STRUCTURE-METALLURGICAL APPROACH TO DESIGN OF STEELS WITH HIGH CRACK PROPAGATION RESISTANCE

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ABSTRACT

The influence of various microstructural factors on fatigue thresholds ΔK_{th} is evaluated with respect to heat treatment conditions and chemical composition of steels. Grain size, cold deformation and alloying are proved to affect ΔK_{th} level of ferritic steels. Experimental evidence suggests that the shape and the size of the cementite in ferritic eutectoid steels, carbon content and tempering temperature in Cr-Mn-Si-Ti steels, relative ferrite/martensite content in dual phase steels excert influence on ΔK_{th}. It is shown that ΔK_{th} - fatigue limit σ_{w} and ΔK_{th} - fracture toughness K_{1c} dependencies may acquire two types of realization.

KEY WORDS

Steel; microstructure; fatigue crack growth rate; threshold; fracture toughness.

Serviceability and fatigue life of a major number of different structural elements functioning in unfavourable conditions is determined to a great extent by the material resistance to crack nucleation and propagation under short-time single and long-time cyclic loading. In such cases the most important material characteristics (besides yield strength) are: fatigue limit, fracture toughness, fatigue crack propagation rate (in the first place in the near-threshold region). Considerable means for increasing the strength and crack resistance of steels are provided by technological methods, which change their structure-metallurgical state [1-3]. Elucidation of complex interrelation between the material structure and its crack resistance and fracture micromechanism allows determination of the optimal ways to control the material properties by changing its microstructure. The aim of the present paper is to establish the general influence of the structure-metallurgical state of engineering steels on their crack resistance. Special attention is paid to subcritical fatigue crack resistance and formation of threshold alternating stress intensity factor ΔK_{th} values, characterising non-propagation of cracks under cyclic loading.

Taking into consideration the variety of structural states of steels due to their chemical composition and heat treatment conditions, analysis of the influence of various structural factors was conducted separately for single-phase materials, high- and low-temperature tempered steels and pearlitic steels. Grain size (d), being the principal parameter for single-phase materials, excerts considerable influence on their mechanical properties, including their cyclic crack resistance [4-8]. Ferritic grain size increase in mild steel (0.035%C, 0.02%Cr, 0.14%Mn) leads to ΔK_{th} increase - this tendency being described in earlier works [4-7]. ΔK_{th} dependency from d is well described by an expression similar to the Petch-Hall equation for yield strength differing, however, by the sign of the power exponent: $\Delta K_{th} = 5.85 + 230d^{1/2}$. If for low carbon steels intergranular fracture under amplitudes close to ΔK_{th} becomes possible, the above described grain size effect undergoes inversion [7]. The ΔK_{th} level and crack propagation resistance of single-phase materials may be considerably changed by dislocation structure alteration and alloying [9,10]. It has been determined, that cold-work of ferrite under rolling deformation promotes a simultaneous yield strength and ΔK_{th} increase. Thus, creation of ordered dislocations structure in iron effectively increases subcritical crack

propagation resistance (Fig. 1a). The influence on ΔK_{th} of alloying iron with elements forming substitutional solid solutions requires further investigation. It is only known [10], that in high-chromium ferrite (~25%Cr) the low rate fatigue crack propagation resistance is considerably higher than in pure iron. The ΔK_{th} level may also be increased by tempering high chromium ferrite at 480°C. In this case globular particles of α'- phase enriched in chromium and coherently bound to the matrix are found, which promotes succeptibility of the steel to mechanical twinning (Fig. 1b) [10].

Considerable possibilities for crack resistance control are provided for medium alloyed carbon steels. Testing of 0.8C-Cr-Mn-Si-Ti lamellar pearlitic steel showed, that for low ΔK_{th} values its fatigue crack propagation resistance increases with an increase in pearlite interlamellar spacing and pearlite colony size. Special attention is drawn to the analogous effects of pearlite colony size and ferritic grain size on ΔK_{th}. Spheroidized pearlitic structures exhibited higher crack resistance levels as compared with lamellar pearlite; however they showed ΔK_{th} insensivity to the increase of spacing between particles from 0.45 to 1.23 μm. Transition to spheroidized structures (Fig. 2a) with large globules (~2.5 μm) is accompanied by a decrease in ΔK_{th} - in this case carbide particles act not only as a strengthening factor but also as potential sources of crack nucleation. It has been shown for Cr-Mn-Si-Ti steels with carbon content from 0.32 to 0.93% that for pearlitic structures the highest ΔK_{th} values as well as optimal $\sigma_{0,2}$ and ΔK_{th} combination is achieved when a dispersed structure is formed (Fig. 2b) at the stage of carbide spheroidization during high-temperature tempering. A positive effect of such structure formation on ΔK_{th} is intensified by the increase of carbide phase volume due to a carbon content increase in steel (Fig. 3) [8].

