Intergranular Fatigue in Interstitial-Free Steels

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ABSTRACT. *Interstitial-free (IF) steels are known to exhibit intergranular (IG) fatigue under certain conditions. This paper explores two issues; alloy conditions under which IG fatigue occurs and whether the fatigue performance in the presence of an IG crack path is lower than similar IF steels which exhibit a transgranular crack path. To explore this latter issue fatigue performance is presented as a function of yield strength.*

INTRODUCTION

Intergranular fatigue is a relatively scarce phenomenon in ductile metals but is known to occur under certain conditions. One such condition occurs in body-centred cubic (bcc) steels, which have a strongly temperature dependent component to plastic deformation mechanisms arising from the slip behaviour of extended screw dislocations, which makes cross-slip more difficult. As noted by Daniélou et al [1] this can lead to slip asymmetry which gives rise to shape changes of bcc crystals undergoing fully reversed cyclic deformation. Stress concentrations can then form along grain boundaries which favours the nucleation of intergranular cracks as shown for α-iron by Mughrabi et al [2].

Interstitial-free (IF) steels constitute one of the major groups of steels used in the automotive industry for forming thin gauge sheet body panels. In performance terms, these alloys have to balance several conflicting requirements, i.e. deep-drawing capability, fatigue resistance, tensile strength and light weight. The conflict arises because increased formability and deep-drawing capability are assisted by very low amounts of interstitial elements such as carbon, boron and nitrogen, typically 10-200 wt ppm (0.001 wt % to 0.02 wt %). In bcc metals, however, plastic deformation behaviour is strongly dependent on temperature, strain rate and level of interstitial atoms. Work by Sommer et al [3] has shown that low carbon contents and "low" temperatures (up to ambient temperature for low levels of plastic strain range) lead to decreased mobility of screw dislocations and hence promote initiation of intergranular (IG) fatigue cracks. An additional influence may arise from the use of high strength IF steel grades with increased levels of P and Mn in solid solution. Susceptibility to brittle fracture has been shown to be higher in the presence of increased P and lower interstitial C, B or N [3].

Thus at temperatures around ambient (293K) and lower, cyclic loading of certain very low carbon and ultra-low carbon interstitial-free alloys at low levels of plastic strain range leads to intergranular crack initiation and crack growth [4]. As noted by Sommer et al [5] “brittle” crack initiation occurs only between surface grains that show evidence of plastic deformation, while in further growth along extended crack paths plastic deformation is often strongly localised within grain boundaries (in the present work sometimes evidenced by striation-like markings due to localised glide deformation and shape changes on intergranular facets, see Figure 1). One of the consequences of this mode of crack initiation and growth is that intergranular fatigue is more prevalent during high cycle fatigue, which is the design regime of interest for automotive components. The question arises as to the relative performance of alloys which show IG fatigue under high cycle fatigue conditions, versus those that do not.

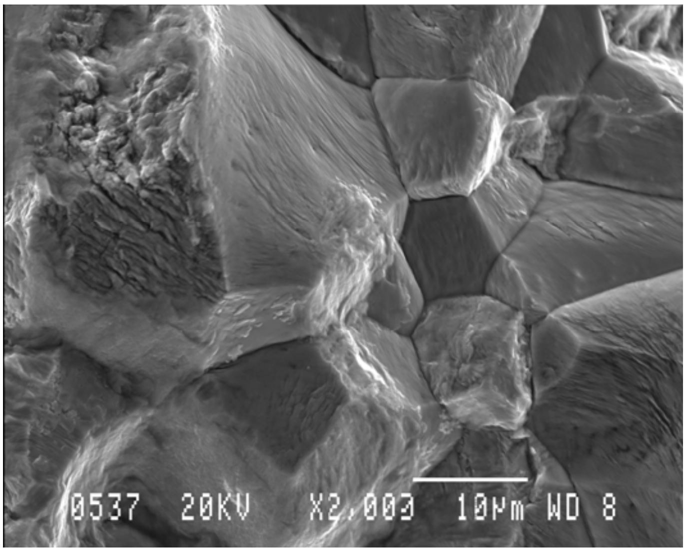

Figure 1. Evidence of localised glide deformation and shape changes on IG facets of an IF (TiNb) GA steel specimen. $N_f = 4.16 \times 10^6$ cycles.

