EFFECTS OF MICROSTRUCTURE ON FATIGUE SHORT CRACK BEHAVIOUR IN MULTIPHASE MATERIALS

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ABSTRACT

The fatigue resistance of metallic materials is influenced by microstructural parameters which govern short crack nucleation and propagation. The microstructure plays a decisive role through the presence of interfaces which decelerate or arrest propagating short cracks and act as “microstructural barriers”.
In the present work, the nature of barriers and the behaviour of short cracks in three classes of multiphase alloys: duplex steels, Al-Si alloys and metastable austenitic steels, are investigated.
The physical nature of barriers is more complex than observed in single phase materials. The properties of the grain (particle) adjacent to the element of microstructure containing a short crack are considered more important than the properties of interfaces themselves. Slight modifications of the degree of plastic deformation in grains of different constitutive phases can thus lead to a strong reduction of the fatigue life as shown on the example of α/γ duplex steels. Similar effects can occur as a consequence of small changes of microstructural parameters as illustrated by the cyclic behaviour of binary Al-Si alloys. In a metastable austenitic steel, a particular case of indirect barriers has been observed. In such alloys, short cracks nucleate and grow exclusively in transformed martensitic regions continuously created ahead of the crack tip. The austenite grain boundaries deviate the transformation path and arrest temporarily crack propagation. This explains the increasing fatigue resistance with decreasing initial grain size.

INTRODUCTION

Cyclically deformed metals and alloys undergo complex modifications of the dislocation structures which lead to the localisation of plastic deformation and consequently to the nucleation of short cracks. The damage stage between the nucleation of first short cracks and the beginning of fatal crack propagation in the bulk, according to Paris’ law, extends typically over more than 80% of the fatigue life N_F, both in high cycle (HCF) and in low cycle fatigue (LCF) [1,2]. In low cycle fatigue, multiple cracking often takes place due to the high density of possible crack nucleation sites and to the presence of interfaces (grain, twin or phase boundaries) which act as “microstructural barriers” for crack propagation [1-3].

The influence of interfaces on fatigue resistance has been principally investigated in single phase polycrystalline alloys in which the possible types of nucleation sites and of barriers are very limited. In this elementary case, short cracks typically nucleate at the intersections of slip bands with the sample surface and their growth at surface is limited by grain or/and twin boundaries [1,4]. Recently, it has been shown that grain and twin boundaries also act as microstructural barriers with respect to the short crack propagation in the bulk [4]. For these reasons, it is understandable that interfaces are generally considered as microstructural barriers. On the basis of this approach, some essential features of fatigue damage, in particular in low cycle regime, could be explained and a numerical simulation of LCF multicracking has been proposed [1,5]. However, the physical mechanisms leading to the short crack propagation across microstructural barriers are still subject to discussions. Existing models consider the misorientation effect between adjacent grains as the essential one, without taking into account the evolution of the state of plastic deformation in the neighbouring grain during cyclic straining. In this approach, plastic zones ahead of the crack tip are arrested at boundaries and they induce stresses in the vicinity of the barrier in the adjacent grain. If the corresponding stress value (dependent, among many factors, on the crack length and on the value of applied stress) exceeds the local yield stress, dislocation sources are activated leading to the localisation of plastic deformation and making possible short crack propagation across the barrier [6].

The above mentioned model [6] predicts either a crack deceleration, or a total blocking in front of a barrier, but not a temporary crack arresting as currently observed in fatigue. Moreover, it does not allow to explain the fatigue behaviour of more complex microstructures in which the nature of short crack nucleation sites and consequently that of microstructural barriers depend not only on morphological and topological parameters which characterise the microstructure but also on the respective responses of each phase with respect to the cyclic straining. Finally, the propagation criterion itself (slip activation in the neighbouring grain) appears to be insufficient, in particular in low cycle fatigue, for which most grains are plastified, but short cracks are still arrested at interfaces, often during thousands of cycles [1-3].