A drop in the crack resistance of quenched steels is observed under a tempering temperature decrease. While analysing the structure-metallurgical aspects of the crack resistance of quenched carbon steels it is necessary to take into consideration such factors as martensite morphology, prior austenite grain size, residual austenite presence and internal microstress effects. As for quenched low-temperature tempered steels thresholds become somewhat lower with an increase of carbon content (Fig. 3), which is probably due to martensite morphology changes. A number of works deal with the prior austenite grain size effect on ΔK_{th} in martensitic steels [6,11], though data available are frequently contradictory (Table 1). In general, these effects on ΔK_{th} manifest themselves to a considerably smaller degree as compared with the effect of d on ΔK_{th} in low-carbon steels. An austenite grain size increase leads to an effective increase of fracture toughness K_{1c} in quenched low-temperature tempered 0.45%C steel [11, 12].

Research, carried out on 0.2%C-14%Cr steel specimens quenched with various cooling rates showed, that martensite crack resistance is to a great extent determined by the internal microstress level, decreasing with their increase. It has been shown that near-threshold crack resistance may be increased up to two times for a decrease in internal microstress. Application of various techniques of melting and subsequent refining which provide the change of form, size and type of non-metallic inclusions without alteration of the martensitic carbon steels chemical composition, exerted no influence on ΔK_{th}, though it had a certain effect under medium and high fatigue crack propagation rates. Higher crack resistance is promoted by liquid metal blow-through by argon in the ladle, which leads to the formation of "composite" inclusions with soft sulphide shells [14].

Recently, special attention has been paid to low-carbon alloy steels with a dual phase ferritic-martensitic structure, which is obtained by quenching from the intercritical ($A_{c1}-A_{c2}$) temperature region [15]. Testing carried out for 0.06%C steel shows the possibility of simultaneous realization of both high ΔK_{th} and σ_{ω}, K_{1c} values (Fig.4,5). However, it has been shown (for 0.4%C-Cr steel [16]) that for steels with a high carbon content the formation of a dual-phase structure has minor effect.

Experimental data available allow the correlation of the most important mechanical properties ($\sigma_{0.2}$, σ_{ω}, K_{1c}, ΔK_{th}). Taking into consideration the fact that there exists an almost proportional correlation between $\sigma_{0.2}$ and fatigue limit σ_{ω} (noticeably violated only in the high-strength region) a diagram of ΔK_{th} vs. σ_{ω} (shown in Fig. 4) is worth consideration. It follows from the diagram, that an inversely proportional relation is observed between ΔK_{th} and σ_{ω}. This confirms once more, that threshold ΔK_{th} as a characteristic of crack non-propagation has a different physical meaning from that of σ_{ω}, which in the first place is the threshold macrocrack initiation indicator under cyclic loading. However, due to structural control possibilities arise which permit deviation from the above given relation and the opportunity to optimize the properties of steels by simultaneous ΔK_{th} and σ_{ω} increase. For the low-strength region (A) dual-phase ferritic-martensitic structures formation provides high ΔK_{th} values. In the medium strength region (B) precipitation hardened ferritic structure have favourable prospects. Figure 5 shows the influence of various strengthening methods on the interrelation of ΔK_{th} and fracture toughness K_{1c}. Strengthening by ferrite cold working, alloying and dual-phase structure formation (testing temperature decrease) in low carbon steels lead to an inversely proportional ΔK_{th} - K_{1c} correlation. At the same time, strength decrease of medium and low carbon steels by increase of tempering temperature leads ro a ΔK_{th} value increase as well as to higher threshold crack growth resistance. However, possibilities for the fatigue threshold increase of high-strength steels with lowest ΔK_{th} values require further investigation.

