HOT DUCTILITY AND FRACTURE MECHANISMS IN A DUPLEX STAINLESS STEEL

M. Kilebu*, H. Herø** and O. Andersen*

INTRODUCTION

It is a well established fact that duplex stainless steel consisting of austenite with some ferrite suffers from poor ductility in the hot working range with a minimum at about 30 volume percent ferrite [1, 2, 3]. With regard to fracture mechanism no detailed studies have been reported, but it is believed to be phase boundary cracking [3, 4]. It is the purpose of the present work to investigate the initiation and propagation of cracks in these types of materials in the hot working range.

EXPERIMENTAL

The composition of the alloy studied was as follows in weight %: 0.045 C, 16.4 Cr, 5.1 Ni and 1.1 Mo. The material was produced by continuous casting and contained 15 vol. % ferrite in this state. The typical as-cast structure consisted of randomly oriented blocks, each containing parallel laths or rods of ferrite reflecting dendritics spacings in the primary grains. By rolling, this ferrite became more and more parallel with the rolling direction (r.d.). The two types of structure are illustrated in Figure 1. The elements Cr, Ni and Mo diffuse very slowly, and heat treatment results show that more than two hours in the temperature range 1200-1450°K are needed to affect the ferrite-amount or morphology in as-cast structure. In the present case, less than 1 hour above 1220°K was applied prior to rolling, and the ferrite content in the as-rolled structure was thus found to be only slightly smaller than in the as-cast.

Four series of high strain rate tensile tests were carried out. Ductility was measured as strain to fracture based on reduction in area measurements. The experimental conditions are given in Table 1.

In test series No. 4 the specimens had pre-machined necks. The calculated ratios of σ_T/σ_Y in the centre of the plane of the neck were 0.74 for the shape applied, using the Bridgeman solution [5]. Here σ_T is the hydrostatic stress and σ_Y the yield stress. The shape of the neck profile was machined so that the plastic strain increments should be constant within any plane perpendicular to the specimen axes in the case of non-strain-hardening material [6].

At room temperature the alloy contains small carbides at some of the ferrite-martensite (austenite in the hot working range) interface. To dissolve these carbides all tensile specimens were held for 30 minutes at 1370°K before the actual test temperature was attained. Ferrite content and morphology were not found to be influenced by this treatment.

* Stavanger Staal A.S., Jørpeland, Norway.
**Central Institute for Industrial Research, Oslo, Norway.

RESULTS AND DISCUSSION

In the test series No. 1, in Table 1, the ductility was measured by means of strain to fracture ε_f, and as a function of temperature T and rolling stage. The results are plotted in Figure 2. From this figure two significant observations can be made:

1) The ductility increases with the degree of hot rolling.

2) The ductility of as-cast structures goes through a minimum at approximately 1370°K.

These results indicate that the fracture mechanism for the as-cast and the as-rolled structures could be different and are therefore discussed separately.

As-cast Structures

In the second test series the specimens were pulled to various strains, but stopped prior to necking and cooled rapidly outside the heating zone of the furnace. The grooved sides resulted in a deformation mode more like plane strain than in uniaxial tension. A typical example of a specimen strained to $\varepsilon = 0.20$ is shown in Figure 3a and b with a SEM of the pre-polished surface and a micrograph of a cross-section parallel to the tensile axis respectively. It can be seen from these figures that slip has occurred preferentially in the dark etched ferrite and produced serrations in the surface. Further, slip takes place mainly in the ferrite which is both oriented approximately parallel to the maximum shear stress τ_{max} and running out to the free surface. This phenomena was accompanied by the initiation of numerous cracks along such ferrite lamellas close to the surface. These cracks quickly developed along ferrite lamellas as shown in Figure 4.

The observations described above led to the third series of tests where the tensile axis is at 45° to r.d. and hence also 45° to the ferrite rods. When pulled to failure these specimens normally developed a strongly shear-dominated fracture along the ferrite as shown in Figure 5. In this test the fracture mode was thus widely different from the cup-and-cone type fracture normally found in this as-rolled structure when pulled along r.d. Further, strain to fracture was reduced to about half of the value observed with the tensile axis parallel to the ferrite needles.

