SCIENTIFIC AND TECHNICAL SECTION

PRELIMINARY COMPRESSION OF A MATERIAL AS A FACTOR IN CHANGING THE BRITTLE FRACTURE MECHANISM FOR BCC METALS

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We have investigated the effect of preliminary plastic compression on the characteristics of brittle fracture in bcc metals using 15Kh2MFA pearlitic steel as an example. We have experimentally established that preliminary plastic compression of the material leads to a change in the mechanism of brittle fracture from transcrystallite to intercrystallite. We have shown that the critical stress for brittle fracture is significantly lower for the intercrystallite mechanism than for the transcrystallite mechanism. We suggest mechanisms for embrittlement of material as a result of its preliminary plastic compression. We present schemes for the transition from intercrystallite fracture to transcrystallite, depending on the thermal and mechanical loading conditions for the precompressed material.

INTRODUCTION

Beginning with the work of A. F. Ioffe (1924) [1], the concept of brittle fracture of bcc metals has continued to be developed. This concept is based on some critical stress S_c, the so-called resistance to tear [2-5] or microspalling [6], or the critical stress for brittle fracture [7,8]. We know that based on the independence of the parameter S_c relative to the temperature and rigidity of the stressed state, a number of phenomena have been explained connected with prediction of the brittle-to-ductile transition temperature, the effect of notches on embrittlement of metal [1-6], and also prediction of the temperature dependence of the fracture toughness [7-8]. Along with investigations devoted to the phenomenological aspects of brittle fracture, in many papers the physical nature of this process has been studied [6-11].

The listed series of papers led to a rather inflexible point of view, connected with the formulation of the brittle fracture criterion in the form of two conditions:

$$\sigma_i \geq \sigma_y; \tag{1a}$$

$$\sigma_1 \geq S_c(\varepsilon^p), \tag{1b}$$

where σ_i is the stress intensity; σ_y is the yield stress; σ_1 is the maximum principal stress; S_c is the critical stress for brittle fracture; ε^p is the plastic strain.

The first equation expresses the necessary condition for brittle fracture: initiation of microcracks. The second equation expresses the sufficient condition for brittle fracture: displacement and propagation of microcracks on a scale greater than the grain size.

We have shown that the criterion (1) does not always yield adequate predictions, so in [12-14] we proposed a more general formulation of the criterion for brittle fracture:

$$\sigma_1 + m_{T\varepsilon}(\sigma_i - \sigma_0) \geq \sigma_d; \tag{2a}$$

$$\sigma_1 \geq S_c(\chi), \tag{2b}$$

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TABLE 1. Mechanical Properties of Steel 15Kh2MFA at Different Temperatures*

T, °C	σ_y, MPa	σ_B, MPa	δ_p, %	ψ, %	ε_f, %	S_k, MPa
−196	1070	1100	11,0	17,0	18,6	1310
−140	790	892	10,0	54,2	78,2	1488
−100	670	794	9,9	67,5	112,3	1544
−60	627	754	10,0	71,1	124,1	1643
20	560	665	7,4	74,1	135,0	1512

*Averaged over three specimens.

TABLE 2. Values of Parameters of the Stress–Strain Diagrams for Steel 15Kh2MFA at Different Temperatures

T, °C	σ_y, MPa	A_0, MPa	n
−196	1070	515	0,542
−140	790	567	0,514
−100	670	602	0,472
−60	627	656	0,512
20	560	592	0,505

where

$$m_{T\varepsilon} = m_T m_\varepsilon; \tag{3}$$

$$m_\varepsilon = \left[C_1^* + C_2^* \exp(-A_g \chi)\right]^{1/2}. \tag{4}$$

(Note that criterion (1) is a special case of (2).)

In the dependences (2)-(4), we use the following symbols: σ_0 is the dislocation frictional stress, which is the sum of the Peierls–Nabarro stresses and the resistance to slip, due to interaction between the dislocations and impurity atoms, point defects, and the original dislocations (σ_0 corresponds to the onset of plastic flow in the grain); $m_{T\varepsilon}$ is a parameter which can be interpreted as the local stress concentration coefficient at the head of a dislocation pileup; σ_d, depending on the specific mechanism for generation of microcracks, determines the strength of the matrix or the inclusion or the matrix–inclusion joints; $\chi = \int \overline{d\varepsilon_i^p}$ is the Odqvist parameter; $\overline{d\varepsilon_i^p}$ is the intensity of the plastic strain increment; $m_T = m_T(T)$ is a temperature-dependent parameter of the material; C_1^*, C_2^*, A_g are material constants.

