HEC-2746

UNIVERSITY OF CALIFORNIA
INSTITUTE OF ENGINEERING RESEARCH
BERKELEY, CALIFORNIA

PRINCIPLES OF SOLUTION HARDENING

Twelfth Technical Report

By

Earl R. Parker and Thomas H. Hazlett

Contract No. AT-11-1-Gen-10, Project #1

SERIES NO. 28

ISSUE NO. 20

DATE October 1953

DISCLAIMER

This report was prepared as an account of work sponsored by an agency of the United States Government. Neither the United States Government nor any agency Thereof, nor any of their employees, makes any warranty, express or implied, or assumes any legal liability or responsibility for the accuracy, completeness, or usefulness of any information, apparatus, product, or process disclosed, or represents that its use would not infringe privately owned rights. Reference herein to any specific commercial product, process, or service by trade name, trademark, manufacturer, or otherwise does not necessarily constitute or imply its endorsement, recommendation, or favoring by the United States Government or any agency thereof. The views and opinions of authors expressed herein do not necessarily state or reflect those of the United States Government or any agency thereof.

DISCLAIMER

Portions of this document may be illegible in electronic image products. Images are produced from the best available original document.

Wo- 38425

AECU - 2745

PRINCIPLES OF SOLUTION HARDENING

By

Earl R. Parker and Thomas H. Hazlett

Twelfth Technical Report

under

U. S. Atomic Energy Commission

Contract AT-11-1-Gen-10, Project 1

October 1953

Series No. 28.

Issue No. 20

0

✓

Introduction

Since ancient times, metals have been alloyed to improve their strength. It was not until well after the turn of the 20th century, however, that reasonable theories of solution hardening were advanced. Rosenhain^(1,2) in the early twenties discussed the properties of solid solution alloys and proposed that the presence of solute atoms "- - - roughens the slip planes", thus causing hardening.

The properties of many solid solutions have been investigated for both single and polycrystalline materials, and the general effects of alloying are well known. These are summarized briefly below:

1. Solid solutions are invariably stronger than the pure parent metals.
2. The yield strength always increases with increasing amounts of an alloying element.
3. At large strains, the slope of the stress-strain curve is usually, but not always, increased when solute atoms are added to a metal.
4. In dilute ternary alloys, each element seems to contribute the same increment of strength that it did in the binary alloy.

A few examples will serve to illustrate the general changes in properties brought about by the addition of a solute element. Fig. 1 shows the results obtained by Sachs and Weerts⁽³⁾ for single crystals of silver-gold alloys. These two elements are mutually soluble over the entire range of composition, and so the effect of composition on the critical stress for slip can be clearly seen. The maximum effect occurs at 50 atomic per cent, and the curve falls in a parabolic manner to minima at the zero per cent alloy points. For dilute alloys, the strengthening effect of solute elements is nearly linear with compositions and is generally so treated.⁽⁴⁾ In some alloy systems, the critical stress for slip in single crystals is markedly affected by small amounts of the added element. Fig. 2, taken from the work of Greenland⁽⁵⁾, shows the effect of small amounts

of silver on the critical shear stress of mercury crystals. However, such large effects due to minute additions are rare.

Data reported in the literature show that in general the effect of alloying in the solid solution range is to increase the rate of work hardening over that observed for the pure metal. However, certain results are at variance with this observation. When zinc is added to cadmium, the rate of strain hardening is lowered.⁽⁴⁾ Similar results were also found for copper crystals containing zinc⁽⁴⁾. The cause of the observed differences in behavior is unknown, but it is obvious even from such preliminary observations that solution hardening is not a simple phenomenon.

Systematic investigations of solution hardening have been made by a number of people. In 1922, Goebels⁽⁶⁾ reported the change in Brinell hardness caused by various additions of Bi, Sn, Hg, and Cd to lead. The following year, Norbury^(7,8) reported the effects of Ag, Si, Mn, Al, Ni, and Zn on the hardness of copper. It was obvious from this early work that the elements causing the greatest change in the parent metal lattice produce the largest change in hardness. However, the increase in hardness was not uniquely determined by the change in atomic volume. The scatter of data obtained when the change in hardness is plotted against the change in lattice parameter is shown in Figs. 3 and 4. Data taken from the work of Frye and Hume-Rothery⁽⁹⁾ are shown in Fig. 3 where the ultimate Meyer hardness is plotted against the square of the change in lattice parameter produced by the addition of various elements to silver. The data in Fig. 4, taken from the work of Brick, Martin, and Angier⁽¹⁰⁾, show the scatter of results obtained when the change in Vickers hardness is plotted against the lattice parameter change for one atomic per cent of various elements added to copper. In a similar investigation on copper alloys, French and Hibbard⁽¹¹⁾ measured the effects of various elements on the yield strength (stress at a strain of 0.01) and lattice parameter. Their results are reproduced in Figs. 5 and 6.

The effects of alloying elements on the strength of binary ferrites have been reported by Bain⁽¹²⁾, Austin⁽¹³⁾, Lacy and Gensamer⁽¹⁴⁾, and by Gensamer⁽¹⁵⁾. These data again confirm the conclusion that the change in lattice parameter is not the sole factor affecting solution hardening.

Dorn, Pietrokovsky, and Tietz⁽¹⁶⁾ have suggested that the second factor influencing solution hardening is the difference in valence between the solute and solvent metals. When they plotted the flow stress for various aluminum alloys against an empirical parameter, including changes in lattice parameter and difference in valence $\{102|\Delta a| + 15|\Delta V| + 7000|\Delta a||\Delta V|\}$, they obtained a single curve correlating the behavior of all alloys tested. (This correlation, however, was based upon the assumption that each aluminum atom has only two free electrons.) The data shown in Fig. 7 were taken from their work. Appreciation of the electronic contribution to solution hardening is certainly a step forward, but inclusion of a valence term is not sufficient to explain the complex behavior of solid solution alloys.

The contributions by Dorn, Pietrokovsky, and Tietz⁽¹⁶⁾, and by Sherby, Anderson and Dorn⁽¹⁷⁾ have been invaluable additions to the literature. These investigators reported the results of extensive and precise work on the influence of alloying elements on the plastic properties of aluminum. Fig. 8 shows the changes found in the lattice constant of aluminum produced by various alloy additions.⁽¹⁸⁾ With the grain size maintained substantially constant, these investigators obtained the stress-strain curves shown in Figs. 9, 10, 11, 12, 13 & 14. They found that the effect of each of the alloying elements could be reduced to an equivalent amount of copper, as shown in Fig. 15.

