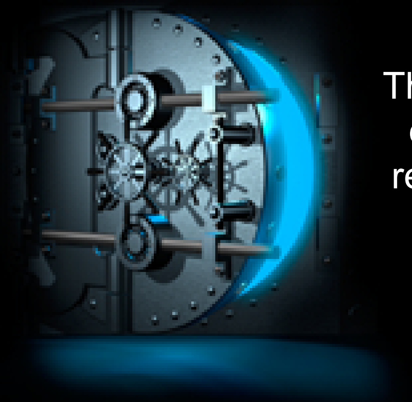

DEPARTMENT OF DEFENSE
UNITED STATES OF AMERICA

DEPARTMENT OF THE AIR FORCE
NATIONAL AIR & SPACE INTELLIGENCE CENTER
WRIGHT-PATTERSON AFB OHIO

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APR 25 2016

John Greenewald
[redacted]

Dear Mr. Greenewald,

This letter is in reference to your Freedom of Information Act (FOIA) request dated 25 February 2016 for a copy of a document entitled *The Accelerating Mechanism in Cold Fusion*. We received your request from the 88 CS/SCOKI FOIA office and assigned case number 2016-02311-F to it.

A review was conducted to determine if the record you requested may be released in whole or in part. After reviewing the document it has been determined that it may be released.

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Sincerely,

TRISHA M. SEXTON, Colonel, USAF
Vice Commander

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②

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AIR FORCE SYSTEMS COMMAND

THE ACCELERATING MECHANISM IN COLD FUSION

by

P.I. Golubnichiy, V.A. Tsarev, V.A. Chechin

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By: P.I. Golubnichiy, V.A. Tsarev, V.A. Chechin

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Д д	*Д д*	D, d	Ф ф	*Ф ф*	F, f
Е е	*Е е*	Ye, ye; E, e*	Х х	*Х х*	Kh, kh
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И и	*И и*	I, i	Ш ш	*Ш ш*	Sh, sh
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М м	*М м*	M, m	Ь ь	*Ь ь*	'
Н н	*Н н*	N, n	Э э	*Э э*	E, e
О о	*О о*	O, o	Ю ю	*Ю ю*	Yu, yu
П п	*П п*	P, p	Я я	*Я я*	Ya, ya

*<u>ye</u> initially, after vowels, and after ъ, ь; <u>e</u> elsewhere.
When written as ё in Russian, transliterate as yё or ё.

RUSSIAN AND ENGLISH TRIGONOMETRIC FUNCTIONS

Russian	English	Russian	English	Russian	English
sin	sin	sh	sinh	arc sh	$\sinh^{-1}$
cos	cos	ch	cosh	arc ch	$\cosh^{-1}$
tg	tan	th	tanh	arc th	$\tanh^{-1}$
ctg	cot	cth	coth	arc cth	$\coth^{-1}$
sec	sec	sch	sech	arc sch	sech^{-1}
cosec	csc	csch	csch	arc csch	csch^{-1}

Russian	English
rot	curl
lg	log

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THE ACCELERATING MECHANISM IN COLD FUSION

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Preprint 149
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1. Introduction

The report of the observation of a "cold fusion" reaction accompanying the saturation of palladium and titanium by deuterium ~~[1,2]~~ appeared entirely unexpectedly and has attracted enormous interest worldwide. Meanwhile, however, references to the possibility of such phenomena go back some years. Apparently the first to call attention to such phenomena were Peneth and Peters in [3], work which was then completely forgotten and only recently resurrected in the memory and made widely known in the wake of the recent "thermonuclear boom." The application filed by Stanyukevich, Burlakov and Kharoshkin in 1973 [4] in connection with a discovery of fusion reactions associated with cavitation is another, less well known, example. It was then more than 10 years later, in 1986, that the Deryagin group published findings from observations of the emission of neutrons accompanying the mechanical breakdown of heavy ice and LiD single crystals, which the authors interpreted as evidence of a fusion reaction [5]. There may be other, similar, examples. The mechanism which triggers these reactions is still not understood. What is more, that such a phenomenon as cold fusion in fact exists can clearly not now be regarded as a conclusively established fact. It nevertheless appears to us desirable to take a look at some of the thinking on the nature of the phenomenon involved here, since this could prove useful when it comes to charting more promising directions for future research. We are going to be looking here at the "accelerating mechanism" of cold fusion proposed in [6], where it is described in qualitative terms. The present paper discusses this model in greater detail and present a number of assessments in quantitative terms.

Russian translations.
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The hypothesis concerning the accelerating mechanism involved in cold fusion rests on two experimental observations: a) the development of microcracks in association with the absorption of hydrogen by Pd and Ti, which absorb hydrogen and its isotopes readily and b) the mechanoemissions accompanying the breakdown of various crystalline substances and adhesion layers. The first of these phenomena is well known from practical experience gained from work with hydrides and is due to an increase in the dimensions of the initial metal lattice which occurs as the hydrogen dissolves within it, the development of mechanical stresses and the loss of plasticity in the hydride phase. For purposes of the following discussion we are going to be focusing in Sections 2-4 on a number of questions concerning the structure, formation and breakdown of metal hydrides. We will be looking primarily at the transition metals, which dissolve hydrogen and its isotopes readily, and of the metals in this category, we will be giving our attention chiefly to palladium and titanium, in which, according to [1, 2], cold fusion reactions have been observed. More detailed discussions can be found in the numerous monographs, surveys and original articles dealing with metal-hydrogen systems (see [7-11], for example).

The mechanoemission of electromagnetic waves accompanying the breakdown of crystals and adhesion layers has also been observed and studied in many experiments (see, for example, [12, 13]). Interpretations of these phenomena will ordinarily develop the idea that charges build up along the edges of the cracks which create an electrostatic field. It has been hypothesized that the latter can then accelerate electrons and ions in the gap to energies of $\sim 1\text{-}10^2$ keV and trigger the field emission of electrons. This generates bremsstrahlung and characteristic radiation, which can be observed experimentally.

