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Phil. Trans. R. Soc. Lond. A 1980 **295**, 210-211
doi: 10.1098/rsta.1980.0102

Phil. Trans. R. Soc. Lond. A **295**, 210–211 (1980) [210]
Printed in Great Britain

Interaction effects between trace element impurities and environment in fatigue of high strength steels*

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The interaction of air environments and trace element impurities during fatigue of quenched and tempered steels has been examined. Fatigue crack growth tests have been performed in laboratory air, dry air (90 μg/g H_2O) and vacuum (0.133 mPa) on a steel produced to two levels of purity. The steels were to BS 817M40 (0.4 C, 1.4 Ni, 1.0 Cr, 0.3 Mo, 0.6 Mn, 0.2 Si). One steel contained trace elements (in micrograms per gram) of 300 S, 160P, 190Sn, 320As, 65Sb and 1600 Cu. The other had reduced concentrations of these elements: 30S, 10P, 20Sn, 16As, 14Sb, 100Cu.

The steels were austenitized at 840 °C, oil quenched and tempered at 200 °C for 1 h, followed by water quenching. This treatment produced a 0.2 % proof strength of 1570 MPa. Impurity concentrations on the prior austenite boundaries were determined by fracturing specimens of the impure steel inside an Auger apparatus. Impurity concentrations on the exposed grain boundaries were determined to be 8 % of a monolayer of phosphorus, with other elements less than 2 %.

Fatigue crack growth rates were compared in air and in vacuum environments. In vacuum, there was little difference between the steels, except that at the highest stress intensity (K) levels, the impure steel had faster growth rates than the pure. Growth rates at high levels of K were sensitive to mean stress. In air, growth rates were up to 10 times faster, especially at low K values, and the impure steel showed enhanced mean stress sensitivity. Growth rates were always fastest in this steel.

Measurements of the percentage of the fracture face covered with intergranular fracture (i.g.f.) were made on a scanning electron microscope. These showed that the percentage of i.g.f. varied considerably with ΔK and material purity. I.g.f. peaked at ΔK values of 12–16 MN $m^{-\frac{3}{2}}$ and declined to less than 10 % at both lower and higher values of ΔK. For the pure steel, the maximum was about 35 %, but in the impure steel 90 % of the fracture face was covered with i.g.f. For a given ΔK value there was always a higher proportion of i.g.f. in the impure steel, which could be correlated with faster growth rates, but there was generally no direct correspondence between i.g.f. and crack growth rate.

Fracture faces produced in vacuum had no i.g.f. in the pure steel, and it was much reduced in the impure steel. Tests conducted in dry air in the high purity steel also failed to produce any i.g.f.

It is concluded that water vapour is responsible for most of the i.g.f. in these steels, and that it acts by reacting with the clean metal surface at the crack tip, producing hydrogen which

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* Extended abstract; the full paper appears in *Metal Sci.* **12**, 495 (1978).

diffuses into the hydrostatic stress region ahead of the crack tip. Decohesion, proportional to the hydrogen concentration, will occur. For the impure steel this will be in addition to the decohesion already caused by phosphorus on the grain boundary. The latter is sufficient to cause i.g.f. in vacuum, and its action is enhanced in the presence of hydrogen derived from water vapour.

The complicated relation between i.g.f. and ΔK can be explained by considering the rates of formation and diffusion of hydrogen. At low growth rates, rate of hydrogen formation is low, and concentrations ahead of the tip will be low. At high growth rates, hydrogen production rates are adequate, but diffusion rates are not sufficient to allow the hydrogen to reach the hydrostatic stress region. Between the two extremes will be a peak in hydrogen concentration in the hydrostatic stress region.