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LA RESISTANCE A LA RUPTURE DUCTILE DES COMPOSANTS REP. ETUDE DE L'EFFET DE GEOMETRIE DES EPROUVETTES ET DU MODE DE CHARGEMENT SUR LES COURBES DE RESISTANCE J-ΔA

DUCTILE CRACK GROWTH RESISTANCE OF PWR COMPONENTS. APPLICATION FOR STRUCTURAL INTEGRITY ASSESSMENT

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SYNTHÈSE :

L'étude de la tenue mécanique des composants nucléaires nécessite de connaître la résistance à la déchirure ductile des matériaux concernés. Cette résistance se traduit par une courbe, fondée sur la théorie de l'intégrale J, donnant l'évolution de J en fonction de la propagation de la fissure Δa. Les courbes étant déterminées en laboratoire à partir d'essais effectués sur éprouvettes CT, leur application aux composants industriels pose question. En effet, on sait bien que les valeurs mesurées dépendent de paramètres tels que les dimensions des éprouvettes et du type de chargement appliqué (tension ou flexion) : les courbes dépendent en fait de l'état des contraintes et des déformations. Mais l'influence de ces paramètres dépend des mécanismes d'endommagement des matériaux qui se produisent avant l'amorçage de la rupture et durant la propagation.

La comparaison de courbes de résistance, d'acier de cuve, obtenues sur éprouvettes CT et sur composants de grandes dimensions soumis à des chocs thermiques, confirme que ces courbes sont identiques pour un même état de contraintes et de déformations. Dans le cas contraire, l'effet du mode de chargement a pu être calculé par la méthode de l'approche locale qui modélise la coalescence de trous initialement formés autour des inclusions caractéristiques de la rupture ductile.

Des essais effectués sur de l'acier au C-Mn des tuyauteries secondaires montrent que les courbes de résistance dépendent des dimensions des éprouvettes CT : les valeurs les plus basses sont obtenues à partir des éprouvettes les plus épaisses. Cet effet provient vraisemblablement de l'apparition de décohésions le long des inclusions allongées observables en surface de rupture.

Avec des aciers austéno-ferritiques vieillis les différences entre les courbes de résistance déduites d'essais sur éprouvettes de laboratoire et de tubes droits sous pression interne n'ont pas pu être correctement interprétées avec le modèle de rupture ductile. Un modèle a donc été développé, adapté aux mécanismes spécifiques de rupture, qui se caractérise par le clivage de la phase ferritique suivi du cisaillement des ligaments d'austénite.

De même, avec des joints soudés, présentant des zones à caractéristiques mécaniques différentes, les courbes de résistance peuvent être correctement évaluées par une modélisation fine.

Les valeurs de tenacité de l'acier de cuve dans le domaine de la transition dépendent des dimensions des éprouvettes : cet effet a été interprété par une modélisation qui permet de simuler numériquement le début de propagation par déchirure ductile suivi par le déclenchement du clivage. L'effet observé est lié à l'aspect statistique du déclenchement du clivage.

L'ensemble de ces résultats démontre que les courbes de résistance à la rupture ductile des matériaux dépendent bien de l'état des contraintes et des déformations mais il montre aussi l'influence des mécanismes de rupture réellement mis en jeu dans le matériau.

EXECUTIVE SUMMARY :

Integrity assessment of nuclear components implies knowledge of the ductile crack growth resistance of the materials concerned. This resistance is defined by J resistance curves, based on the J integral theory, which gives the J versus Δa crack propagation. Since the curves are determined in the laboratory, using tests on CT test specimens, their application to industrial components is questionable. We know that the measured values depend on parameters such as the size of the specimens and the type of load applied (tensile or bending), so that the curves in fact reflect the stress-strain condition produced. But the influence of these parameters depends on material damage mechanisms which occur prior to failure initiation and during crack propagation.

A comparison between vessel steel resistance curves obtained on CT test specimens and on large components subjected to thermal shocks confirms that these curves are identical providing the stress-strain condition is the same. In cases where this is not so, the effect of the loading mode has been calculated by the local approach method which models the coalescence of holes initially formed round the inclusions characterizing ductile cracking.