It should be emphasized here, however, that attempts to carry an accelerating mechanism which may accompany the breakdown of ionic crystals, dielectrons and semiconductors directly over to transition metals (Pd, Ti etc.) and their hydrides will encounter difficulties associated with the metallic type of bond characteristic of these materials. In Section 5 we will be looking at a number of physical phenomena which nevertheless could also function as an accelerating mechanism in the case of the transition metal hydrides of interest to us. In Section 6 we will offer a number of estimates for an accelerating mechanism of the simplest type. Section 7, finally, looks at a number of different products of the model proposed here together with possibilities for testing it experimentally.

2. Structure and bonds in the metal-hydrogen system

Hydrogen and metals form compounds with different types of bond. Hydrides of the most active alkali and alkaline-earth metals follow the halide pattern and will characteristically have an ionic bond. Rare-earth metals and hydrogen form compounds with a covalent and a metallic bond, whereas in the case of the transition metals the metallic bond predominates. The latter compounds are frequently regarded as interstitial solutions, in which the hydrogen atoms, being of relatively small dimensions, occupy the lattice spaces between the atoms of the metal. This alters the properties of both the metal and the hydrogen substantially. That is, chemical interaction occurs. The properties of the hydrides of the transition metals are characteristic of, on the one hand, metals (high electric and thermal conductivity, the low thermoelectric coefficients, weak Hall effect) and, on the other, of nonmetallic materials containing electronegative hydrogen (the evidence here would include the volume relationships, the data from x-ray analysis and the possibility of combining hydrogen and oxygen within the crystal lattice, in Ti for example [9].

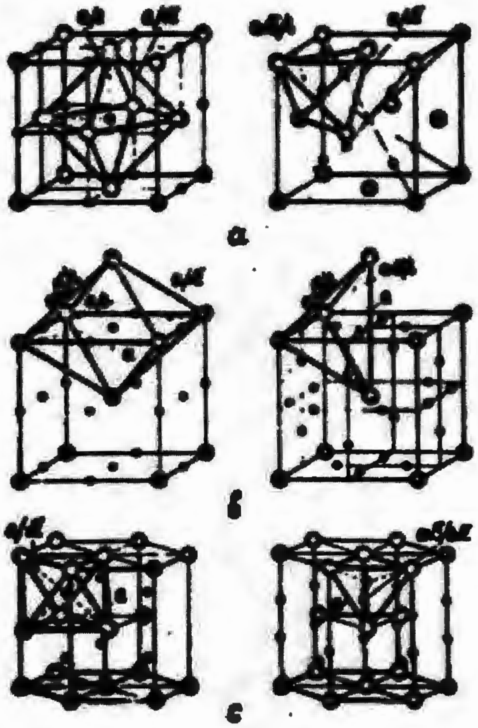

Fig. 1. Interstitial sites in metallic lattices bcc (a), fcc (b) and GPU*: left - octahedrons, right - tetrahedrons

The introduction of hydrogen atoms into the interstices of the metal lattice deforms the latter and creates a displacement field. Hydrogen in metal may exist in a variety of phase states: a gas-liquid-solid substance. The first two differ only in terms of the density of the hydrogen and the value of the lattice constant of the metallic matrix. Characteristic of the solid phase will be the ordering of the interstitial hydrogen atoms with respect to equivalent interstices.

Within the cubic lattices the hydrogen may occupy two types of interstice: octahedral, in which the hydrogen atom is surrounded by six atoms of metal, and tetrahedral interstices, where the H atom is enclosed by four atoms of metal (see Fig. 1). For each atom of metal there will be one octahedral and two tetrahedral spaces, the dimensions of which in the case

---

[*] expansion not given

of the face-centered cubic lattice will be related to the atomic radius of the metal by the following relationship:

$$R_o \simeq 2R_T \simeq 0{,}44\ R_{Me} \qquad (1)$$

Depending on the type of bond in which the hydrogen is participating, the ionic radius may take a number of values: $R_H^{(K)} \simeq 0{,}30 - 0{,}35$ Å for a covalent bond, $R_H^{(И)} \simeq 1{,}3 - 1{,}5$ Å in the case of an ionic bond and $R_H^{(Me)} \simeq 0{,}41$ Å for a metallic bond. The radius of electronegative hydrogen diminishes continuously with increases in the electronegativity of the cation and in the case of the transition hydrides will be ~ 1,23 Å for Pd, ~ 1,25 Å for Ti (and ~ 1,33 Å in the case of La). The mean value for the transition metals will be $R_{H^-}^{(Me)} \simeq 1{,}29 \pm 0{,}5$ Å [7] .
Depending on the location of the partner element within the period system, the effective charge on the hydrogen atom can vary from +1 to -1. Accordingly, two alternative models are frequently employed to describe the electronic characteristics of the metal-hydrogen system: an anionic model, in which the hydrogen attracts an electron of the metal and is transformed into the negative ion H^-, and a proton model, in which the hydrogen releases its electron into the conductivity region, which transforms it into a positively charged proton. In real-world systems we may observe an intermediate situation (in palladium, for example, hydrogen will evidently carry a mean charge of $\approx +0{,}45\,e$ [14]). This means that in the transition metals hydrogen does not exist in the form of a lone proton, but rather carry an average of ~ 0.1-0.6 electrons per atom. The remaining electrons not associated with the H will participate in a residual metallic bond, which is responsible for the electric conductivity of these compounds.

Let us now take a closer look at titanium and palladium. The hydrides of these metals are among those which have been studied most thoroughly. Let us recall here that the ability of palladium to absorb large amounts of hydrogen was established more than 150 years ago, specifically, by the investigations of Graham in 1866. Systematic study of the diffusion of gases in metals began with Pd [15]. As far, now, as the Ti hydrides are concerned, detailed investigation of these materials was encouraged by the extensive use of this metal in industry and the prospect of being able to store hydrogen in a hydride phase.

