FRACTURE TOUGHNESS OF HETEROGENEOUS MARTENSITE IN HYPOEUTECTOID LOW ALLOYED STEEL

I. DLOUHÝ, J. ŠKAREK and M. HOLZMANN

Institute of Physical Metallurgy,
Academy of Sciences of the Czech Republic,
61662 Brno, Czech Republic

ABSTRACT

A model of the fracture initiation mechanism in heterogeneous martensitic structures of medium carbon low alloy steel was discussed in terms of several concurrent micromechanistic phenomena involving mainly the intrinsic toughness of the martensitic matrix and the carbide-matrix interface boundary decohesion susceptibility.

Compared to homogeneous martensite the low carbon martensitic matrix containing the remainders of undissolved carbides exhibits a significant increase in fracture toughness without loss of strength. The presence of microregions with enhanced content of carbide-forming elements results in the increase of strength and fracture toughness. On the contrary, at the presence of remainders of undissolved carbides in a martensitic matrix with higher carbon content, a decrease in fracture toughness was stated at the maintained strength level.

KEYWORDS

Low alloy steel, martensite, heterogeneity, fracture toughness

INTRODUCTION

The problems of low-alloy ultra-high strength steels have got, in the recent years, into the shadow of works devoted to specific features of fracture behaviour in advanced materials. The absence of an unambiguous fracture toughness nature explanation in martensitic steels can be considered as one of the main reasons of this situation. Although knowledge on this field made it possible to describe some features of their fracture behaviour (Knott 1980, Kumar & Pandey 1984, Handerhan 1988, Di Russo 1986) no generally acceptable micromechanistic failure model has been found up to now.

The fundamental problem may be seen in the fact that the majority of strengthening factors in a martensitic matrix is associated with the risks of lowering its fracture resistance (Lee 1989, Dlouhý 1992b)- the strengthening effect being, at the same time, the embrittling one. Certain possibilities in increasing both the strength and thoughness are offered by producing some structural and chemical microheterogeneities in martensitic matrix (Tomita 1987, Padmanabhan

1984, Dlouhý 1992 ab), and namely in the cases in which homogeneous martensite has been considered as the optimal structure. In these cases the martensite strengthening was accompanied by an increase in toughness. These non-conventional heterogeneous microstructures have not to be taken only as a conglomeration of microstructural constituents the plastic and fracture behaviour of which is described by the global characteristics but as the second phase interacting with martensite matrix having its own (intrinsic) toughness. The fracture resistance of such a natural composite might be then analyzed from the point of view of the behaviour of individual phases.

The aim of this contribution is to present the results of an experimental failure micromechanism investigation of martensite with dispersed carbidic phase. At the same time, some results are presented which explain the positive influence of several kinds of martensite microheterogeneities on strength and fracture properties.

EXPERIMENTAL PROCEDURE

The experimental MnSiCrMoV steel used had the chemical composition (wt. %): 0,30 C; 1, 66 Mn; 1,50 Si; 1, 08 Cr; 0,40 Mo; 0,16 V; 0,14 Ni; 0, 022 P; 0,020 S. Square and cylindrical rods with a cross-section of 10,5x10,5 mm and 6 mm in diameter, respectively, both with the length of 120 mm for the purpose of rapid heat treatment were prepared. The steel was heat treated by austenitization for 2 hours at 950°C followed by oil quenching (OQ) and subsequent tempering for 1, 6, 24, 100 and 200 hours at 710°C with air cooling, and initial ferritic-carbidic mixture (marked as A, B,C,D, and E) with various dispersity of globular cementite was thus obtained. Based on analyses of phase transformations (Dlouhý 1992 b), the processing conditions for two sets of experiments were specified: (i) Rapid resistance heating of ferritic-carbidic mixture, C, with the rate of 75 K/s to austenitizing temperatures ranging from 850 to 1100°C followed by direct oil quenching. Some specimens were rapid heated to 1100°C and quenched from this temperature after the holding time (τ) of 0,3,10,30, and 90 seconds, respectively. (ii) Rapid heating (100 K/s) to temperatures specific for individual initial microstructures (Table 2). Obtaining low carbon martensitic matrix independently on initial carbide phase dispersity could be enabled under these conditions.

