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ON POSSIBLE CAUSES OF BRITTLE FRACTURE

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The process of constructing high-pressure vessels, protective shells of chemical and nuclear reactors, ship hulls, tankers, as well as other large-scale objects includes the stage of material selection as one of the stages.

Until now the problem of brittle fracture has been solved "... by taking traditional empirical measures..." [1], "by trial and error..." [2]. Material selection includes testing of specimens of full-scale thickness by using brittle fracture mechanics methods (FM) and a transitional temperature [3, 4]. It is assumed that if the specimens (including even welds) do not fracture brittlely, brittle fracture of the full-scale object should also not be expected. Such a method of eliminating brittle fracture is most reliable because of the lack of a single brittle fracture criterion in FM [1], and as a rule, the impossibility of direct testing of full-scale objects up to fracture. Because of the indisputable successes of FM the number of brittle fractures of objects computed in conformity with existing strength norms, although sharply reduced, have not been eliminated completely. Data have appeared at this time that raise doubts about the reliability of the method of handling brittle fractures by testing materials specimens of full-scale thickness. The crux of the doubts reduce to two questions. 1. How much is the transfer of test results of standard specimens of full-scale thickness of a carrying metal to real structures representative and full-founded? 2. To what degree are the critical values of the stress intensity factors found in experiments on specimens by FM methods under plane strain K_{Ic} and the specific energy of material separation per unit surface 2γ in correspondence with their values for the self-similar stationary mode of brittle separation of a material?*

The answer to the first question is closely connected to the possibility of the appearance of geometric scale effects (SE) of an energetic nature during fracture [7, 8] independently of whether the object defects in addition to their dimensions are similar as the FM requires [9, 10] and consists of seeking the necessary conditions for which brittle fracture is possible (or impossible).

Necessary conditions for the nonfracture of pipelines are found in [11] by FM methods using I-integrals under certain simplifying assumptions and the solution of an additional problem. Analysis of this solution in [12] showed that the fundamental meaning of the solution found in [11] is the existence of SE of energetic nature, that follow directly from the

*Precisely in such a fracture mode are plane strain conditions satisfied strictly and the value 2γ reaches its minimum and is independent of the specimen thickness [5, 6].

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integral energetic approach to the object as a whole or its characteristic part [7, 8] and that the contribution of the inertial and temperature terms to the solution found is negligibly small.

Therefore, determination of the elastic energy balance of an object and the work expended in its brittle fracture permits the necessary brittle fracture (nonfracture) condition, missing in FM [1, 8], to be obtained.

The brittle nonfracture condition found mathematically in [11, 12] for pipes of small relative thickness after replacement of the internal pressure in the pipe by a circumferential stress σ by means of the Mariott formula and discarding inessential terms is

$$\sigma < \left[\frac{2\gamma E}{\pi R\,(1-\nu^2)}\right]^{0.5}, \tag{1}$$

where E is the Young's modulus, ν is the Poisson ratio, and R is the pipe radius. It follows from (1) that the boundary value of σ at which the absence of brittle fracture is assured depends substantially on R: the greater the R the smaller the values of σ for which brittle fracture is possible (other conditions being equal).

To what degree should the dependence (1) be taken into account in selecting the dimensions of standard specimens when seeking K_{Ic} in FM? Taking into account that the pipe circumference is equivalent to the extent of the specimen L in a direction perpendicular to a crack, by substituting $R = L/2\pi$, writing the necessary brittle fracture condition by changing the < sign to $\geqslant$ in (1) and solving it for L, we obtain

$$L \geqslant 2(K_{Ic}/\sigma)^2. \tag{2}$$

The constraint $t \geq 2.5(K_{Ic}/\sigma_{0.2})^2$ is imposed on the thickness t of standard specimens in FM to determine K_{Ic}. We are interested just in the necessary fracture (nonfracture) condition; consequently, we do not examine the requirement on the dimension of the initial crack.

The value of L should either equal t (HRS specimens of type X) or L = 2.5t (HRS specimens of type T or TST specimens) [13, 14] so that the last inequality is written as

$$L \geqslant (2.5 \ldots 6.25)(K_{Ic}/\sigma_{0.2})^2. \tag{3}$$

The essential distinction between inequalities (2) and (3) is σ and $\sigma_{0.2}$. Thus, for material specimens satisfying (3), the brittle fracture condition is satisfied only for $\sigma \sim \sigma_{0.2}$. However, there is a real possibility of brittle fracture even for a σ significantly less than $\sigma_{0.2}$, which indeed cannot be realized during testing of such specimens.

For the same reason, results of testing material specimens or smaller-radius pipes of the same thickness as for a real pipeline do not yield the correct answer to the question of the pipeline reliability. This deduction is apparently valid even for other objects listed earlier and indicates the necessity to take account of the SE of energetic nature to avoid brittle fracture. The reasoning elucidated above is evidently also valid for welded structures and permits understanding of why "... in certain cases the stress level for which a crack is formed is so low that there is doubt even of the possibility of the origination of brittle fracture..." [15], and also answering the question of why in a number of cases "...fracture is propagated through whole sheets, and not along seams..." [16].