The palladium hydride structure exhibits an isotropically extended face-centered cubic matrix lattice. Below 300°C, the homogeneous solid solution breaks down into an α phase with low hydrogen content ($c_\alpha^{max} = 0{,}008$) and a β phase with a high concentration of hydrogen H ($c_\beta^{min} = 0{,}607$) and a higher lattice constant. Fig. 2 is a phase diagram of the PdH_c system. At room temperature, the values of the lattice constant for pure Pd and the accompanying α and β phases will be

$$a_{Pd} = 3{,}890\ \text{Å}\ ;\ a(c_\alpha^{max}) = 3{,}894\ \text{Å}\ ;\ a(c_\beta^{min}) = 4{,}025\ \text{Å}$$

The $\alpha \to \beta$ transition is accompanied by an increase in the volume of the hydride $\Delta V = 1{,}87\ cm^3$ / g. at. H [11]. When the concentration of hydrogen rises above c_β^{min} we will begin to observe a continuous expansion of the homogeneous phase. Taking account of the fact that the atomic radius of Pd is $R_{Pd} = 1{,}34$ Å , we find

$$R_o^{(Pd)} \simeq 0{,}6 \text{ Å} \ ; \ R_T^{(Pd)} \simeq 0{,}3 \text{ Å}$$

from (1) for the dimensions of the octahedral and tetrahedral vacancies. It is clear that when $R_H^{(ne)} = 0{,}41$ Å the hydrogen can be easily accomodated in the octahedral vacancies of the Pd matrix and will not "slip into" the tetrahedral vacancies. It is precisely because of the occupation of the octahedral sites that we see the formation of the α and β hydride phases, the full population of which corresponds to the C→1 transition. We have so far not taken account of the differences between hydrogen and its isotopes - deuterium and tritium. It should be noted here that the hydrides of palladium which contain D and T exhibit an "inverse" isotope effect. That is, they have higher D and T equilibrium pressures as compared with H with equal temperatures and identical compositions. Accordingly, as compared with hydrogen, deuterium will be more advantageously found in a gas, hydrogen in the metal [11]. For purposes of our discussion here, however, these differences are not really of critical importance, so unless stated specifically, we will not be distinguishing between H, D and T in the metals.

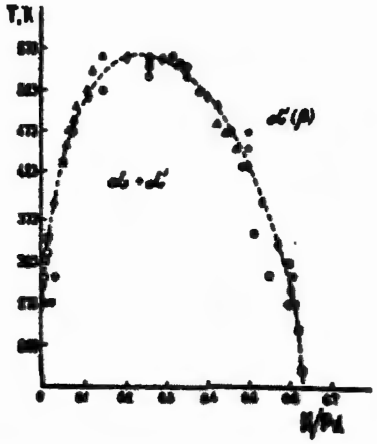

Fig. 2. Phase diagram of Pd-H

For the most interesting (from the cold fusion point of view) hydride phases of titanium with maximum concentration we have the compound TiH_2. Here, the γ phase, which occurs at temperatures of $T < 310K$, has an fct lattice with $a = 4.528$ Å and $c = 4.279$ Å, the γ' phase ($T>310$ K) an fcc lattice with $a = 4.454$ Å for $TiH_{1.99}$ (for $TiD_{1.99}$ $a = 4.440$ Å) [7]. Let us note here that the atomic radius of titanium $R_{Ti} = 1.46$ Å. The dimensions of the vacancies in the fcc lattice will be $R_o \approx 0{,}64$ Å, $R_T \approx 0{,}32$ Å $< R_H^{(Me)} = 0{,}41$ Å. Nevertheless, as structural studies have shown, in both ordered γ and γ' phases of titanium hydride the hydrogen is going to occupy the tetrahedral vacancies. So the titanium hydrogenation possibilities - and this applies to the analogs of titanium as well - are going to be limited to two atoms of hydrogen to one atom of metal, this despite the fact that, generally speaking, three interstitial sites are going to be available. (Let us note here that a similar situation will be observed in the case of Group V metals (V, Nb, Ta) as well.) We will be returning to this question later in Section 5.

3. Hydrogen transfer in metal and formation of the hydride layer

Investigators studying the metal-hydrogen systems can induce hydrogen (or hydrogen-isotope) saturation in a number of ways: from a pressurized medium, by electrolysis, ion implantation and by gas or spark discharge. Among a good many cold fusion researchers the first two methods are most popular, what with the fact that they are fairly uncomplicated and permit the development of high concentrations of deuterium. Let us look now at some of the fundamental mechanisms involved in the diffusion of hydrogen in metals. As is commonly known, the penetration of the metal by the hydrogen is preceded by the absorption of hydrogen molecules on the surface of the metal and the transition to a chemisorbed, near-atomic, state, but with a nonzero effective charge on the hydrogen atoms. Once the hydrogen has interacted with the surface layer of the metal, which is accompanied by its dissociation, the hydrogen atoms begin to diffuse into the crystal lattice. The hydrogen in the metal is highly mobile, and its diffusion coefficient in transition metals approaches the value for liquids. In describing the diffusion process, we can assume the atoms of the matrix to be stationary atoms forming the crystal foundation, within which the hydrogen "lattice gas" propagates. The diffusion process can be regarded as a migration of hydrogen atoms from one interstitial site to another, which involves the negotiation, so to speak, of an energy barrier created by the compression of the interstitial atom by the surrounding atoms of the metal. Digressing here from the quantum effects, we can assume that this diffusion is a product of the thermal vibration of the interstitial atom, magnitude of the barrier $\mathcal{E}$ depending on the elastic energy of the matrix and the dimensions of the interstitial atom. The diffusion coefficient D conforms to the Arrhenius law:

$$D = D_0 \exp\left(-\mathcal{E}/RT\right) \tag{2}$$

The values of the parameters of D_0 and $\mathcal{E}$ for the α hydride of Pd are as follows: $D_0 \simeq 6 \cdot 10^{-3}$ cm^{2}s^{-1} and $\mathcal{E} \simeq 25$ KJ mole^{-1}; for the α phase of the hydride of Ti: $D_0 \approx 10^{-2}$ cm^{2}s^{-1}, $\mathcal{E} \simeq 52$ KJ mole^{-1}, whereas in the case of the TiH_{2x} phase the value of D_0 is roughly an order of magnitude lower $(0{,}98) \cdot 10^{-3}$ [16] .