In recent years the amount of research on IF steels has increased, primarily focussed at studies of precipitation and segregation behaviour related to cold work embrittlement (CWE) during deep drawing [6, 7]. Relatively few studies have focussed on fatigue performance [1, 8-11] and most of those have considered issues other than cracks path and the occurrence of IG cracking during fatigue. Daniélou et al [1] explicitly considered the mechanisms of cyclic plasticity at several different plastic deformation amplitudes and strain rates. In their steel initiation of fatigue cracks was observed to be both intergranular and transgranular at ambient temperature (293K) but the intergranular region was reportedly confined to cracks < 50 μm in length. Islam and Tomota [11] reported the occurrence of extensive IG faceting on the fatigue fracture surfaces but ascribed this to environmentally-induced hydrogen embrittlement and to grain boundary segregation of phosphorous. Most studies have considered a fairly confined range of IF steels and experimental conditions.

It would seem a useful addition to the literature to report the results of an extensive programme of fatigue testing on IF-steels undertaken by the author as part of a research project sponsored by the European Coal and Steel Community (ECSC) during 1998-2000. It considered the reversed bend fatigue performance over a range of lives from 10^4 to 10^7 cycles of a number of thin sheet alloys. The matrix of experimental conditions included alloys with several types of surface condition, stress concentration values of either 1.0 or 3.0 and pre-strain of either 0% or 10%. The primary intention in this paper is to explore two issues; the conditions under which IG fatigue occurs and whether the fatigue performance in the presence of an IG crack path is lower than similar IF steels which exhibit a transgranular crack path. To explore this latter issue fatigue performance is presented as a function of yield strength.

ALLOYS AND EXPERIMENTAL CONDITIONS

Composition and mechanical properties are given in Tables 1 and 2 for four grades of steel sheet used in vehicle manufacture. ELC is an extra-low carbon grade, IF (TiNb) is a standard IF steel stabilised with titanium and niobium, IF (TiNbP) is a high strength rephosphorised IF steel stabilised with titanium and niobium and IF (TiNbPB) is a high strength rephosphorised boron-containing IF steel stabilised with titanium and niobium. Alloying additions to achieve high strength levels include Mn, Si and P. Rephosphorised steel has P added to sheet steel as a work hardening agent. It increases hardness and stiffness with a corresponding loss in ductility. The level of P is kept <0.08wt% because the heating of steels with phosphorus contents between 0.080–0.160 wt% in the temperature range of 450°C–550°C (the galvanising temperature range) produces the maximum segregation of phosphorus to the grain boundaries and promotes temper and cold work embrittlement.

Table 1. Chemical composition of the alloys

Element *wt%*	ELC	IF (TiNb)	IF (TiNbP)	IF (TiNbPB)
C	0.017	0.001	0.002	0.004
Mn	0.186	0.110	0.380	0.378
Si	0.006	0.004	0.004	0.090
P	0.011	0.008	0.070	0.045
S	0.006	0.005	0.005	0.007
Nb	-	0.013	0.0018	0.028
Ti	-	0.018	0.0015	0.045
B (ppm)				10

The Ti and Nb additions stabilise the residual nitrogen and carbon and prevent interstitial formation. Table 3 gives information on the surface treatment, stress concentration factor, and pre-strain values used in the test programme, and also specifies grain size and grain aspect ratio for the cases without pre-strain.

GA denotes a galvannealed specimen, HDG one that has been hot dip galvanised and UC an uncoated specimen. Galvannealed products show improved spot-weldability and coating adhesion, and are easier to paint than galvanized products. In terms of fatigue resistance, galvanized coatings are soft and easily scratched, while galvannealed coatings are hard and relatively brittle. The surface condition would therefore be expected to influence fatigue strength. Both products are made by a hot-dip coating process. The main difference is that for a galvannealed coating the hot-dipped steel is subsequently heated in the range 538°C to 565°C by passing it through a furnace directly above the coating bath with a holding time of around 10s. Alloying occurs by diffusion between the zinc coating and the iron to give a final coating with a composition of approximately 90% zinc and 10% iron [12].