The fundamental difference between the formulated criterion (2) and criterion (1) involves the following. In criterion (1), it is assumed that microcrack initiation occurs at the yield stress of the material, while from criterion (2) it follows that the moment of initiation of microcracks depends on the thermal and mechanical loading conditions. Therefore according to (2), depending on the temperature and rigidity of the stressed state, the process of microcrack initiation can begin both at stresses less than σ_y ($\sigma_0 < \sigma_i < \sigma_y$) and at $\sigma_i > \sigma_y$. The dependence of S_c on the plastic strain in (1) is usually connected only with tensile strains (the deformation history is not considered). In (2), the connection between S_c and the Odqvist parameter (i.e., with a parameter reflecting the deformation history of the material) is substantiated experimentally and theoretically.

The function $S_c(\chi)$ in (26) is a monotonically increasing function. The increase in S_c with an increase in accumulated plastic strain χ is connected with formation of a deformation substructure in the material, the boundaries of which serve as barriers to propagation of transcrystallite spalling microcracks [12-14]. Our previous experimental investigations of the effect of preliminary deformation on S_c showed that both in plastic deformation by tension and in cyclic plastic deformation with variable sign, there exists a single monotonically increasing $S_c(\chi)$ dependence. At the same time, individual papers are available [5, 15] in which a reduction in S_c has been obtained after preliminary plastic deformation by compression; no satisfactory explanation has been given so far for this phenomenon.

Accordingly, the goal of this work was to investigate the characteristics and mechanisms of brittle fracture of a material (using the pearlitic steel 15Kh2MFA as an example) after preliminary plastic deformation by compression.

Experimental Technique and Results. Mechanical tests. The effect of preliminary compression on S_c was investigated in the steel 15Kh2MFA. In a series of preliminary experiments for the metal in the as-delivered (original) state, by testing smooth cylindrical specimens (length of the gauge section equal to 25 mm, diameter 5 mm) under tension in the

TABLE 3. Results of Tests on Precompressed Specimens (averaged over four to five specimens)

Preliminary strain ε_{com}^{p}, %	Plastic strain at the moment of break under tension ε_f, %	True stresses at the moment of break S_c', MPa	S_c'/S_0
10	1,0	1160	0,95
20	0,0	1027	0,84
30	0,0	600	0,49
40	0,0	535	0,44

Note. S_0 is the minimum (for $\varepsilon^p = 0$) critical stress for brittle fracture of the material in the original state.

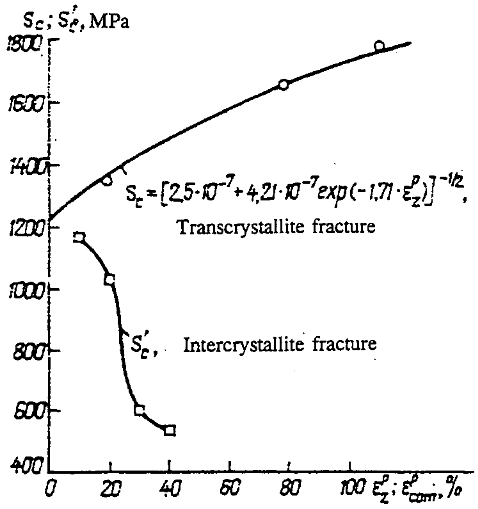


Fig. 1. The critical stress for brittle fracture under tension (along the z axis) vs. plastic strain for specimens in the original state (S_c) and precompressed (S_c'): for the $S_c(\varepsilon_z^p)$ dependence, ε_z^p is the tensile strain in the z direction, testing temperature $-100°C$ to $-196°C$; for the $S_c'(\varepsilon_{com}^p)$ dependence, ε_{com}^p is the preliminary compressive strain along the z axis, testing temperature $-196°C$; the points indicate the experimental data, averaged over three to five specimens.

range $-196°C$ to $20°C$, we determined the following characteristics: yield stress $\sigma_y = \sigma_{0.2}$, breaking stress σ_B, uniform elongation δ_p, relative narrowing ψ, true breaking stress S_k, strain limit ε_f. The values of the mechanical characteristics obtained for the steel 15Kh2MFA in the original state are presented in Table 1. Using the results of the indicated experiments, according to the technique outlined previously in [14] we determined the function $S_c(c)$ (Fig. 1), which is described well by the dependence

$$S_c = \left[C_1^* + C_2^* \exp(-A_g \chi)\right]^{-1/2}, \tag{5}$$

where $C_1^* = 2.5 \cdot 10^{-7}$ MPa^{-2}, $C_2^* = 4.21 \cdot 10^{-7}$ MPa^{-2}, $A_g = 1.71$. In this case, $\chi = \varepsilon_z^p$, where ε_z^p is the plastic strain in the direction of tension.

