THE INFLUENCE OF MICROSTRUCTURE ON STAGE II FATIGUE CRACK GROWTH IN A 13 pct CHROMIUM STEEL

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ABSTRACT

In this study fatigue crack growth rates have been determined for different microstructural conditions in a 13 pct chromium steel. Although all treatments resulted in varying microstructures, it was possible to obtain two pairs of treatments that had two different yield strengths but the same value within each pair. It has been concluded that neither microstructure nor the yield strength alone governs the fatigue crack growth rates in stage II. The operative fracture mechanisms play an important role in determining the material resistance to fatigue crack growth.

KEYWORDS

13 pct chromium steel; fatigue crack growth; microstructure; yield strength; micromechanisms of failure.

INTRODUCTION

The effect of microstructure on fatigue crack propagation in steels has been studied extensively (Benson and Edmonds, 1979; Jones, 1981; Minakawa, Matsuo and McEvily, 1982; Gray and coworkers, 1982; Po-Wekao and Byrne, 1982). Microstructure has been varied to obtain pearlite of different degrees of fineness, bainite, martensite etc. for comparison of crack propagation rates. In a 1018 steel Minakawa, Matsuo and McEvily (1982) found an order of magnitude difference in crack growth rates in two different microstructures. On the other hand, in the same study no difference in crack growth rates has been reported in 10B35 steel for similar structures. Benson and Edmonds (1979) found negligible difference in crack growth rates between a ferritic and a bainitic structure in 0.5 Cr-0.5 Mo-0.25 V steel. Jones (1981) in his work on a martensitic steel has observed enhanced crack growth rate in the furnace cooled condition as compared to the water quenched condition, which was explained on the basis of grain boundary embrittlement in the former. Gray and coworkers (1982) and Po-Wekao and Byrne (1982) have observed decreasing crack growth rates by reducing the pearlitic inter-lamellar spacing in eutectoid steels. On the other hand, a number of researchers (Bathias and Pelloux, 1973; Van Swam, Pelloux and Grant, 1975; Ritchie, 1979; Wilson, 1980) have reported no effect of microstructure on fatigue crack growth rates in steels.

The present study has been undertaken to investigate the effect of microstructure and yield strength of a 13 pct chromium steel on stage II fatigue crack propagation.

EXPERIMENTAL

The chemical composition of the steel investigated is given in Table I. The different heat treatments and the corresponding microstructures and the yield strength values are listed in Table II. Crack propagation experiments were conducted on a servo-hydraulic 250 kN M.TS Materials Testing System. The samples were of SEN type. The stress ratio was maintained at R=0.1 and the crack growth was monitored using a travelling microscope to an accuracy of 0.01 mm. The fractured surfaces of the tested samples were examined in a Cambridge Stereoscan S-150 SEM.

RESULTS AND DISCUSSION

The microstructures obtained in the different heat treated conditions are shown in Figs.1a to e. The austenitising treatment was maintained constant at 1050°C for 1h in all the five cases. Tempering at 700°C and at 800°C (Figs.1a and b) gives rise to $M_{23}C_6$ carbide precipitates (Gooch, 1982) at prior austenitic or martensitic lath boundaries as well as within the grains. In the former case (Fig.1a) these precipitates are irregular particles when formed at austenite grain boundaries and needles when they occur at the martensite lath boundaries (Hede and Aronsson, 1969). When tempered at 800°C the precipitates are coarser, their shape near spherical while the martensitic lath structure disappeared. Tempering at 550°C (Fig.1c) results in M_7C_3 type carbide precipitates and segregation of impurities along the prior austenite grain boundaries (Prabhu Gaunker, Huntz and Lacombe, 1980). Air cooling from the austenitising temperature results in martensite (Fig.1d) with small amount of dispersed carbides of the type M_3C (Gooch, 1982). A pearlitic network along prior austenite grain boundaries in a ferritic matrix resulted on furnace cooling from the austenitising temperature (Fig.1e).

The effect of the microstructure on yield strength of the material has been such that B and E treatments gave approximately equal yield strength of 451 and 431 MPa and C and D treatments gave similar yeild strength of 764 and 784 MPa respectively. The treatment A gave rise to an intermediate yield strength of 686 MPa.