Stress-strain curves were determined at various temperatures ranging from 78°K to 750°K for each alloy. The effect of temperature on the flow stress (stress at 0.05 strain) are shown in Figs. 16, 17, 18, 19, and 20. From these data, it immediately becomes evident that solution hardening produced by a given

amount of a single element is not a specific value but varies markedly with the temperature. Apparently, at two different temperatures a single element can act in two entirely different ways to cause solution hardening. This complication will be discussed more fully in a later section.

Data for some solid solution nickel alloys have been reported by Zackay and Hazlett⁽¹⁹⁾; other unpublished data by the present authors on nickel alloys have also been included in the following figures. Fig 21 shows the changes in lattice parameter associated with additions of various elements to nickel. Stress-strain curves for these alloys are shown in Figs. 22 through 31, and a cross plot of flow stress vs. temperature is shown in Figs. 32 and 33. The relationships between the room temperature flow stress and the atomic per cent alloy addition are shown for the various alloys in Fig. 34; a plot of the increase in flow stress vs. the lattice parameter change is presented in Fig. 35. Values for nickel-titanium have also been included for comparison.

Results on copper base solid solutions reported by Allen, Schofield, and Tate⁽²⁰⁾ are also of interest. They found that 8.78% Zn, 4.26% Ga, 2.95% Ge, and 2.12% As produced the same change in the stress-strain curve as shown in Fig. 36. In each of these alloys, the electron concentration was 1.087 so they concluded, ".....high charges in some way are the cause of the increased resistance to plastic flow." Even though they neglected the size factor, their correlation was unusually good, as is indicated by Fig. 37. It appears that solution hardening cannot be ascribed simply to the lattice strains or electronic changes introduced by randomly distributed solute atoms. In the following section, the various theories of solution hardening will be presented and discussed in relation to the available data.

MECHANISMS OF SOLUTION HARDENING

Until rather recently, solution hardening has been thought to be due to the

lattice disturbances caused by randomly distributed solute atoms. Theoretical treatments by Cottrell⁽²¹⁾, Suzuki⁽²²⁾, Fisher⁽²³⁾, and others has shown that solution hardening is very likely due to small groups of solute atoms concentrated in the lattice, and not due to randomly dispersed atoms.

Cottrell has pointed out that internal stresses due to randomly dispersed solute atoms should have almost no hardening effect. If a dislocation line is considered as a rigid linear discontinuity, then along its length there will be equal numbers of positive and negative stress fields due to solute atoms. Half of these act to move the dislocation line forward and half urge it in the backward direction. Thus, these randomly alternating stress fields would nearly cancel, and, therefore, dislocations could move through the alloy crystal almost as easily as they could move through a crystal of the pure metal. As Mott and Nabarro⁽²⁴⁾⁽²⁵⁾ have shown, however, dislocation lines are not rigid. They may assume a smoothly curved form, but only on a scale that is large compared with atomic dimensions. When the regions of lattice distortion are far apart, the dislocation line may follow a path around the highly stressed areas. When the distance between stressed regions approaches atomic dimensions, the dislocation line can no longer curve around the stressed regions because of the high energy required to produce sharp curves in a dislocation line.

The energy of a dislocation has been estimated by Mott and Nabarro⁽²⁶⁾ as about Gd^3 per unit length of a dislocation line, where G is the shear modulus and d is the atomic spacing (Gd^3 is of the order of a few electron volts). As

Cottrell⁽²¹⁾ has pointed out, a dislocation may be considered to have a tension

$$\tau = G d^2$$

acting along its length and tending to shorten it. The shape assumed by a dislocation in an internal stress field is given by

$$\tau/\rho = \sigma_i d$$

where ρ is the minimum radius of curvature in the region of the internal stress, σ_i . Combination of the two equations yields

$$\rho/d = G/\sigma_i$$

The radius of curvature in atomic distances is equal to the shear modulus divided by the internal stress. For a radius of curvature approaching atomic dimensions, the internal stress would have to be about as large as the shear modulus. Reasonable estimates indicate that the maximum stress around a solute atom is of the order of $G/100$. Thus, it appears fairly clear that solution hardening cannot be due to stress fields surrounding individual randomly distributed solute atoms.

Since solution hardening cannot be accounted for on the basis of randomly distributed solute atoms, it must be due to collections of these atoms in clusters or in groups having short range order. Several mechanisms of hardening based upon this hypothesis are possible, and each of these will now be discussed in detail.

Cottrell Mechanism: Cottrell has proposed a mechanism of hardening in which the solute atoms will form an atmosphere around dislocations. Atoms above

a positive dislocation are compressed and those below are expanded. The strain energy due to this distortion can be reduced by the collection of large substitutional solute atoms in the expanded region or small ones in the compressed volume. Similarly, interstitial solute atoms causing local lattice expansion will tend to collect in the expanded region of the dislocation. Atoms producing unsymmetrical distortion may relieve shear stresses⁽²⁷⁾ as well as the hydrostatic stresses due to the volume change. Substitutional atoms generally produce symmetrical distortion and hence can only relieve hydrostatic stresses.

Only very small quantities of solute elements are required for the formation of a Cottrell "atmosphere". Since the maximum dislocation density in a highly cold worked metal is of the order of 10^{12} per square centimeter, only 0.1 per cent of solute atoms is sufficient to form a concentration of one solute atom per atom distance along the dislocation lines. In annealed metals, the concentration of dislocations is much lower; even in 99.999 per cent pure metals there may be enough solute atoms to form an atmosphere.

The effect of an atmosphere around dislocations is to raise the stress required to make the dislocation move. When a dislocation is moved away from its atmosphere, the energy of that region of the lattice containing the atmosphere is raised. Since this energy must be supplied by the external load, it is evident that the yield strength must be higher when Cottrell "atmospheres" have formed around dislocations responsible for plastic flow. Once freed from the atmosphere, the dislocations can move freely at lower stresses than those required to initiate plastic flow. The sharp yield point phenomenon found in iron containing a small amount of carbon is apparently due to the formation of an atmosphere of carbon atoms around dislocations⁽²⁸⁾. The theory of solution

hardening due to the interaction of solute atoms with dislocations has been developed to a high state by Cottrell⁽²⁸⁾, Cottrell and Jaswon⁽²⁹⁾, and by Cottrell and Bilby⁽³⁰⁾. They have made calculations of interaction energies, have shown how temperature should influence yield strength, how strain aging in iron is accounted for, and the probable effect of such an atmosphere on slow creep wherein the creep is controlled by diffusion of the atmosphere along with the slow moving dislocations. Consequently, there is little to be gained by developing this subject further here. It is sufficient to say that this phenomenon seems to be well established as a mechanism of solution hardening.