Let us note that above, as in [11, 12], energy dissipation by the internal friction in the metal was not taken into account during transfer of the elastic energy to the developing crack. It is not excluded that this effect will become noticeable for large L so that at a certain stage the appearance of the SE of energetic nature can be limited.

Let us turn to the second question. Several methods exist to determine the specific work of material separation per unit surface: a) testing standard material specimens (for example, Sharp-type specimens) by the impact fracture viscosity a_n; b) determination of the coefficient K_{Ic} by FM methods (see above) with a subsequent unique conversion to the desired value 2γ (or G_{Ic}):

$$2\gamma\,(\text{or } G_{Ic}) = K_{Ic}^2\,(1-\nu^2)/E \tag{4}$$

(ν is the Poisson ratio); c) determination of the critical stress of material separation σ_S under one-dimensional strain and multilateral tension in tests on shockwave spall with subsequent conversion to the specific work in material separations λ. In an acoustic approximation, the value of λ is found according to the expression [8]

$$\lambda = \sigma_s^2 \delta (1 + \nu)(1 - 2\nu)/[2\beta E(1 - \nu)], \tag{5}$$

where $\beta = 3$ for a triangular loading pressure pulse and $\beta = 1$ for a rectangular pulse, δ is the thickness of the spalling layer of material. Since the determination of λ is performed in tests on high-velocity loading of the material under the effective impulse times of $\tau \sim 10^{-5}$-10^{-6} sec (the characteristic loading pulse length is ~1 cm), the question arises of the necessity of introducing corrections taking account of $\lambda(\tau)$. However, a study of the possible dependence $\lambda(\tau)$ executed in [17] showed that it is quite weak and its examination at this stage can be neglected.

In experiments on spalling separation of materials a strictly one-dimensional strain is realized that is equivalent to ideal "contraction" of the material which is actually not achievable under static conditions, which is the indubitable advantage of this method, as also small possible elastic energy expenditures on plastic flow [8]. Preliminary shock-wave compression of the material, which is capable of changing its structure and heating up, before rupture should be among the specifics of the spall method in addition to the high rate of deformation. However, these factors can be taken into account. Moreover, the author of [18], on remarking the closeness between the physical mechanism of fracture during spall and crack propagation, poses the question of using spall to determine K_{Ic} in plastic materials.

A comparison, performed for soft steels, of values of a_n found experimentally in the cold brittleness domain, 2γ (or G_{Ic}) for T ~ −200°C and λ indicates their nearby values that lie in the range 5-20 J/cm² [7, 8, 18, 19]. Taking account of the multivariety of factors able to exert influence on the final result of the measurement (texture, rolling direction, heat treatment, specimen size, etc.) and the not-so-great volume of investigations, such a spread of data is totally natural and in a specific sense permits considering the characteristics under consideration as fundamental material fracture parameters just as K_{Ic} in FM. The closeness of the values a_n, 2γ, and λ is even traced for other materials at the low T for which there are appropriate data (titanium and aluminum alloys).

The situation is abruptly made complicated upon comparing the behavior of the temperature dependences of a_n, 2γ, λ. From general physics considerations, it is natural that a drop in the specific work on material rupture should have been expected as T rises to melting. Indeed, the dropping dependence $\lambda(T)$ is observed on tests on spall rupture of materials [20] and in systematic investigations of $\sigma_p(T)$ [21, 22]* for not only solid but also for liquid materials [23] if certain sections of $\lambda(T)$ are not taken into account that depart from the general behavior of the curves and, as is shown in [23], are related to phase transitions. Thus, according to [20], the dependence $\lambda = (0.4-0.55\ T/T_m)\cdot 10^5$ J/m² is valid for copper in the range $T/T_m = 0.05$-0.6.

Turning to $a_n(T)$ as well as the dependences $2\gamma(T)$ obtained from $K_{Ic}(T)$ [14, 24], we arrive at the conclusion that they grow at T increases, in contrast to $\lambda(T)$. What is the reason for such a discrepancy?

It is easy to understand the nature of the growing dependence $a_n(T)$. Indeed, as a rule, as T rises $\sigma_{0.2}$ drops and since the sizes of the specimens being tested are unchanged, then the elastic energy reserve of the specimen also drops, which turns out to be insufficient for brittle fracture of the material even for $\sigma \sim \sigma_{0.2}$. Consequently, as T rises a greater part of the communicated energy is expended in the work of plastic deformation of the specimen as a whole in the bulk of the material, which has no direct relation to the work of material separation on a certain surface. In principle, the "cold-shortness" domain could be shifted to the domain of higher T by increasing the specimen size and using the effect of material "embrittlement" as the object size grows, a SE of energetic nature. So that a_n can be utilized only in the cold-shortness domain as the specific work of material separation.