Tests performed on the secondary piping C-Mn steel show that the resistance curves depend on the size of the CT test specimens, with the lowest values corresponding to the thickest test pieces. This effect is probably due to decohesions occurring along elongated inclusions and observable at the fracture surface.

With age-hardened austeno-ferritic steels, the differences between resistance curves derived from laboratory specimen tests and those corresponding to straight tubes subjected to internal pressure could not be correctly interpreted with the ductile crack growth model. So a model was developed adapted to the specific fracture mechanisms involved, characterized by cleavage of the ferritic phase followed by shearing of the austenite ligaments.

Similarly, in the case of welds between metals featuring dissimilar mechanical characteristics, resistance curves can be correctly assembled by fine modelling.

Vessel steel toughness values in the transition region depend on the size of the test specimens and this effect was interpreted in the model by numerical simulation of incipient ductile crack propagation followed by cleavage initiation.

All these results confirm that ductile crack growth resistance curves for materials depend on the stress-strain condition but also evidence the influence of the fracture mechanisms at work within the material.

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M. Bethmont*, C. Eripret*, P. Le Delliou*, J.M. Frund*

J-resistance curves deduced from different materials and test specimens are compared. It is shown that the specimen geometry and the mode of loading effects are dependent on the physical mechanisms of fracture of the materials.

INTRODUCTION

Structural integrity assessment of PWR components, as pressure vessel and piping, needs to evaluate the ductile crack growth resistance which is generally characterized by J resistance curves (or J-R curves) based on the path-independent J Integral. These curves are more often obtained from laboratory tests with small specimens as CT-specimens and their application to large component safety analysis could be questionnable. Indeed, it is well known that J-R curves could depend on the specimen size and on the loading mode (i.e. bending stress versus tensile stress) but this dependency could be different from one material to another. This means that it would depend not only on the stress-strain state but also on the actual local fracture mechanisms (i.e. the damage) occuring before the crack initiation or during the crack propagation.

The purpose of this paper is to gather some results of crack growth resistance measurement studied at EDF with different materials in order to show how the effect of the parameters, as specimen geometry and mode of loading, is directly related to the local fracture mechanisms or the microstructure of the materials. For that a number of results are analysed by means of the local approach of fracture which is a very useful tool to predict quantitatively the J-R curve dependency, related to fracture mechanisms.

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J RESISTANCE CURVES OF PRESSURE VESSEL STEEL

Several experimental programs on large scale specimens were organized to evaluate capabilities of the fracture mechanics concepts employed in structural integrity assessment of PWR pressure vessels. Most of them aimed at investigating the upper shelf toughness fracture behaviour of low alloyed steels, and to assess the validity of the J-integral and J-resistance curve concepts regarding to ductile crack propagation.

One of the thermal shock experiments performed at MPA-Stuttgart (NKS3 test) has been analysed by Bethmont et al (1). This test was carried out on large hollow cylinder (thickness = 200 mm and inside radius = 200 mm) containing a circumferential crack at inside surface. After the thermal shock (between 300°C and room temperature), the crack grows in ductile regime up to 3.5 mm. The comparison between experimental crack growth measurements and numerical predictions from J-resistance curves, determined on CT specimens, is satisfactory. That means that no "specimen effect" is observed on J-resistance curves. This can be easily explained by the similarity of the stress state at crack tip in the large cylinder and in CT-specimen. For this test, the local approach method has been applied, based on the Rousselier's method which simulates the growth of cavities characterizing the ductile damage. The results of this method give also good prediction of crack growth in comparison with the experimental results (1).

Another study concerns the first spinning cylinder test, performed by AEA Technology. This test involved the rotation of a 200 mm thick cylinder (fabricated from a modified A508 Cl 3 steel) containing a full-length axial flaw to an angular speed of 2600 rpm at the temperature of 290°C (Eripret and Rousselier (2)). This test aimed at generating ductile crack growth by increasing progressively the rotational speed, that created a membrane hoop stress across the thickness. The first objective of this test was to provide experimental data that would permit the construction of a J-resistance curve. A geometry effect (thickness, load, or size effect) was experimentally pointed out and exhibited the problem of transferability of toughness data from small scale to large scale specimens (figure 1). An analysis of this test, by means of local approach to fracture based on the Rousselier's model, has been carried out. This modelling is able to account for local effects of the crack area loading factors, such as stress triaxiality. The prediction of crack growth by this model is in a good agreement with experimental results (figure 1). As the stress triaxiality around the crack tip is larger in the CT specimen than in this hollow structure (figure 2), the material damage (i.e. the cavity growth) increases earlier and the steel resistance to ductile tearing is lower in a CT specimen.