Recently, Stolarz and Foct have suggested [7] that the effectiveness of an interface as microstructural barrier depends mainly on the degree of plasticity developed in the grain (particle) adjacent to the one containing a short crack. Indeed, the stresses induced by the interaction of the plastic zone with interfaces must act over a period (number of cycles) long enough to produce pronounced slip bands along which the crack can propagate. If the neighbouring grain is already plastified, the required number of cycles can be supposed lower than in the case where no dislocation slip has taken place before. Thus, the resistance of the same interface with respect to the propagation of a short crack can be radically different, even for the identical value of the applied stress/strain as schematised in Figure 1.

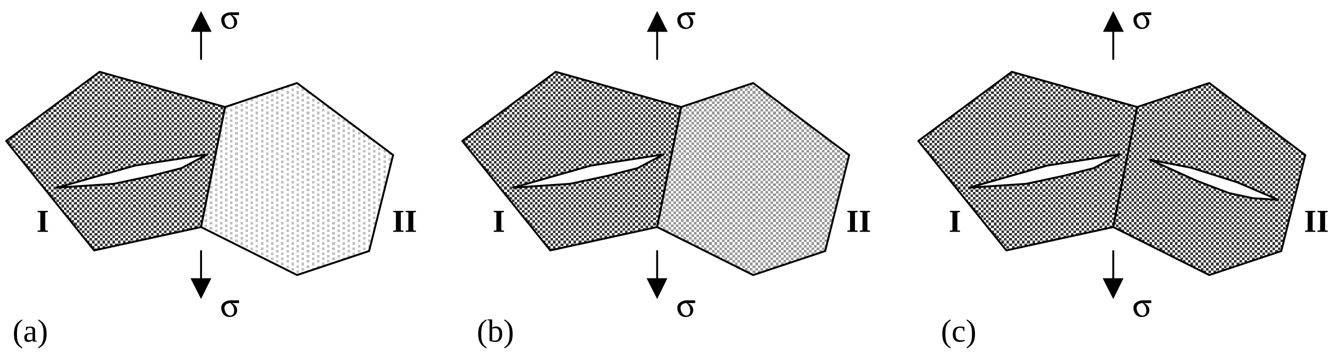

Figure 1: Decreasing resistance of the same grain boundary under same applied stress σ :
(a) grain II non deformed – high resistance; (b) grain II slightly deformed – medium resistance;
(c) grain II strongly deformed and cracked – low resistance (crack coalescence)

In multiphase materials, the complexity of the problem increases since the nature of crack nucleation sites can depend not only on the microstructure but also on the amplitude of applied stress or strain as it has been demonstrated in the case of duplex austenitic-ferritic stainless steels [8,9]. In addition to the factors which play a role in single phase alloys, the effectiveness of an interface will depend also on the physical properties of the adjacent grain (particle). The resistance of a barrier can be supposed to be radically different depending on the short crack nucleation site and it can evolve rapidly, even for little changes of operating parameters [4,8,9].

The present work deals with three different aspects of fatigue short crack growth in multiphase alloys:
- the influence of microstructural parameters on the cracking mode (single or multiple cracking) and on the fatigue life (binary Al-Si alloys);
- the influence of the nature of crack nucleation sites and of the degree of plasticity developed in the grains of second phase on fatigue resistance (binary Al-Si alloys and duplex α/γ stainless steels);

- the behaviour of short cracks related to the dynamic microstructure evolution during cyclic straining (metastable austenitic stainless steels).

INFLUENCE OF SOME MICROSTRUCTURAL PARAMETERS ON THE RESISTANCE OF BARRIERS

The microstructure of binary non-modified Al-Si alloys with composition close to the eutectic one is composed of an acicular Al-Si eutectic and of primary Al or Si crystals [10]. The fatigue damage in such alloys (symmetrical push-pull tests under plastic strain control at 20°C in air) has been found to start through decohesion of interfaces between elongated brittle Si and a very ductile Al-base phase [11]. The “short cracks” created in this way have the surface lengths equal to the linear dimension of Si particles at surface. They are arrested in front of layers of the ductile phase before they can propagate towards neighbouring Si particles, cracked or not, see Figure 2.

