CORRELATION BETWEEN APPARENT TOUGHNESS K_A AND CHARPY IMPACT ENERGY IN HIGH AND LOW STRENGTH STEELS

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ABSTRACT

A new graphical representation of Spink 'and others' relationship, relative to apparent toughness K_A evolution with notch radius, permits microcracks initiation detection whether closely to nocth root or at ρ_o from the notch root. Also, in the brittle region, Charpy impact energy evolution has been found the same as that of K_A with notch radius.

KEYWORDS

Apparent toughness, Charpy impact energy (KCV), notch radius, notch root, K_{Ic}.

INTRODUCTION AND THEORETICAL BACKGROUND

Many efforts have been made to understand notch constraint effects on the apparent fracture toughness of materials "K_A". Unfortunately little agreements have been found between different models.
Wilshaw 'and others' (1968) show that there is a linear dependence between the square root of the notch radius ρ and the apparent fracture toughness when fracture occurs in a brittle manner by cleavage crack propagation, and provided that ρ is greater than some critical value, below which the measured value of K_A is constant and equal to K_{Ic}. Wilshaw 'and others' (1968) relationship is given as following :

$$K_A \simeq 2.89\, \sigma_Y \left\{ \exp\left[\left(\sigma^*_f / 1.15\, \sigma_Y\right) - 1\right] - 1 \right\}^{1/2} \rho^{1/2} \qquad (1)$$

where σ^*_f is the critical fracture stress, σ_Y the tensile yield stress. In this relation, we consider that the material obeys the Von Mises yield criteria.
However, Oates and Datta (1967, 1980) do not support the linear dependence. In addition none of results either those of Irwin (1964) or of Wilshaw 'and others' (1968) is for a root radius greater than one mm. Furthermore, it is difficult to justify this model when failure results from the formation of a microcrack at crack root. On the other hand, Spink 'and others' (1972, 1973)

have shown that even in the post yield region the plane strain fracture toughness controls the mechanisms of fracture. They then assumed that Smith's relationship (1967) calculated for a state of anti-plane strain deformation may be approximately valid for plane strain state. Smith's relation may be written as :

$$\sigma_F^*/\sigma_U = \frac{1}{1+(\rho/c)^{1/2}}\left[\frac{2}{\pi}\sec^{-1}\left[\exp\frac{\pi K_{Ic}^2}{8\sigma_U^2 c}+(\rho/c)^{1/2}\right]\right] \qquad (2)$$

with σ_U the ultimate stress, c the blunt notch length and K_{Ic} the toughness of material. In the case of small scale yielding situation and when $K_{Ic}/\sigma_U \ll 1$, equation (2) reduces to :

$$\frac{\sigma_F^*(\pi c)^{1/2}}{K_{Ic}}\left[1+(\rho/c)^{1/2}\right] \simeq 1+\frac{\sigma_U(\pi c)^{1/2}}{K_{Ic}}(\rho/c)^{1/2}$$

Assuming an apparent fracture toughness K_A in the post yield region and relating it to the plane strain fracture toughness, K_A is then equal to $\sigma_F^*(\pi c)^{1/2}$

$$\frac{K_A}{K_{Ic}}\left[1+(\rho/c)^{1/2}\right] = 1+\alpha(\rho/c)^{1/2} \qquad (3)$$

where $$\alpha = \frac{\sigma_U(\pi c)^{1/2}}{K_{Ic}}$$

The appropriate value of the ultimate stress in bend is difficult to estimate, but as equation (3) is formulated we have no need now of this value. In any case, this equation relates K_A, measured on a specimen with a blunt notch with root radius greater than one mm, to the material property K_{Ic}. Equation (3) has been modified by Townley and Heald (1975) into a semi-empirical fracture formula for plane strain involving the finite dimensions of the test specimen and its collapse load This formula correlates well with fracture data obtained by Spink and Others (1973) on small blunt-notched specimens broken in three-point bend test.