J RESISTANCE CURVES OF C-Mn STEELS OF PIPING

The general use of 1T-CT specimens, in laboratories, comes from previous study results which have shown that the measures are generally conservative as the specimens are thinner than the components - see for example De Roo et al (3) -. However, Hiser and Terrel (4) have shown different results : the lowest J resistance curves are obtained from largest specimens of 100 mm thickness. The investigated material, a A302-B steel, presents a high sulphur content and a low Charpy upper shelf energy. Furthermore, such equivalent results are also presented by Roos et al (5).

Some C-Mn steels of PWR secondary piping components, present some characteristics (sulphur content and Charpy upper shelf) quite similar to those of the material of the study described in (4). So, a study has been undertaken at EDF in order to verify if and why a dimension effect may be observed in the case of these steels of relatively low upper shelf energy.

Tests were conducted on 1/2T-, 1T- and 2T-CT specimens. The test results show that the J0.2 values (J value corresponding to the crack onset from French procedure) deduced from the different specimens are similar. But the propagation resistance of the ductile crack in 2T-CT specimens is lower than in 1/2T- and 1T-CT specimens (figure 3). The fracture surfaces of the different specimens display some splits which are perpendicular to the crack plane and parallel to the plate surface (figure 4). These splits are produced by the separation of the manganese sulphur inclusions. The effect observed on the J resistance curves seems to be relevant to these splits. These results are similar to those presented in (4). It seems that splits, initiated along manganese sulphur inclusions by the through thickness stress (roughly estimated up to 1000 MPa), would modify strongly the local stress-strain state and consequently the resistance to crack growth of the material. A study is in progress to investigate the effect of the splits by the local approach of fracture which could take into account not only the typical ductile damage (i.e. cavity growth) but also split initiation perpendicular to the crack front line.

J RESISTANCE CURVES OF CAST DUPLEX STAINLESS STEELS

The previous example indicates that the local fracture mechanisms (or the material microstructure) is a very important factor which leads to different effects of the specimen geometry on J-resistance curves.

A study was undertaken on an aged duplex stainless steel with a high ferrite content (about 20%), which makes it very sensible to thermal aging effects. This material with coarse microstructure (grain size of about 1 mm) causes a large scatter in material response. The mechanisms of damage preceding the cracking are quite different from typical ductile fracture : cleavage cracks take place in the ferrite phase at the onset of the damage. Then the tearing of the austenitic phase leads to the macrocracking.

A J-resistance curve derived from CT specimens has been compared to the resistance curve from a pressurized pipe burst test (internal and external diameters are respectively 300 and 400 mm) with the same material at room temperature (figure 5). A sub-surface external defect was machined then precracked in the pipe before the test. During the internal pressure increasing, the crack initiates and grows through the thickness of the pipe until the rupture of the pipe. By measurement of the stable crack growth (by electrical potential drop method) it has been possible to determine the J-resistance curve of the pipe material.

It seems still, even if the scatter is large with this material, that the geometry effect could explain the difference between the curves.

In order to simulate this effect, a first tentative of a numerical simulation based on the assumptions of pure ductile fracture (i.e. only cavity growth) has been carried out by Bethmont et al (6). The results showed that this model was not appropriate for this material, particularly the experimental J-resistance curve from CT specimens could not be correctly fitted with the numerical simulation (figure 6).

Some investigations conducted to develop new models which have to take into account the actual mechanisms of fracture.

Such a model has been developed by Eripret and Sun (7) simulating the crack initiation and also the crack propagation. Initiation and propagation occur quite naturally when the softening effect, due to microcracking and growth of microcracks, becomes greater than hardening due to plasticity of the austenitic phase. The results are attractive : the J-resistance curve is correctly simulated (figure 6). The first evaluation of the geometry effect has been carried out by comparison of numerical J-resistance curves from CT and CCT specimens. The study is now in progress to evaluate the J-resistance curve from pressurized pipe test.

