STRUCTURAL CHANGES AND MATERIAL FAILURE UNDER HYDROGEN INFLUENCE

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ABSTRACT

New idea on a possible use of hydrogen and hydrogen induced phase transformations for a *controllable* hydrogen phase naklep and for a controllable failure of metals are discussed.

KEYWORDS

Hydrogen, hydride transformations, hydrogen phase naklep (HPN), hydrogen failure, palladium, niobium, intermetallic compound, hydrogen treatment.

INTRODUCTION

Hydrogen influence on materials is one of the fundamental controllable effects side by side with temperature, pressure, fields and particle flows (Goltsov *et al.*, 1981, 1982, 1985, 1990). Under the influence of hydrogen and hydrogen-induced phase transformations rather large internal forces and stresses occur in metallic materials. Their values and consequences are determined by the conditions of the hydrogen treatment and by the state of a solid (elastic-plastic, plastic or brittle). The most important constituent of a new conception on a fundamental character of the hydrogen influence on metallic materials is its controllability. Great opportunities appear to use hydrogen and hydrogen-induced phase transformations to control materials structure, properties, deformation degree and failure.

PALLADIUM - HYDROGEN

The hydrogen treatment is based on the use of hydride transformations in metal-hydrogen systems. In this connection our knowledge about the peculiarities of hydride transformations is of a great importance.

With a help of a specially made experimantal device we directly observed hydride transformations taking place in a classical palladium-hydrogen system. We had performed experiments to find out those conditions under which an isothermal development of a hydride transformation might be observed. With this purpose hydrogen was introduced into the chamber up to 0.1 MPa at 200°C. Under these conditions a specimen was charged with hydrogen for 0.5 h and then cooled at the rate of 0.5 K/min. At this cooling rate hydrogen quantity dissolved in a specimen is close to equilibrium at each temperature value, and a hydride transformation temperature should be in an agreement with the equilibrium diagram, that is close by 140°C.

Under these conditions first hydride precipitations of 5 mcm were observed on a specimen polished surface at 120°C. As soon as the first hydride precipitations were fixed, a specimen cooling was stopped and the development of a hydride transformation under the isothermal conditions was observed then.

A principal result is that a hydride transformation proceeds according to the mechanism of nucleation and growth, that is according to the classical theory of phase transformations of the first type. From the viewpoint of the classical theory it is also important that a transformation starts developing at some overcooling of an alloy below the critical point corresponding to the equilibrium diagram.

As a massive hydride precipitation grows, the rate of its growth is gradually decreasing. This process can even stop though the thermodynamic motive forces of a phase transition continue acting according to the experimental conditions ($T=const<T_c$, $P_H = const>P_c$).

An equilibrium shift (or a situation close to it) can not be any longer within the same mechanism of growth, that is by the simultaneous growth of the whole massive hydride boundary a break-down of a local equilibrium followed by an "ejection" and a quick growth of a needle hydride formation takes place. Having reached some definite sizes this daughter formation stops its growing and a new temporary local equilibrium comes. This equilibrium is then taken away by a new mechanism of growth, - this formation starts gradual reforming and becomes a part of an initial massive precipitation. Just this regular changing of the morphological form and the mechanism of a hydride transformation development is maybe a common regularity of diffusive - cooperative phase transitions in Pd-H alloys.

Let us now consider an influence of the hydrogen phase naklep on mechanical properties of palladium at hydride transformations. Mechanical properties of palladium specimens as a function of the completeness of $\alpha \to \beta$ hydride transformations are shown in Fig. 1. As shown, an initial meaning of the ultimate strength (curve 1) of the annealed palladium was 150 MPa. When increasing a β-phase content up to 25%, the ultimate strength increases up to 380 MPa. Then, as an $\alpha \to \beta$ transformation completeness increases, the ultimate strength begins decreasing and then it practically stays on one and the same level (245 - 255 MPa), which is, however, higher than that of palladium in its initial state. As an $\alpha \to \beta$ transformation completeness increases, the yield limit of palladium (curve 2) gradually increases from 65 up to 165 MPa. A meaning of the relative elongation (curve 3) first increases from 21% up to 44% and then falls down abruptly up to 9% at 75% of the β-phase. When an $\alpha \to \beta$ transformation was completed, a meaning of the relative elongation turned to be 5%.

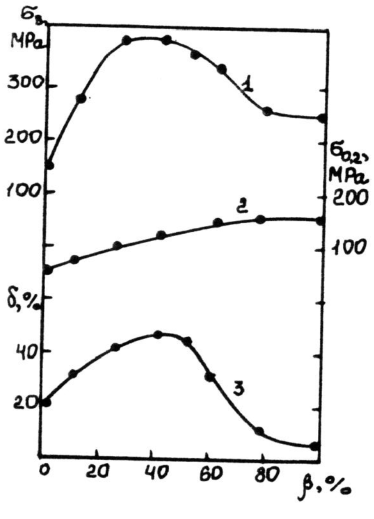

Fig. 1
Mechanical properties of palladium as a function of the completeness of $\alpha \to \beta$ transformations:
1-ultimate strength, MPa;
2-yield limit, MPa;
3-relative elongation, %.