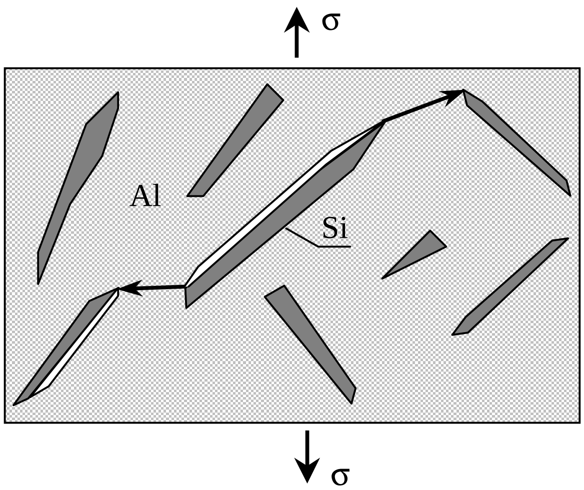

Figure 2: Initial stages of LCF cracking in a non-modified Al-Si eutectic

Two Al-Si alloys, AS12 (containing 25% of primary Al) and AS13 (purely eutectic), exhibit radically different behaviours with respect to the cyclic straining at the same applied plastic strain amplitude, even if the initial stages of damage are the same as schematised in Figure 2. While in AS12 an extended multiple cracking is observed, a single crack develops in AS13 [5,11]. The multiple cracking has in this case a beneficial influence on the fatigue life which is roughly ten times longer in AS12 than in AS13, at same $\Delta\varepsilon_p/2$, see Figure 3. Similar effect has also been observed in other binary Al-Si alloys [5].

Taking into account the above presented damage mechanism, it is obvious that interfaces cannot be considered as microstructural barriers for crack propagation in Al-Si alloys. The role of barriers is played by layers of the Al-rich phase (Figure 2) which is practically non deformed in AS12 and in AS13. The physical properties of barriers are thus the same in both alloys (the case of AS12S will be discussed in the following section). Their resistance to crack propagation is therefore proportional, for the same short crack length, to the distance λ between neighbouring Si particles. On the other hand, the stress concentrations at crack tips are proportional to their surface lengths. Since decohesions at Si/Al interfaces are expected to occur first along the longest particles l_{max}, the resistance of Al layers with respect to the crack propagation can be supposed to be proportional to the microstructural parameter $\mathbf{l_{max}/\lambda}$. Indeed, in two investigated alloys, AS12 and AS13, the respective l_{max}/λ values are very different one from another: 3.8 and 9.0. This suggests a better resistance of microstructural barriers to short crack propagation in AS12 than in AS13 even if the propagation mechanism across layers of non deformed Al-rich phase remains unchanged in both cases. Moreover, it appears that the energy necessary to cross the first barrier in AS13 is lower from that required to nucleate new short cracks because no multiple cracking takes place. The situation is inversed in AS12 in which, due to a better barrier resistance, numerous short cracks are nucleated before any propagation can occur. Thus, the transition from single to multiple cracking results in a strong improvement of the fatigue resistance of binary Al-Si alloys, at applied plastic strain amplitudes.

