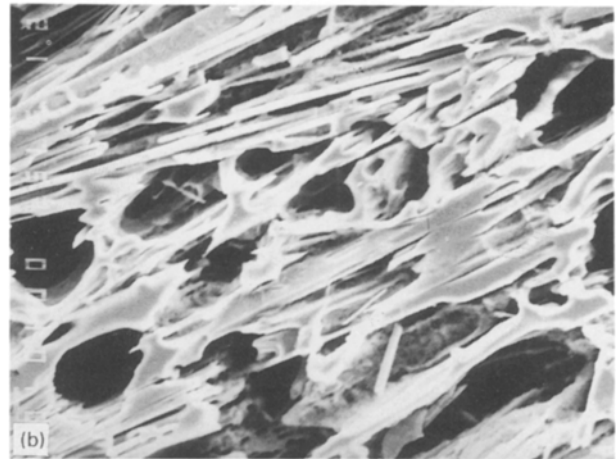
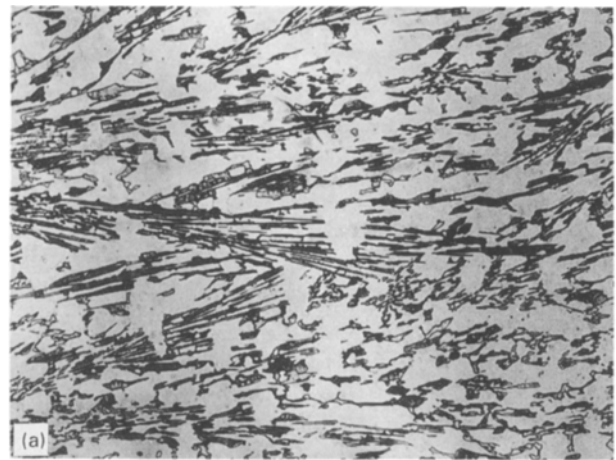


Figure 11 Morphology of eutectic carbides: Fe-C-Cr: (a) OM and (b) SEM Fe-C-Cr-V with 1.19% V; (c) OM and (d) SEM Fe-C-Cr-V with 4.73% V; (e) OM and (f) SEM (magnifications: (a), (c) and (e),  $\times 156$ ; (b), (d) and (f),  $\times 624$ ).

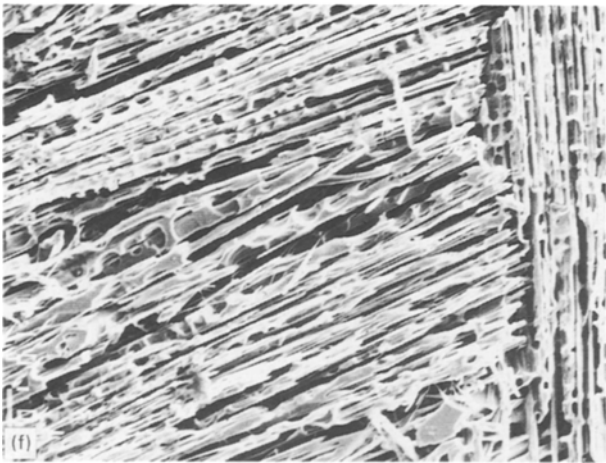
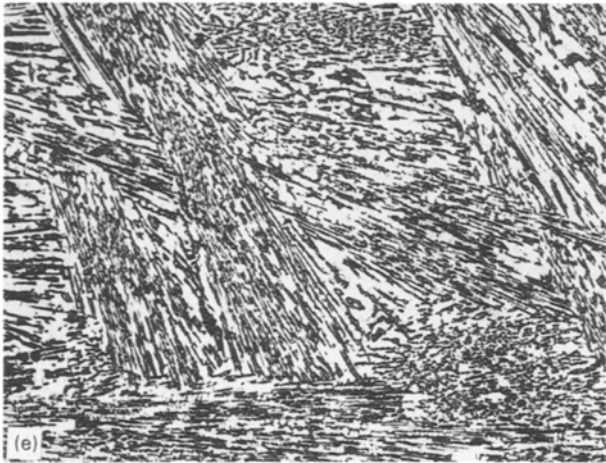
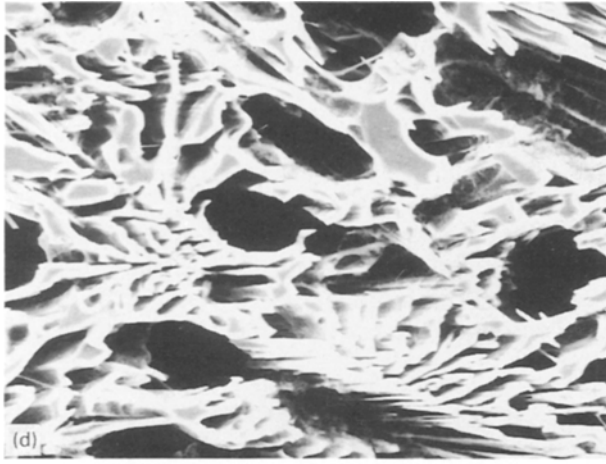


