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ENVIRONMENTAL REACTIONS AND THEIR EFFECTS ON MECHANICAL BEHAVIOR OF METALLIC MATERIALS

Technical Progress Report

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ABSTRACT

The results of research performed under DOE Contract No. DE-AC02-79ER10359.A000 during the period February 1, 1980-January 31, 1981 are presented. Results are described in five project areas: (A) implantation softening of niobium; (B) the crack path in hydrogen embrittlement of 4340 steel; (C) hydrogen embrittlement and stress corrosion cracking (report of a conference); (D) solution hardening and softening of niobium by interstitial solutes; (E) solution hardening and softening in oxide-coated niobium and niobium alloys.

CURRENT RESULTS

A. Implantation Softening of Niobium (D. DeMiglio, J. Ratka)

The goal of this project has been to examine the effects of steep near-surface solute gradients produced by ion implantation on the mechanical behavior of body-centered cubic metals at low temperatures $T \lesssim 0.15T_m$, where T_m is the melting temperature. The intent has been to compare such results with ones obtained previously$^{(1)}$ on near-surface solute gradients produced by diffusion annealing and by surface films. This project has been a cooperative one with Sandia Laboratories, Albuquerque, N. M. A first project has been completed and was first presented at the 1980 AIME Annual Meeting in Las Vegas in the Symposium on Surface Effects in Refractory Metals and Compounds.

The experimental details of the project were described in our last Technical Progress Report (C00-1676-49). Briefly, high purity oriented niobium single crystals were implanted with $^{16}O^{+}$ ions at 50 keV to produce fluences in the range $1 \times 10^{15} - 10^{17}$ O^{+} ions/cm^2. A typical profile is shown in Fig. 1. Materials mechanically tested at $77^{o}K$, in the thermal hardening regime, exhibited reduced flow stresses and increased ductilities relative to unimplanted materials. The reduced flow stress is particularly large over the first 3-6% shear strain. Figure 2 shows some of the results. The same materials exhibited the normally expected surface hardening at athermal temperatures $T \gtrsim 0.15T_m$.

In the write-up of these results, we have compared the softening behavior to our previous work on effects of solute gradients produced by diffusion annealing$^{(2)}$ and by oxide coatings$^{(3)}$. The results of the comparison are given in Fig. 3. One can see that implanted niobium is softened by amounts similar to those achieved by diffusion-annealed gradients and by oxide coatings. Only when oxide-coated materials are prestrained at temperatures $T \gtrsim 0.15\ T_m$ is the effect of oxide films much larger. We have analyzed the large initial softening by ion implantation over the first 3-6% shear strain as an effect of the surface stresses generated by the high local interstitial concentration at the surface(4). These stresses act in the same manner as tensile residual stresses: they reduce the initial flow stress but are relieved during deformation as dislocations are generated in a manner to compensate for the residual stress pattern.

A second generation project is planned in which softening due to heavy ion (non-interstitial) implantation will be examined.

B. The Crack Path in Hydrogen Embrittlement of 4340 Steel (D. DeMiglio, S. S. Wu, J. Talia)

In the last progress report, we reported in summary form the experimental aspects of an examination by transmission and scanning electron microscopy of the crack path during hydrogen-assisted fracture of 4340 steel(5). It was reported that the primary crack path observed by SEM was predominantly intergranular in high strength materials, i.e. materials quenched or quenched and tempered at temperatures $T \lesssim 300^{\circ}C$. With higher tempering temperatures, hence lower strengths, the primary crack path become predominantly transgranular. Examination of secondary cracks by TEM disclosed predominantly interlath, intercolony or intergranular paths in the higher strength materials and predominantly translath paths in the lower strength materials.

Our research in the past year has completed an entire matrix of results on crack paths in these materials and has been able to present a rationale for the results in terms of hydrogen trapping by microstructural imperfections and its effect on the crack path. The detailed analysis is presented in Appendix A, which is a preprint of our paper to appear in the proceedings of the Third International Conference on the Effect of Hydrogen on the Behavior of Materials, held at Jackson Hole, Wyoming, August 25 - 30, 1980. Briefly, the analysis shows that materials tempered at lower temperatures (with higher strengths) contain a relatively high density of irreversible traps (high binding energy, high number density, large specific saturability) and tend to exhibit primary and secondary crack paths along irreversible interfacial traps. On the other hand, materials tempered at higher temperatures (with lower strengths) contain a relatively high density of reversible traps and exhibit more random trans-boundary primary and secondary crack paths. These results are expected ones, based on complex kinetic analyses of hydrogen embrittlement(6,7).

We have now begun to do in-situ tensile experiments in the Hitachi HU-650B electron microscope to compare the above results with cracking obtained on specimens deformed in the microscope in hydrogen-containing atmospheres by use of an environmental cell. Mr. S. S. Wu and Dr. J. Talia are doing these experiments.

C. Hydrogen Embrittlement and Stress Corrosion Cracking (R. Gibala)

During the past year the principal investigator was heavily involved in the organization and implementation of a conference on Hydrogen Embrittlement and Stress Corrosion Cracking that was held in Cleveland on June 1-3, 1980 in honor of Professor Alexander R. Troiano of Case Western

Reserve University. The program for the conference is given as Table I. The effort has included serving as editor for the forthcoming proceedings of the conference and preparation of a review paper with Dr. A. J. Kumnick on Hydrogen Trapping in Iron and Steels for the proceedings. See Paper A.4 in Table I. The work involved in preparing the conference, the paper and the proceedings have been done as part of the time commitment of the principal investigator to this contract. This work has been very instrumental in permitting the detailed analysis involved in the write-up of DeMiglio's thesis.

D. Solution Hardening and Softening of Niobium by Interstitial Solutes (J. Ratka)

Investigations on the role of interstitial solutes in causing solution hardening or solution softening in high purity bcc metals continue to leave the problem unresolved. Both hardening(8) and softening(9) by single interstitial solutes have been observed by many different investigators. Our observations of solution softening in ternary niobium and tantalum alloys but not in the binary alloys of these metals have shown that solute interactions represent at least one of the fundamental causes of solution softening. However, many investigators have reported solution softening by interstitial and substitutional solutes in binary alloys believed to be of high enough purity that solute interactions, sometimes termed extrinsic softening, should be unimportant(9). These results have been analyzed in terms of a solute-induced reduction of the Peierls stress of screw dislocations in the bcc structure, sometimes termed intrinsic softening.

Recently Meshii and co-workers(11) have observed solution hardening or solution softening in Nb-O alloys depending upon the tensile orientation and operative slip system. Solution hardening is observed in center-of-the-triangle orientations (which exhibit "anomalous slip"), as our work has shown for many years(10). However, solution softening is observed for "hard" orientations, such as those along the (011)-(111) boundary, which exhibit primary slip.

As part of our final work on mechanical behavior of bcc refractory metals, we have done this same experiment and have apparently gotten results different from those of Meshii et al. In our experiments, high purity single crystals of Nb were grown by electron-beam zone melting with an orientation 11°from (011) along the (011)-(111) boundary. The crystals were centerless ground to ~2mm diameter, chemically polished, decarburized and finally vacuum degassed at 1×10^{-11} torr near the melting point for 24 hours. All specimens were subsequently exposed to high purity oxygen gas at room temperature to prevent hydrogen contamination. Nb-O alloys with 50-800 ppma oxygen were prepared by anodic oxidation at room temperature plus diffusion annealing for three hours at 1300°K at a pressure of 10^{-8} torr. Mechanical testing was done at 77°K on an Instron machine at a tensile strain rate of 3×10^{-4} s^{-1}. Several of the alloys were tested with ~20nm thermal oxide films applied or with tensile prestrains at room temperature of up to 3.5%.

Our preliminary results are given in Figure 4, along with similar data taken from stress-strain curves given in the paper by Bang, Nagakawa and Meshii(11). Although our current results are too limited to make firm conclusions, it may be noted that: (a) the flow stress of our Nb-0 materials are much smaller than those of Bang et al; (b) we do not observe the same pronounced solution softening that they have. More data are being obtained in order to establish the actual behavior of oxygen in this material. The results will be reported at the United States-Japan Seminar on Mechanical Properties of BCC Metals, to be held in Honolulu on March 23-27, 1981. The principal investigator has worked as co-organizer of this program with Professor Meshii.

E. Solution Hardening and Softening in Oxide-Coated Niobium and Niobium Alloys (R. Gibala)

During the past year we have completed research in which we have used the surface-oxide softening that was first reported in our laboratories(12) to analyze other important problems concerning the plastic flow of bcc metals. One problem reported on briefly in our last progress report and still under detailed analysis is the problem of interstitial solute-edge dislocation versus interstitial solute-screw dislocation hardening in bcc metals. Briefly, by applying oxide coatings we can examine interstitial-edge hardening, whereas the interstitial hardening in uncoated metals is by the less mobile screw dislocations. These results will be reported in detail as part of John Ratka's M.S. thesis.

A second problem is the one represented by the data in Figure 5. In these experiments, done originally by V. K. Sethi as part of his post-doctoral research, solid solution hardening and softening in the Nb-0, Nb-Mo and Nb-Mo-0 systems were examined in both uncoated and oxide-coated materials. The idea was to see if the presence of predominantly edge dislocations in the dislocation substructure (in the oxide-coated materials) alters the solution hardening and softening behavior that is observed in the uncoated binary and ternary alloys, respectively. The result was that there are only small changes in behavior of the materials when edge versus screw dislocations are the deformation carriers. These results will be presented at the U.S. - Japan Seminar discussed in section D.

