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Mechanical model for fracture toughness assessment

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1. Introduction

The fracture toughness assessment for equipment in operation presents a real technical challenge. This is of great importance for objects which tend to exhaust their design service life due to material properties degradation [1]. Among such objects are nuclear reactor vessels, transit oil and gas pipelines and many other high risk facilities. The significance of this problem is evident. That is why specialists all over the world spare no effort to develop reliable methods of monitoring metal degradation.

Since the year 1983 six symposia dedicated to this problem took place under the auspices of the ASTM [2-7]. The primary impetus was the need of the fusion reactor materials research community to assess effects of the very high levels of irradiation expected in the first wall of a fusion reactor. The limited volume of materials which can be irradiated in test reactors to high levels of embrittlement results in the need for small specimen technology [4]. Further symposia have shown that techniques thus obtained have spawned interest in non-nuclear applications [6]. For example direct fracture toughness testing of pipeline steels according to standard linear elastic fracture mechanics method (ASTM E1820) within the operational temperature range is not possible. The matter is that the pipe wall thickness, t, typically does not meet the ASTM E1820 thickness criteria for plane strain:

$$t \geq 2.5(K_{\mathrm{IC}} / \sigma_{0.2})^2. \quad (1)$$

At these temperatures only indirect approximate methods for K_{IC} estimation can be used. Many correlations have been established between fracture toughness and V-notch Charpy impact energy [2-7]. The huge publications quantity dedicated to this problem is known today. Due to small specimen's size, impact strength is widely used as an intermediary for indirect K_{IC} assessment. Strict relationships between stress intensity factors and Charpy impact energy do not exist because of a qualitative feature of the latter characteristic.

The goal of the present work is building up a physical model which makes it possible to simplify fracture toughness assessment within a wide temperature range. The method suggested excludes the necessity of a full-size specimens testing at the equipment operational temperatures. Our approach is based on the fundamental physical laws of a thermoactivated plastic strain. Its essence is discussed below.

2. Analysis

Previously [8] the model for steels with body-centered cubic lattice (BCC) was seen to predict the variation of K_{IC} within temperature range 77-300 K:

$$K_{IC} = K^*_{IC} + \sigma_{0.2i} (K_{IC} / \sigma_{0.2}); \quad (2)$$

$$K^*_{IC} = K_{IC}\sigma^*_{0.2} / \sigma_{0.2} = \text{const}. \quad (3)$$

In formulas (2) and (3) $\sigma_{0.2i}$ and $\sigma^*_{0.2}$ are the athermic structure sensitive (inner) and temperature sensitive (effective) components of the yield stress, $\sigma_{0.2}$, respectively. The ratio $K_{IC} / \sigma_{0.2}$ is regarded as a parameter of the plastic zone length ahead of a crack tip. Every BCC steel has its own constants $\sigma_{0.2i}$ and K^*_{IC}, which values are not similar for different materials.

Taking into account the linear relationship between $\sigma_{0.2}$ and HB within the temperature range $77 \leq T \leq 293$ K [9], the Eq. (2) can be rewritten as:

$$K_{IC} = K'_{IC} + HB_i (K_{IC} / HB), \quad (4)$$

where slope HB_i refers to the athermic component of a Brinell hardness, intercept $K'_{IC} = K_{IC} HB^* / HB$ is a constant and HB^* is the thermoactivated component of a Brinell hardness. Obviously that formula (4) is much more convenient than the Eq. (2) due to simplicity of a hardness test. But its practical use is difficult because generalized linear equation for different pipeline steels doesn't exist. The intercept, K'_{IC}, and slope, HB_i, parameters are not similar for different steels. That's why the graph for Eq. (4) in $K_{IC} - K_{IC} / HB$ coordinates for different steels represents a family of straight lines (Fig. 1).

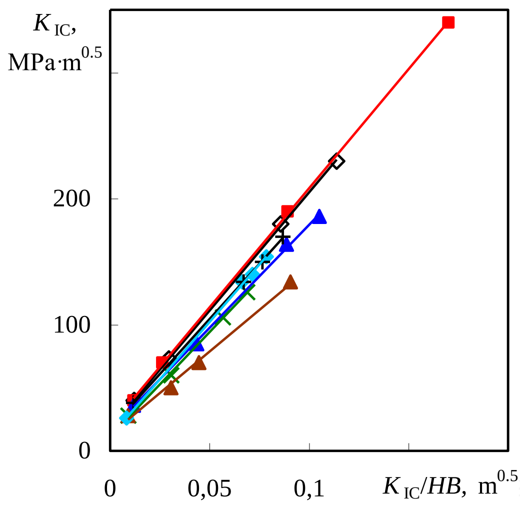

Fig. 1 Correlation $K_{IC} - K_{IC} / HB$ for investigated steels within the temperature range of $77 \leq T \leq 243$ K: ◇ – 10G2FB; ■ – 10G2FB-U; ▲ – 17G1S-U; ✕ – 17GS; ▲ – V st. 3 kp; ◆ – 10HGNMAYu; + – 06G2NAB; the symbols apply also to Fig. 2

