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High-Density, High-Strength Alloys
Having Some Ductility at
Ambient Temperature
A State of the Art Survey

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scientific laboratory
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E. G. Zukas
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HIGH-DENSITY, HIGH-STRENGTH ALLOYS HAVING SOME DUCTILITY AT AMBIENT TEMPERATURE

A State of the Art Survey

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ABSTRACT

The literature was surveyed in an attempt to define a program which would be successful in producing an alloy having high strength, high density, and low-temperature ductility. Rhenium and its alloys, tungsten and its alloys, tantalum and its alloys, platinum and its alloys, metal additions to the gamma-phase 8 wt % Mo - U alloy, and composites were considered. Work is already in progress on other uranium alloys by CMB-6 and Rocky Flats, and the possibilities could be better evaluated by them. The possible areas for study are: (a) the evaluation of a commercial composite manufactured by P. D. Mallory and Company which has high density, high strength, and good ductility; (b) using the 8 wt % Mo - U alloy as the matrix of a composite strengthened with tungsten or tungsten alloy wire. We would expect high density, high strength and good ductility; (c) studying tungsten composites with a matrix of copper, silver, gold, tin, or their alloys. The techniques for producing composites should be developed for we consider these the materials of the future; (d) the study of the mechanical deformation at low and elevated temperatures of tantalum-base alloys to increase their strength. Tantalum is ductile down to liquid hydrogen temperatures; (e) the evaluation of the low-temperature behavior of the W - 35 wt % Re - 18 wt % Mo alloy. There appears to be no way in which the rhenium content can be lowered, so the cost will remain high; (f) the study of the addition of other alloying elements to tantalum, and larger additions of some elements which have already been surveyed. The hope would be to obtain an alloy with sufficient strength by solid solution hardening that could be simply cast to useable shape; (g) ignoring cost, cold work or explosively form pure rhenium or rhenium-base alloys to increase their yield strength. These already have good low-temperature ductility, and are strengthened by very little cold deformation; (h) ignoring cost, strengthen platinum by additions of tungsten, molybdenum, nickel or something else; (i) the study of W - Re alloys, but this should be discouraged since not only is the cost high, but there are many associated problems; (j) the study of factors which cause embrittlement in tungsten and tungsten-base alloys; and (k) the addition of silicon, zirconium and/or vanadium to the 8 wt % Mo - U alloy.

We recommend: (1) the evaluation of the commercial high-density, high-strength composites; (2) a study of producing high-density composites probably with tungsten reinforcement and perhaps with platinum, gold, copper, silver, or other metals or alloys as matrix; and (3) the evaluation of the Mo-W-Re ternary alloys which may possess adequate strength with good low-temperature ductility even though their cost will remain high.

Fiber reinforcement of the 8 wt % Mo-U alloy with tungsten wire or tungsten-alloy wire should be explored, but we do not have facilities for producing uranium alloy castings. We recommend that such a program be discussed with CMB-6 since they have the casting facilities; were first in producing uranium-molybdenum alloys, and possess the technological ability to ensure success in

such a program. Alloying additions to the 8 wt % Mo-U alloy should also be seriously considered.

I. INTRODUCTION

For certain applications, it is desirable to have a material of high density and high strength accompanied by some degree of ductility for use at room temperature and slightly below. If the lower limit on density is arbitrarily placed at 15.5 g/cm^3 , then tantalum, tungsten, rhenium, osmium, iridium, platinum, gold, uranium, and alloys of these metals are the only materials that can meet the density requirement. Very little has been done with osmium and iridium, and there has been no systematic study to produce high-strength alloys with gold and platinum. A number of investigations on uranium alloys are already underway and there will be no attempt here to summarize and evaluate the available data.

In this report, the metals tantalum, tungsten, platinum, and rhenium, and their alloys as well as composites will be discussed. Evaluations will be made and guides for future research suggested.

II. RHENIUM AND RHENIUM BASE ALLOYS

Rhenium is one of the Group VII B transition metals appearing in the third long period of elements between tungsten and osmium. It is hexagonal-close-packed with lattice constants a of $2.760 \text{ \AA}$ and c of $4.458 \text{ \AA}$. It has a density of 21.0 g/cm^3 , melts at 3180°C , and has an elastic modulus of $67 \times 10^6 \text{ psi}$.

Rhenium sheet is formed by sintering powder compacts in vacuum at 2400°C or higher followed by cold rolling with appropriate annealing or recrystallization treatments. Such material⁽¹⁾ contains from 50 to 500 ppm iron, 20 ppm carbon, 25 ppm oxygen, 1 ppm hydrogen, 2 ppm nitrogen, and less than 100 ppm of other elements.