The rods were cut, tempered for 1 hour at 300°C *), and then machined down to the geometry and dimensions of the mechanical test specimens. The room temperature tensile properties (YS by 0,2 % offset method, UTS) were measured using cylindrical tensile specimens with a gage diameter of 3,5 mm, and a gage length of 40 mm. The Charpy V notch energy (CVN) was evaluated using Charpy bars. Plane strain fracture toughness (K_{IC}) was measured by static three-point bending method on 10x10x55 mm fatigue precracked specimens.

The fracture surfaces of the K_{IC} specimens for each microstructure were examined in SEM. To determine the void sizes quantitatively, fractographs were taken within a distance from the crack tip of 3 times of the critical crack tip opening displacement in the crack growth direction and near the mid-thickness position of the specimen. The size distribution of all voids was determined. In addition, individual measurements of particles were done using the EDS system.

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*) To avoid the delayed fracture in the as-quenched state, such tempering conditions were chosen that for all the martensitic microstructures investigated the influence of tempering carbides and of retained austenite on fracture behaviour was insignificant.

RESULTS

Table 1 gives the data on microstructural parameters obtained by means of quantitative metallography. The rapid austenitizing temperatures t_{RA} as values representing the heat treatment were used to differentiate three types of microstructures investigated: Low carbon martensite with globular cementite particles (LCM+CARB - see **Fig. 1a**), medium carbon martensite with cementite particles (MCM+CARB), and martensite containing microregions with increased content of carbide-forming elements (M+MICCE, see **Fig. 1b**, Dlouhý 1992a).

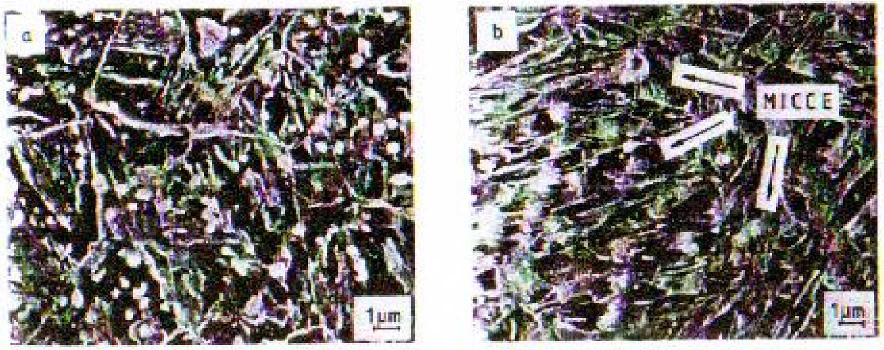

Fig. 1: SEM micrographs of low carbon martensite with remainders of secondary cementite (a) and martensite with MICCE (b).

Table 1: Heat treatment condition and characteristics of microstructures obtained

t_{RA} [°C]	N_A [100 µm^{-2}]	d_C [µm]	v_f	c_{Cm}(x-) [wt.%]	d_{AG} [µm]	RA [vol.%]
initial C	68±15	0,223	0,0461			
900	67±20	0,277	0,0457	0,151	8,5	3,5±0,7
1000	40±15	0,221	0,0310	0,270	8,9	3,7±1,1
1100	30±20⁺)	0,0	0,0	0,300	12,8	4,0±1,0
conv.treat.				0,300	12,1	3,5±1,2

N_A - number of cementite particles per unit area, d_C - mean diameter of cementite particles, c_{Cm} - mean matrix carbon content derived from x-ray measurements, d_{AG} prior austenite grain size, RA - retained austenite content; $^{+}$) area density of MICCE.

On the left part of Fig. 2, with open points, yield stress and fracture toughness of the structures obtained by rapid austenitizing to different temperatures, followed by immediate oil quenching, are presented. The points with dots represent the data of the martensite with MICCE. In the right part of the figure, the yield stress and fracture toughness of microstructures obtained by rapid heating up to 1100°C with immediate quenching and with gradually increasing holding time at this temperature until 90 s, are illustrated. The full points represent the properties of the microstructures produced by conventional treatment (900°/30′/OQ, tempered at 300°C).
In comparison to the microstructure obtained by conventional