The latter two test series clearly indicate that the ferrite is considerably softer than the austenite in the hot working range, since this is the only reasonable explanation of the observed strain inhomogenity and fracture characteristics. This is in contrast to the observations made by Müller [7] for austenitic stainless steel containing ferrite, but consistent with Järvinen's tensile testing of stainless steels ranging from pure austenitic to pure ferritic composition [8], and with the general findings in carbon steels [9].

The shear strain in the ferrite associated with the serrations in Figure 3b is of the order of 5. In the present case, the first cracks to appear were observed along the ferrite laths running out to free surface. The extremely localized deformation and associated stress concentrations at these phase boundaries are probably responsible for this early crack initiation close to the free surface.

Away from the surface, shear deformation in ferrite lamellas will be restricted by differently oriented neighbouring blocks, and stress- and strain-concentrations must be expected to arise at block boundaries. The situation resembles, to some extent, the one with grain boundary sliding and crack initiation at triple points. In accordance with this, crack nucleation and propagation in the interior were found to occur along the ferrite phase at the block boundaries as shown in Figure 6. During crack propagation shear bands were occasionally formed across ferrite laths to link up cracks already formed along block boundaries. This is demonstrated in Figure 7.

As should be expected from these results, decreasing and spherodizing the ferrite phase by long - time annealing (5 - 29 h) is found to improve hot ductility of as-cast structure.

Järvinen [8] has found that strain inhomogenity in duplex stainless steel increases with temperature up to about 1370°K, due to falling ferrite flow stress to austenite flow stress ratio. From 1370°K to 1520°K a constant flow stress ratio of 0.3 was observed. Combined with the above described fracture characteristics and the general upgrading of ductility with increasing temperature, this seems to be a plausible explanation of the ductility minimum showed up by as-cast structure at about 1370°K.

As-rolled Structures

In test series No. 4, with pre-machined necks and hydrostatic stresses, the ferrite was oriented parallel to the tensile axis. The fractured surface had dimples which corresponded with the ferrite spacing. A polished section vertical to the tensile axis and approximately 0.3 mm away from the fracture surface is shown in Figure 8a ($\varepsilon_f = 1.4$). It can be seen that the material is highly porous. A higher magnification in Figure 8b and c reveals that cracks are initiated both in the interior of the ferrite and along the austenite/ferrite interface, but predominantly the former way. From Figure 8c it can be seen that the cracks in the ferrite partially make up a network. The spacing of this network turned out to be of the same order as the sub-grain size in the ferrite observed by TEM on rapidly quenched specimens, indicating that the cracks may be initiated on sub-grain boundaries.

The disappearance of the ductility minimum and general upgrading of ε_f by increasing the amount of hot rolling is most likely due to the fact that the ferrite rods are made more and more parallel to the r.d., whereby the fracture mechanisms found in as-cast structure are effectively eliminated.

Both the size and amount of voids shown in Figure 6a were observed to diminish from the centre of the cross-section towards the surface. The same decreasing tendency was found going along the tensile axis away from the fracture surface. This is in accordance with the magnitude of hydrostatic stresses which is known to enhance void growth substantially [10].

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Table 1 Tensile Tests

Test series No.	Specimen type	Processing stage	Strain rate sec^{-1}	Temp. °K	Orient. rel. to rol. dir.	Protection against oxidation
1	Round, 5 mm Ø	1. As-cast, sq. cross-section with 165 mm sides 2. Rolled to sq. cross-section with sides 140, 110 or 86 mm	1-2 (constant load)	1120 to 1470	0°	Enamelling
2	Flat with grooved sides, 2 mm thickness	As-cast, 165 mm sq. sides	1 (constant crosshead speed)	1170 to 1270	0°	Ar 99.997 % purity
3	Round 5 mm Ø	Rolled to 110 mm sq. sides	1-2 (constant load)	1120 to 1420	45°	Enamelling
4	Round with pre-machined necks	Rolled to 110 mm sq. sides	0.5-1 (constant crosshead speed)	1270	0°	Ar 99.997 % purity