In Figs.2a and b are presented the data relating to da/dN vs ΔK in the crack growth range of 10^{-6} to 10^{-3} mm/cycle. The former depicts the crack growth rate for conditions A and B falling on one line while the later shows similar data for conditions C,D and E falling on one but a different line. The relation between da/dN and ΔK under the two sets of conditions can be described as

$$\frac{da}{dN} = 2.8 \times 10^{-11} \; (\Delta K)^{3.21} \text{ for A and B} \tag{1}$$

and

$$\frac{da}{dN} = 6.6 \times 10^{-11} \; (\Delta K)^{3.63} \text{ for C, D and E} \tag{2}$$

conditions of heat treatment. These relations clearly reveal that tempering at 700 and 800°C (A and B) contributes to better resistance to crack propagation in comparison to the other treatments. The analysis also clearly indicates that though B and E conditions give the same yield strength of around 450 MPa, the B condition is better in terms of fatigue resistance. In a similar way though the microstructural conditions of C, D and E are different from each other, all the three show similar crack growth behaviour.

The SEM fractographs of the fractured surfaces corresponding to the five heat treatments are shown in Figs.3a to e. While ductile fracture type could be identified in A and B conditions, embrittlement type and microcleavage fractures can be seen in C, D and E conditions. This explains why the treatments A and B impart better resistance to fatigue fracture than the treatments C, D and E. For the sake of clarity the microstructural features developed during the heat treatments used in this investigation are schematically represented in Fig.4.

Though literature presents conflicting results pertaining to the effect of microstructure on stage II crack growth in steels, some clear statements can be made on the basis of the present work. Heat treatments C, D and E, altough resulted in different microstructures, gave the same crack growth rates. The same statement applies to conditions A and B. Conditions forming different microstructures but possessing nearly the same yield strength (C and D) can give idential growth rate plots. However, conditions giving different yield strengths, namely (A and B) or (D and E), can also result in the same stage II crack growth rates. These results therefore lead to the conclusion that it is neither the microstructure alone nor the yield strength alone that governs the fatigue crack growth rate in stage II. The difference we get with A and B heat treatments on the one side and C, D and E heat treatments on the other is mainly because of the type and distribution of carbide precipitates in A and B which contribute to ductile type failure and show lower crack growth rates than conditions C, D and E where intergranular mode of fracture and microcleavage prevail which lead to higher crack growth rates.

CONCLUSION

For the 13 pct chromium steel investigated it is inferred that it is neither the yield strength alone nor the microstructure alone that governs the crack propagation rate in stage II. The higher but similar crack growth rates obtained in C, D and E conditions of heat treatment are attributable to grain boundary cracking and microcleavage while the lower crack growth rates in conditions A and B correspond to ductile fracture features.

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Table-I: Chemical composition of 13 pct Chromium steel investigated

C	Mn	Si	S	P	Cr	Mo	V
0.22	0.42	0.35	0.01	0.016	13.0	0.42	0.22

Table II: Heat treatments, the resultant microstructures and the yield strength values

Designation	Heat treatment	Microstructure	Yield Strength (MPa)
A	1050°C/1h, WQ 700°C/5h, AC	Tempered martensite, fine precipitates of $M_{23}C_6$	686
B	1050°C/1h, WQ 800°C/5h, AC	Coarse precipitates of $M_{23}C_6$ in a ferritic matrix	451
C	1050°C/1h, WQ 550°C/5h, AC	Tempered martensite, grain boundary segregation	764
D	1050°C/1h, AC	Auto-tempered martensite	784
E	1050°C/1h FC to 600°C, AC to 25°C	Pearlitic grain boundary net-work in a ferrite matrix	431

