CHAPTER 6

THE RELATIONSHIP BETWEEN SPLITTING IN TRANSVERSE CHARPY SPECIMENS AND LAMELLAR TEARING DURING WELDING OF 3CR12 PLATE

SYNOPSIS

In the first part of this study the mechanisms which are responsible for splitting in Charpy and edge-sheared specimens were identified. In the second part an attempt was made to establish the relationship between splitting and the ferrite factor and splitting and the susceptibility of 3CR12 to lamellar tearing during and after welding.

Planar oriented splits, parallel to the rolling plane, which develop during transverse impact testing and shearing, result in a decrease of the impact shelf energy as well as the ductile-brittle transition temperature. Transverse Charpy ductile-brittle transition (30J) temperatures within the range of -57°C to -68°C were measured for three experimental 3CR12 steels. The degree of planar splitting is mainly controlled by the degree of dimensional anisotropy of the grain structure. It was shown that splitting cannot be attributed to a single microstructural feature alone, but that it is governed by different mechanisms which depend on both the microstructure and the stress distribution at the crack tip. Splitting at sheared edges, which are subject to intense shear deformation, are nucleated by planar oriented inclusions. Splitting in Charpy specimens occurred by transgranular cleavage as well as by decohesion of ferrite-ferrite and ferrite-martensite grain boundaries and is not related to the presence of inclusions in the microstructure. The excellent toughness of 3CR12 arises from its fine grained microstructure and the fact that the splitting phenomenon is not related to the occurrence of inclusions in the microstructure.

The tendency for splitting to occur in transverse Charpy specimens was affected to a limited extent by normalising from 1000°C followed by tempering at 750°C. This resulted in a considerable increase in the transverse ductile-brittle transition temperature.

A direct relationship between the ferrite factor and the occurrence of

splitting in transverse Charpy specimens could not be established. Although the through-thickness ductile-brittle transition temperatures ranged from 107°C to 116°C, through-thickness tensile tests indicated the absence of a correlation between splitting observed in transverse Charpy specimens and the through-thickness ductility. It is concluded that splitting in transverse Charpy specimens is not expected to result in lamellar tearing, in the absence of planar oriented defects or inclusions, during welding.

THE RELATIONSHIP BETWEEN SPLITTING IN TRANSVERSE CHARPY SPECIMENS AND LAMELLAR TEARING DURING WELDING OF 3CR12 PLATE

1. INTRODUCTION

Splitting or laminated type fractures in Charpy, tensile, bend and sheared-edge specimens have been reported in the literature as a phenomenon which manifests itgself in high-strength low alloy steels(24, 25, 26, 27), some ferritic stainless steels(28, 29), ausformed(30, 31), marformed(30) and low carbon(32) steels. Lamellar type fractures (or splits), occur parallel to the rolling plane in the plate when hot rolling is terminated below the A_3 temperature. Splitting or lamellar fractures have also been found in Charpy, tensile and edge-sheared specimens of 3CR12. These fractures occurred when hot rolling was terminated at temperatures low in the duplex ferrite-austenite phase field. An example of this type of fracturing in transverse 3CR12 Charpy specimens is shown in figure 6.1, in which secondary splitting occurred, in the rolling plane, at right angles to the main fracture surface.

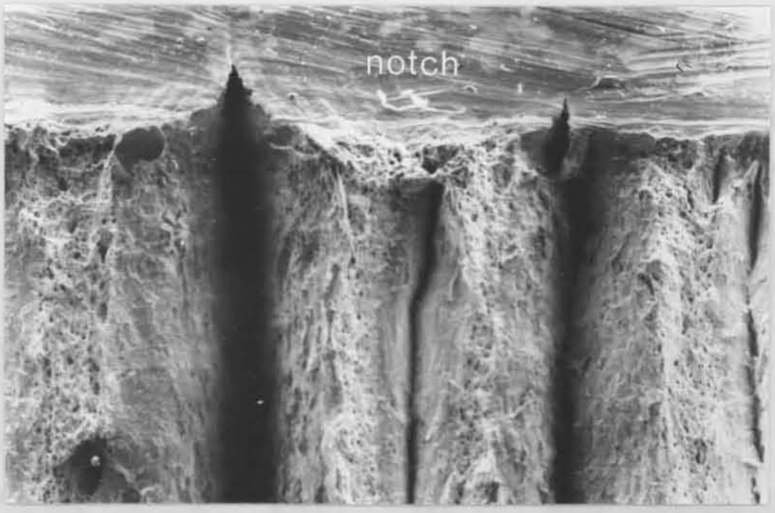


Figure 6.1: Splitting in a fractured transverse 3CR12 Charpy specimen. The splits are oriented in the rolling plane.

Although Hoffman does not consider these splits to have an adverse effect on the properties of 3CR12, the possibility of a correlation between splitting and the susceptibility of 3CR12 to lamellar tearing during and after welding has not yet been investigated(8). With 3CR12 being used in more and more higher strength structural applications this aspect of the weldability of 3CR12 becomes even more important.

3CR12 is currently produced with ferrite factors ranging from 7.8 to 11.5. It has already been shown in Chapter 4 that the phase compositions of both wrought 3CR12 plate and the HT HAZ are among other factors dependent on the ferrite factor. The ferrite factor may also influence the susceptibility of the steel to splitting. Before any correlation between splitting and the phenomenon of lamellar tearing during and after welding can be established it is of importance, first, to identify the mechanism which is responsible for splitting, and, secondly, to establish whether the susceptibility of the steel to splitting is influenced to any extent by the ferrite factor.

This study consists of two parts:

1. In the first part the mechanisms which are responsible for splitting in Charpy and edge-sheared specimens were identified. The influence of a normalising and tempering heat treatment on splitting was also assessed. Normalising and tempering was also used to simulate the effect of the weld HAZ thermal cycles on the occurrence of splitting.

2. In the second part an attempt was made to establish the relationship between splitting and the ferrite factor and splitting and the susceptibility to lamellar tearing during and after welding.

2. EXPERIMENTAL RESULTS AND DISCUSSION

2.1 A study of the mechanism by which splitting occurs along sheared edges and in transverse Charpy specimens of 3CR12

2.1.1 Introduction

A considerable amount of work was carried out in recent years in order to investigate the origin of splits or fissures. Many investigators have attributed the phenomenon of lamellar fracture in HSLA type steels either to the distribution and morphology of carbides(24,27), the dimensional anisotropy of the microstructural unit which controls fracture(25,27), polygonal or acicular ferrite structures resulting in transgranular cleavage, or to intergranular failure along prior austenite grain boundaries. Ausformed steels appear to split by decohesion along grain-boundaries as a result of fine alloy carbides dispersed along austenite grain boundaries while marforming seem to promote some kind of texture-banding which leads to splitting. Bramfitt and Marder attributed splitting in a high-purity single phase Fe-1%Mn alloy, which was hot rolled below the A_1 temperature, to decohesion along grain boundaries, independent of the texture of the material(32).

For ferritic stainless steels different mechanisms of laminar decohesion have been suggested. Mintz has shown that grain boundary shape, rather than texture, is the important factor controlling lamellar type fractures in a ferritic stainless steel containing 13% Cr(28). Hung-Chi Chao suggested that in the case of AISI type 430 ferritic stainless steel sheet, the banded cube-on-face texture component imbedded in other matrix orientations, tends to cause lamellar fracture of the sheet at low temperatures or high strain rates(29). Although splitting appears to be associated with microstructures exhibiting a high degree of dimensional anisotropy, there appears to be no single consistent metallographic or crystallographic feature uniquely responsible for splitting.

The purpose of this first part of this study as a component of a broader assessment of the weldability of 3CR12 was to:

a. identify the mechanisms responsible for splitting in Charpy and edge-sheared specimens;

b. assess the effectiveness of a normalising heat treatment to decrease splitting tendencies.

2.1.2 Materials and experimental techniques

2.1.2a Composition and mechanical properties

The chemical composition of the experimental 3CR12 steel is given in table 6.1. The steel was hot rolled by the producer into plate 12 mm thick, at a temperature within the two phase ferrite-austenite field, air cooled and finally subcritically annealed (750-780°C) to obtain the mechanical properties shown in table 6.2.

Table 6.1: Chemical composition of experimental 3CR12 steel , wt-%.

	C	N	Mn	Si	Cr	Ni	S	Ti
Specification	0.03 max	0.03 max	1.5 max	1.0 max	11-12	1.5 max	0.03 max	4(C+N) min 0.6 max
Experimental steel	0.023	0.016	1.22	0.33	11.23	0.57	0.008	0.45

Table 6.2: Mechanical properties of experimental 3CR12 steel.

