ON STRUCTURAL ASPECTS OF FRACTURE TOUGHNESS OF MATERIALS WITH LIMITED PLASTICITY

S.A. FIRSTOV

Frantzevich Institute for Problems of Materials Science 3, Krjijanivski Str., 252680 Kiev, Ukraine

ABSTRACT

The influence of different structural factors and test conditions on fracture toughness of materials with a strong temperature dependence of yield stress is discussed and a necessity to take account of fracture micromechanisms is shown. It was found that the change of fracture mechanisms is resulted in the non-monotonous dependence of fracture toughness on structural parameters. Physical schemes are suggested to describe effects of dislocation substructure and porosity on fracture toughness.

KEY WORDS

Fracture toughness, micromechanisms of deformation and fracture, structural parameters, previous deformation, brittle-ductile transition.

INTRODUCTION

The wide range of new structural and tool materials with limited plasticity, materials based on refractory metals, metallic or intermetallic compounds is developing now. The sharp temperature and strain rate dependences of the yield stress are indicative of this group of materials. This peculiarity as well as structural parameters stipulate considerable variety of micromechanisms of deformation and fracture under different test conditions including temperature, strain rate and stress state variation.

A number of semi empirical dependences between fracture toughness and characteristics sensitive to microstructure (the yield point, parameters of work hardening, critical deformation) were established by Hahn (1968,1971) and Barsom (1973). Some of them are in a good agreement with experimental data and given below:

$$K_{1c} = B/\sigma_y^2 \qquad (1)$$

$$K_{1c} = A(\sigma_y)^{1/2}(\varepsilon_{cr})^{1/m} \qquad (2)$$

$$K_{1c} = (\tfrac{3}{2}E\sigma_y\varepsilon_{cr}n^2)^{1/2} \qquad (3)$$

In these formulae K1c - is the critical stress intensity factor; E - is an elastic modulus; n - is a work hardening law exponent; σ_y - is a yield stress; ε_{cr} - is the critical deformation value, being realized in the pre-fracture zone. A, B, m - are constants. The most full selection of formulae was done by Panassyuk (1988).

Such differences and even controversy (for example, expressions (1) and (2)) are apparent and reflect the very fact ıt different dependences correspond to different fracture mec ı-nisms.

So it is necessary to point out the main physical aspects of fracture mechanisms variety in materials under consideration.

THE INFLUENCE OF TEST TEMPERATURE AND STRAIN RATE

Transition metals with BCC lattice and their alloys reveal the utmost variety of fracture mechanisms. Under low temperatures they behave like brittle materials. The full set of all possible fracture mechanisms manifests itself sequentially in accordance with test temperature and material structure changes.

The extent of each temperature stage could be large enough and amounts to hundreds of degrees for certain cases. This situation has impeled us to elaborate the scheme of the ductile - brittle transition so that temperature intervals associated with fracture mechanism changes should be specified as Trefilov et al (1988) showed (Fig. 1).

This enables to determine the temperature bounds of the ductile - brittle transition more physically. As far as the low - temperature limit T_b^l is determined by Ioffe-Davidenkov's condition as temperature which provides zero ductility, it is close to the temperature associated with transition from brittle cleavage to cleavage with brittle relaxation.

$$\sigma_y(T_b^l) = \sigma_f(T_b^l) \qquad (4a)$$

here σ_y - the yield stress, σ_f - the fracture stress.
In its turn the high temperature limit T_b^h may be determined according to the condition when effective specific fracture energies for cleavage with plastic relaxation and for ductile failure are equal.

$$\gamma_{eff}^{cleav}(T_b^h) = \gamma_{eff}^{dimpl}(T_b^h) \qquad (4b)$$

In the case of cleavage with plastic relaxation, for example, for molybdenum, according to Vasilev (1979)the following dependence is realized:

$$\gamma_{eff}^{cleav} = \gamma_o(1+\alpha/\sigma_y^2) \qquad (5)$$

where γ_o - is specific surface fracture energy for absolutely brittle materials, α-is const.