NIOBIUM - HYDROGEN

For niobium, whose mechanical properties are very sensitive to interstitial impurities, it is important that the HPN mechanism ensures an active binding of hydrogen. Thus, the $NbH_{0.03}$ alloy, which is a homogeneous solid solution at room temperature, after the HPN-treatment showed a considerable decreasing of the lattice parameter from 3.30640 A up to 3.30430 A, that corresponded to the "binding" of ~ 1 at.% of the solid solution hydrogen. The following many-houred annealing in the high-vacuum chamber at 350°C did not reset the meaning of the lattice parameter, and this is quite clear, as such temperature is much lower than the recrystallization temperature of niobium. Thus, hydrogen "binding" resulted in a growth, though small, of the alloy plasticity and in the simultaneous growth of the strength properties. It is interesting that within the experimental error the specific electrical resistance of the alloy remained unchangable. Data of the HPN treatment influence on the $NbH_{0.03}$ properties are given in Fig. 2.

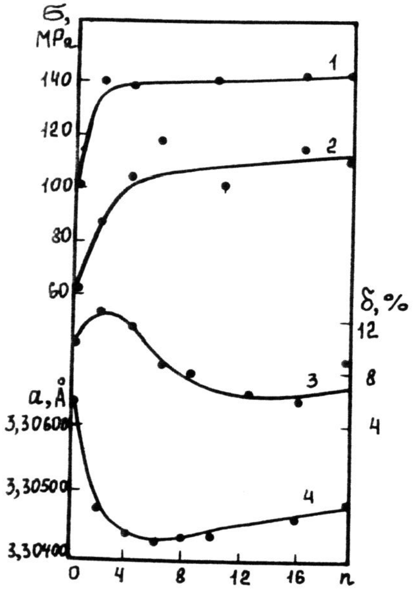

Fig. 2
Influence of the HPN - treatment on the mechanical properties of the $NbH_{0.03}$ and the lattice parameter of its matrix (α-phase):
1-σ_B; 2-$\sigma_{0.2}$; 3-δ; 4-a.

AN INTERMETALLIC COMPOUND-HYDROGEN

As known, hydride-forming intermetallic compounds (IMC), $LaNi_5$ in particular, fail into a fine-dispersive powder under hydrogen influence. An analysis of the surfaces of the particle fractures, made by the scanning electron microscope, confirmed a brittle character of failure. Distribution of microcracks was accompanied by their division into two and more branches. The preferential angles of branching are 60° and 90°. A fine structure of the IMC was studied by the transmission electron microscope. Specimens in the form of a fine-dispersive powder were fastened in a colloid film and subjected to TEM-testing at the accelerating voltage 150 kV.An analysis of the $LaNi_5$ structure, which was not influenced by hydrogen , showed that in a case of brittle failure under a mechanical action plastic deformation took place in the material. Under hydrogen influence these defects are traps and centres of a hydride phase formation. Then after some hydrogen sorption-desorption cycles we analysed the structure of the IMC. A dislocation structure, observed after cycling, is a "memory" of hydrogen and hydride phase influence on the material, as in the result of a final desorption hydrogen was removed from the IMC. So, the hydrogen treatment accompanied by the formation of a hydride phase,having a larger specific volume than a matrix has, results in the formation of the regions of a higher defects, in the dislocation pile-up. The latter are slowed down near the barriers. Such regions of a higher defects can apparently serve as places where microcracks appear; in the final analysis these microcracks results in the failure of a material under hydrogen influence.

CONCLUSIONS

Under hydrogen and hydrogen-induced diffusive-cooperative (hydride) phase transformations large internal "hydrogen" stresses appear in metallic materials.These stresses rearrange and relax by different mechanisms. Plastic (or hydrogen plasticized) metals suffer from a controllable hydrogen phase naklep. As the natural result a material is first strengthened, but under the following hydrogen influence it can be failed through the loss of plasticity by the "superdefect" state. The latter is characterized by a very high dislocation density and by large changes in the fine structure. Under the influence of hydrogen and hydride transformations initially brittle materials at once undergo failure and the formation of fine-dispersive powders whose particle matrixies contain some small quantity of dislocations took place. However this process is also controllable, and allows to get a final arrangement of particles of the failed material specified in advance.

Thus, hydrogen influence on metallic materials is as fundamental as temperature, pressure, fields and particles flows. This influence is the basis of a new treatment, that is hydrogen treatment of metallic materials in order to get structures and properties needed.These fundamental phenomena serve as an experimental basis for the theoretical estimation of the efficiency and failure of metallic materials in the hydrogen containing media.

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