TEMPER EMBRITTLEMENT SUSCEPTIBILITY AND TOUGHNESS OF A 508 CL 3 STEEL
Il a été montré dans une étude précédente, que la susceptibilité à la fragilité de revenu des aciers forgés A508C1 3, utilisés dans la fabrication des enceintes des réacteurs à eau pressurisée, est en relation avec la teneur en impureté, et particulièrement du phosphore. Cette étude a été étendue ici pour prendre en compte d'autres aciers d'enceintes sous pression, et en particulier un matériau d'enceinte présentant une dispersion significative de ses propriétés de ténacité.

Une corrélation a été observée entre la dégradation de la résilience et des propriétés de ténacité et l'apparition de rupture intergranulaire sur la surface de rupture. Il a été trouvé que la rupture intergranulaire s'accroissait sous l'influence d'un vieillissement pour des températures comprises entre 500 et 600 °C. La rupture intergranulaire et la fragilité de revenu associée sont réduites ou éliminées après une exposition à une température légèrement plus élevée (640-650 °C), confirmant la réversibilité du phénomène. Des examens en spectroscopie Auger menés sur des échantillons présentant des ruptures par clivage et intergranulaire ont montré dans ce dernier cas la ségrégation du phosphore au joint de grain, mais également celle du soufre.

Dans le but de caractériser la dispersion de la ténacité ($k_I$ ou $k_{JC}$), une analyse statistique a été appliquée aux données obtenues à -50 °C, en utilisant le modèle de Weibull. Les résultats qui ont été obtenus montrent que la probabilité de rupture cumulée des éprouvettes présentant une rupture par clivage ont une dépendance en puissance 4 de leur ténacité, alors que celles présentant de la rupture intergranulaire ont une dépendance en puissance 8 de leur ténacité.
TEMPER EMBRITTLEMENT SUSCEPTIBILITY AND TOUGHNESS OF A508-CLASS 3 STEEL

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ABSTRACT: A correlation has been established between the scatter in charpy and fracture toughness results of an A508 CL3 pressure vessel forging steel and the appearance of intergranular facets on the rupture surface of specimens tested at low temperatures. Extent of the intergranular rupture has been found to increase with prolonged aging at 500-550°C or following an embrittling treatment known as "step cooling". Both, intergranular rupture and resultant increase in ductile to brittle transition temperature have been found to be reduced or eliminated after a de-embrittling treatment at temperatures ranging from 630°C to 660°C, with optimum results obtained at 640°C. Auger analyses carried out on specimens exhibiting cleavage and intergranular rupture have shown segregation of carbon, molybdenum, sulfur and in particular phosphorus at the grain boundaries.

A statistical analysis based on local rupture criterion is used to describe probability of intergranular and cleavage failures. Based on the observations made it is concluded that the main cause of scatter in low temperature results is heterogeneous segregation of impurities in thick sections with a corresponding shift in ductile to brittle transition temperature. Slight increase in austenitization and de-embrittling temperatures are suggested in order to reduce segregation and improve toughness.

KEY WORDS: pressure vessel, forging, plate, pressurized water reactor (PWR), A508-Class 3, A533B, temper embrittlement, residual elements, segregation, cleavage, intergranular, charpy, $K_{IC}$, $K_{JC}$, toughness, axisymmetric specimens, auger spectroscopy.
1. INTRODUCTION

More recent pressurized water reactor (PWR) vessel designs have incorporated extensive use of large manganese-molybdenum-nickel steel ring forging (ASTM A508 CL3).[1-4] This is due to the fact that such designs offer a number of advantages over the conventional plate (A533 B) designs. Specifically they allow elimination of longitudinal seam welds which in turn improves structural integrity and eases in service inspection. However, with increased size of the initial ingot (several hundred tons) and the subsequent slow cooling imposed by thick sections, comes inevitably the risk of micro- and macrosegregation of some elements with an associated variation in mechanical properties [5-10]. For instance, the steel's susceptibility to temper embrittlement could increase due to local segregation and build up of nonmetallic residual impurities, such as P, As, Sn and Sb, at the grain boundaries [5-12]. As a result control of minor and residual element concentrations and their distributions has a significant effect on the mechanical behavior of pressure vessels.