Suzuki Mechanism: A similar mechanism of solution hardening that is not so well known has been proposed by Suzuki⁽²²⁾. Suzuki hardening is also based upon the concentration of solute atoms around dislocations, but in a manner that differs from that proposed by Cottrell. This theory relies upon the dissociation of dislocations in the close packed planes of the face centered cubic and close packed hexagonal crystal structures. Shockley⁽³¹⁾ has shown that undissociated dislocations are unstable when the dislocation line lies in a close packed plane because the energy is less when it is dissociated into two half dislocations of shorter Burgers vector, each of which forms the boundary of a stacking fault. Such an extended dislocation is shown by Fig. 38 in which the atoms of one plane are represented by squares and those in the plane above by circles. The regular hexagonal array is indicated by the lines. During the normal process of slip, the atoms in the plane above would traverse the path indicated by the arrows on the right hand side of the figure. Moving toward the left, the displacement along the slip path gradually diminishes until a region is reached wherein the displacement is one-half of the slip distance. This is a

metastable location which persists for a distance estimated to be of the order of $100^{\circ}\text{A}^{(31)}$. Within this region, a face centered cubic crystal assumes a close packed hexagonal structure (or in a hexagonal crystal it becomes face centered cubic) having a thickness of about two atom planes. The two half dislocations will move apart because of the repulsion due to the elastic energy. The farther they move apart, the more extensive becomes the hexagonal region. Consequently, the half dislocations will cease to move apart when the gain in energy due to separation is insufficient to extend the stacking fault. The lattice to the left of the hexagonal region shown in Fig. 38 gradually assumes the normal configuration.

Thermodynamic reasoning leads to the conclusion that the concentration of solute atoms in the hexagonal region will differ from the average when equilibrium is established between this region and the surrounding face centered cubic material. This heterogeneous distribution of solute atoms will then cause hardening just as the Cottrell atmosphere does in body centered cubic metals (where half dislocations do not form). In the face centered cubic crystals it is rather unlikely that the formation of a hexagonal layer would cause appreciable volumetric disturbance. Hence the concentration of solute atoms therein would depend upon "chemical" (Suzuki) interaction rather than upon interaction energies due to lattice distortion (Cottrell's mechanism). Around each of the half dislocations, however, the lattice would be distorted. Therefore, both types of interaction are theoretically possible in face centered cubic and hexagonal metals. Suzuki cited the work of Allen, Schofield and Tate⁽²⁰⁾ in support of his theory (see Figs. 36 and 37). Quoting from Suzuki⁽²²⁾ "These facts would be understood by the present theory in taking account of Hume-Rothery's law. Namely, the relative free energy of these alloys with close-packed hexagonal structure against those with face centered cubic structure depends only upon the electron concentration, and the free energy of face-centered cubic structure

depends only upon the electron concentration, except for a term proportional to the concentration. The critical shear stress - - - - - is, therefore, a function of electron concentration and is independent of the chemical species of constituents." Suzuki also pointed out (1) that the chemical type interaction is considerably weaker than the interaction due to lattice distortion, and (2) that thermal fluctuations have little effect upon the behavior of extended dislocations because such dislocations would have to migrate fifty or more atom distances to escape from their atmosphere, whereas migration of only a few atom distances would free a dislocation in iron from its atmosphere of carbon atoms.

Fisher Mechanism: Fisher has proposed that short range ordering provides a substantial contribution to solution hardening, particularly at high solute concentrations. When plastic flow occurs, the local ordered regions intersecting the slip planes are disordered by the flow process. If γ is the energy of the disordered interface, τ , the resolved shear stress, and b the magnitude of the Burgers vector, then the energy change associated with the movement of a unit length of a dislocation through a unit distance is given by

$$\Delta E = \gamma - b\tau$$

ΔE is zero at the stress level that will just disorder a region by plastic flow. Hence,

$$\tau_0 = \gamma/b$$

Fisher⁽²³⁾ estimates the misfit energy, γ , to be about the same as the twin boundary energy for copper, i.e. about 20 ergs/cm². Therefore, $\tau_0 \approx 10^9$ dynes/cm², or 15,000 psi. At low temperatures where a fixed degree of order is frozen in, hardening due to short range ordering should be nearly independent of temperature. It follows from Fisher's analysis that clustering would provide a similar strengthening effect.

Another Possible Mechanism: There is another mechanism of solution hardening which has not been discussed previously which seems to be important. As mentioned earlier, a dislocation line in a solid solution alloy containing randomly dispersed solute atoms would be acted upon by the nearby local stress fields in such a manner that the forces acting in the forward direction balanced those in the reverse direction and, since the distance between regions of stress reversal are only a few atom distances, the dislocation line would tend to remain straight and would move almost as freely through the alloy crystal as it would through a pure metal crystal. However, when the solute atoms cluster, the distance between regions of distortion increases and the dislocation line may curve and follow a path around the highly stressed areas. The farther these regions are apart, the easier it is for the dislocation line to curve around them. This condition is illustrated in Fig. 39. The yield strength may thus be expected to vary with the degree of clustering (or short range ordering), ranging from a minimum with random distribution through a maximum at an optimum dispersion and then down again as the clusters become separated too far. This is similar to the precipitation hardening phenomenon except that the "overaging" stage may never actually be reached with solid solution alloys.

Little data could be found in the literature to check the "degree of dispersion" theory outlined above. If this mechanism of solution hardening is operative, the yield strength of the 50 atomic per cent copper-gold alloy solution quenched from above the ordering temperature should vary with the quenching temperature. The size of the short range ordered regions should decrease with increasing temperature, and hence the yield strength should be less the higher the quenching temperature. The contribution to strengthening due to Fisher's disordering mechanism should remain substantially constant for all quenching temperatures because the area of intersection of the ordered

regions and the active slip plane would be about the same for a unit number of large groups or, for example, for twice as many groups of half the size. In this approximation, of course, the variation in ordering energy with group size is neglected. The authors have started a program of research to evaluate the effect of quenching temperature on the plastic properties of copper-gold and other alloys. However, no data are available at present.

There is one experimental result that is of interest in evaluating the "dispersion" effect and the "disordering" factor. Fig. 40 shows data on single crystals of Cu_3Au , taken from the work of Sachs and Weerts⁽³²⁾. The yield strength of the completely ordered crystal was low (indicating that the dislocation lines were straight). However, the rate of work hardening was abnormally high, as Fisher's theory predicts that it should be. The solution quenched crystal which was disordered had a higher yield strength. This could not have been so had the atoms been randomly distributed, and hence it may be concluded that this specimen contained separated regions wherein short range order existed, as indicated by sketch (b) in Fig. 39.

