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# On the microstructure and mechanical properties of high-speed steels

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The microstructure of high-speed steels consists of a martensitic matrix with a dispersion of two sets of carbides. These carbides are usually known as primary and secondary carbides. The role of the primary carbides has been reported to be of no importance in strengthening the steels, due to their large size and large interparticle spacing. The present authors have studied the role of the primary carbides on the wear of high-speed steels and found them to be of no importance, and under certain conditions contributing to higher wear rates. It has been shown analytically and experimentally that in quenched and tempered high-speed steels, the precipitation of the secondary hardening carbide (cubic $M_2C$ type) is the main reason for the improved strength and wear resistance. This shows that the secondary hardening phenomenon of high-speed steels is a direct result of the hardening caused by the precipitation of the cubic $M_2C$-type carbide. The present study has estimated that at peak hardness the volume fraction of secondary hardening carbides is approximately 20%. The measured strength of high-speed steels was found to be lower than the theoretically calculated strength due to non-homogeneous precipitation of the secondary hardening carbides. Areas which were observed to be free from secondary hardening carbides are real and are not artefacts. It has been shown that the strength of high-speed steel in the region of peak hardness depends primarily on the precipitation of the secondary hardening carbide and secondarily on martensitic strengthening.

## 1. Introduction

It is now well established that the secondary hardening of high-speed steels is caused by precipitation hardening due to the precipitation of special alloy carbides [1, 2]. Several carbides have been claimed to be responsible for the secondary hardening of high-speed steels. Kuo [3] reported hexagonal $M_2C$ carbide. White and Honeycombe [4] reported $M_{23}C_6$ carbide and Mukherjee [5] claimed that MC carbide was the secondry hardening carbide in high-speed steels. Recently, the present authors have for the first time successfully extracted the secondary carbides of high-speed steels. The study of these carbides proved that the cubic $M_2C$ carbide was the carbide responsible for the secondary hardening of high-speed steels [1, 2].

Analytical studies of the strengthening of high-speed steels based on dislocation theory have been attempted by previous workers [6–10]. However, because data on shape, size, distribution and volume fraction of secondary hardening carbides was not available, values were assumed by these workers. The successful extraction and characterization of secondary hardening carbides has recently made available such data. The present paper is a further step in the analytical study of the strengthening mechanism of high-speed steels.

## 2. Microstructure of hardened and tempered high-speed steels

It is well known now that the microstructure of hardened and tempered high-speed steels consists of a martensitic matrix with a dispersion of two sets of carbides, primary carbides and secondary carbides.

### 2.1. Primary carbides

Primary carbides are those carbides formed during solidification of the steel. They are observable in all the heat-treated steels (as-quenched, as-tempered and as-annealed). The particle size of the primary carbides is of the order of a few micrometres [1, 11]. Such a large particle size of the primary carbides has enabled observation and intensive investigation of these carbides by previous workers [2, 12–14]. Primary carbides are observable even by optical microscopy as shown in Fig. 1.

It is now agreed that the primary carbides in different high-speed steels are type $M_6C$ and MC carbides. Full data on both species of carbides found in some high-speed steels, such as volume fraction, average particle size, particle size distribution, chemical composition and crystallography have been reported by the present authors in previous work [1, 11].

It has been shown that as the volume fraction of the primary carbides increases, their particle size increases. These carbides did not contribute to the strength of the steel due to their large particle size and large interparticle spacing [1, 8, 11]. Also, the primary carbides did not show any variation in their size, distribution, volume fraction, chemical composition

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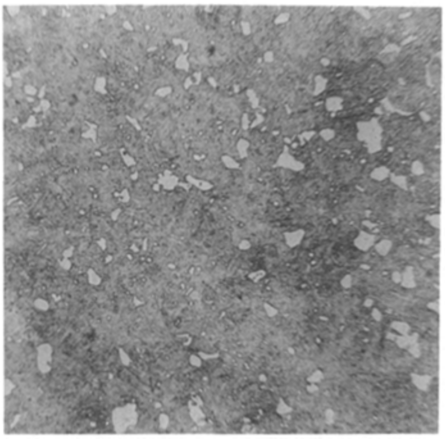