Table 2. Mechanical properties of the alloys

Material	Yield *MPa*	UTS *MPa*	Elongation *%*	$n_{10\text{-}20}$	r_{20}	Thickness *mm*
ELC HDG	209 (lower) 215 (upper)	330	37.7	0.173	2.17	1.00
IF (TiNb) UC	174	255	-	-	-	1.00
IF (TiNb) HDG	150	285	-	-	-	1.00
IF (TiNb) GA	156	280	45.8	0.227	2.57	1.00
IF (TiNbP) HDG	258	377	36.5	0.2	2.15	1.00
IF TiNbP) GA	236	367	37	0.196	1.88	0.80
IF (TiNbPB) GA	210	378	36.7	0.22	2.08	1.00

Table 3. Experimental conditions used in the test programme

Alloy	Surface Condition	Stress Conc. k_t	Pre-Strain %	Grain Size ASTM *G No.*	Grain Aspect Ratio
ELC	HDG	1 3	0 10	8.5	1.04
IF (TiNb)	GA	1 3	0 10	8.4	1.03
IF (TiNb)	HDG	3	0	-	-
IF (TiNb)	UC	3	0	-	-
IF (TiNbP)	GA	1	0	10.3	1.03
IF (TiNbP)	HDG	1	0	9.85	1.02
IF (TiNbPB)	GA	1	0	10.1	1.03

Pre-strain can be either beneficial or detrimental to fatigue strength of sheet steel alloys, depending on the resulting microstructural effects [13] and 10% pre-strain is commonly used in industrial practice. Pre-strained specimens had $k_t = 1$.

These steels are all used in deep drawing and stretch forming applications and for thin textured sheets the modulus of elasticity, tensile strength and deformation are different in different directions. The anisotropy of deformation is important in deep drawing alloys and essentially governs the resistance to thinning during deep drawing. It is represented by the r_{20} value in Table 2, which is defined as the ratio of the true strain across the width to the true strain in the thickness of a strip specimen strained by a uniaxial tensile load to 20% deformation and is termed the "vertical anisotropy". Using this definition, the r value is 1 for an isotropic alloy and a high value of r denotes a material that has a very good deep drawing capability. The strain hardening exponent $n_{10\text{-}20}$ measured over the range 10-20% deformation is also important in deep drawing with high n-values giving good formability. The values of r and n in Table 2 indicate alloys with good stretch forming capability.

High strength IF steels are used for the manufacture of complex parts requiring high formability and strength, e.g. door inner panels, wheel arches, spare wheel wells and floor panels. They contain low levels of C and N, and achieve their high strength from solid solution strengthening by P, Si and Mn, and precipitation hardening from additions of Ti, Nb and B. This is apparent in Table 1 for the two high strength rephosphorised grades and grain sizes in these alloys also tend to be small conferring strength via the Hall-Petch effect.

Service loading on thin sheet steels is primarily bending and torsion and is likely to be deflection controlled, rather than load controlled. These conditions can be simulated using deflection-controlled fully reversed bending (R = -1) on purpose-designed fatigue testing machines such as the Avery 7303. These machines control specimen displacement via a cam-operated bending arm and a coil spring. A calibration curve is obtained between grip deflection and bending moment which can be converted to applied nominal stress via the simple bending equation:

$$\frac{M}{I} = \frac{\sigma}{y} \tag{1}$$

In Eq. 1 M is the applied bending moment, I is the second moment of area of the specimen, σ is the nominal applied bending stress and y is the distance from the neutral axis to the surface of the specimen.

Specimen designs are shown in Figure 2 and have the dimensions indicated. Tests were carried out in laboratory air at ambient temperature and were stopped at crack lengths equivalent to a 5% decrease in displacement (e.g. a crack of around 3 mm deep from the notch root in the k_t = 3 specimens). Specimens were then pulled to fracture using a tensile testing machine and the fracture surfaces examined using a scanning electron microscope.

S-N DATA

Figure 3 presents S-N data plotted against stress amplitude for both values of k_t, while Figure 4 presents equivalent data as a function of the ratio of stress amplitude to yield strength. It could be argued that the cyclic yield strength should be used, but such data is available for only two of the alloys the ELC HDG (172 MPa) and the IF (TiNb) GA (119 MPa).

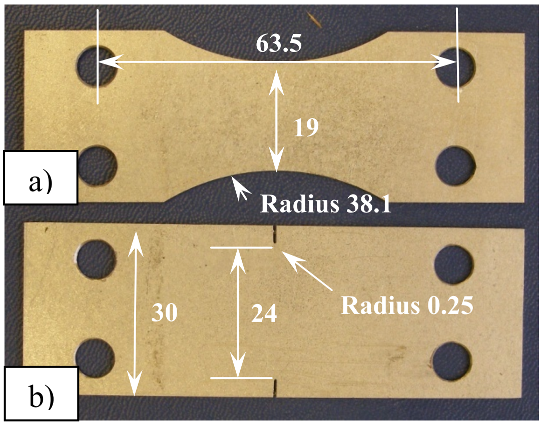

Figure 2. Specimen design for specimens with a) k_t=1.0 and b) k_t=3.0; mm dimensions.