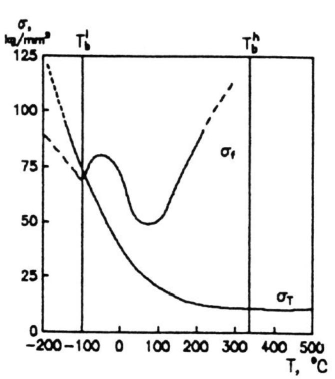
In the case of ductile failure, since a dimple size remains constant within some temperature interval above T_b^h, one may be suppose that

$$\gamma_{eff}^{dimp} \geq h\cdot\sigma_y, \qquad (6)$$

where σ_y - is the yield stress, h - is a dimple size.

The expressions (5) and (6) enable us to analyze the temperature dependence of γ_{eff} as well as K_{1c} or G_{1c} quite easily. According to Gridneva (1969), Trefiliv (1975), the temperature and strain rate dependence of the yield stress (without a non-thermal component), particularly for refractory metals with BCC lattice, can be described within interval 0,1-0,2T_m as following:

$$\sigma_y = A\,(\dot{\varepsilon})^{1/3}\exp(Uo/3KT) \qquad (7)$$



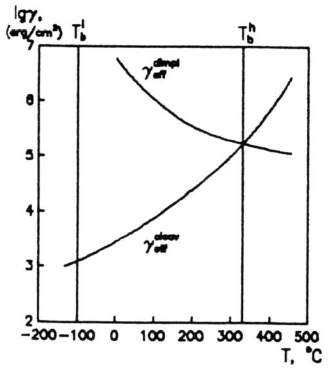


Fig. 1 On the determination of lower T_b^l and upper T_b^h temperature limits of a brittle-ductile tansition of molybdenum.

where Uo - is activation energy for dislocation motion, which is 0,2 eV for chromium; 0,19 eV for molybdenum and 0,49 eV for tungsten, A - is constant of material.

Taking into account (7), we can analyze the temperature dependence of γ_{eff}^{cleav} and γ_{eff}^{dimp} (Fig.1,b). A scheme presented is useful for investigation of a temperature dependence of fracture toughness since $K_{1C} \propto E\cdot\gamma_{eff}$.

In the Igolkina's work (1990) concerning polycrystalline chromium the significant T_b^h shift was observed (about 100...300 degrees) which was accounted for by increase of activation energy for dislocation motion due to interaction between dislocations and impurities in the temperature interval where dynamic strain aging occurred. Fracture toughness proved to be sensitive to such shift in realization of the fracture mechanism change.

THE INFLUENCE OF DISLOCATION STRUCTURE

The main types of dislocation structure arising under deformation and thermal treatment are studied quite satisfactory up to now. Homogeneous high dense distribution of dislocations (under temperatures below $0,2T_m$) or slightly misoriented cellular structure arise at the early stages of plastic deformation. The characteristic peculiarity of those structure types is the presence of dislocations with opposite charges without their space separation. When strain increases dislocations of single charge accumulate in some parts of volume forming misoriented cellular structure. The average misorientation angle rises vs strain as $\theta \sim e^{3/2}$, while misorientation angles spectrum approaches toward that for polycrystalline state (Firstov et al (1991)).

The critical strain required to change the structure, from slightly misoriented to misoriented, depends on the grain size and temperature as

$$e_{cr} = \mathrm{const} \cdot d \cdot e^{U_1/kT} \qquad (8)$$

where d - is the grain size, $U_1 \approx 0,2\text{-}0,3$ eV

According to analysis made in (Firstov (1991), Trefilov (1976)) this transition is accompanied by the deep change in deformation mechanism: it transforms from one characterized by work hardening law $\Delta\sigma(\varepsilon) = \alpha G b\, [\rho(\varepsilon)]^{1/2}$ (controlled by dislocations "forest") to that characterized by $\Delta\sigma(\varepsilon) = K d^{-\alpha}(\varepsilon)$ (controlled by grain boundaries), here $\alpha=1/2$ for boundaries formed under "warm" deformation and $\alpha=1$ for "cold" deformation.