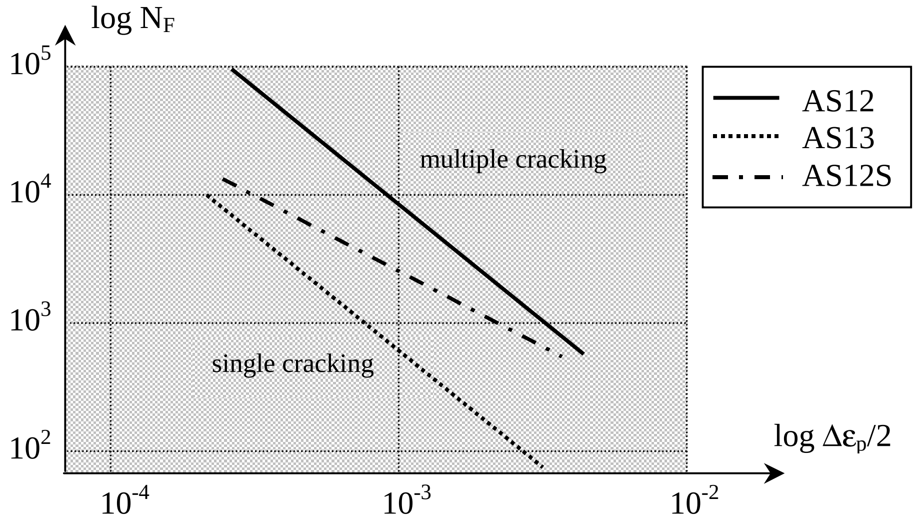

Figure 3: Coffin-Manson plot for three binary Al-Si alloys with a non-modified eutectic

MICROSTRUCTURAL BARRIER STRENGTH RELATED TO THE PLASTIC DEFORMATION

The effect of heat treatment on fatigue resistance of the binary AS12 alloy

As shown in the previous section, the AS12 alloy undergoes a fatigue damage through multiple cracking, in low cycle fatigue at applied plastic strain amplitude. The heat treatment at 560°C during 20 hours (AS12S) results in some roundening of sharp edges of Si particles in the eutectic without any significant modification of the l_{max}/λ parameter value. Surprisingly, the fatigue resistance of AS12S decreases compared with AS12, at same applied plastic strain amplitude (Figure 3), and the damage mode becomes single cracking. Since the short crack nucleation mechanism remains unchanged compared with AS12 [5], the morphological factors alone cannot explain the radical modification of the cyclic behaviour of the material. In fact, the essential difference between the fatigue behaviours of both analysed microstructures consist in the cyclic response of the Al-rich phase in the eutectic. While the maximum tensile stress remains constant with respect to the number of cycles in AS12, it increases continuously all over the test in AS12S, its maximal value reaching nearly the double of the initial one (Figure 4).

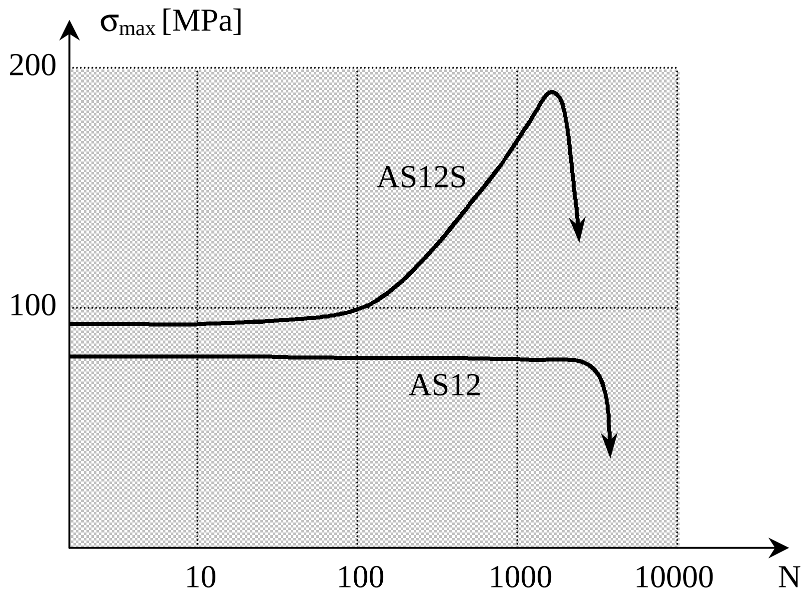

Figure 4: Cyclic hardening in AS12 and in AS12S at $\Delta\varepsilon_p/2 = \pm 10^{-3}$

This result indicates that considerable structural modifications of the Al-rich phase take place as it has been confirmed by TEM observations [5]. Consequently, the decrease of the resistance of barriers for the identical value of the l_{max}/λ parameter occurs. In fact, a barrier composed of a layer of a non-deformed Al phase (AS12) is more difficult to be crossed by a short crack than in a similar layer in which the dislocation density is high (AS12S). The validity of the generalised approach to the problem of microstructural barriers taking into account the degree of plastic deformation behind the interface which arrests crack propagation, is thus enhanced.