Figure 11 Continued.

fine particles were present in the austenite of the Fe-C-Cr-V alloys examined. On the basis of previous investigations [24], transmission electron microscopy established that these particles were carbides of  $M_{23}C_6$  type (ca. 40–100 nm). Apart from eutectic austenite, primary austenite had also, more or less, transformed into martensite in these alloys (Figs 2a, 3a, 4a and 5a). Therefore, an increase in the concentration of vanadium in the alloy (Fig. 14) would reduce the volume fraction of retained austenite.

After heat treatment there were no notable changes in the structure of Fe-C-Cr alloys (Fig. 1b). However,

during isothermal holding at 500 °C for 4 h, secondary carbides precipitated in the austenite of the Fe-C-Cr-V alloys examined. Martensite present in the as-cast structure was also tempered at this temperature [27]. Austenite was then partly transformed into martensite during the cooling process (Figs 2b and c, 3b, 4b and 5b). The amount of precipitated carbides and the volume fraction of martensite which formed depended on the basic as-cast structure, i.e. on vanadium content in the alloy.

The matrix microhardness, and hardness of the alloy on the whole, were improved by increasing the vanadium content in both the as-cast and heat treated state (Figs 15 and 16). The influence of vanadium on abrasion resistance and dynamic fracture toughness are presented in Figs 17 and 18. Wear resistance improved as the concentration of vanadium increased up to 3.28%, and fell off thereafter. Abrasion wear resistance was slightly higher after heat treatment at 500 °C (Fig. 17). An alloy containing 1.19% V in the as-cast condition showed the highest fracture toughness (Fig. 18).

#### 4. Discussion

Vanadium changed the microstructure parameters of phases present in the structure of the alloys examined, including volume fraction, size, morphology and distribution. The degree of martensitic transformation also depended on the content of V in the alloy. Vanadium therefore altered the microstructure characteristics of high chromium white iron and affected its properties.

The improved hardness of the Fe-C-Cr-V alloys examined by increasing vanadium content in the as-cast condition (Fig. 16) was the result of an increased volume fraction of the carbide ( $M_7C_3$ ) and  $V_6C_5$  phase (Fig. 6), and structural refinement (Figs 7–10). Nevertheless, an increase in the amount of precipitated carbides in austenite and the formation of a larger amount of martensite, combined with a reduction in the volume fraction of retained austenite (Figs 1–5), improved the matrix microhardness (Fig. 15) and alloy macrohardness (Fig. 16) in both the as-cast and heat-treated state.

Wear resistance improved as hardness increased in Fe-C-Cr-V alloys containing up to 3.28% V. However, alloys containing 4.73% V showed poorer wear resistance (Fig. 17) but greater hardness (Fig. 16) than alloys containing 3.28% V. This suggests that hardness was not a sufficient indicator of a material's wear resistance.