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Appendix A

HYDROGEN IN IRON AND STEELS: INTERACTIONS, TRAPS AND CRACK PATHS*

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A number of defect sites interact significantly enough with hydrogen in ferrous alloys to act as traps under circumstances in which the material exhibits hydrogen-induced loss of ductility. These sites include point defects (solute atoms, vacancies), dislocations (edge and screw), interfaces and surfaces (grain boundaries, particle-matrix interfaces, cracks, external surfaces), and volume defects (voids, second phase particles). Experiments performed under controlled conditions involving careful microstructural characterization have generated quantitative estimates of number densities, binding energies and saturabilities of defects, which in turn can be used to imply probable susceptibilities to hydrogen embrittlement. Current information on hydrogen-defect interactions is reviewed and applied to results on the nature of hydrogen-assisted crack paths in a high strength steel. It is shown that both the primary and secondary crack paths can be explained in terms of the types of microstructural traps that predominate in a given material.

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Introduction

It is now well-established that hydrogen embrittlement of steels correlates closely with hydrogen-defect interaction processes (trapping) which have been examined by a number of experimental techniques (1-3). Figure 1 illustrates this correlation in an ideal manner with diffusion data as normally obtained in many investigations (4). A steel can be embrittled by hydrogen if it is stressed at temperatures high enough for diffusion of hydrogen to occur to potential embrittlement sites (T_A in Figure 1), but also low enough that the hydrogen will not be depleted in some way from embrittlement-producing traps (T_B in Figure 1). In the range of increasing temperatures from $T<T_A$ to $T>T_B$, a given material is expected to undergo a ductile-brittle-ductile transition determined largely by the nature of hydrogen transport among the populations of various hydrogen traps.

A defect trap for hydrogen is well-characterized if its binding energy E_B, its number density N_x and its specific saturability (proportional to the number of hydrogen atoms trapped per defect or defect site) are known. Research in the past decade has resulted in a significant increase in knowledge of the hydrogen trapping ability of various imperfections in steels. Hirth (3) has summarized most of the current literature. Table I gives value of E_B and N_x for a number of traps, along with a few other recent results not included in his compilations.

Examination of Table I shows that most traps can be classified into one of three categories relative to E_s, the heat of solution of hydrogen in iron, which is $\sim$ 29 kJ/mol: weak traps with $E_B < E_s$, moderate traps with $E_B \sim E_s$ and strong traps with $E_B > E_s$. Each category of trap can affect or control the mechanical behavior of the hydrogenated material. Strong traps behave irreversibly in the sense that the rate of escape of hydrogen is relatively small. Pressouyre and Bernstein (5) have argued that $E_B \gtrsim 60$ kJ/mol dictates irreversibility in hydrogen embrittlement situations. Moderate or weak traps have appreciable escape rates by comparison and act reversibly under the same experimental conditions. Strong traps with large specific saturabilities for hydrogen, e.g. incoherent precipitates and other interfaces (5), appear to have the largest detrimental effects, although the detailed kinetic behavior can be complicated and confusing when the various categories of traps are present in differing densities (5,6).

By the previous arguments, the nature of the crack path in hydrogen embrittlement should be determined by the extent to which strong traps present in adequate numbers can collect hydrogen from external sources or reversible traps. Table II summarizes this statement in a general way. Briefly, one would expect materials with relatively large densities of reversible traps to be resistant to well-defined interfacial cracking. The same result could be obtained with a high density of well-dispersed strong traps characterized by small specific saturabilities (5,6). In the present paper, we have examined the nature of the crack path in a commercial 4340 steel as a function of strength level by scanning and transmission electron microscopy. It is shown that the classification scheme given in Tables I and II can be used to explain the nature of both primary and secondary crack paths.

Experimental

The 4340 steel used in this investigation was obtained from The Timken Corporation as 38 mm diameter hot-rolled bars approximately 1 m long. The composition in wt. pct. was: C-0.40, Mn-0.73, Si-0.30, Ni-1.72, Cr-0.75, Mo-0.28, S-0.013, P-0.010. Sharp-notched stress rupture specimens were machined

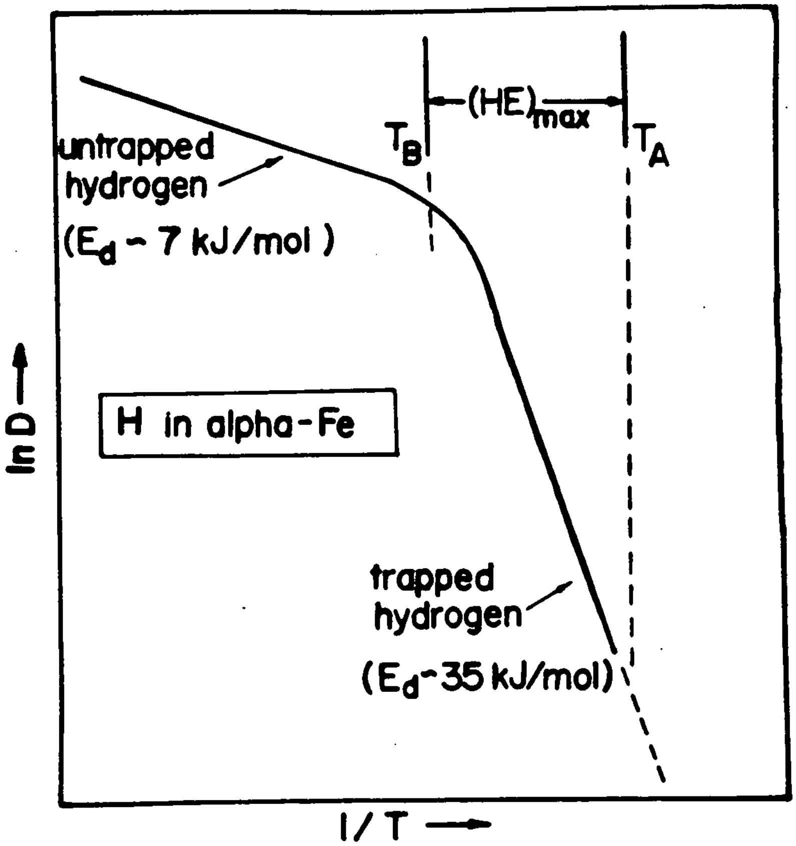

Fig. 1. Schematic illustration of the diffusivity of hydrogen in α-iron. Room temperature typically falls in the range between T_A and T_B.

Table I. Hydrogen-Trap Interactions

Trap	E_B (kJ/mol)	N_x (no/m^3)
Interstitial solutes (N,C)	∿3-15	10^{25}
Si atom	>20	10^{27}
Ti atom	26	10^{27}
Vacancy	46	$<10^{23}$
Y-vacancy	126	$\sim 10^{23}$
⊥elastic stress field	20	10^{19}-10^{26}
⊥core (screw)	20-30	10^{19}-10^{26}
⊥core (mixed)	59	10^{19}-10^{26}
½ H_2 (vapor/void)	29	--
(n boundary	∿59	10^{19}-10^{23}
Free surface	70-95	10^{21}
AlN interface	65	10^{24}-10^{25}
Fe_3C interface	84	10^{24}-10^{25}
TiC interface	96	10^{24}-10^{25}

Table II. Trap Classification for Hydrogen-Assisted Cracking

Principal Microstructural Traps	Nature of Trap	H.-A.C. Susceptibility
Solute atoms Microvoids Low angle boundaries Dislocations Coherent precipitates	Weak/moderate Reversible $E_B \leq 60$ kJ/mol	Cracking along well-defined interfaces is not favored.
Pre-existing } microcracks Propagating } High angle boundaries Incoherent { precipitates { interfaces	Strong Irreversible $E_B \gtrsim 60$ kJ/mol	Cracking along interfaces that are irreversible traps is favored.

from these bars, austenitized for 1 h at 870°C, oil-quenched to room temperature, and tempered for 1 h at various temperatures in the range 200°C-600°C or left untempered.

Hydrogen charging was done cathodically in 4% H_2SO_4 at 30 A/m^2 for 300 s with no catalytic poison and was followed immediately by cadmium plating. Specimens were baked for 0.5-3.0 h in air at 150°C to homogenize the hydrogen distribution and then tested to failure (usually) in static loading at stresses below the yield stresses of the unnotched, unhydrogenated materials. Table III gives some of the typical results.

Scanning electron microscopy (SEM) was used to characterize the primary crack paths. Secondary crack paths were determined by optical microscopy and transmission electron microscopy (TEM). TEM specimens were back-thinned from sections cut ≤ 1 mm below the fracture surface (7). Conventional bright field and dark field TEM was done on either a 125 kV Siemens 102 or 650 kV Hitachi 650B microscope. Mass thickness imaging was also done to locate fine detail of the crack paths.

Table III. Results of Hydrogen-Assisted Cracking Experiments

Tempering Treatment	Unnotched Yield Stress (MN/m^2)	Notched Tensile Stress (MN/m^2)	Baking Time at 150°C (h)	Applied Stress (MN/m^2)	Time to Fracture (h)
600°C	950	-----	1	860	no failure
500°C	1035	1930	1	860	no failure
400°C	1205	2140	1	860	0.75
°C	1380	2220	3	690	0.75
200°C	1550	2290	3	515	0.75
As-quenched	>1550	>2290	0.5	240	0.25

Results

Detailed microscopical documentation of the results on the crack path determinations in the form of SEM and TEM micrographs is presented elsewhere (8,9). Table IV summarizes the findings for the various materials, which fall into three categories according to the tempering temperatures.

Tempering Temperatures 500°C and 600°C

These materials did not undergo hydrogen-assisted fracture at stresses below the yield stress. The fracture surfaces of the uncharged and the charged materials that were deformed to fracture consisted primarily of ductile tearing as observed by SEM. The charged materials exhibited a decreased reduction in area in tensile tests and a larger dimple size than their uncharged counterparts, consistent with observations by other investigators (10). TEM observations of the secondary crack path in these materials were limited, but appeared to consist of predominately random translath cracks that often intersected cementite particles. Occasional cracking occurred along lath and lath colony boundaries, but mainly as transition paths between larger translath segments.