The influence of ageing on low cycle fatigue damage of austenitic-ferritic duplex steels

The ageing treatment of duplex austenitic-ferritic stainless steels (at 475°C during 200 h) is applied in order to increase the mechanical resistance of the alloy due to the spinodal decomposition in the ferrite. However, in fatigue, the microstructural modification induced by ageing leads to significant changes in short crack nucleation mechanisms [4,12].

In a non-aged commercial SAF 2205 duplex steel, tested at high levels of applied plastic strain amplitude ($\Delta\varepsilon_p/2 = \pm 6 \cdot 10^{-3}$), plastic deformation takes place in grains of both phases α and γ. Short cracks can thus nucleate on slip bands in α or in γ grains and they are stopped at α/γ interfaces (Figure 5). Due to the intense plastic deformation, the density of crack nucleation sites is very high and a generalised multiple cracking at surface is observed (Figure 6).

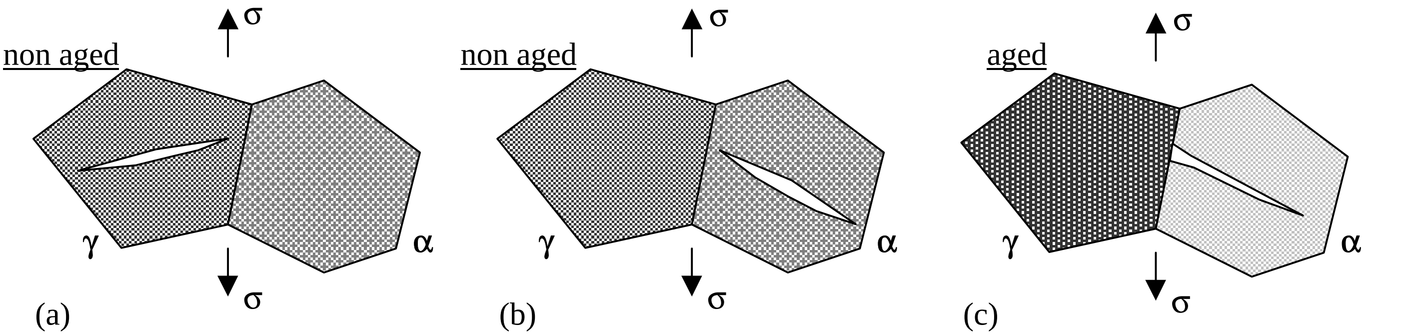

Figure 5: Short crack nucleation in a α/γ duplex steel – the influence of ageing : (a) non aged : plasticity in α and in γ, crack nucleation in γ; (b) non aged : plasticity in α and in γ, crack nucleation in α; (c) aged : plasticity in γ, crack nucleation close to γ/α interface, propagation in α

