# FRACTURE OF STEELS IN THE BRITTLE TO DUCTILE TRANSITION REGION

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The inherent stochastic variability of J-Integral values at the onset of cleavage fracture for steels in the brittle-to-ductile toughness transition regime is modelled using the weakest link approach, suitably modified to include the effect of ductile tearing, combined with a precise description of the material J-R curve as determined from fractography. The model predictions are verified with test results obtained for specimens of different sizes from two steels, namely a 1CrMoV rotor steel and a SA533B pressure vessel steel. The implications of using simpler approaches which neglect any preceding cleavage in the statistical analysis of test results are discussed.

## INTRODUCTION

Three main factors are responsible for the large variability of toughness data in the transition regime. The first is the intrinsic stochastic nature of cleavage fracture. Global fracture is triggered by an initial event, occurring at the weakest of a population of possible initiators randomly distributed over the volume of material sampled by the stress field. Weibull statistics (1) can be satisfactorily employed under Linear Elastic Fracture Mechanics (LEFM) conditions. Problems, however, arise in the upper part of the transition region, where a consistently increasing amount of ductile tearing can precede instability. Secondly as cleavage fracture probability is highly influenced by lateral constraint, lab data are affected by the so-called "constraint effect". Factors such as specimen type, the initial crack length to specimen width ratio and the specimen thickness have been recognized as important variables affecting the degree of plane strain to plane stress. Proper consideration of constraint effects is important for the transferability of lab data to actual component situations (2). Last but not least heat-to-heat variability is quite important as cleavage fracture is triggered by microstructural features. This aspect

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has been addressed in recent years by a number of EPRI and Japanese reports (3, 4) which provide systematic correlations between material parameters such as the FATT, upper shelf fracture impact energy, flow stress and composition.

Over the last five years the authors of this paper have been working on the development of a predictive approach:

- in 1988 a phenomenological correlation between fractographic parameters and $J_{inst}$ values was established from tests on SA533B pressure vessel steel (5);
- in 1992 a predictive model, based on the coupling of a J-R curve derived from the above fractographic correlation with a stochastic fracture model previously reported the in literature, was successfully verified for 1CrMoV rotor steel data (6);
- in 1993 the analysis by that model was extended also to the previous data on SA533B (7).

The present paper provides a brief description of the model and of the results obtained in the analysis on the two steels above, before going on to compare it with simpler procedures (as presented in the draft ASTM standard (8)) for statistical analysis of toughness data in the transition regime.

## THEORY

Starting from the well accepted use of the weakest link concept to describe the statistical nature of fracture under brittle conditions (with the crack tip stress fields described by the K parameter), an extended model for fracture in the brittle-to-ductile transition regime (EPFM conditions, with a limited amount of ductile tearing) was proposed by Wallin (9):

$$P(K) = 1 - \exp[-\frac{B}{B_0}(\frac{K-K_{min}}{K_0-K_{min}})^4 (1+\frac{2\Delta a\sigma_0^{\,2}}{\beta K^2})] \quad (1)$$

where K is calculated from the value of J at instability via the equivalence principle of fracture mechanics, B is the specimen thickness, $B_o$ is a normalizing coefficient (the value of which may be chosen arbitrarily), $\Delta a$ is the amount of stable tearing, $\sigma_o$ is the flow stress (average of yield and ultimate tensile strengths) and $\beta$ is a measure of the normalized distance to the stress maximum from the crack tip ($\beta = 3.5\ U_t$, $U_t = x_{max}\ \sigma^2/K^2$, see (10)). The exact value of $K_{min}$, the minimum possible value of the toughness, is not easy to estimate (thousands of tests would be needed) bit is also generally not an important issue (6,9).

Eqn.1 itself may be useful for demonstrating the validity of this model by correlating experimental data, but due to the presence of the term $\Delta a$, which is unknown before any given test, it cannot predict toughness values. To obtain predictions, a second equation is provided based on a fractographically derived J-R curve and expressed in terms of the stretch zone width, measured during blunting (SZW) or critical ($SZW_c$), and the crack length ($\Delta a$):

$$for\ J/C_1<SZW_c: \quad J = C_1 \cdot SZW \quad , \quad \Delta a=0 \tag{2a}$$

$$for\ J/C_1>SZW_c: \quad J = C_1 \cdot SZW_c + C_2 \cdot \Delta a^{C_3} \tag{2b}$$

Eqn.2a refers to fracture during blunting and Eqn.2b to fracture after tearing. The accuracy of this equation, as well that as of more complex and therefore subsequently discarded models, for correlating fracture events in the transition regime was the subject of a number of previous studies by the authors (5,6), in which details of the parameter estimation procedures were also given. Here we point out only that the best correlations are obtained when Δa values are measured at the thickness position of the specimens where cleavage originates.

