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CERTAIN GENERAL PATTERNS IN THE OCCURRENCE OF HYDROGEN BRITTLENESS IN THE ALLOYS VT3-1 AND VT15

by

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Discussed is the development of hydrogen brittleness in an ($\alpha + \beta$)-titanium alloy VT3-1 and a β-titanium alloy VT15. It is shown that hydrogen brittleness develops most intensely in VT3-1 at temperatures less than room temperature and at low strain rates. It is discovered that β-titanium alloys at temperatures below room temperature are also prone to hydrogen brittleness, the temperature at which this occurs decreasing with a decrease in the hydrogen content. Hydrogen brittleness in the alloy VT15 is found to occur only at low strain rates in a narrow temperature range, approximately from -30°C to +10°C. Hydrogen brittleness of ($\alpha + \beta$)-titanium alloys is believed to be caused by processes developing in the β-phase during plastic deformation. Since the composition of the β-phase of VT3-1 after isothermal annealing is close to the composition of the β-phase of VT15, it is concluded that hydrogen brittleness develops in approximately the same manner in both of them, provided the conditions for mechanical testing are the same for both.

Of all types of hydrogen brittleness of metals, of greatest scientific interest is the brittleness developing at low rates of plastic deformation at fairly large hydrogen concentrations which, however, are below the ultimate solubility at experimental temperature. This brittleness is not due to the formation of stable hydrides or of any other phase; it is due to processes developing during the plastic deformation in solid solutions of hydrogen in metals.

It was recently established that such a brittleness takes place in many intermediate metals irrespective of their crystalline structure.[1] It is observed in metals having a body-centered lattice: in iron, niobium, vanadium, tantalum,[2] in nickel having a face-centered cubic structure,[1,3] in α-titanium alloys with compact hexagonal structure.[4] Such a brittleness is observed in metals in which hydrogen is occluded endothermically (such as iron and nickel) or exothermically (titanium, vanadium, niobium, and tantalum); in the first metals, hydrogen may develop extremely large internal pressures, which are insignificantly low in the second group of metals.

Therefore, the mechanism of the hydrogen brittleness of these metals must contain some common factor which depends neither on their crystalline structure nor on the internal hydrogen pressure, although these factors must undoubtedly affect substantially the parts of the mechanism of brittleness. The common phenomenon in processes resulting in hydrogen brittleness of metals is the dislocating nature of both the mechanism of plastic deformation and of the fracturing process. Consequently, hydrogen brittleness must be due to the specific influence exerted by hydrogen on the dislocating motion during a plastic deformation and on origin and spread of cracks resulting in fracturing.

At temperatures below a certain critical point T_0, it may be assumed[6] that hydrogen will form Kotterell atmospheres in the dislocations. These atmospheres will accompany the dislocations by lagging a certain distance

behind them, when the rate of deformation is low and the temperature is not too low, so that the mobility of the hydrogen atoms is comparable to the rate of the motion of the dislocations.

The dislocations made to move by the applied stresses and the hydrogen atmospheres reach the boundaries of the grains where they form accumulations. Taking into consideration that the accumulation may consist of hundreds and more dislocations, it can be expected that the segregation of hydrogen near the grain boundaries will be considerable. A crack will appear at the apex of the accumulation near the grain boundary, when the effective stresses are fairly large.

It is often assumed that the cause which favors the formation and development of cracks in case of hydrogen brittleness is due to the reduced surface energy of the cracks by adsorption of hydrogen on its surface.[12, 14] This phenomenon, however, cannot be the decisive cause of fractures due to brittleness. According to modern ideas, the Griffits criterion, which determines the spontaneous development of cracks resulting in fracture, should be used in the following form:

$$\sigma \geq \left[\frac{2E(p + \gamma)}{\pi C}\right]^{\frac{1}{2}} ,$$

where γ is the surface energy of a crack; p is the energy of the plastic deformation of the metal connected with the spreading of the cracks; E is the modulus of elasticity; C designates the linear dimensions of the crack, and σ is the stress required for developing the crack. In plastic metals, the energy p of the plastic deformation is thousands of times larger than the surface energy γ and, if the surface energy is reduced by hydrogen even to zero, it will not change substantially the character of the spread of cracks.

