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STATISTICAL ANALYSIS OF CLEAVAGE-FRACTURE DATA

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ABSTRACT

This paper describes a practical method for dealing with scatter in fracture toughness of steel within the ductile/brittle transition region. The procedure is based on evidence that cleavage-fracture toughness of steel is controlled by local slip-induced fracture of second-phase particles and unstable growth of the resulting cracks. Operation of this mechanism is consistent with K_{Ic} values in the ductile/brittle transition being strongly temperature-dependent, with scatter around the median described by a Weibull distribution.

Research contributing to the above viewpoint picture is first described. Observations of the locations of fracture origins are reviewed and it is suggested that the scatter in locations of the cleavage-triggering particles (relative to the starting crack) are reflected strongly in the fracture toughness scatter.

The scatter is analyzed by means of a statistical algorithm developed to characterize temperature-dependent Weibull distributions. A data partitioning method for dealing separately with upper-shelf and transition-region data is also illustrated, and shown to yield better fits than those obtained using unpartitioned data. The analysis is then applied to evaluating cumulative failure percentiles for crack initiation (K_{Ic}) and crack arrest (K_{Ia}) data, yielding simple practical expressions for steel fracture toughness in the ductile/brittle transition region.

MICROMECHANISM OF CLEAVAGE

Current theories of cleavage fracture of steel assume that there are two controlling materials parameters: a critical stress (S*) and a critical length (x*). Both of these quantities are believed to depend on microstructure and it has been suggested (e.g. 1,2) that they need to reflect statistical variability. Such variability is reflected in the severe scatter in reported values of K_{Ic} in the ductile/brittle transition region (e.g. 3).

The micromechanism involves slip-induced fracture of a brittle particle, which in turn, triggers propagation of cleavage in ferrite. For a pre-cracked specimen these events are nucleated at a distance x* ahead of the crack at a point where the maximum local stress reaches S*. While the brittle particle is usually considered to be a carbide, our fractographic evidence (2,4) has shown that sulfides can play the same role. Calculations reported elsewhere (5) suggest that the sulfides become important factors when the carbide size falls below about one micron.

The fractographic studies also provide measurements of x* (4), which can vary by as much as 2 mm for six nominally identical specimens tested at one temperature. This distance is very large compared to typical microstructural features. A postulated mechanism, consistent with these results, involves growth of the crack-tip plastic zone as the specimen is loaded. With increasing load, the elastic/plastic boundary in the vicinity of the crack tip sweeps through the steel, causing fracture of brittle particles (6). If, when a particle cracks, the local conditions are sufficient to cause the crack to spread into the surrounding matrix, unstable cleavage results. Otherwise, the particle-crack blunts and the local stress concentration decreases. If, in addition, the triggering condition is not met before the dimpled-rupture condition is met, stable crack growth will occur. As the crack grows by dimpled rupture, additional crack-tip plasticity allows the possibility for further particle-induced cleavage to occur leading to final cleavage instability.

Since the local stresses in the plastic zone scale with yield strength, and because S* is generally believed to be temperature insensitive, this mechanism is consistent with increasing difficulty in producing cleavage (i.e. increasing K_{Ic}) with increasing temperature and decreasing strain rate. Since dimpled rupture obeys a separate strain criterion (7), we have assumed that the two fracture micromechanisms of steel have separate K_{Ic} values (2). Finally, the approach implies that cleavage fracture is inherently stochastic, since it requires just the right local conditions to occur to cause an instability. On the other hand dimpled rupture in low-to-medium strength steels is not associated with an instability and should be more reproducible.

A final ingredient in the model is the method of calculating K_{Ic} for specimens whose behavior is not linear elastic. This issue is under discussion by an ASTM Task Group. Our analysis (2) uses the idea that cleavage involves elastic unloading, so that only the the elastic energy stored in the specimen at the onset of instability contributes to driving the crack. Calculating K_{Ic} values using load, crack length at instability, and linear-elastic formulae from ASTM-E399 provides good agreement with results from very large cylinders subjected to thermal shock, whose behavior is presumably elastic (2). This elastic analysis, however, does not completely eliminate scatter in K_{Ic}. What remains is variability in toughness connected with the variability in the location of the trigger particle in relation to the tip of the precrack (2,4).

FRACTURE STATISTICS

The trigger particles combined with the surrounding ferrite can be considered "weak spots" in the context of Weibull's analysis (8). Thus it is appropriate to use the Weibull distribution to describe scatter in K_{Ic}:

$$F = 1 - \exp\{-(K_{Ic}/K_{Ico})^m\} \qquad (1)$$

where F is the probability that any measurement will fall below a specified value of K_{Ic}, K_{Ico} is the 63^{rd} percentile, and m (the Weibull modulus) characterizes scatter. A low value of m (e.g. five or less) reflects broad scatter.