It is more complicated to understand the growth of K_{Ic} and 2γ as T increases. Indeed, despite the constraints imposed on the size of standard specimens in FM [13, 14] of the form (3), which substantially follows from the constancy or nondiminution of the ratio of elastic energy in the specimen to the work of fracture, the value of 2γ grows as T increases. Thus, according to [24] the value of 2γ grows from 5 to 38 J/cm² for A216 steel during the transition from T = −150°C to T = −20°C. An analogous pattern is conserved for the steels 28Kh3SNMVFA [24], 15Kh2MFa [24], A533 [14] as well as other metals [10, 13, 14, 24]. This growth is less definite, it is true, for less plastic metals (as titanium alloys). The form of the dynamic functions $K_{Id}(T)$ [25] also confirms the growing dependence $2\gamma(T)$.

*According to (5), other conditions being equal $\sigma_s^2 \sim \lambda E$.

Without subjecting the important contribution to the fracture problem of the standard material testing methods taken in FM to doubt, it is impossible not to turn attention to certain false circumstances also. The methods of determining K_{Ic} in FM do not assure achievement of the self-similarity mode, stationarity of the plastic flow domain in the crack opening for which K_{Ic} and 2γ, respectively, can be taken as characteristics of the material. Only under conditions of rapidly propagating cracks [5, 6] is one-dimensional strain achieved under multilateral tension, the same stress state as for shockwave spalling fracture and elastic energy expenditures in unloading through the specimen-free boundary are reduced to a minimum.

In the light of the above elucidation it is impossible to exclude that as T grows the self-similar stationary fracture mode will be achieved for laboratory tests on specimens where SE of an energetic nature should be manifested more strongly, as will necessarily result in higher rates of fracture development, localization of the plastic flow domain in the crack mouth, and diminution of 2γ because of the achievement of the stationary fracture mode. The fact that "... for so-called brittle fracture under laboratory conditions, a strain is detected that exceeds 10% (with a certain reserve) at the notch apex while the strain during working conditions during fracture does not exceed 2%..." indicates the possibility of realizing such fracture [2]. Therefore, the value of 2γ during material fracture in the stationary self-similar mode and for elevated T can turn out to be substantially less than follows from K_{Ic} measured by FM methods on standard specimens for T above the cold-shortness T; or in other words, the K_{Ic} themselves can turn out to be exaggerated.

In conclusion, let us note the following:

1. The absence of brittle (quasibrittle) fracture for laboratory tests by FM methods on standard material specimens of full-scale thickness does not yield assurance of the brittle (quasibrittle) fracture of carrying elements of the full-scale object.

2. A sharp difference in the form of the temperature dependences of the specific work in the fracture of metals λ and 2γ found from tests by shockwave spall and from laboratory investigations of K_{Ic} by FM methods yields a foundation for assuming that K_{Ic} and, respectively, 2γ are subtantially exaggerated far from the cold-shortness limit for metals.

3. The strength margin of large-scale objects determined by FM methods can turn out to be exaggerated substantially in comparison with their actual values, which may be a reason for unpredictable catastrophic fractures of such objects.

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PROPAGATION OF RAYLEIGH WAVES IN DISSIPATIVE MEDIA. LINE SOURCE

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Surface acoustic waves and, in particular, Rayleigh waves have attracted the attention of researchers in a number of areas of science and technology, such as seismologists and creators of microelectronic technology. This is due to the specific features of surface waves. Thus, in seismological measurements of a wave field at large distances from the source its most clearly recorded component is often the surface wave, due to the fact that its damping is weaker than that of bulk waves. In microelectronics one uses the fact that surface waves, possessing a low velocity in comparison with the propagation velocity of electromagnetic waves, as well as localization in the near-surface layer, make it possible to create very compact electron devices. In both seismology and microelectronics it is important to know how the wave profile and amplitude change with its propagation in the medium. These variations may be a consequence of the dispersion-dissipative properties of the medium. The necessity of keeping these properties in mind becomes clear if it is taken into account that in seismology one deals with wave propagation at very large distances, while in microelectronics one uses very short pulses. This brings about the possibility of a strong effect of dispersion-dissipative medium properties on the propagation of surface waves.

There exists a number of studies devoted to the study of the effect of nonideal medium properties on propagation of Rayleigh waves [1-4]. For example, a relation was found in [2] between small corrections to the wave vector of a monochromatic Rayleigh wave and similar corrections to the wave vectors of longitudinal and transverse bulk waves. A numerical calculation was carried out in [3] of the profile of a Rayleigh wave, excited by a shock near the surface. The dispersion-dissipative properties of the medium were accounted for by a linear frequency dependence of the imaginary part of the wave vectors of longitudinal and transverse waves. It was noted in [4] that substantial difficulties arise in attempting to account for the effect of dispersion-dissipative properties of the medium on the wave field evolution in half-space by using the mathematical apparatus developed for an elastic medium. A successive method of calculating the wave fields in dispersion-dissipative media was first

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