LOCAL CHARACTERISTICS OF DUCTILE FRACTURE SURFACE IN SPHEROIDIZED STEELS

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Abstract

It is possible to successfully propose the physical-metallurgical and structural conditions for the equilibrium between steel strength and toughness by designing structural parameters which have a positive effect on the relation between local and macroscopic fracture processes. In carbon steels and microalloyed steels for a wide range of technical uses, whose structure after heat treatment consists of a basic matrix of tempered lower bainite with precipitated carbides and sometimes other types of inclusions, toughness is dependent primarily on the size distribution of second phase particles, their volume ratio, and also the strength of the matrix/particle phase boundary and the mechanical properties of structural phases. By modelling and simulating the process of main crack formation during high-energy ductile fracture, it is possible to propose optimum physical-metallurgical and geometric parameters of steel structure in order to achieve the required relation between strength characteristics and toughness. This paper presents an analysis of results achieved in several tasks carried out to predict mechanical properties in ductile fracture, and it outlines potential future developments. The aims are to determine the limit characteristics of mechanical behaviour of structural steels which can be achieved with a view to the current structural situation and technological possibilities, and furthermore to propose future methods for determining relations between microstructure and toughness.

Keywords: ductile fracture, microalloyed steel, second phase particle, void, dimple, fracture surface, fractal dimension

1. SELF-SIMILARITY OF THE DUCTILE FRACTURE SURFACE

One very infrequently used method of studying ductile fracture in relation to designing structural steel structure and properties - a method which is especially applicable to the relation between strength characteristics and toughness - involves quantitative fractographic analysis. From evaluating the roughness of the ductile fracture surface R_s , it is evident that this surface is rougher in comparison with fracture surfaces formed by other mechanisms of crack propagation, and thus we can naturally expect a higher degree of toughness. Quantitative analysis of ductile fracture surfaces has shown that the mean actual dimple surface S_d is directly equal to the product of surface roughness R_s and mean dimple diameter A_d [1], $S_d = R_s A_d$.

Ishikawa [2] attempted to find a relation between the number of dimples Υ and their size d_D, demonstrating that $\Upsilon(d_D) \cong k_0 d_D^{-1.5}$, where k_0 = 280 μm$^{1.5}$ is the constant. However, this evaluation does not enable us to predict the relation between microstructure and fracture characteristics. The results of the first fractal analyses of cleavage fracture surfaces [3,4] were used as the basis for similar studies of ductile fracture [5-7]. Dauskardt et al. [5] clearly demonstrated that the length of the fracture profile line $L(\varepsilon)$ is not only the function of measuring step ε, but also depends on the scale regime during observation,

$$L(\varepsilon) = L_0 \varepsilon^{1-D_L}. \tag{1}$$

Later studies [1, 6] explored the dependence of the fracture surface fractal dimension D_L on the distance from the crack root. These changes in connection with the change of stress state associated with the change of distance from the crack root [7]. The shear fracture zone and the local plastic zone - the first phase of the main crack propagation mechanism - show visibly lower values for the dimension D_L than the subsequent ductile fracture zone. In the ductile rupture zone, controlled by the dimple growth and coalescence mechanism, there is an increase in dimension D_L; however, its values for the investigated Cr-Ni steel are not higher than D_L = 1.22. The final phase of main crack propagation, which occurs by means of the brittle fracture mechanism, shows a sharp decrease in the fracture profile fractal dimension, with the lowest value D_L = 1.07 [7].

2. EFFECT OF DUCTILE TOUGHNESS ON FRACTURE SURFACE FRACTALITY

Applying the rules for void growth in strengthening material with deformation strengthening coefficient n and surface particle density N_A, for the coalescence criterion $2R = \lambda$ and random, $\kappa = N_A\lambda^2 = ½$ and hexagonal $\kappa = 4/\pi$ arrangement of second phase particles, dimension D_A was found to be dependent on the ratio $2\lambda/d_p$ [6]. However, because $2\lambda/d_p = \sqrt{3\pi/(2f_V)}$, dimension D_A increases with increasing secondary phase particle volume ratio f_V. This is indirectly confirmed by analysis of the dependence of toughness J_{Ic} on the fracture line profile dimension D_L and on the volume ratio f_V in the form:

$$J_{Ic} = \frac{\sigma_0}{3} ln\left(\frac{4^{(1-D_L^{-1})}-1}{12 f_V}\right) l_0 \tag{2}$$