Duckworth and Rosenfield (9) have used Eq. (1) to derive a description of scatter in K_{Ic} due to formation of a microcracked process zone in ceramics and predicted that m = 4. For steel, Wallin (1) assumed that the Weibull modulus is four and has derived a relation for the temperature dependence of K_{Ic} in the ductile/brittle transition region. However, it is not necessary to assume a fixed value of m, since researchers at Battelle have developed a very general technique for dealing with data that are both temperature-dependent and strongly scattered (10,11). In their analysis both m and K_{Icq} in Eq. (1) are allowed to be temperature-dependent. The new technique has several advantages: it makes no prior assumptions about the shape of the $K_{Ic}(T)$ curve; linear regression may be used; it is fast, since it can take advantage of highly optimized statistical subroutines; it produces confidence intervals, which give an estimate of error in the fit; and it can give estimates for any desired K_{Ic} percentile, which are important for engineering estimates of reliability.

Bishop et.al. (10,11) used the technique to develop a code denoted CHARPY to provide accurate estimates of exceedance percentiles and temperature-dependent variances of Charpy impact energy in the ductile/brittle transition region. In order to extend the analysis to the full transition curve, a modified code was developed with the acronym SAM MCFRAC. The modifications were necessary because attempts to fit entire Charpy curves with the original code were unsatisfactory. This problem is not surprising since the transition involves varying contributions of two fracture mechanisms (cleavage and shear), while the shelves each involve a single mechanism. Therefore, we added an option to enable the data set to be partitioned. As a minimum, two subsets are recommended: [1] the upper shelf and [2] the transition and lower shelf. Fracture appearance is used to partition the data set by separating the 100% shear results from the balance.

The SAM MCFRAC code was applied to Charpy-energy data for irradiated ASTM-A302B steel from the PSF Blind test (12). Figure 1 compares the results to those obtained with the well-known hyperbolic tangent fit (13). Note that the SAM MCFRAC fit provides a much better representation of the shape of the transition.

APPLICATION TO CRACK INITIATION AND ARREST

The thermal-shock experiments performed at Oak Ridge (14) and at Framatomme (15) have provided an opportunity to use our statistical techniques to derive reference fracture-toughness relations for a bainitic steel (ASTM A508) loaded in the ductile/brittle transition region. As noted above, these experiments involve very large specimens where plasticity effects are expected to be minimal. An initial evaluation of the crack <u>arrest</u> toughness, K_{Ia}, using the earlier CHARPY code provided two results: [1] the ratio K_{Ia}/K_{IR} is quite temperature insensitive (K_{IR} is the lower-bound fracture toughness curve of Section XI of the ASME Boiler & Pressure Vessel Code) and [2] the coefficient of variation of K_{Ia} is also insensitive to temperature (16). The same results are valid for crack-initiation toughness, K_{Ic}, the data being given in Table 1. Note that for K_{Ia} the "lower-bound" toughness is only about 1.4 standard deviations below the mean. These data further suggest that $K_{Ic} = 1.5\ K_{Ia}$, as is shown graphically in Figure 2. The solid line in the Figure is the mean of the crack-arrest data as given in the central column of Table 1, while the distance between the solid and each dashed line represents one standard deviation. As can be seen, multiplying each K_{Ic} value by 2/3 superposes both initiation and arrest data.

Applying Eq.(1) to the first column of Table 1, K_{IR} is found to lie on the 2.4 percentile of the K_{Ic} data and m = 4. In contrast, SAM MCFRAC analysis of compact-tension results suggests that K_{IR} represents the tenth percentile and m<2. This is an illustration of the fact that small-specimen K_{Ic} data are excessively scattered and may not be representative of large-specimen data (17). Correction of the compact results to account for plasticity provides a considerably better representation of the large-specimen data (18).

While the K_{IR} relation was derived empirically and might not seem to have physical significance, it can be written in the form:

$$K_{IR} = K_o\{ 1 + \exp [(T - T_o)/T^*] \} \quad (1)$$

where T is temperature. In equation (1) K_o is the lower-shelf toughness, T_o is the ductile/brittle transition temperature, and T^* describes the breadth of the transition. Little is known about metallurgical effects on either K_o or T^* (19). On the other hand, metallurgical effects on the ductile/brittle transition temperature have been thoroughly investigated, although relation(s) between impact-energy and T_o need to be established. For the data of Figure 1 (ASTM-A508 steel), K_o = 29.5 $MPam^{1/2}$, T^* = 38 C, and $T_o = RT_{NDT}$ + 29 C (RT_{NDT} is the ductile/brittle transition temperature as defined in Section XI of the ASME Boiler and Pressure Vessel Code).