*Figure 1* Optical micrograph of primary carbides precipitated in M15 (×800).

or crystallography during secondary hardening when the strength of the steel increased considerably [1, 11]. The traditional belief that primary carbides contribute to wear resistance has also been shown to be untrue. In a recent study [11] the primary carbides were found to wear out at the same rate as that of the matrix, and abrasive tracks were also observed on the wear surface of primary carbides. In over-tempered steel, primary carbides were even harmful to the steel since they acted as abrasives which ploughed through the matrix, contributing to the observed high wear rate of over-tempered steels. In view of these recent findings it is claimed that further developments in improved high-speed steels should be concentrated on improving the matrix properties of the steel [1, 11].

## 2.2. Secondary carbides

Secondary carbides are those carbides formed during secondary hardening heat-treatment of the steels. The study of secondary carbides is technically more difficult than the study of primary carbides. The difficulties of studying such carbides are due to the small particle size of these carbides, especially in the early stages of their formation and also due to their metastable nature. Recently, the sequence of secondary carbide precipitation during the tempering of high-speed steels and the identification of the secondary hardening carbide has been resolved by the present authors [1, 2]. Analytical electron microscopy (AEM) has been used to perform such studies [15, 16]. Thin-foil techniques were unsatisfactory to study secondary hardening carbides, due to the dense precipitation of such tiny particles which were totally embedded in the matrix. Thin foils were used to obtain the ring pattern from such carbides. Imaging secondary carbides from thin foils showed them as clouds of precipitates as shown in Fig. 2.

It was impossible to measure the size or determine the shape of the carbide particles from thin foils due to the high magnetization of the foil and the tiny size of the particles. Microanalysis of secondary carbide particles from thin foils was also impossible since they were totally embedded in the matrix. This made the extraction of secondary carbides, and the secondary hardening carbide in particular, a necessity. Unsuccessful attempts at extracting such carbides have been reported [17]. However, the present authors, for the first time, were able to extract them on carbon and non-carbon replicas [1, 2, 16]. This allowed imaging, measurement of the size and microanalysis of these carbides as well as obtaining single-crystal diffraction patterns from the carbide by the convergent beam technique. It has also been shown that secondary hardening carbides of high-speed steels were plate-shaped, between 20 and 40 nm across and about 3 to 5 nm in thickness. The interparticle spacing was of the same order as the particle size, as shown in Fig. 3.

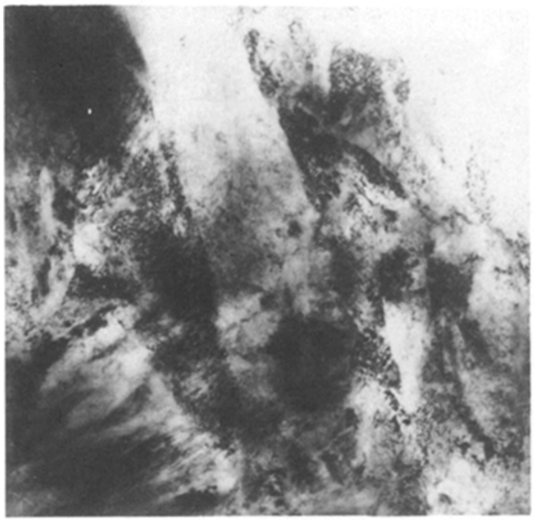

*Figure 2* Thin-foil electron micrograph showing secondary hardening carbides precipitated in M42 [×80 000].

It has been shown that secondary hardening of high-speed steels is a direct result of the precipitation of the cubic $M_2C$ carbides [1, 2].