Table IV. Crack Path Data in 4340 Steel

Tempering Treatment	SEM Fracture Characteristics of the Uncharged Material	SEM Fracture Characteristics of the Primary Hydrogen-Assisted Crack Path	TEM Fracture Characteristics of the Secondary Hydrogen-Assisted Crack Path
600°C	Ductile tearing	Ductile tearing	Translath cracking Interlath cracking
500°C	Ductile tearing	Ductile tearing	Translath cracking Interlath cracking
400°C	Ductile tearing Transgranular cleavage	Ductile tearing Transgranular cleavage	Translath cracking Interlath cracking
320°C	Ductile tearing Transgranular cleavage	Ductile tearing Transgranular cleavage	----
200°C	Ductile tearing Transgranular cleavage	Cracking along prior γ gb's Transgranular cleavage	Translath cracking Interlath cracking Intercolony cracking Cracking along prior γ gb's
quenched	Ductile tearing Transgranular cleavage Cracking along prior γ gb's	Cracking along prior γ gb's Ductile tearing	Translath cracking Interlath cracking Intercolony cracking Cracking along prior γ gb's

Te ring Temperatures 320°C and 400°

Table III illustrates that these materials exhibited delayed hydrogen-induced failure at stresses below the yield stress. The primary crack path observed by SEM is transgranular such that the fracture surface consists of approximately equal amounts of ductile tearing and transgranular cleavage in the hydrogen-charged material, similar to results by Beachem (11). Both fracture modes exist in the uncharged material, but the proportion of transgranular cleavage is smaller. The secondary crack path observed by TEM is predominately translath, with some tendency toward fracture on {110} translath slip planes. Some interlath fracture is also observed, more so than in materials tempered at 500°C and 600°C.

As-quenched Steels and Steels Tempered at 200°C

These materials, which failed at stresses well below the yield stress, exhibited primary crack paths that were more than 50% intergranular fracture along prior austenite grain boundaries, similar to other observations (11-13). Transgranular cleavage constituted another 20-30%, with the balance, often 10% or less, being ductile tearing. The secondary crack paths were varied in these materials. Translath, interlath and intercolony cracks were observed in most foils, with higher percentages of the latter two modes than in materials tempered at higher temperatures. In addition, secondary intergranular cracks along prior austenite grain boundaries were often observed. The general trend was toward more interfacially-oriented primary and secondary crack paths in the stronger materials. It should be noted that the as-quenched material was mildly tempered at 150°C as part of the hydrogen baking treatment. Thus the similarity in results of these two materials is expected.

Discussion

It has become widely accepted in recent years that manipulation and control of microstructure is an effective method of obtaining resistance to hydrogen-assisted cracking in steels (14-16). In view of the nature of hydrogen trapping by many types of microstructural imperfections given in Tables I and II, it is clear that the effect of microstructure on hydrogen embrittlement is an effect of hydrogen trapping on the nature of the hydrogen-assisted crack path. The following discussion is intended to illustrate that the nature of microstructural traps in 4340 steel as a function tempering temperature is consistent with the trap classification scheme given in Table II and the crack path expected from the classification.

Table V lists the major microstructural traps for hydrogen in 4340 steel and attempts to identify which traps are reversible or irreversible in nature. The list and categorization are not intended to be complete or absolute, but they do indicate a simple generalization: materials tempered at the higher temperatures tend to have an abundance of reversible types of traps, whereas materials tempered at the lower temperatures tend to have an abundance of irreversible traps. This observation, plus the recognition in Table II that prediction of the hydrogen-assisted crack path relies mainly dentification of the irreversible trapping sites, allows one to under- d the results on the crack path determinations given in Table IV and summarized schematically in Figure 2. Figure 2 illustrates the crack paths only for the materials examined extensively: those tempered at 320°C or 400°C (in(i)) and the as-quenched materials or those tempered at 200°C (in(ii)). The crack path in materials tempered at 500°C and 600°C resembles that in (i), but with relatively little inter-boundary cracking.

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Table V. Nature of Microstructural Traps for Hydrogen in 4340 Steel

Tempering Treatment	Description of the Microstructure	Reversibile Traps for Hydrogen	Irreversible Traps for Hydrogen
600°C 500°C	Spheroidized, incoherent Fe_3C at low angle lath bdies Dislocation recovery	Low angle lath bdies and lath colony bdies Dislocations Solute atoms Incoherent and coherent Fe_3C	Microcracks Incoherent Fe_3C
400°C	Early-stage Fe_3C spheroidization, dislocation recovery, and high-to-low angle lath bdy transformation	↓	↓
320°C	Very little dislocation recovery Coherent Fe_3C at high angle lath bdies	↓	Microcracks High angle lath bdies and lath colony bdies Prior γ gb's
200°C	No significant dislocation recovery Coherent ε-carbide at high angle lath bdies	Dislocations Solute atoms ε-carbide	↓
As-quenched	Precipitate-free high angle lath bdies Carbon in solution in α'	Dislocations Solute atoms	↓

The prediction of the crack path in the classification scheme in Table II relies on the fact that all irreversible traps bind hydrogen effectively by virtue of the large E_B and N_x and that most irreversible traps are also characterized by large specific saturabilities. A large saturability in general can permit accumulation of hydrogen to a critical concentration necessary to initiate a hydrogen-assisted crack (17). The classification scheme in Table II works because most of the irreversible traps are either extended interfaces of various types or second phase particles that precipitate at such interfaces. The yttrium-vacancy point defect trap with $E_B \sim 126$ kJ/mol and $N_x \sim 10^{23}$ per m^3 reported by Myers et al. (18) is a probable example of irreversible trap with a low specific saturability. So are other point defect traps reported to have large E_B's (6) and small, finely dispersed incoherent precipitates.

The presence of reversible traps can greatly complicate the sensitivity to hydrogen embrittlement (6). Reversible interfacial traps, which predominate in the materials tempered at 500°C and 600°C, do not trap hydrogen for

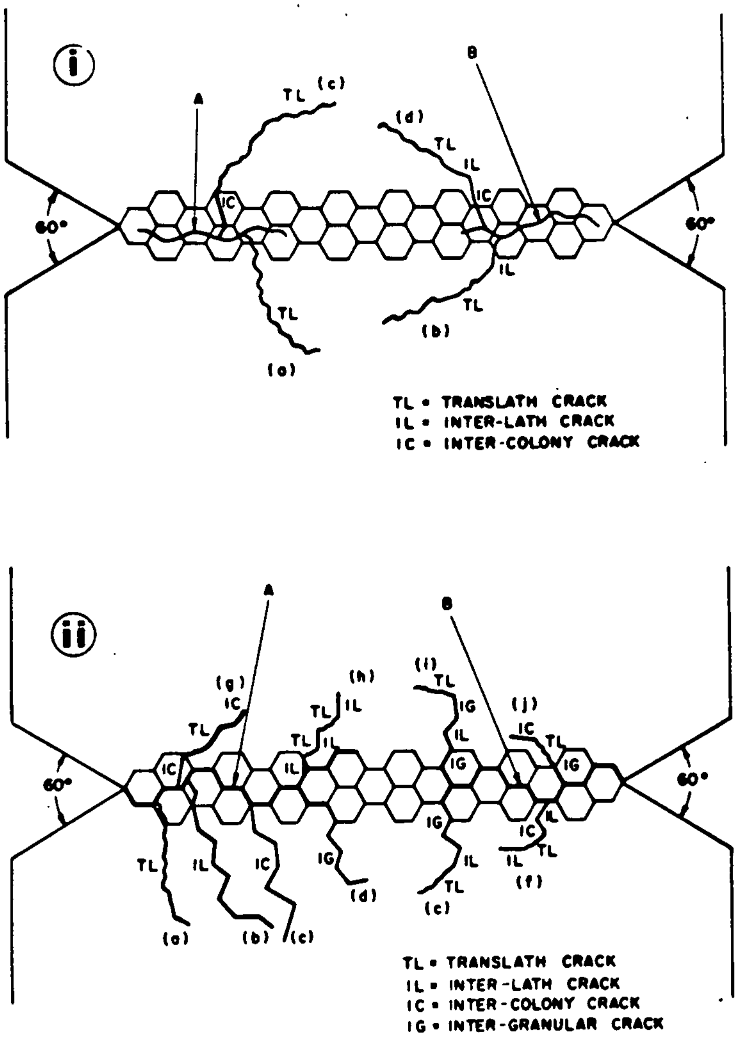

Fig. 2. Schematic illustration of hydrogen-assisted primary (A,B) and secondary (a,b,......) cracking in sharp-notched 4340 steel as observed by SEM and TEM respectively. (i) Summary of results for materials tempered at 400°C and 320°C. (ii) Summary of results for materials tempered at 200°C and as-quenched materials.

long enough times to allow a large hydrogen accumulation. Thus hydrogen-assisted cracking along reversible interfacial traps is not expected. However, these and other reversible traps can help deter irreversible interfaces from trapping adequate hydrogen to become crack initiation sites. On the other hand, mobile reversible traps (dislocations) might actually transport hydrogen to pre-existing or propagating cracks or other irreversible microstructural traps (19) and initiate and accelerate hydrogen cracking at those sites. The same argument applies to the materials tempered at 400°C, except that incomplete recovery and carbide spheroidization leaves these materials with a smaller relative density of reversible traps and a correspondingly higher remnant irreversible trap population in the incompletely recovered tensite substructure. None of the materials tempered at 400°C and above contain a high enough density of major irreversible traps with high specific supersaturabilities (apart from pre-existing or propagating cracks) to promote a large fraction of interfacially-aligned fracture paths.