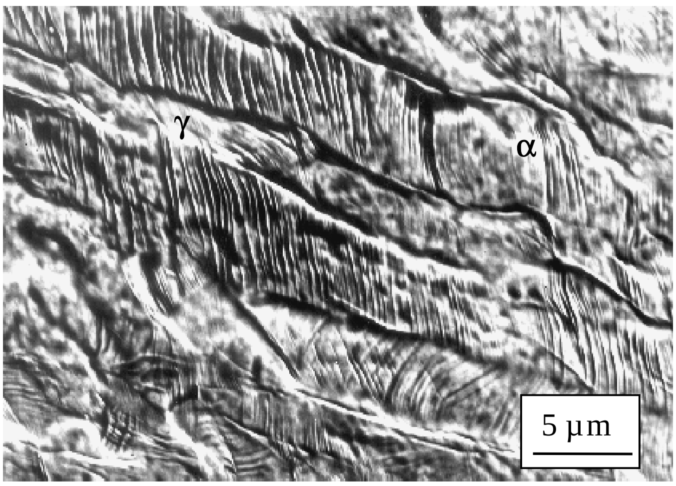

Figure 6: Surface plasticity and short cracks in the non-aged SAF 2205 austenitic-ferritic duplex steel (air, 20°C, $\Delta\varepsilon_p/2 = \pm 6 \cdot 10^{-3}$, N_F = 600 cycles) [12]

The ageing induces a strong embrittlement of the α phase leading, at $\Delta\varepsilon_p/2 = \pm 6 \cdot 10^{-3}$, to the development of plasticity exclusively in grains of austenite. Intense slip bands in γ interact with γ/α interfaces and induce stresses in brittle α grains, close to boundaries. This leads to short crack nucleation through microcleavage in the vicinity of boundaries in α (Figures 5 and 7). Such short crack extend rapidly between two neighbouring

γ/α interfaces and their further propagation must take place across very strongly deformed grains of austenite.

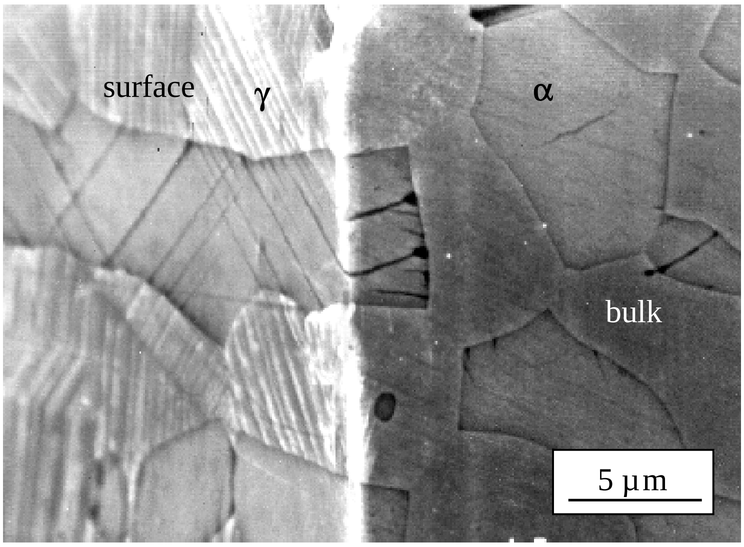

Figure 7: Plastic deformation in austenite and short crack nucleation in ferrite in the aged SAF 2205 duplex steel (air, 20°C, $\Delta\varepsilon_p/2 = \pm 6 \cdot 10^{-3}$, $N_F = 200$ cycles) [12]

For this reason and due to the fact that stresses at the crack tip in the aged steel are higher, at same $\Delta\varepsilon_p/2$, the interface boundaries resist much less to crack propagation than in the previous case. Once the first barrier crossed, further barriers (alternatively very brittle α grains or heavily deformed γ ones) do not oppose any significant resistance and the development of the fatal crack is much faster than in the non-aged material. This explains a marked evolution of the fatigue life, at $\Delta\varepsilon_p/2 = \pm 6 \cdot 10^{-3}$, which decreases from 600 cycles to 200 cycles after ageing.

INDIRECT MICROSTRUCTURAL BARRIERS IN A METASTABLE AUSTENITIC STAINLESS STEEL

Austenitic steels cyclically deformed in the temperature regime between M_s and M_d can undergo a strain-induced martensitic transformation which results in a significant cyclic hardening of the material (at applied strain amplitude) [13]. It is generally admitted that the thresholds of the plastic strain amplitude and of the cumulative plastic strain must be reached in order to set off the $\gamma \rightarrow \alpha'$ transformation under cyclic loading [13,14].