With the presence of a gradient in concentration and external forces F acting on the interstitial atoms, the total flow of hydrogen through the sample material can be expressed in the following form:

$$j = j_{\text{диф.}} + j_{\text{дрейф.}} = -nD\,\mathrm{grad}\,C + ncMF \qquad (3)$$

Here, n is the number of lattice atoms per unit volume and M the mobility of the interstitial atoms. The electric current generated by the field applied $E = \mathrm{grad}\,\phi$, creates a force F, the magnitude of which can be expressed in terms of the effective phenomenological charge Z^* of the interstitial atom: $F = -eZ^*E$. This relationship takes account of the fact that the interstitial atom is being acted upon by two component forces. One is governed directly by the force field and includes statistical screening and polarization effects. The second ("electron wind") is a product of the interaction of an interstitial atom with conduction electrons drifting under the effect of the external field.

Similarly, in the presence of a temperature gradient we observe the appearance of a force $F = -Q^*\mathrm{grad}\,T/T$, where Q^* is heat of diffusion. The time dependence of the local concentration $C(r,t)$ can be described by the following equation:

$$\partial c/\partial t = D\,\Delta C - MF\,\mathrm{grad}\,c \qquad (4)$$

where it is assumed that D, M, F are not functions of r. When $F = 0$, we have the following standard diffusion equation:

$$\frac{\partial c}{\partial t} = D\left(\frac{\partial^2 c}{\partial x^2} + \frac{\partial^2 c}{\partial y^2} + \frac{\partial^2 c}{\partial z^2}\right) \qquad (5)$$

The essential characteristics of the hydrogen transfer observed in cold fusion experiments can be easily grasped if we look at an extremely simple and yet fairly typical example of diffusion when $F = 0$ in a flat plate ($\ell_y, \ell_z >> \ell_x$). In this case, diffusion flow will be a function exclusively of x governed by the concentration gradient. Assuming that when $x = 0$ and $t \geq 0$ hydrogen concentration remains constant at C_0, and that the diffusion coefficient D is not a function of C, we can find the known solution of the problem with a constant source

$$c(x,t) = c_0 \exp(-x^2/4Dt) \qquad (6)$$

Diffusion attenuation time will be

$$\tau_{\text{диф}} \simeq L^2/D$$

where L is the characteristic distance over which C will vary. Let us note here that in this instance, when the drift term $MF\,\mathrm{grad}\,C$, predominates, the time to termination of transfer will be governed by the following relationship:

$$\tau_{\text{ДРЕЙФ.}} \simeq L/MF = L^2/MU$$

This will be a diffusion process if

$$U = LF < KT$$

In the case, now, of our problem, the situation is complicated by the fact that the saturation of the metal by the hydrogen (deuterium) may be accompanied by the formation of different hydride phases with coefficients of diffusion D_{ri} which, generally speaking, will be different from the diffusion coefficient D_M of the metal. The dynamics of these processes is discussed in [17]. The solution is easily found if we assume that in the vicinity of point x the hydride forms as soon as the hydrogen concentration reaches a certain level $C=C_r$. We will be assuming that concentration C_0 at the point of entry in the surface layer remains constant and that the coefficients of diffusion D_M and D_r are not functions of C. Let us note here that C_0 is the smaller of the two values for concentration - concentration in the gas phase whence occurs the penetration of the hydrogen into the metal and the maximum concentration which can be obtained in this metal.

From (6) we can find the hydrate penetration depth as a function of time

$$x_r(t) = 2\sqrt{D_M t \ln(C_0/C_r)} \tag{7}$$

and concentration at the exit surface of the metal plate

$$C_B(t) = C_r \exp\{-[l - x_r(t)]^2/4D_M t\} \tag{8}$$

where l is the thickness of the plate.

If several hydride phases with different diffusion coefficients coexist, solutions such as (7) and (3) should be joined at the phase boundaries. Expansion of the lattice in the transverse plane (y, z), particularly along the x axis, may alter the nature of the diffusion process to some extent. Of greater importance as far as we are concerned, however, is the fact that the increase in the dimensions of the lattice in the hydrate layer generates stresses, which at a certain critical hydride-layer thickness cause the material to crack. Because the cracks which now form fill with hydrogen, the flow of hydrogen along the x axis will initially drop sharply and then gradually reestablish itself as the cracks fill. Then, as the new layer of hydride reaches the critical thickness the cracking cycle begins again. Similar "wavelike" phenomena have been observed experimentally (in association with the diffusion of hydrogen in titanium in a gas-discharge plasma for example [18]).

4. Breakdown associated with hydrogen absorption

Whatever the hydrogen-absorption method involved, the energy of the hydrogen atoms is going to be of the order of several electronvolts, which is enough to escape the surface potential barrier. Continued penetration of the hydrogen into the material due to diffusion and the effect of the external fields will be accompanied by deformation of the lattice and, in some instances, by breakdown of the material. Let us underline here the fact that the possibility of breakdown with zero external mechanical load is not an obvious one. It is determined, among other things, by the chemical interaction between the hydrogen and the material, the characteristics of the developing crystalline phases, the rate of the diffusion and the kinetics of the microcracking. From this point of view, the problem under consideration here differs greatly from the problem of material failure due to radiation, the latter being a situation in which external active particles produce defects in the lattice directly in consequence of the development of the cascade, the formation of intensely ionizing recoil nuclei, "thermal spikes" etc. [21, 26].

The diffusion and the formation of hydride phases in the material together make for a highly nonuniform distribution of both the hydrogen atoms and the deformations and stresses. Because the hydride phases have different lattice dimensions, we will see the stresses develop not only within the phases, but along the boundaries between them. What is more, once the cracking begins, the cracks fill rapidly with hydrogen, thus becoming still another source of stress. And in the case of polycrystalline samples, we will observe very considerable, very intense diffusion of the hydrogen along the grain boundaries, in the dislocations and in the other defects. So in general terms what we see is the development of a highly complex, dynamic picture of diffusion processes acting and reacting on one another and of the formation of hydrides and cracks. Still another factor complicating the picture here is the change which occurs in the properties of the materials under the effect of exposure to the hydrogen (which is particularly significant in the vicinity of the "beak" portion of the crack). Exposed to hydrogen, materials which are originally fairly pliable will typically become harder and more brittle [19].

In [20] the authors formulate the general conditions governing the breakdown of brittle material under the effect of hydrogen.