## 3. Volume fraction of secondary hardening carbide

As shown in the previous section, the primary carbides have a large particle size which enabled easy observation and study of such carbides, even by simple techniques such as optical microscopy and simple wet methods of chemical analysis. Early techniques of X-ray analysis were also suitable in studying the

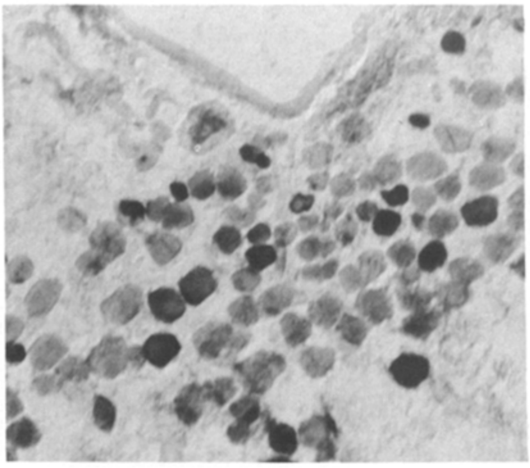

*Figure 3* Carbon replica electron micrograph showing group of secondary hardening carbides precipitated in M42 (×80 000).

crystallography of primary carbides. Data on the crystallography, chemical composition, particle size, volume fraction, etc. have been available in the literature since the early 1950s [3, 12]. Such data are confirmed by recent work [1, 2, 17, 18].

As for the secondary carbides, it has proved quite difficult to find many data in the literature except for crystallographic data, until the recent work of the present authors [1, 2]. Sophisticated analytical techniques were needed to clarify the nature, mechanism of precipitation, size, morphology, crystallography and chemical composition of secondary carbides. AEM was a very suitable technique to study such tiny particles of carbides. The two main methods used to prepare samples for investigation by transmission electron microscopy (TEM) are the thin-foil technique and the replica extraction technique.

Thin foils showed the secondary hardening carbides as clouds of precipitates as shown in Fig. 2. The difficulty of imaging, sizing and chemical analysis of such carbides is reported and the useful outcome from the thin-foil technique was the ring pattern from groups of these precipitates [17]. It was therefore necessary to extract these carbides to study them in isolation from the matrix. The difficulty of extracting these carbides has also been reported recently [17]. However, the present authors developed a method for extraction and were able to extract all the species of secondary carbide from high-speed steels [16]. The study of the secondary hardening carbides of high-speed steels by AEM has provided data on shape, size, crystallography and chemical composition of these carbides [1, 2]. However, the volume fraction of the carbides could not be determined experimentally. From thin foils, it was very difficult to image carbide particles due to their small size and the distortion of the electron beam due to the high magnetization of the foil. As for replica extraction, the complete extraction of all the carbide particles, especially when the particle size becomes as small as that of the secondary hardening carbides of high-speed steels, cannot be guaranteed. Generally, evaluating volume fraction data by the extraction replica method is not satisfactory. Henderer and co-workers [7, 8] tried to estimate the volume fraction of these carbides. Their estimation was based on the assumption that 50% of the amount of (MC + $M_6C$) carbides, dissolved on hardening of the steel, would precipitate on tempering the steel to peak hardness. $M_{23}C_6$ was excluded since it precipitated at higher temperatures. The value of 50% was based on the data reported by Payson [19] concerning the intensity of the diffraction pattern of hexagonal $Mo_2C$ precipitated at a temperature corresponding to the maximum yield strength, relative to the intensity of the pattern of the same carbide when fully precipitated.