Increasing the volume fraction of eutectic  $M_7C_3$  carbides up to 36% in the Fe-C-Cr-V alloys examined (up to 30% in hypoeutectic Fe-C-Cr alloys [1]) improved abrasion resistance. The abrasion resistance of the carbide phase was more effective than the matrix in high chromium white irons since, among other things, the hardness of carbides (1200–1600  $H_V$  [3]) was greater than the hardness of the abrasive used (960  $H_V$  [28]). In high chromium white cast irons the matrix was preferentially worn by a cutting action [29]. Eutectic  $M_7C_3$  carbides were then forced to

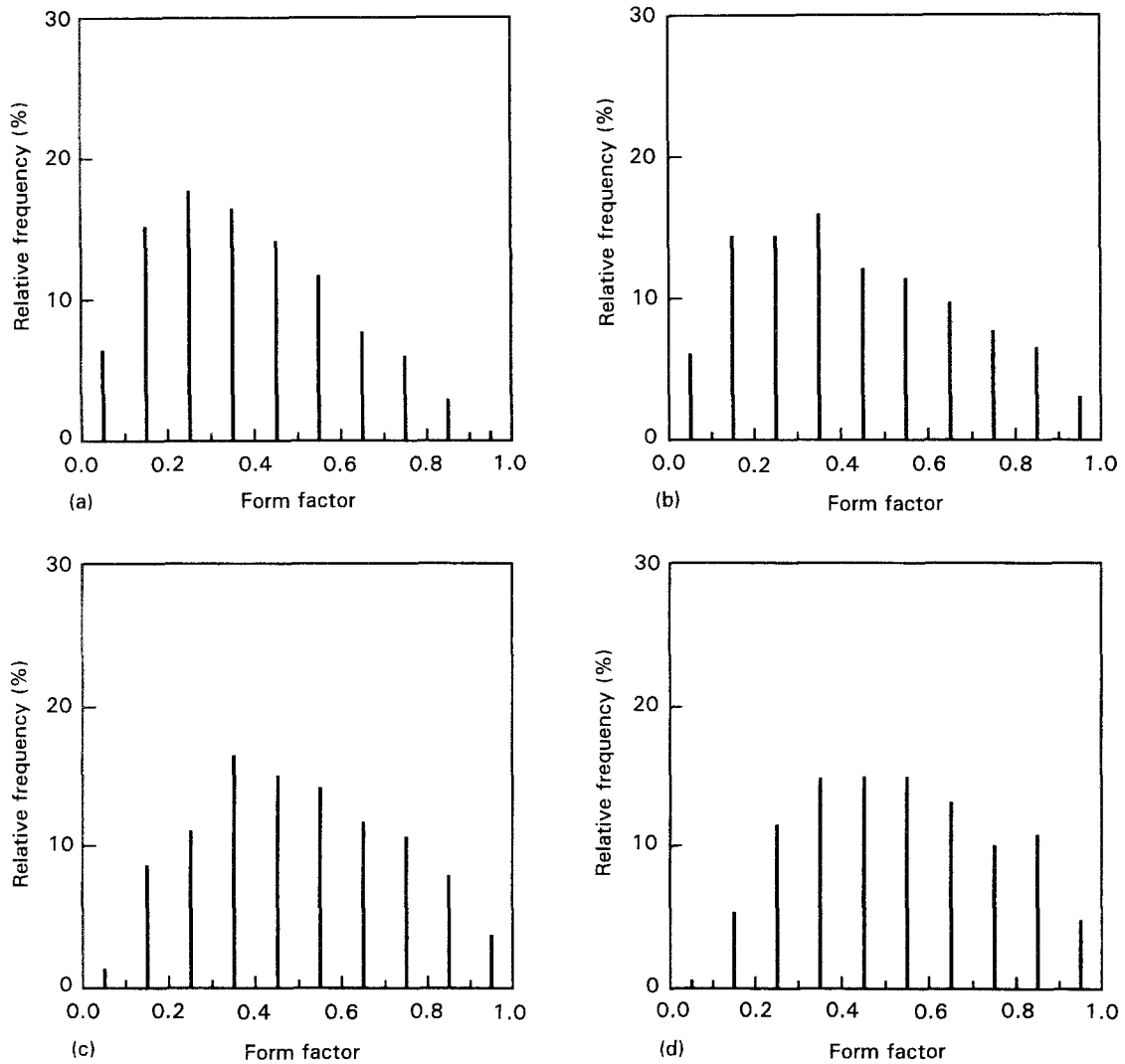


Figure 12 Distribution histograms of form factor: (a) Fe-C-Cr; (b) Fe-C-Cr-V with 1.19% V; (c) Fe-C-Cr-V with 3.28% V and (d) Fe-C-Cr-V with 4.73% V.