Rhenium does not undergo any ductile-brittle transition even in the recrystallized condition.⁽²⁾ As a result, rhenium can be readily welded by tungsten arc inert gas or electron beam techniques with the welds being completely ductile. The deformation mechanisms were investigated by Churchman⁽³⁾ who found slip on basal, prismatic and pyramidal planes and extensive twinning predomi-

nantly on the $(11\bar{2}1)$ planes with secondary twinning on the $(11\bar{2}2)$ and $(10\bar{1}2)$ planes. Approximately 20% reduction by cold work can be attained before annealing is necessary. The addition of 10 wt % W or 5 wt % Mo enabled the amount of cold reduction to be increased to 30% before edge cracking occurred,⁽⁴⁾ however the addition of 5 wt % Ta gave no improvement.⁽⁵⁾ Rhenium does not form a carbide but dissolves about 0.85% C in solid solution⁽⁶⁾ which increases its hardness from 150 to 500 VHN. On slow cooling, graphite precipitates and only about 0.3% is soluble at 2000°C , and the hardness drops to about 350 VHN. In any case, the rhenium is not embrittled by the carbon. The ultimate tensile strength of rhenium⁽⁷⁾ is about 175,000 psi in the annealed condition, 220,000 psi after 9% reduction and about 350,000 psi after 15% reduction. The yield strength in the recrystallized condition at 0.2% offset is about 50,000 psi and the elongation is about 25%. The yield and elongation values after cold work are not given. Softening after cold work becomes appreciable at about 1000°C and recrystallization is well under way at 1500°C .

It is very difficult to hot-work or to hot-cold-work rhenium because of hot-shortness which is caused by small amounts of rhenium oxide in the grain boundaries. This oxide is volatile at 363°C .⁽⁸⁾ Low melting eutectics are also formed with boron, silicon, and phosphorous, and these impurities must be kept low.

Rhenium can be electroplated from aqueous solution using either acid or alkaline baths or vapor deposited from rhenium pentachloride or rhenium hexafluoride. It can also be plasma sprayed but the bond and the resulting properties are not as good as those obtained by vapor deposition.

Very little has been done in studying the properties of rhenium-base alloys. Tungsten, molybdenum, and tantalum have been added in an attempt to improve low-temperature fabricability with some success; the amount of cold work before edge cracking is increased from 20% to 30%.

We recommend the use of pure rhenium where the use of a cold forming operation or explosive forming are used to increase the yield strength to an acceptable value. Additional effort should be expended in a study of rhenium-base alloys, especially those with cobalt, chromium, nickel, platinum, and vanadium.

III. TUNGSTEN AND TUNGSTEN-BASE ALLOYS

Tungsten and tungsten-base alloys have high density and high strength. However, they are plagued by low-temperature brittleness. This section of the report will discuss the brittleness problem in both tungsten and tungsten alloys, and will evaluate the practical aspects and possible solutions.

In tensile testing of brittle materials, the specimens often break at the root of a thread or at some scratch or flaw which acts as a stress concentration point. The same problem exists in the bend test, and in other tests used to determine the temperature of the ductile-brittle transition. Therefore, if one is testing the same material, the surface condition is an important consideration. For example, ⁽⁹⁾ the ductile-brittle transition temperature of chromium was measured as -45°C for an electro-polished sample, +45°C for a surface finished with a fine file, and +120°C for a sample finished with a coarse file. Similar behavior has been observed for tungsten and other brittle materials.

The type of test used to determine the ductile-brittle transition temperature also has an influence on the results obtained. Lowest ductile-brittle transition temperatures are obtained in torsion tests, next lowest are those obtained in tension tests followed rather closely by the popular bend test, then the unnotched impact test, and finally the notched impact tests which give the highest transition temperatures. Comparisons must be made on specimens using the same type test or at least a little "Kentucky windage" should be used if comparisons using different test methods can not be avoided.

The ductile-brittle transition temperature increases as the strain rate of the test increases. Therefore, for comparative results the strain rate during testing must be held constant. Referring back to the previous paragraph, one notes that the different type tests each use a

different strain rate, and this factor may be more influential on the ductile-brittle transition temperature than the actual type of test used.

The grain size is also an important factor, the larger the grain size (all other factors being equal) the higher the ductile-brittle transition temperature. This is attributed to the smaller grain boundary surface area. This factor will be covered in more detail later.

