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THE STRENGTH AND FRACTURE OF TWO-PHASE ALLOYS - A COMPARISON OF TWO ALLOY SYSTEMS

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ABSTRACT

The functional roles of the hard and soft constituents in the deformation and fracture of two-phase alloys are discussed on the basis of two commercially important alloy systems, namely spheroidized carbon steels and cemented carbides, WC-Co. A modified rule of mixtures provides a structural approach to the yield and flow strength. Consideration of the fracture toughness is attempted by means of a phenomenological modelling of the fracture process on the microscale. While there are large differences in properties between the two alloys, the deformation and fracture processes show broad similarities which are associated with the features of the interaction between constituents common to both alloys.

INTRODUCTION

One of the most effective means of strengthening metals is by the introduction of a hard second phase. Unfortunately, in general, the fracture properties (such as fracture toughness) change in opposition to the strength properties (such as yield strength, flow stress and hardness), and the design of alloys which are both strong and tough is a continuing challenge. This article presents several aspects of the problem dealing with the change of properties of two-phase alloys as a function of composition and microstructural variables, with particular emphasis on two commercial alloys at opposite ends of the two-phase composition range, namely spheroidized carbon steels and sintered tungsten carbide-cobalt alloys. The former are classified as dispersion hardened alloys, with particles of iron carbide dispersed in a matrix of ferrite. The latter belong to the class of cemented carbides, with a structure consisting of tungsten carbide embedded in a cobalt-rich binder phase. The natures of the two phases in each alloy are similar, both alloys contain a hard brittle constituent, the carbides, and a soft and ductile constituent, cobalt or iron. However, the relative amounts of the two constituents in each alloy are very different, since the steel contains of the order of 5-16 volume percent of hard particles, and the cemented carbides contain approximately 64-95 volume percent of tungsten carbide. Not surprisingly, the yield strength at room temperature and the effect of temperature on fracture toughness are also different in the two alloys, as presented schematically in Figure 1.

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It will be shown that in spite of the different functional roles of the constituents in each alloy, there are common features to the interaction between constituents in each alloy during deformation and fracture, and it is hoped that their consideration will lead eventually to generalizations applicable to the group of two-phase alloys as a whole.

1. Yield Strength and Flow Stress of Two-phase Alloys - A Structural Approach

A comprehensive theory of the strength of two-phase alloys, which would predict the yield strength and flow stress of multi-phase alloys from the properties of their constituent phases, has not yet been achieved. The relation between local and average stress and strain values in non-homogeneous microstructures is very complex and depends in detail on the spatial variation of such microstructural factors as the volume fraction, the morphology, the degree of dispersion or continuity and the orientation of the structural elements. However, recently published work of H. Fischmeister and B. Karlsson (1) showed that some useful generalizations may be obtained from the consideration of average deformation characteristics and, in particular, that in certain two-constituent steels and Al-Cu alloys with coarse microstructures the distribution of stress and strain among the constituents obeys the empirical mixture rules first used by Tamura, Tomata and Ozawa (2), namely

$$\bar{\sigma} = f_{\alpha} \sigma_{\alpha} + f_{\beta} \sigma_{\beta} \quad (1)$$

$$\bar{\epsilon} = f_{\alpha} \epsilon_{\alpha} + f_{\beta} \epsilon_{\beta} \quad (2)$$

where $\bar{\sigma}$ and $\bar{\epsilon}$ are, respectively, the flow stress and the strain of the aggregate, σ_{α} and σ_{β} are the average in-situ stresses, ϵ_{α} and ϵ_{β} are the average in-situ strains, and f_{α} and f_{β} are the volume fractions of the α and β constituents, respectively.

Equation (1) will be used here as a basis for the discussion of the strength properties of two-phase alloys, following, in part, a more complete presentation published elsewhere (3). When restated specifically for the yield strength of a two-phase alloy (with random or isotropic microstructure, i.e. the area fraction of a phase on any plane is statistically equal to the volume fraction of that phase), Equation (1) becomes

$$\bar{\sigma}_y = \sigma_{\alpha} f_{\alpha} + \sigma_{\beta} f_{\beta} \quad (3)$$

where $\bar{\sigma}_y$ is the yield strength of the alloy.