Additional pertinent data are hardness curves made by Carter on 70 per cent gold, 30 per cent platinum as reported by Vines and Wise⁽³³⁾. The platinum-gold system has a miscibility gap, and the 70 per cent gold alloy is within this region. Specimens were quenched from 1000°C and 1150°C , both temperatures being in the solid solution range. The room temperature hardness of the specimen quenched from 1000°C was significantly higher than for the alloy quenched from 1150°C . It may reasonably be assumed that pronounced clustering of like atoms would occur in this alloy, as will be discussed in the next section, and that the size of the clusters would be larger for the specimen quenched from 1000°C than from the one quenched from 1150°C . These results, therefore, constitute quite strong evidence in support of the "degree of dispersion mechanism" of hardening.

Evidence for Clustering and Short Range Ordering

It has been recognized for some time that solute atoms are frequently not randomly distributed. In fact, deviation from randomness is thought to be the rule rather than the exception.⁽³⁴⁾ In alloys like 50 atomic per cent copper-gold, short range ordering is to be expected in the temperature range above the ordering temperature. Thermodynamic measurements, such as the vapor pressure of gold in gold-copper alloys determined by means of radioactive gold⁽³⁵⁾ have shown that short range order does exist even at temperatures approaching the melting point. Furthermore, Birchenall⁽³⁶⁾ has shown that short range order exists in alpha brass, where no long range order occurs.

More recently a group of research workers at MIT have made a number of excellent contributions to the literature on clustering and short range ordering^(37,38,39). They have shown that short range ordering may even occur in some systems which would be expected to exhibit clustering of like atoms, e.g. the gold-nickel alloys.

There is no doubt at all that the distribution of solute atoms is generally other than random and that the mechanical properties of solid solutions are dependent upon the degree of clustering or short range ordering. It is evident that future investigations of solution hardening must be integrated with thermodynamic and x-ray investigations if progress is to be made in the understanding of the influence of alloying on mechanical properties.

Other Factors Influencing Solution Hardening

It is well known that grain size has a significant effect upon plastic properties of metals. Fig. 41, taken from the work of Dorn, Pietrokovsky and Tietz⁽¹⁶⁾, shows the effect of grain size on the stress-strain curve for aluminum. It is also well known that sub-grain structures either exist in metal crystals or develop as a result of mechanical working and heating^(40,41,42,43).

The effect of sub-structures on plastic properties is, however, not so well known. Since sub-structure boundaries generated by plastic flow are formed by the collection of like dislocations⁽⁴²⁾, and since Cottrell and Suzuki have shown that "atmospheres" of solute atoms should form around dislocations, it is important to know what effect a sub-structure would have on the solution hardening contributed by a given element.

Of interest in this connection are some observations which indicate rather clearly that Cottrell and Suzuki atmospheres do form in dislocation type sub-boundaries. Some recent observations on the stress-induced motion of low angle boundaries⁽⁴³⁾ in zinc have shown that in certain temperature ranges the motion of these boundaries is discontinuous. When the stress is continuously increased, the boundary does not move continuously forward as it does at both higher and lower temperatures but jumps from one position to another at a velocity too rapid to follow by eye. It then remains in the new location until the stress becomes so high that it again jumps. This is exactly the performance predicted by Cottrell's theory for zinc containing nitrogen.

Other observations made on aluminum and nickel alloys seem to confirm the Suzuki theory. Fig. 42 shows a fairly well defined "yield point" phenomenon observed by the present authors in an annealed nickel alloy containing 40 atomic per cent copper. As would be expected from the Suzuki theory, the yielding was not as sharp as that due to Cottrell hardening. Also, this behavior was not observed in the more dilute nickel-copper alloys. Thus, it seems probable that Suzuki hardening was operative in this case.

In Fig. 43 a stress-strain curve is shown for an aluminum alloy containing 3.2 atomic per cent magnesium. This alloy did not exhibit a "yield point" in the annealed state. However, a yield point did appear after straining about $4\frac{1}{2}$ per cent at 194°K and aging for 420 minutes at the same temperature. This behavior

is anticipated by the Suzuki theory.

Sub-Structure Hardening Mechanism: Since there seems to be little doubt that dislocation-type sub-structure boundaries have an effect upon the flow stress, it is important to evaluate rather thoroughly the role of the sub-structure. This has been done by the present authors for nickel and a nickel titanium alloy.

Polycrystalline tensile specimens of both pure nickel and nickel containing one atomic per cent of titanium were annealed at 1150°C for one-half hour to remove all possible effects due to prior cold working. Individual specimens of each material were then strained one, two, four, and six per cent in tension and reannealed at 800°C for one hour. X-ray and microscopic examination revealed that the microstructure was substantially unaltered but that the individual grains contained strain free blocks separated by low angle boundaries. This amount of substructure varied with the degree of prestrain, increasing with higher values of strain. When these specimens were tested in tension, the yield strength was found to increase rapidly with increasing amounts of prestrain, as shown in Figs. 44, 45, and 46. This increase in strength is attributed to the presence of the substructure boundaries. Since these boundaries were generated by plastic flow (and heating), it is reasonable to presume that they are dislocation boundaries. Cottrell⁽⁴⁴⁾ has shown that theoretically such a dislocation wall is a barrier to moving dislocations. This has also been demonstrated experimentally with zinc single crystals.⁽⁴³⁾ When a low angle boundary was placed in the path of moving dislocations, the boundary angle was found to increase from about one degree to fifteen. Furthermore, no dislocations penetrated the barrier provided by the dislocation wall in this particular experiment.

An examination of Fig. 46 reveals that the substructure produced by only two per cent pre-strain was more effective in raising the yield strength than was the addition of one per cent of titanium, which is a notably potent alloying element.

A duplicate series of specimens were strained, recovered, and then examined by x-ray microscopic techniques. The number of substructure boundaries produced by the various amounts of pre-strain were counted for a large number of grains in each specimen. Since the number of sub-boundaries for any given specimen was of course not single-valued but varied from grain to grain, it was necessary to analyze these data by statistical methods. The distribution was determined to be self-consistent among specimens and among groups of data taken from different areas of the same specimen as judged by a modification of the Wilcoxon test. These data were then plotted as cumulative frequency curves. For correlation purposes the area under these curves to the seventy-fifth percentile is used as the "index of density of sub-boundaries" and is designated by the symbol $I_{.75}$. (The seventy-fifth percentile was used as a cut-off point in calculating the areas under the curves because this yielded a maximum sensitivity.) Results of this study are shown in Fig. 47 in which the density index is shown to vary in a regular manner with the amount of pre-strain given the specimens.

