# BRITTLE FRACTURE OF DUPLEX STAINLESS STEELS.

T.J. Marrow*

The fracture of hot-rolled, age-hardened super duplex stainless steel, Zeron 100, was investigated below the brittle/ductile transition. Unstable fracture may be preceded by brittle stable crack growth, particularly in thin sections, which increases the apparent fracture toughness. The most sensitive technique to monitor crack initiation depends on the crack orientation and the tendency for de-lamination in the rolling plane. A simple model predicts the crack initiation toughness for yield strength above 800 MPa.

## INTRODUCTION

Duplex stainless steels have a ferrite matrix, comprising 50% to 60% of the austenite/ferrite microstructure (1). These steels are now being selected for thick-walled structures (2, 3), for which a good understanding of the plane strain fracture toughness is necessary. The ferrite matrix is susceptible to age-hardening between approximately 250°C and 500°C. This induces "475°C embrittlement", which is most rapid around 475°C and is due to decomposition of the ferrite (4). Duplex stainless steels generally have high toughness. Low temperature decreases the fracture toughness (2, 5), which is influenced by the microstructure and crack orientation (5). The fracture toughness is also reduced by age-hardening and $\sigma$-phase precipitation (6, 7).

The brittle fracture mechanism and the criteria for crack initiation in age-hardened super duplex stainless steels have been studied (8, 9). Brittle fracture initiation is due to ferrite cleavage, which is nucleated by deformation twinning. In notched samples at temperatures below the brittle-to-ductile transition, brittle fracture has been observed to require a critical shear stress, $\tau_f$, ahead of the notch. The tensile stress at the notch is sufficient to propagate cleavage, and fracture initiation depends on the availability of cleavage nuclei in the plastic zone ahead of the notch tip. Unstable fracture may be preceded by stable propagation of a bowed crack. This paper investigates the effect of orientation on the measurement of fracture toughness in a super duplex stainless steel below the brittle-to-ductile transition temperature. A model for the effects of temperature and age-hardening on the crack initiation toughness is also briefly presented.

## EXPERIMENTAL DETAILS

Hot-rolled super duplex stainless steel plate (25 mm), Zeron 100, was supplied by Weir Materials, UK. The nominal composition was 25 wt.% Cr, 7 wt.% Ni, 4 wt.% Mo, 0.2

wt.% N. The ferrite matrix was approximately 50% of the microstructure by volume. Colonies of austenite grains with approximate dimensions of 40 μm x 200 μm were elongated in the rolling direction. The austenite thickness normal to the rolling plane was approximately 20 μm. The spacing between austenite colonies was approximately 20 μm to 30 μm, and was comparable to the ferrite sub-grain size. The as-received plate has a Vickers hardness (30 kg diamond pyramid indenter) of 270 $kgmm^{-2}$.

Fracture toughness tests (British Standard BS7448) were performed between -50°C and 215°C in a temperature controlled environmental chamber. Single edge notch bend specimens (thickness 10 mm), with cracks in the T-L and T-S orientations (fig 1) were tested in three point bending with a clip-gauge to record the crack opening displacement. The steel was aged for one week at a nominal temperature of 475°C to a Vickers' hardness of 400 $kgmm^{-2}$ (T-L orientation), and 450 $kgmm^{-2}$ (T-S orientation). This hardness difference was due to an inaccurate furnace temperature controller. Fatigue pre-cracking was performed after age-hardening. The test specimens were then aged at 475°C for two hours to heat-tint the fatigue pre-crack. This had a negligible effect on the hardness and facilitated measurement of the pre-crack length. The crack tip opening displacement (CTOD) for unstable crack propagation was determined from the critical fracture load (type c), pop-in load (type u) or maximum load (type m). Electrical crack monitoring by the direct current potential drop (dcpd) method was also used to record the beginning of stable crack propagation. The crack tip opening displacement, $CTOD_{dcpd}$, was calculated when the voltage change was equivalent to a crack extension of 100 μm.