The untempered materials or ones tempered at relatively low temperatures contain a large density of irreversible interfacial traps in the form of high angle austenite grain boundaries and unrecovered lath and lath colony boundaries. The presence in the microstructure of these major irreversible traps with relatively high supersaturabilities results in predominately interfacial hydrogen-assisted primary and secondary cracking. Evidently, the presence of coherent ε or Fe_3C as immobile reversible traps is not adequate to prevent either interfacial primary or secondary cracking at irreversible traps.

It might also be argued that the nature of the primary and secondary hydrogen-assisted crack path correlates with the strength of the material. Certainly the primary crack path undergoes a transition from intergranular to transgranular cracking with decreasing strength, at the same time that secondary cracking path undergoes a transition from predominately interfacial to random trans-boundary cracking. However, in the same manner that it has been shown that susceptibility to hydrogen embrittlement does not correlate with strength, but is more directly an effect of microstructure (15,16), we would argue that the correlation of crack path with strength is an effect of microstructure in the manner outlined in Tables IV and V.

Summary and Conclusions

The nature of the hydrogen-assisted crack path in quenched and tempered 4340 steels can be correlated with the trapping ability of the major microstructural features of the steels. Materials tempered at lower temperatures contain a relatively high density of irreversible traps (high binding energy, high number density, large specific saturability) and tend to exhibit primary and secondary crack paths along irreversible interfacial traps. Materials tempered at higher temperatures contain a relatively high density of reversible traps and exhibit more random trans-boundary primary and secondary crack paths.

Acknowledgements

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Gibala/DeMiglio

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TABLE I
Program for

HYDROGEN EMBRITTLEMENT AND STRESS CORROSION CRACKING

(Alexander R. Troiano Honorary Symposium)

June 2-3, 1980
Case Western Reserve University
Cleveland, Ohio

PROGRAM

A. Hydrogen Embrittlement

1. Dr. H.H. Johnson, Cornell University
   Overview on Hydrogen Degradation Phenomena
2. Dr. J.P. Hirth, The Ohio State University
   Theories of Hydrogen-Induced Cracking
3. Dr. R.A. Oriani, United States Steel Corporation
   Decohesion Theory of Hydrogen Embrittlement
4. Drs. A.J. Kumnick and R. Gibala, Case Western Reserve University
   Hydrogen Trapping in Iron and Steels
5. Dr. P.G. Lacombe, Université de Paris - Sud
   Investigation of Hydrogen Trapping in Steels by Tritium Autoradiography
6. Dr. R.P. Wei, Lehigh University
   Fracture Mechanics Investigations of Hydrogen Embrittlement
7. Drs. I.M. Bernstein and A.W. Thompson, Carnegie-Mellon University
   Hydrogen Degradation in Lower Strength Materials
8. Dr. H.K. Birnbaum, University of Illinois
   Hydrogen Embrittlement in Hydride-Forming Materials

B. Stress Corrosion Cracking

1. Dr. R.W. Staehle, University of Minnesota
   Overview on Stress Corrosion Cracking Phenomena
2. Dr. E.N. Pugh, National Bureau of Standards
   Environmental Failure Mechanisms
3. Dr. Z. Szklarska-Smialowska, Polish Academy of Sciences
   Films and their Importance in Stress Corrosion Cracking
4. Drs. R.F. Hehemann and A.R. Troiano, Case Western Reserve University
   The Role of Hydrogen in Stress Corrosion Cracking of Austenitic Stainless Steels.
5. Dr. D.J. Duquette, Rensselaer Polytechnic Institute
   Corrosion Fatigue
6. Dr. M.O. Speidel, Brown Boveri and Company
   Stress Corrosion Cracking in Aluminum Alloys
7. Dr. R.M. Latanision, Massachusetts Institute of Technology
   Hydrogen Embrittlement and Stress Corrosion Cracking in Amorphous Metallic Materials
8. Dr. P.R. Rhodes, Shell Development Company
   Hydrogen Embrittlement and Stress Corrosion Cracking in Industrial Circumstances

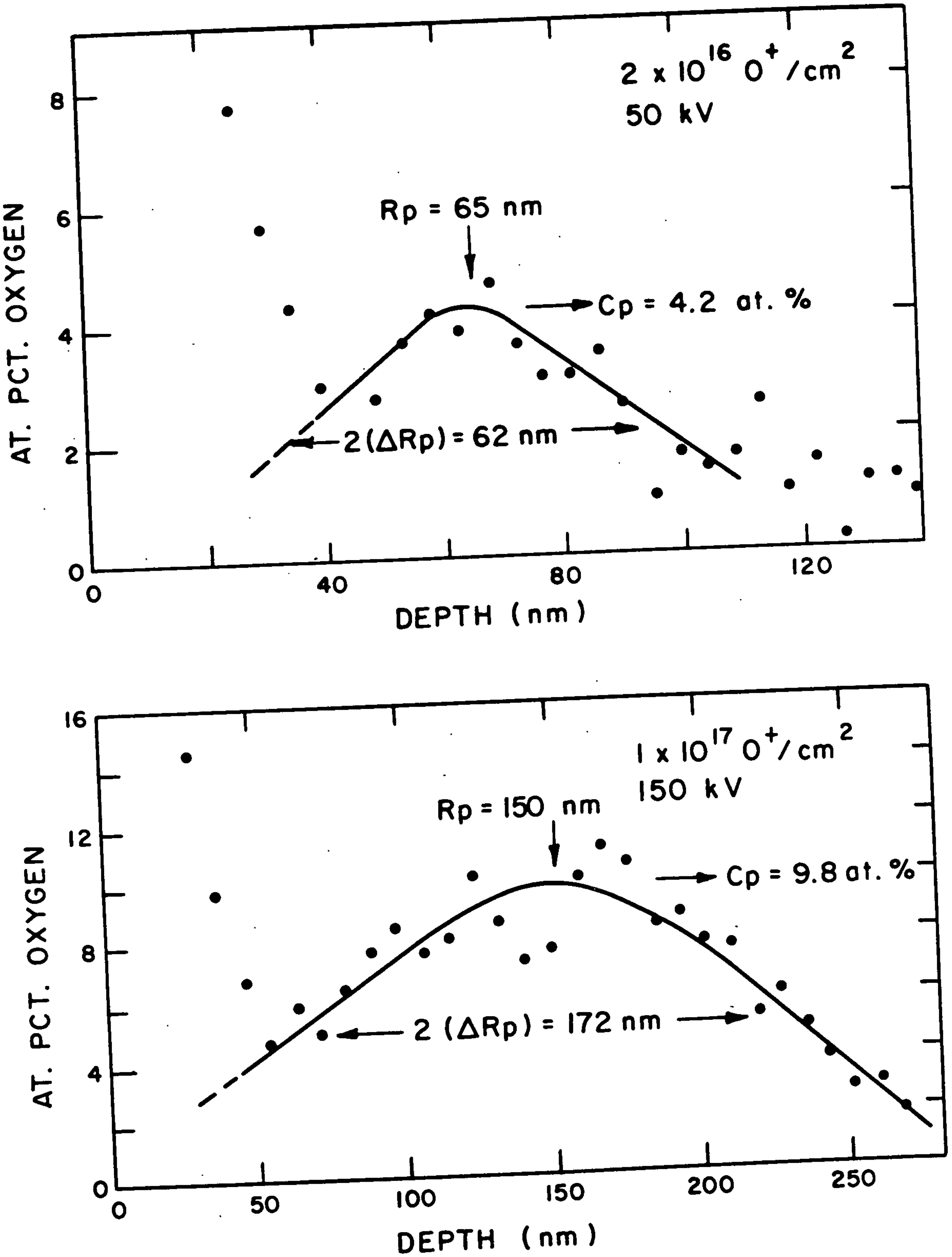

FIGURE 1. DEPTH PROFILES FOR OXYGEN IONS IMPLANTED INTO SINGLE CRYSTALS OF NIOBIUM.