Brittle fractures must be due to causes which reduce the work of plastic deformation (p) during the formation and spread of cracks. In metals, which absorb hydrogen endothermically, this work can be reduced drastically by the internal hydrogen pressure.[15, 16] For metals, such as titanium, this explanation is unacceptable, because their hydrogen pressure is negligible.

In this respect, it is interesting to note that tempered VT15 alloy becomes brittle at room temperature and at any rate of deformation, when its hydrogen content is larger than 0.25 percent by weight. Hydrides begin to appear in it at considerably larger concentrations of hydrogen in the order of 35 percent by weight. The impression is that, at hydrogen concentrations larger than 0.25 percent by weight (10 atomic percent), the lattice of the β-phase is stressed to a degree that makes the development of plastic deformations impossible. Similar

segregations of hydrogen in the β-phase may appear with deformations proceeding at low rate.

It should be also noted that all metals with exothermic occlusion of hydrogen form hydrides. Therefore, it is possible that in certain of these metals, the segregation of hydrogen in the zone of accumulated dislocations caused by plastic deformation will result in local appearance of submicroscopic hydrides with densities, as a rule, considerably less than of the metal and, therefore, resulting in considerable stresses. These hydrides and their related stresses favor the development of plastic deformation at the top of a crack.

In accordance with these concepts, hydrogen brittleness develops at a definite temperature interval which depends on the rate of deformation. The upper temperature threshold of hydrogen brittleness is determined by the temperature at which the Kotterell atmosphere is completely destroyed.

Hydrogen brittleness appears most intensively at temperatures below the temperature T of complete condensation of Kotterell atmospheres and is determined by the equation

$$T = \frac{U_{max}}{k \ln \frac{1}{C_0}} . \qquad (1)$$

where U_{max} is the maximum bonding energy of hydrogen atoms and dislocations; k is Boltzmann constant; and C_0 is the average concentration of hydrogen in the metal. The lower threshold of hydrogen brittleness is determined by the temperature at which the mobility of hydrogen atoms is too small to enable them to follow the dislocations. This temperature interval of hydrogen brittleness decreases with increased rate of deformation until it disappears completely (Figure 1). The critical rate of deformation at which hydrogen brittleness disappears is determined by the equation

$$v_{kp} = \frac{4b\rho kTD}{A} , \qquad (2)$$

where b is the Burgers vector of dislocation; ρ is the dislocation density; T is the temperature; D is the coefficient of diffusion of hydrogen.

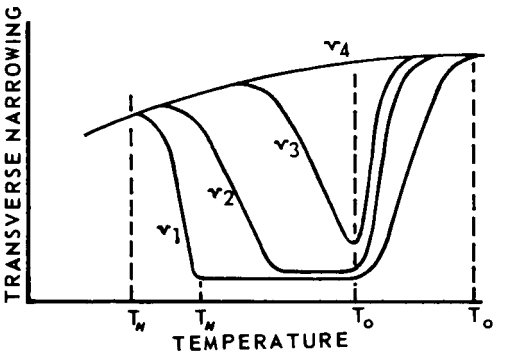


FIGURE 1. EFFECT OF EXPERIMENTAL TEMPERATURE ON TRANSVERSE NARROWING OF TITANIUM ALLOYS CONTAMINATED BY HYDROGEN AT DIFFERENT RATES OF DEFORMATION ($v_1 < v_2 < v_3 < v_4$)

The constant A in Eq. (2) is connected with the energy U_{max} by the relationship

$$U_{max} = \frac{A}{r + d}. \qquad (3)$$

In this relationship, r is the distance between a hydrogen atom and the center of dislocation; and d is the width of dislocation. It is assumed that the dislocation width is close to the distance between the atoms.