Eqns.2a and 2b can be solved to provide values of Δa as functions of J or K in the cleavage probability density function, Eqn.1. In this way cleavage probabilities for specimens of various sizes may be estimated at any given temperature, provided the appropriate values of the model parameters are known.

## MATERIALS AND EXPERIMENTAL

Fracture toughness data for a 1CrMoV forging rotor steel were taken from an experimental activity jointly carried out by CISE in Italy and by GKSS (Forschungszentrun Geesthacht GmbH) in Germany. Test procedures followed the recently proposed ESIS Procedure for Determining the Fracture Behaviour of Materials, ESIS P1-90, 1990. Compact tension (C(T)) specimens of three different thicknesses (from 10 to 100mm), all 20% side grooved, with $a_o/W=0.5$, were used in these tests; details are given in (6).

The subsequent verification of the model was carried out using existing fracture toughness data from an extensive characterization carried out by CISE in the '80s on SA533 gr.B cl.1 pressure vessel steel plates (5,11). In this latter case the C(T) specimens were plain-sided, with the exception of the room temperature specimens which were 20% side grooved; testing procedures followed the ASTM Standard Test Method for $J_{Ic}$, A Measure of Fracture Toughness (E 813-89) and Standard Test Method for Determining J-R Curves (E 1152-87).

The compositions and principal mechanical properties of the 1CrMoV and SA533B steels are reported in Table 1; descriptions of their microstructures may be found in (12) and (11) respectively.

## ANALYSIS

### Determination of the model constants

The capability of the model expressed by Eqn.1 to explain the variability of the $J_{inst}$

TABLE 1 - Composition and mechanical properties of the steels considered.

a) 1CrMoV steel

*Chemical composition (weight percent)*

| C | Si | Mn | P | S | Cr | Mo | Ni | V | Cu | As | Sn |
|---|---|---|---|---|---|---|---|---|---|---|---|
| 0.33 | 0.22 | 0.77 | 0.009 | 0.006 | 1.25 | 1.18 | 0.06 | 0.027 | 0.05 | 0.01 | 0.005 |

*Summary of tensile and impact properties*

| Temperature (°C) | $\sigma_y$ (MPa) | $\sigma_{uts}$ (MPa) | Elongation (%) | Red. of area (%) | FATT Orient. | FATT (°C) |
|---|---|---|---|---|---|---|
| RT | 660 | 806 | 17 | 62 | R-C | 81 |
| 500 | 506 | 563 | 22 | 79 | C-R | 90 |
| | | | | | L-R | 80 |

*Heat treatment: austen. at 970°C + air cooling + temp. at 660°C + furnace cooling*

b) SA533B steel

*Chemical composition (weight percent)*

| C | Mn | P | S | Si | Ni | Mo | Cr | Cu | Al |
|---|---|---|---|---|---|---|---|---|---|
| 0.21 | 1.43 | 0.006 | 0.011 | 0.23 | 0.62 | 0.50 | 0.08 | 0.08 | 0.016 |

*Summary of tensile and impact properties*

| Plate thickness (mm) | Grain size ASTM N° | $\sigma_y$ (MPa) | | $\sigma_{uts}$ (MPa) | | FATT (°C) |
|---|---|---|---|---|---|---|
| | | -80°C | 25°C | -80°C | 25°C | Orient. LT |
| 136 | 9.5 - 10 | 513 | 450 | 690 | 598 | 0 |

*Heat treatments: austenitizing at 880°C for 3.5 h + water quenching + tempering at 650°C for 5 h + stress relieving; furnace cooling for 6 h from 610°C to 315°C*

and $\Delta a$ values measured in tests at different temperatures and with different specimen sizes for a number of steels has been already documented in literature (9). As described above, by including the fractographically derived Eqns.2 the model can be used for making predictions, not merely correlations. Such predictions have been verified, adopting the following procedure: the model, with parameters determined from the small specimen data, was used to derive failure probability predictions for the larger specimens for which the transition region is shifted to higher temperatures (this is typical of life prediction and integrity assessments involving estimates of large component behaviour from small specimen data). Therefore two main types of extrapolation were involved:

- the model coefficients, and especially $K_o(T)$, had to be extrapolated in temperature, and
- the fractographically determined J-R curves had to be extrapolated to larger

crack extensions (to describe cracking in large specimens with extended J validity).