WQ-Water quenched; AC-Air cooled; FC-Furnace cooled

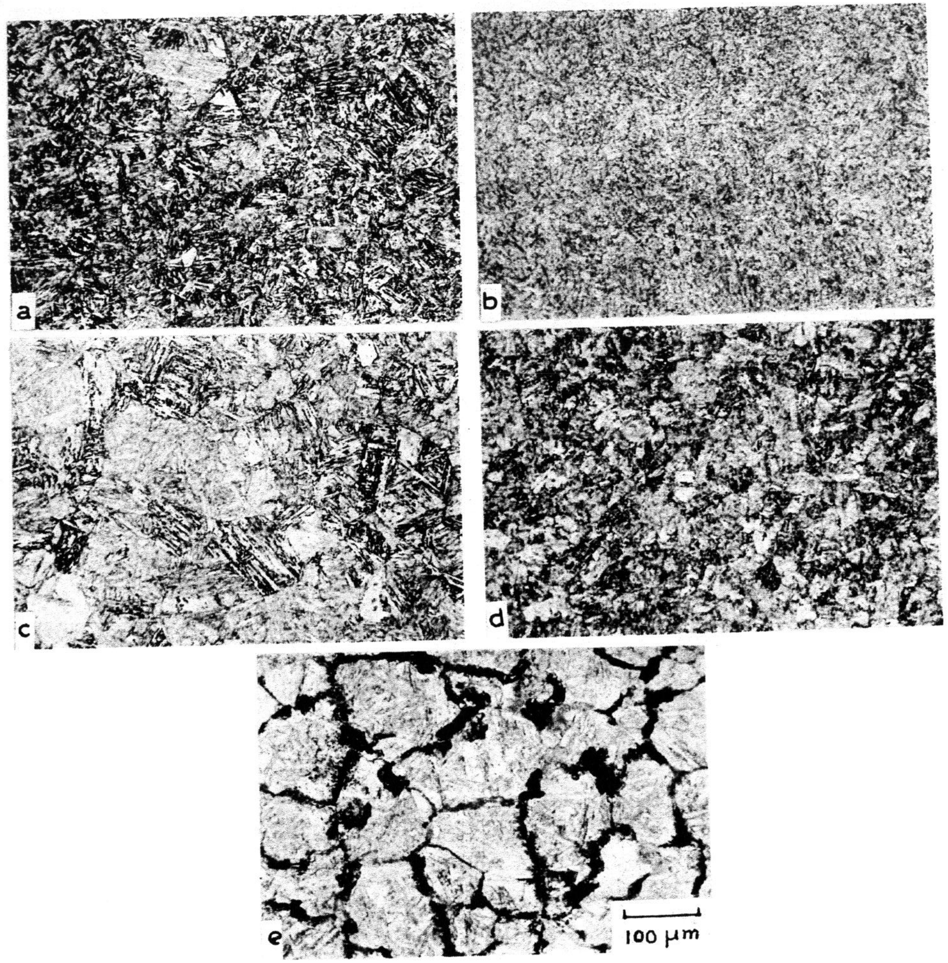

Fig.1. Microstructures of a 13 pct chromium steel.

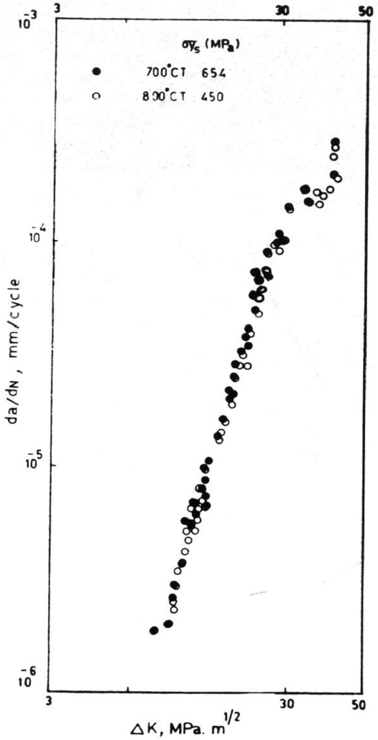

Fig.2a. da/dN vs ΔK plots for microstructures A and B.