where σ_0 is yield strength and l_0 is characteristic distance [8]. A similar relation between toughness J_{Ic} and volume ratio f_V was also found by a recent analysis of ductile fracture in the viscoplastic solid phase strengthened by dispersed second phase particles [9]; the only difference was that the latter analysis is valid only for low volume ratios f_V, when the fracture process is controlled by the coalescence of individual voids. In the multiple void interaction controlling regime, J_{Ic} is practically independent of volume ratio f_V. Increasing surface roughness in the void-by-void dominated crack growth regime also corresponds with an increase in toughness J_{Ic} [9]. However, the search for a relation between fracture toughness of structural steels and fracture surface fractal dimension has not produced unambiguous results. Likewise, Mandelbrot et al. [10] were the first to discover that with increasing toughness the fractal dimension D_S exhibits a decreasing trend. This trend is unexpected to say the least, and it has been confirmed in steels by other authors including Imre et al. [11]. Underwood and Banerji [12], studying fracture surfaces of AISI steel after different heat treatments, found a slight indication that the fractal dimension grows with increasing fracture toughness. A similar result has been shown for steels by other researchers [8,13]. In principle there are two reasons why the character of the dependence of the fracture surface fractal dimension on the fracture characteristics of steels is not unambiguous. Firstly, there are differences between the fractal character on different scaling levels caused by different micromechanisms controlling the fracture process in different stages of development [5]. The second reason is undoubtedly the random character of the initiation of fracture mechanisms, due to the randomness of microstructure and the complex stress-deformation state at microvolumes. One consequence of this is the random microcrack propagation direction, which in steels is often controlled by two or more mechanisms, and the initiation of transitional stages, e.g. the interaction of cleavage facets [14,15].

3. SELF-AFFINITY OF THE DUCTILE FRACTURE SURFACE

The experimental assessment of fractal dimension D_A by the box-counting method [16] involves determining the dependence of the number of boxes in the box network $N(\varepsilon)$ covering the fracture surface on the box size ε. The experimental points of dependence of log $N(\varepsilon)$ on log $(1/\varepsilon)$ using the least squares method approximate a straight line. For two low values of measuring steps ε and $\delta\varepsilon$ where $\delta > 1$, it follows from Eqn. (1) that

$$N(\delta\varepsilon) = \delta^{-D_A} N(\varepsilon) \tag{3}$$

This so-called self-similar fractal fracture surface is statistically isotropic. That means that the scaling transformation of a point on the fracture surface $z(x, y)$ has the same character in all directions: $(x, y, z) \rightarrow (\lambda x, \lambda y, \lambda z)$. However, it has been shown [17] that fracture surfaces are mostly statistically invariant, suiting a self-affine scaling transformation of the forms $(x, y, z) \rightarrow (\lambda x, \lambda y, \lambda^H z)$, where $H < 1$ is the Hurst exponent or also the roughness exponent. That means that for a self-affine fracture surface, the z coordinate characterizing the surface unevenness is statistically dependent on the horizontal coordinates *x, y*, however its sensitivity to change is lower than that of coordinates *x, y*. This is also projected into the change in distance [18],

$$\Delta z \propto (\Delta x^2 + \Delta y^2)^H \quad (4)$$

where $(\Delta x^2 + \Delta y^2)^{1/2}$ is the distance from the fracture plane and Δz is the corresponding difference in the fracture surface heights.

Fractal analysis of fracture surfaces after DWTT of X70 steel specimens showed that the fractal dimension of the fracture surface decreases with increasing ductile fracture percentage, especially from D_A = 2.49 - 2.17. [19] In the investigated fracture surfaces, the Hurst exponent - which is related to the fractal dimension as $H = 3 - D_A$ [20] - ranges approximately from 0.5 to 0.8. This is entirely in accordance with the results of the analysis of crack front waves [20], which found a limit value of 0.8 for the Hurst exponent in ductile fracture. It appears that the value of the Hurst exponent increases with increasing ductile fracture percentage [19]. This means that lower values of the Hurst exponent *H*, corresponding with higher values of the fractal dimension D_A , are represented by a more segmented surface in all details at the same level of resolution. At higher ductile fracture percentages, accompanied by higher plasticity, areas of dimple ductile fracture are often covered by a layer of intensively deformed material with a paradoxically lower roughness. For this reason the Hurst exponent is higher in a fracture surface with a significant ductile fracture percentage than in a fracture surface with a significant percentage of brittle or quasi-cleavage fracture.

CONCLUSIONS

The ductile fracture surface does not have the character of a self-similar fractal but rather of a self-affine fractal; however, the Hurst exponent is dependent on the scale regime during observation.

The Hurst exponent in the ductile fracture surfaces approaches the limit value of 0.8; this corresponds with the large length scale regime of observation.

The physical model of the ductile fracture surface appears to be a suitable tool for evaluating the relation between the fracture surface fractal dimension and geometric structural parameters such as the second phase particle volume ratio, the size distribution of second phase particles, or interparticle spacing. The scale regime during observation is probably the most important parameter in determining ways to evaluate the effect of toughness on the fracture surface fractal dimension.

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