First, hydrogen must be absorbed to the point where concentrations are sufficient to cause significant "swelling" of the entire lattice ($\lesssim 25\%$). These concentrations will ordinarily approach the hydride-formation concentrations.

Second, local stresses associated with hydrogen absorption must approximate breakdown stress.

Third, the hydrogen must saturate the material under conditions in which the relaxation of local stress can be disregarded (i.e., at sufficiently low temperatures).

Typical relative deformations associated with hydrogen saturation $\varepsilon = (a_h - a_l)/a_h$, where a_l and a_h are the lattice parameters in the different phases. Then, according to kinetic fracture theory, fracture time (life) for the material

$$\tau_{\text{разр}} = \tau_0 \exp\{C_a(\varepsilon^* - \varkappa\varepsilon)/\alpha k T\} \qquad (9)$$

where C_a is atomic heat capacity, α the coefficient of linear expansion, T temperature, maximum fracture deformation and ε^* the local load factor, which takes account of the fact that local stresses within the material may exceed mean stress. It is assumed here that, from the fracture point of view, the internal swelling stress (in the initial stage, at the level of microcrack formation) plays the same role as the external stress, the sole difference consisting in the fact that, unlike the external stress, the internal stress is a relaxing stress. Unfortunately, lifetime estimates according to formula (9) are highly uncertain, what with the fact that the parameter τ_0 is extremely small, while the exponent is large. If, for example, we employ a set of parameters which would be typical of electroncompounds such as $FeTi: \tau_0 = 10^{-13}$ s, $C = 3k$, $\alpha = 10^{-5}\ K^{-1}$, $\varepsilon^* = 0{,}2$, $\varepsilon = 0{,}083$, $T = 300$ K, we obtain [20]

$$\tau_{\text{разр}} \sim 10^{-13}\,\text{сек}\, \exp[200(1 - \varkappa/2{,}4)] \qquad (10)$$

The ambiguity in parameter $\varkappa$: $1 \le \varkappa \le 2{,}4$ produces a very high degree of ambiguity in τ : $\tau \sim 10^{38} \div 10^{-13}$ s. So we are forced to look to experimentation, on the assumption that breakdown time is going to be governed by the diffusion of the hydrogen and the formation of the hydride, which begins to crack almost immediately. Then, according to (2), when $L = 10^{-2}$ and $D = 10^{-8}$ cm^{2}s^{-1} (at room temperature), we obtain a characteristic cracking time of $\tau \sim$ $\sim 10^4$ s. As far, now, as the developing cracks are concerned, their dimensions, according to a range of experimental data, are going to be roughly 10^4-10^5 Å. Specifically, this is indicated by the fact that after a prolonged period of hydrogen absorption (~ 1 h), some materials (of the type represented by electron compounds) break down into a powder with these characteristic dimensions.

The breakdown of materials due to hydrogen absorption has been the subject of a great many theoretical and experimental studies. Most of these investigations have focused on the technical problems of the retention of hydrogen in the metal and fatigue failure in materials placed under load in a hydrogen atmosphere. Still lacking, however, is any really comprehensive quantitative theory of this process. It will be sufficient for our purposes here to proceed on the basis of the following assumptions, which conform to the experimental data:

1. The characteristic cracking depth L and time τ for a material are related by the diffusion relationship $L^2 \sim D\tau$. This means that as compared with diffusion time, the lifetime of the hydride phase can be disregarded.

2. The cracking process leaves cracks with characteristic dimensions of $\ell \sim 10^4+10^5$ Å and $d \sim 10^3+10^4$ Å in the longitudinal and transverse directions respectively.

3. Crack density n_T per unit volume is assumed to be such as to produce observable swelling in the material due to hydrogen absorption $\Delta V/V \sim 0{,}1$. If the mean volume of a crack is ϑ_0, $n_T \sim \Delta V/V\vartheta_0$. Microcracks are conventionally represented in the form of a near-plane disk. Then $\vartheta_0 \sim \ell^2 d \sim 0{,}1\,\ell^3 \sim 10^{-13}+10^{-10}\,\mathrm{cm}^{-3}$, $n_T \sim 10^{12}+10^{9}\,\mathrm{cm}^{-3}$.

Let us emphasize here that the cracks we are discussing here should be understood in too literal a sense. For the fact is that the cracking which occurs in the material is much more complex, the values $\ell, d, \vartheta_0, n_T$ describing only the mean characteristics. For example, the total area of the surfaces formed by the cracking (per unit volume) will be of the order of $\sim \ell^2 n_T$, the width of the irregular openings $\sim d$.

5. Possible explanations for the charges which develop in association with breakdown of hydrides of transition metals

Our discussion to this point has shown that the saturation of metals with deuterium to high concentrations could create conditions capable of sustaining cold fusion reactions. A hydride rich in deuterium will form many microcavities, which will also fill with deuterium ions. If it is assumed that the cracking in this instance is accompanied by the formation of intense electric fields such as we observe with the breakdown of ionic crystals, these fields could accelerate the deuterium ions and impart to them the energy required to escape the Coulomb barrier. However, the accelerating mechanism cannot, it would appear, be legitimately extended in this direct way to hydrides of transition metals. As we have observed in Section 2, the investigated properties of stable hydrides such as PdH, TiH_2, for example, indicate that they will characteristically have the metallic type of bond. In this instance, the answer to the fundamental question of how, after all, the surface of the crack becomes charged is unclear. In the ionic crystals the charge is generated by the rupturing of the ionic bonds accompanying the fracturing along the crystal planes, in consequence of which the two edges of a crack are going to carry opposite charges. In the case of crystals with the metallic bonds, this mechanism is absent. What is more, the "microcapacitor" which now develops has to discharge very quickly through the surrounding hydride, what with the fact that resistivity for PdH, TiH_2 etc. at room temperature does not differ all that much from the resistance of the corresponding metal. (Numerical estimates will be given below in Section 6.) Finally, the development of the strong fields (up to $E \sim 10^8$ V/cm) which can appear within the cracks should be accompanied by field emission from the metal, which would also have the effect of charging the "microcapacitor." The same time frame should see the heavier particle, the deuteron, is going to be able to build up only a small amount of energy $W_d \sim (m_e/m_D)W_e \sim$ $\sim (1-10^2)$ eV, which is not sufficient to escape the Coulomb barrier with the necesary efficiency. When it comes to the acceleration of the ions within the microcracks, the "electron wind," is evidently not a factor either. In view of the anticipated crack dimensions and the densities of the particles within the cracks, the probability of multiple collisions with H (D) atoms is not great, while the energy which can be transferred to them by the electrons is only a small fraction $(\sim m_e/m_D \sim 1/4000)\ W_e$.