In an earlier work we showed that microsegregation of residual elements, specially phosphorus, was at the origin of reversible temper embrittlement susceptibility of A508 cl3 [5]. It was also shown that the phenomenon was more marked in the coarse-grained simulated heat affected zones and that it could be deliberately accentuated through an embrittling treatment known as "step cooling". Its main consequences were reported to be a rise in the ductile to brittle transition temperature (DBTT) along with a change in the low temperature fracture mode from cleavage to intergranular. The reversible nature of the phenomenon (RTE) was demonstrated by reheating above the critical temperature range (>600°C) and rapidly cooling. This treatment produced a decrease in the DBTT, together with a reversion in the low temperature fracture mode from intergranular to cleavage. For more information with regards to Reversible Temper Embrittlement (RTE) and the effect of various parameters involved, including chemical composition and microstructure, the reader is referred to a recent review carried out by Eyre et al [9].

In the present work we have examined the role of residual elements in a forged vessel steel (A508 CL3) exhibiting significant scatter in charpy and fracture toughness results. We have paid a particular attention to the rupture mode and have discussed the fracture toughness results using a statistical analysis in conjunction with local criterion for rupture [13-16].

2. MATERIALS AND PROCEDURES

2.1. MATERIALS

The main material investigated is an A508 CL3 type steel taken from a nozzle cut-out of a PWR vessel forging (diameter 1420 mm and thickness 270 mm). Also investigated is a commercial A533B plate which is a similar type steel.

Nominal chemical composition of the two steels are presented in table 1. Both steels satisfy the corresponding ASTM specifications and have low residual element concentrations. Local variations of chemical composition, as obtained with emission spectrometry analysis carried out on several specimens and also on through thickness samples, are shown in table 1. Apart from the arsenic concentration which may be considered high (260-300 ppm) all other values are within the specified limits and do not exhibit a marked variation.

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1 This study is part of an AIEA sponsored program in which a large number of commercial PWR vessel nozzle cut-outs have been investigated.
2.2. HEAT TREATMENTS

The forged steel was tested in several conditions. They are (see also figure 1):

a) initial state; austenitized at about 875°C and then quenched in water,

b) step-cooled; an embrittling treatment voluntarily applied to intensify RTE phenomenon: 1h at 600°C - 15h at 540°C - 24h at 525°C - 48h at 500°C - 72h at 470°C,

c) aged state; 2000h at 500-550°C,

d) de-embrittled state; about 1 hour at 620 to 660°C followed by rapid cooling,

e) reaustenitized at 900°C.

The plate material was investigated only in the quenched and step-cooled states. The main difference between the two types of product is, apart from their size, a slightly higher austenitization temperature (~900°C) used for the plate.

2.3. SPECIMENS AND MECHANICAL TESTS

Two types of specimen were used; standard charpy-V and standard compact tension (CT 25). In addition a few conventional tensile specimens and axisymmetric notched bars (Ø = 18 mm), with different radii (2, 4, and 10 mm) were tested. The main purpose of these tests was to plastify a larger volume of the material but also to obtain necessary data used in local rupture criterion approach.

Mechanical tests were carried out at temperatures ranging from +100 to -100°C, but the bulk of tests were performed at -50°C. In general conventional properties were recorded or calculated after each test. However, instead of $K_{IC}$ we use $K_{JC}$ in order to account for plastic deformation ($P_{max}/P_Q > 1$). of some specimens. $K_{JC}$ was calculated in a similar manner as $J^2$.

2.4. FRACTOGRAPHY

Fractographic observations were carried out using a conventional scanning electron microscope (SEM).

Auger electron spectroscopy (AES) analyses were conducted using an ASC 2000 type microscope, model RIBER, operating in differential mode $dN(E)/dE$. It was equipped with a cylindrical mirror analyser and allowed in-situ ion sputtering and high vacuum (2 to 3. $10^{-10}$ Torr) fracture testing. Primarily, the analysis was restricted to intergranular facets (low toughness specimens), although cleavage facets were also examined for comparison. All the examinations were conducted on small size samples prepared from CT and charpy specimens and broken inside the microscope, except one charpy specimen whose fracture surface was entirely examined as broken in air. In the latter case, about 20 Å of the surface was removed with ion sputtering in order to eliminate surface contamination.

3. RESULTS

Figures 2 and 3 show two sets of low temperature (-50°C) charpy and fracture toughness results obtained with the specimens taken from the forged steel. In both cases the results can be divided into two groups, one which is situated in the normal range and another which is situated in the low toughness range. Fractographic examinations carried out on these specimens revealed that in both

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2 The terms $K_{IC}, K_{JC}$, and $J$ are used in fracture mechanics and are usually expressed in MPaVm or $KJ/m^2$. $P_{max}$ is the maximum (elastic+plastic) load recorded during test while $P_Q$ is the maximum elastic load.
situations the low toughness values were associated with appearance of intergranular facets, see for instance figure 4.

