NATURE OF THE STRENGTH OF TUNGSTEN-NICKEL-IRON ALLOYS

PART 2. RUPTURE MECHANISM OF W–Ni–Fe ALLOYS

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To formulate new constructional materials and study the properties of existing materials, it is essential to know the relationship between their structure and strength. In establishing such a relationship, use can be made of data on the fine structure of fractures which constitute a direct result of a rupture process brought about by the structural characteristics of the material and its deformation conditions.

To establish the connection between the microstructure and strength of W–Ni–Fe heavy alloys, in the present investigation a study was made of the structure of fractures obtained by submitting such alloys to tensile tests. Electron-microscopical fractography was the principal method of investigation employed. Using this technique, a study was made of the effects of cooling conditions and composition on the character of rupture of the alloys.

The compositions of the test alloys and the experimental procedure were described in the earlier communication [1] where it was noted that the rate of cooling from the sintering temperature is responsible for the structural changes in the alloys which govern their properties in tensile tests. When tested in tension, W–Ni–Fe alloy specimens quenched from the sintering temperature or slowly cooled (with the furnace) to 1200-1300°C behave in a brittle manner. It has been shown by us [1] that under such cooling conditions only the early stage of aging – formation on the surface of the refractory grains of lamellas of the precipitated phase – can be completed.

Study of fracture facets in these alloys after the above-mentioned heat treatments indicates that single crystals of the refractory constituent rupture by a cleavage mechanism. Photomicrographs exhibit so-called "river patterns" (Fig. 1). Their appearance has been attributed to the fact that a cleavage crack, instead of being propagated over a single crystallographic plane, is broken up by crystalline-structure defects into separate portions, due to different levels of cleavage surfaces [2]. The formation of cleavage steps has been ascribed to various causes, the most common of which is the presence of screw dislocations piercing the cleavage plane [3]. For the propagation of a cleavage crack, energy of interface generation and kinetic energy are required. It is believed that some additional energy is expended on the formation of cleavage steps, in consequence of which the propagation of a cleavage crack is hindered [2]. Indeed, examination of the fracture facets of W–Ni–Fe alloys indicates that the propagation of such a crack is usually restricted to single grains of the tungsten-base phase.

The brittle cleavage fracture is then arrested and changes in the binder phase into a "tough fracture" or splitting along slip planes. This type of rupture is regarded as being intermediate between cleavage and tough rupture [4]. Splitting along slip planes manifests itself in the form of relatively large, almost plane surfaces, occasionally showing signs of slip bands (Fig. 2).

Transcrystalline fracture in the tungsten grains frequently exhibits a microstructure similar to that illustrated in Fig. 3. Fractures of this type are evidently due to the formation of cellular structure at high degrees of deformation, which is characteristic of metals with a high stacking-fault energy [5].

Examination of microfractograms of various specimens after the initial stage of aging indicates that the lamellar particles of the precipitated phase are subjected to shear during plastic deformation (Fig. 4), the lamella deformation reproducing the fragmentation of the tungsten grain, which is due to its plastic-flow nonuniformity. Such a behavior of the precipitated phase is evidence of coherence between the precipitates and of the presence of an orientational and structural bond between these precipitates and the matrix lattice in this stage of aging.

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Fig. 1

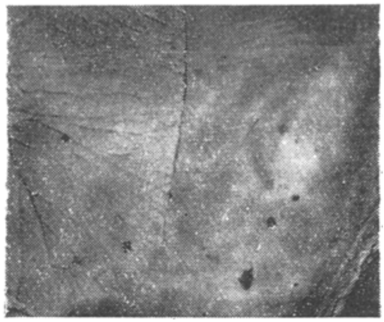

Fig. 2

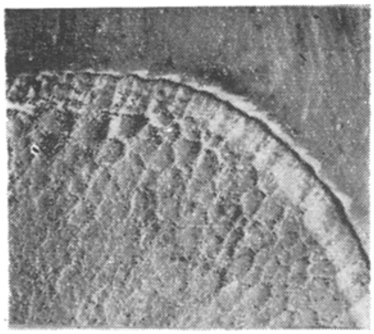

Fig. 3

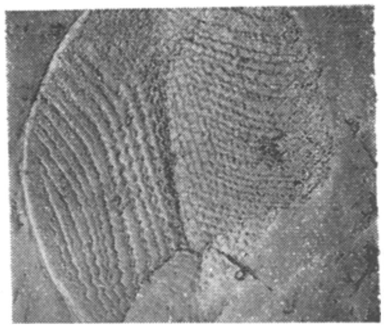

Fig. 4

Fig. 1. Formation of "river patterns" in transcrystalline rupture of tungsten grains. $\times 4100$.

Fig. 2. Brittle cleavage of refractory constituent and "tough fracture" of binder phase in W–Ni–Fe alloy. $\times 4800$.

Fig. 3. "Cellular structure" in tungsten grains. $\times 16{,}000$.