The general problem is to calculate $\bar{\sigma}_y$ for any composition from known or estimated values of σ_{α} and σ_{β} , at yield of the alloy. In order to do this, certain simplifying assumptions are required which idealize the actual behavior of the constituent phases. The idealized composite alloy consists of a relatively soft and ductile matrix α and embedded coarse particles (1-500 μm) of a hard and ductile phase β . The matrix phase is assumed to be continuous and to deform homogeneously in conformity with the aggregate body. The hard β phase may or may not be continuous depending on the structure of the alloy. The in-situ yield strengths of the two phases are $\sigma_{\alpha y}$ and $\sigma_{\beta y}$, respectively, with $\sigma_{\beta y} \geq \sigma_{\alpha y}$. The magnitude of the in-situ yield strength of each phase is a function of the parameters of the microstructure.

The preceding concepts will now be applied to the two types of two-phase alloys under discussion.

a) Dispersed particle strengthening. The alloy is assumed to consist of equiaxed β particles dispersed in a matrix of α . Since the continuous α matrix is postulated to deform in conformity with the alloy as a whole, it must yield when the alloy yields (or vice-versa), and therefore at yield of the alloy $\sigma_\alpha = \sigma_{ay}$. As in other treatments of dispersion hardening, the excess load-carrying ability of equiaxed hard particles is ignored, so that, as a first approximation the stress in the β particles is equal to the stress in the matrix and $\sigma_\beta = \sigma_{ay}$, also. After substitution in equation (3), the yield strength of the alloy is given as

$$\bar{\sigma}_y = \sigma_{ay} \quad (4)$$

i.e. the yield strength of the alloy is equal to the in-situ yield strength of the ductile matrix phase.

This simple and perhaps trivial statement hides a great deal of complexity, since in general, the yield strength of the matrix is a complicated function of its dislocation structure and internal boundary spacing. For instance, the yield strength of spheroidized steels increases markedly with decreasing iron carbide particle spacings between 10 and 0.2 μm (figure 1). However, a detailed study (4) showed that the yield stress of tempered and annealed spheroidized plain carbon steels is not directly determined by the particles themselves but is controlled by a subgrain boundary network, the mesh size of which is stabilized by the iron carbide particle spacing. The yield strength follows a Hall-Petch type equation,

$$\bar{\sigma}_y = 9.5 + 1.33 \lambda_{l,p}^{-1/2} \quad \text{kgf/mm}^2 \quad (5)$$

where $\lambda_{l,p}$ is the spacing of the subgrain boundaries resulting from the heat treatment. In steels in which the subgrain boundaries were eliminated by thermal cycling about the transformation temperature A , the yield stress was found to be predominantly controlled by the ferrite grain size λ_g via the relation

$$\bar{\sigma}_y = 12.4 + 1.87 \lambda_g^{-1/2} \quad \text{kgf/mm}^2 \quad (6)$$

and the intraboundary particles contributed only a small strain hardening term which increased the value of the friction stress (12.4 kgf/mm^2) over that associated with grain boundary strengthening alone (8.8 kgf/mm^2). The Hall-Petch relations of equations (5) and (6) are compared with that for grain boundary strengthening in figure 2.

The effect of the particles increases rapidly with strain due to the buildup of dislocation density and internal stresses associated with the incompatibility between elastic particles and plastically deforming matrix (5,6,7). As an example, figure 3 shows the relative magnitudes of the components of the hardening stress as a function of strain, according to one theory (6). Components $\Delta\sigma_i$, due to internal stress, and $\Delta\sigma_{gg}$, due to geometrical dislocation density, are associated with the particles. Component $\Delta\sigma^s$ is due to the strain hardening of the matrix. The stress on the particles is limited by the fracture strength of the particles or interfaces.

b) Matrix-reinforced hard-particle skeleton. Increasing the relative amount of the hard phase in a two-phase alloy system leads, in general, to clustering of the particles and eventually to their agglomeration into a continuous skeleton. By analogy with fiber reinforcement, a larger share of the load is transferred

to the hard phase as it becomes more continuous, by the action of the matrix shear stress on the particle-matrix interface (3). The stress in the hard phase is limited by the yield strength or fracture strength of that phase.

The yield strength and deformation of the type of alloy with a continuous hard phase can also be considered on the basis of the concepts underlying equation 3, and the strength and hardness of sintered WC-Co alloys will now be discussed from that point of view.