The characteristics obtained in testing the specimens (Table 1) were used to calculate the coefficients in the dependence

$$\sigma_i = \sigma_y + A_0(\varepsilon_i^p)^n \tag{6}$$

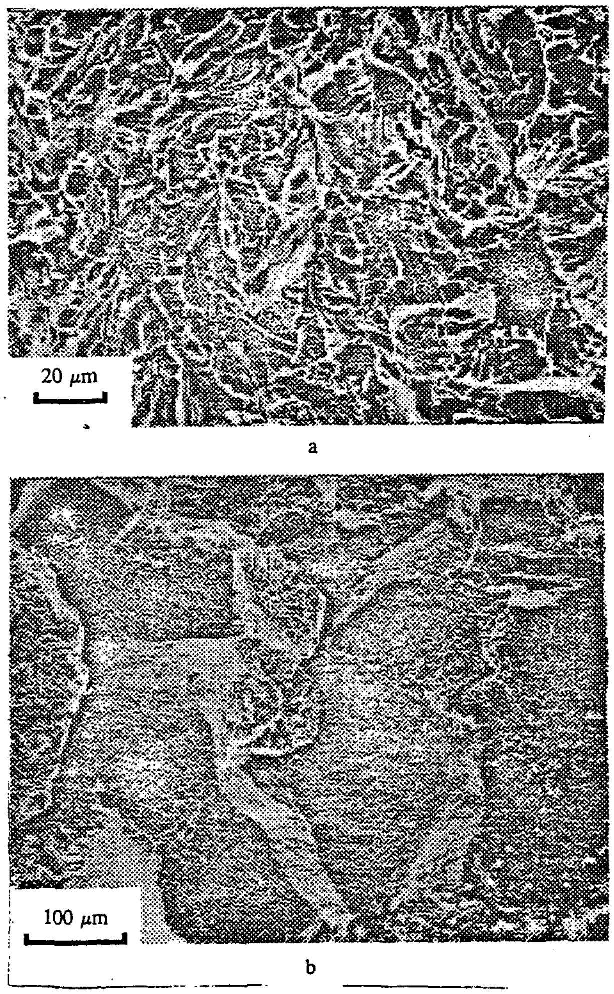


Fig. 2. Structure of the fracture surface for specimens of steel 15Kh2MFA, tested under tension at −196°C in the original state (a) and after preliminary compression to $\varepsilon_{com}{}^{p} = 30\%$ (b): a — transcrystallite microspalling; b — section of intercrystallite brittle fracture.

($\varepsilon_i{}^p$ is the plastic strain intensity), describing the stress−strain diagram (Table 2).

The basic tests were performed on cylindrical specimens (length of the gauge section equal to 20 mm, diameter 4 mm), which were cut from blanks precompressed at T = 20°C down to different percentage strain: $\varepsilon_{com}{}^{p}$ = 10, 20, 30, and 40%. A series of four to five blanks was subjected to the same preliminary deformation. Cylindrical specimens of precompressed metal were tested under conditions of tension at T = −196°C. The directions of tension and preliminary compression were collinear.

The test results are presented in Table 3 and in Fig. 1. From these data we see that preliminary compression leads to significant embrittlement of the material and reduction of S_c compared with the original state.

Fractographic investigations of the fracture surfaces of the specimens in the original state and predeformed by compression, subjected to tension tests at −196°C, were performed on a scanning electron microscope. In this case we established that brittle fracture of specimens of steel 15Kh2MFA in the original state at low temperature occurs according to a mechanism of transcrystallite spalling and microspalling (Fig. 2a). This mechanism is typical not only for pearlitic steels of this class [11, 14], but also for other bcc metals [8, 11]. Preliminary cold working under tension or by variable-sign cyclic deformation does not affect the indicated mechanism of brittle fracture in pearlitic steels [14].