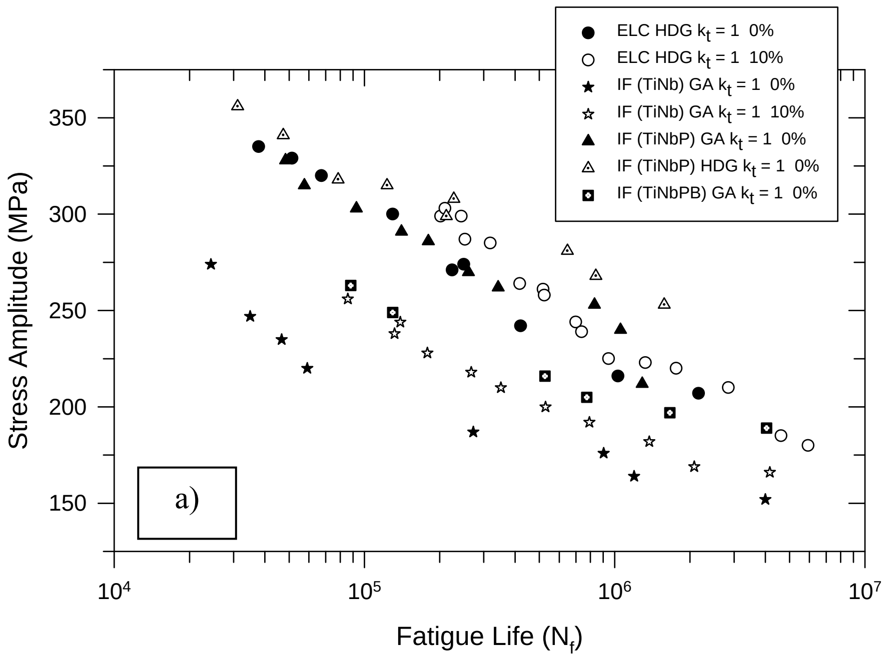

Figure 3. a) S-N data given as stress amplitude for k_t = 1.

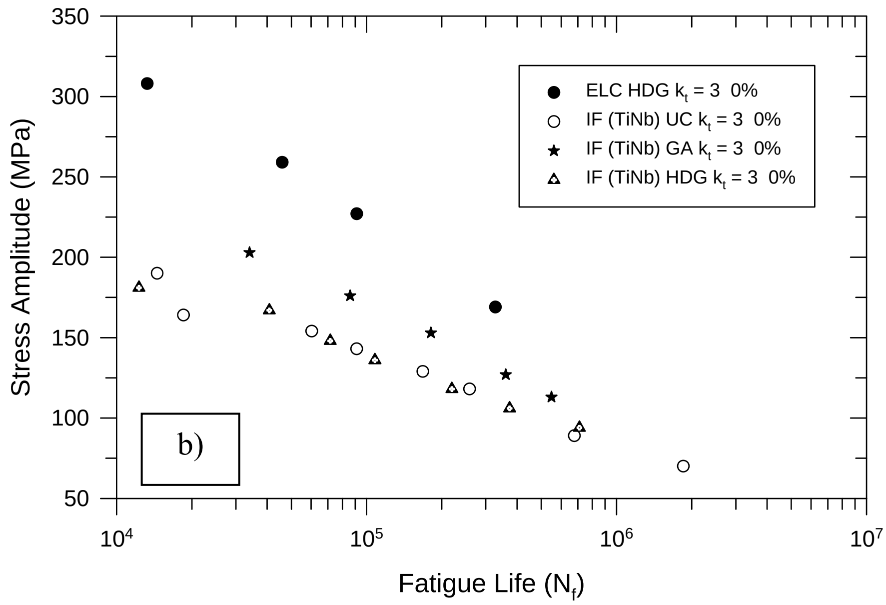

Figure 3. b) S-N data given as stress amplitude for k_t = 3.