It is of significance that the alloy has more substructure in the fully annealed state than does the pure nickel. This has been further substantiated by studies of several other nickel alloys. It is also worthy of note that the difference in the amount of substructure present in the pure metal, and the alloy seems to disappear as the amount of pre-strain is increased. It was not possible to determine the sub-boundary indices for specimens having six per cent pre-strain because the density of the sub-boundaries became too high for accurate counting.

When data from the two previous figures are cross-plotted, the relationship between flow stress and density of sub-boundaries is obtained as shown in Fig. 48. It may be seen that the difference in flow stress between pure nickel and the alloy for a given density of sub-boundaries is essentially a constant. This is more clearly illustrated in Fig. 49 in which it may be seen that the increase in flow stress accompanying an increase in the density of sub-boundaries is the

same for both pure nickel and the alloy. Thus the strengthening due to substructure is additive to that produced by the presence of the alloying addition. It is quite probable that this potent influence of substructure has been an important uncontrolled factor in previous studies undertaken to determine the laws governing solid solution hardening.

Conclusions

1. Solid solutions are invariably stronger than the pure parent metals.
2. The yield strength always increases with increasing amounts of an alloying element.
3. The slope of the stress-strain curve is usually, but not always, increased by the addition of a solute element to a pure metal.
4. In dilute ternary alloys, each solute element seems to contribute the same increment of strength that it did in the binary alloy.
5. The difference in size between the solute and solvent atoms is an important factor influencing solution hardening.
6. The difference in the number of free electrons of solute and solvent atoms is another important factor influencing solution hardening.
7. Solution hardening cannot be explained by the interaction of dislocations and the stress fields surrounding randomly dispersed solute atoms.
8. Solution hardening is almost certainly due to the interaction between dislocations and groups of solute atoms.
9. The Cottrell mechanism of hardening, in which an "atmosphere" of solute atoms forms around dislocations, seems to be established by experimental results beyond any reasonable doubt.
10. The Suzuki mechanism of hardening, which is similar to Cottrell's but depends upon "chemical" rather than "mechanical" interaction forces, also seems to be a hardening mechanism.
11. The Fisher mechanism of solution harden, involving the energy supplied mechanically to disorder the short range ordered regions, undoubtedly contributes substantially to solution hardening in certain alloy systems, particularly in the more concentrated solutions.

12. Clustering can contribute to solution hardening, just as does short range ordering, but probably to a lesser degree.
13. The grouping of atoms, either as clusters or in small regions of short range order, can contribute to solution hardening still further by separating the regions of high local strain in the lattice to distances large enough so that dislocations no longer act as rigid lines but curve around and avoid the highly strained regions.
14. Dislocation-type of substructure boundaries have a marked influence on the strength of a metal or alloy, often far greater than the strengthening due to solution hardening.
15. The increment of strength added by a given number of solute atoms is not a fixed value but varies with temperature, sometimes in a complicated manner.
16. Substructure hardening and solution hardening are apparently additive.
17. Since substructures are almost invariably present in any wrought alloy no matter how processed, and since the amount of substructure in an alloy differs from that in a pure metal processed in the same manner, it is necessary to evaluate the increment of strengthening due to substructure before the contribution from solution hardening can be determined.

ACKNOWLEDGEMENTS

The authors wish to express their sincere appreciation to Mrs. Rosa Hansen and Mr. George Kilbourne for performing the tests and correlating data on the original mechanical test results reported herein and to Dr. Betsy Anker for carrying out the studies on sub-structures.

Bibliography

1. "The Hardness of Solid Solutions", by W. Rosenhain, Proc. Roy. Soc., V 99A, 1921, pp. 196-202.
2. "Hardness and Hardening", W. Rosenhain, Chem. & Met. Eng., V 28, 1923, pp. 899-901.
3. "Zugversuche und Gold-Silberkristallen", G. Sachs and J. Weerts, Z. f. Phys., V 62, 1930, pp. 473-493.
4. "Kristallplastizitaet", E. Schmidt and W. Boas, Julius Springer, Berlin, 1935.
5. "The Critical Shear Stress of Mercury Single Crystals", K. M. Greenland, Proc. Roy. Soc. (London), V 163A, 1937, pp. 28-34.
6. "Binaere Bleilegierungen", J. Goebel, Z. f. Metallkunde, V 14, 1922, pp. 357-366, 388-94, 423-425, 449-456.
7. "The Volumes Occupied by the Solute Atoms in Certain Metallic Solid Solutions and Their Consequent Hardening Effects", A. L. Norbury. Trans. Faraday Soc., V 19, 1923, pp. 586-600.
8. "The Hardness of Certain Alpha-Solid Solutions", A. L. Norbury, Jour. Inst. Metals, V 29, 1923, pp. 423-444.
9. "The Hardness of Primary Solid Solutions with Special Reference to Alloys of Silver", J. H. Frye, Jr. and W. Hume-Rothery. Proc. Roy. Soc., V 181A, 1942, pp. 1-14.
10. "Effects of Various Solute Elements on the Hardness and Rolling Textures of Copper", R. M. Brick, D. L. Martin and P. P. Angier, Trans. ASM, V 31, 1943, pp. 675-698.
11. "Tensile Deformation of Copper", R. S. French and W. R. Hibbard Jr., Trans. AIME, V 188, Jan. 1950, pp. 53-58.
12. "Functions of Alloying Elements in Steel", E. C. Bain. ASM, Cleveland, 1939.
13. "Effect of Elements in Solid Solution on Hardness and Response to Heat Treatment of Iron Binary Alloys", C. R. Austin, Trans. ASM, V 31, 1943, pp. 321-339.
14. "The Tensile Properties of Alloyed Ferrites", C. E. Lacy and M. Gensamer, Trans. ASM, V 32, 1944, pp. 88-110.
15. "Campbell Lecture on Strength and Ductility", M. Gensamer. Trans. ASM, V 36, 1946, pp. 30-60.
16. "The Effect of Alloying Elements on the Plastic Properties of Aluminum Alloys", J. E. Dorn, P. Pietrowsky and T. E. Tietz, Jour. Metals, V 188, 1950, pp. 933-943.