TENTATIVE UNIFYING RELATIONSHIP

Using equation (3) we plotted $(K_A/K_{Ic})\,[1+(\rho/c)^{1/2}]$ versus $(\rho/c)^{1/2}$ for results found in literature (1980, 1973, 1979, 1976) and authers' results. Figure 1 shows that the relations are always linear and at $\rho = 0$, $K_A = K_{Ic}$. Each relation has a slope value which represents the true value of σ_U, and then the microstructure feature of material at crack notch as has been mentioned by Curry and Knott (1976). Datta results (1980) show that 4340 steel austenitized at 870°C (grain size ≃ 25 μm) has a slope value greater than that of the same steel austenitized at 1200°C with grain size about 300 μm, figure 1. Also 4340 steel with small grain size tempered at 175°C, i.e. containing fine precipitates of ε-carbide leads to a slope lower than that of the austenitized steel at 870°C. We applied also equation (3)on Oates results (1967) obtained on mild steel with different grain sizes. Figure 2 shows that the slopes of the relation $(K_A/K_{Ic})[1+(\rho/c)^{1/2}] = f(\rho/c)$ corroborate with the microstructure of the steel, i.e. slope value decreases with increasing grain size.
Continuing plotting $K_A/K_{Ic}[1+(\rho/c)^{1/2}] = f(\rho/c)$ with Ritchie's and Others (1976) and Wilshaw and Others (1968) results, figure 3, we can observe that all

the relations are linear and also equation (3) is modified and may be expressed as :

$$K_A/K_{Ic}\left[1+(\rho/c)^{1/2}\right] = \alpha'(\rho/c)^{1/2} \qquad (4)$$

where $$\alpha' = \frac{1+(\rho_o/c)^{1/2}}{(\rho_o/c)^{1/2}}$$, ρ_o characteristic distance over which the critical stress σ_f must exist to cause failure and is related also to microstructural feature which controls fracture, such as slip or twin band spacings, grain size or inclusion spacing. Indeed, figure 3 shows that α' for 4340 steel austenitized at 870°C (24 → 32 μm grain size) is greater than α' of 4340 steel austenitized at 1200°C followed by a salt quench to 870°C, then oil quenched to room temperature.
Table 1 shows values of α and σ_U calculated from equation (3), and from the results of (1980), (1973), (1976) and (authors, 1984) on aged martensitic Fe-23,6%Ni-0,36%C alloy.

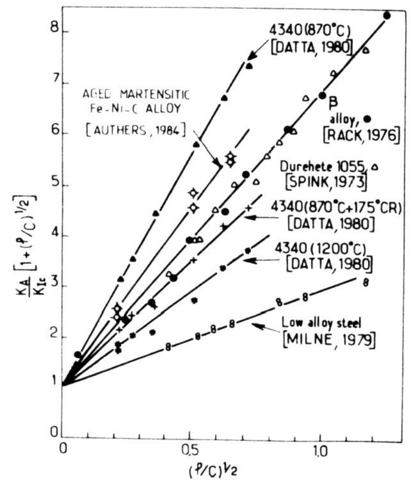

Fig. 1. $K_A/K_{Ic}\,[1+(\rho/c)^{1/2}]$ versus $(\rho/c)^{1/2}$ when $\rho_o = 0$

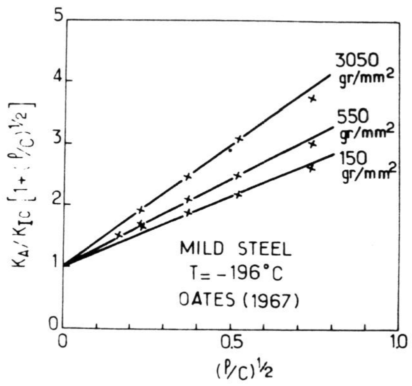

Fig. 2. Slope variation of Spink and Others relation with grain size

TABLE 1 : α and σ_U values calculated from equation (3)

Steel	σ_Y,(MPa)	σ_U*,(MPa)	K_{Ic},(MPa$\sqrt{m}$)	α	σ_U calculated,(MPa)
Durehete	690	1080	30.00	6.00	927
Low alloy steel	580	740	57.00	2.00	587
β Titanium alloy	1734	1734	21.90	6.00	829
4340 (870°C)	-	-	41.29	9.20	2396
4340 (870°C + 175°C)	-	-	71.50	5.00	2255
4340 (1200°C)	-	-	61.78	3.80	1481
Fe-23%Ni-0,36%C** aged martensite	960	2217	62.65	7.08	2798

σ_U* measured by tensile test at room temperature.