The data in Figure 3a indicate that for k_t = 1, 10% pre-strain leads to improved fatigue performance in both the ELC and the IF (TiNb) GA alloy reflecting the improved tensile properties arising from the strain hardening. In stress amplitude terms, the IF (TiNbP) HDG alloy performs best and the IF (TiNb) GA alloy the worst. Data for the IF (TiNbPB) alloy lies between these two limits. The alloys appear to maintain their relative ranking at all lives. Figure 3b indicates that the high cycle fatigue performance (>2x10^6 cycles) of specimens with k_t = 3 is very similar, but that their low cycle fatigue performance is different. The ELC alloy performs best followed by the IF (TiNb) GA alloy. The interpretation of this is that high cycle fatigue life reflects resistance to crack initiation which is similar in these alloys and therefore probably governed by cyclic plasticity at the notch root, while short life behaviour reflects crack growth resistance and hence the yield strength (ease of plastic deformation) as the grain size is fairly constant across these alloys.

Fatigue performance is characterised in terms of the ratio of stress amplitude over yield strength in Figure 4. At lives > 2x10^6 cycles and k_t = 1 (Figure 4a) the endurance limits for all alloys except the IF (TiNbP) GA data fall into a fairly narrow band between (0.9-1.0)σ_y. The IF (TiNbP) GA alloy has an endurance limit value of around 1.07. At shorter lives (<10^5 cycles) there is little difference between the IF (TiNbPB) GA, IF (TiNbP) GA and IF (TiNbP) HDG grades, while the ELC HDG and IF (TiNb) GA grades start to perform substantially better. The IF (TiNb) GA grade appears to performs best at around 2x10^4 cycles.

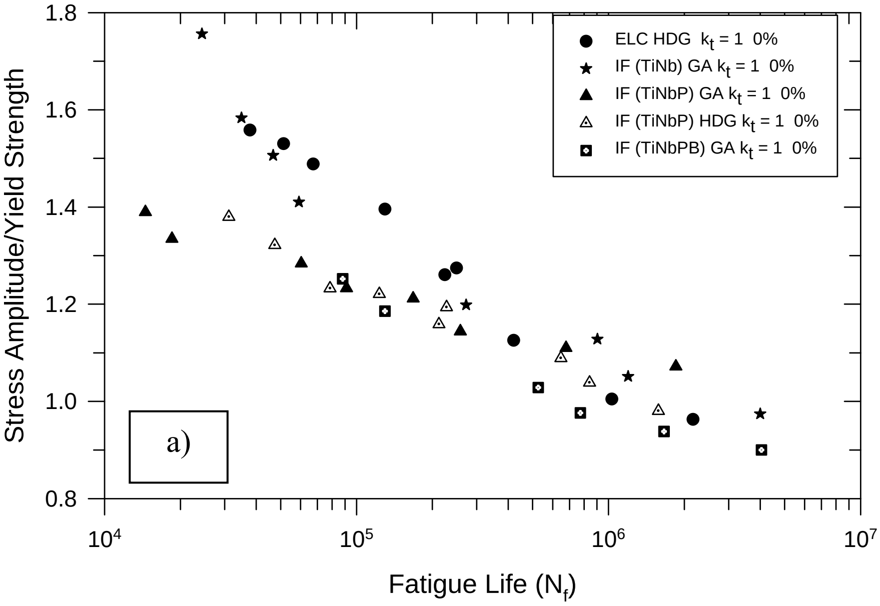

Figure 4. a) S-N data as a ratio of stress amplitude/yield strength for k_t = 1.