Experimental results from previous work [1, 2] have proved that secondary hardening carbides were not hexagonal $Mo_2C$ nor MC nor $M_6C$ carbides. The secondary hardening carbide was the cubic $M_2C$ carbide rich in molybdenum and containing vanadium and chromium. Therefore, the assumption that the secondary hardening carbide was a result of the precipitation of 50% of (MC + $M_6C$) carbides, dissolved on hardening, is questionable. Also, both carbides MC and $M_6C$ were found to be deficient in chromium, while the secondary hardening carbide contained 10 to 20 at % Cr. Moreover, it was difficult to balance the chemical composition of the secondary hardening carbides to 50% of the amount of (MC + $M_6C$) carbides found in the annealed steels, for any value of the amount of secondary hardening carbide. However, the value of 50% was based on the relative intensity of diffraction patterns of hexagonal $Mo_2C$ precipitated in the pure molybdenum steel, which is not the case for complex alloyed high-speed steels. It is more reasonable to believe that each carbide precipitates having its own stoichiometry and quantity according to the thermodynamics and kinetics of the individual carbide reactions.

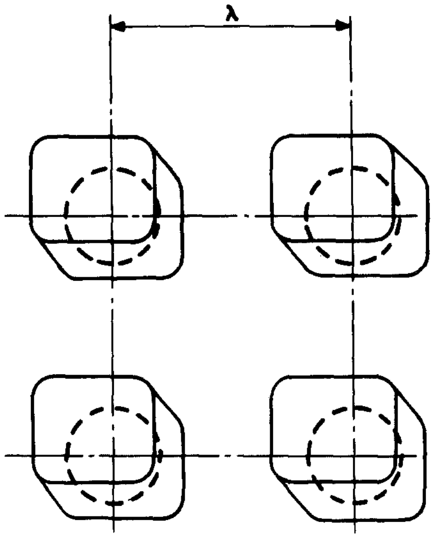

*Figure 4* Schematic drawing showing the idea of replacement of carbide plates by spheres of the same volume.

From the data available, the particles of the secondary hardening carbide are plate-shaped, having a size between 20 and 40 nm across and between 3 and 5 nm in thickness. Since the mathematical relationships between particle size, interparticle spacing and volume fraction are well established for spherical particles, it would seem reasonable to assume the replacement of the plates of the secondary hardening carbides by spheres of the same volume located at the centres of the plates, as shown in Fig. 4.

Since the volume of a plate is 2 to 8 × $10^3$ $nm^3$, then the radius of a sphere of the same volume is 8 to 125 nm. For spherical particles [20]

$$f = \frac{4r}{3\lambda + 4r} \qquad (1)$$

where $f$ = volume fraction, $r$ = radius and $\lambda$ = interparticle spacing. As shown in Fig. 3, the interparticle spacing is of the order of the particle size. Therefore Equation 1 gives $f$ = 20 to 25%, which is slightly higher than the value of the volume fraction of all the carbides dissolved on hardening the steels [14, 21].

## 4. Theoretical analysis of secondary hardening strengthening

As shown previously, the microstructure of high-speed steels consists of a martensitic matrix with a dispersion of primary and secondary carbides. It has been shown that primary carbides make no contribution to the strength of the steel [1, 8, 11]. Therefore, the contribution to the strength of the steel arises from both martensitic strengthening and the dispersion strengthening due to the precipitation of the secondary hardening carbides.

The strength of the martensite is a result of the degree of tetragonality of the martensitic lattice. Such tetragonality is a direct result of the dissolution of the interstitial atoms [22]. In the case of high-speed steels, carbon is the interstitial atom. So, the strength of the martensitic matrix of high-speed steel is dependent on the carbon content in the matrix. The experimental determination of the partitioning of carbon between the matrix and secondary hardening carbides has not yet been reported in the literature, because of the difficulty of the experimental determination of the carbon content in either the matrix or the fine carbide particles. However, it would seem reasonable to agree with the assumption reported by Henderer and Turkovich [8] that the matrix at peak hardness contained about 0.25% carbon. Winschell and Cohen [23] calculated the martensitic strengthening as a result of the interaction between dislocations and the interstitial atoms according to the equation

$$\sigma_c = 1.65 \times 10^{-3} L^{-1/2} (W\%C)^{1/3} \qquad (2)$$

where $L$ = dislocation length and $W\%C$ = percentage of carbon in martensite. Assuming that the fine-scale spacing between twin interfaces limits the effective dislocation length to be approximately 10 nm, Equation 2 yields a value of 1.04 GPa [8].