The effect of the strain-induced martensitic transformation has been studied in a high purity austenitic steel containing 16.8 wt % Cr, 13.5 wt % Ni and < 0.0010 wt % C. Two microstructures with different grain sizes, D1 = 12.5 µm and D2 = 40.5 µm, have been obtained through annealing at 900°C/1h and at 1200°C/2h. Low cycle fatigue tests have been carried out under plastic strain control ($\Delta\varepsilon_p/2 = \pm 4 \cdot 10^{-3}$) and at a constant strain rate of 10^{-3} s^{-1} in laboratory air at 20°C.

The degree of the martensitic transformation and the respective fatigue lives are found to depend strongly on the grain size of the initial non transformed austenitic microstructure. At same plastic strain amplitude and same cumulative plastic strain, the fine D1 microstructure is characterised by martensite contents (up to 20 vol. %) five to ten times higher than in the coarse D2 one (maximum 2.5 vol. %). The fatigue life of D1 (N_F = 7500 cycles) is about 50% longer than that observed in D2, even if both yield stress and plateau stress values are roughly 50 MPa higher in D1, at same plastic strain amplitude. It has been shown [15] that the strain-induced transformation radically modifies the plastic cyclic behaviour of the austenite from the very beginning of loading by retarding short crack nucleation which takes place exclusively in the martensitic islands at surface. The crack propagation is always preceded by the $\gamma \rightarrow \alpha'$ transformation assisted by strains concentrated in the plastic zone around the crack tip (Figure 8).

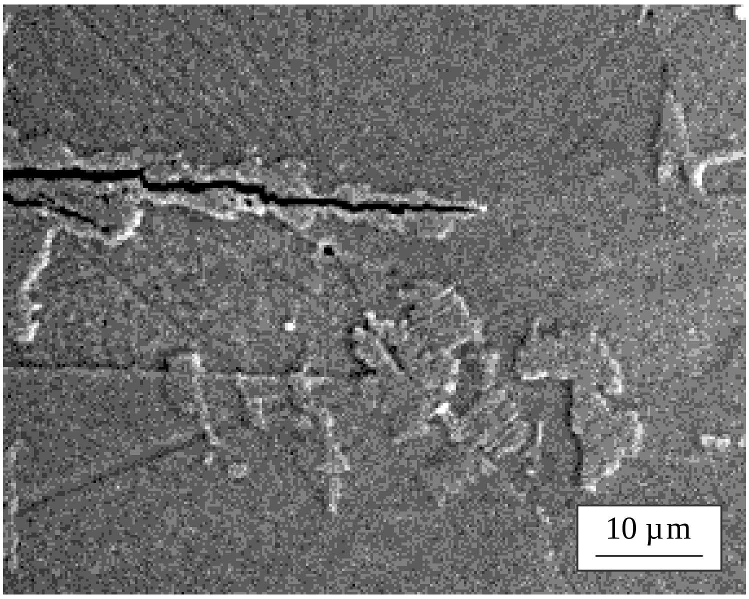

Figure 8: Short crack growth in martensite in a metastable austenitic steel at $\Delta\varepsilon_p/2 = \pm 4 \cdot 10^{-3}$ (coarse D2 microstructure, $N = N_F$, vertical sample section)

Some elements of explanation of the increasing fatigue resistance with decreasing grain size can be given on the basis of an original concept of microstructural barriers in alloys with dynamic microstructure evolution during cyclic straining. It has been pointed out that short crack nucleation and propagation take place exclusively in the martensite. Moreover, short cracks do not change their orientation, which remains perpendicular with respect to the stress axis, when crossing γ/γ grain boundaries [15]. Nevertheless, austenite grain boundaries play indirectly a decisive role in the damage process. As shown in Figure 9, when a short crack, with a transformed zone around it, reaches the proximity of a grain boundary, its propagation is temporarily arrested because the transformation occurs first along the boundary. Since short cracks in the martensite have a tendency to grow perpendicularly with respect to the stress axis, propagation can only occur when the transformation of the intergranular zone has been completed and when further transformation in the neighbouring austenite grain can take place. The number of such obstacles being more important in a fine microstructure, it is understandable that crack propagation is retarded compared with a coarse microstructure. On the other hand, the crack which meets the first grain boundary is shorter in the fine microstructure inducing less stress concentration at the crack tip and making local conditions for martensitic transformation less favourable.

