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RELIABILITY OF HIGH-STRENGTH STEELS

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In Nos. 7 and 8 of our journal two articles [1, 2] were published in which it was recommended that medium-carbon steels (0.25 – 0.40% C) alloyed with various elements (5 – 6% in total) be used. Heat treatment of such steels (quenching + tempering) provides a high strength that depends on the composition and especially the tempering temperature, namely, $\sigma_r = 1500$ N/mm^2 and even $\sigma_r = 2000$ N/mm^2. The admissibility of the use of steels with this strength is explained by the authors by their high ductility ($\delta \cong 8\%$, $\psi \cong 40\%$) and impact toughness ($a_1 = 60 - 80$ J/cm^2). These data were obtained in tensile and impact tests at room temperature.

It is known that the quantitative results of tests depend considerably on their conditions. The results of tests characterize the properties of the specimen and depend on its configuration, dimensions, and form of loading. Uniaxial tensile stress creates conditions for determining the maximum ductility of the specimen. Under actual conditions parts produced from the tested material commonly have a quite different shape and a different deformability.

It should be noted that for the overwhelming majority of actual parts in mechanical engineering a deformation exceeding 1% is inadmissible, and therefore the ductility values obtained in a tensile test of a specimen are not indicative enough (we do not deny their significance for the production engineer).

There are cases where parts produced from soft iron with $\psi \cong 50\%$ undergo brittle fracture.

I am far from neglecting mechanical tests at all. Of course, they are an objective estimation of the quality of the metal, but the results do not characterize its reliability as a structural material. Unfortunate, but true.

The results of standard tests would be different if we were to change the shape of the specimen and the loading conditions. However, it is impossible to take into account all possible shapes of parts and forms of their loading in running monitoring of the metal. In some cases full-scale mechanical tests of the part are conducted although they are very expensive.

Here I should apologize for speaking about routine and universally known (for a metal scientist) matters.

The point is that any steel with any composition, structure, and heat treatment (except for purely austenitic steels) can fracture by *two* different mechanisms, namely, *ductile* and *brittle*. Specific properties of the metal are monitored for each form of fracture. Unfortunately, these two mechanisms are seldom distinguished when analyzing the tensile diagram.

During a tensile test of a metal with a certain ductility the atoms are moved apart in the crystal lattice (their mean position increases by tenths of a percent). After a certain elastic deformation the metal changes structure (shear, motion of dislocations), i.e., undergoes plastic deformation (changes shape). It is known that this is described by a curve of type *1* shown in Fig. 1. If the load is removed in the test, only plastic deformation remains in the specimen (segment 0*a* in Fig. 1). In practice the residual deformation is limited to 0.2%. The stress at which this deformation is observed is called the yield limit ($\sigma_{0.2}$).

The yield limit in static loading is the principal characteristic used for strength calculations. It should be noted that in tensile tests (uniaxial) $\sigma_{0.2}$ commonly has a minimum value compared to similar characteristics obtained by other test methods.

It should be noted that loading is by far not always accompanied by motion of dislocations, i.e., plastic deformation does not always occur. The process is very sensitive to many factors such as the configuration of the object studied,

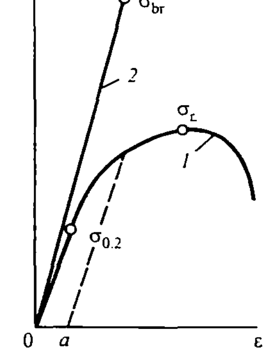

Fig. 1. A tensile diagram.

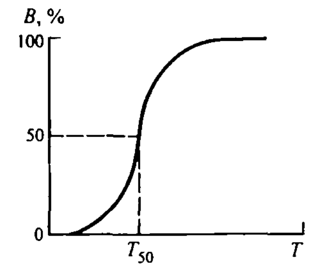

Fig. 2. A sequential curve describing the fraction of fiber component in a fracture.

the form of stress state, the rate of growth of the stresses, the temperature, etc. It often happens in practice that plastic deformation is hindered or even absent. In such cases the deformation can only be elastic (curve *2* in Fig. 1).

It is assumed that if the distance between atoms increases by 10 – 15%, the bond between them will break, causing brittle fracture (without plastic deformation). This will occur only at a very high stress that exceeds $\sigma_{0.2}$ by hundreds of times, i.e., at a stress corresponding to the so-called theoretical strength. In actual practice fracture occurs much before the theoretical strength is attained.