The experimental data obtained by Schwarzberg, Williams, and Jaffee[7] for an ($\alpha + \beta$)-titanium alloy Ti-140A (2-percent Mo, 2-percent Cr, 2-percent Fe) confirm the dislocation theory of hydrogen brittleness as described above.[6] The actual picture, however, may turn out to be more complex. Segregations of hydrogen formed by accumulation of dislocations are thermodynamically unstable. It is known[17] that when a load is removed from a slowly elongated specimen which is quickly broken at the expiration of a certain time, it will show no brittleness. Thermal diffusion eliminates the sources which cause a hydrogen brittleness at low rates of deformation. However, in breaking up the hydrogen segregation at the accumulation of dislocations, the thermal diffusion is effective not only after the load is removed, but also during the loading process. If the rate of deformation is not too low, the hydrogen segregations, caused by dislocations moving the hydrogen atoms to an obstacle, are not significantly destroyed by thermal diffusion. However, if the rate of deformation is extremely small, the thermal diffusion will predominate over the hydrogen segregation in the zone of accumulation and there will be no oversaturation of hydrogen by the metal.

Consequently, the plasticity must be restored not only at very low testing temperatures, but also at very low rates of deformation. In other words, hydrogen brittleness of this type should be observed not only at a definite temperature interval, but also at a definite range of rates of deformation.

Unfortunately, direct experimental data to confirm this hypothesis are still not available. It is possible, however, to cite the results[18] obtained for an ($\alpha + \beta$)-alloy of titanium containing 4-percent Al, 4-percent Mn, and 0.02-percent H_2 by protracted static tests at room temperature of the strength of smooth specimens (Figure 2). It was found that the destruction is rapid and

plastic at stresses of 112 to 116 kg (force)/mm^2. At stresses less than 108 kg (force)/mm^2, the destruction begins only after a long (more than 100 hours) action by the stresses, but it still remains plastic. Brittle destruction is observed only in a very narrow range of stresses, 112 to 108 kg (force)/mm^2, and begins at the expiration of 2 to 50 hours after the application of the load.

The curves illustrating the change in transverse narrowing with temperature (Figure 1) should be modified when an important role is played by thermal destruction of hydrogen segregations during the elongation. At low rates of deformation, an improvement in plasticity may take place not as a result of destruction of Kotterell atmospheres, but by thermal break-up of the segregations. The lower the rate of deformation, the longer will be effective the thermal break-up and, therefore, the more complete will be the sources of brittle destruction eliminated. Consequently, the minimum of plasticity should be displaced to lower temperatures when the rate of deformation is lowered (Figure 3).

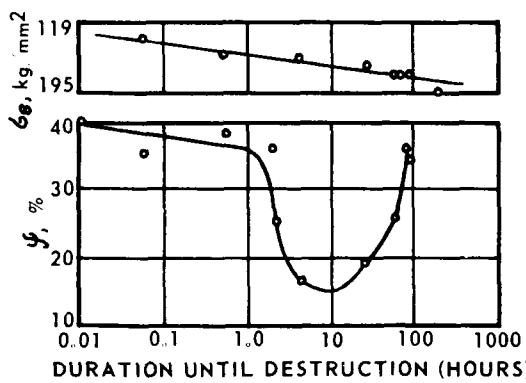


FIGURE 2. TRANSVERSE NARROWING OF ALLOY Ti + 4% Al + 4% Mn WITH 0.02% H_2 DEPENDING ON TIME UNTIL DESTRUCTION UNDER STATIC LOAD

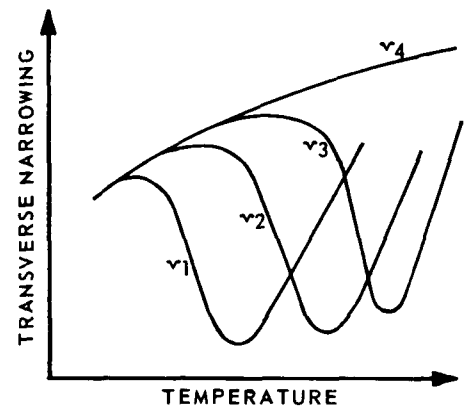


FIGURE 3. EFFECT OF TEST TEMPERATURES ON TRANSVERSE NARROWING OF TITANIUM ALLOYS AT DIFFERENT RATES OF DEFORMATION BY TAKING INTO ACCOUNT THE THERMAL BREAK-UP OF HYDROGEN SEGREGATION ($v_1 < v_2 < v_3 < v_4$)

The dislocation theory of hydrogen brittleness was checked experimentally for an $\alpha + \beta$-alloy of VT3-1 titanium subjected to a wide range of temperatures and rates of elongation.