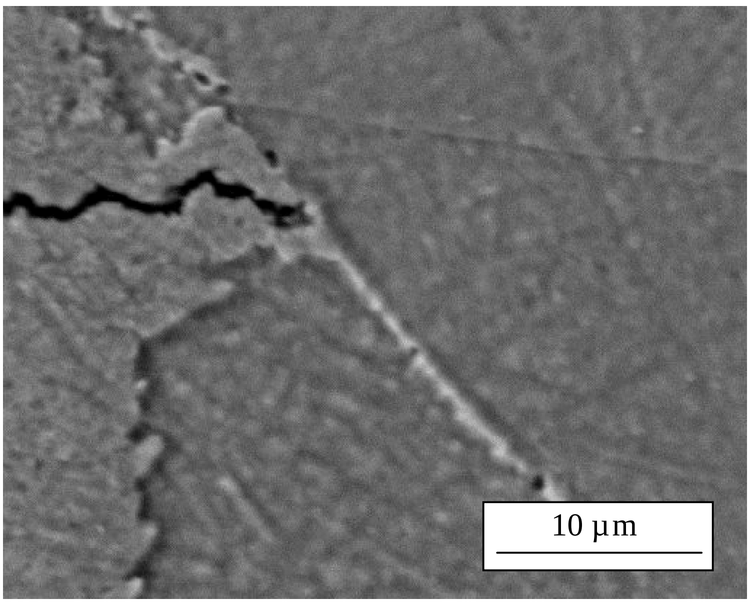

Figure 9: Short crack arresting close to a grain boundary and $\gamma \rightarrow \alpha'$ transformation along the boundary (D2, $N = N_F$, vertical sample section)

CONCLUSION

The results of fatigue short crack investigation in three different types of two phase alloys provide an interesting contribution to the understanding of the role of microstructural barriers in multiphase materials. According to the generalised concept of barriers which takes into account not only the properties of interfaces (grain, twin, phase boundaries) but mainly the physical properties and the degree of plastic deformation in the adjacent grain (particle), the fatigue behaviour of investigated alloys could be qualitatively explained.

In families of two phase alloys in which the properties of constitutive phases are the same (example: binary Al-Si alloys), a slight change of the value of some microstructural parameters can induce radical changes of the cyclic damage mode (single or multiple cracking) with a direct incidence on the fatigue life. The quantitative microstructure analysis cannot be limited to average parameters since the extreme values of some parameters play a decisive role in the fatigue damage (l_{max}/λ parameter).

In alloys in which the morphological parameters do not change but where properties of one or both phases are modified by a heat treatment (Al-Si alloys, duplex stainless steels), the resistance of microstructural barriers and consequently the fatigue life decreases when the degree of plasticity developed in the grain adjacent to the cracked one, increases. Moreover, short crack nucleation sites can be modified when the properties of phases change (aged duplex steels). Consequently, the nature of barriers changes and the cyclic behaviour of the material becomes radically different. The conventional concept of microstructural barriers in which only he properties of interfaces are taken into account appears therefore as insufficient to explain the cyclic behaviour of two phase alloys.

Finally, the existence of a specific kind of microstructural barriers in metastable austenitic steels with dynamic microstructure evolution during cyclic straining, has been demonstrated. In such alloys, short cracks nucleate and propagate exclusively in the transformed martensitic regions and their orientation do not change when crossing previous austenite grain boundaries. However, γ/γ boundaries are found to determine indirectly the damage evolution by deviating the transformation part in front of the crack tip and by retarding in this way further crack propagation. The fatigue life of such metastable alloys tested under plastic strain control can thus be increased by decreasing the grain size, even if the resulting maximum stress values are much higher than in coarse microstructures.

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