Proof stress 0.2% offset (MPa)	Tensile Strength (MPa)	Elongation %	Hardness as received (BHN)	Hardness* heat treated (BHN)
345	512	36	164	157

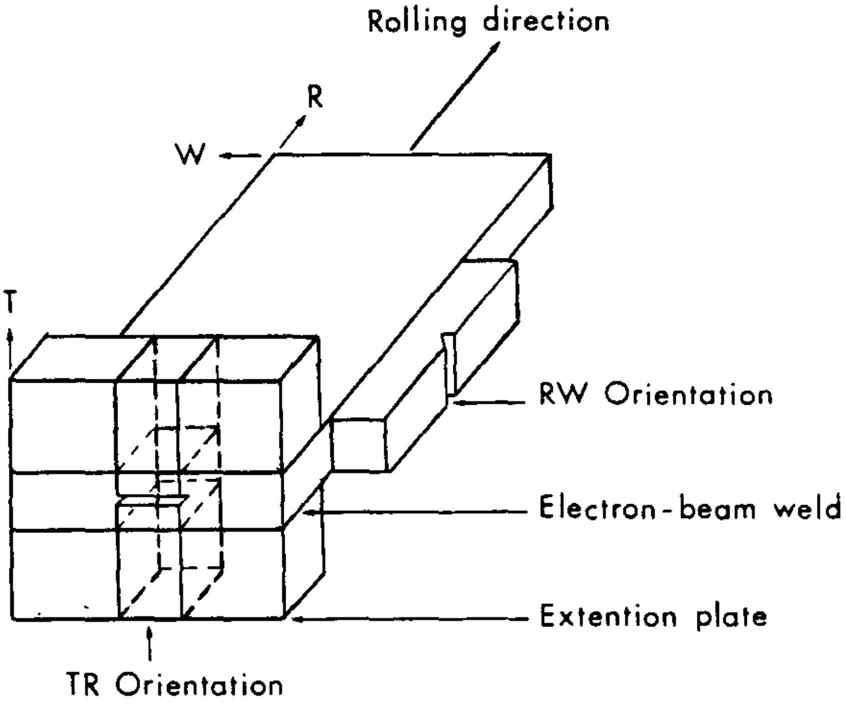
* Normalised from 1000°C followed by tempering at 750°C for 1h.

2.1.2b Heat treatment

The steel was tested in the as-received hot rolled-and-tempered condition as well as after a heat treatment consisting of normalising after heating for a period of 1h at 1000°C followed by subcritical annealing (tempering) for 1h at 750°C and subsequent air cooling. The Brinell hardness values of the steels in the hot rolled-and-tempered and the heat treated conditions are compared in table 6.2.

2.1.2c Specimen preparation

Standard 10 mm x 10 mm Charpy V-notch specimens were prepared according to ASTM E23 from both the conventional transverse (RW), the RT and through-thickness (TR) orientations shown in figure 6.2. The through-thickness impact specimens were prepared by electron-beam welding of extensions of the same material to the top and bottom surfaces of a transverse section of the plate shown in figure 6.2. Figure 6.3 shows an electron-beam welded through-thickness Charpy specimen.



R = rolling direction; W = width direction; T = through-thickness direction

Figure 6.2: Charpy specimen orientation and notation. The first letter refers to the axis of the specimen and the second to the direction in which the fracture propagates.

The 12 mm thick plate was supplied by the producer in 30 cm x 15 cm sheared sections, and sheared surfaces of the sections were evaluated in the as received condition.

2.1.2d Metallographic preparation

Following final metallographical polishing using 0.3 µm alpha alumina abrasive, optical microstructural studies were conducted on specimens etched with Kalling's reagent (5g $CuCl_2$ + 100 HCl + 100ml Ethyl alcohol + 100ml H_2O). It was found that a pseudo-structure, resulting from the presence of a surface layer of disturbed metal, is revealed on the first etch. However, by alternatively polishing and etching several times, the true structure of the specimen was revealed.

Figure 6.3: Electron-beam welded through-thickness Charpy impact specimen (TR orientation).

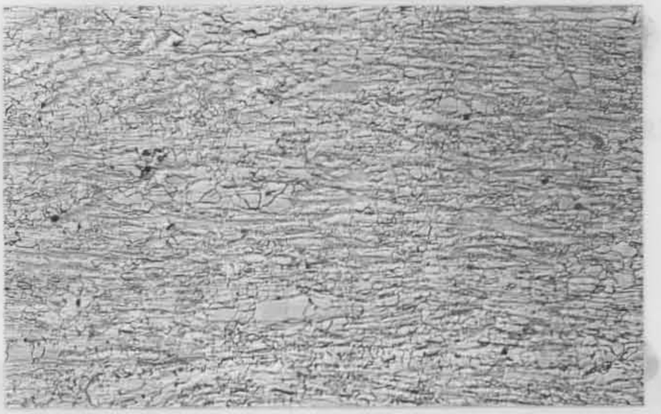
2.1.3 Results and discussion

2.1.3a Microstructures

The banded grain structure of 3CR12, in the hot rolled-and-tempered condition, as it appears in the TR-plane, parallel to the rolling direction (fig. 6.2) is shown in figure 6.4a. Figure 6.4b, on the other hand, shows the grain structure as it appears in the TW-plane, perpendicular to the rolling direction. The microstructure of the steel consists of narrow bands of small recrystalised ferrite grains and pancake shaped tempered martensite grains.

The change in microstructure induced by normalising from 1000°C followed by subcritical annealing at 750°C for 1h is shown by a comparison of the microstructure in figure 6.4a with the microstructure in figure 6.5. It is evident that the dimensional anisotropy of the ferrite-martensite grain structure has not been completely eliminated by this particular heat treatment. The retention of anisotropy is probably due to the low solubility of titanium carbo-nitrides at 1000°C and their ability to act as grain boundary precipitates to effectively restrict the movement of the ferrite-austenite grain boundaries(33). The most striking change that occurs on normalising is the change in the ferrite-martensite grain boundary from straight to undulating.

a.(300X)



b.(150X)

Figure 6.4: Microstructures of 12 mm hot rolled and tempered 3CR12 plate.
a. in plane TR (fig. 6.2) parallel to the rolling direction
b. in plane TW perpendicular to the rolling direction

Figure 6.5: Microstructure of 12 mm 3CR12 plate, in plane TR (fig. 6.2) parallel to the rolling direction, after normalising from 1000°C with subsequent tempering at 750°C for 1h (150X).

2.3.3b Splitting in edge sheared specimens

Visual examination of the sheared surfaces of 12 mm thick plate sections revealed two types of fracture surfaces. Figure 6.6 shows a sheared surface with scaly appearance as well as a smooth sheared surface. Transverse sections through the different sheared surfaces indicated that the scales shown in figure 6.6 were in effect small splits. Figure 6.7 seems to indicate that the fine splits are nucleated by voids which develop around individual coarse TiCN inclusions in the vicinity of the sheared interface. The sheared surface of most steels becomes rougher with an increase in the clearance between the edges of the shear. An increase in the clearance results in an increase of the volume subjected to intense

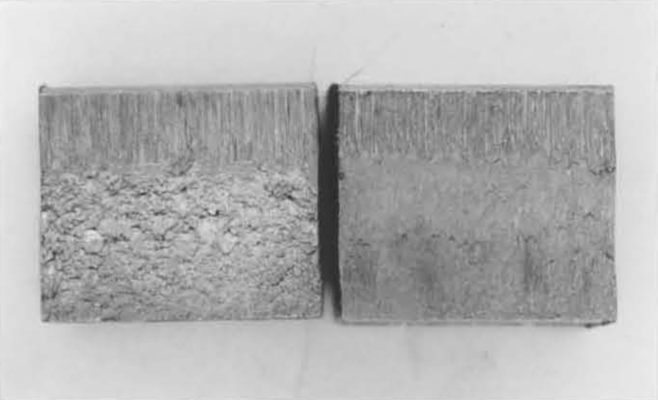


Figure 6.6: Scaly and smooth sheared surfaces of 12 mm 3CR12 plate.

shear strain. In 3CR12 this allows shear cracks which nucleate in the rolling plane to propagate to a greater extent.

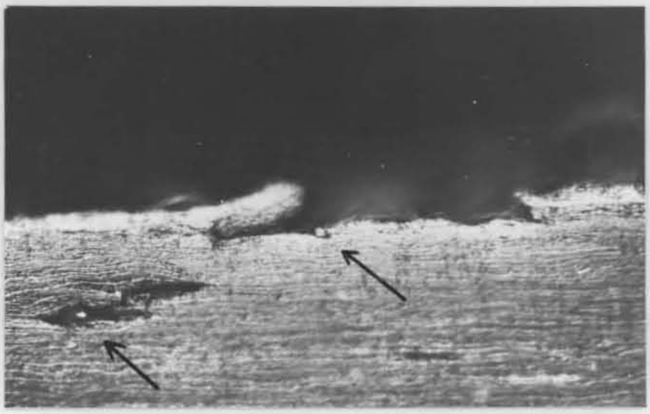


Figure 6.7: Section through the scaly sheared surface in figure 6.6. Arrows indicate voids around TiCN inclusions and a split which nucleated from a void on the sheared surface (270X).

The development and growth of voids around TiCN inclusions can be controlled by decreasing the clearance between the edges of the shear and maintaining sharp cutting edges. With a small clearance, the voids which develop around TiCN inclusions will usually not result in splitting on the shear surface, but internal shear cracks can develop from these voids just below the sheared surface as shown in figure 6.8.