The investigations were carried out for a forged material made of an industrial series-produced ingot which, after its hydrogenation, was annealed as follows: heated to 870°C, soaked for 1 hour, cooled in air to 650°C, held at this temperature for 1 hour, and cooled in air. The elongation was tested with Gagarin-type smooth specimens on a mechanically driven machine.

As an example, Figure 4 shows the effect of the temperature used for testing on the mechanical properties of VT3-1 alloy with 0.05-percent hydrogen at different rates of elongation. The ultimate tensile strength of the alloy increases with decreasing temperature. For example, with a rate of deformation equal to 4 mm/min, the ultimate tensile strength is 122 kg/mm^2 at +20°C and 148 kg/mm^2 at -175°C. However, the difference between the tensile strength at all three rates of deformation is very small and is close to the possible errors made during the experiment.

The transverse narrowing at a rate of elongation equal to 0.4 mm/min is sharply reduced from 35 percent at +22°C to 4 percent at -18°C. At a temperature further reduced to about -40°C, the transverse narrowing remains unchanged at about 5 percent; it increases somewhat at still lower temperatures and reaches 9 percent at -70°C. The elongation at all three rates is approximately the same for temperatures ranging from +20°C to -70°C and decreases somewhat at still lower temperatures.

At a rate of deformation of 4 mm/min, the transverse narrowing varies with temperature nearly in the same manner as at a rate of 0.4 mm/min. The only difference is that the temperature at which the transverse narrowing begins to decrease sharply and the temperature corresponding to a minimum of plasticity are both higher: +12°C and -8°C, respectively. At a rate of elongation of 20 mm/min, the transverse narrowing decreases in a very narrow temperature range from -40°C to +2°C.

The cited experimental data for each temperature used in testing can be presented in the form of coordinates (properties-rate of deformation). Such curves are shown in Figure 5 for two test temperatures: +2°C and -40°C. According to the data, at a temperature of +2°C, hydrogen brittleness is actually present within a definite range of rates of deformation (0.4 - 20 mm/min). At a temperature of -40°C, however, no restoration of plasticity is observed at a rate of deformation of 0.4 mm/min, because the thermal diffusion is very weak at this temperature. A restoration of plasticity at this temperature should be expected at considerably lower rates of deformation.

Consequently, the experimentally obtained data confirm, at least qualitatively, the dislocation theory of hydrogen brittleness in accordance with the pattern shown in Figure 3.

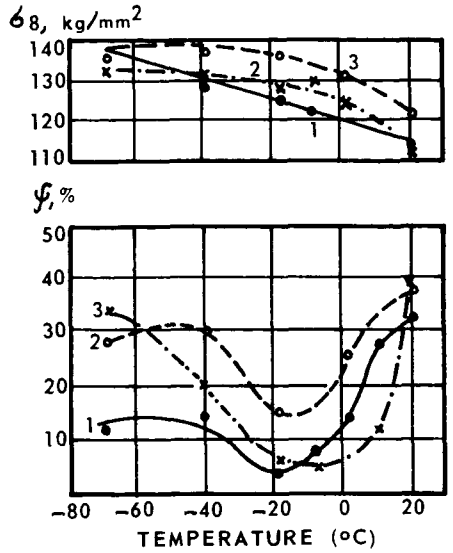


FIGURE 4. MECHANICAL PROPERTIES OF VT3-1-ALLOY WITH 0.05% H_2 DEPENDING ON TEMPERATURE: RATE OF ELONGATION, mm/min: 1 = 0.4; 2 = 4; 3 = 20