Specimens tested after step-cooling also exhibited a scatter in toughness (figure 5), but the difference between the two levels was less pronounced. Here also the low values coincided with appearance of intergranular facets. However, the most significant effect of step-cooling was a shift of about 20-30°C in DBTT, figure 6.

Prolonged aging at 550°C also resulted in similar effects with occasionally giving rise to significantly low toughness values, \( = 1.5 \). De-embrittling treatment, carried out before or after step-cooling or aging, resulted in restoration of toughness and a net decrease in DBTT (fig.6) and in elimination of intergranular facets. Optimum results were obtained after treatment at 640°C, figure 7. Furthermore reaustenitization at 900°C followed by a temper at 640°C gave the forging improved toughness which remained elevated after step-cooling or aging, Table 2. The benefic effect of reaustenitization at 900°C was such that tempering at a lower temperature (625°C) followed by prolonged aging at 500°C did not induce severe embrittlement as it was the case with initial austenitization at 875°C (table 2).

There was not a significant difference between tensile properties of step-cooled and de-embrittled smooth specimens, except a slight increase in Y.S. and UTS due to step-cooling (fig. 8), however, some notched specimens exhibited a significant reduction in rupture elongation and the average cleavage stress to rupture in the quenched and embrittled conditions, table 3. A533B plate steel in general exhibited a superior toughness at -50°C than A508 CL3 forging and was less sensitive to embrittlement, table 4.

Auger analyses showed traces of carbon, oxygen, chlore (probably due to contamination) and molybdenum on the fracture surfaces, but most of all they showed a significant segregation of phosphorus in the zones exhibiting intergranular rupture (fig. 9). Traces of sulfur were also detected on the surface of the specimen broken in air. These traces did not disappear until sputtering was repeated for several hundred Å. In this case also the segregation was more pronounced at the grain boundaries than at the cleavage facets. For instance the average ratio of \( S(152 \text{ ev})/\text{Fe}(631 \text{ ev}) \) peak at the grain boundaries were on the order of three times higher than those at the cleavage facets.

4. DISCUSSION

A simplified explanation to the observed large differences in toughness values of A508 CL3 specimens can be given in terms of local variations in transition curves and the changing mode of rupture. Indeed, charpy specimens taken from regions with more pronounced segregation had a higher DBTT and hence when tested at -50°C broke mainly, if not fully, in fragile mode. This was also confirmed by fracture toughness results where \( K_{IC} \) was valid (\( K_{IC} = K_{JC} = P_{\text{max}}/P_{\text{Q}} \geq 1 \)). Charpy specimens taken from regions with less segregation had a lower DBTT and therefore when tested at -50°C broke with some plastic deformation. Likewise for these specimens \( K_{IC} \sim K_{JC} \) (\( P_{\text{max}}/P_{\text{Q}} \geq 1 \)). On this basis one may assume that if tests were carried out at higher or lower temperatures than -50°C (more precisely in the upper and in the lower shelves) the above mentioned scatter will disappear.

In practice, some segregation is likely to occur in large ingots and thick sections, although its extent can be reduced by faster cooling, or using slightly higher austenitization and de-embrittling temperatures. Therefore, one would like to predict the probability of brittle fracture independent of its
nature, being inter or transgranular. Here below we apply Weibull's statistical analysis to our fracture toughness data after a brief recall of Beremin's local criterion for failure [13].

The notion of critical cleavage stress (see review in [17]) was first proposed by Griffith:

$$\sigma_c = \sqrt{E\gamma m(1-\nu^2)}l_0$$

Here $E$ is Young's modulus, $\gamma$ surface energy, $\nu$ poisson's ratio and $l_0$ length of a microcrack, considered to be coherent with metallurgical features (grains or carbides). Later on Knott [18] and Tetelman [19] calculated $\sigma_c$ using the slip line field theory. Richie, Knott and Rice [20] subsequently predicted $K_{JC}$ values by assuming that the critical stress is reached at a distance of $\lambda_0$ from the fissure. Finally local rupture criteria approach was developed in which $\lambda_0$ substituted by bulk analysis.