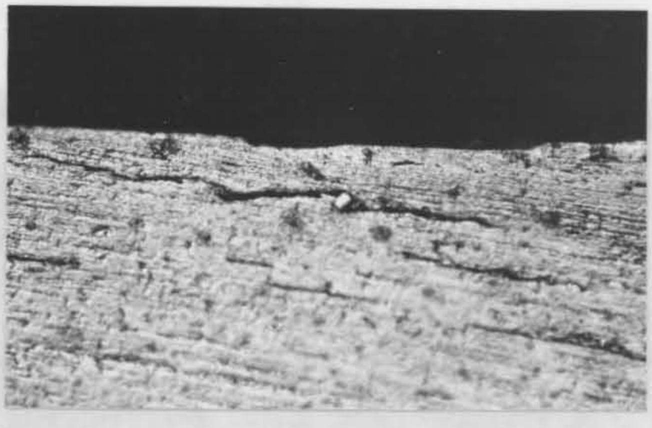


Figure 6.8: Section through the smooth sheared surface in figure 6.6. Internal shear cracks which originated from TiCN inclusions are evident (270X).

A third type of fracture surface was identified in specimens sheared in the laboratory with an excessive clearance of 2.4 mm between the edges of a blunt shear. Figure 6.9 shows the longitudinal sheared surface (parallel to the rolling direction) containing fine parallel splits which extend into the plate parallel to the rolling plane. A scanning electron microscope (SEM) analysis of the sheared surfaces revealed that the splits propagated mainly by transgranular cleavage, while a section through the sheared surface also showed that the splits were not associated with TiCN inclusions. The splits were possibly caused in this particular case by the propagation by cleavage of micro shear cracks as a result of high secondary tensile stresses which developed in the through-thickness direction during shearing.

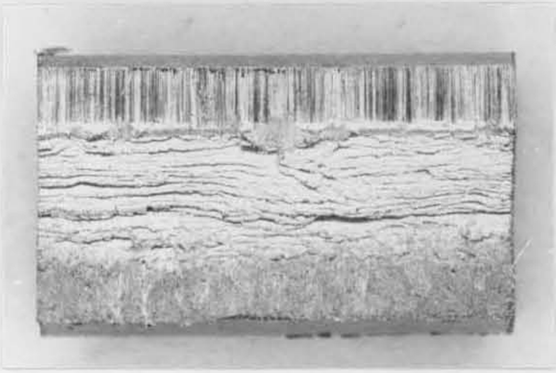


Figure 6.9: Longitudinal sheared surface with fine parallel splits. The specimen was sheared with a clearance of 2.4 mm between the edges of the shear.

It may be concluded that fine splitting of 3CR12 plate, which manifests itself as scales on the sheared surface, occurs by different mechanisms depending on the clearance between the edges of the shear. The quality of the sheared surface can be improved and controlled by close control of the clearance between the edges of the shear.

2.1.3c Splitting or delamination behaviour of Charpy V-notch specimens

Transverse Charpy test

Transverse Charpy specimens (RW orientation shown in figure 6.2) were tested in the as-received as well as in the heat treated condition. The fracture surfaces of the specimens exhibited a characteristic appearance with small transverse splits running parallel to the rolling plane. Figures 6.10 and 6.11 show the Charpy transition curves for the experimental steel and the fracture surfaces of Charpy specimens fractured at different temperatures, respectively. Transverse splits or fissures, which tend to divide the fracture area, occurred at all test temperatures. The number of splits increased as the test temperature was decreased(32).

The curves in figure 6.10 show that the 30J ductile-brittle transition temperature and upper shelf energy were increased by normalising from 1000°C, followed by subcritical tempering. Splitting still occurred after this heat treatment but there was a significant decrease in the number of splits. The development of splits in the heat treated steel is probably due to the dimensional anisotropy of the grain structure which was retained (fig. 6.5). Some proof of this conclusion is to be found in the observation that the hardness was not significantly changed by this heat treatment (table 6.2).

The very low ductile-brittle transition temperature (-64°C) of 3CR12 in the hot rolled-and-tempered condition can primarily be attributed to the development of small transverse plits which effectively release the high triaxial stress condition in the plastic zone at the root of the notch. The ductile-brittle transition temperature and, to a lesser extent, the shelf energy, shift to lower values when fracture is associated with transverse splits. The impact specimen tends to behave like a set of thinner samples(34). In effect these transverse splits resulted in brittle cleavage fracture occurring at lower temperatures(32).

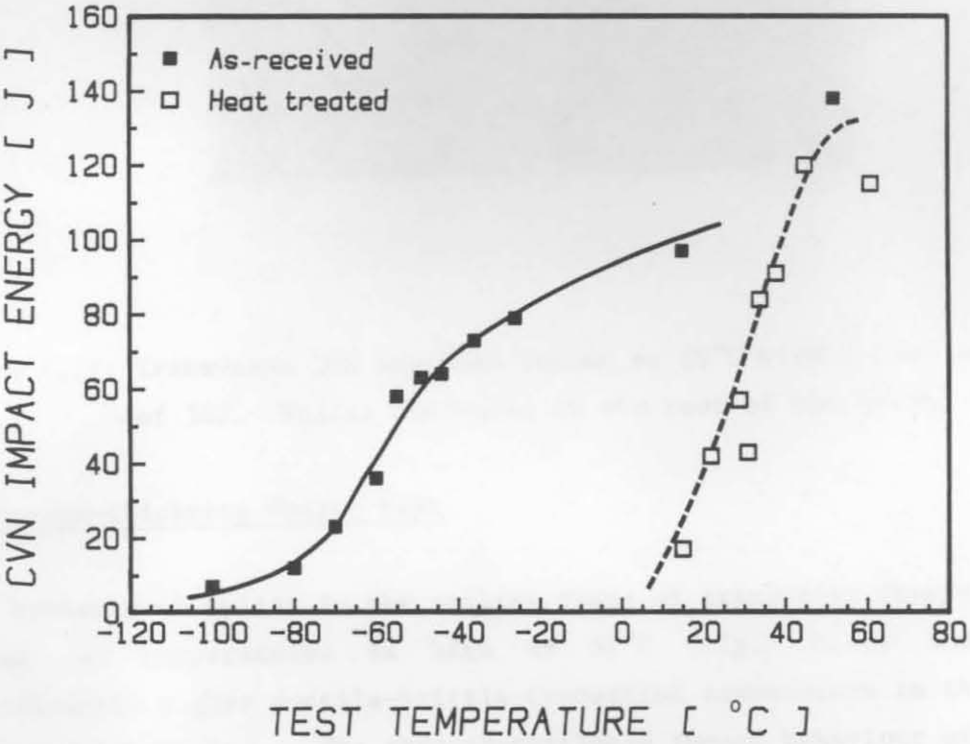


Figure 6.10: Transverse Charpy curves for 3CR12 tested in the as-received and heat treated conditions.

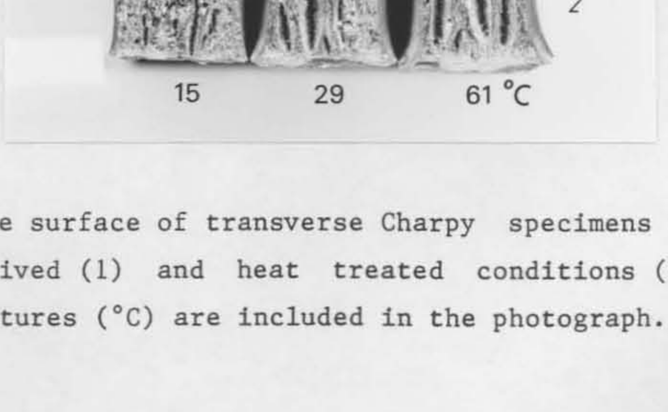


Figure 6.11: Fracture surface of transverse Charpy specimens tested in the as-received (1) and heat treated conditions (2). The test temperatures (°C) are included in the photograph.

Dabkowski suggested that splits may nucleate and propagate ahead of the principal propagating crack during impact testing of high-strength steels used for pipe lines(26). In order to determine whether the splits in 3CR12 steel occur before, after or concurrently with the passage of the principal propagating crack, an impact specimen was tested at 20°C in an impact machine by limiting the energy of the blow to 30J. The specimen was bent slightly and figure 6.12 shows that the root of the notch contains three small splits. These splits had already propagated to some extent without any transverse fracturing taking place. Splitting or laminated fractures in 3CR12 are therefore initiated ahead of the principal propagating crack. The principal fracture occurs by ductile failure of smaller subsized specimens created by splitting. The actual fracture appearance of the split surface will therefore be altered as a result of the large amount of plastic deformation which preceeds ductile failure of the material between the splits.