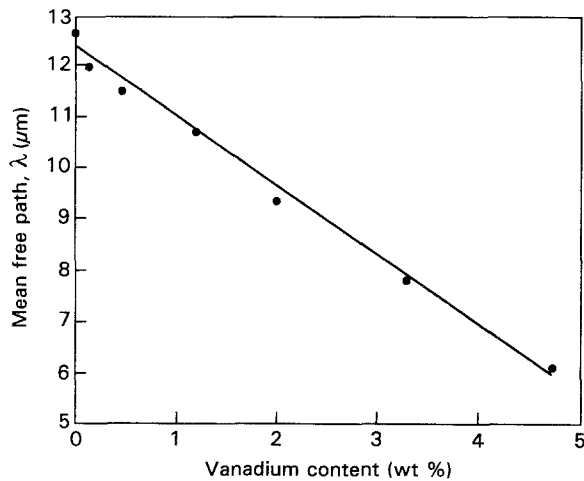


Figure 13 Mean free path of eutectic carbides ( $\lambda$ ) as a function of vanadium content of the alloy.

stand in relief and directly obstruct the cutting action of abrasive particles for a period of time until they were partly or completely removed by a cutting or chipping action. In other words, wear under low-stress abrasion conditions, and using a quartz abrasive, was apparently controlled by the rate of removal of the carbide phase since the protruding carbides protected

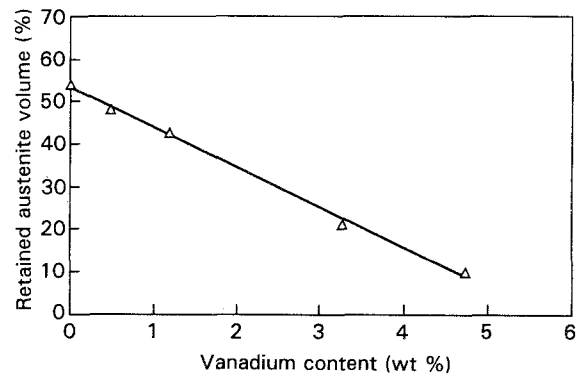


Figure 14 Volume fraction of retained austenite as a function of vanadium content in the alloy.

the matrix from direct attack of abrasive particles. Therefore, increasing the volume fraction of eutectic  $M_7C_3$  carbides, up to 36% in the Fe-C-Cr-V alloys examined, and simultaneously increasing the amount of very hard  $V_6C_5$  carbides (2800  $H_V$  [3]) (Fig. 6) reduced volume loss caused by abrasive wear (Fig. 17). However, a further increase in the volume fraction of the carbide phase (alloy with 4.73% V) (Fig. 6) reduced wear resistance (Fig. 17), which was most likely attributable to intensive spalling of massive carbides during wear. A break in the continuity of the carbide

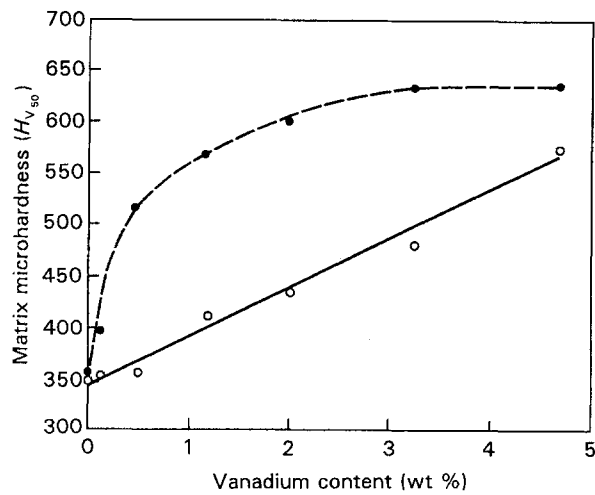


Figure 15 Matrix microhardness as a function of vanadium content in the alloy. ○—○, As-cast; ●—●, heat treated.