17. "Effect of Alloying Elements on the Elevated Temperature Plastic Properties of Alpha Solid Solutions of Aluminum", O. D. Sherby, R. A. Anderson, and J. E. Dorn, Jour. Metals, V 191, 1951, pp. 643-652.
18. "The Lattice Spacing of Solid Solutions of Different Elements in Al", H. J. Axon and W. Hume-Rothery, Proc. Roy. Soc. (London), V 193, 1948, pp. 1-25.
19. "Contribution of d-Level Electrons to the Plastic Properties of Nickel Alloys", V. Zackay and T. H. Hazlett, To be published in Acta Metallurgica.
20. "Mechanical Properties of α -Solid Solutions of Copper with Zinc, Gallium, Germanium and Arsenic", N. P. Allen, T. H. Schofield and A. E. L. Tate, Nature V 168, 1951, pp. 378-379.
21. "Theory of Dislocations", A. H. Cottrell. Progress in Metal Physics, V 1, 1949, pp. 77-126; V IV, 1953, pp. 205-264.
22. "Chemical Interaction of Solute Atoms with Dislocations", H. Suzuki, Science Reports of the Research Inst., Tahaku Univ., Japan, A-Vol. 4, No. 5, 1952, pp. 455-463.
23. "On the Strength of Solid Solution Alloys", J. C. Fisher. Private Communication.
24. "An Attempt to Estimate the Degree of Precipitation Hardening with a Simple Model", N. F. Mott and F. R. N. Nabarro. Proc. Phy. Soc., London, V 52, 1940, pp. 86-89.
25. "The Mechanical Properties of Metallic Solid Solutions", F. R. N. Nabarro, Proc. Phy. Soc. London, V 58, 1946, pp. 669-676.
26. "Dislocation Theory and Transient Creep", N. F. Mott and F. R. N. Nabarro, Rept. Conf. Strength of Solids, Phy. Soc. London, 1948, pp. 1-19.
27. "Effect of Solute Atoms on the Behavior of Dislocations", A. H. Cottrell, Rept. on a Conference on Strength of Solids, Phy. Soc. London, 1948, pp. 30-38.
28. "Mechanical Effects of Carbon in Iron", F. R. N. Nabarro, Rept. on a Conference on Strength of Solids, Phy. Soc. London, 1948, pp. 38-45.
29. "Distribution of Solute Atoms Round a Slow Dislocation", A. H. Cottrell and M. A. Jaswon, Proc. Phy. Soc., London, V 199, 1949, pp. 104-114.
30. "Dislocation Theory of Yielding and Strain Aging of Iron", A. H. Cottrell and B. A. Bilby, Proc. Phy. Soc., London, V 62A, 1949, pp. 49-62.
31. "Study of Slip in Al Crystals by Electron Microscope and Electron Diffraction Methods", R. D. Heidenreich, W. Shockley, Rept. Conference Strength of Solids, Phys. Soc. London, 1948, pp. 57-74.

32. "Untersuchungen an der Legierung $AuCu_3$ ", G. Sachs and J. Weerts, *Z. f. Physik*, V 67, 1931, p. 507.
33. "Age Hardening Precious Metal Alloys", R. F. Vines and E. M. Wise, *Age Hardening of Metals*, AMS, Cleveland, 1940, pp. 190-228.
34. "Clustering in Solid Solutions", R. Smoluchowski, *Int. Conference on Phy. of Metals* (Amsterdam 12-17 July 1948), The Hague, Martinus Nijhoff, 1949, pp. 179-183.
35. "A Thermodynamic Study of Gold and Gold-Copper Alloys", L. D. Hall. Ph.D. Thesis in Metallurgy, Univ. Calif., Berkeley, 1950.
36. "Interaction and Structure in Copper-Zinc Alloys", C. E. Birchenall, *Trans. AIME*, V 171, 1947, pp. 167-175.
37. "Atomic Arrangements in Gold-Nickel Solid Solutions", P. A. Flinn. Sc.D. Thesis, Metallurgy Dept., MIT, Cambridge, Mass. 1952.
38. "The Diffuse Scattering of X-rays", B. E. Warren and B. L. Averbach, Modern Research Techniques in Physical Metallurgy, ASM, Cleveland, Ohio, 1953, pp. 95-130.
39. "Measurements in Clustering in Solid Al-Zn Alloys", P. S. Rudman, P. A. Flinn and B. L. Averbach. *Jour. Appl. Phy.* V 24, 1953, p. 365.
40. "Internal Strains and Recrystallization", R. W. Cahn, Progress in Metal Physics II, Interscience Publ., New York, 1950, pp. 151-174.
41. "Substructures in Crystals", A. Guinier, Imperfections in Nearly Perfect Crystals, Wiley, New York, 1952, pp. 402-440.
42. "Stress-Induced Movement of Crystal Boundaries", C. H. Li, E. H. Edwards, J. Washburn and E. R. Parker, *Acta Metallurgica*, V 1, March 1953, pp. 223-229.
43. "Recent Observations on the Motion of Small Angle Dislocation Boundaries", D. W. Bainbridge, Choh Hsien Li, E. H. Edwards, J. Washburn, E. R. Parker, ONR Project N 7-onr-29516, NR-039-009, 9th Tech. Rept., Series 27, Aug. 1953.
44. "The Yield Point in Single Crystal and Polycrystalline Metals", A. H. Cottrell. *Symposium on Plastic Deformation of Crystalline Solids at Mellon Institute*, Pittsburgh, May 1950, pp. 60-76.

FIG. 1 VARIATION OF CRITICAL SHEAR STRESS WITH
COMPOSITION OF Ag - Au SINGLE CRYSTALS
(SACHS & WEERTS)

FIG.2 EFFECT OF TRACES OF SILVER
ON THE CRITICAL SHEAR STRESS
OF SINGLE CRYSTALS OF MERCURY
(GREENLAND)

FIG.3 EFFECT OF LATTICE STRAIN ON THE HARDNESS OF SOME SILVER ALLOYS

(FRYE & HUME-ROTHERY)

FIG. 4 EFFECT OF VARIOUS ELEMENTS IN COPPER
ON INCREASING THE VICKER'S HARDNESS
NUMBER

(BRICK, ET AL.)