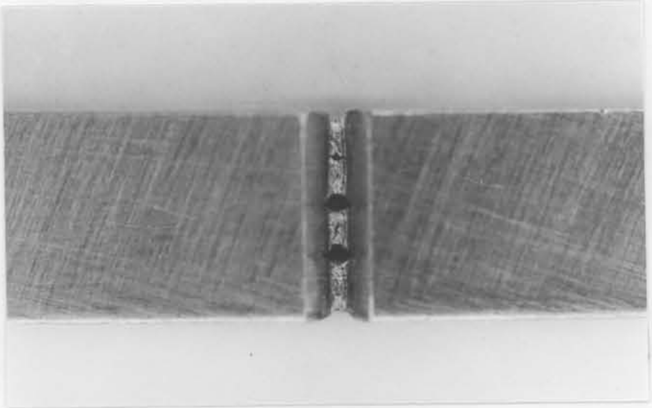


Figure 6.12: Transverse CVN specimen tested at 20°C with a low energy blow of 30J. Splits nucleated at the root of the notch.

Through-thickness Charpy test

The presence of splits in the rolling plane of transverse Charpy specimens tested at temperatures as high as 52°C (fig. 6.11) indicates a significantly higher ductile-brittle transition temperature in the through-thickness direction. The through-thickness impact behaviour of 3CR12 was therefore studied in order to determine the ductile-brittle transition temperature in the through-thickness direction and to characterize the appearance of the fracture surface parallel to the rolling plane.

Figure 6.13 shows the through-thickness Charpy curves for the 3CR12 steel tested in the as-received and heat treated conditions, respectively, while figure 6.14 shows the fracture surfaces of specimens fractured at different temperatures. The 30J through-thickness ductile-brittle transition temperatures are 112°C and 116°C for 3CR12 specimens tested in the as-received and heat treated conditions, respectively. The through-thickness transition temperatures are significantly higher than in the transverse direction (compare figures 6.10 and 6.13). A normalising heat treatment from 1000°C, followed by tempering at 750°C, caused a significant increase in the shelf energy with the ductile-brittle transition temperature being unchanged (fig. 6.13). The increase in shelf energy resulted from a transition from a ductile fibrous fracture to a conventional ductile fracture by micro-void coalescence (fig. 6.14). A notable feature shown in figure 6.13 is the virtual absence of any lower shelf energy for specimens tested at temperatures slightly below the ductile brittle-transitional temperature. The large difference between the transverse (20°C) and through-thickness (116°C) ductile-brittle transition temperatures after heat treatment is a feature which is associated with the dimensional anisotropy of the grain structure (fig. 6.5).

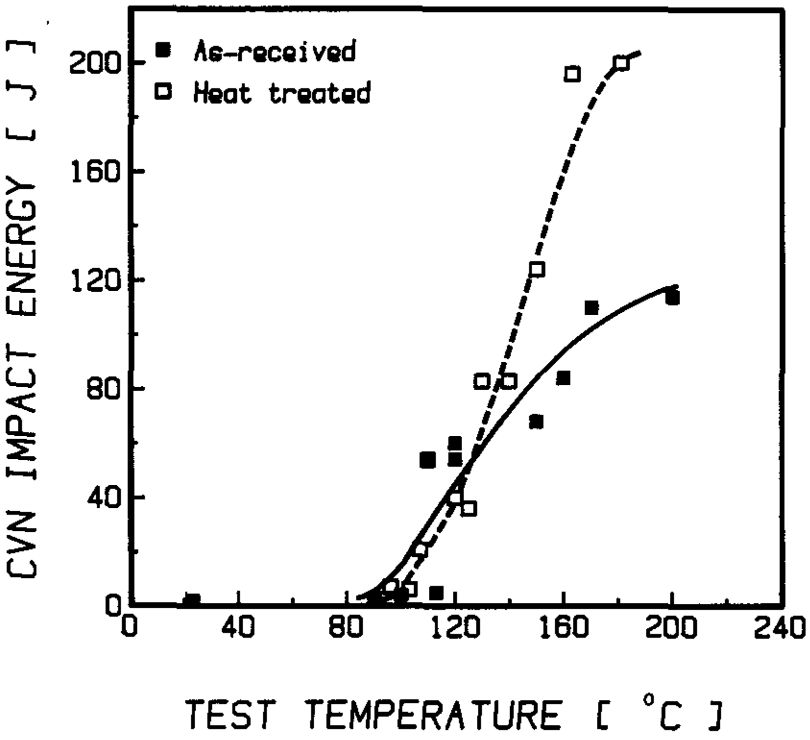


Figure 6.13: Through-thickness Charpy curves for 3CR12 tested as-received and heat treated.

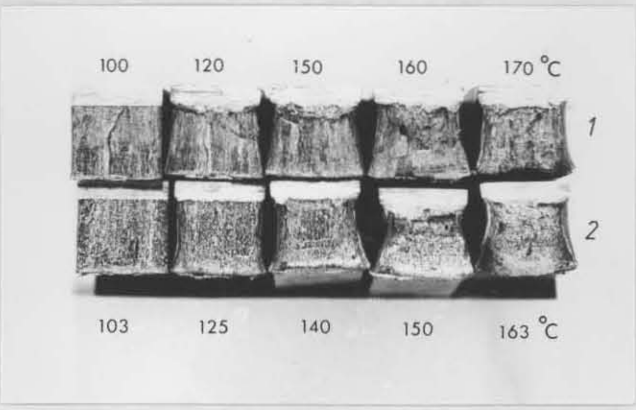


Figure 6.14: Fracture surfaces of through-thickness Charpy specimens tested in the as-received(1) and heat treated conditions(2). The temperatures (°C) at which the tests were conducted are are included in the photograph.

2.1.3d Metallographic and micro fractographic analysis of the splitting behaviour of Charpy specimens.

Splitting which preceeds the fracture of transverse Charpy specimens shown in figure 6.11 results in the subdivision of the fracture surface into an assembly of thinner specimens. Each sub-unit fractures in a ductile manner. This behaviour will result in a relatively high toughness of the steel even at very low test temperatures. In effect the transverse splitting ahead of the principal propagating crack changes the stress condition in the notch from plane strain to plane stress. Figure 6.15 shows the appearance of a fracture surface resulting from splitting. A portion of the fracture surface was relatively smooth and undulating adjoining a cleavage fracture surface with small cleavage steps. The appearance of the undulated fracture surface matches the pancake-shaped grain structure of the steel and probably arose as a result of intergranular decohesion. The undulating surface developed its characteristic appearance as a result of the subsequent tensile-elongation of each subunit prior to ductile fracture.

The original fracture appearance of the split surface has thus been altered by the subsequent large amount plastic deformation. Although Bramfitt and Marder(32) showed similar smooth undulating split surfaces in transverse

Figure 6.15: SEM micrograph of the split surface of a Charpy specimen showing a smooth undulating fracture surface (left) adjacent to a cleavage fracture (right) (1260X).

Charpy specimens of a high-purity Fe-1%Mn alloy and suggested that splitting occurred by decohesion of grain boundaries, the fracture characteristics of the split surface shown in figure 6.15 seems to indicate that splits may occur virtually simultaneously by grain boundary decohesion as well as by transgranular cleavage.

Polished and etched transverse sections through splits of a transverse Charpy specimen (fig. 6.11) tested at -100°C, revealed that the split crack propagated by decohesion of the ferrite-ferrite and ferrite-martensite grain boundaries as well as by transgranular cleavage through the ferrite grains (fig. 6.16). The transverse splits only gradually disappeared with the onset of cleavage fracture. Careful examination of the split surface and sections through splits showed that splitting in transverse Charpy specimens is not related to TiCN inclusions in the steels.

A visual examination of the through-thickness Charpy fracture surfaces of specimens machined from as-received plate and tested at 100°C and 120°C respectively, revealed two different fracture appearances on the same specimen (figs. 6.14 and 6.17A). The first 50% fracture surface adjacent to the notch, exhibited a dull appearance while the rest of the fracture surface had a brighter appearance. The fracture surface of the heat

Figure 6.16: Microstructure showing the propagation of a split by grain boundary decohesion of the ferrite-ferrite and ferrite martensite grain boundaries together with transgranular cleavage through the ferrite grains. Tested at -100°C (270X).

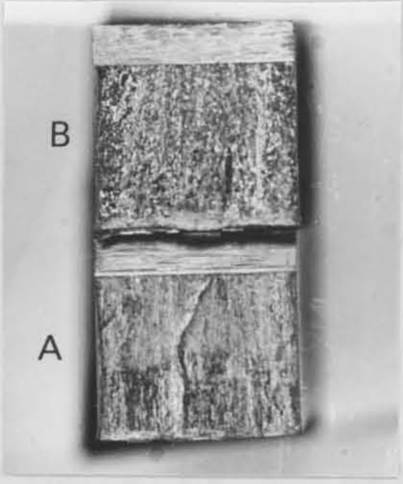


Figure 6.17: Through-thickness Charpy fracture surfaces of 3CR12.

A. Tested as-received at 100°C. Different fracture appearances on the same specimen are evident

B. Tested in the heat treated condition at 100°C. A shiny brittle fracture appearance is evident

treated specimens tested at 103°C and 125°C, respectively, exhibited the characteristic shiny appearance associated with a brittle cleavage fracture (figs. 6.14 and 6.17B).

