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MICROSTRUCTURE, PLASTIC ZONE SIZE, AND CRACK PROPAGATION IN Ni STEELS

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It is well known that by an addition of nickel to ferritic steels the transition temperature is shifted to lower values. Not so well known is whether optimum toughness is obtained for a homogeneous solid solution or some two-phase microstructures. It is the purpose of this work to explore a range of microstructures in respect to crack propagation at room temperature and down to -196°C, in addition to fatigue crack growth velocity to find out whether there is a possibility to reduce the nickel content by producing an optimum microstructure. Secondly the experiment should test a hypothesis in which it is implied that the mechanism of crack propagation should change its qualitative nature if the plastic zone size becomes smaller or larger than the grain and crystallite size of the material [1].

The investigation was conducted with two alloys; composition, heat treatment, and microstructural features are given in Table 1. In Figure 1 the typical microstructures are shown. From the established phase diagram it is evident that the γ-phase which originates during heat treatment in the $\alpha+\gamma$-field transforms martensitically during cooling (see Figure 1a). Therefore the heterogeneous microstructures of the iron 9% Ni-alloy consists of two α-phases. a) low nickel content, defect-free (α_{5Ni}), b) higher nickel content, high defect density (α_{M15Ni}) [2]. Microhardness measurements indicated a ratio of

$$DPH_{\alpha_{5Ni}} : DPH_{\alpha_{M15Ni}} = 0{,}6 : 1$$

The fatigue experiments at room temperature provided the possibility to observe crack propagation for a variety of cyclic plastic zone sizes, because plastic zone size increases linearly with crack length: $r_p \propto a$.

In Figure 2 the crack growth per cycle is plotted as a function of the stress intensity. The range of stress intensities at which the cyclic plastic zone size is equal to grain size is indicated in this diagram. It is evident that different crack velocities for different microstructures are observed in the range of small plastic zone sizes $r_p \leq d$. Figure 2 shows clearly a much higher crack velocity for all heterogeneous microstructures irrespective of grain size and shape (duplex or Widmanstätten-structure). The 9% Ni-alloy has its lowest crack velocity in the homogeneous martensitic condition, but even a 5% Ni-alloy in the same condition has a lower crack growth rate than the 9% Ni heterogeneous microstructures ($\alpha_{5Ni} + \alpha_{M15Ni}$) in the range of very low stress intensities. A microscopic investigation of the crack path showed that it followed exclusively the α_{5Ni}-phase, while α_{M15Ni} is always by-passed, (Figure 5b).

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For high amplitude fatigue ($r_p >> d$) in Figure 2 and under the conditions of application of static load in fracture toughness tests (Figure 3) the extent of the plastic zone prevents that the crack propagation is exclusively controlled by the α-phase. Figure 3 shows that at room temperature the heterogeneous microstructure and the homogeneous martensitic material have a similar fracture toughness. Corresponding values have been measured for the homogeneous martensitic 5% Ni-alloy at room temperature. But the fracture toughness of this alloy drops down to about a half of the room temperature value if it is tested at -196°C, while the fracture toughness of the 9% Ni-alloy in the homogeneous as well as in the heterogeneous condition does not decrease with decreasing temperature. This is surprising because approximately 50 per cent of its volume consists of a phase with 5% nickel, which should behave at -196°C similar to the homogeneous alloy of the same composition. (Impact tests showed that an Fe 5% Ni-alloy behaves brittle at -196°C in the recrystallized as well as in the martensitic condition). Scanning electronmicrographs taken from the fracture surfaces of the homogeneous martensitic 5% Ni-alloy and the heterogeneous duplex structure ($\alpha_{5Ni} + \alpha_{M15Ni}$) show that the former almost exclusively fails by <100> cleavage, while this type of fracture is rare in the heterogeneous alloy, in spite of the fact that approximately 50% of its area should consist of the brittle phase, containing 5% Ni (Figure 4a, 4b). A corresponding experiment with the heterogeneous Fe 5% Ni-alloy confirmed that toughness is improved by a mixture of a ductile and a brittle phase, as compared to the homogeneous martensitic condition (see Figure 3).

The results can be generalized in the following way. At a given nickel content the heterogeneous microstructure is much more sensitive to fatigue crack propagation at low amplitudes than the homogeneous martensitic material. For fracture toughness testing, i.e., for $r_p >> d$, there is no effect of microstructure at room temperature, because both phases are ductile. There is a tendency that at low temperatures materials with the heterogeneous microstructure have better fracture toughness than the homogeneous ones. A phase which fails by cleavage in the single phase alloy still behaves ductile in combination with a second ductile phase of the two phase structures. Presently we are investigating the mechanical properties of alloys with lower nickel contents, especially for those conditions where the microstructure consists of a ductile and a brittle phase.

An understanding of these phenomena is possible if the correlation of plastic zone size and crystallite size is considered. For $r_p \leq d$ the crack will exclusively pass through the softer phase, if the difference in strength between the harder and softer one is larger than the work hardening produced by the slip processes during crack propagation. The crack velocity da/dN will be determined by the partial velocity in the softer phase ($da/dN \mid \alpha_{5Ni} >> da/dN \mid \alpha_{M15Ni}$) and a geometrical factor C. This factor comes in because the crack is forced to deviate from its optimum direction in order to by-pass α_{M15Ni} particles (see Figure 5b)

$$r_p \leq d \; : \; \frac{da}{dN}\bigg|_{\alpha_{5Ni} + \alpha_{M15Ni}} = \frac{da}{dN}\bigg|_{\alpha_{5Ni}} \; B \cdot C \qquad (1)$$

B considers the volume portion of the harder phase and would be one for an alloy which consists completely of the soft phase. It will decrease with increasing volume portion of α_{M15Ni}. C is depending on the morphology of the hard phase, and becomes $C \sim \sin \phi$ (see Figure 5b). This mechanism can be only valid for volume portions and microstructures for which the crack

can pass exclusively through the α_{5Ni}-phase. If this phase were dispersed in an α_{M15Ni}-matrix the much lower value $da/dN|\alpha_{M15Ni}$ should control crack propagation.