The experimental results show that for such a material, the very large scatter comes from the damage mechanisms. The macrocracking depends on the probability to break by cleavage the ferrite phase. This statistical aspect of fracture of duplex stainless steel has been investigated by Pineau and Joly (8).

J RESISTANCE CURVES OF WELDED JOINTS

A typical problem for integrity assessment of components is the evaluation of the ductile resistance of welded joints. In such structures, the inhomogeneity of the materials introduces large variations of the local mechanical properties.

A great number of studies have been undertaken in reference with the "mismatch" effect. A typical stucture, investigated at EDF, is the bimetallic welds made of ferritic and austenitic materials.

The ductile fracture resistance of austenitic material is high in case of specimens containing only one material but in case of bimetallic specimens, the resistance of the same austenitic material could become lower (see Table 1).

It appears clearly, that the interpretation of the tests needs finite elements calculation to model the crack initiation and propagation. The experiments showed that the fracture mechanism responsible for ductile tearing was the growth of small cavities initiated on oxide inclusions in the austenitic weld material. Thus, the study was undertaken with the model developed by Beremin (9) linking ductility of the material to its microstructure through a cavity growth law. The growth of cavities is related to stress triaxiality and plastic strain. Failure is assumed to occur when the cavity growth reaches a critical value. The mismatch effect, i.e. the way the harder ferritic material restricts the development of plasticity from the crack tip to the structure, was investigated for a normally diluted bimetallic weld, as well as for a highly diluted weld which enhance the mismatch effect by creating an austenitic martensitic zone which concentrates even more plasticity in the softer austenitic weld.

The first numerical simulation of toughness tests were performed at EDF by Chas (10), - similar to the previous work described by Devaux et al (11)-, on bimetallic ferritic-austenitic weld and have been conducted to investigate the fracture behaviour of a crack located in the softer austenitic material and close to the bimetallic interface. The main goal of this work was to investigate how the overall toughness of the bimetallic weld could be affected by the mismatch between austenitic and ferritic steel, and how the shear stresses generated near the interface could affect the stress triaxiality.

The different results obtained by applying the local approach to fracture model are gathered in the Table 1. They are in good agreement with experimental results.

TABLE 1 - Comparison of experimental and numerical $J_{0.2}$ values of bimetallic welds

J 0.2 (kJ/m2)	Austenitic material	Bimetallic weld (Normal dilution)	Bimetallic weld (High dilution)
Experimental result	98	54	32
Numerical result	100	61	28

These results show that undermatching can affect the overall toughness of the welds (Rp0.2 = 450 MPa for ferritic and 350 MPa for austenitic materials), according to the enhanced concentration of plastic strain at the crack tip and with respect to the rise of stress triaxiality. It depends on the way by which the surrounding harder materials will restrict the plasticity development pattern when compared to the homogeneous material configuration (figure 7).

Moreover, the local approach to fracture model give informations about the possible crack path deviation in bimetallic weld, showing that the crack would rather propagate in the softer austenitic material for a highly diluted weld, while it would probably jump to the bimetallic interface for a normally diluted weld. Those informations are also in good agreement with the experimental observations.

This study will be carried on a numerical simulation of ductile crack resistance of welds with different geometry and mode of loading.

J RESISTANCE CURVES FOR LARGE CRACK PROPAGATION IN PIPES

Bending experiments have been conducted on carbon and stainless steel straight pipes containing circumferential machined cracks. The purpose of the program was to develop a better understanding of pipe fracture behaviour in order to evaluate the leak-before-break (LBB) approach and to improve in service flaw assessments.

The experiments were carried out on 6" (168 mm) and 16" (406 mm) diameter pipes containing through-wall cracks, with total angles ranging from 30 to 120°. All the tests were performed at 300°C, in four point bending. The determination of crack initiation was done using the d-c electric potential drop method whereas crack propagation was estimated using the marks made by occasional unloadings. During these tests it was observed that large propagation occurs between initiation and maximum load, with crack turning out from the original crack plane.

Three-dimensional finite element analyses of seven pipe experiments were performed. The calculations were carried out with load-line displacement and measured crack growth (projected crack length), as input. The crack extension was simulated by successive meshing of the pipe. The energy release rate G was calculated using the THETA method, developed by EDF which has been demonstrated as equivalent to the J parameter derived from Parks method.