The reason behind this phenomenon was established by A. Griffith in 1925. He assumed (for glass, but later this concept was generalized to metals) that a material possesses defects, i.e., very sharp cracks, at the tip of which a stress equal to the theoretical strength appears. In other words, the stress in a concentrator attains the theoretical strength [3], namely,

$$\sigma_r K = \sigma_{theor},$$

where σ_r is the rated (nominal) stress and K is the concentrator, characterized by the length of the crack l and its sharpness r, or

$$\sigma_r \frac{\sqrt{l}}{\sqrt{r}} = \sigma_{theor}.$$

In other words, the presence of crack-type defects in a material gives rises to a local stress, and when the theoretical strength is attained, brittle fracture occurs (begins).

Based on these considerations, we can determine approximately the brittle strength (σ_{br}), which has earlier been called (A. Ioffe, G. Uzhik) the rupture strength ($\sigma_{br} = \sigma_{rup}$).

σ_{br} is a characteristic that is independent of the structure and the composition because these parameters do not affect the theoretical strength either (the latter is mainly connected with the melting temperature).

Using l and r, we can predict, to a certain degree, and control (approximately) the quantity σ_{br} (for details see [3] and my earlier works on this problem).

Since we cannot yet measure the length l or the sharpness r of a crack, the value of the brittle strength has not been used as a parameter determining the resistance to brittle fracture.

So far we have not had to resort to the limiting value of the admissible stress but rather have had to avoid factors that have a negative effect on the capacity of the metal for plastic deformation. Tests at a high deformation rate with simultaneous variation of the stress, tests with multiaxial loading, and other hard-to-realize tests are only placed on the agenda.

However, there is a factor that affects considerably the plastic deformability of the metal and, hence, characterizes its susceptibility to embrittlement. As the temperature in the fracture of a specimen decreases, areas of brittle fracture appear and finally the fracture becomes fully brittle (Fig. 2). The form of the fracture changes most abruptly at a 1 : 1 proportion of brittle and ductile regions. Therefore, the cold-brittleness threshold should be determined by the presence of a 50% ductile component in the fracture and can be called the *semibrittleness temperature* T_{50} [4].

A direct estimate of the cold-brittleness threshold from the appearance of the fracture is quite satisfactory (the difference in the visual determinations of the proportion of fibers in a fracture by experienced specialists does not exceed 5%). However, this is possible only for steels with a low strength (at most 1000 N/mm^2). For stronger steels an electron or scanning microscope is required.

The impact ductileness should not be used to determine the temperature of the transition to a brittle state, because a change in the toughness properties is connected with a change not only in the amounts of the fracture components but also in the properties of the components themselves, which can differ for steels differing in structure and composition.

In accordance with the diagram in Fig. 1 the value of $\sigma_{0.2}$, even in simple stretching should not exceed σ_{br}, because the stress will attain σ_{br}, passing by $\sigma_{0.2}$, i.e., without plastic deformation of the metal.

Since we cannot yet determine σ_{br} with the requisite accuracy and objectivity (we cannot determine the sharpness r of the crack and can only assume for its length that for a starting crack it is equal to the grain diameter), we should limit ourselves, in choosing the admissible load, to determining the susceptibility of the steel to brittle fracture by estimating the semibrittleness temperature T_{50}.

Here we should proceed from elementary concepts known from textbooks [5].

If T_{50} of the steel lies in the domain of positive temperatures, the steel should not be recommended for use.

A steel with T_{50} ranging between 0 and –20 °C can have limited use because there exists a danger of brittle fracture. A steel with T_{50} ranging between –20 and –40 °C can be assumed to resist "brittle" fracture reliably enough. For a steel with T_{50} ranging between –40 and –60 °C the probability of brittle fracture at + 20°C is virtually zero and it can be used for critical parts.

Unfortunately, it is often assumed that if a steel does not operate at a negative temperature there is no need to test it under these conditions.

In his time G. Pogodin-Alekseev has shown convincingly that such low-temperature tests are important for steels operating at room temperature, and the reasoning above confirms this opinion.

Thus, the main criterion of the reliability of a steel consists in preventing brittle fracture and, in practice, attaining a sufficiently low cold-brittleness threshold.

Ya. Potak has studied causes of fracture and collected data on 2000 such cases. In all the cases the fracture was brittle (although in one case a bolt underwent ductile fracture because it has not been quenched).

The authors of [2] work at the same enterprise where Ya. Potak once worked and it is difficult to explain how without studying the possibility of brittle fracture they consider it possible to increase the strength without guaranteeing the absence of brittle fracture.

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