It seems to be natural, that different types of dislocation substructure affect differently the failure mechanism of materials and, particularly, materials inherent to brittle fracture. According to Trefilov (1975) the ductile - brittle transition temperature reaching a maximum at strains 20-30% which correspond to misoriented structure (actually the structure with fine grains) subsequently decreases, reaching the level of -150 °C for molybdenum, chromium and -100 °C for tungsten.

The behaviour of the K_{1c} vs ε dependence differs somewhat from considered above. In the case of ductile "dimple" failure

(Tuchinski (1991)) previous deformation reduces K_{1c}. This result seems to be expected because deformation considerably promotes structural preparation to failure by pores coalescence. In the case of cleavage the K_{1c} vs ε dependences are non-monotonous (Fig. 2). The nature of maximum at strains 20-30% is unclear. Presumably, the structure which corresponds to the early stages of misoriented cells formation is indicative of special instability under the recurrent loading. This process can favour the increase of K_{1c}.

Within the interval of large plastic deformation (ε > 80%) the fracture toughness rises again if initial notch (crack) is oriented perpendicularly to the plane where grains (cells) are elongated.

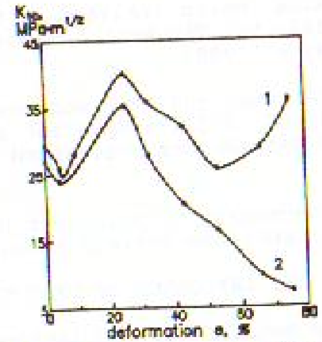


Fig. 2 The K_{1c} vs deformation of powder iron for transversal (1) and longitudinal (2) introduction of the notch.

Fractographic investigations have shown that such a behaviour was connected with the delamination along grain (cells) boundaries. The phenomenon of delamination is stipulated by disordered distribution of dislocations in the boundaries under deformation temperatures below 0,35 Tm and, consequently, by significant decrease of energy of the boundaries which can be evaluated as $2\gamma_o$ (where γ_o - is true specific surface energy of boundaries, Korniushin (1976)). If impurities segregate into boundaries, the value given might be lower.

If the notch (crack) under the fracture toughness tests is directed along the elongated grains (cells), the crack propagation along boundaries is facilitated so that Kic decreases when degree of previous deformation increases.

THE INFLUENCE OF POROSITY

The problem of the porosity influence on elastic modulus and characteristics of strength was thoroughly discussed in the literature. It was shown that the mechanical properties of sintered materials are drastically reduced when the porosity increases. In particular, the following formulae have been derived in a number of publications of Balshin (1972), Skorokhod (1982), Andrijevsky (1982).

$$\sigma_y = \sigma_{yo}(1 - \theta)^m \qquad \sigma_p = \sigma_{po} e^{-b\theta}$$

$$\sigma_y = \sigma_{yo}(1 - \theta)^2/(4 - 3\theta)^{1/2} \qquad \sigma_p = K_y d^{-1/2} e^{-b\theta}$$

$$E = E_o(1 - \theta)^m$$

These dependences agree with the experiment as to the porosity values of $\theta \le 30\%$. Some more complicated and more exact dependences of the properties on porosity have been proposed for the range of $\theta \le 60\%$.

In the works of Drachinsky (1982), Firstov (1987) the nonmonotonous K1c vs porosity dependences for porous iron have been first established under test temperature -196 ^oC.