## RESULTS

Significant pop-ins were observed for all tests in the T-S orientation, and also for the majority of tests at all temperatures in the T-L orientation. The first significant pop-in was used to calculate the CTOD. In the remaining tests, no significant pop-ins were observed and the CTOD was calculated at the maximum load. The significance of each pop-in was determined according to BS7448. In the T-L orientation, the pop-in or maximum load CTOD was always greater than or equal to $CTOD_{dcpd}$. In the T-S orientation, $CTOD_{dcpd}$ was always greater than the pop-in CTOD (figure 2). The fracture path in the T-L orientation was parallel to the plane of the fatigue pre-crack, whereas in the T-S orientation, fracture propagated out of the plane of the fatigue pre-crack by an effective de-lamination in the rolling plane. Fractography, using scanning electron microscopy, confirmed that crack propagation occurred by cleavage of the ferrite matrix.

## DISCUSSION

### Crack Initiation Toughness Measurement.

A previous study of the same duplex stainless steel, in the L-T orientation, showed that the fracture toughness, $CTOD_{dcpd}$, measured by the dcpd technique corresponded to the lowest toughness value recorded at pop-in or unstable fracture (9). Heat tinting of arrested cracks confirmed that $CTOD_{dcpd}$ corresponded to the onset of stable brittle crack growth near the centre of the specimen, from which a severely bowed crack front could develop. This was defined as the crack initiation toughness, $CTOD_i$. The CTOD at maximum load depended on the stability of the bowed crack, and was sensitive to constraint from specimen thickness below at least 16 mm. $CTOD_i$, however, was invariant for specimen thickness between 4 mm and 16 mm. The BS7448 criteria for selecting the first significant pop-in occasionally gave a toughness significantly greater than $CTOD_{dcpd}$, which implied that clip-gauge methods may not always be sufficiently sensitive to detect stable crack propagation by small pop-ins (9). The crack initiation toughness, $CTOD_i$, is important, since its measurement in small specimens may have more relevance to the fracture behaviour of thick components than the maximum load or pop-in CTOD.

The T-L data presented here is consistent with the data for the L-T orientation. $CTOD_{dcpd}$ records the onset of stable brittle crack propagation in these orientations. However, the T-S data implies that the dcpd technique can be less sensitive than the clip gauge detection of pop-in for this orientation. This can be understood by considering the effect of the textured duplex microstructure on crack propagation. In the T-S orientation, the crack propagation direction is perpendicular to the rolling plane, unlike the T-L and L-T orientations. Brittle crack propagation in duplex stainless steels is stable due to the ductile behaviour of the austenite which tends to resist crack growth. However, the rolled texture has an easy crack propagation path through the ferrite matrix parallel to the rolling plane. In the T-S orientation, pop-in and stable crack growth occur by deflection at a large angle to the fatigue crack. The decrease in specimen compliance is detected by the clip gauge at pop-in. However, the dcpd crack monitoring technique is quite insensitive to this mode of crack propagation, since the de-lamination has little effect on the flow of electrical current through the test specimen. The rolling plane is not available as a crack path in the T-L and L-T orientations, and although de-lamination in the rolling plane does occur due to tri-axial stresses in the specimen, it appears to have no significant effect on crack propagation. Previous results for the T-S orientation in material aged to a hardness of 370 $kgmm^{-2}$ found agreement between CTOD at pop-in and $CTOD_{dcpd}$ (9). Although the stable cracks deflected from the fatigue pre-crack, there was a smaller tendency for propagation in the rolling plane, which suggests that pop-in by de-lamination and the associated error in $CTOD_{dcpd}$ may be encouraged by high hardness.

The fracture toughness for brittle fracture propagation in duplex stainless steel should therefore be measured with some care. Stable crack growth may precede unstable fracture. The condition for unstable brittle crack propagation depends on the behaviour of bowed cracks, and this is not considered in this paper. The best technique for measurement of the crack initiation toughness, i.e. for the onset of stable crack propagation, depends on the crack orientation. In some orientations and conditions, stable crack growth can occur without any obvious pop-in behaviour. This is detectable using the electrical dcpd method if the crack does not propagate by deflection parallel to the rolling plane.