It is well known [10] that under standard test conditions (ASTM E1820) small scale yielding zone length, r, ahead of a crack tip is proportional to $(K_{IC} / \sigma_{0.2})^2$. Due to the linear relationship between $\sigma_{0.2}$ and HB for all the steels investigated within the temperature range 77–293 K [9], the K_{IC} / HB ratio also can be regarded as a parameter of the small scale yielding zone length ahead of a crack tip.

Table 1

Properties of investigated steels, [9, 12]

Steel and heat treatment	T, K	K_{IC} or K_{JC}, MPa m$^{0.5}$	HB, MPa
10G2FB, controlled rolling	77	40	3330
	163	73	2470
	213	180*	2100
	243	230*	2020
	293	240*	1870
10G2FB-U, controlled rolling	77	40	3410
	153	70	2660
	203	190*	2130
	253	340*	2000
	293	170*	1890
17G1S-U, hot rolling	77	36	3080
	193	85	1950
	213	164*	1850
	243	186*	1770
	293	164*	1700
17GS, hot rolling	77	28	3050
	193	60	1950
	213	106*	1870
	243	126*	1830
	293	136*	1720
V st. 3 kp, hot rolling	77	28	3100
	193	50	1640
	213	70	1570
	243	134*	1480
	293	170*	1420
10HGNMAYu, normalizing	77	26	3160
	193	134*	2020
	213	154*	1960
	243	140*	1960
	293	124*	1740
06G2NAB, normalizing	77	38	3210
	193	134*	2000
	213	170*	1960
	243	150*	1960
	293	130*	1780

*Values with asterisk: K_{JC}

Let's accept the assumption that for all the steels investigated exists the general linear relationship between the ratios $K_{IC,T} / K_{IC,243}$ and $\sqrt{r_T / r_{243}}$. Here r_T and r_{243} refer to the small scale yielding zone lengths at a certain temperature $T < 243$ K and 243 K, respectively. As was shown in our previous work [9], most of the steels investigated reveal the phase transition at the temperature of 243 K. The plastic flow below this temperature point is controlled by the Peierls-Nabarro force [11].

For check of our assumption, both parts of the linear Eq. (4) were normalized to fracture toughness $K_{IC,243}$ at a phase transition temperature.

It is obvious that:

$$\frac{K_{IC,T}}{K_{IC,243}} = \frac{K_{IC,T}}{K_{IC,243}} \times \frac{HB^*}{HB_T} + \frac{K_{IC,T}}{K_{IC,243}} \cdot \frac{HB_i}{HB_T}. \tag{5}$$

After the multiplication and division of the last item in the expression (5) by HB_{243} we get:

$$\frac{K_{IC,T}}{K_{IC,243}} = \frac{K_{IC,T}}{K_{IC,243}} \times \frac{HB^*}{HB_T} + \left(\frac{K_{IC,T} \cdot HB_{243}}{K_{IC,243} \cdot HB_T} \right) \times \frac{HB_i}{HB_{243}}. \tag{6}$$

The analysis of the Eq. (6) was performed for the pipeline steels presented in Tables 1 and 2. The plane strain fracture toughness and tension tests of the steels investigated were carried out by Krasowsky and Krasiko [12] at the Institute for Strength Problems (Kiev, Ukraine). After testing the specimens were offered to the author of this paper. Hardness measurements (Table 1) were then carried out at the Volgograd State Technical University (Russia). The experimental procedure for the Brinell hardness, HB, determination was described in our previous work [9].

The results of our calculations, presented in Table 3, demonstrate a very minor ratio $\frac{HB_i}{HB_{243}}$ (slope) deviation from the average value. It keeps almost constant values within all the temperature range for all the steels investigated.

The ratio $\frac{K_{IC,T}}{K_{IC,243}} \cdot \frac{HB^*}{HB_T}$ (intercept) has more significant scatter but its values in itself are very small.