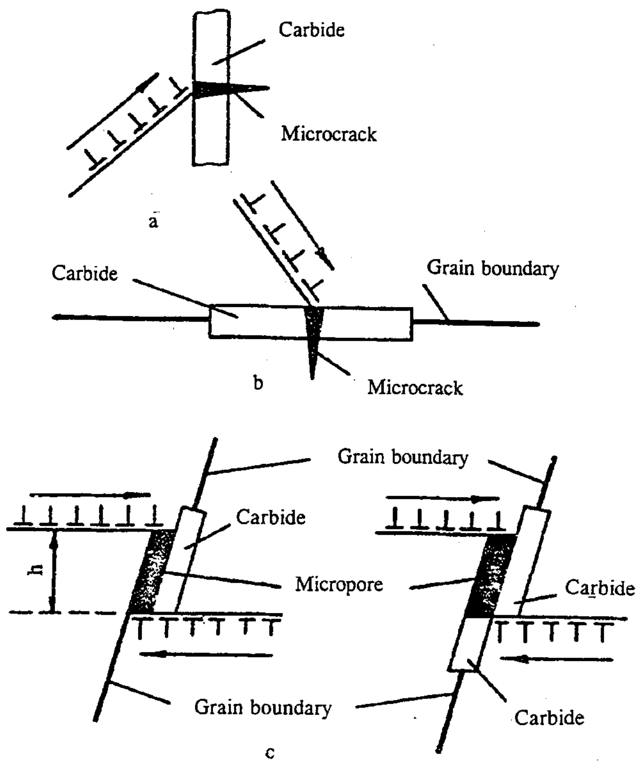


Fig. 3. Scheme for initiation of transcrystallite microcracks in carbides located within the grain body (a) and along the grain boundary (b), and also intercrystallite micropores along the grain boundary (c).

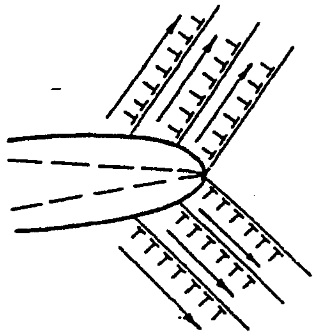
A fundamentally different structure for the fractures is observed for precompressed specimens (Fig. 2b), the fracture surface of which is characterized by the presence of intercrystallite brittle sections. The fraction of such sections in the fracture increases with an increase in the preliminary deformation by compression. Thus for $\varepsilon_{com}{}^{p}$ = 10% and 20, it is about 10-20% of the total fracture surface; for $\varepsilon_{com}{}^{p}$ = 30% and 40%, it is about 50%. The fracture of the rest of the fracture surface represents facets of transcrystallite microspalling. Thus brittle fracture under tension of precompressed specimens occurs according to a mechanism of transcrystallite spalling and intercrystallite brittle tearing fracture.

The fractographic analysis data presented allow us to conclude that a change occurs in the fracture mechanism of pearlitic steel due to preliminary compression.

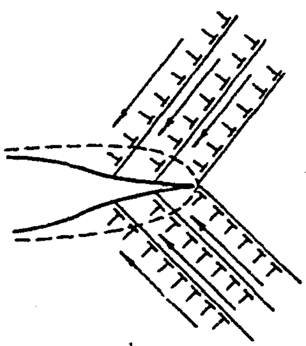
We must note that we do not know of any papers providing evidence for such a change in the mechanism of brittle fracture (from transcrystallite to intercrystallite) as a result of preliminary plastic deformation of the material.

Discussion of Results. Description of the process of brittle fracture. As follows from the experimental investigations, preliminary deformation by compression of steel in which brittle fracture in the original state occurs according to a transcrystallite spalling mechanism leads to a change in the fracture mechanism from transcrystallite to intercrystallite. Papers published earlier by us [12-14] and by other investigators [6] show that preliminary tension does not change the brittle fracture mechanism (transcrystallite spalling), and the value of S_c significantly increases with an increase in preliminary deformation. The transcrystallite nature of the brittle fracture of the material in the original state suggests that the material under consideration has very strong grain boundaries and does not tend toward intercrystallite fracture due to low cohesive strength of the grain boundaries. In the following, we will consider the brittle fracture of specifically such a material. Considering the indicated circumstances, and also the fact that initiation of discontinuities is controlled by shear deformation and does not depend on the sign of the plastic deformation (tension or compression), let us propose the following model explaining the results obtained.