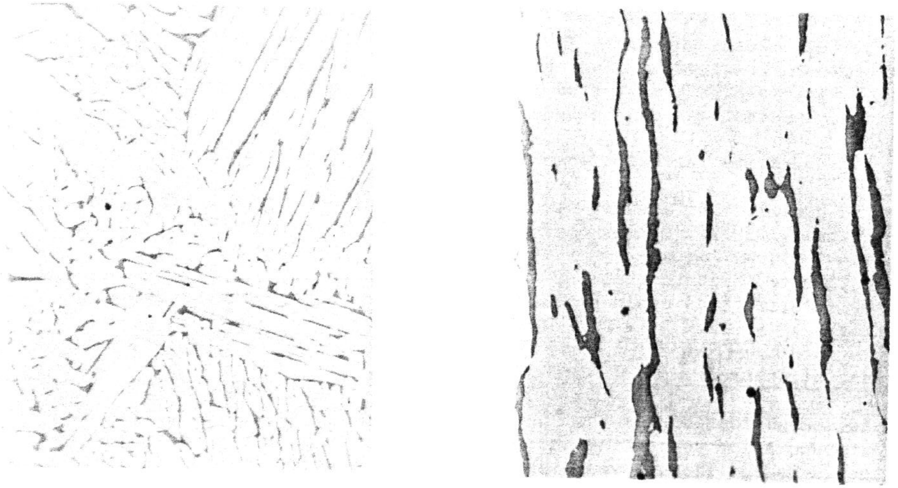


Figure 1 Microstructure of the Examined Material, As-Cast (left) and As-Rolled ε = 1.3 (right) 500X

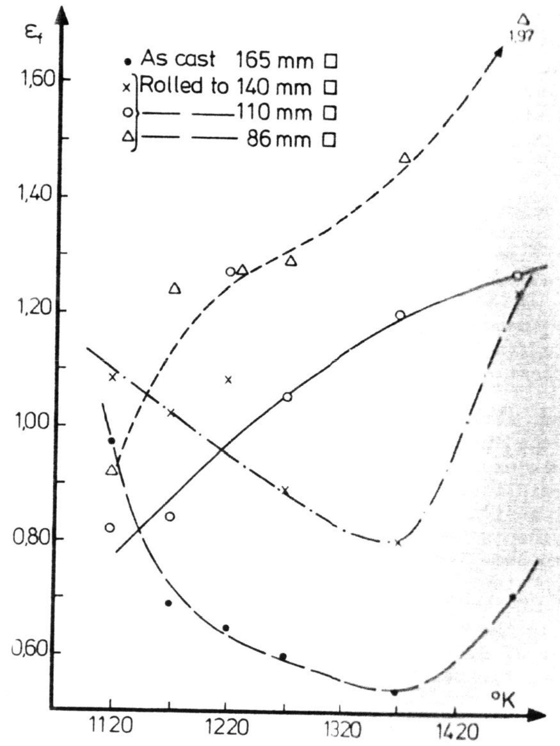
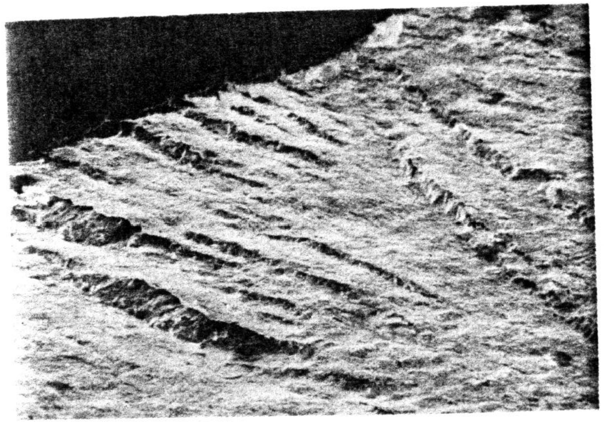
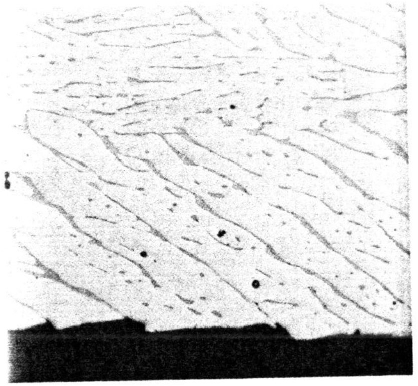