In commercial tungsten carbide-cobalt alloys (WC-Co) both phases are considered to be continuous for the purposes of this discussion, which means that both phases are assumed to conform to the yield strain of the specimen as a whole. However, the stress is not necessarily uniform in the WC since the load is transmitted through the restricted joints between grains, creating more highly stressed regions near the grain joints and less highly stressed regions elsewhere. The non-uniform stress distribution was specifically taken into account by Lee and Gurland (8) who, in brief, considered the WC divided into two parts as indicated schematically in figure 4: 1) a continuous volume fraction, $f_{\beta C}$, which links the load transferring contact areas in WC, and is loaded uniformly to the yield strength of WC, namely $\sigma_{\beta y}$, and 2) a non-continuous volume fraction $f_{\beta} - f_{\beta C}$, in which the average stress is equal to the flow stress in the binder phase, namely $\sigma_{\alpha y}$. The continuous volume fraction of WC is given by $f_{\beta C} = Cf_{\beta}$, where C is the contiguity ratio, i.e. the ratio of WC-WC grain boundary area and total surface area of WC in the alloy (9). For this particular case, then, equation (3) becomes

$$\sigma_y = \sigma_{\alpha y} f_{\alpha} + \sigma_{\beta} f_{\beta} \quad (7)$$

$$\text{where } \sigma_{\beta} f_{\beta} = \sigma_{\alpha y} (f_{\beta} - f_{\beta C}) + \sigma_{\beta y} f_{\beta C}$$

and, after substitution,

$$\sigma_y = \sigma_{\alpha y} (1 - Cf_{\beta}) + \sigma_{\beta y} Cf_{\beta} \quad (8)$$

Lee and Gurland (8) applied their model to the hardness of WC-Co. The experimentally determined change of hardness with composition is shown in figure 5. For hardness, the equivalent equation is

$$\bar{H} = H_{\alpha} (1 - Cf_{\beta}) + H_{\beta} Cf_{\beta} \quad (9)$$

where $\bar{H}$, H_{α} and H_{β} are the hardnesses, respectively, of the aggregate, the cobalt-rich binder phase and the continuous portion of the WC phase. From limited available data, the in-situ hardness of the matrix follows the equation

$$H_{\alpha} = 304 + 12.7 \lambda_{\alpha}^{-1/2} \quad \text{kgf/mm}^2 \quad (10)$$

and for the carbide phase, again from limited data,

$$H_{\beta} = 1382 + 23.1 \bar{\lambda}_{\beta}^{-1/2} \quad \text{kgf/mm}^2 \quad (11)$$

where λ_{α} is the mean free path of the binder and $\bar{\lambda}_{\beta}$ is the mean intercept grain size of the carbide. As shown in figure 6, good agreement was found between the measured hardness of the aggregate and the hardness values calculated from Equation 9, 10 and 11, for this alloy system. It is of interest to note that the

hardness of the matrix in this cemented carbide follows a Hall-Petch type relation (eq. 10) with respect to mean free path. This may indicate that the structural dependence of the flow stress or hardness of the binder in WC-Co is governed by similar strain hardening mechanisms as the structural hardening of the matrix of the spheroidized steel (eqs. 5 and 6). A conclusion of this kind had been reached by Doi et al. (24) on the basis of dislocation density determination in WC-Co as a function of strain.

2. Fracture and Fracture Toughness of Two-phase Alloys - A Phenomenological Approach.

A rule of mixtures applies only to certain fracture properties of two-phase alloys in which the failure criterion is directly related to the stress distribution among the phases and involves large regions of the specimen. Examples are the ultimate tensile strength of dual phase steels (10) and that of fiber reinforced metals (11). For other fracture properties, such as the fracture toughness, which involve deformation and fracture processes in very localized regions, a comprehensive theoretical framework has not yet been formulated for a general relationship between fracture strength and microstructure. The best one can do, perhaps, at the present stage of development of the science, is to examine the microscopic fracture processes of various alloy systems for similarities, and consider whether these can be developed into predictive relations between microstructural parameters and macroscopic measurements of fracture toughness. In this spirit, the fracture of the two alloy systems, spheroidized carbon steels and sintered tungsten-carbide, will now be considered.