The model coefficients $C_1$, $C_2$, $C_3$, $SZW_c$, $\beta$, $K_{min}$, $K_o$ and the tensile properties E and $\sigma_o$ were evaluated from the smallest specimen tests on the two steels: B = 10 mm for 1CrMoV and B = 25.4 mm for SA533B. The temperature dependence of all these coefficients was explicitly taken into account. It may, however, be useful to point out that for the most part the temperature effect in this transition fracture model is provided by the temperature dependence of $K_o$. For example, in the case of the less ductile, higher strength steel 1CrMoV, the influence of the temperature dependence of all the other coefficients turns out to be negligible, thus making their numerical evaluation much simpler. The values of all the coefficients for both steels may be found in (7).

## Fracture behaviour prediction

Having calibrated the model constants for a particular group of specimens, the model then provides estimates of the failure probability and ductile crack extension at the instant of final fracture, for any values thickness, temperature or applied K (or J) level. Two distinct forms of presentation can be considered, with equivalent information content: failure probability densities at discrete temperatures plotted as a function of K and values of K at discrete fracture probabilities plotted as a function of temperature. Examples of such results are given in Figs. 1 and 2 for the 1CrMoV and SA533B steels respectively (the individual experimental data points in the left hand diagram in Fig. 1 are plotted along the corresponding probability density curves to solely indicate their test temperatures). From close examination of the first type of plot in Fig. 1, the sharp change in the shape of the probability density curve marks the border between the region where consistent probability exists that the specimens fail without any partial tearing and the region where the J levels are large enough to initiate tearing; mathematically this corresponds to the appearance of the term containing $\Delta a$ in Eqn.1.

The effects of temperature and specimen size on shape and position of the cleavage probability curves generated by the model are clearly seen in the toughness vs. temperature diagrams, which show the usual trend of increasing toughness with increasing temperature and decreasing specimen size. The positions of the K values corresponding to the LEFM limit, to $J_{IC}$ and to the last J-valid point (limit of EPFM) allow identification of the different fracture modes most likely to occur for various temperatures and specimens sizes:

- at low temperatures $K_{IC}$-valid data under LEFM conditions for specimens of sufficient size;
- a temperature region in which cleavage failure occurs during blunting;
- a temperature range of typical transition behaviour, with cleavage after some ductile tearing;
- at high temperatures cleavage probability is negligible at realistic K (or J) values and fully stable J-R curve behaviour is obtained.

Moreover from Figs.1 and 2 it is seen that the experimental results for both steels fall within the predicted 5% and 95% probability limits of the model. This is also true for the three test results on the 1CrMoV specimens with B=100 mm and hence the largest extrapolations: a 10-fold size increase, crack extensions 10 times larger and temperatures up to 50°C higher than in the calibration tests. For the SA533B steel it should be noted that the range of specimen sizes is lower and consequently the prediction capability check is less significant. In addition, the fact that these old tests were conducted using different test procedures and test machines and that the specimens taken from plates with different mechanical properties has restricted the population of SA533B data available for analysis. In any case the capability of the model to correctly predict the toughness trends is substantially confirmed, as in the more detailed analyses discussed in (6,7). These also include a verification of Eqn.2 which is shown to maintain its validity even for large extrapolations of $\Delta a$. This is indeed a key requisite for Eqn.1 (containing the $\Delta a$ term) to give reliable predictions for large specimens for which cleavage is likely to occur after significant stable crack extension.

Neglecting ductile tearing

In the literature it is well established that simple Weibull-type statistical model ($\Delta a = 0$ in the Eqn.1) is not appropriate if, as in the upper transition, cleavage occurs after a considerable amount of tearing. However it should provide a reasonable approximation as $\Delta a$ approaches zero and therefore, apart from any theoretical considerations, one might ask what is the advantage of the more complex model presented here over simpler approaches in typical practical situations?. In particular a draft fracture toughness testing standard for the transition regime which does not consider the role of ductile crack growth and which is based on a Weibull-type statistical model is in an advanced phase of preparation by ASTM. In view of the large scatter and the small number of tests available here for verification, the commonly used statistical significance tests are not possible. Nonetheless some comparative analyses can be made. Neglecting ductile tearing leads to two differences with respect to the present approach. In addition to using Eqn.1 with $\Delta a = 0$, to limit the role of ductile tearing the ASTM proposal considers test results yielding ductile crack extensions larger than $0.05\ b_o$ ($b_o$ = initial ligament) invalid; these are consequently ignored in the subsequent statistical analysis (derivation of mean values, confidence limits etc.).