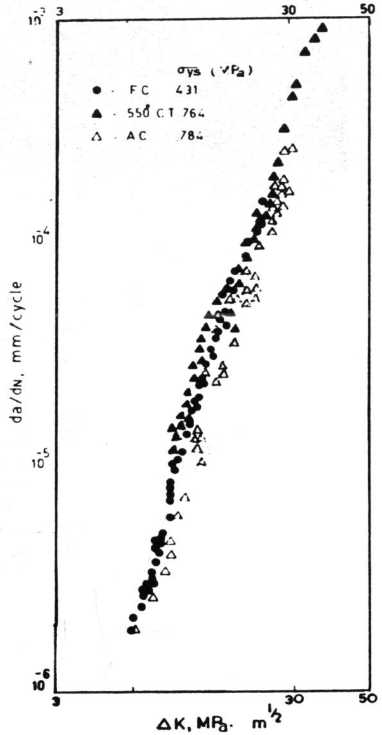

Fig.2b. da/dN vs ΔK plots for microstructures C, D and E.

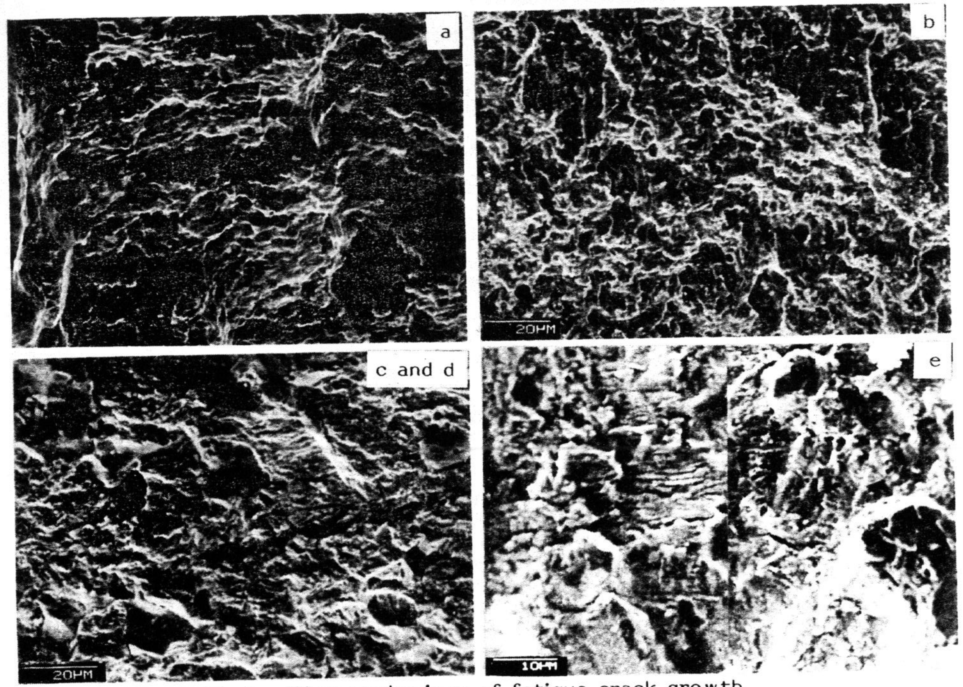

Fig.3. Micromechanisms of fatigue crack growth

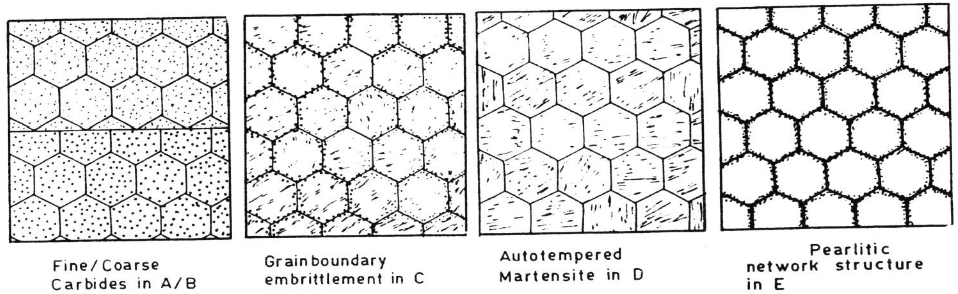

Fig.4. Schematic representation of microstructures studied