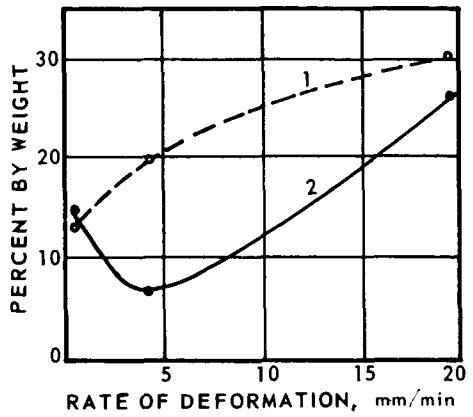
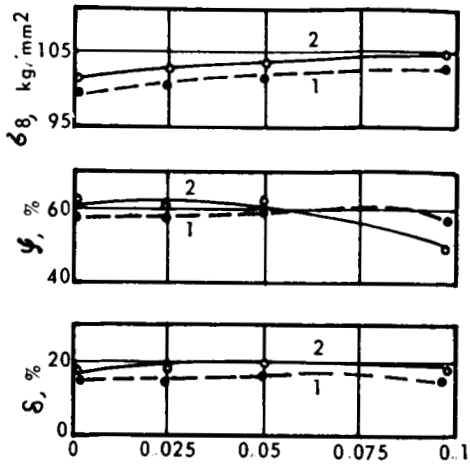


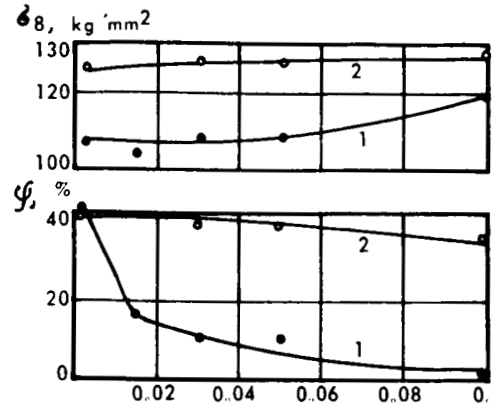
FIGURE 5. TRANSVERSE NARROWING OF VT3-1-ALLOY WITH 0.05% H_2 DEPENDING ON RATE OF DEFORMATION AT -40°C (1) AND AT +2°C (2)

It is commonly assumed that titanium alloys with a structure represented by the β-phase have no tendency to hydrogen brittleness. In fact, Jaffee, Lenning, and Craighead[8] found no reduced plasticity in alloys of Ti + 13-percent Mo and Ti + 20-percent Mo tempered to a β-phase, even when these alloys contained 0.2-percent hydrogen by weight and were tested at 25°C with the arm of the breaking machine moving at the rate of 0.1 mm/min. According to their data, an alloy of Ti + 9-percent Mn has no tendency to hydrogen brittleness with concentrations of hydrogen ranging from 0.0015 to 0.075 percent, when tested mechanically under the same conditions.

As shown by A. S. Moroz and Yu. D. Khesin,[9] an alloy containing 3-percent Al, 13-percent V, and 11-percent Cr with a stable β-phase after annealing and belonging, therefore, to β-alloys, has no tendency to hydrogen brittleness at room temperature after its annealing. We obtained similar data for the domestic β-titanium alloy VT15 (Figure 6).

For concentrations of 0.002 to 0.1 percent by weight, the properties of the VT15-alloy are not affected substantially by hydrogen, irrespective of whether the rate of deformation is large or small.





H_2 PERCENT BY WEIGHT

FIGURE 6. MECHANICAL PROPERTIES AT ROOM TEMPERATURE OF VT15-ALLOY ANNEALED FROM 780°C, DEPENDING ON CONTENT OF HYDROGEN IN ALLOY: RATES OF ELONGATION 1 = 0.4; 2 = 4 mm/min

FIGURE 7. MECHANICAL PROPERTIES OF VT15-ALLOY AT -18°C AS A FUNCTION OF THE CONTENT OF HYDROGEN IN THE ALLOY: RATES OF ELONGATION: 1 = 0.4; 2 = 20 mm/min

It should be expected, however, that β-titanium alloys should have a tendency to hydrogen brittleness at a definite range of temperatures. The VT15-alloy, just as other metals with a space-centered cubic lattice, becomes brittle when the temperature is fairly low. In testing the elongation at a rate of 6 mm/min, the temperature of transition from plastic to brittle state is about -70°C.