While not considering the run-outs i.e. neglecting ruptures at high toughness values, is theoretically questionable (more appropriately a censored variable method with maximum likelihood minimisation is generally used, as in the analyses of the present data mentioned above), the method should yield conservative results. (One might observe that it is not for the experimentalist to introduce safety margins in experimental curves; ideally labs should provide accurate unbiased data and it is then up to the user of those data to decide how much conservatism to assume in the analysis.) On the other hand the consequence of dropping $\Delta a$ in Eqn.1 cannot be easily imagined. When used for "correcting" small specimen toughness data for large components via the relation:

$$B_s^{1/4}(K_s-K_{min})(1+\frac{2\Delta a_s\sigma_0^2}{\beta K_s^2})^{1/4} = B_l^{1/4}(K_l-K_{min}) \quad (3)$$

(subscripts: s = small, l = large) the fact that a multiplicative factor greater than unity is neglected in the right hand term might suggest that unconservative predictions are obtained for $_{KI}$ if ductile tearing occurs in the component. This does not necessarily mean that the Weibull model is unconservative: $K_o$ is an empirical parameter in a formula that has to be fitted to actual data (small specimen data in the present case) and therefore the best fit of the Eqn.1 provides a value for $K_o$ which is not identical to the value of $K_o$ that would be obtained by best fitting with the Eqn.1 with $\Delta a = 0$. Without performing a detailed numerical analysis it is difficult to predict how the two toughness curves with and without the $\Delta a$ term would compare, as the respective values of $K_o$ are different.

It was decided to perform a thorough re-analysis of all the data for the 1CrMoV and SA533B steels using the analysis procedure recommended in the ASRM draft. From several comparisons of the type shown in Fig.3, the following comments could be made.

- The differences in the models can not be appreciated in correlation analyses (model self-consistency: curves should obviously be well-fitted to the data used for calibrating the model constants) but only when large extrapolations are involved. This is the case here particularly for 1CrMoV (factor of 10 in size), and to a lesser extent for SA533B (factor of 4).
- For the 1CrMoV steel the simplifying assumptions of the ASTM analysis always yield toughness curves which are consistently below those of the present model which includes consideration of crack growth (Fig.3a).
- For the SA533B steel the toughness curves for the lower probability of fracture tend always to overlap (Fig.3b) as a consequence of the lack of higher temperature data needed to provide consistent extrapolations. For the higher fracture probabilities the ranking of two model generated curves is not clear (in particular in Fig.3b the ductile tearing correction model curve actually lies slightly below the ASTM curve, in contrast to the situation for 1CrMoV).
- Apart from this indication that Weibull predictions tend to underestimate fracture resistance of large components, particularly for a high strength, high tearing modulus steel, the amount of test data to demonstrate this important conclusion is presently poor and many more tests on large specimens fractured at temperatures close to the upper limit of the transition regime would be needed. At least it can be reported that in the case of the three tests on 100 mm thick 1CrMoV specimens, all the fracture values were fell within the 90% confidence limits predicted by the present approach, whereas one

data point was outside the corresponding 90% band of the Weibull model. Interestingly, this incorrect prediction was for one of the two large specimens which failed after significant ductile crack extension. The result therefore supports the self-evident logic that the usefulness of including consideration of ductile tearing is appreciable when large extrapolations in size and/or temperature are involved and significant ductile tearing occurs.