So there is at first glance no basis for anticipating that the breakdown of transition-metal hydrides is going to be accompanied by the formation of electrical fields able effectively to trigger a cold fusion reaction.

In our discussion below we are going to be looking at a number of physical phenomena which nevertheless could provide an accelerating mechanism for the case of interest to us here, namely, transition metal hydrides. Unfortunately, however, we are unable to offer theoretical substantiation of the mechanisms proposed here which could be regarded as to any degree reliable and so will be limiting ourselves to the discussion of a number of supporting qualitative arguments. In Section 7 we will look at the possibility of verifying these hypotheses experimentally.

The fundamental proposition which we would like to discuss here can be formulated in the following terms: the nonequilibrium process involving the saturation of transition metals by hydrogen (D, T) generates an unstable hydride phase in which the metal is transformed into a dielectric (or semiconductor).

As we have observed in Section 2, this is an altogether phenomenon when it comes to the hydrides of the rare earth metals (see [10], for example) and can be explained by the presence on the hydrogen atoms in these hydrides of an effective negative charge generated by the diversion of part of the valence electrons associated with the atoms of the metal from the conduction band, which then "settle" on the hydrogen atoms. From this point of view, increases in the hydrogen (D, T) content of the hydride MeH_c should, on the one hand, weaken its metallic properties (electrical conductivity, for example) and enhance those associated with the ion compound $Me^+H^-_c$. A similar effect is in fact observed, for example, in the lanthanum hydride series $LaH_2 \rightarrow LaH_{2.1-3.0}$, which, like PdH and TiH_2, have a face-centered lattice. Increases in concentration from C=2 to C=3 will be accompanied by decreases in the conductivity of the hydride LaH_c by a factor of roughly 100, while the phase with the maximum concentration approaches the ionic compound $La^{+++}H^-_3$. The hydride of the La is formed with the filling of all tetrahedral vacancies, the trihydride with the filling of the octahedral to vacancies to the LaH_3 level. Exhibiting similar behavior are the hydrides of other rare-earth metals, where the first hydride MeH_2 is characterized by a metallic conductivity (metallic hydride), the more highly saturated hydride MeH_3 by semiconduction (ionic or ionic-covalent hydride).

Can we now expect to see a similar intensification of dielectric properties for the hydrides of transition metals accompanying increases in the concentration of hydrogen (D, T) in these materials? To answer this question, let us take a closer look at the coordination of the hydrogen atoms in the metal. As we observed earlier, in the structures typical of metals (fcc, hcp and bcc*), each atom of the metal is going to be associated with one octahedral and two tetrahedral vacancies and in the bcc lattice with three and six vacancies respectively. From the geometrical criterion based on considerations of solid-sphere packing stability [22], we should expect that in the case of the hydrides, an octahedral coordination would be to advantage when $0{,}41 \le R_H/R_{Me} \le 0{,}73$, a tetrahedral coordination when $0{,}22 \le R_H/R_{Me} \le 0{,}41$ [10]. For the transition metals, the ratio R_H/R_{Me} will range between 0.22 and 0.41, and the hydrogen atoms in the hcp and fcc metals should occupy the tetrahedral vacancies. This is in most instances in fact the case, although in Ni and Pd, for example, the hydrogen will occupy the octahedral vacancies, which, of course, flies in the face of what we have just said here.

Analysis of data available on the hydrides of transition metals [10] reveals a certain pattern in the change in the coordination of the hydrogen (type of interstitial site) as a function of the radius of the atoms of the metal. Small atoms will characteristically exhibit octahedral, large atoms tetrahedral coordination, the change in coordination occurring when $R_{Me} \doteq 1{,}34 \div 1{,}37$ Å.

It is important here to stress the fact that it is precisely for the stable hydrides, which may persist even after nonequilibrium saturation conditions no longer exist, that this pattern holds. When it comes to the formation of these hydrides, the dimensions of the vacancies should not be so small as to be unable to accommodate the hydrogen atoms or, on the other hand, so large as to permit the escape of the hydrogen atoms once nonequilibrium saturation conditions no longer exist.

So in light of the considerations discussed, a logical hypothesis would appear to be the following: nonequilibrium saturation may be accompanied by the formation in the surface layer through which the hydrogen (D, T) penetrate the metal of unstable phases containing elevated concentrations of H (D, T) up to the maximum level possible for this type of lattice or modification of it (e.g., $C_{max}=3$ for fcc and hcp, 3-6 for bcc). This may in turn enhance dielectric properties in the unstable phase in consequence of either the diversion of some of the electrons from the conduction band (as in the case of the hydrides of the rare-earth elements) or the expansion of the crystal lattice.

Let us look, for example, at the γ' phase of TiH_2 with the highest concentration of H. It has an fcc lattice with tetrahedral vacancies occupied by H atoms. Estimation of the dimensions of the octahedral vacancies gives

$$R_0^{TiH_2} = 0{,}5(a_{\gamma'} - 2R_{Ti}) \simeq 0{,}77\ \text{Å}\ ,$$

which greatly exceeds the dimensions of the hydrogen atom ($R^H \simeq 0.53$ Å). This means that in the hypothetical (unstable) TiH_3 phase, excess hydrogen atoms may divert some of the electrons from the conduction band, which should have the effect of enhancing the dielectric properties (electronegativity H and Ti: $\chi_H = 2{,}1$; $\chi_{Ti} = 1{,}5$).

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* face-centered cubic, hexagonal close-packed and body-centered cubic - TR

In the case, now, of palladium hydride, in which the octahedral vacancies are filled ($R_o^{PdH} = 0.5\,(a_\beta - 2R_{Pd}) \simeq 0{,}705$ Å), the increase in concentration due to penetration of remaining unoccupied tetrahedral vacancies (where $R_T \simeq R_o/2$) may result in either a restructuring or a substantial deformation (and subsequent destruction) of the lattice.