Inasmuch as metals with a space-centered lattice have a greater tendency to brittle destruction, they are even more sensitive to contamination by impurities. In fact, nitrogen and oxygen reduce sharply the plasticity of β-alloys at low temperatures. Hydrogen is hardly an exception.

In this respect, it should be noted that the above data were obtained by testing the mechanical properties only at room temperature. Generally, such data cannot be used as proof that β-titanium alloys, as a rule, have no tendency to hydrogen brittleness.

Actually, mechanical tests carried out at a temperature of -18°C had shown that even the first addition of hydrogen (0.015 percent by weight) to the alloy results in a sharply reduced transverse narrowing at low rate of deformation (Figure 7). A VT15-alloy, which is first annealed in vacuum and then quenched

from 780°C in water, has a transverse narrowing of more than 40 percent at -18°C; but the narrowing of the alloy containing 0.015 percent by weight of hydrogen is reduced to 15 percent, although the alloy received the same heat treatment.

There is a difference between this reduction in plasticity and the brittleness of the alloy at temperatures below the point of transition to a brittle state. While "its own" brittleness of the VT15-alloy appears at all rates of elongation, the plasticity of the alloy at a temperature of -18°C increases with the increase in rate of elongation. With the arm of the breaking machine moving at the rate of 20 mm/min, the transverse narrowing decreases only slightly with the increase in content of hydrogen, but no typical brittle destruction is observed even at 0.1 percent by weight. The specimen breaks with a formation of a neck at about 35 percent of transverse narrowing.

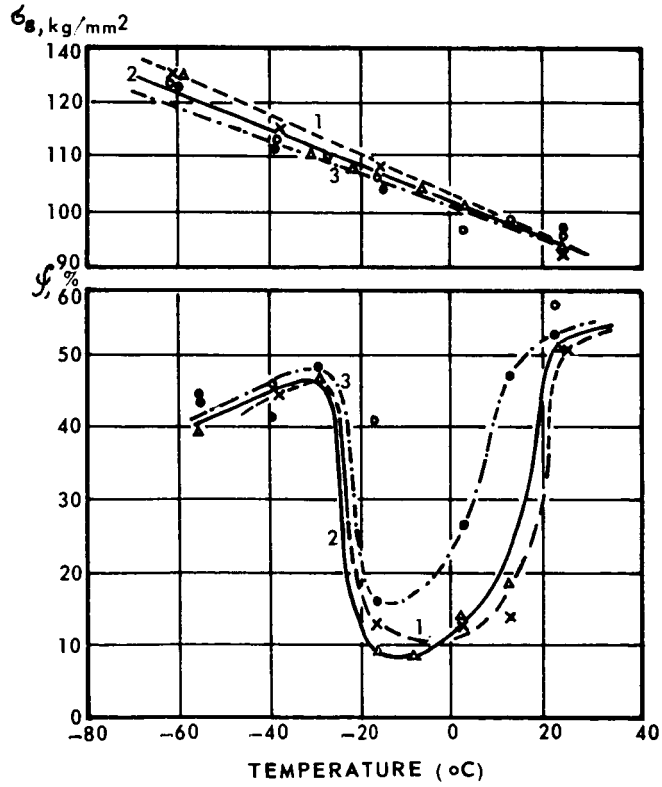


FIGURE 8. MECHANICAL PROPERTIES OF VT15-ALLOY CONTAINING (1) 0.05%, (2) 0.03%, and (3) 0.015% HYDROGEN, DEPENDING ON TEMPERATURE OF THE TEST: RATE OF ELONGATION = 0.4 mm/min

Just as in the (α + β)-titanium alloy VT3-1, the hydrogen brittleness of the VT15-alloy appears within a definite temperature range (Figure 8). At a

deformation rate of 0.4 mm/min, the transverse narrowing decreases sharply with a decrease in temperature. This reduction takes place within a narrow range of temperatures. Using the point at which the transverse narrowing is reduced to half as the upper temperature for the appearance of hydrogen brittleness, this temperature will be equal to +8°C, +2°C, and -8°C for hydrogen concentrations of 0.05, 0.03, and 0.015 percent, respectively. Consequently, the higher the content of hydrogen the higher will be the upper temperature for the appearance of hydrogen brittleness.