It is generally agreed that the fracture of alloys with hard inclusions starts with the cracking of the inclusions or their interfaces (12,13,14). The failure of inclusions and interfaces is associated with the accumulation of stress and strain in the ductile matrix during strain hardening. The particle or interface stresses increase either because of direct load transfer across the particle-matrix interface or due to the indirect effects of the increasing incompatibility between elastic particle and plastically deforming matrix, until particle fracture or interface decohesion occurs. However, the initiation of cracks or voids at the hard particles is not generally the critical event (although it is often a necessary pre-condition of fracture), but it is the unstable growth of these cracks into the surrounding microstructure which produces the failure of the specimen. The specimen fracture is brittle if the crack growth is by cleavage. The fracture is ductile, if particle or interface cracking is followed by void growth and void coalescence.

The conventional measures of fracture toughness in a thick, deeply-cracked test piece subjected to mode I loading are K_{IC} , the critical stress intensity factor, and G_{IC} , the critical energy release rate. A number of failure criteria have been developed on the basis of models of the fracture events at the crack tip. For instance, for fracture propagation entirely by cleavage, Ritchie, Knott and Rice (15) used the attainment of a critical stress level, σ_f^* over a minimum size scale, X_0 , as a local criterion for unstable cleavage fracture. Under brittle conditions, such as low temperature, high strain rate or high concentration of the brittle constituent, when σ_f^* is of magnitude comparable to or smaller than the yield stress, this criterion reduces to

$$K_{IC} = \sigma_f^* \sqrt{2\pi X_0} \quad (12)$$

The authors (RKR) found good agreement between theory and experiment for a high nitrogen mild steel using $X_0 = 2$ grain diameters and a macroscopically measured fracture stress.

With increasing ductility, at higher temperatures for instance, the microscopic mode of fracture initiation changes from cleavage to ductile tearing. An approximate failure criterion was stated by Rice and Johnson (16), namely that crack extension proceeds when the heavily deformed region in front of the crack tip is comparable in dimension to the width of unbroken ductile ligaments in the microstructure, at the onset of ductile rupture. Since the severely deformed region is approximately equal to the critical crack tip opening displacement δ_c , where $\delta_c = 0.5 K_{IC}^2 / \sigma_y E$, and furthermore δ_c is of the same order as the interparticle spacing Δ , one obtains

$$K_{IC} = [2E \Delta \sigma_y]^{1/2} \quad (13)$$

where σ_y and E are, respectively, the yield stress and Young's modulus of the metal. The experimental data for a number of steels and aluminum alloys have been found to agree with equation 13 (17).

The fracture of spheroidized steels at low temperatures and that of cemented carbides at room temperature will now be examined in the context of the Ritchie, Knott and Rice, and Rice and Johnson criteria. It should be noted here that these two theories are among a number of proposed relations between the fracture toughness parameter and unidirectional tensile properties and microstructural variables which have met with varying degrees of success and are not necessarily generally accepted (13).

a) Spheroidized carbon steel. The observation of the fracture process in notched and fatigue cracked specimens (0.13 - 1.46%C) by Rawal and Gurand (14) revealed two different modes of fracture in the region near the crack tip, at low temperatures. At very low temperatures (-196° to -150°C) the fracture is almost completely brittle, with cleavage crack initiation sites at cementite particles often observed. At somewhat higher temperatures (-150° to -110°C) a fibrous (ductile) fracture region is formed between the fatigue crack tip and the final brittle fracture. Examination of crack tip blunting showed broken particles and voids. At even higher temperatures the specimens fail to meet the plain strain conditions for fracture toughness testing due to the extension of the zone of extensive plastic deformation.

All indications strongly support the conclusion that both the brittle and the ductile fracture processes start with particle cracking or interface decohesion. In effect, the comparison of the stress required for particle fracture or decohesion (1400 - 1670 MN/mm²) with the estimated maximum stress intensity ($\sim 3 \sigma_y$) ahead of the crack shows that the stress conditions for particle or interface cracking exist in these specimens within a few grains or particle spacings from the crack tip.

The critical condition for fracture is the condition for crack growth or propagation from the particles into the ferrite matrix. The RKR model was applied to the brittle initiation case. The critical tensile stress σ_f^* (eq. 12) was equated to the fracture propagation stress of a crack nucleated within a grain-boundary carbide particle as given by Smith (12).

$$\sigma_f^* = \left(\frac{4E\gamma}{\pi(1-\eta^2)\bar{d}} \right)^{1/2} \quad (14)$$

where γ is the effective surface energy of ferrite, η is the Poisson's ratio and $\bar{d}$ is the mean carbide thickness. The size scale X_0 (eq. 12), i.e. distance within which the principal stress at the crack tip equals or exceeds σ_f^* , was empirically established to be equal to 1.3 grain diameters. The values of K_{IC}

calculated by the RKR procedure are compared with the observed values of K_{IC} in figure 7, they show a fair degree of agreement for all specimens with observed fracture initiation by cleavage.