According to fractographic data the fracture micromechanism changes from cleavage to dimple failure that occurs due to a transition from open to closed porosity. In total correspondence with this fact the fracture toughness depends nonmonotonously on the porosity (Fig. 3). A correct theoretical explanation of this phenomenon cannot be given at present. It may be supposed that the right-handed branch of the curve corresponds to an abrupt reduction of the deformable volume in connection with a localization of the deformation in the contact bridges.

The left-handed branch of the dependence of K_{1c} on θ shows that for cleavage the fracture toughness increases in comparison with the compact material due to the plastic relaxation enhanced by the porosity. As far as for such a failure mechanism $K_{1c} \sim \sigma_F^3/\sigma_y^2$, as Panassyuk (1988) found, it may be supposed that the behaviour of $K_{1c}(\theta)$ results from the more abrupt reduction of the yield stress with porosity in comparison with the fracture stress.

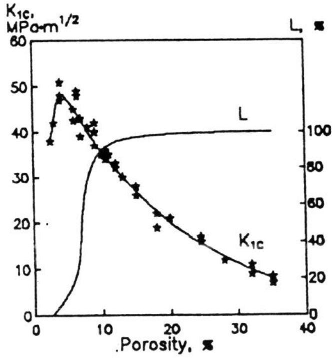


Fig. 3 Influence of the porosity on the fracture toughness K_{1c} and the fraction of dimple surface L in iron at -196 ^oC.

Data obtained for porous glass Coronel (1990) are of great interest. As follows from Fig. 4a, which represents data of Coronel (1990) et al the dependence of Young's modulus on porosity and the K_{1c} vs porosity dependence are both decreasing, but show distinct curvatures. Due to this fact the function $G_{1c}(\theta) = K_{1c}(\theta)/E(\theta)$ (where θ - is volume fraction of pores) has well - defined maximum at $\theta \approx 25\%$.

It is interesting to note that the fracture toughness value estimated from G_{1c} at $\theta = 25\%$ exceeds for 3 times the K_{1c} of non-porous material. In our investigations of glass with nano-porous structure a maximum of the K_{1c} vs porosity curve has been obtained also. The maximum has corresponded to the similar porosity interval (Fig. 4b).

CONCLUSION

Data concerning the influence of different factors on fracture toughness of materials having been discussed in the present short review enable us to draw the conclusion, that the considerable variety of fracture mechanisms do not allow to suggest quite simple theory which would describe the influence of some single factor on fracture characteristics.

Such a theory could be created for a comparatively narrow interval of manifestation of one specified fracture mechanisms. In the work Trefilov et al (1988) a thermodynamical conception of the fracture micromechanism changes was suggested.

It is obvious that the changes of fracture toughness characteristics caused by structural factors should be expected mainly in the intervals of test parameters variations, which entail fracture mechanism changes (see, e.g., fracture mechanisms maps given by Ashby (1983) for different materials).

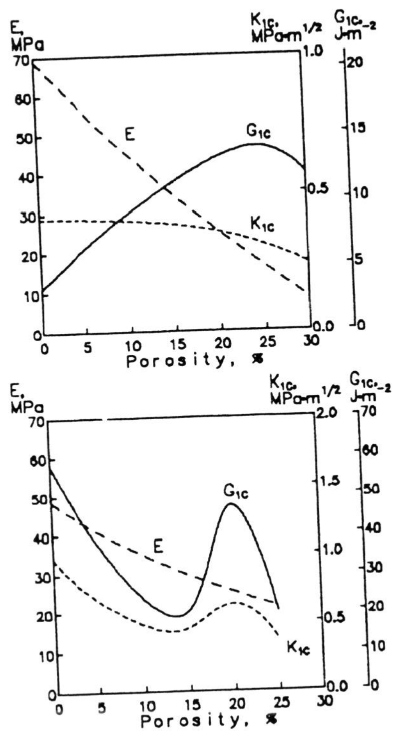


Fig. 4 The fracture toughness (K_{1c}, G_{1c}) and the elastic modulus (E) vs porosity dependencies for glass with coarse (a) and nanosized (b) pores.

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