For the condition $r_p >> d$ no selective crack growth is possible if compatibility of the material is preserved (Figure 5a). Therefore as the first approximation, assuming a straight crack front, the property of an alloy can be obtained from the partial toughness of crack growth values of the microstructural components and the volume portion.

$$\frac{da}{dN}\bigg|_{\alpha+\alpha_M} = \frac{da}{dN}\bigg|_{\alpha} f_\alpha + \frac{da}{dN}\bigg|_{\alpha M} \cdot f_{\alpha_M} \qquad (2)$$

$$G_{\alpha+\alpha_M} = G_\alpha \cdot f_\alpha + G_{\alpha_M} \cdot f_{\alpha_M} \qquad (3)$$

The low temperature fracture toughness tests have shown that these partial values are not necessarily identical with the values of the homogeneous alloy (Figures 3 and 4), because a brittle phase can behave ductile if it is surrounded by another ductile phase (Figure 5c). However, equation 2 and 3 may approximate the behaviour of the alloys reasonably well if the mechanical properties of the two components are not too different. Further experiments to test this model are in progress.

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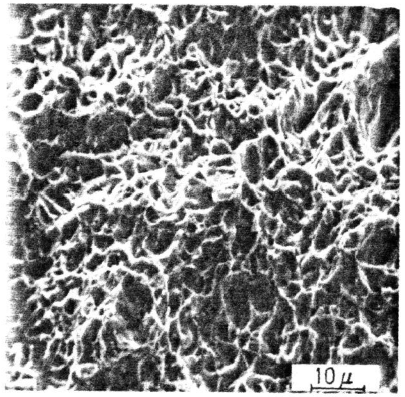
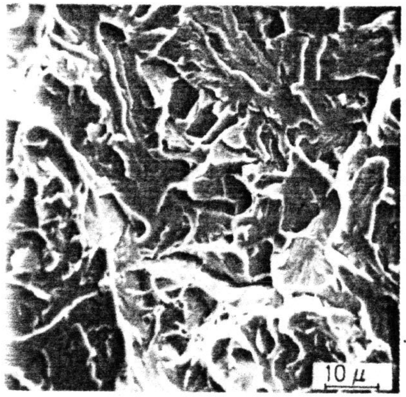
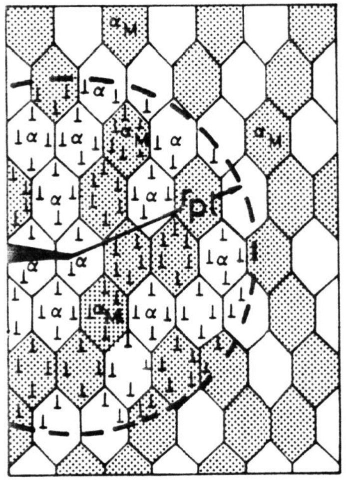


a) Heterogeneous microstructure of the 9% Ni alloy (No. 2 in Table 1)

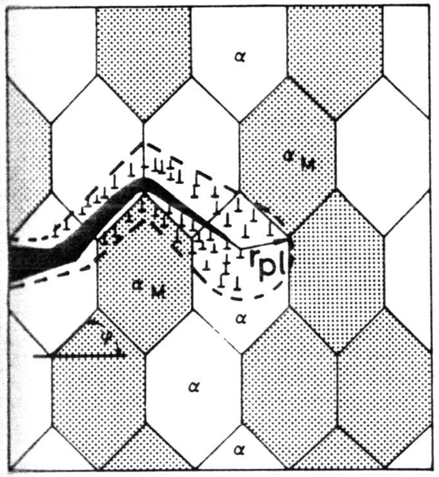


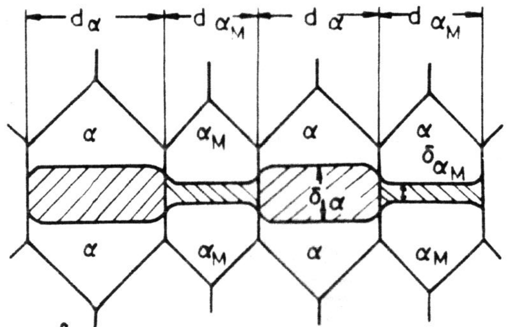
b) Homogeneous microstructure of the 5% Ni alloy (No. 4 in Table 1)

Figure 4 Fracture surface (SEM) after K_c-testing at -196°C



a) $r_p >> d$: Both phases are deformed in the plastic zone which precedes the crack tip





$$\frac{\sum_{i=1}^{n} d_{i\alpha}}{\sum_{i=1}^{n} d_{i\alpha_M}} = \frac{f_\alpha}{f_{\alpha_M}}$$

b) $r_p \leq d$: The crack path follows the softer zone unless it is deviated to an angle $\phi > \phi_{crit}$.

c) Model of a crack tip which passes through a duplex structure showing two different partial crack opening displacements.

Figure 5 Schematic representation of the crack path