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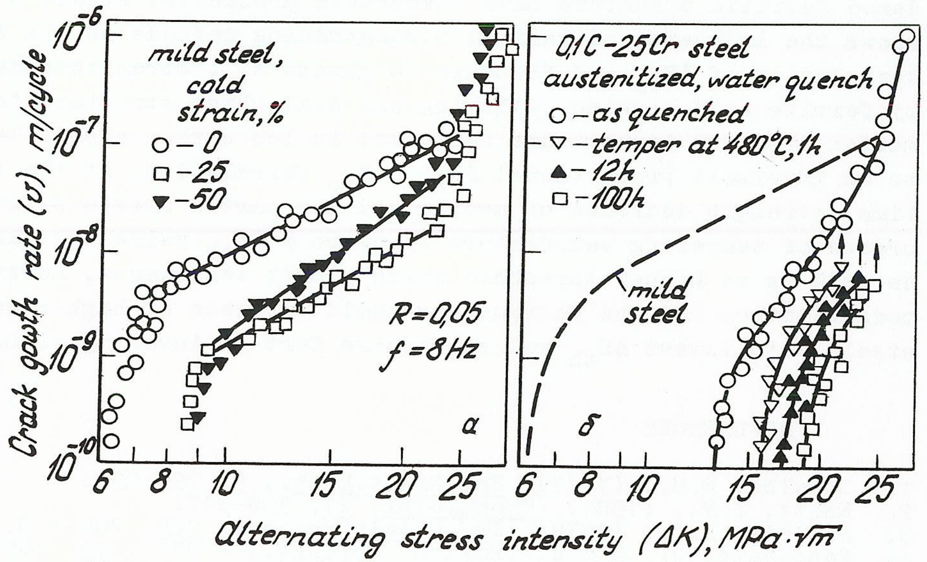

Fig. 1. Fatigue crack growth curves for mild steel specimens subjected to plastic deformation with reduction degrees ε (a) and for 0.1C-25Cr steel specimens, quenched and tempered under 480°C for τ hours (b).

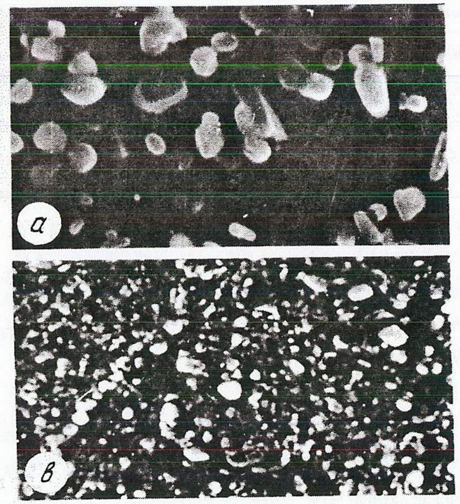

Fig. 2. Microstructure of 0.8C-Cr-Mn-Si-Ti steel specimens quenched and tempered at 700°C, 60 h(a) and 600°C, 2 h(b).

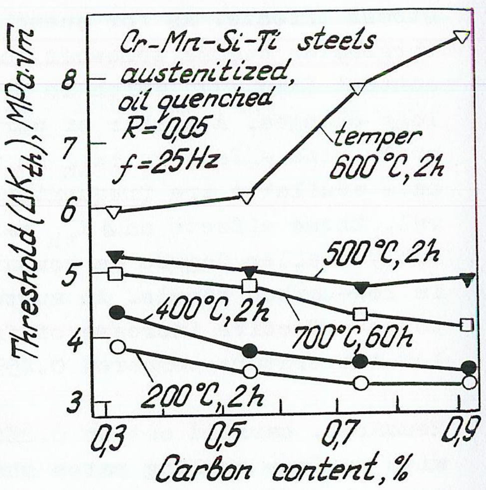

Fig. 3. Threshold ΔK_{th} values dependence on carbon content in quenched and tempered specimens.