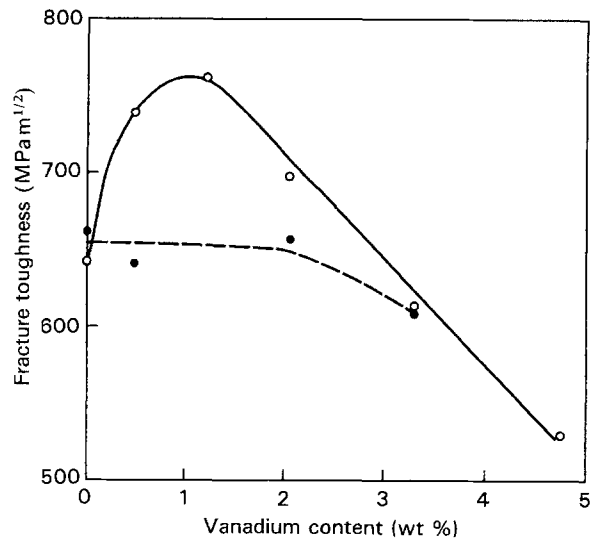


Figure 18 Fracture toughness as a function of vanadium content in the alloy. ○—○, As cast; ●—●, heat treated.

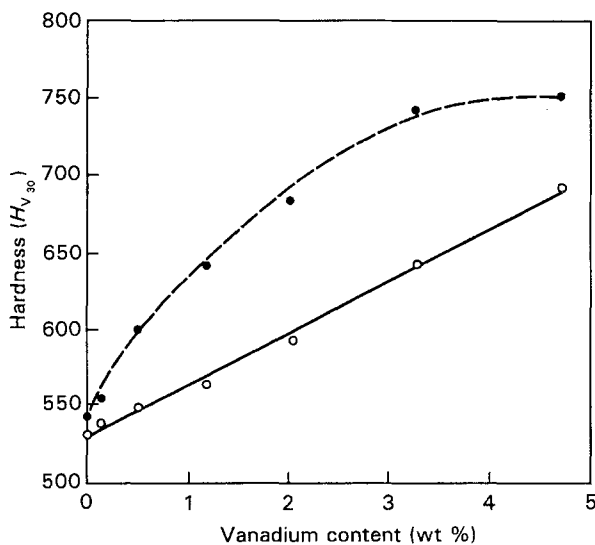


Figure 16 Hardness as a function of vanadium content in the alloy. ○—○, As cast; ●—●, heat treated.

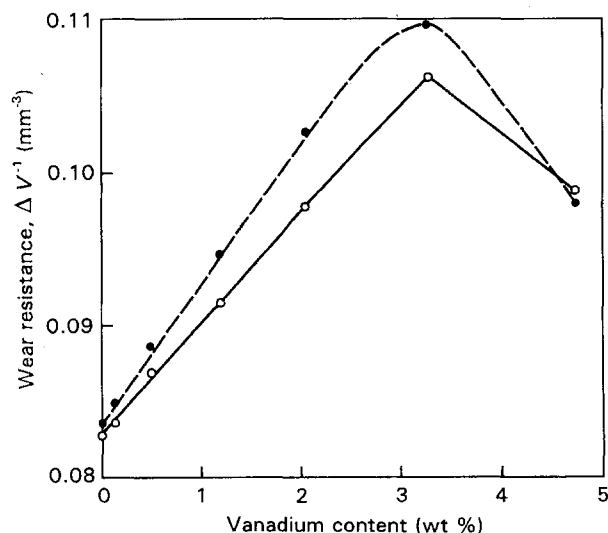


Figure 17 Wear resistance as a function of vanadium content in the alloy. ○—○, As-cast; ●—●, heat treated.

network observed in this alloy (Fig. 11f) also contributed to reduced wear resistance.