## Crack Initiation Toughness Model.

The crack initiation toughness can be modelled by considering the requirements for stable brittle crack propagation. The duplex microstructure is treated as a uniform material, whose strength is dominated by the ferrite matrix hardness. Brittle fracture in the ferrite is nucleated by deformation twins. Twin/twin interactions nucleate cleavage cracks, whose propagation requires a critical tensile stress. The tensile stress requirement is assumed to be satisfied in the constrained plastic zone ahead of the crack tip. The cleavage crack nuclei are therefore unstable, and fracture is initiation controlled.

The fracture criteria can be expressed as (9),

$$\sigma_{12} > \tau_f \text{ for } x < x_0 \qquad (1)$$

where $\sigma_{12}$ is the shear stress ahead of the crack tip at distance x, $\tau_f$ is the critical shear stress for cleavage crack nucleation and $x_0$ is the critical distance describing the availability of unstable crack nuclei. Applying the fracture criteria to the stress distribution within the crack tip plastic zone, which is characterised by the stress intensity factor, K, gives the fracture toughness, $K_{IC}$, as,

$$K_{IC} = \beta^{-(N+1)/2} \sqrt{x_0} \frac{\tau_f^{(N+1)/2}}{\sigma_y^{(N-1)/2}} \qquad (2)$$

where N is the Ramsberg-Osgood strain hardening exponent, $\beta$ is the amplitude of the stress singularity and $\sigma_y$ is the tensile yield stress. The crack tip opening displacement, $CTOD_i$, can be calculated for plane strain small scale yielding as,

$$CTOD_i = 0.717 \frac{K_{IC}^2}{E\sigma_y} \qquad (3)$$

The following constants were used; E=200 GPa, $\beta$=0.59 and N=13. The shear stress, $\tau_f$, at fracture initiation has been measured as a function of yield stress in this material using smooth and notched test specimens (8). Equations 2 and 3 were then used to calculate the $CTOD_i$ for a critical distance, $x_0$=30 μm. The crack initiation toughness data, recorded by pop-in or dcpd methods as appropriate, is presented as a function of the yield stress at the test temperature (figure 3). The yield stress was determined from measurements of the empirical relationship between hardness, temperature and yield stress (8). There is good agreement between the model and the data over a broad range of conditions below the brittle/ductile transition. It is interesting to note that $x_0$ is of the same order as the size of ferrite regions between austenite in the duplex microstructure for the orientations tested. This is consistent with a critical distance that is related to the microstructure. Agreement between the data and the model becomes poor near to the brittle/ductile fracture transition, below a yield strength of approximately 800 MPa.

This simple model provides parameters to describe the lower shelf brittle fracture behaviour of a duplex stainless steel. In principle, the model may be extended to include the effects of ferrite content, grain size and hardness in the duplex microstructure. The model ceases to be valid once crack propagation from the fatigue pre-crack has occurred. A complete model for the fracture behaviour of duplex stainless steels requires understanding of crack initiation, crack propagation and the stability of bowed cracks.

## CONCLUSION

Unstable brittle fracture of duplex stainless steel can be preceded by stable brittle crack growth. This occurs at a crack tip opening displacement, $CTOD_i$, which is the crack initiation toughness. The most sensitive technique for measurement of $CTOD_i$ depends on the crack orientation. Clip-gauge methods to determine pop-in are most suitable when crack deflection occurs, whereas electrical methods are most sensitive in other orientations.

The variation of the crack initiation toughness with yield stress can be modelled below the brittle/ductile transition, using a criteria for the nucleation of unstable cleavage cracks by deformation twinning in the process zone ahead of the crack tip.

## ACKNOWLEDGEMENTS

The author is grateful to Weir Materials Ltd, Manchester, for the provision of materials.

## REFERENCES


(1) J.-O. Nilsson, (1992), Mater. Sci. and Tech., 8, 685-699.