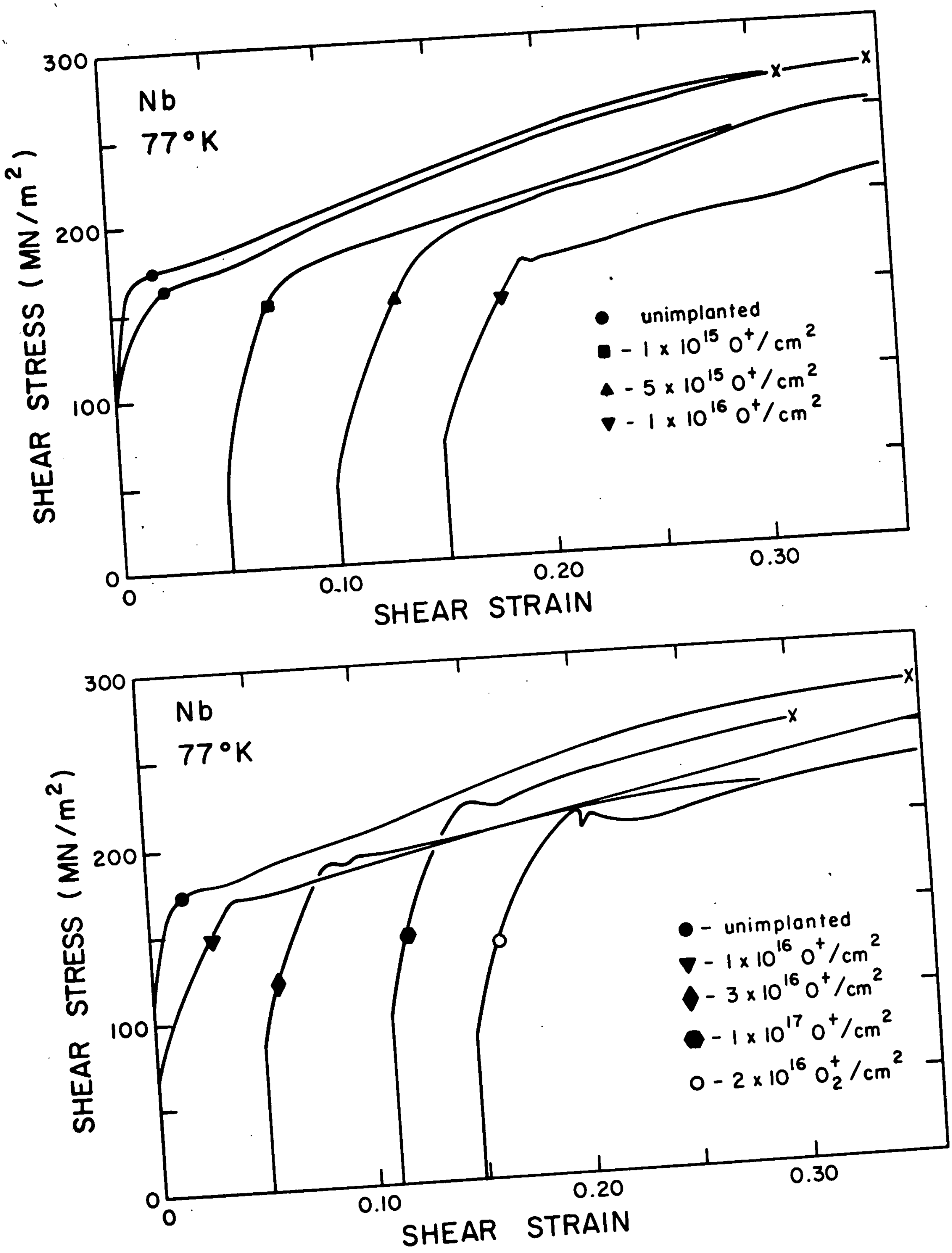

FIGURE 2. RESOLVED SHEAR STRESS-SHEAR STRAIN CURVES FOR [213] ORIENTED SINGLE CRYSTALS OF NIOBIUM IMPLANTED WITH 50kV OXYGEN IONS TO THE FLUENCES GIVEN.

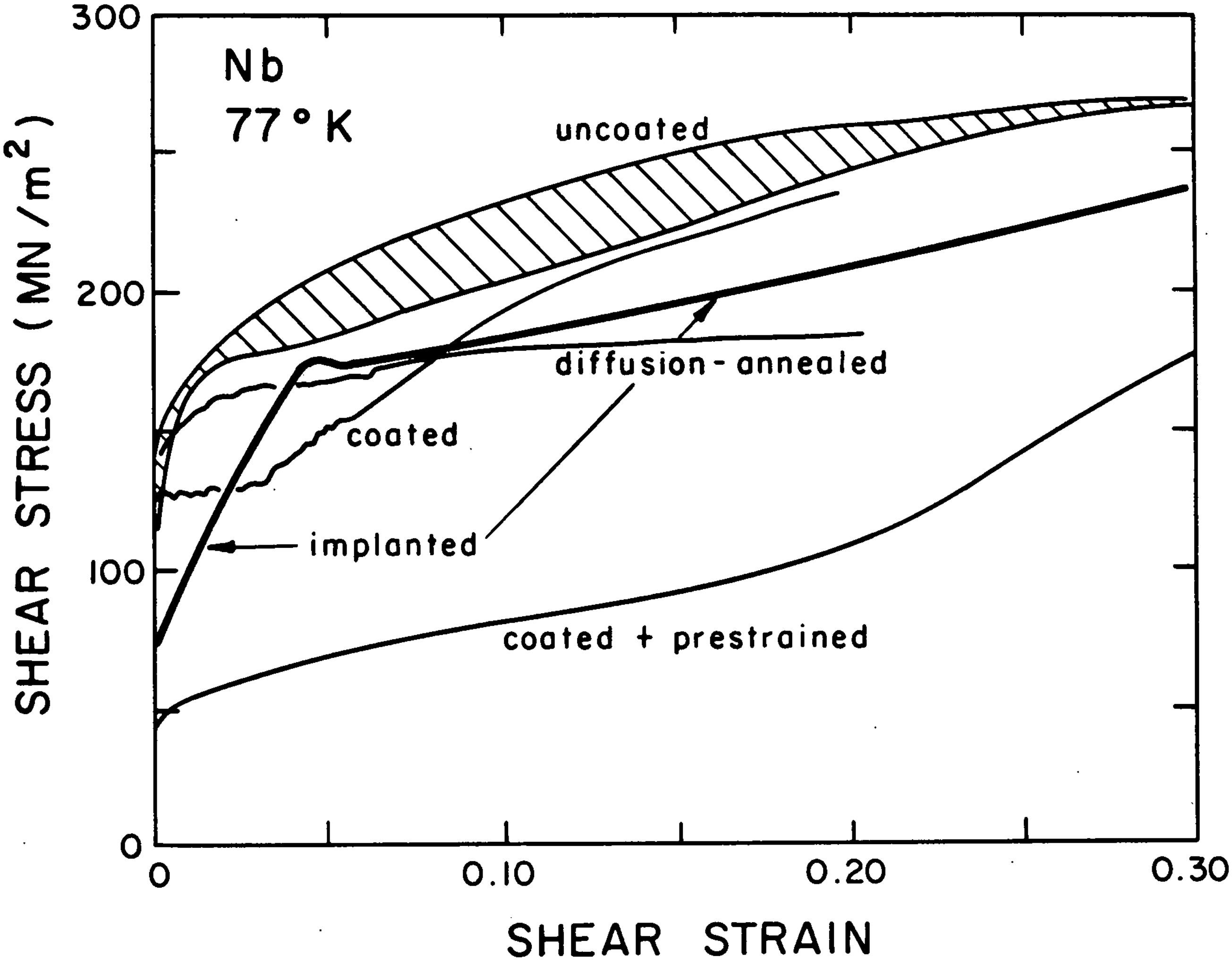

FIGURE 3. A COMPARISON OF SURFACE SOFTENING OF NIOBIUM SINGLE CRYSTALS WITH A [213] ORIENTATION BY OXYGEN ION IMPLANTATION, BY SURFACE OXIDE FILMS, AND BY DIFFUSION-ANNEALED SURFACE OXIDE FILMS. EACH CURVE REPRESENTS THE MAXIMUM SOFTENING ACHIEVED FOR EACH SURFACE TREATMENT.

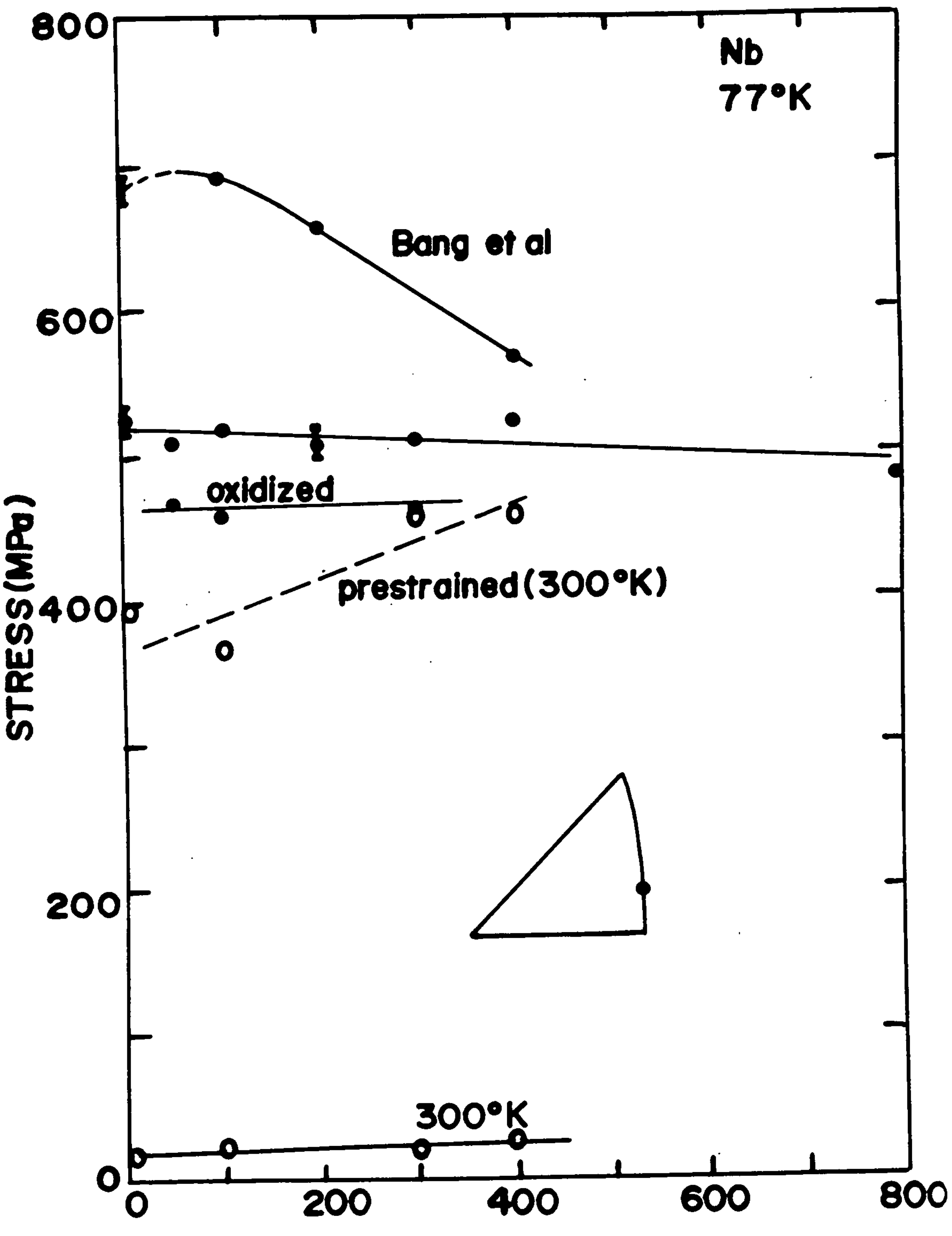

FIGURE 4. EFFECT OF OXYGEN ON THE FLOW STRESS AT 2% PLASTIC STRAIN OF NIOBIUM SINGLE CRYSTALS OF THE ORIENTATION SHOWN. TRUE STRAIN RATE $\dot{\epsilon} \approx 3 \times 10^{-4}\ S^{-1}$.