The restoration of plasticity at sufficiently low temperatures (below -30°C) does not reach its full extent because of the brittleness possessed by the alloy. As should have been expected from the dislocation theory of hydrogen brittleness, the VT15-alloy is destroyed in a plastic manner at the investigated range of temperatures and at 20 mm/min as the rate of deformation. Brittle destruction appears only at temperatures below -70°C.

Since the widely held opinion about the absence of hydrogen brittleness in β-alloys was not confirmed experimentally, it would be interesting to return to the hypotheses on hydrogen brittleness of $(\alpha + \beta)$-alloys advanced earlier by several investigators.[8,9,10] In the final analysis, hydrogen brittleness was linked with processes taking place at the surface separating the α- and the β-phases, since it was assumed that neither the α- nor the β-alloys have a tendency to hydrogen brittleness at a low rate of deformation.

Generally speaking, there were no good reasons for this conclusion, since the composition of the β-phase in the $(\alpha + \beta)$-alloys in which hydrogen brittleness was observed at room temperature did not coincide with the composition of the investigated β-alloys. It is entirely possible that hydrogen brittleness could be observed at room temperature in β-titanium alloys having the same β-phase as the investigated $(\alpha + \beta)$-titanium alloys.

As far as the β-phase in the VT3-1 alloy is concerned, the latter has almost the same composition as the VT15 alloy which contains on the average 3-percent Al, 11-percent Cr, and 7-percent Mo versus the content of the β-phase in VT3-1 alloy of 2-percent Al, 11-percent Cr, and 10-percent Mo after an isothermic annealing.[11] Therefore, it is not surprising that hydrogen brittleness in VT3-1 alloy is observed at practically the same range of temperatures as in the VT15 alloy, although there is no complete analogy as, for example, the presence of hydrogen brittleness in VT3-1 alloy at a deformation rate of 20 mm/min and its absence in the VT15 alloy at the same rate of deformation.

The incomplete coincidence in the ranges of hydrogen brittleness in the VT3-1 and VT15 alloys is due primarily to the fact that the β-phase in the VT3-1 alloy is equal to about 20 to 30 percent of the total volume of the alloy. Since

it is in the β-phase where the hydrogen is concentrated in $(\alpha + \beta)$-alloys, therefore, for the same average concentration its content in the β-phase of the VT3-1 alloy is approximately 3 to 4 times larger than in the β-phase of the VT15 alloy. Therefore, for the same average concentration of hydrogen, the β-alloys are less sensitive to hydrogen brittleness than the $(\alpha + \beta)$-alloys.

Although incomplete, the hydrogen brittleness ranges coinciding in VT15 and VT3-1 alloys indicate that the boundaries separating the α- and β-phases are not responsible for the brittleness of $(\alpha + \beta)$-alloys. The intergranular character of destruction by hydrogen brittleness is not due to the specific nature of the processes taking place at the boundaries separating the two different phases, but is due to the formation of accumulations of dislocations at the grain boundaries, irrespective of the phase to which the grains belong. Since in $(\alpha + \beta)$-alloys the concentration of hydrogen is in the grains of the β-phase, therefore, these are the grains where hydrogen brittleness will develop. Also, the rate of hydrogen diffusion in the α-phase is considerably less than in the β-phase; therefore, even when the hydrogen concentration in the α-phase is sufficiently large, hydrogen brittleness in it should develop at considerably lesser rates of deformation, perhaps less by several orders.

It would be interesting to compare the experimentally obtained data with what could be expected from the dislocation theory of hydrogen brittleness. Suitable for experimental direct checking is the critical rate of deformation determined by Eq. (2). The value of A in this equation can be estimated from the relationship (3) which determines the bonding energy. In its turn, the value of the bonding energy can be determined from the relationship (1). The minimum plasticity of the VT15-alloy containing 0.05-percent hydrogen by weight is observed at low rates of deformation at temperatures of about -10°C. From this it follows that according to Eq. (1) the bonding energy should be equal to 0.15 electron volts. This value is somewhat larger than the energy bonding the hydrogen atoms in iron and nickel (0.1 and 0.08 ev, respectively). Assuming that the energy bonding the hydrogen atoms with the dislocations in a VT15-alloy is equal to 0.15 ev, A is equal to 1.1×10^{-20} ergs/cm.