The contribution to strength from the precipitation of a second phase is a result of the interaction of the dislocations with such particles, since these particles act as dislocation barriers. In the case of high-speed steels, particles of the secondary hardening carbides are incoherent and plastic deformation occurs with dislocation bowing between the carbide particles.

Among the models for such a strengthening mechanism, Orowan's model was recommended [24] such that for spherical particles

$$\tau_0 = \frac{G\boldsymbol{b}}{L} \qquad (3)$$

where $\tau_0$ = shear strength due to second-phase precipitation, $G$ = shear modulus, $\boldsymbol{b}$ = Burgers vector and $L$ = spacing between particles. Taking $L$ = 30 nm, Equation 3 yields $\tau_0 = 0.83$ GPa. According to Von Mises' criterion, $\sigma_y = 3^{1/2}\tau_0$. This gives a contribution of $\sigma_y = 1.44$ GPa.

Kelly [25] has shown that particle shape affects the dispersion strengthening and the effect of plate-shaped particles depends on their volume fraction and the ratio $D/t$, where $D$ is the major dimension of the plate and $t$ the plate thickness. From data reported by Kelly [25], for plates having a volume fraction of 20 to 25% and $D/t$ = 8 to 10 which is the ratio for secondary hardening carbides in high-speed steel, the ratio of stress caused by plates to the stress caused by spheres of the same number and volume fraction is given by

$$\tau_{\text{plates}}/\tau_{\text{spheres}} = 1.8$$

Therefore, the contribution to the yield strength by the precipitation of plate-shaped secondary hardening carbides would be

$$\sigma_y = 1.44 \times 1.8 = 2.6\,\text{GPa}$$

Adding this value to the contribution of the martensitic strengthening gives a total yield stress to the steel equal to 1.04 + 2.6 = 3.64 GPa.

## 5. Discussion

The experimental work carried out by the present authors on the secondary hardening carbides of high-speed steels has provided data on the crystallography, chemical composition, shape and size of these carbides. However, the available techniques are not capable of measuring the volume fraction of these carbides. Previous workers assumed the volume fraction of the secondary hardening carbides in high-speed steel to be 5%. Such assumptions were based on the postulate that 50% of the carbides dissolved on