** σ_Y and σ_U* are measured by tensile test at -196°C.

Table 2 shows also values of α' and ρ_0 calculated from equation (4), from the results of (1968) and (1976).

TABLE 2 : α' and ρ_0 values calculated from (4)

Steel	σ_Y (MPa)	calc. σ_U (MPa)	K_{Ic} or K_{Id} (MPa$\sqrt{m}$)	α	from (4) ρ_0, µm	ρ_0 from (1) µm
High nitrogen mild steel	840	1400	21.50	6.5	66	51
4340 (870°C)	1593	2217	36.63	12.1	16	25
4340(1200-870°C)	1593	2193	56.65	3.7	274	284

As has been mentioned elsewhere (1976), when K_{Ic} is improved there is an unexplained and perplexing reduction in the Charpy impact energy. These contradictory results are due to important differences between fracture induced at sharp (K_{Ic} at $\rho = 0$) and blunt notch (KCV, $\rho > 0$). These differences are :

1) Charpy test measures total energy (initiation + propagation) required to cause complete failure of the specimen ; K_{Ic} test measures a critical energy at the crack tip necessary to initiate plane strain unstable fracture (1976).
2) Charpy test strain rate is several order of magnitude greater than in a K_{Ic} test (1976).
3) Charpy test specimen contains a V-notch with $\rho \simeq 0,25$ mm, whereas K_{Ic} test specimen contains a fatigue precrack with $\rho \simeq 0$ (1976).

Then there is no physical reason to compare KCV evolution with K_{Ic} values in function of microstructural modifications. However, in brittle region, on (KCV - T) curves, we can neglect propagation energy in Charpy test and we have to compare KCV and K_A values at each notch radius whether on applying equation (3) or (4). By this way we plotted from Rithie results (1976) relative to 4340 steel $(KCV/KCV_0)^{1/2}[1 + (\rho/c)^{1/2}] = \beta(\rho/c)$ in figure 4. KCVo is the impact energy in specimen containing fatigue precrack and KCV the impact energy in specimen containing V-notch with $\rho > 0$. We can observe that for the two heat treatments, the relations are also linear and may be expressed by equation (4) as has been shown with K_A measurements on the same steel and heat treatments, figure 4. ρ_0 values are respectively 21 and 238 µm for 870°C and 1200°C - 870°C heat treatments. These values are equivalent

to grain sizes and approximatively equal to those obtained in K_A measurements. The new empirical relation may be written :

$$\left(\frac{KCV}{KCV_0}\right)^{1/2}\left[1+(\rho/c)^{1/2}\right] = \beta(\rho/c)^{1/2} \qquad (5)$$

with

$$\beta = \frac{1+(\rho_0/c)^{1/2}}{(\rho_0/c)^{1/2}}$$

The purpose of this paper is to verify equations (3), (4) and (5) validity in the brittle region, i.e. at liquid nitrogen temperature for two materials mild steel with and without internal hydrogen and no-aged martensitic Fe-24%Ni-0,3%C alloy.

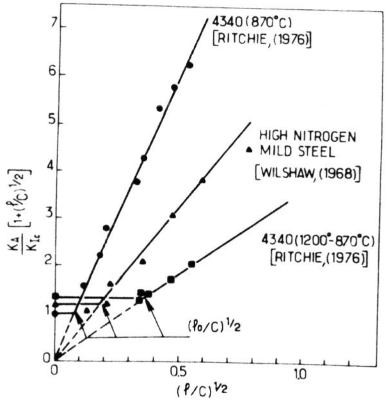

Fig. 3. $K_A/K_{Ic}\ [1 + (\rho/c)^{1/2}]$ versus $(\rho/c)^{1/2}$ when $\rho_0 > 0$

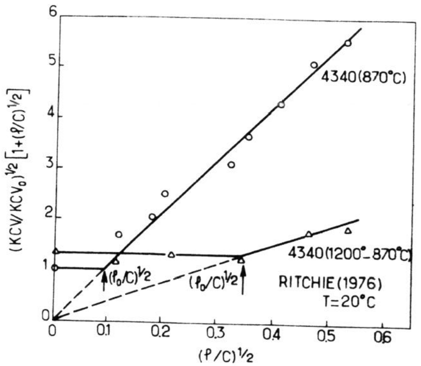

Fig. 4. $(KCV/KCV_0)^{1/2}\ [1 + (\rho/c)^{1/2}]$ versus $(\rho/c)^{1/2}$

FURTHER RESULTS

Alloys and experimental procedure

The materials used in this investigation are normalized mild steel with the following chemical composition, Table 3.