The coefficient of hydrogen diffusion in VT15-alloy at low temperatures was determined experimentally. For this purpose, specimens of VT15-alloys were electrolytically saturated with hydrogen at 18°C, 40°C, and 80°C in a fully normal solution of sulfuric acid followed by determination of distribution of hydrogen by mass-spectrographing the cross section of the specimen. The coefficient of diffusion of hydrogen in VT15-alloy at given temperatures was found by adjusting the Gauss curve for the experimental data. The extrapolation of data for a temperature of -18°C provides a value of about 10^{-10} cm^2/sec.

Using $10^8 cm^{-2}$ as the density of dislocations in VT15-alloy and b = 2.8 x 10^{-8} cm as the Burgers vector, we find that the critical rate of deformation should be close to 0.04 1/min. This, for a 30-mm long effective part of the specimen, corresponds to moving the arm of the breaking machine at a rate of 1.2 mm/min.

Actually, the VT15-alloy with 0.05-percent hydrogen at a temperature of -18°C and at rates of deformation of 20 and 4 mm/min is destroyed plastically, but in a brittle manner at a rate of 0.4 mm/min. This conformity with the experimental data confirms once more the expediency of using concepts of dislocation for solving the problems of hydrogen brittleness.

CONCLUSIONS

1. It was shown that hydrogen brittleness of (α + β)-titanium VT3-1 alloy develops most intensively at temperatures below room temperatures and at low rates of deformation.

2. It was found for the first time that below the room temperature, the β-titanium alloys also have a tendency to hydrogen brittleness. The temperature for the appearance of hydrogen brittleness decreases with the decrease in content of hydrogen. The hydrogen brittleness of VT15-alloy appears only at low rates of deformation in a narrow range of temperatures, approximately from -30°C to +10°C.

3. Hydrogen brittleness of (α + β)-titanium alloys is due to processes developing in the β-phase during plastic deformation.

4. Since the composition of the β-phase in VT3-1 alloy after an isothermic annealing is close to the composition of the β-phase of VT15 alloy, the hydrogen brittleness in them develops nearly identically during the same conditions for the mechanical tests.

5. The main laws governing the appearance of hydrogen brittleness in VT3-1 and VT15 alloys are satisfactorily explained by the dislocation theory of hydrogen brittleness based on the assumption that hydrogen forms Kotterell atmospheres in the dislocations which in a definite temperature range are carried away by the dislocations. The dislocations carry the hydrogen to the grain boundaries, which results in hydrogen segregations favoring the formation and spread of cracks.

6. It is assumed that the formation and spreading of cracks is helped not by the adsorption of hydrogen on the surface of the cracks, but by the formation of dislocation accumulations enriched with hydrogen, by submicroscopic emissions of hydrides which hinder the plastic deformation at the apex of the cracks, or by strengthening the metal at the zone of accumulations due to the hydrogen segregation reaching a degree at which plastic deformation of the metal at the top of a crack becomes impossible.

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13. ABSTRACT

Discussed is the development of hydrogen brittleness in an (α + β)-titanium alloy VT3-1 and a β-titanium alloy VT15. It is shown that hydrogen brittleness develops most intensely in VT3-1 at temperatures less than room temperature and at low strain rates. It is discovered that β-titanium alloys at temperatures below room temperature are also prone to hydrogen brittleness, the temperature at which this occurs decreasing with a decrease in the hydrogen content. Hydrogen brittleness in the alloy VT15 is found to occur only at low strain rates in a narrow temperature range, approximately from -30°C to +10°C. Hydrogen brittleness of (α + β)- titanium alloys is believed to be caused by processes developing in the β-phase during plastic deformation. Since the composition of the β-phase of VT3-1 after isothermal annealing is close to the composition of the β-phase of VT15, it is concluded that hydrogen brittleness develops in approximately the same manner in both of them, provided the conditions for mechanical testing are the same for both.

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