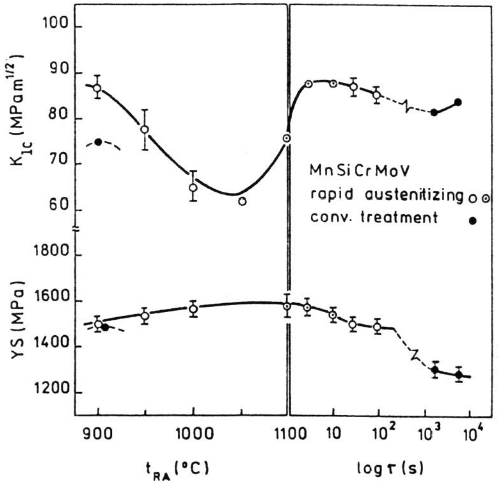

Fig. 2: Plane strain fracture toughness K_{IC} and yield stress YS in dependence on austenitizing condition

treatment (Tab. 1), low carbon martensite with undissolved cementite particles (t_{RA} - 900°C) has a relatively high fracture toughness. The yield stress values of both microstructures are nearly comparable. When rapid heating to 950-1050 °C is applied the increase in the mean matrix carbon content is stated (Tab. 1). In the presence of undissolved cementite particles the significant drop of toughness is observed. The yield stress of the structure with MICCE is by almost 90 MPa higher compared to homogeneous martensite. The fracture toughness is only slightly higher for t_{RA} 1100°C and more considerably, for holding time 3-10s at this temperature. If the holding time is longer than 10 second both the strength level and fracture properties decrease.

Table 2: Heat treatment conditions and characteristics of microstructures obtained

initial	t_{RA} [°C]	d_C [μm]	v_f	c_{Cm}(x-) [wt.%]	d_{AG} [μm]	RA [vol.%]
A	895	0,11	0,040	0,154	7,8	2,7±1,5
B	895	0,15	0,039	0,145	7,5	2,9±2,0
C	905	0,22	0,044	0,151	8,5	3,5±0,7
D	935	0,28	0,042	0,155	8,4	3,2±1,1
E	945	0,41	0,040	0,130	8,7	2,9±1,5

Dependence of yield strenght and fracture toughness on globular cementite particles (**Fig. 3**) dispersity is specific mainly by the matrix with comparable intrinsic properties rising from comparable carbon content (open points). Except for data in Table 2 changes in distribution of secondary cementite, it is obvious from a series of microstructures on **Fig. 4**. The optimal combination of strength and fracture toughness is reached for cementite particles dispersity obtained for the initial mixture B.

The micrographs in **Fig. 5** show the surface morphology in the region between fatigue precrack and overload fracture. Under the stretched zone (SZ) there is the cliff (C) the plane of which has an orientation of about 70-90° to the static fracture region (for low carbon martensite with carbides and partly for martensite with MICCE). The latter is covered partly by dimples originating from

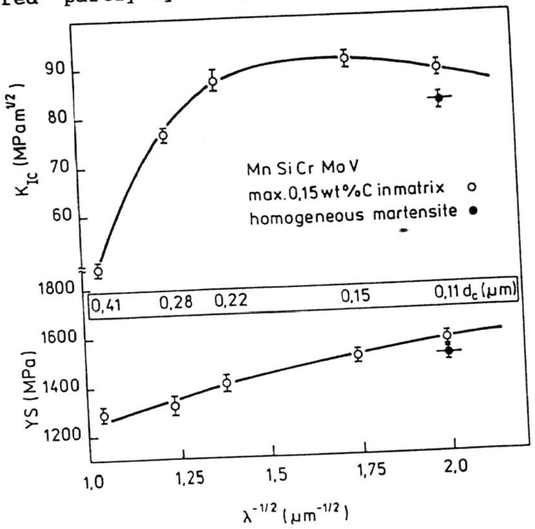

Fig. 3: Yield stress and fracture toughness as a function of inverse square root of carbide interparticle spacing.

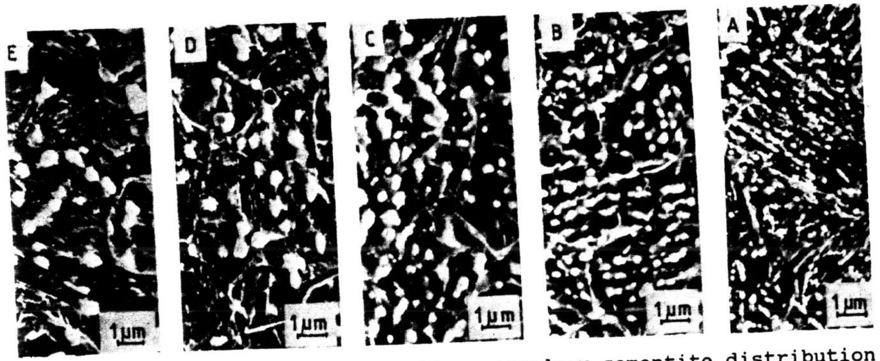

Fig 4: SEM micrographs illustrating secondary cementite distribution in low carbon martensitic matrix.

secondary phase particles (inclusions and carbides), and partly by microdimples nucleated from tempering carbides. The cliff's height tends to be greater in the tougher microstructures. The stretch zone is smaller and the cliff may also absent, especially in the less tough microstructures (for martensite with 0,27 wt % C and carbides).