With surface condition, the type of test, the strain rate during testing, the grain size and other variables affecting the measurement of the ductile-brittle transition temperature, it is not surprising then that there is considerable scatter in the reported data.

The presence of impurities at the grain boundaries, especially hydrogen, carbon, nitrogen, and oxygen are suspect in promoting brittle fracture. The low-temperature solubility of these impurities was studied ⁽¹⁰⁾ and estimates of the interstitial contents likely to be maintained in solid solution after moderate cooling rates were made for a number of the refractory metals. The interstitial contents likely to be maintained in tungsten were much less than 0.1 ppm for hydrogen, carbon, and nitrogen and 1 ppm for oxygen. The evidence for interstitials at the grain boundaries causing brittle fracture is pretty convincing. First, brittle fracture follows the grain boundaries. Secondly, a single crystal produced by electron beam melting had a ductile-brittle transition temperature of -150°C whereas the same material after warm working and recrystallization was brittle below +150°C. Until recently, carbon, nitrogen, hydrogen, and oxygen were the only interstitials considered important. However, a recent study using Auger spectroscopy ⁽¹¹⁾ showed that phosphorous, which also segregates at the grain boundaries was very effective in increasing the ductile-brittle transition temperature. For example, a specimen containing enough phosphorous (estimated at about 50 ppm) to give an Auger peak height of 40 had a ductile-brittle transition temperature of approximately 800°C, 250°C at a peak height of 12, and 50°C at zero peak height. These results indicate that while phosphorous is undoubtedly an embrittling contaminant, others must also be present since the single crystal results showed that tungsten can be duc-

tile to -150°C .

In June of 1955, Geach and Hughes⁽¹²⁾ reported that they had produced alloys of rhenium with tungsten and/or molybdenum which had good high-temperature properties and that could be fabricated at low temperatures; in fact some of these could be fabricated at room temperature with reductions of up to 90% without intermediate annealing. The compositions showing maximum ductility were those containing the maximum amount of rhenium in solid solution without producing excessive sigma phase, i.e., slightly more than 25 wt % rhenium in tungsten⁽¹³⁾ and to almost 50 wt % rhenium in molybdenum.⁽¹⁴⁾ The binary molybdenum-rhenium alloys are too low in density to meet the requirements set forth in the introduction of this report. The maximum rhenium addition to molybdenum before excessive sigma phase formation is 50 wt % (35 at.%), which gives an alloy having a density of about 13.8 g/cm^3 . If the density requirements could be relaxed, this alloy would then appear promising. In the recrystallized condition⁽²⁾ it has a tensile strength of 150,000 psi, a yield of 123,000 psi and about 18% elongation in a 2" gauge length. Its elastic modulus is 59×10^6 psi. The ductile-brittle transition is less than -200°C . Cold work increases the tensile properties; specimens reduced 50% by cold work have a tensile strength of 240,000 psi and a yield strength of 210,000 psi.

Getting back to tungsten and tungsten-rhenium alloys, the ductile-brittle transition temperature is a real problem and will be discussed in much more detail from a practical standpoint. If the embrittlement is caused by interstitial impurities migrating to the grain boundaries, then for a given impurity content, the smaller the grain size (the larger the grain boundary surface area per unit volume) the thinner the impurity thickness at the boundaries. Figure 1 shows a plot of grain size versus the ductile-brittle transition temperature plotted from data of Campbell and Dickinson⁽¹⁵⁾ for electron beam melted tungsten. The same ductility data are also plotted against the ratio of the volume of a surface layer one Angstrom thick to the total grain volume. These data also produce a straight line relationship just slightly displaced from the grain size plot. This relationship is shown because

Fig. 1. The effect of grain size and surface-to-volume ratio on the ductile-brittle transition temperature in tungsten. Data from reference (15) used.

we feel that the important variable is the grain boundary surface where the impurities concentrate and not the grain size. Of course this argument is purely academic when equiaxed grains are present and the ratios are equivalent, but what number do you choose for grain diameter when the grain shape is cylindrical after rolling and its length is several times its diameter?