Fractographic examination with the SEM indicated that the dull fracture surface, adjacent to the notch of the specimen machined from as received material and tested at 100°C was, of an intergranular nature. From a comparison of the equi-axed grains of this fracture surface with the microstructure of the steel shown in figure 6.4a, which contained pancake shaped tempered martensite grains (recrystalised), separated by small recrystalised ferrite grains, it can be inferred that fracture occurred mainly by decohesion along ferrite-ferrite grain boundaries. The grain diameter measured by both SEM and light micrographs are comparable. The other half of the fracture surface exhibited a combination of intergranular and transgranular cleavage fractures. The change in the fracture appearance was therefore associated with a change in the mechanism from intergranular decohesion to a combination of intergranular decohesion and transgranular cleavage. Brittle fracture of the heat treated through-thickness Charpy specimens occurred exclusively by transgranular cleavage in the rolling plane. The cleavage cracks are probably limited to the ferrite grains which are separated by tougher pancake shaped tempered martensite grains.

It can therefore be concluded that the mechanism of splitting or brittle crack propagation in the rolling plane of through-thickness Charpy specimens is dependent on the microstructure (heat treatment) and the stress state at the crack tip.

2.2 A study of the relationship between splitting and the ferrite factor and lamellar tearing during and after welding .

2.2.1 Introduction

A prerequisite for splitting is a large component of stress acting in the short transverse or through-thickness direction. Splitting in transverse Charpy, tensile and edge-sheared specimens occurs frequently in various low alloy and stainless steels including 12% chromium duplex ferrite-martensite steels. The degree of planar splitting in 12% chromium steels is mainly controlled by the degree of dimensional anisotropy of the grain structure. Splitting cannot be attributed to a single microstructural feature alone, but is attributed to different mechanisms which depend on the crack tip stress state as well as the microstructure. The type of lamellar tearing which occurs in transverse Charpy specimens is not related to inclusions and will be referred to as splitting.

In order to obtain the required properties, the chemical composition of 3CR12 is adjusted by carefully balancing the ferrite and austenite stabilising elements. For this purpose the ferrite factor proposed by Kaltenhauser is used(1).

$$\text{Ferrite Factor (F.F.)} = Cr + 6Si + 8Ti - 40(C + N) - 2Mn - 4Ni + 4MO + 2Al$$

Although the Kaltenhauser relationship was initially derived to calculate the phase composition of ferritic stainless steel weld metal, it can also be used as a qualitative parameter to predict the relative balance between the ferrite and austenite phase stabilisers in wrought steel. The phase composition of 3CR12 which depends on the thermomechanical treatment is therefore characterized by the ferrite factor. A low ferrite factor will yield a higher martensite content and vice versa. Duplex ferite-martensite structures are obtained in 3CR12 when the values of the ferrite factor range from 7.8 to 11.5.

The mechanical properties of flat-rolled products which traditionally have been of interest, are those in which stress is applied in the rolling plane. The short transverse properties in which stress is applied in the thickness direction are of particular interest during welding in weld configurations susceptible to lamellar tearing. This applies especially in thick plate with high through-thickness restraint(28). Lamellar tearing is sometimes also referred to as splitting. Lamellar tearing, however, is usually related to the planar orientation of highly segregated inclusions. The splitting encountered in transverse Charpy specimens of 3CR12 suggests a susceptibility to lamellar tearing not related to inclusions. De Ardo studied the influence of splitting, in high strength low alloy steels, on the susceptibility to lamellar tearing(28). He concluded that the occurrence of splitting in impact specimens is neither related to, nor a good indication of, the probability of lamellar tearing during welding.

The purpose of this second part is to report on results of a study in which an attempt was made to establish, first the relationship between the ferrite factor and the phenomenon of splitting in transverse Charpy specimens of hot rolled-and-tempered 12mm thick 3CR12 plate, and, secondly, the contribution of splitting to the susceptibility to lamellar tearing during welding.

2.2.2 Materials and experimental techniques

The compositions of the three 3CR12 steels used in this study are shown in table 6.3. The ferrite factors of these steels span the range of 12 mm thick 3CR12 plate currently produced. The plate was produced by completing hot rolling at a temperature of about 800°C. Subsequently the plate was subcritically annealed or tempered at 750°C-760°C. The mechanical properties of the steels so produced are given in table 6.4. All the tests were performed on the steel in the as-received condition.

Standard 10 mm x 10 mm Charpy V-notch specimens, with the longitudinal direction in the rolling direction, were prepared with the notch in the short transverse direction. Also Charpy V-notch specimens with the notch oriented to result in a fracture path parallel to the rolling plane, were prepared by electron beam welding extensions to the top and bottom plate surfaces.

Through-thickness tensile specimens were specially prepared from a composite which was manufactured by furnace brazing mild steel extensions to the top and bottom surfaces of square sections cut from the rolled plate. (The maximum temperature during brazing was 650°C while the time at temperature varied between 1-3 seconds). Miniature tensile specimens with a 9 mm gauge length and 3.8 mm diameter were prepared from this composite as shown in figure 6.18. The through-thickness ductility was determined by measuring the total percentage reduction in area.

Table 6.3: Chemical analysis of 12 mm 3CR12 plate, wt-%.

Steel	C	N	S	P	Mn	Si	Ti	Cr	Ni	F.F.
A	0.027	0.022	0.008	0.020	1.32	0.44	0.32	11.34	0.61	9.70
B	0.023	0.016	0.008	0.017	1.22	0.33	0.45	11.23	0.57	10.75
C	0.021	0.010	0.013	0.017	1.22	0.48	0.36	11.20	0.53	11.26

Table 6.4: Mechanical properties of 12 mm 3CR12 plate material.

Steel	Proof stress (0.2%) MPa	Tensile Strength MPa	Elongation, Percentage	Brinell Hardness HBN
A	336	509	30	169
B	345	512	36	164
C	390	526	29	164

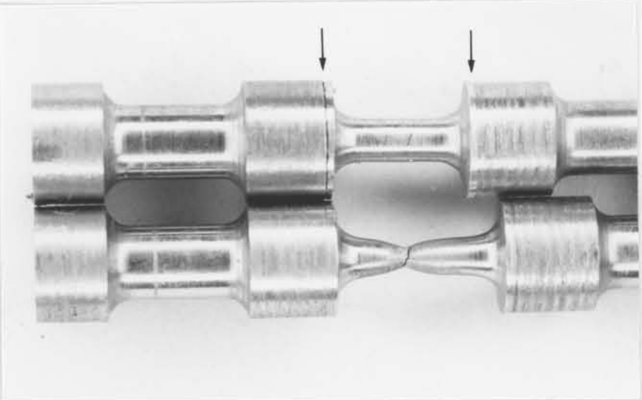


Figure 6.18: Composite through-thickness tensile specimen. Brazed joints are indicated by arrows.

2.2.3 Results and Discussion

2.2.3a Microstructures

The microstructures of steels A and C (table 6.3) with ferrite factors 9.70 and 11.26 respectively are shown in figures 6.19 and 6.20. The

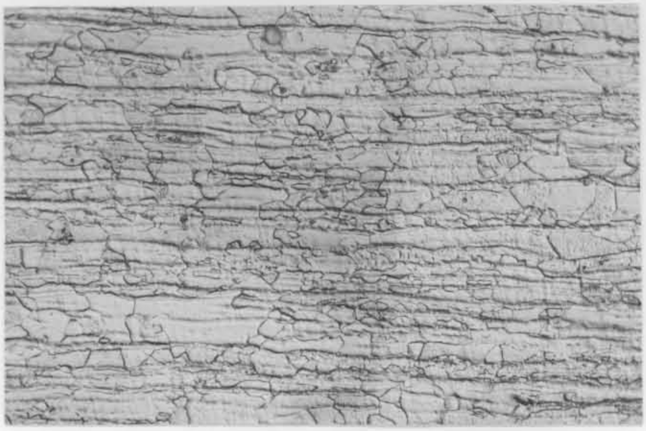


Figure 6.19: Microstructure of steel A (F.F. = 9.70), in the as-received condition. The banded structure consists of pancake shaped tempered martensite grains separated by recrystalised ferrite bands which lie in the rolling plane (300X).

Figure 6.20: Microstructure of steel C (F.F. = 11.26). The higher percentage ferrite is evident when this microstructure is compared to the microstructure in figure 6.19. The ferrite bands are also slightly thicker due to the higher percentage ferrite (300X).

microstructures of the steels consist of narrow bands of recrystalised ferrite and pancake shaped tempered martensite grains. It is evident from figures 6.19 and 6.20 that an increase in the ferrite factor results in an increase in the average width of the recrystalised ferrite bands (with a concurrent decrease in the percentage tempered martensite). The pancake shaped ferrite and martensite grains are oriented in the rolling plane.