In this approach the stressed volume of the metal, $V$, is divided into $n$ smaller volumes $V_0$, chosen to be on the order of several grains. The probability of failure $P_R$ of the volume $V$ is defined as that of one of the $V_0$. In its simplified form the expression derived is [13]:

$$P_R = 1 - \exp \left( - \frac{\sigma_w}{\sigma_u} \right)^m$$

where $\sigma_w$ is Weibull's stress$^3$ and $\sigma_u$ a material constant found to vary with grain size. It can be seen that the probability of rupture will be $=63\%$ when $\sigma_w$ reaches $\sigma_u$. A similar equation has been proposed by Beremin [13] with $\sigma_w$ and $\sigma_u$ being replaced by toughness values.

We have used as a first approximation Beremin's proposed equation for cleavage fracture of nuclear pressure vessel steel. The results obtained are presented in fig. 10. Here, the curve predicting the probability of cleavage fracture ($P_R$) is given by:

$$P_R = 1 - \exp \left( - \frac{K_{JC}}{K_{UC}} \right)^4$$

where $K_u$ is material's constant. It is independent of temperature and its value is taken to be $=125$ MPa$\sqrt{m}$. Experimental points, corresponding to the 29 fracture toughness tests carried out at $-50^\circ C$ are also shown in this figure. The probabilities of these tests are assumed to be $1/29$ for lowest toughness value, ..., and $29/29$ for highest toughness value.

It can be seen that by taking $K_u=125,3$ MPa$\sqrt{m}$ the theoretical curve obtained provides a good prediction of results in the upper region but fails to follow low $K_{JC}$ values. For this reason we then divided our results into two groups according to their mode of rupture. Although the number of tests with intergranular facets is small (9 tests) the results can be satisfactorily represented if the exponent is taken as equal to 8 (fig. 10):

$$P_{R1} = 1 - \exp \left( - \frac{K_{JC}}{K_{UC}} \right)^8$$

$K_{UL}$ is calculated to be about 73 MPa$\sqrt{m}$.

Likewise the 20 tests with cleavage facets can be represented by:

$$P_{R2} = 1 - \exp \left( - \frac{K_{JC}}{K_{UC}} \right)^4$$

with $K_u = 137$ MPa$\sqrt{m}$.

Finally one may combine the two equations to represent the probability of rupture in all the population:

$$P_R = 1 - \left[ \hat{\xi} \exp \left( - \frac{K_{JC}}{K_{UC}} \right)^8 + (1-\hat{\xi}) \left( \frac{K_{JC}}{K_{HC}} \right)^4 \right]$$

In this equation $\hat{\xi} = 9/29 = 0.31$. The equation derived is similar to that of Service et al [21] which has been used for analyzing brittle fracture of ceramic materials with two types of defects. However it

$^3$ Calculated through finite element analysis and whose main variables are deformation and maximum principal stress in each $V_0$. 
should be pointed out that in our case this assumes that only one of the two modes of rupture, intergranular or cleavage, occur at one time, while in most cases both modes could occur simultaneously.

5. CONCLUSIONS

Forgoing results clearly demonstrate that the principal cause of the scatter in toughness of the forged steel is local variations in DBTT with a change in low temperature rupture mode from cleavage to intergranular. The mechanism involved is reversible temper embrittlement and is provoked by heterogeneous segregation of residual elements, in particular P, at the grain boundaries. A slight increase in the austenitization temperature, from 875°C to 900°C, and/or the de-embrittling temperature, from about 630°C to 640°C, can significantly reduce temper embrittlement susceptibility of the steel and consequently reduce variations in toughness.

7. ACKNOWLEDGEMENTS

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8. REFERENCES


7


TABLE 1. NOMINAL CHEMICAL COMPOSITIONS OF PLATE AND FORGING STEELS AS WELL AS LOCAL VARIATIONS IN THE LATTER.

<table>
<thead>
<tr>
<th>STEEL</th>
<th>C</th>
<th>S</th>
<th>P</th>
<th>Si</th>
<th>Mn</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
<th>Cu</th>
<th>V</th>
<th>Nb</th>
<th>Vn</th>
<th>As</th>
<th>Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>PLATE (A533B)</td>
<td>0.23</td>
<td>0.002</td>
<td>0.008</td>
<td>0.23</td>
<td>1.46</td>
<td>0.66</td>
<td>0.03</td>
<td>0.51</td>
<td>0.04</td>
<td>0.002</td>
<td>0.005</td>
<td>0.005</td>
<td>0.019</td>
<td>0.028</td>
</tr>
<tr>
<td>FORGING (A508 CL3)</td>
<td>0.15</td>
<td>0.009</td>
<td>0.009</td>
<td>0.24</td>
<td>1.36</td>
<td>0.68</td>
<td>0.24</td>
<td>0.46</td>
<td>0.063</td>
<td>0.008</td>
<td>0.01</td>
<td>0.008</td>
<td>0.022</td>
<td>0.031</td>
</tr>
</tbody>
</table>