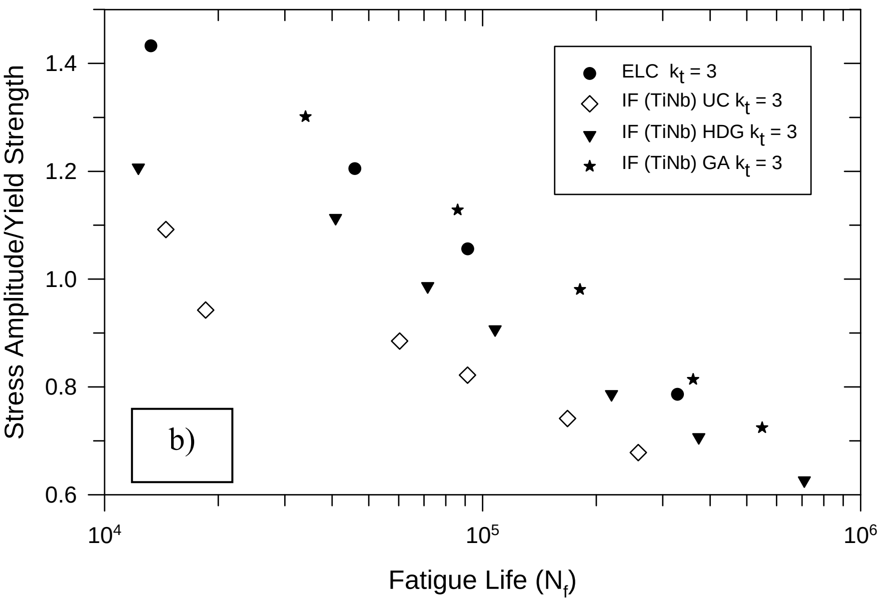

Figure 4. b) S-N data as a ratio of stress amplitude/yield strength for k_t = 3.

In the presence of a notch with k_t = 3 (Figure 4b) the IF and ELC grades perform best in terms of fatigue performance as a function of yield strength. It is observed for the IF (TiNb) alloy that both the HDG and GA conditions perform better than the uncoated condition, reflecting increased surface protection and hardness rather than yield strength (which is reduced in both these conditions compared with the uncoated case).

CRACK PATHS

Table 4 summarises the observed crack paths in all these alloys grades. It allows conclusions to be drawn regarding the mechanisms of crack initiation and growth and provides a rationale to explain and support the fatigue life data given in Figures 3 and 4. On the basis of the ELC HDG and IF (TiNb) GA data in Table 4, increasing the k_t value from 1 to 3 has no influence on crack initiation mechanism.

Table 4. Crack path observations for these steel grades

Alloy	Surface Condition	Stress Conc. k_t	Pre-Strain %	Crack Path	
				Low cycle	High cycle
ELC	HDG	1 & 3	0	Ductile	Ductile
ELC	HDG	1	10	Ductile	Infrequent IG facets near initiation and in interior
IF (TiNb)	GA	1 & 3	0	IG facets ~100-200 µm from surface, ductile interior	Extensive IG faceting
IF (TiNb)	GA	1	10	IG facets ~100-200 µm from surface, ductile interior	Extensive IG faceting
IF (TiNb)	HDG	3	0	Infrequent IG facets ~10-30 µm from surface	Mixed IG/ductile
IF (TiNb)	UC	3	0	Infrequent IG facets ~10-30 µm from surface	Mixed IG/ductile
IF (TiNbP)	GA	1	0	Infrequent IG facets ~10-50 µm from surface, ductile interior	Infrequent IG facets ~10-50 µm from surface, ductile interior
IF (TiNbP)	HDG	1	0	Ductile	Mixed IG/ductile
IF (TiNbPB)	GA	1	0	Ductile	Ductile

Considering the IF (TiNb) alloy, IG faceting occurs at all lives in all three surface conditions tested, although to variable extents reflecting the interaction of several factors. The crack initiation mechanism discussed in references [1-3] can be identified at shorter lives in the IF (TiNb) UC and HDG grades. A typical fractograph of such "brittle" crack initiation between surface grains with IG facets confined to within 50 μm of the surface is shown in Figure 5a. Decreasing the level of plastic strain, i.e. increasing the fatigue life, in the IF (TiNb) UC and HDG grades increases the IG contribution to crack growth both in percentage terms and in extent below the surface (Figure 5b), although the IG faceting reduces at the crack length increases. Figure 6 demonstrates the difference between crack initiation at long lives in the IF (TiNb) GA and HDG conditions.