In addition to the volume fraction of carbides, the size of phases present in the structure was another microstructure variable which affected the abrasion resistance of the Fe-C-Cr-V alloys examined. The smaller size of primary austenite dendrites (Fig. 7) and the mean free path, i.e. the average distance between carbide particles (Fig. 13) caused by increasing the content of vanadium in the alloy, better protected the matrix from direct attack by abrasive particles. The shorter rods of  $M_7C_3$  carbides (Fig. 9) also tended to be less severely cracked during wear, thereby improving the wear resistance of the alloys. It must be specified, however, that the effect of controlling wear rate by refining the microstructure depended on abrasive grit size, i.e. the ratio of mean free path to mean abrasive asperity size [30]. The smaller the ratio, the less likely that grit would undermine the hard particles by penetrating the matrix.

The improved wear resistance of the Fe-C-Cr-V alloys examined after heat treatment, compared with the as-cast condition (Fig. 17), indicated that wear resistance under low-stress abrasion conditions also depended on the matrix microstructure. In addition to the fact that the matrix helped control the penetration depth of abrasive particles, it also played an important role in preventing bodily removal of smaller carbides and cracking of massive ones [31]. Experimental results indicate that the martensitic or martensite-austenitic matrix microstructure more adequately reinforced  $M_7C_3$  carbides to minimize cracking and removal during wear than the austenitic matrix. Furthermore, a matrix of the Fe-C-Cr-V alloys examined containing secondary carbide precipitation within the austenite promoted higher abrasion resistance. The secondary carbide precipitation increased the yield strength of the matrix, thereby giving the eutectic carbides greater mechanical support.

Brittle failure involved crack initiation and propagation, the latter of which was controlled by fracture toughness. Previous investigations [32] of the fracture



surfaces of white cast irons by scanning electron microscopy revealed three main locations of crack propagation that resulted in fracture: in the matrix, along the interfaces of massive carbides and the matrix, and on the cleavage planes of massive carbides. Since the crack moved easily through eutectic carbides [29], due to a higher content of vanadium in the alloy (Fig. 6), increasing the volume fraction of the carbide phase reduced fracture toughness. On the other hand, reducing the size of massive carbide particles (Fig. 9) improved toughness.

Eutectic carbide–matrix interfaces and eutectic carbides themselves offered little resistance to crack growth. As the crack moved from one eutectic carbide particle to the next it was blunted by the tougher matrix. Therefore, the fracture toughness of high chromium white cast iron depended significantly on the average spacing between eutectic carbides [33]. By increasing the vanadium content in the alloys examined, the mean free path was reduced (Fig. 13). The smaller the interparticle spacing between carbides the smaller the austenite ligament length through which a growing crack had to pass. Hence, a smaller spacing provided greater toughness.

However, the results of fracture toughness tests in both as-cast and heat-treated conditions (Fig. 18) show that the dynamic fracture toughness in the Fe–C–Cr–V alloys examined was determined mainly by the properties of the matrix. A primary role of the matrix in the fracture process in high chromium white cast irons was to prevent brittle cracks from propagating from one carbide particle to another. The matrix therefore had a crack-blunting effect, which subsequently increased the critical stress-intensity factor required to continue crack propagation [30]. Austenite was more effective in this respect than martensite [1]. By increasing the vanadium content the amount of retained austenite decreased (Fig. 14), which subsequently reduced toughness. However, the Fe–C–Cr–V alloys examined containing up to 2.02% V in the as-cast condition, showed greater dynamic fracture toughness (approximately 20%, alloys containing 1.19% V) when compared to the basic Fe–C–Cr alloy (Fig. 18). Fracture toughness was determined mainly by the energy that had to be consumed in extending the crack through ligaments of matrix [34, 35]. Since the austenite in these alloys contained very fine coherent or partly coherent  $M_{23}C_6$  carbide particles and very fine precipitations [24], the higher fracture toughness was attributed to a strain-induced strengthening of the austenite during fracture, and thereby an additional increment of energy. Hence, the contribution to improving the fracture toughness of alloys containing up to 2.02% V, due to the presence of fine  $M_{23}C_6$  carbides within the austenite, was considerably higher than the reduction caused by, on the one hand, increasing the amount of massive carbides and the reduced inter-carbide distance, and on the other hand, reducing the volume fraction of retained austenite. Where the content of vanadium exceeded 2.02%, fracture toughness decreased (Fig. 18), since the matrix microstructure of those alloys was mainly martensitic.