In deformation of the steel, microdiscontinuities are mainly initiated in the carbides located both within the body of the grain and along the grain boundaries [16]. Analysis of the possible mechanisms for initiation of microdiscontinuities [16, 17] leads to such a conclusion. The initiated microdiscontinuities may be both transcrystallite and intercrystallite in nature.



a



b

Fig. 4.. Scheme for blunting of a sharp microcrack under tension (a) and tapering of the micropore under compression (b).

The transcrystallite nuclei are formed according to a mechanism involving splitting of the carbides by dislocation pileups and represent sharp microcracks (Fig. 3a, b). Intercrystallite nuclei represent pores (Fig. 3c).

The different geometry of the initiated discontinuities is due to the following circumstances. In most cases, formation of microdiscontinuities in steels occurs by cracking of the carbides, and not as a result of delamination along the carbide–matrix boundary [8, 16]. Accordingly, we can suppose that the strength of the carbide σ_d^c is less than the strength of the carbide–matrix bond s_d^{c-m}. For a dislocation pileup (Fig. 3a, b), the local stresses at the head of the pileup are less than σ_d^{c-m}, but reach σ_d^c. Higher local stresses (for the same effective stresses) are realized in formation of antiparallel pileups (Fig. 3c) [17]. In this case, rupture of the carbide–matrix bond and initiation of a discontinuity are possible, but the latter will represent a pore.

Thus for any plastic deformation (compression, extension, shear), initiation of discontinuities of different types occurs: sharp transcrystallite microcracks and intercrystallite pores.

As shown in [12-14], only sharp microcracks, capable of developing in an unstable manner, can be initiators of brittle fracture; pore development will occur according to a plastic mechanism. Therefore brittle fracture of the material under consideration in the initial state will be transcrystallite, and the critical S_c will be determined by the condition when the running transcrystallite microcracks can overcome the barriers created by the deformation substructure. With an increase in plastic strain, a higher stress is needed for breaking through the barriers of deformation origin [12-14]. Therefore with an increase in ε^p, the value of S_c increases.

Fracture of specimens predeformed by tension will be accomplished analogously to the fracture of specimens without preliminary deformation. In the case of preliminary tension, the intercrystallite pores initiated in the early stages grow plastically; the transcrystallite microcracks are blunted as a result of the dislocation sink at their tips (Fig. 4a). Simultaneously with initiation of microdiscontinuities, a deformation substructure will be formed. Thus in a material subjected to preliminary tension, there are none of the sharp microcracks needed for realization of brittle fracture. Therefore for such fracture of specimens predeformed by tension (as for samples in the original state), three conditions must be satisfied: initiation of new sharp microcracks, displacement of these new microcracks, and their propagation through the deformation barriers. Since, as for the original specimens, new sharp microcracks will be only transcrystallite, the fracture will also be transcrystallite in

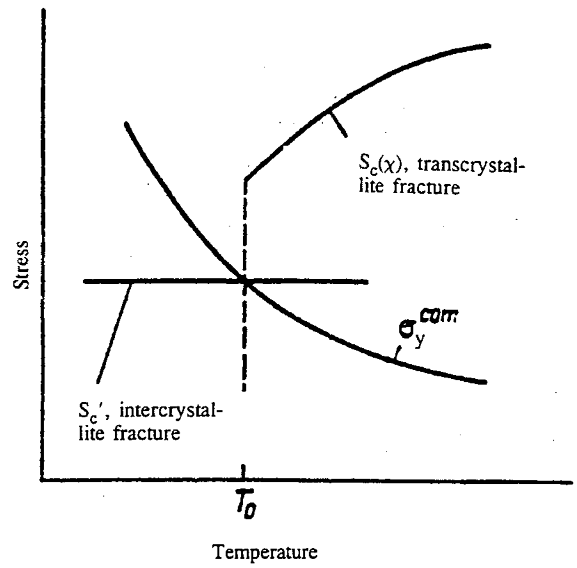


Fig. 5. Brittle fracture under tension for different temperatures of preliminary compression of the specimen: $S_c(\chi)$ and S_c' are the critical stress for fracture; c is the accumulated plastic strain during preliminary deformation and breaking tests; σ_y^{com} is the yield stress under tension of a precompressed material.