At higher temperatures or reduced carbon contents, the measured toughness values deviate appreciably from the values calculated by the RKR criterion. The specimens for which the cleavage fracture stress appreciably exceeds the available stress intensification (experimentally, if $\sigma_f/\sigma_y > 3.86$) show ductile rupture surface characteristics in the initiation zone. The stress level is sufficient to break the particles or interfaces, but is insufficient to cause cleavage propagation. The fracture here proceeds by void growth and coalescence over a distance more or less equal to the nearest neighbor interparticle distance, Δ , in Equation 13. The nearest neighbor particle distance for randomly distributed uniform spherical particles is $\Delta_2 = 0.36 \text{ df}^{-1/2}$ on a plane, and $\Delta_3 = 0.45 \text{ df}^{-1/3}$ in a volume (9). Hahn and Rosenfield (17) used the volume spacing Δ_3 in applying Equation 13, but, as shown in figure 8 we found better agreement between experiment and theory when using the planar spacing Δ_2 , for the specimens with ductile initiation (solid line). It is surprising to note that the results are insensitive to the average particle size, also that equation 13 with volume spacing Δ_3 agrees with the results of cleavage initiated fracture (broken line). These aspects of the model are further discussed in reference 14.

b) Cemented carbide alloy WC-Co. The fracture toughness of sintered WC-Co alloys (5.1 - 36.9 volume percent Co) was measured at room temperature on single-edge notched beam (SENB) specimens precracked by electron discharge machining (18). The fracture toughness increases with both binder content and WC particle size, and it is a function primarily of the matrix mean free path, as shown in figures 1 and 9. In general agreement with other published observations (19,20,21), it was found that the fracture path involves three main modes of rupture: i) interfacial decohesion along WC-WC and/or WC-Co interfaces, ii) cleavage of large WC grains, and iii) ductile rupture of binder ligaments.

Although there are several published fracture theories for cemented carbides based on dislocation mechanisms (19) and fracture mechanics (21,22,23), the present approach attempts to test a particular phenomenological criterion against the measured results. The observed fracture process, namely interfacial decohesion or WC cleavage followed by extensive plastic deformation and rupture of the binder ligaments, is compatible with the ductile fracture criterion of Rice and Johnson. If the ligament width Δ (Eq. 13) is taken as equal to the binder phase mean free path λ , and using $G_{IC} = K_{IC}^2 (1-\nu^2)/E$, one obtains

$$\sigma_y \lambda \approx \alpha G_{IC} \quad (15)$$

where α is a constant, and $(1-\nu^2)$ is taken as unity. A plot of the calculated values of the left hand side of equation 15 against the measured critical strain energy release rate G is shown in figure 10. The linearity is fairly good, although the value of the slope (0.24) is appreciably less than the theoretically expected value (between 1 and 2). This discrepancy can be accounted for partially by correcting for a possible overestimate of the ductile ligament thickness and an overestimate of the in-situ yield strength of the binder phase in the calculations of Pickens and Gurland (18).

COMMENTS AND SUMMARY

The resistance to plastic deformation of the two-phase alloys follows a modified rule of mixtures, and, in principle, the yield strength and flow stress of the

alloys can be accounted for by the summation of the separate contributions of each phase. In practice, the properties of each phase are subject to the effects of interaction between the phases, and the interaction effects are not always well known. The following general statements apply to the two alloys under discussion.

In dispersion hardened or particle strengthened alloys, the flow strength generally increases as an inverse function of the spacing of structural barriers to the propagation of slip, as, for example, grain boundaries, subgrain boundaries and hard particles. In the dispersion-hardened spheroidized carbon steels, the direct load-carrying contribution of the hard particles is negligible, but they exert a strong indirect effect on the yield and flow strength of the matrix, due to the interaction of the particles with subgrain boundaries and with the dislocation structure of the matrix during plastic deformation.