Of great interest in this connection are the findings presented in [23], the authors of which have discovered the existence of a new hydride of palladium with a chemical composition of $PdH_{1.33}$. It has been found that this hydride has a tetragonal lattice with parameters of a=2.896 Å, c=3.330 Å. The ordered arrangement of the hydrogen in the tetrahedral vacancies of the body-centered lattice combined with the presence of a large number of vacancies in the metal sublattice (25%, or $2 \cdot 10^{22}$ cm^{-3}) together constitute a remarkable characteristic of this phase. It has been possible to stabilize this phase by rapid cooling of a thin palladium film (~10^3 Å) saturated with gas-phase hydrogen at high temperature or by ion implantation.

The presence of such a large number of statistically distributed vacancies should clearly cause a sharp decrease in the electrical conductivity of this phase.

Still another factor which in the general case can diminish electrical conductivity and the associated discharge of the "microcapacitor" through the surrounding hydride involves the nature of the breakdown the surface layer of the hydride undergoes. As is generally known, this breakdown may result in the development of a "hydride sponge" and then disintegration of the surface layer. What we in fact see as a result is a loss of continuity in the material at the surface and, accordingly, a dramatic diminution of electrical contact between the individual particles into which the surface layer breaks up.

As far, now as the expansion of the crystal lattice of the hydride phase is concerned, we know that it accompanies increases roughly linearly with increases in concentration. We could anticipate that in the case of the high concentrations characteristic of the unstable phases, the deformation becomes very pronounced and ultimately breaks up the hydride layer. At the same time, as we know from the band theory of solids (see [24], for example), an increase in the distance between atoms in a crystal should result in a constriction of the allowed bands on the one hand and an expansion of the forbidden bands. This means that for the unstable phases with high concentrations of hydrogen (D), we can, in the general case, expect to see an enhancement of the dielectric properties of the material.

In concluding this section let us underscore once again the fact that what we have advanced here is not to be regarded as any rigorous proof of the hypothesis that there is an intensification, or enhancement, of dielectric properties in the unstable hydride phase of transition metals. At the same time, if we proceed on the basis of the hypothesis that the factor enabling the D nuclei to escape the Coulomb barrier is the acceleration of these nuclei within local electric fields in the microcracks, we are also going to have to introduce another, secondary (and very powerful) hypothesis that in terms of characteristics, the decomposing deuteride approaches those of an ionic crystal. Let us note here in passing that as applied to the type of fusion involved here, the term "cold" is used in only a relative sense. It reflects simply the fact that the crystal in which the reaction occurs remains at room temperature and that the incident deuterium ions carry only low (eV) levels of energy. As far as the microscopic conditions for the fusion of the accelerated deuterons are concerned, they are of course going to correspond to high effective temperatures ($\sim 10^7$).

6. The accelerating mechanism associated with crack propagation

Interpretations of the mechanoemission of electromagnetic waves accompanying breakdown traditionally work with the idea that charges, which generate a powerful magnetic field develop along the edges of the cracks.

We are now going to look at some estimates for the simplest model, one in which the crack is regarded as a plane capacitor of area ℓ^2 and gap $d = 0.1\,\ell$.

Taking account of the fact that the lattice constant for hydrides is typically going to be 3-4, let us estimate the maximum surface density of the charge:

$$\Sigma \sim e/a^2 \sim e \cdot 10^{15}\,\text{cm}^{-2} \approx e\mathcal{J}$$

where $\mathcal{J}$ is the density of the hydrogen or deuterium ions. Due to neutralization of some of the ions on the surface, the density of the charge may in many instances be substantially lower. In [25], for example, the authors give an estimate of $\mathcal{J} \sim 10^{10} \div 10^{14}\,\text{cm}^{-2}$ for adhesion layers.

The electrical field in this capacitor $E \sim \Sigma \sim 10^{8}\,\text{V/cm}$, the potential difference when $d \sim 10^{-5} \div 10^{-4}\,\text{cm}$ $U \sim Ed \sim 10^3 \div 10^4$ V. So in this capacitor an ion can develop an energy of W ~ 1-10 keV. Ion acceleration time, however, is going to be much longer than the lifetime of the charge on the capacitor. In fact, $t_{acc} \sim d\sqrt{m_{\text{д}}/W} \sim (0{,}5 \div 1{,}5)10^{-12}$ s, capacitor discharge time being a function of the resistance of the surrounding medium. If we assume that the hydride phase created by the cracking is filled with microcracks, so that the characteristic dimensions of the conducting walls are of the same order of magnitude as the micropores, and specific resistance approaches the metallic, $t_{\text{разр}} \sim \rho_{\text{мигр}} \sim 10^{-17}$ s $\approx 10^{-5}$ Ω·cm. Field emission time (emission of electrons from the metal) $t_{\text{а.э.э.}} \sim \Sigma/j_{\text{а.э.э.}} \sim 10^{-10}$ s, where $j_{\text{а.-э.}} \sim 10^{6}\,\text{а.см}^{-2}$ is the density of the field emission current. Accordingly, given the parameter values used here, field emission is evidently not going to be playing any substantial role.

So for the accelerating mechanism in cold fusion to play an effective role, it is going to be critically important to realize the conditions discussed in Section 5 (strengthen the dielectric properties of the hydride and break down the continuity of the continuity of the hydride, which substantially degrades electrical contact between the individual charged particles of the microcracks). So we will therefore be assuming that the result is going to be a substantial increase in the crack-capacitor discharge time, so that $t_{acc}<t_{dis}^{(eff)}$. Diffusion time for ions within the newly formed crack will be great as compared with t_{acc}:

$$t_{diff} \sim d^2/D$$

Now to estimate neutron yield we are going to be assuming that cold fusion is a product of the coalescence of these accelerated deuterium ions with deuterium ions along the opposite side of the crack. The effective width of this layer will be determined by the path of the deuterium ions over which a fusion reaction is still possible. Assuming a deuterium-target surface density of $\mathcal{J}$, we obtain