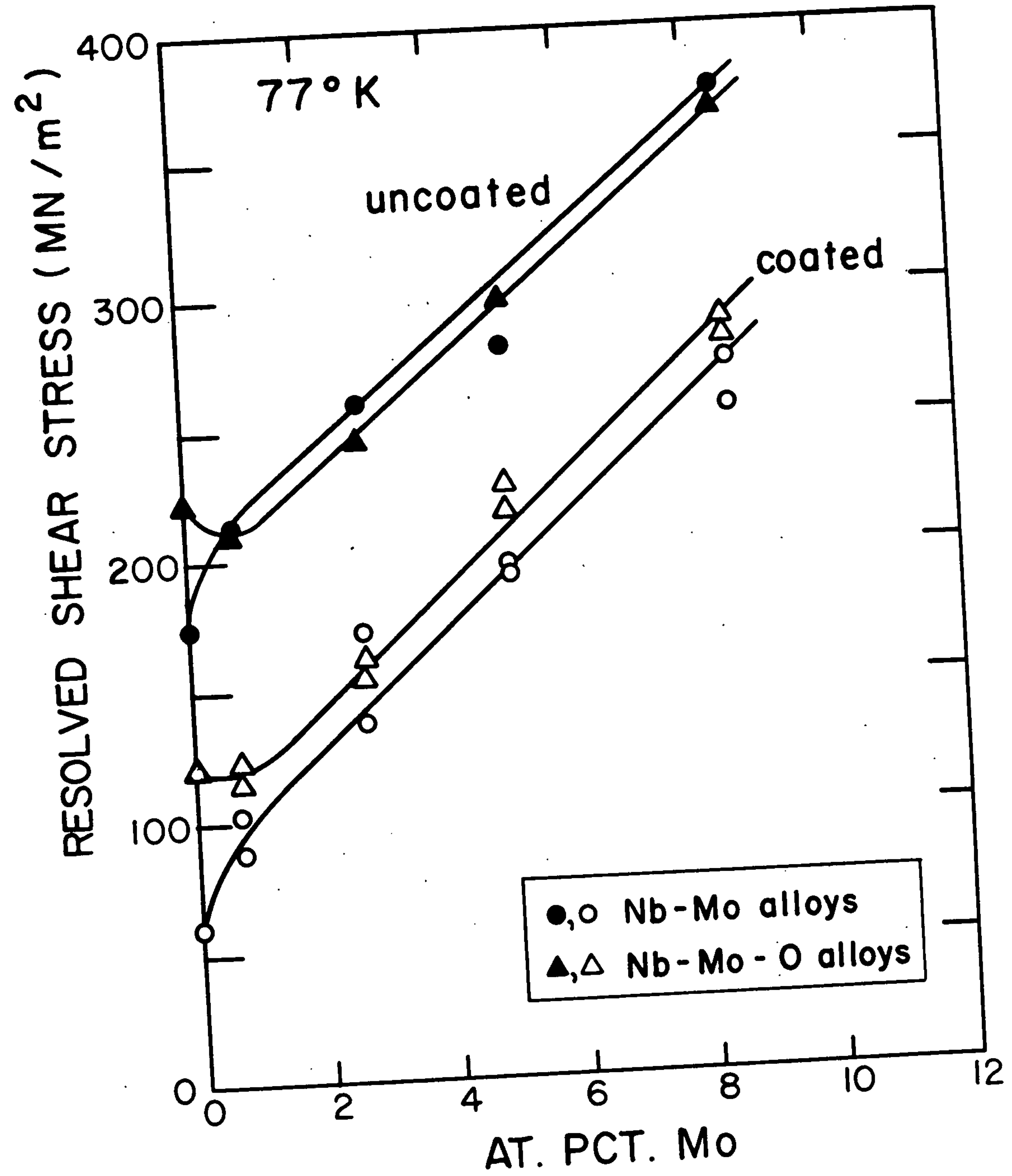

FIGURE 5. SOLID SOLUTION HARDENING AND SOFTENING OBSERVED IN Nb-O, Nb-Mo, AND Nb-Mo-O SINGLE CRYSTALS. THE OXYGEN CONCENTRATION IS 50 ppma IN THE Nb-O AND Nb-Mo-O ALLOYS.

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A SUMMARY OF REPORTS SUBMITTED TO THE UNITED STATES ATOMIC ENERGY COMMISSION

Report No.	
C00-1676-1	R.Gibala, F. Povolo, M.G. Ulitchny, and K.V. Ravi, Technical Progress Report, Dislocation-Solute Atom Interactions in Alloys, October 31, 1967.
C00-1676-2	R. Gibala, Comments on the Cold-Work Peak in Alpha Iron, October 31, 1967.
C00-1676-3	R. Gibala and C.A. Wert, Interstitial Alloys of BCC Metals, October 31, 1967.
C00-1676-4	R. Gibala, F. Povolo, M.G. Ulitchny, and K.V. Ravi Technical Progress Report, Dislocation-Solute Atom Interactions in Alloys, October 31, 1968.
C00-1676-5	F. Povolo, Comment on the Influence of Hydrogen Content on the Relaxation Spectrum of α Titanium, November 7, 1968.
C00-1676-6	F. Povolo and R. Gibala, A Marx Three Component Oscillator for Internal Friction Measurements at Low and High Temperatures in High Vacuum, January 8, 1969.
C00-1676-7	K.V. Ravi and R. Gibala, The Strength and Alloy Softening of BCC Metals, May 30, 1969.
C00-1676-8	R. Gibala, F. Povolo, M.G. Ulitchny, K.V. Ravi, and G.J. Klems, Technical Progress Report, Dislocation-Solute Atom Interactions in Alloys, October 31, 1969.
C00-1676-9	R. Gibala, Hydrogen-Dislocation Interaction in Iron, December 24, 1969.
C00-1676-10	K.V. Ravi and R. Gibala, The Strength of Niobium-Oxygen Solid Solutions, December 20, 1969.
C00-1676-11	K.V. Ravi and R. Gibala, Low Temperature Strengthening in Niobium-Hydrogen Single Crystals, May 27, 1970.
C00-1676-12	K.V. Ravi and R. Gibala, The Strength Differential Effect in Niobium-Oxygen Solid Solutions, September 9, 1970.

C00-1676-13	R. Gibala, M.G. Ulitchny, G.J. Klems, A.K. Vasudevan, and A.A. Sagues, Technical Progress Report, Dislocation-Solute Atom Interactions in Alloys, October 31, 1970.
C00-1676-14	K.V. Ravi and R. Gibala, Dislocation Dynamics in Niobium-Oxygen Alloys, May 18, 1971.
C00-1676-15	M.G. Ulitchny, A.A. Sagues and R. Gibala, Alloy Softening in Niobium- and Tantalum-base Solid Solutions, June 10, 1971.
C00-1676-16	A.A. Sagues and R. Gibala, Substitutional-Interstitial Solute Interactions in Ta-Re-N Solid Solutions, July 14, 1971.
C00-1676-17	R. Gibala, A Report on the Discussion Meeting on Defects in Refractory Metals, at Mol, Belgium on September 20-22, 1971, October 31, 1971.
C00-1676-18	R. Gibala, M.G. Ulitchny, G.J. Klems, A.A. Sagues and A.K. Vasudevan, Technical Progress Report, Dislocation-Solute Atom Interactions in Alloys, October 31, 1971.
C00-1676-19	M.G. Ulitchny and R. Gibala, Solute Interactions Involving Hydrogen and Their Influence on Mechanical Behavior of Niobium Single Crystals, December 27, 1971.
C00-1676-20	M.G. Ulitchny and R. Gibala, Internal Friction and Strain Aging of Austenite, April 10, 1972.
C00-1676-21	F. Povolo and R. Gibala, An Analysis of Amplitude Dependent Internal Friction and Modulus Defect of Materials Subjected to Inhomogeneous Stresses, June 19, 1972.
C00-1676-22	M.G. Ulitchny and R. Gibala, Mechanical Properties of Interstitial Alloys of Niobium, June 19, 1972.
C00-1676-23	R. Gibala, A.K. Vasudevan, J.A. Slane, R.P. Krupitzer, Technical Progress Report, Dislocation-Solute Atom Interactions in Alloys, October 31, 1972.
C00-1676-24	R. Gibala, A Report on the International Congress on Hydrogen in Metals, at Paris, France on May 29-June 2, 1972, December 31, 1972.
C00-1676-25	M.G. Ulitchny and R. Gibala, The Effects of Interstitial Solute Additions on the Mechanical Properties of Niobium and Tantalum Single Crystals, February 15, 1973.
00-1676-26	R. Gibala, Hydrogen-Defect Interactions in Iron-Base Alloys, March 1, 1973.