FIG. 5 EFFECT OF SOLUTE CONCENTRATION ON THE
YIELD STRENGTH ($\sigma_{0.01}$) OF COPPER ALLOYS
(FRENCH & HIBBARD)

FIG. 6 EFFECT OF CHANGE IN LATTICE PARAMETER ON
THE CHANGE IN YIELD STRENGTH ($\sigma_{0.01}$)
AS PRODUCED BY THE ADDITION OF THE INDICATED
SOLUTE ELEMENTS TO COPPER.
(FRENCH & HIBBARD)

FIG. 7 VARIATION OF FLOW STRESS WITH $\left\{102|\Delta a| + 15|\Delta v| + 7000|\Delta a|\Delta v|\right\} \times$ ATOMIC PERCENT FOR STRAIN OF $\epsilon = 0.05$, AT 295 °K

FIG. 8 EFFECT OF ALLOYING ELEMENTS ON THE LATTICE CONSTANT OF ALUMINUM (AXON & HUME-ROTHERY)

FIG. 9 TENSILE STRESS-STRAIN CURVES
FOR ALUMINUM-MAGNESIUM ALLOYS
AT 293° K (20° C.)

(DORN, PIETROKOWSKY & TIETZ)

FIG. 10 TENSILE STRESS-STRAIN CURVES
FOR ALUMINUM-COPPER ALLOYS
AT 296° K (23° C)

(DORN, PIETROKOWSKY & TIETZ)

FIG. II TENSILE STRESS-STRAIN CURVES
FOR ALUMINUM-ZINC ALLOYS
AT 290° K (23° C)

(DORN, PIETROKOWSKY & TIETZ)

FIG.12 TENSILE STRESS-STRAIN CURVES
FOR ALUMINUM-GERMANIUM ALLOYS
AT 296° K (23° C)

(DORN, PIETROWKOWSKY & TIETZ)

FIG. 13 TENSILE STRESS-STRAIN CURVES
FOR ALUMINUM-CADMIUM ALLOYS
AT 248° K (25° C)

(DORN, PIETROKOWSKY & TIETZ)

FIG.14 TENSILE STRESS-STRAIN CURVES
FOR ALUMINUM-SILVER ALLOYS
AT 295° K (22° C)

(DORN, PIETROKOWSKY & TIETZ)

FIG.15 THE STRAIN HARDENING EQUIVALENCY OF THE VARIOUS SOLUTE ELEMENTS AS BASED ON THE STRAIN HARDENING OF THE COPPER ALLOYS

(DORN, PIETROKOWSKY & TIETZ)

FIG.16 EFFECT OF TEST TEMPERATURE ON THE FLOW STRESS OF MAGNESIUM-ALUMINUM ALLOYS FOR A TRUE STRAIN OF 0.05

(SHERBY, ANDERSON & DORN)

FIG.17 EFFECT OF TEST TEMPERATURE ON THE
FLOW STRESS OF COPPER - ALUMINUM
ALLOYS FOR A TRUE STRAIN OF 0.05

(SHERBY, ANDERSON & DORN)

FIG.18 EFFECT OF TEST TEMPERATURE ON THE FLOW STRESS OF ZINC-ALUMINUM ALLOYS FOR A TRUE STRAIN OF 0.05

(SHERBY, ANDERSON & DORN)

FIG. 19 EFFECT OF TEST TEMPERATURE ON THE FLOW STRESS OF GERMANIUM-ALUMINUM ALLOYS FOR A TRUE STRAIN OF 0.05

(SHERBY, ANDERSON & DORN)

FIG. 20 EFFECT OF TEST TEMPERATURE ON THE
FLOW STRESS OF SILVER - ALUMINUM
ALLOYS FOR A TRUE STRAIN OF 0.05
(SHERBY, ANDERSON & DORN)

FIG. 21 CHANGE OF LATTICE PARAMETER WITH ALLOY CONTENT FOR NICKEL ALLOYS

FIG. 22 TENSILE STRESS-STRAIN CURVES FOR NICKEL-TITANIUM ALLOYS AT 25 °C

FIG.23 TENSILE STRESS-STRAIN CURVES FOR
NICKEL-IRON ALLOYS AT 25 °C

FIG. 24 TENSILE STRESS-STRAIN CURVES FOR
NICKEL-COBALT ALLOYS AT 25 °C

FIG.25 TENSILE STRESS-STRAIN CURVES FOR NICKEL-COPPER ALLOYS AT 25°C

49

FIG.26 TENSILE STRESS-STRAIN CURVES FOR
 NICKEL-COBALT ALLOYS AT 700 °C

FIG. 27 TENSILE STRESS-STRAIN CURVES FOR
 NICKEL-IRON ALLOYS AT 700 °C

FIG.28 TENSILE STRESS-STRAIN CURVES FOR NICKEL -
COPPER ALLOYS AT 700° C

FIG. 29 TENSILE STRESS-STRAIN CURVES FOR
NICKEL - COBALT ALLOYS AT -195°C

FIG. 30 TENSILE STRESS-STRAIN CURVES FOR
NICKEL - IRON ALLOYS AT -195°C

FIG. 31 TENSILE STRESS-STRAIN CURVES
FOR NICKEL AND 10% Ni-Cu ALLOY
AT -195 °C

FIG. 32 VARIATION OF FLOW STRESS WITH TEMPERATURE
FOR NICKEL-COBALT ALLOYS

FIG.33 VARIATION OF FLOW STRESS WITH TEMPERATURE FOR NICKEL-IRON ALLOYS

FIG. 34 EFFECT OF ALLOY ADDITIONS ON FLOW STRESS OF NICKEL ALLOYS

FIG. 35 RELATIONSHIP BETWEEN INCREASE IN FLOW STRESS AND CHANGE OF LATTICE PARAMETERS OF SEVERAL NICKEL ALLOYS

FIG. 36 TRUE STRESS-STRAIN CURVES FOR COPPER-ZINC, COPPER-GALLIUM, COPPER-GERMANIUM AND COPPER-ARSENIC ALLOY WIRES OF APPROXIMATELY THE SAME ELECTRON ATOM RATIO

(ALLEN, SCHOFIELD & TATE)

FIG. 37 EFFECT OF ELECTRON CONCENTRATION ON
THE YIELD STRENGTH OF COPPER ALLOYS

(ALLEN, SCHOFIELD & TATE)

EXTENDED DISLOCATION

FIG. 38 EXTENDED DISLOCATION IN FACE CENTERED CUBIC CRYSTAL
CONSISTING OF A STRIP OF HEXAGONAL CLOSE PACKING
BOUNDED BY TWO EXTENDED HALF DISLOCATIONS.