DISCUSSION

Possible static fracture model. As the onset of fracture initiation the stage is considered at which the blunted crack front links up with the void that has nucleated in the local plastic zone (Knott 1980, Kumar 1984, Di Russo 1986). Two groups of factors are responsible

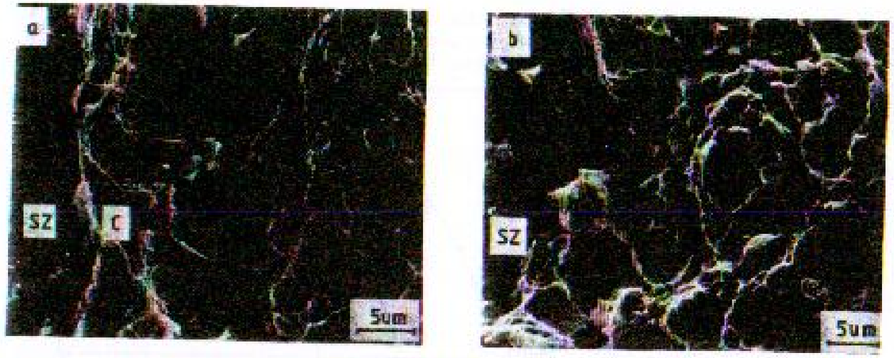

Fig. 5: Fracture surface of K_{IC} testpieces showing the transition between fatigue (left) and overload fracture (right); a - low carbon martensite with carbides, b - medium carbon martensite with carbides.

- both the factors deciding the nucleation of the primary voids (Briant 1982) and those controlling the linking up of crack front with this void (Knott 1980). For the nucleation of primary voids, the interface between the nonmetallic inclusions and the martensitic matrix represents the preferable sites. It is the appropriate combination of sites, forms and types of inclusions with the matrix properties and local stress deformation condition which is decisive for the nucleation site activation. The effect of nonmetallic inclusions on fracture initiation mechanism seems to be satisfactorily solved, a parametr among those characterizing their distribution represents a part of the majority of quantitative models. Linking up of crack front with primary voids by the microvoid nucleation is being immediately controlled by the plastic properties of the matrix, by the maximum stress level σ_y ahead of the crack front, and by the cohesion strength σ_c of the particle - matrix interface (Knott 1988, Lee 1989). The deformation ability of the matrix (intrinsic matrix toughness) is decisive for the rise of dislocation clusters and local stresses which can overcome the interface cohesion strength σ_c.

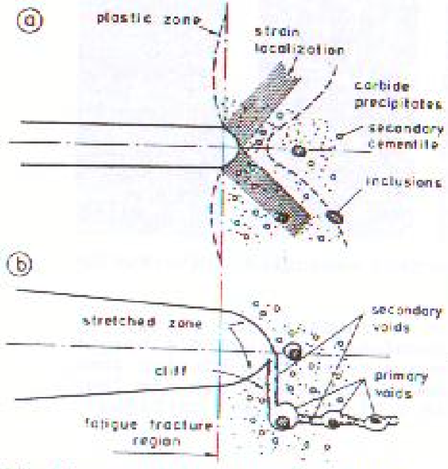

Fig 6: Possible micromechanism at the onset of unstable crack propagation for high plastic martensitic matrix

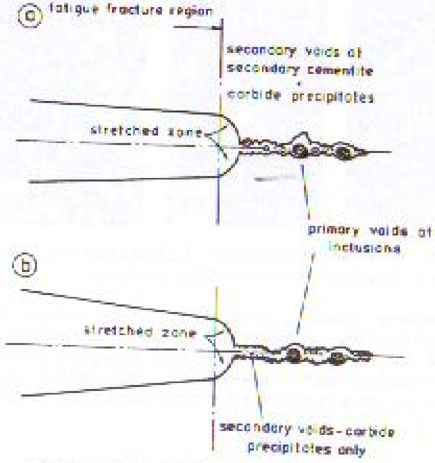

Fig.7: Iniciation mechanisms for lower plasticity of martensite a) with and b) without secondary cementite remainders