$$N^{(1)} \sim \mathcal{J}^2 \ell^2 \sigma(W) \tag{11}$$

for the number of neutrons from a single crack-"accelerator," where σ is the cross section of the reaction $d+d \to {}^3He + n$. Where W ~ 1-10 keV, σ increases very rapidly. For purposes of our estimates here, let us use the characteristic value $\bar{\sigma} \sim 10^{-32}\,\text{см}^2$. Assuming, now, that $\Delta V/V \sim 0{,}1$, $n_T \sim 10^{12} \div 10^{9}\,\text{см}^{-3}$, we obtain the estimated neutron yield per unit volume of hydrogen-charged material:

$$n_H \sim N^{(1)} n_T \sim (10^2 \div 10^1)\,\text{см}^{-3}$$

Let us observe here that, in a more general form, these assumptions yield the following formula:

$$n_H \sim \mathcal{J}_{\text{уск}}\, \mathcal{J}_{\text{миш}} \frac{\Delta V}{V} \frac{1}{d}\, \bar{\sigma}(W \simeq 4\pi e^2 \mathcal{J} d) \tag{12}$$

where $\mathcal{J}_{\text{уск}}, \mathcal{J}_{\text{миш}}, \mathcal{J}$ are the surface densities of the accelerated ions, the target atoms and of the charges generating the accelerating field.

This estimate can now be compared with the findings of the experiment described in [2], where in the case of a hydrogen-charged titanium electrode investigators observed a yield of 10^{-23} neutrons per pair of deuterium ions per seconds, or per atom of titanium (for the TiD_2 hydride). In the case of an electrode density of 10^{22} atoms $Ti \cdot cm^{-3}$ and a hydrogen absorption time of 10^4 s (i.e., several hours), we obtain $n_H^{(\text{эксп})} \sim 10^3$ neutrons$\cdot cm^{-3}$. A neutron yield of this order of magnitude was reported from the experiment described in [5] as well.

Here, the breakdown of an LiD crystal of 0.04 cm^3 volume was observed to yield roughly 10 neutrons; i.e., $n_H^{\text{экп}} \sim 250$ neutrons·cm^3. Let us note here that the energy released by cold fusion is going to be ~MeV·$10^3 cm^{-3} \dot{\sim} 10^{-3}$ erg·g^{-1}, which is many orders of magnitude less than the heat associated with the formation of the hydride (1-10)kcal$(gmole)^{-1} \sim 10^{10}$-$10^{11}$ erg·$(gmole)^{-1}$.

Conclusion

The assumptions and hypotheses on which the proposed model is based together with the conclusions deriving from it permit direct experimental verification. At this point let us restate our basic assumptions and hypotheses.

a) The nonequilibrium saturation of transition metals with hydrogen (D, T) may be accompanied by the formation of unstable hydride phases containing high concentrations of hydrogen (i.e., higher than concentrations in the stable phases).

b) Hydrides in these phase states may exhibit properties resembling those of dielectrics or semiconductors.

c) The expansion of the crystal lattice, the concentrated stresses and the brittleness accompanying the growth of the hydride layer ultimately result in the breakdown of the hydride and development of a network of microcracks.

d) The formation of these cracks will be accompanied by the development of electric charges. The fields thus generated can accelerate the deuterium ions and very effectively trigger a cold fusion reaction. Whence follow a number of conclusions and predictions.

1. Cold fusion is essentially a surface phenomenon. That is, it develops within the surface layer of the hydride which is breaking down during this time frame. The deeper the process penetrates the material, the greater the number of hydride layers involved.

2. Nonequilibrium deuterium saturation of the metal is a condition critical to the development of the process.

3. To create proper conditions requires the use of metals capable of dissolving deuterium (hydrogen) in large amounts. (Because of the rapid development of hydrogen storage technology, we now know of a large number of such metals and alloys.)

4. Saturation conditions should ensure a concentration gradient sufficient to generate internal concentration stresses which locally exceed the stress associated with the breakdown of the material.

5. The hydride formed cannot exhibit a high degree of plasticity.

6. Saturation time and temperature must exclude the effect of stress relaxation.

7. The stochastic nature of the hydride fracture process with a certain degree of quasi-periodic time modulation means that the cold fusion reactions are going to be similar in nature.

We are now in a position to formulate some fairly self-evident recommendations concerning experimental verification of the mechanism proposed here.

a) In view of the fact that the assumption of the existence of unstable phases containing high concentrations of hydrogen is a critical feature of the model, it is going to be important to analyze both structure and composition *in situ*, i.e., during the hydrogen absorption process itself. Specifically, this would call for the application of x-ray, neutron-diffraction and electron-microscopic techniques to analysis of the surface and subsurface layers.

b) Information concerning the intensification of dielectric properties in the unstable phase could be derived from measurements of electrical conductivity (also *in situ*). It would clearly be best to conduct such measurements with thin films similar in thickness to the characteristic critical (for fracturing) thickness ($\sim 10^3$-10^5 Å) of the hydride layer. It would also be possible to use hydrides known to have an ionic bond, or a variable bond as in the case of the rare-earth metals.

c) The hypothesis that cold fusion reactions are associated with the appearance of cracks can be checked by studying the correlation between the recordable products of cold fusion (neutrons and protons) and the acoustic emission accompanying crack propagation. It would also be possible to study the correlation with the onset of cracking using, for example, ultrasonic pulses, thermo/cryoshocks etc. [6].

d) The role of acceleration in the electric fields can be studied by correlating the products of cold fusion and the electromagnetic radiation in different wavelength ranges [26, 27].

Recommendations for verification with respect to conclusions (1-7) would be just as readily apparent.

(1) The surface nature of the phenomenon can be checked by comparing results for test material with different area-volume ratios.

(2) The critical importance of nonequilibrium as a feature of the saturation process has already been underscored in [2] and can be easily verified by varying the conditions of the experiment. Items 3 and 6 can also be easily checked by varying both the conditions and the materials employed. Let us note here that the ability of different metals and alloys to undergo hydrogen fracturing may vary greatly and permits variation of this material characteristic within broad limits.

(3) Finally, the stochasticity and quasiperiodicity [4] of cold fusion signals can be tested by analyzing the development of events over time with extremely small time frames (to values of the order of characteristic cracking times s).

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