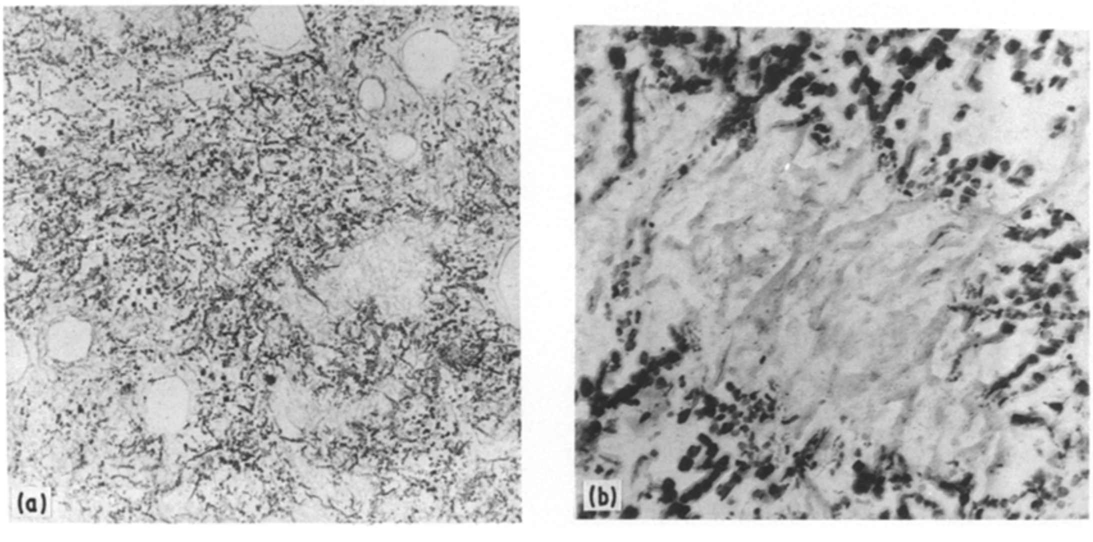

*Figure 5* Carbon replica micrographs showing areas free from precipitation of secondary hardening carbides in M42 tempered to its peak hardness: (a) low magnification (× 15 000), (b) high magnification (× 50 000).

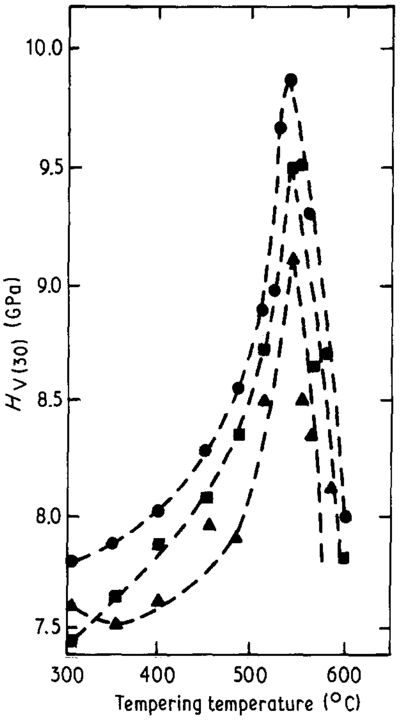

*Figure 6* Variation of Vickers hardness with tempering temperature for some high-speed steels: (▲) M1, (■) M15, (●) M42.

hardening the steel would precipitate as secondary hardening carbides. However, no verification of such an assumption based on the quantity and the chemical composition of the carbides was possible.

The present work estimated the volume fraction of the secondary hardening carbides to be 20 to 25%. This estimation was based on homogeneous precipitation of the secondary hardening carbides throughout the microstructure. However, the electron micrographs of replica-extracted carbides showed areas free from precipitation as shown in Fig. 5. Those areas without precipitation formed about 10 to 15% of the total area. Therefore the volume fraction of the secondary hardening carbides may be estimated as 20%, which is about the same value as the volume fraction of carbides dissolved on hardening the steels and not 50% as claimed by previous workers.

The evaluation of the volume fraction of secondary hardening carbides in the present work was based on the experimental data available on size, shape and distribution of secondary hardening carbides. However, a suitable technique is still required in order to measure the volume fraction of those carbides experimentally in order to confirm the present estimation. The theoretical analysis of the strengthening of the steels gave a value of 3.64 GPa. The measured peak hardness of several high-speed steels, as shown in Fig. 6, showed a value of 8.8 to 9.7 GPa. Since the hardness equals three times the yield strength of a steel, i.e. $H_V = 3\sigma_y$ [26]. So, the actual yield stress of high-speed steels has a value of 2.9 to 3.20 GPa, which is smaller than the theoretically calculated strength. This may be explained as a result of the non-uniform precipitation of the secondary hardening carbides, which confirms that the areas which appeared free from precipitation were not an artefact but were real areas without precipitation. However, more work is needed on the mechanism of precipitation of the secondary hardening carbides in high-speed steels.

## 6. Conclusions

1. The volume fraction of secondary hardening carbides in high-speed steels was estimated as 20%.
2. The actual strength of high-speed steels was found to be lower than the theoretically calculated strength due to the non-homogeneity of the precipitation of the secondary hardening carbides.
3. The strength of high-speed steel in the region of peak hardness depended primarily on the precipitation of the secondary hardening carbide and secondarily on martensitic strengthening.
4. Areas which appeared free from secondary carbide precipitate in the electron micrographs are real and are not artefacts.

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