The previous discussion serves to introduce our assumption that the grain boundary surface area for a particular impurity content is the important factor and that the decreases in the ductile-brittle transition temperatures brought about by cold rolling or by the drawing of wire is related to this surface area increase and not to the deformation per se, although it is possible that the defects and dislocations produced may act as deposition sites for the impurities. This could be proven experimentally by shock loading tungsten at temperatures above its ductile-

brittle transition temperature (making sure that the strain rate factor is taken into account) to different shock pressures and measuring the change in ductility. However, we are not proposing such a program at this time. If the grain boundary surface area is the important factor, then we should know how it varies with reduction in area either from rolling or drawing. This is shown in Fig. 2 where the surface area increase for a starting cylinder 1 cm diameter x 1 cm long is plotted versus the reduction in area. The important thing to note is that the reduction in area must really be large to get any appreciable increase in grain boundary surface area. For example, to create a 50% increase in grain boundary surface area requires a reduction in cross-sectional area of almost 80%. We believe that this is the most important factor involved primarily because the new surface area formed by the deformation is free of the impurity contaminants. The temperature at which the new surface area is formed and to which it is subsequently exposed becomes very important because the impurities will diffuse into the newly formed boundaries, and rapidly, because its impurity concentration is zero and the concentration gradient is high. The ductile-brittle transition temperature for tungsten reduced different amounts is shown in Fig. 3. The scatter band is fairly large since data from several sources are included. There are several features of this plot that are very important. First, there is essentially no decrease in the transition temperature for reductions below 60%.

Fig. 2. The increase in grain boundary surface area as a function of the reduction in area where the grain shape is assumed to remain a right circular cylinder.

Fig. 3. The upper and lower limits of the ductile-brittle transition temperature for tungsten as a function of the reduction in area effected either by rolling or by drawing.

Referring back to Fig. 2, a 60% reduction in area increases the new grain boundary surface area only by a mere 20%. Secondly, when the wire is drawn sufficiently so that its cross section consists of only one grain, it behaves as a single crystal and has the same transition temperature as a single crystal.

Rhenium is the only alloying addition tried so far that lowers the ductile-brittle transition temperature in tungsten. The changes in transition temperature with rhenium content for both tungsten and molybdenum are shown in Fig. 4 drawn from data presented by Hahn et al.⁽¹⁰⁾ Note that there is a limited composition range where rhenium is effective in lowering the ductile-brittle transition temperature when alloyed with tungsten, (25 to 28 wt % rhenium) and even under the best circumstances does not

Fig. 4. The ductile-brittle transition temperature as a function of rhenium content for either tungsten-rhenium or molybdenum-rhenium alloys.

Fig. 5. The upper and lower temperature limits of the ductile-brittle transition temperature as a function of the reduction in area for W-25 wt % Re made either by arc-casting or by powder metallurgy techniques.

make tungsten ductile below 0°C. However, rhenium is effective in any concentration below 50 wt % in molybdenum. At rhenium contents greater than 30 wt % in tungsten and 50 wt % in molybdenum, appreciable amounts of the brittle sigma phase is formed, and this destroys the low-temperature ductility.

There are several reasons given for the rhenium ductilizing effect. Wukusick⁽¹⁶⁾ believes that the low-temperature ductility is caused by removal of interstitial elements from the matrix as a result of rhenium-induced clustering. In 1958, Jaffee et al.⁽¹⁷⁾ found that in molybdenum, an eutectic between molybdenum and MoO₂ formed and migrated to the grain boundaries. There, it has a low dihedral angle with the grain boundary indicating that it wets the boundaries. When rhenium was added, a complex rhenium-molybdenum oxide formed and these inclusions were shaped as globules, and furthermore these inclusions remained within the grains. A similar complex oxide was postulated for the tungsten alloys, but the amounts formed were too small for positive identification. In 1964, Booth, Jaffee, and Salkovitz⁽¹⁸⁾ reported that the mechanisms of the rhenium-ductilizing effect remain elusive and that the factors which might bear on the explanation were dislocation velocity-stress relationship, effect of interstitial solubility, surface energies including liquid surface tension, grain boundary energy, twin boundary energy and stacking-fault energy, electron-

ic structure and/or other factors.

From Fig. 4, it is readily apparent that if low-temperature ductility is the important factor, then the obvious choice of composition for the W-Re binary system is W-25 wt % Re. Earlier experimentalists recognized this, and a concerted effort was made to determine the properties and behavior of this particular composition. Since the surface area of the grain boundaries is an important factor, the ideal solution would seem to be one in which the grain size was kept as small as possible. This can be done by making pieces by powder metallurgy methods. A comparison of the ductile-brittle behavior for the W-25 wt % Re alloy made by powder metallurgy and by arc-melting as a function of reduction of area is shown in Fig. 5. The data^(19,20,21) are not as extensive as that for pure tungsten and may be subject to some slight revision. The powder metallurgy alloy, even though it has a small grain size, does not have as good low-temperature ductility as the same alloy made by arc casting. From these results, it appears that the influence of impurities outweigh those of grain size. In the powder metallurgy process, the large surface area of the powder particles must be purified during sintering. Also, since the artifact surface reaches temperature first, it is difficult to keep from trapping impurities in the alloy even though they may be readily reduced by hydrogen. Another important factor is that the as-sintered density is lower (86-94%)⁽²²⁾ thereby allowing internal contamination to take place during exposure to the atmosphere while mechanical deformation is taking place.