TABLE 3 : Chemical composition of mild steel.

C%	Mn %	Si %	P %	S %	Cu %	Ni %	Cr %	Mo %
0.080	1.520	0.370	0.024	0.001	0.340	0.150	0.200	0.010

Steel hydrogen charging is performed in a molten salts bath (El Kholy, 1976) at 300°C, -2 volts/Ag and during one hour. The chemical composition of Fe-Ni-C alloy is given in Table 4. M_s temperature of this alloy is $\simeq$ -40°C.

TABLE 4 : Chemical composition of Fe-Ni-C alloy.

C%	Mn %	Si %	P %	S %	Ni%
0.300	0.530	0.360	0,019	0.016	24.090

Charpy impact energies and K_A values were determined using standard sized ASTM Charpy V-notch specimens (c = 2 mm). Pendulum type impact machine of hammer velocity 3,3 $m.s^{-1}$ is employed to measure KCV. K_A values are determined with a three-point bend device mounted on an Instron machine at a cross-head displacement rate 1 $mm.min^{-1}$. The failure stress σ^*_f is defined as the maximum value of the elastic center fiber stress $\sigma = 6\ PS/_{4BW^2}$ with P the applied load, S the span, B the specimen thickness and W its width.

RESULTS AND DISCUSSION

Figures 5 and 6 show the variation of $K_A/K_{Ic}[1 + (\rho/_c)^{1/2}]$ and $(KCV/KCV_o)^{1/2}.[1 + (\rho/_c)^{1/2}]$ with $(\rho/_c)^{1/2}$ for the no-aged martensitic Fe-Ni-C alloy and the normalized mild steel. In the two cases, the relations KA/K_{Ic} or $KCV/KCV_o)^{1/2}$ versus $(\rho/_c)^{1/2}$ are similar and linear. ρ_o are also equivalent. When mild steel is hydrogen charged at high temperature, micro cracks ahead of the fatigue crack tip or notch root may be created, the maximum tensile stress σ_{YY}^{max} is then very close to the notch tip and the characteristic distance is very small. Figure 7 shows that $\rho_o = 0$ in K_A and KCV measurements without hydrogen, σ_{YY}^{max} is located at or just behind the plastic-elastic interface (Ritchie, 1976) and thus the characteristic distance is larger. Table 5 gives the different values of K_{Ic}, KCV_o, α or α', β and ρ_o measured by the two methods for the materials studied in this paper and by Ritchie and others (1976).

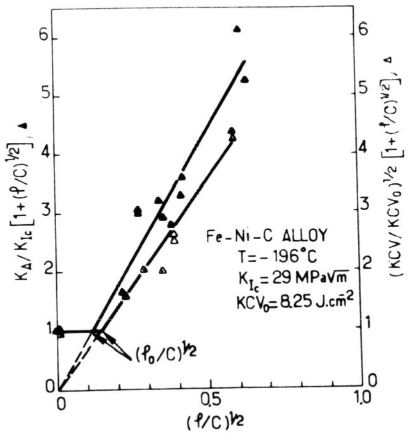

Fig. 5. K_A/K_{Ic} $[1 + (\rho/_c)^{1/2}]$ and $(KCV/KCV_o)^{1/2}$ $[1 + (\rho/_c)^{1/2}]$ versus $(\rho/c)^{1/2}$ for a no-aged martensitic Fe-24%Ni-0.3%C alloy.

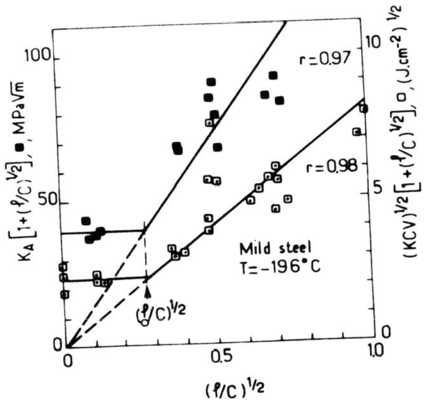

Fig. 6. K_A/KI_c $[1 + (\rho/c)1/2$ and $(KCV/KCV_o)^{1/2}$ $[1 + (\rho/c)^{1/2}]$ versus $(\rho/c)^{1/2}$ for mild steel.