Figure 2 Strain to Fracture by Tensile Testing versus Temperature and Processing Stage, $\dot{\varepsilon}$ = 2 - 3 s^{-1}



(a)



(b)

Figure 3 Flat and Polished Specimen Strained to $\varepsilon = 0.20$, $\dot{\varepsilon} = 1\ s^{-1}$, $T = 1170°\ K$

(a) SEM of the Surface 200X

(b) Cross-Section Through the Terraced Surface in Figure 3a 300 X

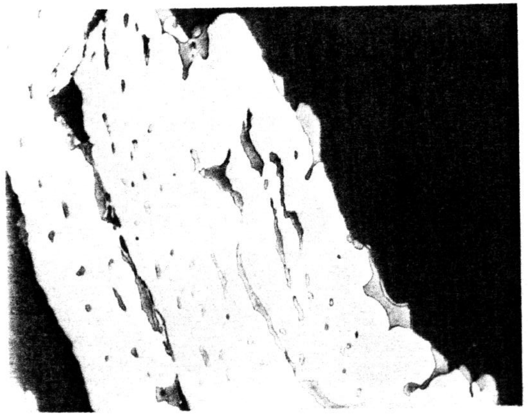


Figure 4 Crack Propagation Along Ferrite Lamellas 200X

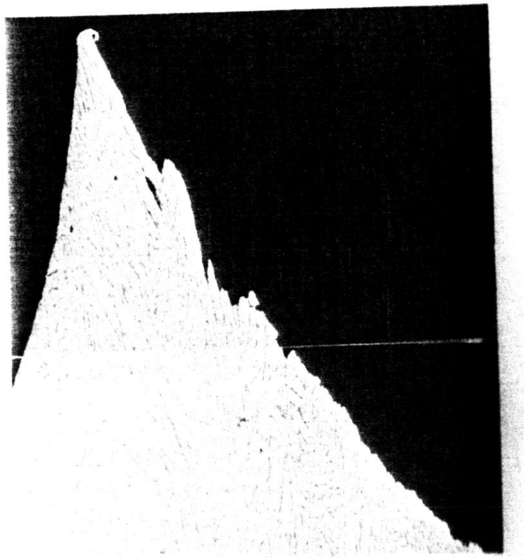


Figure 5 Entire Cross-Section of a Fractured Specimen with its Length Axis 45° to the Rolling Direction and Ferrite Rods.
$\dot{\varepsilon} = 2\ s^{-1}$. $T = 1370°\ K$ 24X

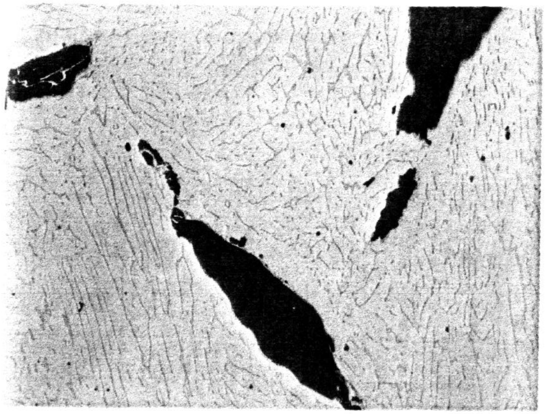


Figure 6 Crack Propagation and Interlinking Along Block Boundaries and Ferrite Lamellas 150X