Table 2

Composition of investigated steels (wt%), [12]

	10G2FB	10G2FB-U	17G1S-U	17GS	V st. 3 kp	10HGNMAYu	06G2NAB
C	0.10	0.10	0.16	0.15	0.17	0.12	0.08
Mn	1.60	1.55	1.39	1.31	0.59	1.20	1.50
Si	0.33	0.33	0.51	0.51	0.22	0.26	0.25
S	0.004	0.004	0.018	0.016	0.025	0.01	0.01
P	0.020	0.020	0.015	0.017	0.016	0.01	0.02
Cr	—	—	0.02	—	—	0.04	—
Al	—	—	0.042	—	—	0.13	0.03
Ti	0.021	0.020	0.064	—	—	—	—
As	—	—	0.010	0.004	0.002	—	—
V	0.097	0.096	—	—	—	—	—
Nb	0.025	0.025	—	—	—	—	0.15
Mo	—	—	—	—	—	0.30	—
Ni	—	—	—	—	—	1.30	0.70

Table 3
The comparison of intercept and slope values in Eq. (6) for every steel with generalized ones

Steel	Intercept $(K_{IC,T} / K_{IC,243})(HB^* / HB_T)$	Slope HB_i / HB_{243}
10G2FB	0.0774	0.9275
10G2FB-U	0.0572	0.9456
17G1S-U	0.0858	0.9238
17GS	0.0901	0.9067
V st. 3 kp	0.0956	0.8927
10HGNMAYu	0.0800	0.9296
06G2NAB	0.1137	0.8931
Generalized Eq. (7)	0.0813	0.9232
Highest deviation from generalized Eq. (7) for steels investigated	±39%	±3%

One can therefore conclude that the general linear relationship has to exist for all the materials investigated. As one can see from Fig. 2, this assumption is fully confirmed. The data points ideally lay down on a straight line which can be approximated by the next equation:

$$K_{IC,T} / K_{IC,243} = 0.0813 + \\ + 0.9232(K_{IC,T}\, HB_{243}) / (K_{IC,243}\, HB_T). \qquad (7)$$

The correlation coefficient for the last relationship equals 0.9992.

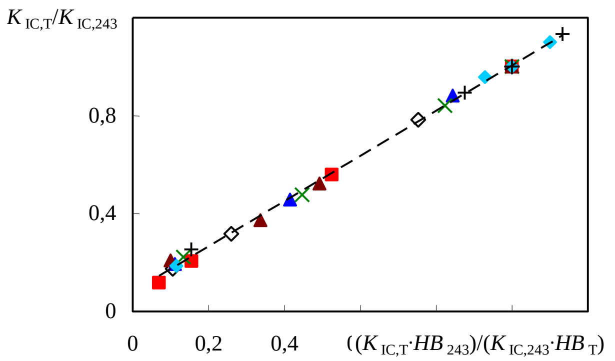

Fig. 2 The unified linear fracture toughness diagram for pipeline steels within temperature range $77 \le T \le 243$ K. Symbols - as in Fig. 1.

Let's consider the meaning of the variables in the Eq. (7). The data obtained leads to a reasonable conclusion that there is a stable relationship between the ratios of stress intensity factors $K_{IC,T} / K_{IC,243}$ and parameters of small scale yielding zone lengths $(K_{IC,T}\, HB_{243}) / (K_{IC,243}\, HB_T)$. Here $K_{IC,T}$, HB_T, $K_{IC,243}$ and HB_{243} are the stress intensity factors and Brinell hardness at a certain temperature, T, and at the phase transition temperature 243 K, respectively. Taking into account the linear relationship between $\sigma_{0,2}$ and HB within temperature range of 77–293 K [9], we reach the conclusion that the generalized linear relationship exists between the $K_{IC,T} / K_{IC,243}$ and $\sqrt{r_T / r_{243}}$.

Let's address to the further analysis of the Eq. (6). It is obvious that at a phase transition temperature of 243 $K_{IC,T} = K_{IC,243}$. In this case terms HB^* / HB_T and HB_i / HB_T in the right part of the Eq. (6) characterize respectively a contribution of the thermal and athermic hardness components. Theoretically their sum has to equal one. As can be seen from the Eq. (7), at a phase transition temperature $HB^* / HB_T = 0.0813$ and $HB_i / HB_T = 0.9232$ that in the sum makes 1.0045. The deviation from one makes only 0.45% what is the additional testimony of the model validity.

Variations in temperature lead to change of the both components HB_i and HB^* contribution into the hardness value. Constancy of the first term (intercept) for every steel in the Table 3 at any temperature within the range of 77 – 243 K testifies that fracture toughness temperature sensitivity is caused by the thermal component of a yield stress (hardness).

The analysis of the slope term HB_i / HB_{243} in the Table 3 and the Eq. (7) leads to the conclusion that at a phase transition temperature it is almost constant and doesn't depend on the hardness of steel. In other words, the athermic component, HB_i, contribution to HB hardness at a phase transition temperature is almost similar for all the steels investigated.