TABLE 5 : K_{Ic}, KCV_o, α or α', β and ρ_o measured in three-point bend and Charpy Impact pendulum.

Steel	K_{Ic} MPa√m	KCV $J.cm^{-2}$	α or α'		$\rho_o(K_A)$ µm	$\rho_o(KCV)$ µm	d µm
4340 (870)	36.63	1.63	12.1	10.7	16	21.3	24 to 32
4340 (1200-870°C)	56.65	2.64	3.7	3.9	274	238.0	254 to 360
non aged martensitic Fe-Ni C alloy	29.00	8.25	8.8	7.5	33	54.0	60 to 70
Mild steel	40.00	2.50	3.6	3.3	146	157.0	10
Mild steel with hydrogen	41.80	1.75	2.7	2.3	0	0,0	10

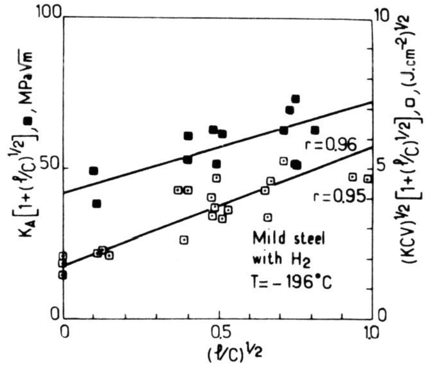

Fig. 7 . K_A/K_{Ic} $[1 + (\rho/_c)^{1/2}]$ and $(KCV/KCV_o)^{1/2}$ $[1 + (\rho/_c)^{1/2}]$ versus $(\rho/_c)^{1/2}$ for hydrogenated mild steel.

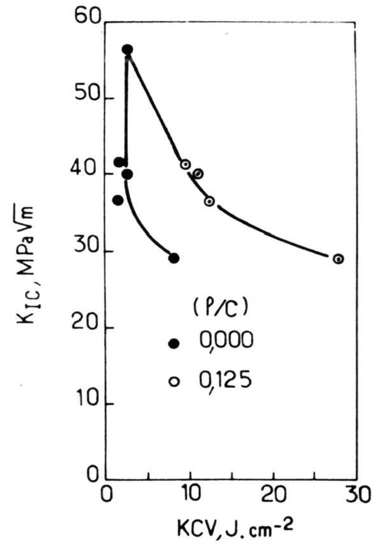

Fig. [illegible]. K_{Ic} versus KCV for noth radius 0.00 [illegible]d 0.25 mm.

From table 5, we notice that the values α or α' and β or ρ_o (K_A) and ρ_o (KCV) agree well. We observe also that when K_{Ic} increases, KCV_o and KCV at ρ = 0.25 mm decrease, figure 8. These results are in good agreement with results mentioned by Ritchie and others (1976) and with those reported by Curry and Knott (1976).

It is evident that we have to support this finding with many results achieved on other materials with different microstructural parameters. In any way, we can conclude, in spite of few results in our possession, that the mechanism of failure at notch root in three-point bending or in Charpy test are the same and this correlates well with Green and Hundy conclusions (1956).

CONCLUSION

Applying Spink and others (1973) relationship in a special representation on many steels and in various microstructural conditions, we found that $(KA/K_{Ic}).[1+(\rho/c)^{1/2}] = f(\rho/c)$ is always linear and we can then detect whether microcracks are initiated close to the notch root or at a characteristic distance ρ_o from the crack root. This new representation does not need the knowledge of σ_U value. Also, we showed that the slope of each linear relation is a function of the microstructure feature of the material studied. Furthermore when we plot $(KCV/KCV_o)^{1/2}.[1+(\rho/c)^{1/2}] = g(\rho/c)$, KCV impact Charpy energy when $\rho > 0$ and KCV_o when $\rho = 0$, the relation is also linear with slope and characteristic distance which are equivalent to those found by K_A method, for materials studied in this paper and in the brittle region on the curve (KCV - T). In addition, when K_{Ic} increases, KCV_o and KCV at ρ equal to zero and 0,25 mm decrease.
These findings have to be confirmed by studying further materials in other structural conditions and also in transition and ductile regions.

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