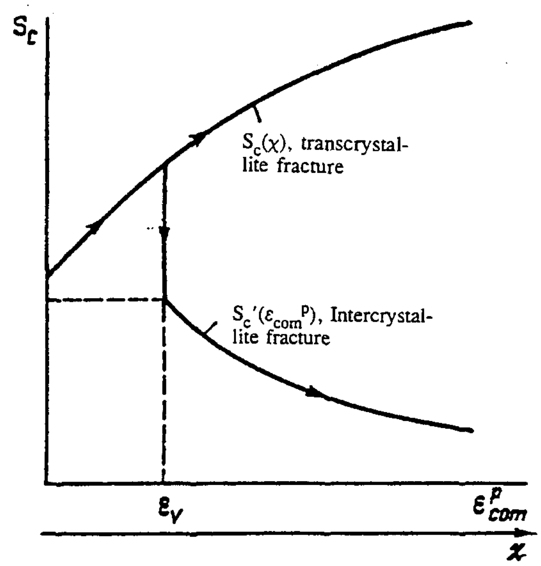


Fig. 6. Brittle fracture under tension as a function of the degree of preliminary plastic strain by compression ε_{com}^P: ε_n is the strain corresponding to the onset of formation of intercrystallite pores. (The rest of the symbols are the same as in Fig. 5.)

nature. We note that the substructure formed in preliminary deformation leads to an additional increase in the parameter S_c compared with its values for the original material.

In the case of preliminary compression, as for preliminary tension, sharp transcrystallite microcracks and pores along the grain boundaries are initiated. However, in contrast to tension, during compression the pores along the grain boundaries after their initiation will become tapered (compressed) as a result of the dislocation sink at the tips (Fig. 4b). As a result, after preliminary compression the specimen will have sharp transcrystallite and intercrystallite microcracks, oriented perpendicular to the direction of compression (Fig. 4b) and also to the intragranular deformation substructure formed.

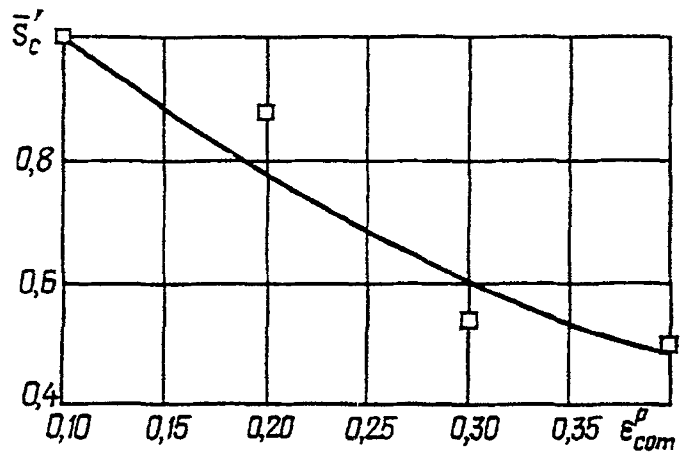


Fig. 7. Relative critical stress for brittle fracture $\bar{S}_c'$ vs. preliminary plastic compressive strain $\varepsilon_{com}{}^p$. (Solid line — the calculation results; points — experimental data, averaged over four to five specimens.)

When such a specimen is subsequently subjected to tension, fracture will be of the brittle intercrystallite type if the condition for beginning of growth (the Griffith's criterion) for the sharp microcracks existing in the material is satisfied earlier than their blunting begins as a result of the dislocation sink at the tips. The intercrystallite nature of the fracture is due to the fact that the microcracks along the grain boundaries may develop freely (there are no barriers for them), while propagation of transcrystallite microcracks will be blocked by the deformation substructure.

The condition for the absence of blunting of microcracks may be formulated as the condition for the absence of plastic deformation in the material ($\sigma_i < \sigma_y$) or as restriction to plastic deformation within some tolerance ($\sigma_i{}^p < \sigma_{bl}{}^p$, where $\sigma_{bl}{}^p$ is the plastic strain at which the sharp microcracks begin to be blunted). From the data in Table 3, we see that the condition $\sigma_i < \sigma_y$ is rather conservative, since for $\varepsilon^p = 1\%$ we observe the presence of an intercrystallite component in the fracture; in other words, for this strain some microcracks remain sharp and capable of unstable development according to an intercrystallite mechanism.

In the following, for concreteness we will assume the condition for the absence of blunting of the microcracks in the form $\sigma_i < \sigma_y$.