Fig. 4. Deformation of precipitated phase. $\times 12{,}000$.

Precipitate comminution and particle coagulation in the course of aging during slow cooling to 800–900°C introduce a marked change in the character of rupture. Figure 5a illustrates the simultaneous appearance of tough rupture and cleavage. Figure 5b shows a characteristic case of tough rupture of the binder alloy. The microfractogram exhibits pits typical of a fibrous fracture, in the bottom of which it is frequently possible to see precipitated particles or their traces. This helps to understand the mechanism of formation of fibrous fractures. Crussard and coworkers [4] consider that microcracks can form during deformation as a result of the difference in the elastic and plastic properties of the particles and the matrix (separation of the particles from the matrix). In the region of high volume stresses, a large number of such cracks may form in front of a spreading fracture or the existing cracks may grow, leading to the generation of microregions visible in fractograms in the form of pits. The distribution of precipitates in the course of aging determines the crack-propagation paths.

During the aging of W–Ni–Fe alloys, intense formation of the precipitated phase occurs at the interface between the phases. The opening up of cracks along the phase boundaries in tensile tests makes it possible to observe the consecutive stages of aging of these alloys [1]. The formation of a fracture similar to that shown in Fig. 6 may be explained by referring to literature data on rupture processes. An over-all picture of rupture may be formed by considering its principal stages (after Cottrell [6]), namely, 1) crack initiation and 2) crack growth and propagation.

The initiation of a brittle crack takes place as a result of a gliding dislocation becoming transformed into a void (unit) dislocation [6]. The several initiation mechanisms proposed [7] are usually divided into three groups. The first group comprises models based on the concept of a pile-up of edge dislocations in

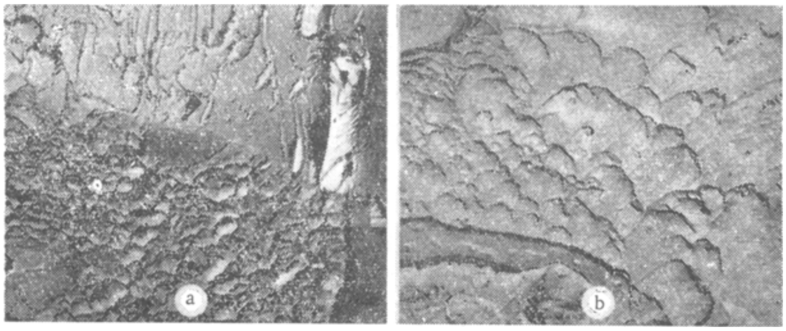

Fig. 5. Structure of fracture facets in aged alloy: a) combination of brittle cleavage and "dimpled" fracture; b) fibrous, "dimpled" fracture. $\times 8500$.

a slip band before a firm obstacle (boundaries of grains with a large angle of disorientation, strong incoherent inclusions, etc.). Externally-applied stress joins dislocations and a crack is formed under their merged half-planes (Fig. 7a). The Mott-Stroh initiation mechanism also belongs to this group [8, 9]. The second group contains models linking crack initiation with the formation of pile-ups in the course of propagation of plastic deformation. Dislocations moving at the same time on two intersecting slip planes merge together at the intersection line and become transformed first into void dislocations and then into a crack (Fig. 7b). Finally, the third group includes barrier-free models in which cracks form as a result of the reaction of crystalline-lattice defects (dislocations, vacancies) in the absence of firm barriers.

At the present time, an examination is being made in the literature of the applicability of dislocation models of brittle rupture to the determination of tough-rupture conditions. According to Mott [10], tough cracks are initiated as a result of dislocations piling up before an obstacle, so that the mechanism of their formation is essentially analogous to that describing the initiation of brittle cracks. Stresses required for the generation of a crack nucleus and determined in accordance with the pile-up model have been compared with stresses corresponding to experimentally observed values of pile-ups for the case of ductile metals and were found to be in good agreement with each other [7]. This indicates that the models of brittle-crack initiation can be employed to analyze the generation of tough cracks. A crack generated by one of the schemes described in the foregoing grows in width and extends into the material, leading to its rupture.

Cottrell and Petch [6] almost simultaneously reached the conclusion that the rupture process is controlled by crack growth. According to this hypothesis, transformation of gliding dislocations into void dislocations is easier to accomplish than crack growth.

The question of the mechanism governing crack propagation remains under discussion. Several authors [11-13] concerned with the brittle rupture of metals have cited dislocation reactions responsible for the drawing apart of cleavage planes to such an extent that the resulting crack attains a size at which it can be propagated by Griffith's mechanism under the action of the elastic energy being released [14]. Orowan considers [15] that the propagation of a crack within a grain need not necessarily depend on the prior existence of a crack of the critical Griffith size or on the generation of such a crack as a result of some dislocation reactions. The fine cracks present in each grain can grow by the mechanism of plastic propagation until they reach this critical size. However, even the mechanism of elastic propagation, involving crack displacement in parallel cleavage planes, calls for a large number of tough cleavages.