A predictive analysis aimed to calculate the maximum moment must take into account the crack extension. However the large amounts of crack propagation observed at the maximum moment (10 to 30 mm) greatly exceed the capacity of standard CT specimens (2 to 3 mm). Furthermore, the size of the specimens is limited by the curvature of the pipe wall. Three solutions are proposed to derive a J-R curve for large crack extension values. The first is to calculate J directly from the pipe test results, but it is no longer a predictive method ! The second is to extrapolate results from CT specimens with a conservative method : i.e. reducing the slope of the J-R curve for large crack extension values. The third solution is to construct the curve by using the G values calculated by the FE method together with the experimental crack growth, for the same load-line displacement. Figure 8 presents an application of this method. The comparison between laboratory specimen and pipe results is limited to the crack initiation and to the first millimeters of ductile crack growth. Nevertheless, a fair agreement is obtained on this limited range of propagation.

This result clearly shows that if the dimension (thickness) and the mode of loading (bending) are similar, the J resistance curves are identical. In this case, CT specimens can be used to predict crack initiation. But, the evaluation of the crack propagation beyond few millimeters is not possible : the resistance to crack propagation has to be directly determined on pipes.

TOUGHNESS IN THE TRANSITION DUCTILE/BRITTLE REGIME

The integrity of some components has also to be demonstrated at the temperature near the ductile-brittle transition temperature. In this case, the crack initiates and propagates in ductile regime followed by cleavage conducting to a brittle fracture.

Some studies have been carried out in view to evaluate the influence of such parameters as specimen geometry and loading mode on the toughness of materials in the transition regime.

The modelling of this type of fracture has been performed by Eripret et al (12) by means of a coupled model in the framework of the local approach of fracture : the ductile damage is simulated by Rousselier model and the cleavage by Weibull model.

The first results of the study are concerning on one hand the influence of prior ductile tearing on cleavage initiation, and on the other hand, the effect of size of the specimens on toughness properties. It appears clearly that larger is the specimen sooner occurs cleavage (figure 9). By means of local approach to fracture modelling, it was demonstrated that a lower bound of toughness properties could be predicted and then compared to the design transition curve. Safety margins have been exhibited.

CONCLUSION

Structural integrity assessment of PWR components, as pressure vessel and piping, needs the evaluation of the J-resistance curves of the materials. These curves could depend on the specimen size and on the loading mode (i.e. bending stress versus tensile stress).

The results gathered in this paper show that these effects depend on the material behaviour (i.e. on its local mechanisms of fracture) and will be different if the fracture is only dependent of cavity growth or if it includes some other mechanisms as cleavage.

It is shown that the specimen size and the loading mode effects on J-resistance curves can be explained and evaluated by the local approach of fracture which models the actual local damage in relation with the local stress-strain state.

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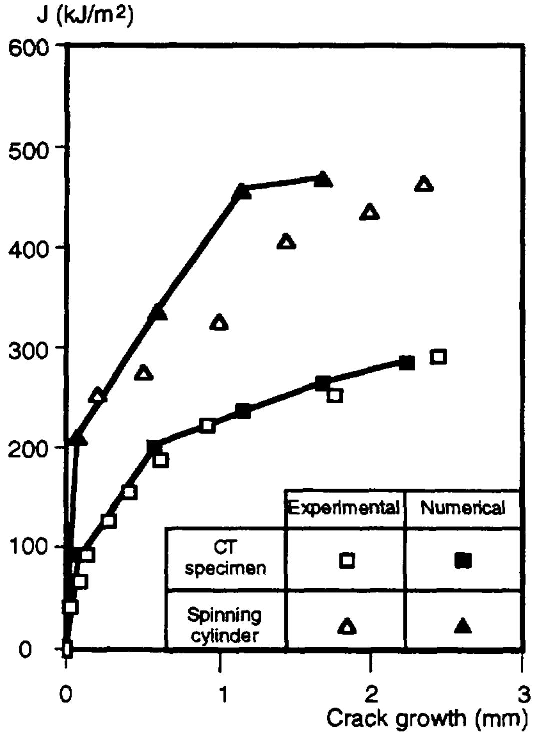

Figure 1 : Comparison of J-resistance curves derived from first spinning cylinder and CT-specimen tests.