In heat-treated Fe–C–Cr–V alloys with varying contents of vanadium, lower  $K_{Ic}$  values were obtained, compared with as-cast alloys (Fig. 18). Several processes which occurred during tempering affected fracture toughness. When held at 500 °C for 4 h, extensive carbide precipitation probably occurred in the martensite present in the as-cast structure, and retained austenite was decomposed; similar to the behaviour of high-carbon stainless steels tempered at the secondary hardening temperature [27]. Both effects combined to produce a significant drop in fracture toughness. Furthermore, during the cooling process, after it had been held at the tempering temperature, martensite formed, which also contributed to lower  $K_{Ic}$  values.

Finally, it is important to note that the as-cast matrix microstructure of Fe–C–Cr–V alloys containing in excess of 3.28% V was similar to the microstructure of heat-treated Fe–C–Cr alloys. The fracture toughness of these alloys was greater than high-temperature treated Fe–C–Cr alloys (27.2 MPa m<sup>1/2</sup> [33]), although they contained a larger amount of massive  $M_7C_3$  carbides. This may be attributed to a larger amount of retained austenite in Fe–C–Cr–V alloys. However, retained austenite was also present in heat treated Fe–C–Cr alloys [36], as much as 40% [37]. It can therefore be concluded that the difference in size, and probably the amount, distribution and behaviour of secondary carbide particles present in the matrix microstructures of the alloys examined had a decisive influence. Compared to Fe–C–Cr–V alloys, secondary carbides within the dendritic regions of Fe–C–Cr alloys were generally too large (0.5–1.0 µm [32]) to play a strengthening role and simply acted to restrict ductility. In this case, voids were nucleated around secondary carbide particles during stressing and fracture proceeded due to growth and coalescence of these voids [34].

Apart from better fracture toughness, as-cast Fe–C–Cr–V alloys with a high vanadium content offered better wear resistance under low-stress abrasion conditions (Fig. 17) than high-temperature treated Fe–C–Cr alloys (0.100 mm<sup>-1</sup> [36]), mainly because of a larger volume fraction of the carbide phase, a finer and more uniform structure, and because the morphology of eutectic colonies had been altered.

It was therefore possible, by adding approximately 3% V to high chromium white iron, to obtain an alloy that will have, in the as-cast condition or after tempering at 500 °C, better properties than high-temperature treated alloys containing no vanadium. This may be very important from a practical point of view, especially for castings with a complex configuration, since the possibility of high residual stresses during high temperature heat treatment and problems related to this process can be avoided.

## 5. Conclusions

Vanadium influenced the microstructure characteristics of high chromium white cast iron. By increasing the vanadium content in an alloy, the structure became finer, i. e. the spacing of austenite dendrite arms

and the size of massive  $M_7C_3$  carbides were reduced. The distance between carbide particles was reduced, the volume fraction of eutectic  $M_7C_3$  and  $V_6C_5$  carbides increased and the morphology of eutectic colonies changed. The amount of very fine  $M_{23}C_6$  carbide particles precipitated in austenite and the degree of martensitic transformation also depended on the content of vanadium in the alloy.

Because it changed microstructure characteristics, vanadium improved the wear resistance and fracture toughness of high chromium white cast iron. The addition of 1.19% V improved toughness by approximately 20% and wear resistance by 10%. The higher fracture toughness was attributed to strengthening during fracture and thereby an additional increment of energy, since very fine secondary carbide particles were present mainly in an austenitic matrix.

An Fe-C-Cr-V alloy containing 3.28% V showed the highest abrasion resistance, i. e. 27% greater than a basic Fe-C-Cr alloy. A higher carbide phase volume fraction, a finer and more uniform structure, a smaller distance between  $M_7C_3$  carbides and changes in the morphology of eutectic colonies all contributed to greater wear resistance.

Experimental results therefore indicate that the volume fraction of the carbide phase, carbide size and distribution, as well as the morphology of eutectic colonies, had an important influence on the wear resistance of Fe-C-Cr-V alloys under low-stress abrasion conditions, whereas fracture toughness was determined largely by the matrix microstructure.

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