SPECIMENS TAKEN FROM THE FORGING:

- **504**: C=0.15, S=0.007, P=0.009, Si=0.26, Mn=1.36, Ni=0.76, Cr=0.26, Mo=0.49, Cu=0.066, V=0.012, ≤0.002, ≤0.007, ≤0.024, ≤0.032
- **505**: C=0.15, S=0.008, P=0.009, Si=0.26, Mn=1.37, Ni=0.78, Cr=0.27, Mo=0.49, Cu=0.068, V=0.012, ≤0.002, ≤0.006, ≤0.026, ≤0.040
- **(1/4)**: C=0.16, S=0.007, P=0.010, Si=0.27, Mn=1.38, Ni=0.78, Cr=0.27, Mo=0.49, Cu=0.068, V=0.012, ≤0.002, ≤0.006, ≤0.026, ≤0.037
- **(1/2)**: C=0.16, S=0.009, P=0.010, Si=0.28, Mn=1.39, Ni=0.79, Cr=0.28, Mo=0.50, Cu=0.069, V=0.013, ≤0.002, ≤0.006, ≤0.026, ≤0.037
- **610**: C=0.18, S=0.008, P=0.011, Si=0.29, Mn=1.45, Ni=0.81, Cr=0.29, Mo=0.52, Cu=0.072, V=0.014, ≤0.002, ≤0.007, ≤0.030, ≤0.037
- **618**: C=0.18, S=0.007, P=0.011, Si=0.30, Mn=1.48, Ni=0.83, Cr=0.29, Mo=0.52, Cu=0.073, V=0.015, ≤0.002, ≤0.009, ≤0.031, ≤0.034

† N2<0.009 and Cu<0.073

TABLE 2: Effect of reaustenitization and temper on fracture toughness of A508-Class 3 forging.

<table>
<thead>
<tr>
<th>SPECIMEN</th>
<th>TEMP.℃</th>
<th>KIC (MPa)</th>
<th>HEAT TREATMENT</th>
</tr>
</thead>
<tbody>
<tr>
<td>607</td>
<td>-50</td>
<td>265</td>
<td>reaustenitized at 900℃ → quenched in oil → tempered at 640℃ → step-cooled.</td>
</tr>
<tr>
<td>606</td>
<td>-50</td>
<td>320</td>
<td></td>
</tr>
<tr>
<td>752</td>
<td>-50</td>
<td>127.3</td>
<td>reaustenitized at 900℃ → quenched in oil → tempered at 625℃ → aged 2000 h at 500℃.</td>
</tr>
<tr>
<td>758</td>
<td>-50</td>
<td>139.7</td>
<td></td>
</tr>
<tr>
<td>755</td>
<td>-50</td>
<td>103</td>
<td></td>
</tr>
<tr>
<td>753</td>
<td>-50</td>
<td>130.5</td>
<td></td>
</tr>
</tbody>
</table>

TABLE 3: Effect of step-cooling on average stress and strain to rupture, as well as the critical fracture stress, of notched axisymmetrical specimens.

<table>
<thead>
<tr>
<th>NOTCH / SPECIMEN TEMP.</th>
<th>σR MPa</th>
<th>σs MPa</th>
<th>σc MPa</th>
<th>σc (corrected)</th>
</tr>
</thead>
<tbody>
<tr>
<td>2 mm (quenched)</td>
<td>-50</td>
<td>1384</td>
<td>41</td>
<td>1633</td>
</tr>
<tr>
<td>2 mm / 530</td>
<td>-50</td>
<td>1261</td>
<td>27.9</td>
<td>1488</td>
</tr>
<tr>
<td>2 mm / 528</td>
<td>-50</td>
<td>1156</td>
<td>32.5</td>
<td>1364</td>
</tr>
<tr>
<td>4 mm / 531</td>
<td>-50</td>
<td>1150</td>
<td>44.3</td>
<td>1426</td>
</tr>
<tr>
<td>4 mm / 609</td>
<td>-50</td>
<td>1219</td>
<td>28.5</td>
<td>1512</td>
</tr>
<tr>
<td>4 mm / 611</td>
<td>-50</td>
<td>1210</td>
<td>30.6</td>
<td>1504</td>
</tr>
<tr>
<td>4 mm / 613</td>
<td>-50</td>
<td>1209</td>
<td>31.1</td>
<td>1499</td>
</tr>
<tr>
<td>4 mm / 614</td>
<td>-50</td>
<td>1293</td>
<td>16.7</td>
<td>1603</td>
</tr>
<tr>
<td>4 mm / 612</td>
<td>-50</td>
<td>1235</td>
<td>17.3</td>
<td>1531</td>
</tr>
<tr>
<td>4 mm (quenched)</td>
<td>-80</td>
<td>1283</td>
<td>23.1</td>
<td>1591</td>
</tr>
<tr>
<td>4 mm (quenched)</td>
<td>-80</td>
<td>1213</td>
<td>14.5</td>
<td>1504</td>
</tr>
<tr>
<td>4 mm / 536</td>
<td>-80</td>
<td>1281</td>
<td>26.5</td>
<td>1588</td>
</tr>
<tr>
<td>4 mm / 618</td>
<td>-80</td>
<td>1186</td>
<td>15.8</td>
<td>1470</td>
</tr>
<tr>
<td>4 mm / 610</td>
<td>-80</td>
<td>1209</td>
<td>21.7</td>
<td>1499</td>
</tr>
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</table>
TABLE 4. Comparison between charpy-V toughness of forged and plate steel at -50°C.