Influence of intrinsic toughness of the matrix. When assessing the parameters that condition a material's fracture toughness, all the microstructural factors should not be considered without regard to the plasticity characteristics of the matrix at the crack tip. Such a material in which high plastic deformation occurs before microvoids nucleate can be taken as implicitly tough. The results presented in Fig.8 give evidence of significant influence of carbon content in solid solution of martensite, especially at the presence of undissolved cementite particles:

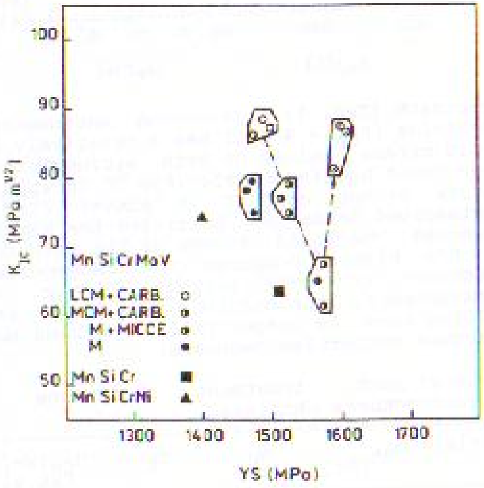

Fig.8: Comparison of strength and fracture properties for microstructures investigated

(i) Open points (on Fig. 8) represent the low carbon martensite with remainders of undissolved cementite particles. For this microstructure increasing toughness (K_{IC}) is caused by the ultra-low carbon content in solid solution in as-quenched state, as the case may be, by low carbon concetration and low volume fraction of tempering carbides in the matrix under the low tempering temperature conditions. The increased initiaction resistance is conditioned by high intrinsic toughness of martensitic matrix

(ii) Halved points show the state, in which the partial carbide dissolution alrerady occurred. An increase of matrix carbon content results in its reduced plasiticity. In the presence of the remainders of undissolved cementite with discrete interface boundary it means an increased frequency of void nucleation sites in earlier stages of mechanical loading which results in drop of fracture toughness. This drop of fracture resistance is connected with change in initiation micromechanism, the situation like on Fig.7a) is observed.

(iii) Further carbide dissolution results in gradual "removing" of critical carbide-matrix interface boundary from the structure (on

Fig. 8 points with dots). In spite of further increasing of carbon content in solid solution, the reduction of void nucleation sites results in expressive increasing of fracture toughness. The nucleation of voids occurs on relatively smaller tempering carbides in later stages of plastic deformation. Fracture initiation mechanism was changed in that, observed in case (i) and/or on a mechanism like that on Fig. 7b).

(iv) Further rearrangement of substitutional alloying elements under conditions, corresponding nearly to conventional austenitization, results in a change of properties, marked with full points.

The role of secondary cementite particles. The dispersion of undissolved cementite particles in the martensitic matrix thus involves, apart from its positive effect on strength properties, also the risk of a higher probability that some structural discontinuities will be formed during mechanical loading. A decrease in ductility and also in fracture resistance against low energy tearing is therefore observed, especially when the volume fraction of carbides is higher (Dlouhý 1992b). The reason of it is connected with the interaction between cementite particles and dislocations in the nucleation stage of voids, and in local plasticity of the martensitic matrix during void growth and coalescence. From this point of view, the carbide-matrix interface boundary is a critical locality controlling the level of ductility. Excelent evidence on deleterious effect on carbide-matrix interface boundary is given by the fracture behaviour of microstructure from which this boundary was removed, i.e, comparison of medium carbon matrix with secondary cementite and those with MICCE (Fig. 8).

CONCLUSIONS

Fracture micromechanims in heterogeneous martensitic structures were discussed in terms of several concurrent micromechanistic phenomena involving intrinsic toughness of the martensitic matrix, and the carbide-matrix interface boundary decohesion susceptibility. Increasing carbon content in matrix at the present second phase particles act as strengthening but embrittling factor. The removing from medium carbon martensitic matrix of carbide-matrix interface boundary acts as the toughening factor without loss of strength.

Compared to homogeneous martensite the matrix of low carbon martensite containing the undissolved carbides exhibits a significant increase in fracture toughness without loss of strength. Microregions with enhanced content of carbide-forming elements results in the increase of both strength and fracture toughness. On the contrary, at the presence of remainders of undissolved carbides in a martensitic matrix with higher carbon content a decrease in fracture toughness was stated at the maintained strength level.

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