2.2.3b Charpy impact testing

Charpy impact tests were conducted at various temperatures on the experimental steels in the as-received condition. The results of the impact tests are summarized in figure 6.21. Splitting occurred in all of the specimens tested in the temperature range -100°C to 55°C. Some of the Charpy fracture surfaces are shown in figure 6.22. The number of splits increases with a decrease in test temperature and the onset of cleavage fracture is suppressed to temperatures below -60°C. In specimens tested below -20°C the number of splits which developed was the same for all three steels. Specimens of steel B tested at temperatures above -20°C exhibit a smaller number of splits than steels A and C. This is reflected by the slightly higher impact energy in figure 6.21. The 30J ductile-brittle transition temperatures of the three steels are shown in table 6.5. Although the ductile-brittle transition temperature decreases slightly with a decrease in the ferrite factor, this does not necessarily indicate a correlation between the ferrite factor and splitting. It has already been shown that the transverse toughness of 3CR12 is greatly enhanced by the occurrence of splitting. The very low transition temperatures of the three 3CR12 steels are at least partially due to splitting which releases the high triaxial stress state at the root of the notch. Figure 6.23 shows a comparison of the impact energy, testing temperature relationship for steel B with the fracture plane in the long and short transverse planes, respectively.

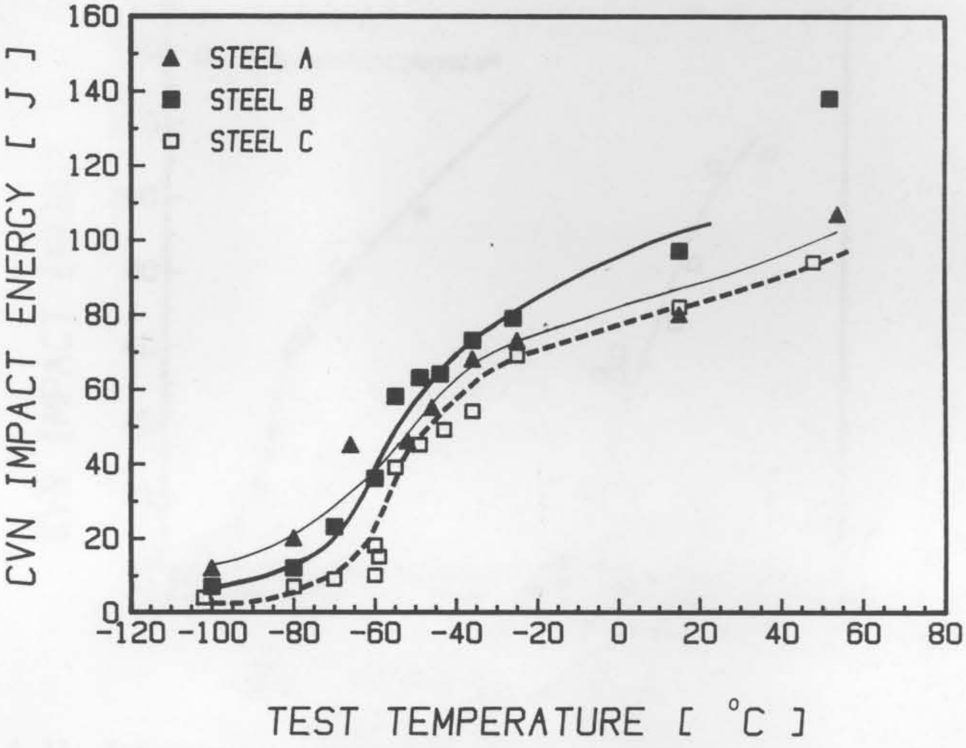


Figure 6.21: Transverse CVN impact curves for steels A (F.F. = 9.70), B (F.F. = 10.75) and C (F.F. = 11.26) tested as-received.

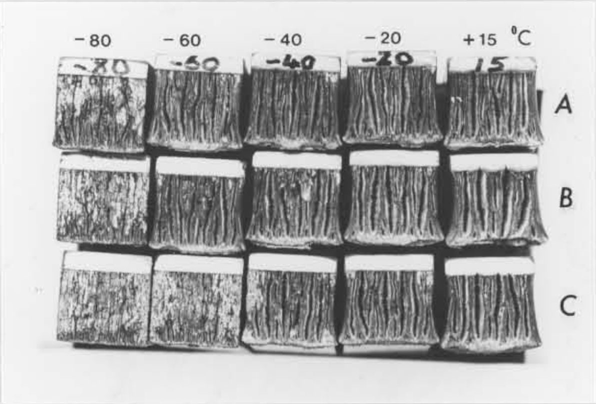


Figure 6.22: Charpy fracture surfaces of 3CR12 steels A, B, and C tested at different temperatures (°C).

Table 6.5: Transverse ductile-brittle transition temperatures (°C) at 30J for 3CR12 steels tested in the as-received condition

Steel	Ferrite Factor (F. F.)	Transition Temperature (°C)
A	9.70	-68
B	10.75	-64
C	11.26	-57

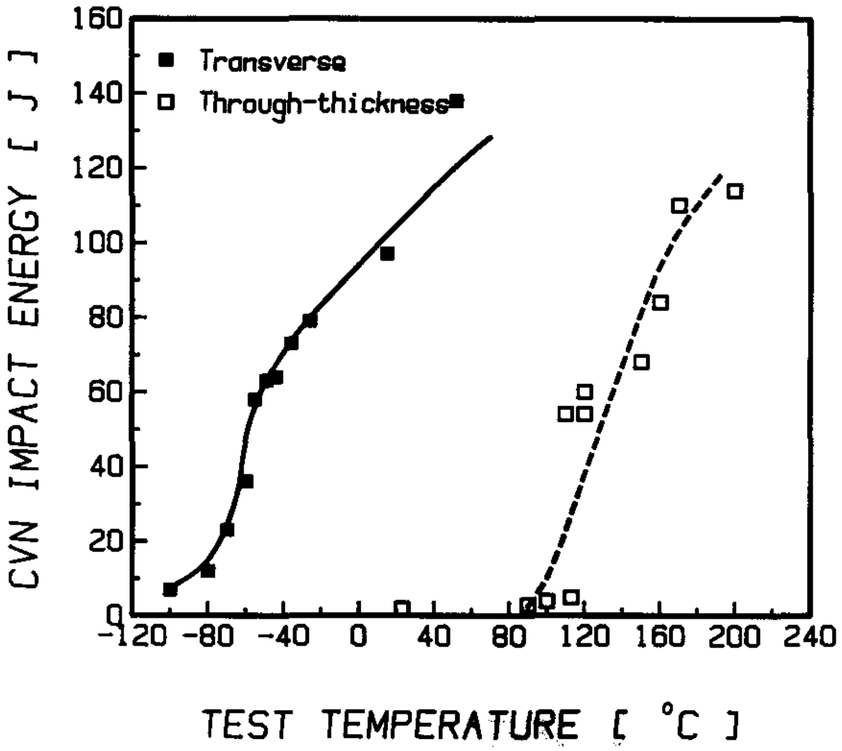


Figure 6.23: Transverse and through-thickness CVN impact curves for steel B

2.2.3c Through-thickness tensile tests

The presence of splits parallel to the rolling plane in transverse Charpy specimens and the high through-thickness ductile-brittle transition temperature (112°C for steel B) raises the issue whether 3CR12 steel is not highly susceptible to lamellar tearing during welding. Lamellar tearing is usually an inclusion-related phenomenon. Farrar, Ginn and Dolby have shown a good correlation between the short transverse ductility and the susceptibility of a steel to lamellar tearing during welding(29). The susceptibility to lamellar tearing increases with a decrease in the short transverse ductility due to the presence of elongated inclusions. However, the type of splitting which occurs in transverse Charpy specimens of 3CR12, is not an inclusion related phenomenon and the relationship between splitting in transverse Charpy specimens and the short transverse ductility of duplex ferrite-tempered-martensite, 12% chromium steels has as yet not been determined.

A study by De Ardo, on the mechanism of splitting in four high strength low alloy steels, revealed that cementite, either as grain boundary precipitate or as a constituent of pearlite in a particular distribution, appeared to be responsible for splitting(25). No direct correlation between splitting and the short transverse ductility could be found and it was concluded that splitting in transverse Charpy specimens of HSLA steels is not related to the phenomenon of lamellar tearing, which sometimes attends welding.

In the present study the short transverse ductilities of three 3CR12 steels, table 6.1, were determined and the results are summarized in table 6.6. All the specimens exhibited elliptical fracture surfaces (fig. 6.24)

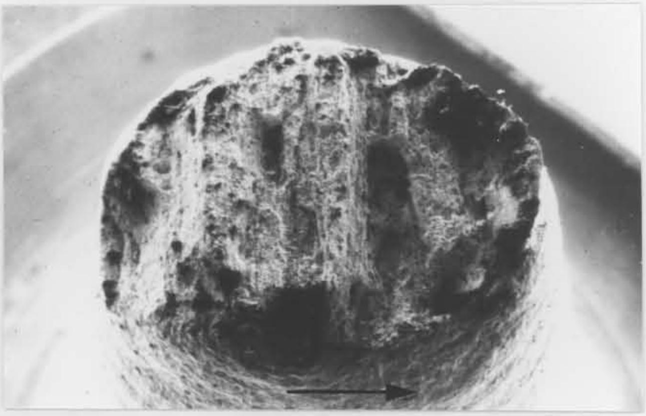


Figure 6.24: Micrograph of the elliptical through-thickness tensile fracture surface of specimen 3 of steel C in table 6.6. The arrow indicates the long transverse direction.