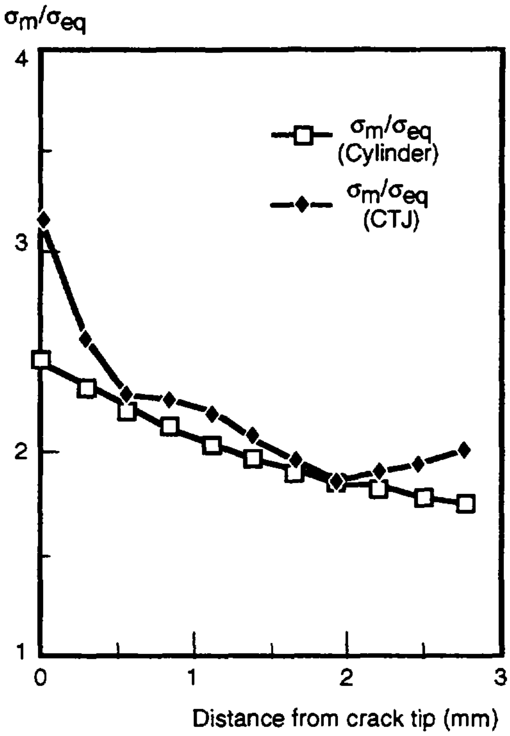

Figure 2 : Comparison of stress triaxiality at crack tip in the first spinning cylinder and CT-specimens.

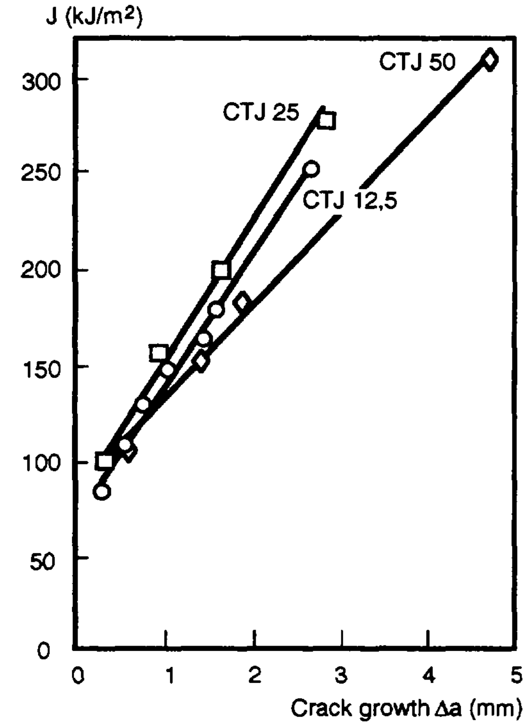

Figure 3 : Comparison of J-resistance curves of low upper shelf energy C-Mn steel obtained with 1/2T-. 1T- and 2T-CT specimen.

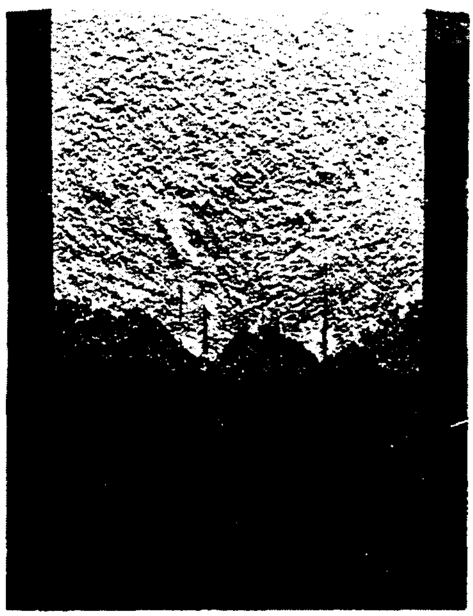

Figure 4 : Fracture surface of C-Mn steel.