C00-1676-27	M.G. Ulitchny, A.K. Vasudevan and R. Gibala, Solution Hardening and Softening in Niobium-Base Solid Solutions, April 1, 1973.
C00-1676-28	A.A. Sagues and R. Gibala, Crystallographic Orientation Dependence of Interstitial Anelasticity in Ta-Re-N and Ta-Re-O Alloys, October 31, 1973.
C00-1676-29	G.J. Klems, R.E. Miner, F.A. Hultgren and R. Gibala, Internal Friction of Ferrous Martensites, October 31, 1973.
C00-1676-30	R. Gibala, A Report on the Third International Conference on the Strength of Metals and Alloys and on the Fifth International Conference on Internal Friction and Ultrasonic Attenuation in Crystalline Solids, October 31, 1973.
C00-1676-31	R. Gibala, A.K. Vasudevan, V.K. Sethi, J.A. Slane and R.P. Krupitzer, Technical Progress Report, Dislocation-Solute Atom Interactions in Alloys, October 31, 1973.
C00-1676-32	R. Gibala and T.E. Mitchell, Solid Solution Hardening and Softening, October 31, 1973.
C00-1676-33	A.A. Sagues and R. Gibala, Substitutional-Interstitial Interactions in Ta-Re-N and Ta-Re-O Alloys, September 3, 1974.
C00-1676-34	R. Gibala, V.K. Sethi, and A.A. Sagues, Technical Progress Report, Dislocation-Solute Atom Interactions in Alloys, October 31, 1974.
C00-1676-35	V.K. Sethi and R. Gibala, Surface Oxide Softening in Nb and Ta Single Crystals, March 26, 1975.
C00-1676-36	G.J. Klems, R.E. Miner, F.A. Hultgren and R. Gibala, Internal Friction in Ferrous Martensites, September 24, 1975.
C00-1676-37	A.A. Sagues, M.G. Ulitchny and R. Gibala, Hydrogen Strengthening in Niobium and Niobium Base Alloys, October 31, 1975.
C00-1676-38	V.K. Sethi and R. Gibala, The Effect of Anodic Oxide Coatings on the Low Temperature Mechanical Behavior of Niobium Single Crystals, October 31, 1975.
C00-1676-39	R. Gibala, V.K. Sethi and R. Fournier, Technical Progress Report, Dislocation-Solute Atom Interactions in Alloys, October 31, 1975.

COO-1676-40	V.K. Sethi and R. Gibala, The Effect of Anodic Oxide Coatings on the Mechanical Behavior of Niobium and Tantalum Single Crystals, February 4, 1976.
COO-1676-41	V.K. Sethi and R. Gibala, Surface Oxide Softening of Niobium Single Crystals, February 10, 1976.
COO-1676-42	V.K. Sethi and R. Gibala, Effect of Oxygen Distribution on the Low Temperature Mechanical Behavior of Niobium Single Crystals, October 31, 1976.
COO-1676-43	V.K. Sethi and R. Gibala, Surface Oxide Softening in Single Crystals of Niobium and Tantalum, October 31, 1976.
COO-1676-44	R. Gibala, V.K. Sethi and R. Fournier, Technical Progress Report, Obstacle-Induced Hardening and Softening in Metallic Materials, October 31, 1976.
COO-1676-45	R. Fournier and R. Gibala, Effect of Purity, Prestrain and Cooling Rate on Hydrogen Strengthening in Niobium Single Crystals, July 15, 1977
COO-1676-46	V.K. Sethi and R. Gibala, The Strength-Differential Effect in Surface Oxide Softening of Niobium Single Crystals, July 15, 1977.
COO-1676-47	R. Gibala, V.K. Sethi, R. Fournier and I. Rusakova, Technical Progress Report, Environmental Reactions and Their Effects on Mechanical Behavior of Metallic Materials, September 30, 1977
COO-1676-48	R. Gibala, R. Fournier and D. DeMiglio, Technical Progress Report, Environmental Reactions and Their Effects on Mechanical Behavior of Metallic Materials, October 31, 1978.
COO-1676-49	R. Gibala, R. Fournier, D. DeMiglio and J. Ratka, Technical Progress Report, Environmental Reactions and Their Effects on Mechanical Behavior of Metallic Materials, October 31, 1979.
COO-1676-50	R. Gibala, J. Ratka, J. Talia, S. Wu, Technical Progress Report, Environmental Reactions and Their Effects on Mechanical Behavior of Metallic Materials, October 31, 1980.

DOE Contract No. DE-AC02-79ER10359.A000

ORAL PRESENTATIONS

1. R. Gibala, "Internal Friction of BCC Metals", Purdue University, April 1968.
2. K.V. Ravi and R. Gibala, "The Strength of Niobium-Oxygen Solid Solutions", AIME Meeting, Pittsburgh, May 1969.
3. M.G. Ulitchny and R. Gibala, "Internal Friction and Strain Aging of Austenite", AIME Meeting, Philadelphia, October 1969.
4. K.V. Ravi and R. Gibala, "Thermally Activated Deformation of Niobium and Niobium Base Interstitial Solid Solutions", AIME Meeting, Philadelphia, October 1969.
5. R. Gibala, "Strength and Structure of Solid Solutions", Cornell University, February 1970.
6. K.V. Ravi and R. Gibala, "The Effects of Aging on the Mechanical Behavior of Niobium-Oxygen Solid Solutions", AIME Meeting, Los Vegas, May 1970.
7. R. Gibala, "Strength and Structure of Solid Solutions", Argonne National Laboratory, August 1970.
8. K.V. Ravi and R. Gibala, "The Strength Differential Effect in Niobium-Oxygen Solid Solutions", Second International Conference on the Strength of Metals and Alloys, Pacific Grove, California, August 1970.
9. F. Povolo and R. Gibala, "Low Temperature Dislocation Damping in BCC Metals", AIME Meeting, Cleveland, Ohio, October 1970.
10. R. Gibala, "Interstitial Solid Solutions", Case Western Reserve University, February, 1971.
11. R. Gibala, "The Effect of Solute and Point Defect Interactions on Mechanical Properties," AIME Meeting, Atlanta, May 1971.
12. K.V. Ravi and R. Gibala, "The Influence of Interstitials on the Dislocation Velocity Exponent in Niobium-Oxygen Solid Solutions", AIME Meeting, Atlanta, May 1971.
13. M.G. Ulitchny, A.A. Sagues and R. Gibala, "Alloy Softening in Niobium- and Tantalum-base Solid Solutions", Discussion Meeting on Defects in Refractory Metals, Mol, Belgium, September 1971.
14. R. Gibala, "Alloy Softening in Body-Centered Cubic Materials", Cleveland Section, AIME Meeting, October 1971.

15. T.E. Mitchell and R. Gibala, "Solution Softening in BCC Metals: The Importance of Solute Interactions", AIME Meeting, Detroit, October 1971.

16. A.K. Vasudevan and R. Gibala, "Alloy Softening in Nb-Mo-O Single Crystals", AIME Meeting, Detroit, October 1971.

17. M.G. Ulitchny and R. Gibala, "Mechanical Behavior of Niobium-base Interstitial Solid Solutions", AIME Meeting, Detroit, October 1971.

18. C.J. Klems, R.E. Miner, F.A. Hultgren and R. Gibala, "Internal Friction in Twinned Fe-Ni-C Martensites", AIME Meeting, Detroit, October 1971.

19. R. Gibala, "Point Defects in Refractory Metals", Case Western Reserve University, October 1971.

20. M.G. Ulitchny and R. Gibala, "Internal Friction and Strain Aging Behavior of Ferrous Austenites", AIME Meeting, Boston, May 1972.

21. M.G. Ulitchny and R. Gibala, "Thermally Activated Deformation Behavior of Nb-O-N Single Crystals", AIME Meeting, Boston, May 1972.

22. R.E. Miner, F. Hultgren and R. Gibala, "An Analysis of the Cold-Work Internal Friction Peak in Iron", AIME Meeting, Boston, May 1972.

23. M.G. Ulitchny and R. Gibala, "Solute Interactions Involving Hydrogen and Their Influence on the Mechanical Behavior of Niobium Single Crystals", International Congress on Hydrogen in Metals, Paris, France, May 1972.

24. R. Gibala,"The Strength Differential Effect in Niobium-Oxygen Solid Solutions", ONR Workshop on the Strength Differential Effect in Metals and Polymers, Boston, June 1972.

25. R. Gibala, "The Strength and Structure of Solid Solutions", General Electric Research and Development Laboratory, Schenectady, N.Y., November 1972.

26. R. Gibala, "The Strength and Structure of Solid Solutions", University of Illinois, Urbana, November 1972.

27. A.K. Vasudevan and R. Gibala, "Alloy Softening in Nb-Mo-O and Nb-Mo-N Solid Solutions", AIME Meeting, Philadelphis, May, 1973.

28. A.A. Sagues and R. Gibala, "Substitutional-Interstitial Solute Interactions in Ta-Re-N and Ta-Re-O Solid Solutions", AIME Meeting, Philadelphia, May, 1973.

29. R. Gibala, "Hydrogen-Defect Interactions in Iron-Base Alloys", International Conference on Stress Corrosion Cracking and Hydrogen Embrittlement of Iron Base Alloys, Unieux-Firminy, France, June, 1973.

30. M. G. Ulitchny, A.K. Vasudevan and R. Gibala, "Solution Hardening and Softening in Nb-Base Solid Solutions", Third International Conference on the Strength of Metals and Alloys, Cambridge, England, August, 1973.

31. A.A. Sagues and R. Gibala, "Crystallographic Orientation Dependence of Interstitial Anelasticity in Ta-Re-N and Ta-Re-O Alloys", Fifth International Conference on Internal Friction and Ultrasonic Attenuation in Crystalline Solids, Aachen, Germany, August, 1973.