## CONCLUSIONS

1. A method for describing cleavage fracture events in steels in the transition regime has been discussed which combines a reported statistical cleavage fracture model with a fractographically determined J-R curve relationship. The model equations can be conveniently calibrated using a limited number of conventional laboratory tests (20 to 25 specimens seem sufficient in authors' experience). These can be subsequently used to predict failure probabilities for experimental conditions different to those used for the model calibration. The good agreement of the model predictions with experimental data confirmed the potential of this approach. Of particular engineering importance is the fact that the method can provide an estimate of the temperature level beyond which the risk of cleavage fracture of the component drops below any specified probability level.
2. The use of model equations (calibrated on small specimen test results) to predict larger specimen data at higher temperatures suggests that the use of a simpler Weibull model may be strongly conservative under conditions of extensive tearing. Interestingly, however, the lower bound toughness curves predicted by Weibull approach and the present more complex model were similar for the lower strength steel analyzed, SA533B. In any case the usefulness of considering ductile tearing appears most appreciable when (a) large extrapolations in size and/or temperature are involved, (b) significant ductile tearing occurs, (c) for high strength and high tearing resistance materials and (d) when dealing with average and upper bound probability data for the material toughness.
3. The proposed model aims to solve the problem of the description of the inherent stochastic variability of cleavage on fracture events, including systematic size effects, in the transition regime using the weakest link concept. Given that the heat-to-heat variability of toughness behaviour of steels can be tackled using appropriate empirical models available in the literature, only the variability due to constraint effects remains explicitly unaccounted for. In the authors' opinion a simple way to introduce constraint effects in the present model might be through an appropriate modification of the fractographically determined J-R curve (Eqn.2), whereas the weakest link statistical equation (Eqn.1) could remain unchanged. Work is presently in progress to consider the possible redefinition of Eqn.2 to include established constraint theories.

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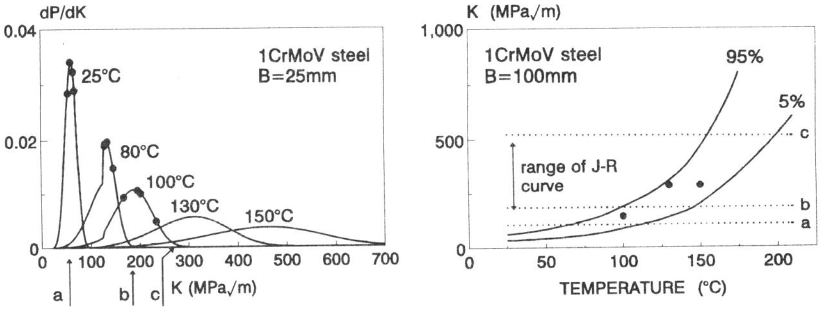


Figure 1 Model output examples [*a: K limit for LEFM; b: K from* $J_{Ic}$*; c: K at* $J_{max}$ *or* $\Delta a_{max}$*; in the left diagram the slight temperature dependences of a, b, c, are neglected.*]

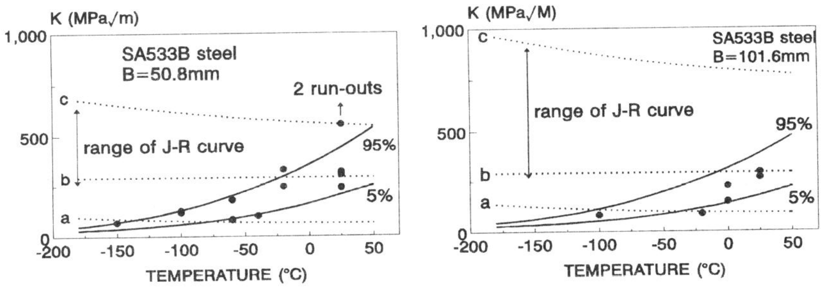


Figure 2 Toughness predictions for SA533B steel [*a,b,c = as above.*]

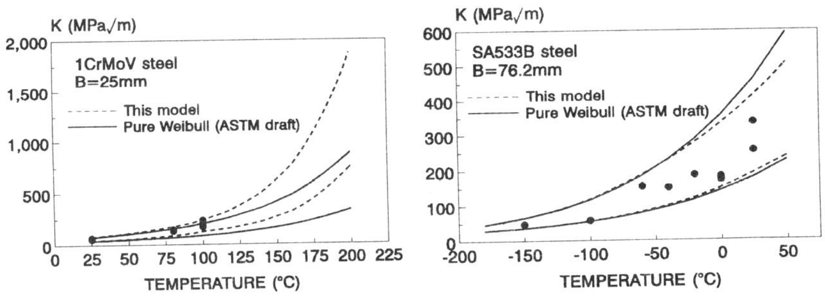


Figure 3 Comparisons of toughness curves predicted by the two models.