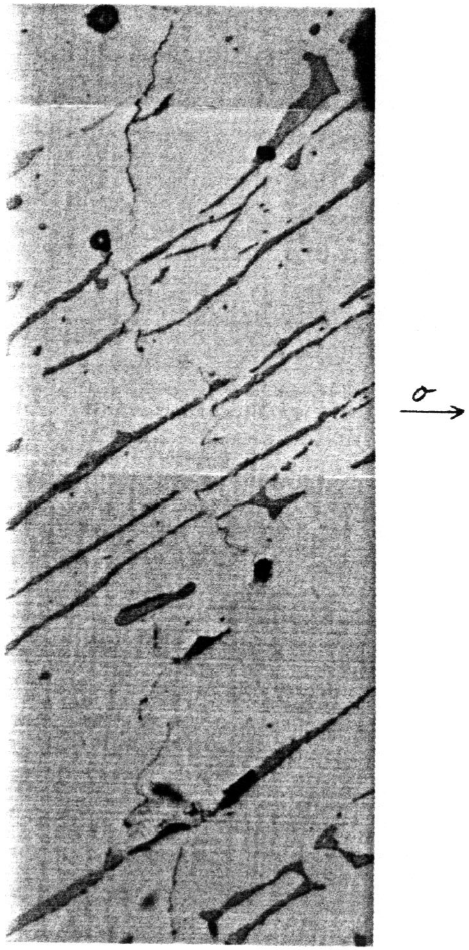
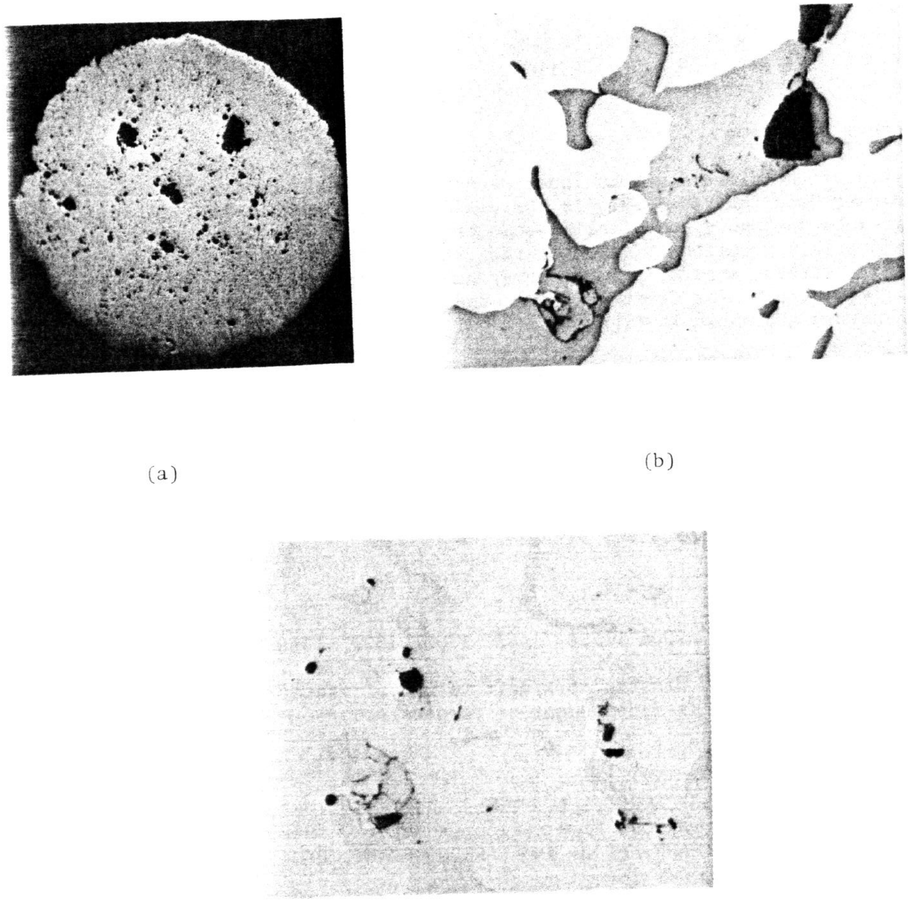


Figure 7 Cracks Formed Along a Shear-Band, $\dot{\varepsilon} = 1\ s^{-1}$, $T = 1270°\ K$. 1100



(a) (b)

(c)

Figure 8 Cross-Section Vertical to the Tensile Axis 0.3 mm from the Fracture Surface. Pre-Machined Neck. $\varepsilon_f = 1.4$, $\dot{\varepsilon} = 1\ s^{-1}$, $T = 1270°\ K$

(a) 25 X
(b) 1100 X
(c) 1100 X