TABLE 1: STATISTICAL PARAMETERS ASSOCIATED WITH FRACTURE TOUGHNESS VALUES DERIVED FROM THERMAL-SHOCK EXPERIMENTS

Parameter	K_{Ic}/K_{IR}	K_{Ia}/K_{IR}	(2/3) K_{Ic}/K_{IR}
Mean	2.31	1.59	1.54
Standard Deviation	0.73	0.42	0.48

SUMMARY AND CONCLUSIONS

While excessive data scatter is a disturbing feature of fracture-toughness of steel in the ductile/brittle transition region, the fractographic evidence suggests that variability is an inherent feature of cleavage behavior. Since the analytical techniques used in this paper are based on weakest-link theory, they provide a sound approach to describing this scatter. The improvements in methodology described above, particularly the technique of data partitioning with the aid of fracture-surface appearance, provide a particularly useful means of improving the efficiency of the statistical algorithm. Advantages of the current SAM MCFRAC code include flexibility in describing the temperature dependence of fracture toughness, provision for temperature-dependent variances, and ability to define both low percentiles and their confidence intervals. Finally, the code is not limited to fracture toughness but can be applied to other degradation mechanisms such as fatigue, stress rupture, and wear.

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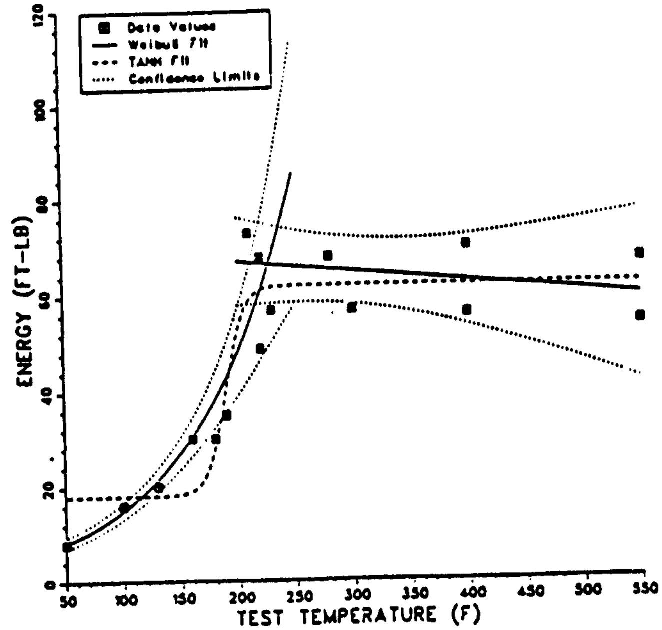

FIGURE 1. COMPARISON OF TEMPERATURE-DEPENDENT WEIBULL DISTRIBUTION AND HYPERBOLIC TANGENT FIT OF TYPICAL CHARPY IMPACT ENERGY DATA

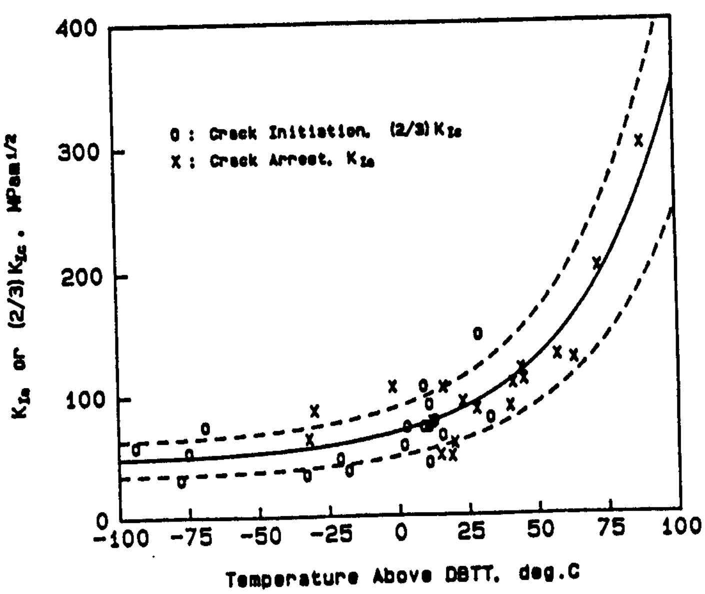

FIGURE 2. COMPARISON OF CRACK-INITIATION AND CRACK-ARREST DATA, THERMAL-SHOCK EXPERIMENTS IN THE DUCTILE/BRITTLE TRANSITION REGION