32. G.J. Klems, R.E. Miner, F.A. Hultgren and R. Gibala, "Internal Friction in Ferrous Martensites", Fifth International Conference on Internal Friction and Ultrasonic Attenuation in Crystalline Solids, Aachen, Germany, August, 1973.

33. M.G. Ulitchny and R. Gibala, "Mechanical Properties of Nb Single Crystals Containing H and H+O", International Conference on "Hydrogen in Metals: Effects on Properties, Selection and Design", Seven Springs, Champion, Pa., September, 1973.

34. A.K. Vasudevan and R. Gibala, "Adoucissement par Alliage du Niobium par Addition de Molybden et Oxygene", Colloque Plasticité, Toulouse, France, March, 1974.

35. R. Gibala, "Dislocation Relaxation in BCC Metals", Max-Planck Institut fur Metallforschung, Stuttgart, Germany, April, 1974.

36. R. Gibala, "Internal Friction in Ferrous Alloys", Université de Paris, Colloque Internationale, Paris, France, April, 1974.

37. R. Gibala, "Relaxation de Dislocation dans Metaux Cubiques Centres", Laboratoire de Mechanique et Physique des Materiaux, Université de Poitiers, Poitiers, France, April, 1974.

38. R. Gibala, "Interactions Hydrogene-Defaut dans Metaux", Centre d'Etudes Nucléaires de Grenoble, Grenoble, France, April, 1974.

39. R. Gibala, "Relaxation de Dislocation dans Metaux Cubiques Centres", Ecole Polytechnique Federale de Lausanne, Lausanne, Switzerland, July, 1974.

40. R. Gibala, "Interstitials in BCC Metals", General Electric Company, Lamp Division, Cleveland, Ohio, October, 1974.

41. R. Gibala, "Effects of Interstitials on Mechanical Behavior of BCC Metals", Rice University, Houston, Texas, December, 1974.

42. R. Gibala, "Anelastic and Magnetic Relaxation in Ni-Co Alloys", Case Western Reserve University, February, 1975.

43. R. Gibala, "Internal Friction of Ferrous Martensites", Conference on Martensite Transformations, M.I.T., Cambridge, Mass., April, 1975.

44. R. Gibala, M. Wuttig, W. Kunz and P. Moser, "Anelastic and Magnetic Relaxation in Face-Centered Cubic Ni-Co-C Alloys", AIME National Spring Meeting, Toronto, Canada, May, 1975.

45. R.P. Krupitzer and R. Gibala, "Effects of Preferred Orientation on Snoek Phenomena in Commercial Steels", AIME National Spring Meeting, Toronto, Canada, May, 1975.

46. V.K. Sethi,and R. Gibala, "Low Temperature Mechanical Behavior of Oxide-Coated Single Crystals of Niobium", AIME National Spring Meeting, Toronto, Canada, May, 1975.

47. A.A. Sagues, M.G. Ulitchny and R. Gibala, "Hydrogen Strengthening in Niobium and Niobium Base Alloys", International Conference on Effects of Hydrogen on Behavior of Materials, Jackson Hole, Wyoming, September, 1975.

48. V.K. Sethi and R. Gibala, "Effect of Oxide Coatings on the Mechanical Properties of Niobium Single Crystals Deformed at Low Temperatures", NATO Advanced Study Institute on Surface Effects in Crystal Plasticity, Hohegeiss, W. Germany, September, 1975.

49. V.K. Sethi and R. Gibala, "Low Temperature Mechanical Behavior of Oxide-Coated Single Crystals of Niobium and Tantalum", ASM/AIME Materials Science Symposium, Cincinnati, Ohio, November, 1975.

50. V.K. Sethi and R. Gibala, "The Effect of Anodic Oxide Coatings on the Mechanical Behavior of Niobium and Tantalum Single Crystals", International Conference on Metallurgical Coatings, San Francisco, California, April, 1976.

51. V.K. Sethi and R. Gibala, "The Effect of Oxygen Distribution on the Low Temperature Mechanical Behavior of Niobium Single Crystals", Second International Conference on Mechanical Behavior of Materials, Boston, Mass., August, 1976.

52. V.K. Sethi and R. Gibala, "Surface Oxide Softening in Single Crystals of Niobium and Tantalum", Fourth International Conference on Strength of Metals and Alloys, Nancy, France, August, 1976.

53. R. Gibala, "Effects of Interstitial Distributions on Mechanical Behavior of BCC Refractory Metals", AIME Fall Meeting, Niagra Falls, N.Y., September, 1976.

54. T.E. Mitchell, V.K. Sethi and R. Gibala, "Interstitial Solution Hardening in BCC Refractory Metals", AIME Fall Meeting, Niagra Falls, N.Y., September, 1976.

55. R. Gibala, "Effects of Interstitial Distributions on Mechanical Behavior of BCC Metals", General Motors Corporation, Warren, Michigan, January 1977.

56. V.K. Sethi and R. Gibala, "An Analysis of Surface Oxide Softening in Niobium Single Crystals", AIME Annual Meeting, Atlanta, Ga., March 1977.

57. R. Gibala, "Hydrogen-Defect Interactions in Metals", Polish Academy of Sciences, Commemorative Symposium on Contributions of Michal Smialowski, Hydrogen Embrittlement and Corrosion of Metals, Warsaw Poland, May 1977.

58. R. Gibala, "Surface Film Softening in BCC Metals", Max-Planck-Institut fur Eisenforschung, Dusseldorf, West Germany, May 1977.

59. R. Gibala "Effects of Interstitial Distributions on Mechanical Behavior of BCC Metals", Institut fur Festkorperforschung der Kernforschungsanlage, Julich, May 1977.

60. R. Gibala, "Effect of Interstitial Distributions on Mechanical Behavior of BCC Metals", Universitat Gottingen, Institut fur Physik, Gottingen, West Germany, May 1977.

61. R. Gibala, "Recent Experiments and Interpretations on Hydrogen-Defect Interactions in Metals", Max-Planck Institut fur Metallforschung, Stuttgart, West Germany, May 1977.

62. R. Gibala, "Effect of Purity, Prestrain and Cooling Rate on Hydrogen Strengthening in Niobium Single Crystals", Second International Congress on Hydrogen in Metals, Paris, France, June 1977.

63. R. Gibala, "Plasticite des Metaux Cubiques Centres a Basse Temperature" Centre d'Etudes Nuclearies de Grenoble, Grenoble, France, June 1977.

64. R. Gibala, "Effect of Interstitial Distributions on Mechanical Behavior of BCC Alloys", Michigan Technological University, Houghton, Michigan, August, 1977.

65. V.K. Sethi and R. Gibala, "Anisotropy of Surface Oxide Softening of Niobium Single Crystals", AIME Fall Meeting, Chicago, Illinois, October, 1977.

66. V.K. Sethi, I. Rusakova and R. Gibala, "Dislocation Substructures in Surface Oxide Softened Niobium", AIME Annual Meeting, Denver, Colorado, March, 1978.

67. V.K. Sethi, J.E. Talia, L. Fernandez and R. Gibala, "Surface Oxide Softening in BCC Refractory Metals", AIME Fall Meeting, St. Louis, Missouri, October, 1978.

68. J. Talia, L. Fernandez, V.K. Sethi and R. Gibala, "Surface Oxide Softening of Body Centered Cubic Refractory Metals", Fifth Int. Conf. on Strength of Metals and Alloys, Aachen, W. Germany, August, 1979 (poster paper).

69. D. DeMiglio, D. Follstaedt and R. Gibala, "The Effect of Implanted Oxygen on the Mechanical Properties of Niobium Single Crystals", AIME Annual Meeting, Las Vegas, Nevada, February, 1980.

70. A. J. Kumnick and R. Gibala, "Hydrogen Trapping in Iron and Steels", Alexander R. Troiano Symposium on Hydrogen Embrittlement and Stress Corrosion Cracking, Cleveland, Ohio, June, 1980.

71. R. Gibala and D. S. DeMiglio, "Hydrogen in Iron and Steels: Interactions, Traps and Crack Paths", Third International Conference on the Effect of Hydrogen on Behavior of Materials, Jackson, Wyoming, August, 1980.

72. R. Gibala, "Hydrogen Embrittlement: Fundamentals of Causes and Cures", Case Institute of Technology, Research Day, Cleveland, Ohio, October 1980.

73. D. S. DeMiglio and R. Gibala, "Observations of the Hydrogen-Assisted Crack Path in 4340 Steel", AIME Annual Meeting, Chicago, Illinois, February, 1981.

74. R. Gibala, "Solid Solution Softening in Body-Centered Cubic Metals", U.S. - Japan Seminar on Characteristics of Mechanical Properties of Body-Centered Cubic Metals", Honolulu, Hawaii, March, 1981.

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THESES

1. M.G. Ulitchny (M.S. , 1969), "Internal Friction and Strain Aging of Austenite".
2. K.V. Ravi (Ph.D., 1969), "The Strength of Niobium-Oxygen Solid Solutions".
3. G.J. Klems (Ph.D., 1971), "Internal Friction of Ferrous Martensites."
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EFFORT DISTRIBUTION OF THE PRINCIPAL INVESTIGATOR

During the contract period February 1, 1980 to January 31, 1981, Professor Ronald Gibala, principal investigator for Contract No. DE-AC02-79ER10359.A000, has allotted his time as follows:

1. February 1, 1980 - May 31, 1980.

25% of these four months was spent on Contract research. 25% of this time was charged to the Contract.

2. June 1, 1980 - August 31, 1980.

65% of this two month period during the summer session was spent on Contract research, and 65% of the time was charged to the Contract.

3. September 1, 1980 - January 31, 1981.

25% of these five months of the academic year was spent on Contract research. The full 25% was financed by the Contract.