<table>
<thead>
<tr>
<th>CONDITION</th>
<th>FORGED (J)</th>
<th>PLATE (J)</th>
</tr>
</thead>
<tbody>
<tr>
<td>AS-RECEIVED (QUENCHED)</td>
<td>27.5</td>
<td>33</td>
</tr>
<tr>
<td></td>
<td>10</td>
<td>24</td>
</tr>
<tr>
<td></td>
<td>8.5</td>
<td>15</td>
</tr>
<tr>
<td>STEP-COoled</td>
<td>24</td>
<td>28</td>
</tr>
<tr>
<td></td>
<td>14</td>
<td>38</td>
</tr>
<tr>
<td></td>
<td>3</td>
<td>46</td>
</tr>
<tr>
<td>DE-EMBRITTLED (1 h at 640°C)</td>
<td>80.5</td>
<td></td>
</tr>
<tr>
<td></td>
<td>48</td>
<td></td>
</tr>
<tr>
<td></td>
<td>128</td>
<td></td>
</tr>
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</table>
FIG. 1. HEAT TREATMENTS APPLIED TO A508 CL3, EXCEPT REAUSTENITIZATION WHICH WAS CARRIED OUT AT 900°C.

FIG. 2. OBSERVED SCATTER IN A508 CL3 CHARPY V-NOTCH TEST RESULTS OBTAINED AT -50°C.
FIG. 3. OBSERVED SCATTER IN FRACTURE TOUGHNESS RESULTS OBTAINED AT -50°C.

FIG. 4. FRACTOGRAPHIC ASPECTS OF RUPTURE AT -50°C, SHOWING a) PRESENCE OF INTERGRANULAR FACETS IN A QUENCHED SPECIMENS EXHIBITING LOW-TOUGHNESS, b) HIGHER DENSITY OF INTERGRANULAR FACETS IN A STEP-COOLED SPECIMEN.
FIG. 5. SHOWING SCATTER IN RESULTS FOLLOWING STEP-COOLING. WITH LOW TOUGHNESS VALUES CORRESPONDING TO SPECIMENS EXHIBITING INTERGRANULAR FACETS.

FIG. 6. CHARPY TRANSITION CURVES OBTAINED FOR A508 CL3 SPECIMENS TESTED IN THE AS RECEIVED (QUENCHED), STEP-COOLED AND DE-EMBRITTLED CONDITIONS.
FIG 7. EFFECT OF DE-EMBRITTLE TEMPERATURE ON CHARPY TOUGHNESS OF FORGED STEEL.

FIG. 8. TENSILE PROPERTIES IN DE-EMBRITTLED AND STEP-COOLED CONDITIONS.
FIG. 9. AUGER ELECTRON SPECTROSCOPY RESULTS SHOWING SEGREGATION OF P AT THE GRAIN BOUNDARIES (TOP) AS COMPARED WITH CLEAVAGE FACETS (BOTTOM).
FIG. 10-EXPERIMENTAL DATA PRESENTED IN TERMS OF CUMULATIVE PROBABILITY TO FAILURE AS A FUNCTION OF $K_{JC}$ AND COMPARED WITH PREDICTED CURVES DERIVED FOR ALL DATA AND SEPARATED DATA ACCORDING TO THEIR MODE OF RUPTURE.