3. Practical applications

The obtained results allow proposing the following technique for K_{IC} estimation. First, the relationship $HB(T)$ has to be established. After that, the $K_{IC,77}$ value at a liquid nitrogen temperature has to be determined and $K_{IC,243}$ can be calculated. It should be noted that at the liquid nitrogen temperature the ASTM thickness criteria for materials considered is satisfied at thicknesses of 3 – 5 mm.

Then $K_{IC,243}$ has to be found from the Eq. (7):

$$K_{IC,243} = \frac{(1 - 0.9232 \times HB_{243} / HB_{77}) \times K_{IC,77}}{0.0813} . \qquad (8)$$

Further it is simple to find $K_{IC,T}$ at any temperature as:

$$K_{IC,T} = \frac{0.0813 \times K_{IC,243}}{1 - 0.9232 \times HB_{243} / HB_T} . \qquad (9)$$

The comparison of the experimental K_{IC} data (Table 1) with the prediction of the Eq. (9) is presented in Fig. 3.

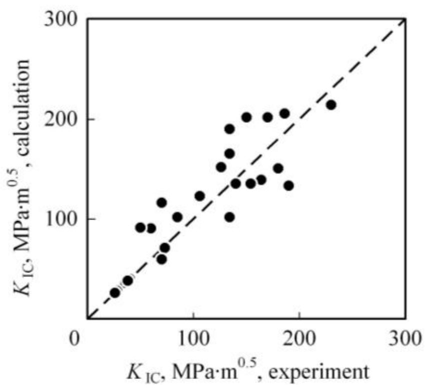

Fig. 3 Results of the experimental check of the generalized Eq. (7)

As can be seen, the prediction accuracy as a whole is satisfactory. Only in one case for the V st. 3 kp steel the error reaches 65.9%. The reason is that the intercept term deviation from average accounts for ± 39% (Table 3). This may increase the calculated K_{IC} scatter band. Rigorous intercept term analysis is very complicated because of different strengthening mechanisms existing in different steels. The exact contribution of every mechanism is not known. These important questions make a subject of physics of metals and aren't discussed in this work.

For better prediction accuracy the calculation from separate equations (Table 3) for every steel can be recommended (Fig. 4). The calculations procedure is the same as described in this section, Eqs. (8) and (9). The only difference is that separate intercept and slope values from Table 3 were used for every steel.

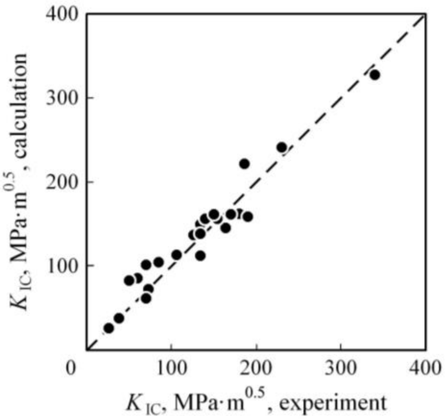

Fig. 4 Comparison of the K_{IC} values calculated from the separate equations for every steel (Table 3) with the experimental data (Table 1)

4. Conclusions

1. Many high risk facilities tend to exhaust their design service life due to material properties degradation. The fracture toughness assessment for equipment in operation is associated with significant complications. The limited volume of materials which are available for testing, results in the need for small specimen technology.

2. The physical model for fracture toughness assessment is proposed. This model is based on the known fact that yield stress/hardness consists of athermic and thermoactivated components. It is shown that below the phase transition temperature of 243 K the general linear relationship between the ratios $K_{IC,T} / K_{IC,243}$ and $\sqrt{r_T / r_{243}}$ exists for all the steels investigated. Here r_T and r_{243} refer to the small scale yielding zone length at a certain temperature T < 243 K and phase transition temperature 243 K, respectively. This model is valid within the temperature range where the plastic flow is controlled by the Peierls-Nabarro force.

3. The model described above makes it possible to suggest the simplified methodology for fracture toughness assessment. It is based on the direct fracture toughness determination, according to the ASTM E1820 procedure, using very small specimens, at the liquid nitrogen temperature. After that, the K_{IC} values at the operational temperatures can be calculated using the proposed in the present work relationships between fracture toughness and Brinell hardness.

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MECHANICAL MODEL FOR FRACTURE TOUGHNESS ASSESMENT

S u m m a r y

Nuclear reactor vessels, transit oil or gas pipelines and many other high risk facilities tend to exhaust their design service life due to material properties degradation. The fracture toughness assessment for such equipment in operation poses a real technical challenge. The limited volume of materials which are available for testing, results in the need for small specimen technology. The method suggested is based on direct fracture toughness determination using very small specimens, at the liquid nitrogen temperature. After that, the K_{IC} values at the operational temperatures can be calculated using the model proposed by the author.

Keywords: fracture toughness, Brinell hardness; material properties degradation; low temperatures; thermoactivated plastic strain.

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