The strength of a two-phase alloy with an appreciable amount of the hard constituent does not depend only on the strength of the ductile matrix, but is also affected by the load carried by the hard phase. The fraction of the total load carried by the hard phase increases generally with the amount of that phase, as it becomes more continuous. The mechanism is one of load transfer between ductile matrix and hard inclusion. In the case of WC-Co alloys of commercial composition, the carbide phase is continuous, for all practical purposes, and it must therefore deform in compatibility with the alloy as a whole and carry the major portion of the load. At yield, the strength of the alloy is determined by both the in-situ yield strength of the binder phase and that of the continuous portion of the carbide phase. The binder phase is subject to similar indirect strengthening processes as the matrix in the spheroidized steel.

As the plastic deformation proceeds, the increasing plastic incompatibility between the soft and hard constituents causes cracks to form as a relaxation process. The major stages of the fracture processes in spheroidized steels at low temperatures and in WC-Co alloys at room temperature are similar, in spite of their different microstructures. In both alloys, the initial cracking takes place in the carbide or at the carbide-matrix interface, and is followed by crack growth through the matrix. The ductile fracture process in both alloys may correspond to the fracture model underlying the Rice-Johnson criterion, although the plastic deformation during fracture occurs on very different size scales in the two microstructures, and the fracture toughness values are comparable only within a narrow temperature range. At first glance it may seem surprising that the rather brittle cemented carbide and the more ductile steel are compatible with the same ductile fracture model. However, it should be realized that the materials' fracture behavior was compared only under conditions of small scale yielding. In the case of the spheroidized steel, the small scale yielding condition is obtained only at very low temperatures. In cemented carbides, it holds over a wide range of temperatures (including room temperature) because the extensive plastic deformation is restricted to the ductile binder phase which constitutes only 5 to 30 volume percent of the alloy. In the case of the cemented carbides, the localized plastic deformation of the binder phase layers at the crack tip contributes the dissipative work, which accounts for most of the fracture surface energy without greatly affecting the deformation of the bulk of the material which deforms elastically. In both steels and cemented carbides, the linear elastic fracture mechanics are applicable only if the plastic yielding is highly restricted. The conclusions about the fracture process are very tentative because the ductile fracture criteria are not yet sufficiently developed to account quantitatively and with confidence for the fracture behavior of a group of alloys over a range of microstructural parameters.

In summary, the respective role of each constituent is succinctly stated as follows: the hard carbide contributes the strength and hardness; the soft metal provides the toughness.→ The expected functions of the minor constituents of each type of alloy are aptly described by current terminology, namely "dispersion-hardened" and "cemented" alloys. The hard particles in steels serve to strengthen the matrix in which they are dispersed. The main function of the binder phase in cemented carbides is to cement the aggregate, i.e. to hold together and to transmit load. The matrix or binder phase has the additional important function to maintain the carbide in a desirable state of dispersion such as to maximize the resistance to fracture propagation.

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Figure 1a. Schematic comparison of the yield stress σ_y of spheroidized steels (Fe-C) and cemented carbides (WC-Co)

Figure 1b. Schematic comparison of the fracture toughness K_{IC} of spheroidized steels (Fe-C) and cemented carbides (WC-Co).

Fig. 2. Yield stress σ_y versus spacing parameter λ . Line refers to equation 5, line 2 refers to equation 6, and line 3 is the Hall-Petch plot for grain-size strengthening in mild steels. (After ref. 4). [Reproduced by kind permission of the Metallurgical Soc. of AIME and Am. Soc. for Metals].

Fig. 3. Magnitudes of the three components of the hardening stress for a spheroidized steel (After ref. 6). [Reproduced by kind permission of Pergamon Press].

Fig. 4. Structure model for the calculation of the continuous volume in WC-Co alloys (After ref. 3). [Reproduced by kind permission of Elsevier Sequoia SA.].

Fig. 5. Hardness versus carbide volume fraction in WC-Co (After ref. 3). [Reproduced by kind permission of Elsevier Sequoia SA.].

Fig. 6. Measured hardness versus calculated hardness in WC-Co, according to equation 4 (After ref. 3). [Reproduced by kind permission of Elsevier Sequoia SA.].

Fig. 7. Measured K_{IC} versus calculated K_{IC} in spheroidized steels, for small scale yielding conditions at $X_o = 1.3 \lambda_y$ (After ref. 14). [Reproduced by kind permission of The Metallurgical Soc. of AIME and Am. Soc. for Metals].

Fig. 10. Critical strain energy release rate G_{IC} versus $\sigma_y \lambda$ (After ref. 18). [Reproduced by kind permission of Elsevier Sequoia SA.].