Thus if microcracks begin to grow at $\sigma_i < \sigma_y$, then intercrystallite fracture is realized. If the thermal and mechanical loading of the precompressed specimen is such that there is no brittle fracture for $\sigma_i < \sigma_y$ (the Griffith's criterion is not met), then for $\sigma_i \geq \sigma_y$ blunting will occur for both intercrystallite and transcrystallite microcracks. Therefore in order to realize brittle fracture, we should have initiation of sharp microcracks, beginning of their growth, and propagation through the deformation structure formed. In this case, the process of realization of brittle fracture becomes identical to fracture of a specimen subjected to preliminary tension, where the condition for development of transcrystallite fracture is expressed in the form $\sigma_1 = S_c(\chi)$.

Devising schemes for the process of brittle fracture of precompressed material. The process of brittle fracture described above can be represented as schemes. In Fig. 5, we show the scheme for fracture of a specimen precompressed up to some strain $\varepsilon_{com}{}^p$ and subjected to breaking tests at different temperatures. In the temperature range $T < T_0$, intercrystallite fracture is realized for $\sigma_1 = S_c'$ ($\sigma_i < \sigma_y{}^{com}$), the initiators of which are sharp intercrystallite microcracks. If $T > T_0$, then fracture occurs according to a transcrystallite mechanism for $\sigma_1 = S_c(\chi)$ ($\sigma_i > \sigma_y{}^{com}$) from new sharp microcracks, initiated during the breaking tests. The temperature at which the fracture mechanism changes from intercrystallite to transcrystallite T_0 and the critical stress for intercrystallite tearing $S_c{}^c$ depend on the preliminary compressive strain $\varepsilon_{com}{}^p$.

In Fig. 6, we present the scheme for brittle fracture depending on the degree of preliminary compression of the sample. In the case when $\varepsilon_{com}{}^p < \varepsilon_n$ (ε_n is the strain corresponding to the onset of formation of intercrystallite microdiscontinuities), intercrystallite discontinuities are not formed and fracture occurs according to a transcrystallite mechanism for the condition $\sigma_1 = S_c(\chi)$. For $\varepsilon_{com}{}^p \geq \varepsilon_n$, the following situations are possible, depending on the testing temperature. If for $\sigma_1 = S_c'(\varepsilon_{com}{}^p)$ we have $\sigma_i < \sigma_y$, then the sharp intercrystallite microcracks are not blunted and intercrystallite fracture of the specimen develops from them. If for $\sigma_1 = S_c'(\varepsilon_{com}{}^p)$ we have $\sigma_i > \sigma_y$, then new sharp microcracks must form in order for fracture to occur. In breaking tests, the sharp microcracks are only transcrystallite; consequently the fracture will be transcrystallite from satisfaction of the condition $\sigma_1 = S_c(\chi)$.

Prediction of the $S_c'(\varepsilon_{com}{}^p)$ dependence. Let us consider the condition for initiation of pores on the grain boundaries according to the antiparallel pileup mechanism (Fig. 3c). In [17] it is shown that the maximum local stress in the scheme involving antiparallel pileups $\sigma_m{}^{loc}$ is greater than the maximum stress in a single pileup $\sigma_s{}^{loc}$ and may be determined from the dependence

$$\sigma_m^{loc} = \sigma_s^{loc}\left(1 + K\frac{L}{h}\right), \tag{7}$$

where L is the length of the dislocation pileup; h is the distance between dislocation pileups; K is some coefficient, according to the data in [17] K = 0.1.

Condition (2a) was obtained in considering initiation of a microcrack at the head of a single pileup [13, 14]. Taking into account (7), initiation of a pore on the grain boundary according to the antiparallel pileup mechanism may be written in the form

$$\sigma_1 + m_{T\varepsilon}\left(1 + K\frac{L}{h}\right)(\sigma_i - \sigma_0) = \sigma_d^{c-m}. \tag{8}$$

From the latter equation, we see that as the degree of plastic deformation increases and accordingly σ_i increases, initiation of a pore may occur for small values of L/h, i.e., for large h. Consequently, as $\varepsilon_{com}{}^p$ increases, the initiated pores will be more stretched out, as a result of which when they are flattened out sharp microcracks will occur. Obviously longer microcracks will begin to grow at lower stresses. From this it follows that as $\varepsilon_{com}{}^p$ increases, the parameter S_c' should decrease.

