THE ROLE OF HYDROGENATION IN THE CORROSION FATIGUE OF STEEL

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Abstract

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THE ROLE OF HYDROGENATION IN THE CORROSION FATIGUE OF STEEL

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In previous works (1) we developed adsorption–electrochemical concepts of the mechanism of corrosion fatigue of steel and drew attention to the impossibility of relating the intensity of general corrosion to the development of corrosion fatigue. Here we shall present several propositions that supplement our views on the corrosion fatigue of steel at high stress amplitudes.

When a corrosive medium and repeatedly alternating stresses act simultaneously on a metal, adsorption, diffusion, and corrosion (electrochemical) phenomena occur. A corrosion-aggressive medium causes corrosion damage to all anodic areas of the metal surface accessible to it, but before the corrosion process begins, adsorption of surface-active components of the medium occurs (specific adsorption of ions or molecules). This adsorption causes a thermodynamically inevitable decrease in the strength of the metal (2), which under cyclic loading conditions manifests itself in a decrease in endurance in the phenomena of adsorption fatigue (3).

At cathodic areas hydrogen cations are adsorbed. Atomic hydrogen reduced on the metal diffuses into the metal lattice, hydrogenates it (with possible formation of hydrides), and causes the appearance of hydrogen embrittlement. Hydrogen embrittlement of cathodic areas of the metal may influence the corrosion-fatigue strength of steel in the case of hydrogen attack on regions of the metal of considerable depth, which becomes possible during deformation of the metal, when hydrogen penetrates into the depth of the metal along slip planes.

Experiments by M. I. Chaevsky, carried out in our laboratory, showed that hydrogenation of steel during its deformation by statically applied forces occurs along slip planes at a very high rate.

The experiments were arranged so that specimens of mild steel-3, 10 mm in diameter, served as cathodes and anodes* in a bath with an acidic electrolyte and were simultaneously stretched on an IM-12 tensile-testing machine at a constant rate (v = 16 mm/min) until rupture within 1-2 min. This led to brittle fracture of the cathode made of plastic steel along slip planes, which is clearly seen in Fig. 1a; moreover, the strength decreased by about 10%, while the true stresses decreased approximately twofold. The anodic specimens did not lose their ductility or strength and fractured in the same way as specimens

Fig. 1. Place of rupture of a steel-3 specimen. a –specimen is the cathode; b –specimen is the anode; c –rupture in air

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in air (cf. Fig. 1b and 1c). In the case of a copper anode and a steel cathode, under conditions in which no hydrogen was evolved at the cathode, the indicated phenomena were not observed.

In the experiment described, hydrogen ions were reduced on the surface of the deforming steel specimen, and hydrogen in the atomic state penetrated deep into the metal along slip planes developing during deformation, and then, diffusing into the metal lattice, brought it into a brittle state. After this, the tensile forces caused brittle fracture of the specimens by separation along the hydrogenated regions of the metal, coinciding with the slip planes.

What is striking in the experiments is the enormous rate at which the process of hydrogenation of the metal takes place. This can be explained only if it is assumed that along the slip planes not only is the metal lattice distorted, but there also develop numerous microcracks; in the course of their formation hydrogen is absorbed into them, penetrating deep into the metal at sufficiently high rates, considerably exceeding the rates of bulk and even grain-boundary diffusion.

Fig. 1. Place of rupture of a steel-3 specimen. **a** –specimen is the cathode; **b** –specimen is the anode; **c** –rupture in air

The portions of metal between defects, which under ordinary conditions account for the increased strength of the slip zone, have a distorted lattice, which is readily permeable to diffusion as a result of activation. It is known that the rate of diffusion into a distorted lattice is several orders of magnitude higher than the rate of diffusion into the bulk with a regularly constructed lattice. S. T. Konobeevskii (4) points to an increase in the rate of diffusion of nickel into deformed copper by more than 1000 times.

Our experiments allow us to suppose that in such polycrystalline metals as steel, which have at the metal-medium interface many microcathodic and microanodic areas, in the process of cyclic loading in a corrosion-active medium very rapid hydrogenation of the cathodic areas occurs. Obviously, these areas are weak points along which brittle fracture occurs at large stress amplitudes, when this fracture sets in soon after the specimens are loaded. In this case the time is insufficient for appreciable corrosion damage of the anodic areas, i.e., for the appearance of weak points in the specimen under the influence of their weakening by corrosion attack.

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* Current 20 A, current density 55.5 A/dm^2.

This is confirmed by the fact that, at high stress amplitudes, the fracture surface of steel specimens in corrosion-active media does not

has oxidized regions, and corrosion products are not found in fatigue cracks. At the same time, fatigue failure in corrosive media, at all values of the stress amplitudes, occurs at stresses considerably lower than in air.

At lower stress amplitudes, when the time to failure of the cyclically loaded metal is considerable, all the factors that reduce endurance have time to manifest themselves—namely adsorption, diffusion, and corrosion factors. In this case the fatigue cracks are filled with corrosion products, and the fracture is usually strongly oxidized.

I consider it my duty to express my gratitude to Academician P. A. Rebinder and Doctor of Physico-Mathematical Sciences V. I. Likhtman for their critical comments.

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