(2) G. Warburton, M. Spence and A.W. Stevenson, (1992), Duplex Stainless Steels '91, Vol. 2, pub. Les Editions de Physique, France, 1225-1247.

(3) A. Dhooge and E. Deleu, (1995), International Welding Conference, 359-366.

(4) M. Guttman, (1992), Duplex Stainless Steels '91, Vol. 2, pub. Les Editions de Physique, France, 79-109.

(5) A.O. Humphreys, X. Li, M. Strangwood, P. Bowen, M.A. Spence and G.R. Warburton, 6th International Offshore and Polar Engineering Conference, pub. ISOP, Vol. 4, 263-268.

(6) L. Iturgoyen, J. Alcalá and M. Anglada, (1992), Duplex Stainless Steels '91, Vol. 2, pub. Les Editions de Physique, France, 757-764.

(7) E. Erauzkin and A.M. Irisarri, (1992), Fat. Frac. Engng. Mater. Struc. 15, 129-137.

(8) T.J. Marrow and C. Harris, (1996), Fat. Frac. Engng. Mater. Struc. 19, 935-947.

(9) T.J. Marrow, A.O. Humphreys and M. Strangwood, (1997), Fat. Frac. Engng. Mater. Struc. 20, 1005-1014.


## FIGURES

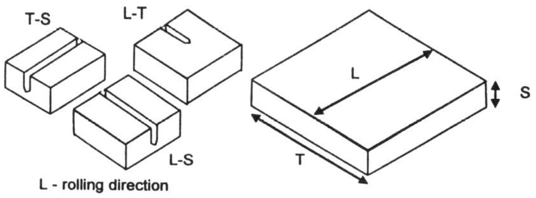


Figure 1: Test Specimen Orientations.

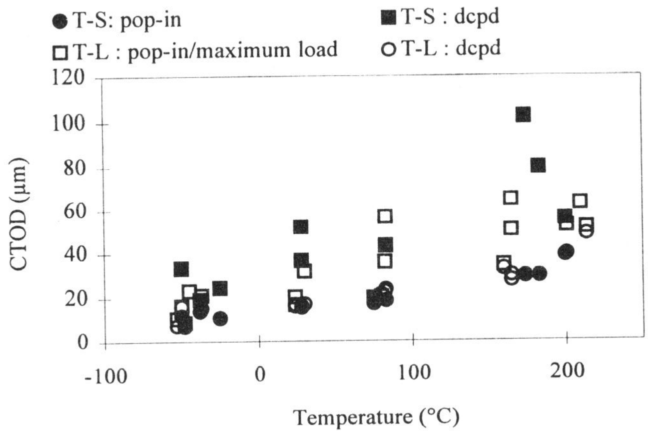


Figure 2: The effect of temperature on CTOD measured by clip-gauge (pop-in/maximum) and electrical (dcpd) techniques for orientations T-S and T-L.

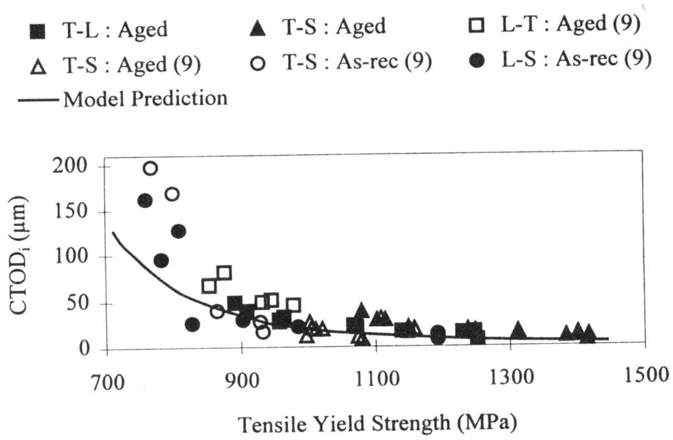


Figure 3 : The effect of yield strength (at test temperature) on the crack initiation toughness for orientations T-S, T-L, L-S and L-T. Data included from reference (9).