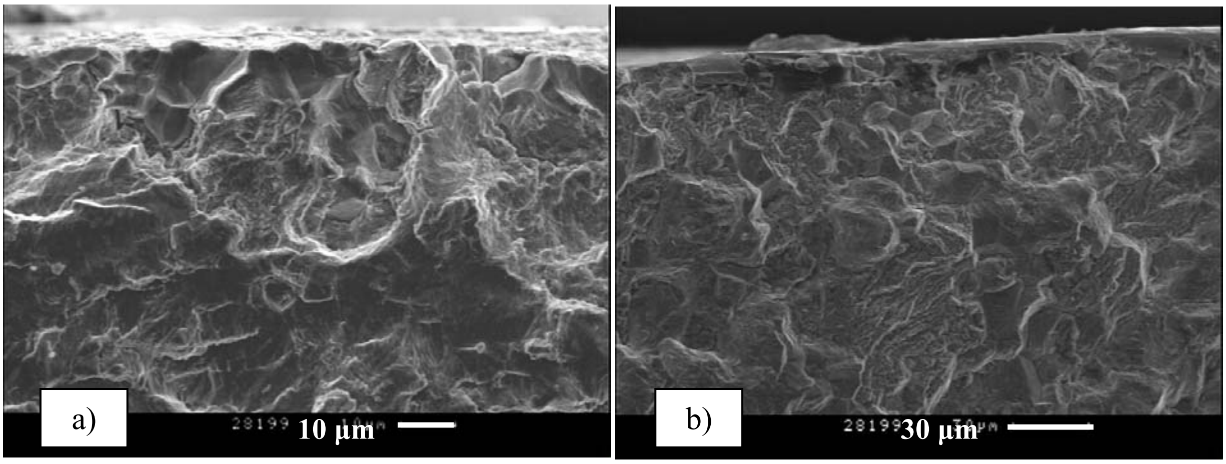

Figure 5. a) "Brittle" crack initiation at surface grains in the IF (TiNb) UC grade with some further localised growth within grain boundaries – N_f = 8,706 cycles. b) Crack path in same grade at N_f = 8,204,640 cycles.

For the IF (TiNb) GA grade with either 0% or 10% pre-strain the extent of an IG crack path is greater (100-200 μm) at short lives, reflecting the diffusion of zinc into the surface, and becomes much more extensive as the level of plastic strain decreases at longer lives. In the IF (TiNb) grade of steel applying a pre-strain of 10% does not appear to affect the crack growth mechanisms as there is already a substantial IG contribution (see Figure 7a). For the ELC grade, which has a ductile transgranular crack path with 0% pre-strain, increasing this to 10% induces infrequent IG facets near the surface and in the interior of the specimen at low levels of plastic strain (long lives) as indicated in Figure 7b.

The IF (TiNbP) grade is less susceptible to IG cracking than the (TiNb) grade (this is particular clear at shorter fatigue lives) reflecting the higher C content, but the HDG and GA conditions show reverse behaviour to the (TiNb) grade at low levels of plastic strain (long lives). It is clear that tensile properties, as well as surface condition, influence the propensity for IG faceting during fatigue in this steel. These data also imply that the P content in these IF alloys has little influence on the occurrence of IG fatigue.

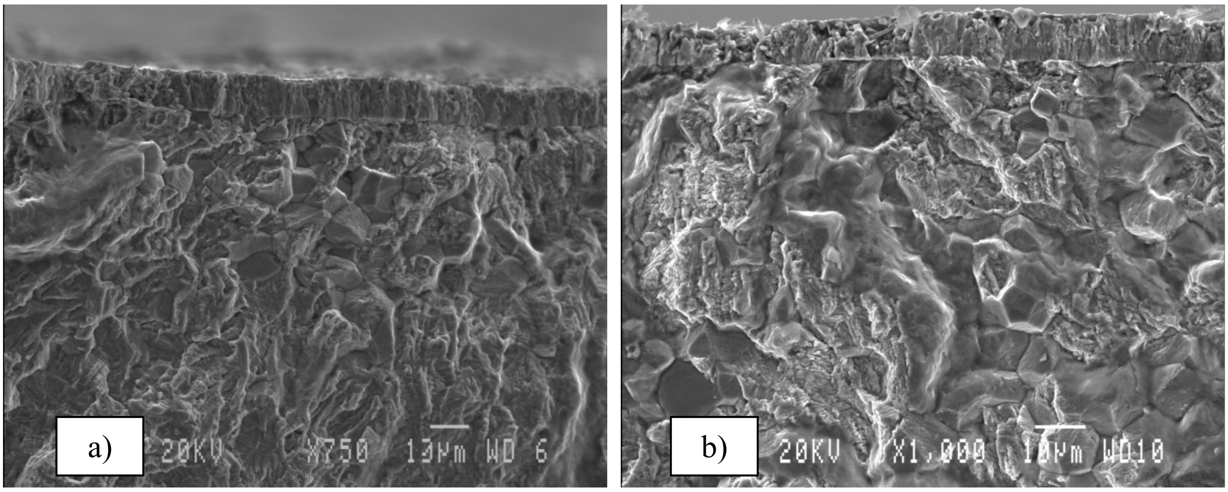

Figure 6. a) Crack initiation in the IF (TiNb) GA grade – N_f = 1,288,127 cycles. b) Crack path in the IF (TiNb) HDG grade at N_f = 1,577,158 cycles.