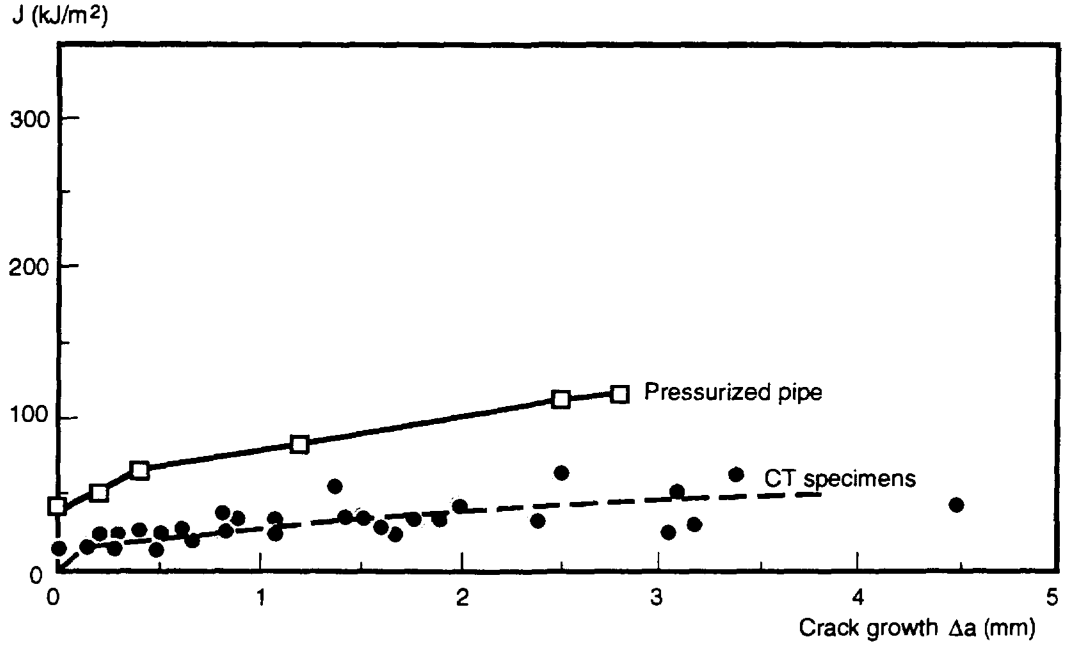

Figure 5 : Comparison of experimental J-resistance curves of aged duplex stainless steel obtained with burst test pipe and CT specimen.

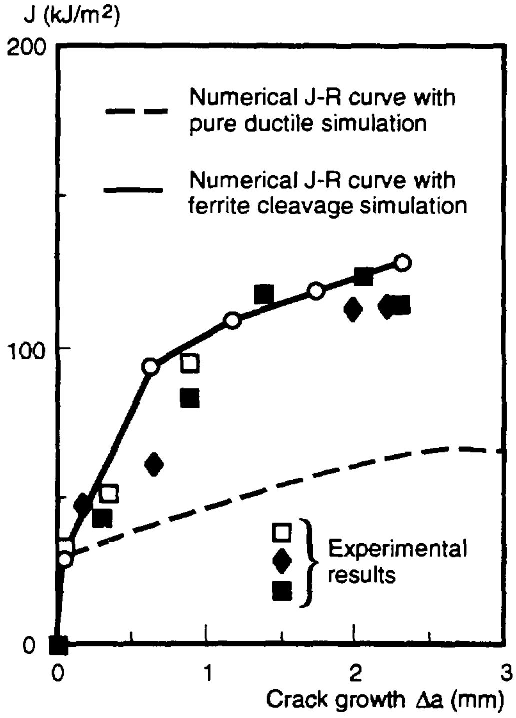

Figure 6 : Comparison of experimental and numerical J-resistance curves of aged duplex stainless steel.

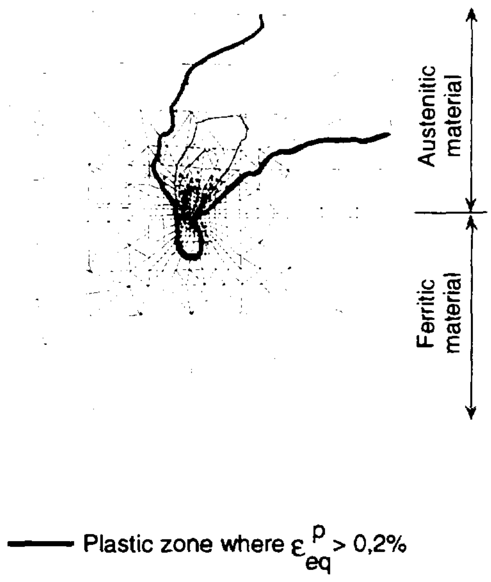

Figure 7 : Plastic zone extension at crack tip in a bimetallic CT specimen.