Let us determine the $S_c'(\varepsilon_{com}{}^p)$, based on Eq. (8). Since we will consider plastic strains which are significantly greater than the strain corresponding to the yield stress of the material (≈0.2%), in (8) the parameter σ_0 can be replaced by σ_y.

According to the data in [13, 14], the length of the dislocation pileup decreases in proportion to the plastic strain according to the dependence

$$L = \frac{m_\varepsilon^2 P_M^2}{\pi}, \tag{9}$$

where P_M is some material constant; m_ε is determined from Eq. (4), in this case $\chi = \varepsilon_{com}{}^p$.

Solving Eq. (8) relative to h taking into account the Eqs. (3), (6), and (9), and also considering that on uniaxial loading $\sigma_1 = \sigma_i$, we obtain

$$h(\varepsilon_{com}^p) = \frac{KP_M^2}{\pi}\left[\frac{\sigma_d^{c-m} - \sigma_T - A_0(\varepsilon_{com}^p)^n}{m_T m_\varepsilon A_0(\varepsilon_{com}^p)^n} - 1\right]^{-1} m_\varepsilon^2. \tag{10}$$

Using the Griffith's equation in [8] for analysis of the beginning of growth of microcracks of length h, the critical stress of intercrystallite fracture in relative form can be determined from the formula

$$\bar{S}_c' \equiv \frac{S_c'(\varepsilon_{com}^p)}{S_c'(\varepsilon_{com}^p = 0{,}1)} = \left[\frac{h(\varepsilon_{com}^p = 0{,}1)}{h(\varepsilon_{com}^p)}\right]^{1/2}. \tag{11}$$

Substituting (10) into (11), we obtain

$$\bar{S}_c' = \frac{1}{B}\frac{1}{m_\varepsilon}\left[\frac{\sigma_d^{c-m} - \sigma_T - A_0(\varepsilon_{com}^p)^n}{m_T m_\varepsilon A_0(\varepsilon_{com}^p)^n} - 1\right]^{1/2}, \tag{12}$$

where $B = P_M(K/\pi h)^{1/2}$ is a parameter determined by Eq. (10) for $\varepsilon_{com}{}^p = 0.1$.

The calculation using the dependence (12) was done based on data presented in Table 2 (for T = 20°C) and also the parameters $C_1{}^*$, $C_2{}^*$, A_g (Eq. (5)). The value of the temperature-dependent parameter $m_T(T)$ for steel 15Kh2MFA is taken from the previous investigations in [14]: $m_T(T = 20°C) = 21830$ MPa. The parameter $\sigma_d{}^{c-m}$ was specified based on the best agreement between the calculated and experimental data; in this case, we considered that $\sigma_d{}^{c-m} > \sigma_c{}^c = 6300$ MPa. The value of $\sigma_d{}^c$ was taken from [14].

Comparison of the experimental data with the calculated prediction of the $\tilde{S}_c'(\varepsilon_{com}{}^p)$ dependence for $\sigma_d{}^{c-m} = 10000$ MPa (Fig. 7) suggests rather good agreement. Thus we can conclude that the proposed schemes for the process of brittle fracture when precompressed material is subjected to tension, based on which we make a prediction of the $\tilde{S}_c'(\varepsilon_{com}{}^p)$ dependence, correctly reflect the basic physical mechanisms for brittle fracture of such a material.

CONCLUSIONS

1. We have experimentally established that in contrast to preliminary tension (which does not change the transcrystallite nature of fracture and increases the critical stress for brittle fracture of the material), preliminary plastic compression may lead to a change in the mechanism for brittle fraction of bcc metals: from transcrystallite microspalling (when testing the material in the original state) to intercrystallite fracture (when testing the plastically precompressed material). A consequence of the change in the fracture mechanism is a significant reduction in the critical stress for brittle fracture S_c.

2. Realization of the intercrystallite mechanism for brittle fracture is connected with initiation of intercrystallite pores and their conversion during plastic compression to sharp microcracks. Under subsequent tension, such microcracks may be initiators of intercrystallite brittle fracture.

3. We propose schemes for the transition from intercrystallite fracture to transcrystallite fracture, depending on the thermal and mechanical loading conditions for precompressed material.

4. We have shown that reduction of the critical stress of intercrystallite brittle fracture S_c' with an increase in the preliminary deformation by compression is connected with initiation of large pores. We have obtained equations allowing us to predict the dependence of S_c' on the preliminary deformation by compression.

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