TABLE 1

Material	Prior Austenite Grain Size, μm	Monotonic Yield Stress, MPa	ΔK_{th} MPa$\sqrt{m}$	K_{1c} MPa$\sqrt{m}$	Ref.
0.35C-4Cr steel	30	1324	4.38	58	6
- " -	90	1324	3.40	76	6
- " -	180	1324	3.0	79	6
0.45C-Cr-2Ni-Mo-V steel	9	1805	4.62	41	11
- " -	140	1780	4.9	60	11
0.45C-Cr-2Ni-Mo-V steel with sawtooth shape of prior austenite grain boundaries	140	1952	5.8	68	11

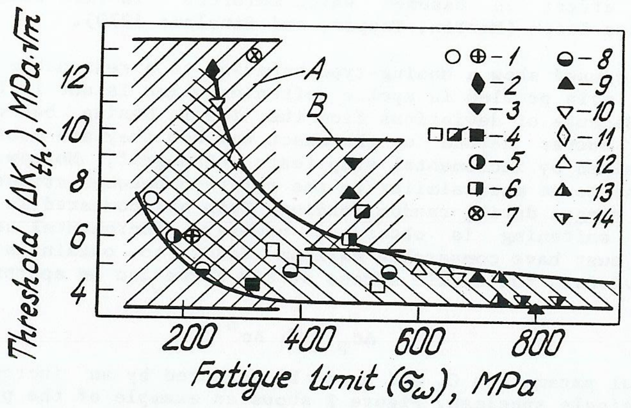

Fig. 4. Variation of threshold ΔK_{th} with fatigue lumut for steels: 1 - mild steels, ferrite[9, 17]; 2 - 0.1C-25Cr steels, ferrite; 3 - 1/2Cr-1/2Mo-1/4V steel, precipitation hardened ferritic microstructure[18]; 4 - 0.8C-Cr-Mn-Si-Ti steel, pearlite; 5 - 0.2C-14Cr steel, annealed; 6 - Cr-Mo steel[17]; 7 - 0.06C-2Cr-2Ni-Mo steel, ferrite-martensite; 8 - HT80, SM50, SB22 steels[19]; 9 - 0.8C-Cr-Mn-Si-Ti steel, quenched and tempered at 400-600°C; 10 - 0.2C-14Cr steel, as quenched; 11 - G40.11 steel[20]. 12 - 0.3C-Cr-Mn-Si-2Ni steel, quenched and tempered at 250°C; 13 - 0.4C-Cr-Mo-V steel, quenched and tempered at 200°C; 14 - 1C-1.5Cr steel, quenched and tempered at 150°C.

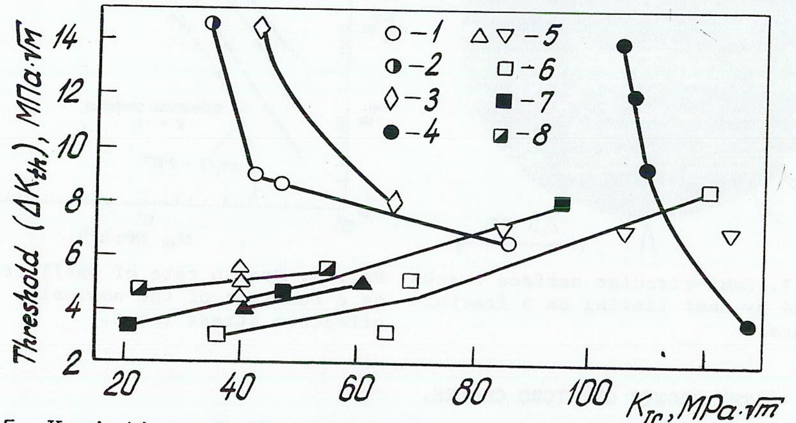

Fig. 5. Variation of threshold ΔK_{th} with fracture toughness for steels: 1- mild steel, cold working; 2- 0.1C-25Cr steel; 3- mild steel, low temperature tests[21] ; 4- 0.06C-2Cr-2Ni-Mo steel, duplex microstructure; 5- 0.8C-Cr-Mn-Si-Ti steel, pearlite; 6- 300M steel, quenched and tempered at 100-650°C[3]; 7- 0.4C-Cr-Si steel, quenched and tempered at 200-500°C; 8- 0.8C-Cr-Mn-Si-Ti steel, quenched and tempered at 200-600°c.