Table 6.6: Through-thickness tensile test results of 3CR12 steel A, B, and C tested in the as-received condition.

Steel	Specimen No.	Tensile strength(MPa)	Reduction in area, pct.	
			L*	Z*
A	1	477	34	55
	2	480	27	27
	3	490	36	52
	4	488	42	56
	Average	484	35	48
B	1	460	56	73
	2	475	57	78
	3	473	50	73
	4	-	17	68
	Average	469	45	73
C	1	465	40	69
	2	473	33	63
	3	-	42	67
	4	456	17	43
	Average	465	33	61

*L : Longitudinal ductility : reduction in area (fig. 6.24)

*Z : Long transverse ductility : reduction in area (fig. 6.24) as a result of the dimensional anisotropy of the grain structure in the rolling (longitudinal, L) and long transverse (Z) directions. The test results in table 6.6 show that the three 3CR12 steels exhibit a reasonably high short transverse ductility. An SEM analysis of the fracture surfaces revealed complete dimpled ductile fracture surfaces with small patches of inclusions which have been identified as mainly titanium carbonitrides. The few exceptions in table 6.6 were specimens which contained excessive amounts of titanium carbonitride stringers. No relationship between the ferrite factor and the short transverse ductility could be found from the limited number of tests performed.

The high through-thickness ductility of the 3CR12 steels, together with the absence of any brittle transgranular or intergranular fracture in the short transverse tensile specimens, indicate that there is no apparent relationship between planar oriented splitting in transverse Charpy specimens and the through-thickness ductility. This observation indicates that the occurence of splitting in transverse Charpy impact specimens is neither related to, nor a good indicator of, the probability of lamellar tearing that may occur during and after welding. The presence of an abnormally high amount of segregated, planar oriented inclusions can result in lamellar tearing during welding. The absence of splitting during tensile testing of specimens with a through-thickness orientation may probably be attributed to the fact that the fracture stress necessary for splitting is in excess of the true stress when ductile fracture due to micro-void coalescence occurs. Splitting will therefore only be possible when plastic deformation is constrained in such a way that the local stress exceeds the fracture stress required for splitting.

3. GENERAL DISCUSSION

3.1 Mechanism of splitting in transverse Charpy specimens

Various researchers have attempted to relate the splitting or delamination phenomenon in transverse Charpy specimens to specific microstructural features(24-30). It appears, however, that splitting of transverse Charpy specimens is not related to a specific microstructural or crystallographic feature. Careful examination has shown that it is not related to any type of inclusion in the steel.

A model for the phenomenon of splitting in transverse Charpy specimens of 3CR12 can be formulated by comparing the results obtained in this study with those obtained by Emburry(34) on so-called 'crack-divider' laminated Charpy specimens (fig. 6.25). He used standard-size laminated Charpy specimens, containing up to six layers of annealed mild steel, which were prepared by machining transverse Charpy specimens from a composite manufactured by silver soldering thin annealed mild steel plates. The transition temperature and shelf energy were lowered as the number of crack-dividing layers increased. Examination of the fracture surfaces revealed that a pair of shear lips formed on each subunit in contrast to the single set of shear lips associated with a homogeneous specimen. The fracture mechanism of a specimen which allows crack division can be considered to be equivalent to that of a thin plate-like Charpy specimen laterally supported to prevent buckling.

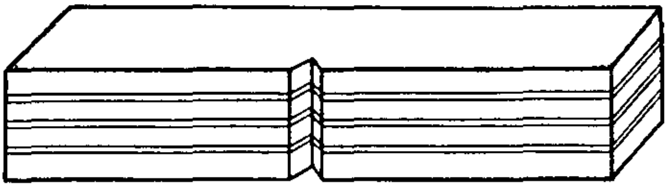


Figure 6.25: A schematic diagram of a 'crack-divider' Charpy specimen. (After Emburry)

Considering the pancake-shaped grain structures (fig. 6.4) normally present in 3CR12 steel, transverse Charpy specimens can also be considered as laminated, 'crack-divider' specimens. During Charpy impact testing, plastic deformation in the plastic zone at the root of the notch is highly restrained. Consequently the stress state just ahead of the root of the notch is characterised by a high degree of triaxiality. As a result of the plastic constraint the tensile bending stress can in extreme conditions be as high as three times the yield stress and the tensile stress induced at right angles to the direction of splitting in transverse Charpy specimens can reach a value of approximately 2.5 times the yield stress(17). If the tensile stress during transverse Charpy testing is below the fracture stress necessary to induce transverse fracture, and the tensile stress at right angles to the direction of splitting exceeds the fracture strength required for splitting, splitting will preceed transverse fracturing. Depending on the fracture toughness of the material in the direction in which the split propagates, the distance that a split will extend will be governed by the size of the plastic zone which develops at the root of the Charpy specimen. Any local split will cause a relaxation of the local triaxial stress condition in the vicinity of the split. The distance over which the stress is relaxed will depend on the length of the split. Further parallel splits can, therefore, only nucleate some distance from an existing split. The number of splits which develop during testing of a given material will, therefore, depend on the size of the plastic zone at the root of the transverse Charpy specimen. In the brittle transition temperature regime where the size of the plastic zone decreases as the testing temperature is lowered, the distance between splits can also be expected to decrease. Ample experimental proof for this model is to be found in figure 6.11.

When splitting occurs, the triaxial stress condition at the root zone of the notch is relaxed and the stress state altered from mainly plane strain to plane stress(17). The reduction of the stress state in the notch as a result of splitting will therefore reduce the temperature at which cleavage fracture can be expected. Splitting can, therefore, be expected to reduce the brittle transition temperature in accord with Emburry's results. The reduction in the stress state will, however, also cause a reduction in the upper shelf impact energy.

Emburry's laminated 'crack-divider' Charpy specimen can, therefore, be considered as a valid model for laminated fracture of transverse 3CR12 Charpy specimens. Splitting in transverse 3CR12 specimens will probably thus not have a crystallographic origin per se but will occur rather by delamination along planes or interfaces in a laminated structure which has a lower fracture stress than the stress required for transverse fracture. Although a micro analysis showed no significant partition of alloying elements between the ferrite and tempered martensite phases, different possible low fracture-stress planes or interfaces have been identified. Delamination of transverse Charpy specimens occurred by transgranular cleavage and decohesion of the ferrite-tempered-martensite and ferrite-ferrite grain boundaries (fig. 6.16), whereas through-thickness Charpy specimens fractured by transgranular cleavage (probably through ferrite grains) and decohesion of the ferrite-ferrite grain boundaries.

An elongated grain structure (laminated) and the consequent anisotropic mechanical properties appear to be the most important prerequisite for splitting to occur in transverse Charpy specimens cut from rolled plate. The excellent toughness (low transverse ductile-brittle transition temperature) of hot rolled and tempered 3CR12 plate is, apart from the small grain size, also due to splitting which promotes partial relaxation of the plastic constraint. The very favourable influence of the laminated grain structure on the fracture toughness of 3CR12 is further illustrated by a fractured Charpy specimen in figure 6.26, prepared with the notch in the rolling plane, which absorbed 288J energy at -30°C.

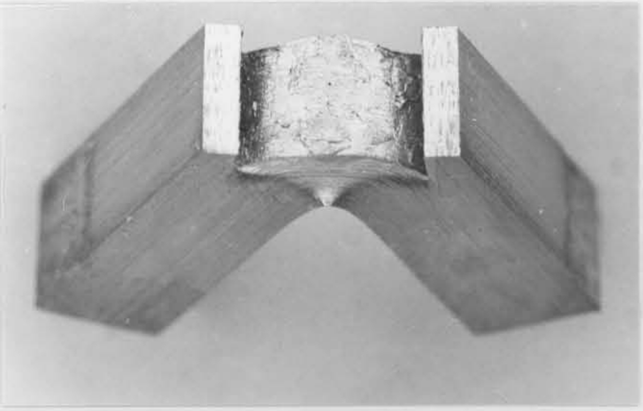


Figure 6.26: Extensive delamination in a longitudinal 3CR12 Charpy specimen prepared from hot rolled and tempered 12 mm plate with the notch in the rolling plane. The specimen absorbed 288J at -30C.

3.2 Influence of heat treatment

Normalising from 1000°C, followed by tempering at 750°C, partially eliminated the dimensional anisotropy of the grain structure (fig. 6.5) without influencing the tensile mechanical properties significantly. It can be expected, however, that the heat treatment reduced the difference in the transverse fracture stress and the fracture stress required for splitting. In terms of the model of the splitting mechanism, normalising can be expected to increase the distance bertween splits. The ductile-brittle transition temperature can therefore be expected to be increased by normalising (as also predicted by Emburry's results). In fact the ductile-brittle transition temperature was increased considerably from -64°C to +20°C by a normalising and tempering heat treatment. Following the normalising heat treatment, there was still some anisotropy in the fracture stress sufficient to cause some splitting. This is confirmed by the large difference in transition temperature following normalising for specimens with a transverse orientation (20°C) and specimens tested in the through-thickness direction (114°C).