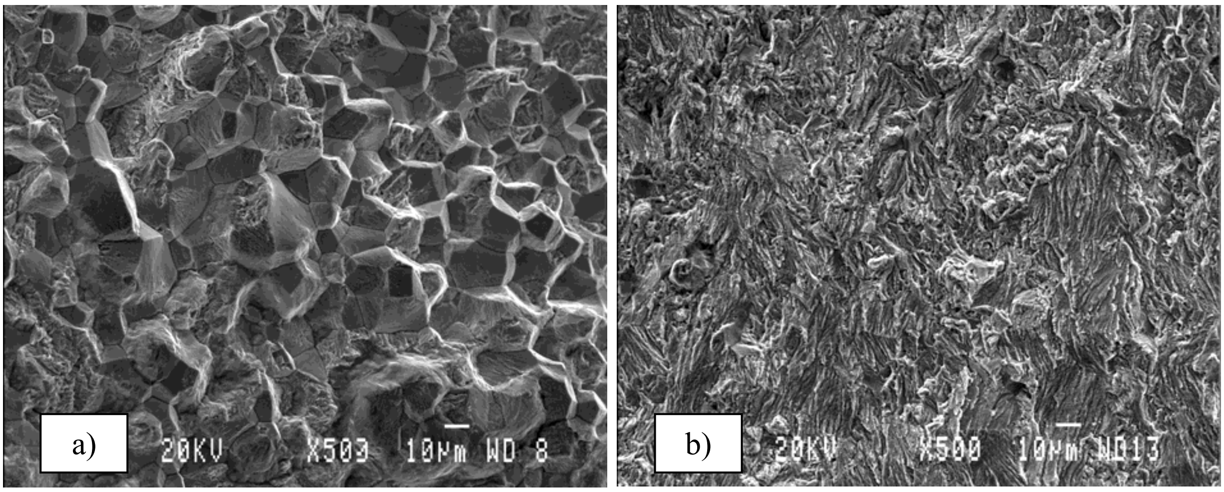

Figure 7. a) Interior of an IF (TiNb) specimen with 10% pre-strain; N_f = 4,165,923 cycles. b) Crack path in ELC specimen with 10% pre-strain; N_f = 5,928,779 cycles.

No IG cracking was observed with the IF (TiNbPB) grade but despite this it does not show improved fatigue performance as a ratio of sustainable stress amplitude compared with any other alloy tested. Even when fatigue performance as a function of stress amplitude is considered only the IF (TiNb) GA grade shows lower fatigue resistance. Similar comments apply to the ELC alloy which also exhibits ductile cracking at all lives. On a yield strength ratio basis long life fatigue resistance of the ELC alloy is comparable with the other alloys, although it performs better at short lives than any alloy except the IF (TiNb) GA.

CONCLUSIONS

IG fatigue can occur at ambient temperatures in these thin sheet automotive alloys across a fairly broad range of composition when carbon contents are below 40 ppm. This mechanism of crack growth arises from the plastic deformation behaviour and hence temperature, strain range and rate, and interstitial element content are key influences on its occurrence. Although plastic deformation is then strongly localised at grain boundaries the fatigue performance is little different to other comparable alloys which do not exhibit such IG faceting during crack initiation and growth.

10% pre-strain can induce occasional IG facets at long lives in alloys which do not otherwise show this phenomenon (ELC HDG grade) and this is probably an environmental effect which is known to occur in some steels when plastic zone sizes are small and growth rates are low. Phosphorous content appears to have little influence on IG fatigue in these alloys as the IF (TiNbPB) alloy shows no IG faceting and the IF (TiNbP) alloy exhibits behaviour consistent with interstitial control of the IG mechanism. GA surface treatments, which give rise to hard relatively brittle coatings, appear to promote IG faceting in lower strength, low carbon IF steels but have little effect on higher strength, high carbon IF alloys. Stress concentration values in the range k_t = 1-3 do not appear to influence the occurrence of IG facets in these steels.

Acknowledgements
Steel for this work was supplied by the partners in the ECSC sponsored research project “Metallurgical Aspects of the Fatigue Resistance and Cold Work Embrittlement of Thin Sheets” F4.4/98. The project report is available from the EU Research Directorate-General with Eur No. 20500, ISBN No. KI-NA-20500-DE-S. Particular thanks are due to W.J. Vortrefflich of Hoogovens Research & Development, IJmuiden.

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