In the case of tough rupture [6], plastic deformation is essential not only as a preparatory stage prior to rupture, but also as a component part of the actual rupture process. The function of this deformation, however, is not to rupture the material ahead of the growing crack, but to remove it from the rupture zone by plastic shaping (opening up of the crack through the emergence onto its surface of dislocations attracted by stress concentration).

On the basis of the foregoing considerations of a general, theoretical nature, it is possible to suggest a probable mechanism of rupture of the alloys under consideration and, consequently, to explain the observed changes in the character of rupture caused by variation in alloy composition and cooling conditions.

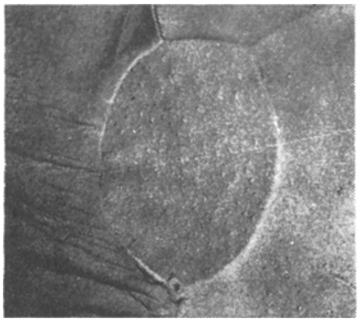

Fig. 6. Open cracks in aged alloy. ×12,700.

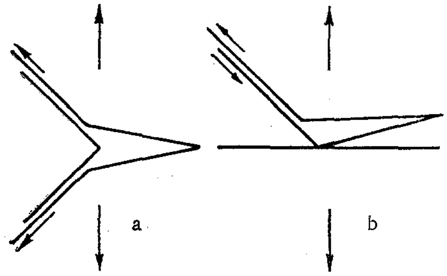

Fig. 7. Dislocations and cracks: a) crack resulting from slip in two bands; b) crack generated from pile-up of dislocations before an obstacle [6].

1. The structure of these alloys corresponds to the early stages of aging; in particular, lamellas coherently bonded to the refractory constituent were found to form on the interface between the phases.

The rupture process of these alloys clearly commences as a result of the formation of a pile-up, for instance, at a dislocation-intersection line in the low-melting-point phase, i.e., the binder. The crack generated in this manner grows, at first by the mechanism of plastic propagation and then, when the critical Griffith size is reached, under the action of the elastic energy being liberated. An alloy in this condition exhibits symptoms of brittle rupture, which manifest themselves in the formation of "river patterns." The phase produced in the early stage of aging is broken up by the growing crack.

2. The subsequent course of the aging process changes the mechanism of rupture of the alloys. The crack opens up as a result of the piling up of dislocations at the interface between the phases, which is studded with particles of the precipitated phase. If a crack nucleus is greater than the distance between these particles, it will round the tungsten grains and penetrate into the binder phase. Rupture occurs within the low-melting-point component of the alloy; the interphase zone constantly changes the direction of travel of the crack, thereby confining its propagation to the alloy binder. The presence of the precipitated phase at the interface between the structural constituents of the alloy prevents transgranular rupture of the refractory constituent. An alloy cooled with the furnace from the sintering temperature to 800°C is ductile. A characteristic feature of its rupture surface is the presence of regions of purely fibrous structure. Thus, the ductility of an alloy may be determined, after Gilman [13], as its resistance to rupture by cleavage. Cooling an alloy to lower temperatures introduces no significant changes into the structure of its fracture and its properties; this is evidently due to the fact that the mobility of atoms at a temperature below 800°C is not sufficient to affect significantly the size and spacing of the precipitates. Consequently, the critical crack size remains greater than the distance between the precipitated particles. This fact determines the optimum, from the viewpoint of mechanical properties, conditions of cooling from the sintering temperature, namely, cooling with the furnace to 700-800°C (cooling rate 10 deg C/min), followed by ejection into a cooler.

Study of the fracture surface of W−Ni−Fe alloys as a function of alloy composition also indicates that the character of rupture changes from brittle to tough with increasing binder-phase content of the alloys. Varying the amount of this phase determines the extent to which aging processes can proceed. Clearly, it is this circumstance that explains why there is a distinct analogy between the effects of cooling conditions and alloy composition on the character of rupture of these alloys. For instance, an alloy with 3 wt.% nickel and iron (4.2 vol.% of the low-melting-point constituent) in which the aging process cannot take place fully after slow cooling from the sintering temperature exhibits brittle, transgranular rupture when tested in tension.

All that has been said in this study leads to the conclusion that from the point of view of mechanical properties the optimum alloy is one in which aging processes lead to the formation of a dispersed, incoherent phase. The strength and ductility properties of such an alloy are governed by the precipitation-hardened binder alloy formed as a result of aging.

SUMMARY

A study was made, by electron-microscope fractography of the effects of composition and conditions of cooling from the sintering temperature on the character of rupture of W−Ni−Fe heavy alloys. It has been established that the change in the mechanism of rupture from brittle to tough is due to structural changes in the alloy linked with the aging process.

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