The large increase in the transverse ductile-brittle transition temperature of 3CR12 after a normalising and tempering heat treatment, with the hardness unaffected, indicates that the toughness of this steel is greatly enhanced by the degree of dimensional anisotropy of the grain structure. Hot rolling should therefore be terminated at temperatures low in the two-phase ferrite-austenite phase field in order to obtain the desired toughness levels.

3.3 Relationship between the ferrite factor and splitting in transverse Charpy specimens

It has already been shown that Emburry's laminated 'crack-divider' Charpy specimen is a valid model for the splitting encountered in transverse Charpy specimens of 3CR12. It was also shown that a laminated microstructure, with planes or interfaces parallel to the rolling plane with a low fracture stress, is an important prerequisite for splitting to occur in transverse Charpy specimens. Splits form during the initial stages of the Charpy test, before the initiation and propagation of the main crack, when the lateral stresses which develop as a result of lateral constraint, exceed the brittle fracture stress on the low fracture toughness planes or interfaces. This results in a relaxation of the stress triaxiality in the vicinity of the split.

The number of splits, or the distance between the splits at a given test temperature, will be related to the number of sub-specimens in the composite specimen, or the distance between the low fracture stress planes in the rolling plane and the amount of stress relaxation in the vicinity of a split. If the transverse size of the stress relaxed zone associated with a split (it is that zone in which the transverse stress is lower than the fracture stress of the low fracture stress plane) is smaller than the distance between the low fracture stress planes, the number of splits will be determined by the number of low fracture stress planes (fig. 6.27).

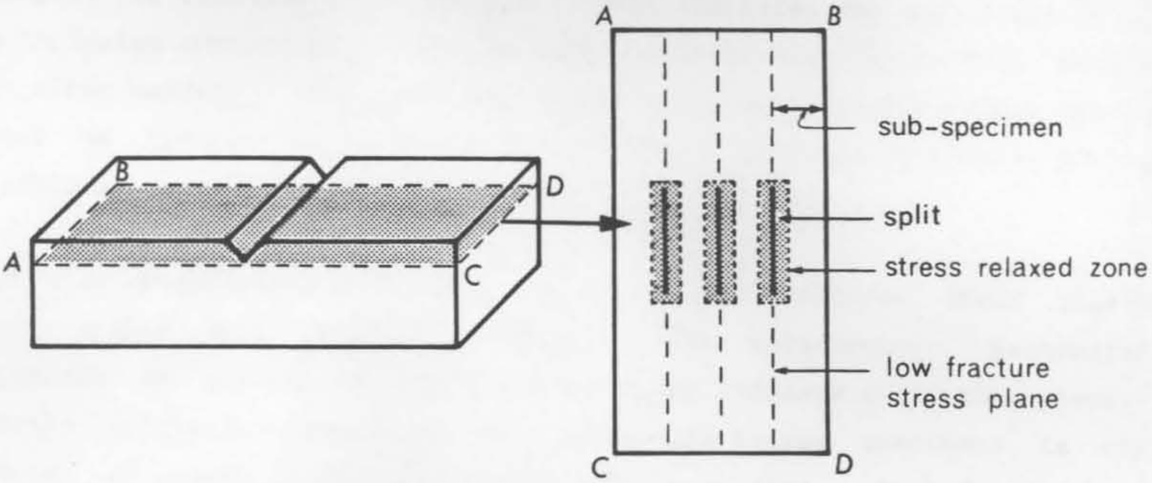


Figure 6.27: Section through the splits which have developed in the rolling plane, ahead of the notch tip of a transverse Charpy specimen prior to initiation of the principal crack at the notch tip during impact loading.

With the distance between the splits much larger than the distance between the low fracture stress planes, the number of splits at a given test temperature will rather be determined by the size of the stress relaxed zone associated with a split. The size of the stress relaxed zone will then approximately be equal to the distance between the splits. Measurements of the distance between the splits at the ductile-brittle transition temperatures of steels A, B and C gave an average value of 0.85 mm. If this is compared with the thickness of the pancake shaped grains (sub-specimens) of steels A (fig. 6.19), B and C (fig. 6.20), which varied between 0.003-0.009 mm, 0.006-0.010 mm and 0.006-0.010 mm respectively, it is evident that the number of splits will not be influenced by the ferrite factor within the range of 9.7 to 11.26 since the thickness of the pancake shaped grains, which decreases with a decrease in the ferrite factor, is much smaller than the size of the zone in which the lateral stresses are relaxed.

A comparison is made in figure 6.21 of the transverse Charpy curves for steels A, B and C. The differences between the curves for steel B (F.F. = 10.70) and C (F.F. = 11.26) are probably related to the differences in the ferrite factors (table 6.5). The slightly higher transition temperature of steel C is due to the lower percentage tempered martensite in the structure. The ductile-brittle transition temperature of steel A is slightly higher than that expected from the value of the ferrite factor alone, and is probably due to the lower Ti:(C+N) ratio of 6.5 for this steel, compared to a ratio of 11.5 for steels B and C. The results show therefore that there is no direct relationship between the ferrite factor and the incidence of splitting. The small difference between the impact curves can be accounted for by the ferrite factor and by the ratio Ti:(C+N).

3.4 Relationship between splitting and lamellar tearing during welding

The low short transverse ductility of three of the twelve tensile specimens of the three 3CR12 steels shown in table 6.6, was mainly due to segregated stringers of titanium carbonitrides. This indicates the importance of a low inclusion content to prevent inclusion-related lamellar tearing during and after welding. The number of titanium carbonitride inclusions can of course be reduced by limiting the carbon and nitrogen contents during steelmaking.

High through-thickness ductilities were found for the three 3CR12 steels which necked down without splitting. The satisfactory mechanical properties in the through-thickness direction indicate that the planar-oriented splitting associated with transverse Charpy specimens is not expected to result in lamellar tearing during welding. De Ardo reported similar results for high-strength, low-alloy steels(28). The degree of anisotropy in this instance was not as marked as that of the present material. It must, however, be emphasized that the brittle transition temperatures in the through-thickness direction of most other HSLA steels are much higher than the transverse values. The fracture toughness in the through-thickness direction at room temperature will be quite low. Fortunately 3CR12 also has a fairly low yield strength so that the critical crack length due to planar oriented defects, such as that due to segregated inclusions, will be tolerable. In critical applications where plane strain, rather than plane stress, conditions prevail, such as weld configurations with a high thickness constraint, care should be exercised.

4. SUMMARY

a. A variety of microstructural features are responsible for splitting during shearing of 12 mm thick plate and for the development of laminated fracture surfaces in transverse Charpy specimens.

b. Splitting of 12 mm thick 3CR12 plate during shearing occurs by different mechanisms which are governed by the clearance between the edges of the shear. Fine scaly splitting originated mostly from voids which formed around coarse TiCN inclusions, evenly distributed in the plate. A large clearance of 2.4 mm between the edges of the shear, resulted in long parallel splits, parallel to the rolling plane, by transgranular cleavage.

c. Fine scaly and parallel splits on sheared edges can be controlled, and probably eliminated, by maintaining small clearances between the edges of the shear, and by maintaining sharp edges.

d. Splitting along the rolling plane in transverse Charpy specimens is a phenomenon which arises from the anisotropy of the fracture stress.

e. The mechanism of splitting in transverse and through-thickness Charpy specimens (decohesion of the ferrite-ferrite, ferrite-martensite grain boundaries or transgranular cleavage) is governed by the microstructure (prior heat treatment) and the stress state at the crack tip during fracture.

f. The very low transverse ductile-brittle transition temperature of the hot holled and tempered 3CR12 steels is a feature which is associated with splitting. In order to obtain low ductile-brittle transition temperatures, hot rolling should be terminated at low temperatures in the two-phase austenite-ferrite phase field in order to obtain a banded grain structure.

g. The tendency for splitting to occur in transverse Charpy V-notch specimens was affected to a limited extent by normalising from 1000°C followed by temperaturing at 750°C. This resulted in a considerable increase in the transverse ductile-brittle transition temperature.

h. The difference between the transverse and through-thickness ductile-brittle transition temperatures decreased from 176°C to 134°C on normalising from 1000°C, followed by subcritical tempering at 750°C.

i. There is no direct relationship between the ferrite factor (in the range 9.7 - 11.26) and splitting in transverse Charpy specimens of 3CR12.

j. The high through-thickness tensile ductility of 12 mm thick plate indicates that splitting is dependent on plane strain rather than plane stress conditions.

k. The through-thickness tensile ductilities of the experimental steels were high in spite of the phenomenon of splitting which is encountered in transverse Charpy specimens. The radius of the plastic zone at the tip of a split is probably fairly large and it is not expected to result in lamellar tearing in the absence of planar oriented inclusions during welding.