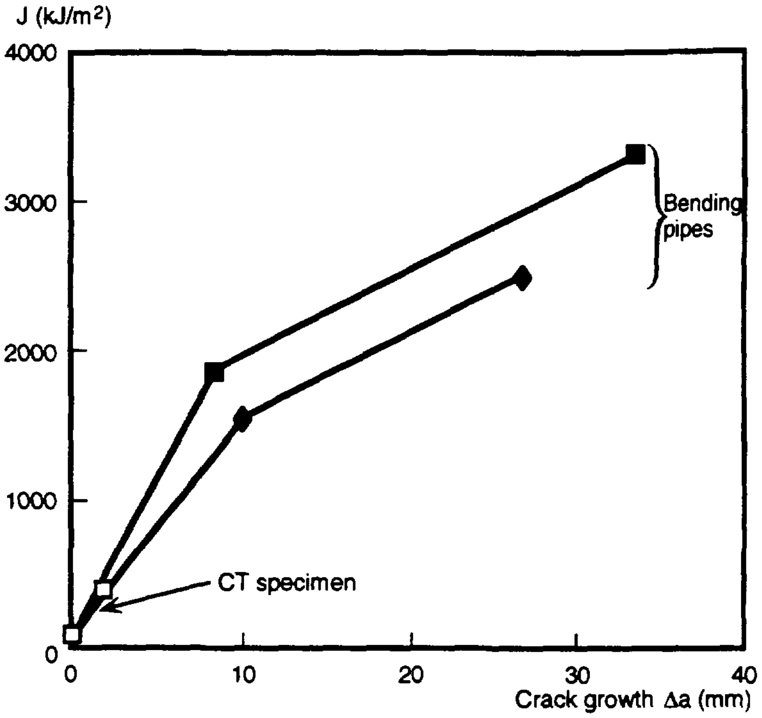

Figure 8 : Comparison of J-resistance curves of A42 carbon steel obtained with bending pipes and CT specimens.

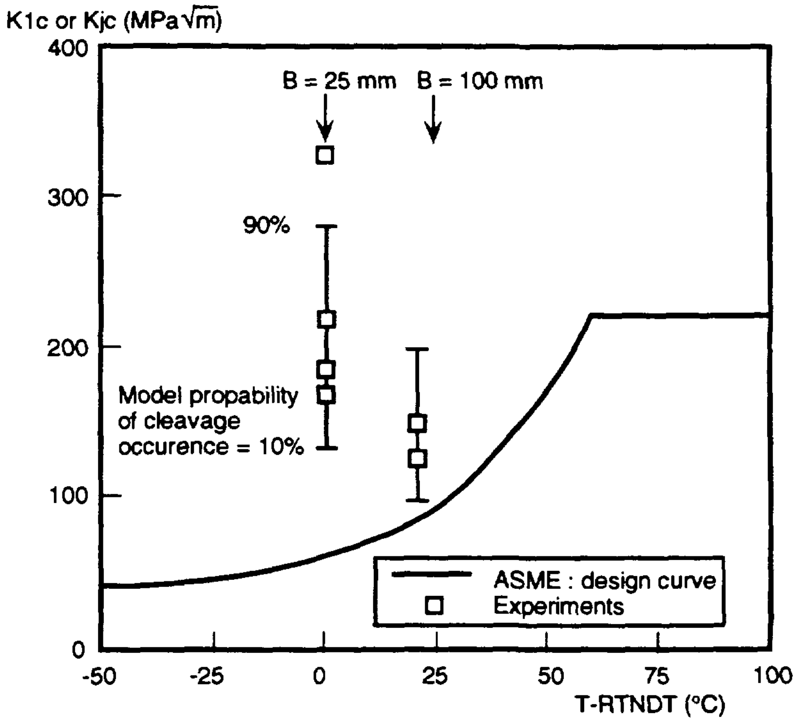

Figure 9 : Comparison of the experimental results and prediction of the toughness of A508 Cl 3 steel in the transition regime.