FIG.39 EQUILIBRIUM SHAPE OF A DISLOCATION LINE IN A SOLID SOLUTION ALLOY DEPENDS UPON THE DEGREE OF DISPERSION OF SOLUTE ATOMS. FOR RANDOM DISTRIBUTION, LINE IS STRAIGHT AS IN (a); FOR CLUSTERING (OR SHORT RANGE ORDERING), LINE ASSUMES CURVED SHAPE AS IN (b)

FIG.40 STRAIN HARDENING CURVES FOR SINGLE CRYSTALS
OF Cu_3Au IN THE ORDERED AND DISORDERED
STATES
(DATA BY SACHS & WEERTS)

FIG.41 THE EFFECT OF GRAIN SIZE ON THE TENSILE STRESS-STRAIN CURVE FOR PURE ALUMINUM AT 292°K (19°C)

(DORN, PIETROKOWSKY & TIETZ)

FIG. 42 YIELD POINT PHENOMENA IN ANNEALED
NICKEL-COPPER ALLOY

FIG.43 EFFECT OF A STRAIN-AGING TREATMENT
ON THE TENSILE STRESS-STRAIN CURVE
OF ALUMINUM CONTAINING 3.2 ATOMIC
PERCENT MAGNESIUM AT 194° K (-79° C)

(SHERBY, ANDERSON & DORN)

FIG. 44 EFFECT OF AMOUNT OF SUB-STRUCTURE
ON STRENGTH OF PURE NICKEL

FIG.45 EFFECT OF AMOUNT OF SUB-STRUCTURE
ON STRENGTH OF NICKEL-TITANIUM ALLOY

FIG. 46 CHANGE OF FLOW STRESS PRODUCED BY INDICATED AMOUNT OF PRE-STRAIN FOLLOWED BY AN 800 °C ANNEAL

FIG. 47 CHANGE IN DENSITY OF, SUB-BOUNDARIES PRODUCED BY INDICATED AMOUNT OF PRE-STRAIN FOLLOWED BY AN 800 °C ANNEAL

FIG. 48 RELATIONSHIP BETWEEN DENSITY OF SUB-BOUNDARIES AND FLOW STRESS FOR PURE NICKEL AND A NICKEL - TITANIUM ALLOY

FIG. 49 INCREASE IN FLOW STRESS ACCOMPANING INCREASE IN DENSITY OF SUB-BOUNDARIES FOR PURE NICKEL AND Ni-Ti ALLOY

DISTRIBUTION LIST

Argonne National Laboratory.....	4
Knolls Atomic Power Laboratory.....	3
Mound Laboratory.....	2
Battelle Memorial Institute.....	1
Carbide & Carbon Chemicals Division (Y-12).....	2
Carbide & Carbon Chemicals Division (K-25).....	2
Ames Laboratory.....	1
National Advisory Committee for Aeronautics, Dr. R. F. Selden.....	1
North American Aviation, Inc., Dr. Chauncey Starr.....	1
Brookhaven National Laboratory.....	3
Los Alamos Scientific Laboratory.....	2
Oak Ridge National Laboratory.....	4
General Electric Company, Richland.....	2
Massachusetts Institute of Technology, Dr. A. R. Kaufmann.....	1
National Bureau of Standards, Dr. Taylor.....	2
Sylvania Electric Products, Inc.....	2
Sandia Corporation.....	1
University of California, Berkeley, Radiation Laboratory.....	3
Dow Chemical Company, Denver.....	1
Bureau of Ships.....	1
Office of Naval Research, Code 423.....	1
Naval Research Laboratory.....	2
Dr. Cyril S. Smith, University of Chicago.....	1
Dr. B. S. Old, Arthur D. Little, Inc.....	1
U. S. Atomic Energy Commission, Chicago Operations Office.....	1
Westinghouse Electric Corporation.....	2
E. I du Pont de Nemours & Co.....	2
Aircraft Nuclear Propulsion Proj., Oak Ridge.....	1
Office of Naval Research, Code 252.....	1
Office of Ordnance Research, Duke University.....	1
Dr. J. B. Johnson, Wright-Patterson Air Force Base.....	1
Dr. Maxwell Gensamer, Columbia University.....	1
U. S. Atomic Energy Commission, Hanford Operations Office.....	1
U. S. Atomic Energy Commission, Idaho Operations Office.....	1
U. S. Atomic Energy Commission, New York Operations Office.....	3
U. S. Atomic Energy Commission, Savannah River Operations Office.....	1
U. S. Atomic Energy Commission, Oak Ridge Operations Office.....	1
U. S. Atomic Energy Commission, Washington, Librarian.....	2
U. S. Atomic Energy Commission, Metallurgy & Materials Branch.....	1
U.S.A.E.C. New York Operations Office, National Lead Company.....	1
U.S.A.E.C. Schenectady Operations Office.....	1
U. S. Atomic Energy Commission, Library Branch, Oak Ridge.....	15
U. S. Atomic Energy Commission, Patent Branch, Washington.....	1
U. S. Atomic Energy Commission, San Francisco Operations Office.....	2
Dr. J. H. Hollomon, G. E. Research Laboratory.....	1
Dr. R. Smoluchowski, Carnegie Institute of Technology.....	1
Dr. Morris Cohen, Massachusetts Institute of Technology.....	1

DISTRIBUTION LIST

Dr. B. L. Averbach, Massachusetts Institute of Technology.....	1
Dr. Frederick Seitz, University of Illinois.....	1
Dr. E. E. Stansbury, University of Tennessee.....	1
Dr. H. B. Huntington, Rensselaer Polytechnic Institute.....	1
Dr. T. A. Read, Columbia University.....	1
Dr. R. M. Brick, University of Pennsylvania.....	1
Dr. W. M. Baldwin, Jr., Case Institute of Technology.....	1
Mr. S. J. Sindeband, American Electro Metals Corporation.....	1
Dr. D. J. Clark, California Institute of Technology.....	1
Dr. A. J. Herzig, Climax Molybdenum Corporation.....	1
Dr. Harry B. Goodwin, Battelle Memorial Institute.....	1
Dr. A. J. Shaler, The Pennsylvania State College.....	1
Dr. R. Maddin, Johns Hopkins University.....	1
Dr. H. Eyring, University of Utah.....	1
Dr. W. J. Moore, Indiana University.....	1
Dr. P. A. Beck, University of Illinois.....	1
Dr. T. J. Neubert, Illinois Institute of Technology.....	1
Office of Naval Research, Code 421.....	1
Dr. A. G. Guy, Dept. of Chem. & Metallurgical Eng., Purdue University...	1
U. C. Institute of Eng. Research, Bldg T-3, Campus	13
California Research & Develop. Co., Livermore, Research Lab.....	1
Dr. B. W. Gamson, Great Lakes Carbon Corporation.....	1

0-74

THIS PAGE
WAS INTENTIONALLY
LEFT BLANK