From Figs. 3 and 5, it appears that we could achieve satisfactory low-temperature ductility simply by some mechanical process which would effect a 75 to 80% reduction in area thereby creating enough new grain boundary surface area to promote ductility (about 50% more grain boundary surface). To do this at room temperature would be very difficult for at least two reasons. First, the material is below its ductile-brittle transition temperature and would probably fracture. Second, if we substitute what appears to be reasonable values (i.e., yield = 140,000 psi, reduction of area = 75%, etc) into the equations for cold extruding,⁽²³⁾ a die material having a yield strength

Fig. 6. The ultimate strength and yield strength of the W-25 wt % Re alloy as a function of test temperature.

in excess of 400,000 psi would be required. Since the tensile strength decreases as the temperature is increased, then perhaps pieces could be made by extruding the hot alloy through cold or relatively cold dies. The tensile and yield strengths^(17,24,25,26) versus temperature for the 25 wt % Re-W alloy are shown in Fig. 6. At 1000°C, the yield strength is less than half that at room temperature, and at 1600°C, the yield strength is less than one-fourth the room temperature value. Therefore, extrusion is feasible at an elevated temperature. However, at the elevated temperature, the impurities would become mobile and diffuse to the grain boundaries, the rate of diffusion being dependent upon temperature and the time at temperature. In tungsten the electrical resistivity of heavily worked wire begins to decrease at about 400°C⁽²⁷⁾ and according to Gripshover et al.,⁽²⁸⁾ over 80% of the quenched-in resistance increase annealed away at slightly over 600°C indicating defect mobility at low temperatures. However, the diffusion of impurities to the grain boundaries is not too rapid below 1000°C. The ductility transition temperature for rolled tungsten, W-6.5 wt % Re and W-26 wt % Re as affected by one hour holds at various annealing temperatures is shown in Fig. 7. Note that the transition temperature of the test samples had been lowered by the cold working process. These data are from the work of Klopp and Witzke.⁽²⁹⁾ For times of one hour at 1000°C, the ductility transition temperature is raised by about 30°C, while at 1600°C, the increase is about 135°C for the W-25 wt % Re alloy and it is even more for

Fig. 7. The effect of an one-hour hold at an elevated annealing temperature on the ductile-brittle transition temperature for tungsten and two tungsten-rhenium alloys.

tungsten containing less rhenium.

A number of other alloying elements have been added to obtain solid solution strengthening⁽³⁰⁾ and still others to obtain carbide dispersions.⁽³¹⁾ These additions were not successful in lowering the ductile-brittle transition temperature significantly.

Hafnium nitride and thorium in amounts to 4.5% have been added to tungsten and tungsten rhenium alloys.^(32,33,34) These additions are somewhat effective in improving high temperature strength, but they do not improve low-temperature ductility. A suggested explanation was that these alloys have to be made by powder metallurgy methods, and so it is impossible to lower the impurity content sufficiently for good ductility.

A study has been made on tungsten-technetium alloys⁽³⁵⁾ and an improvement in lowering the ductile-brittle transition temperature noted. For example, the addition of 3.5 at. % Tc lowered the ductile-brittle transition temperature from 675°C to 280°C. Such results need verification with a higher purity tungsten as the base material.

Rhenium improves the low-temperature ductility and workability of molybdenum, and to a somewhat lesser degree does the same for tungsten. It is not surprising then that the Re-W-Mo ternary alloys were studied with the view that perhaps the amount of the very expensive rhenium could be reduced with little loss in low temperature ductility and high-temperature strength. The ternary

Fig. 8. Workability diagram for W-Re-Mo alloys with the commercial alloys designated.

system was investigated quite extensively by Sims and Jaffee.⁽³⁶⁾ They plotted a diagram showing the isoworkability lines which produced visible cracking by 10% and 80% reduction on rolling at room temperature. These same data have been replotted on the basis of weight percent composition and the commercial alloys designated. This plot is shown in Fig. 8. The Cleveland Refractories metals data sheet⁽³⁷⁾ gives some properties of the W-35 Re - 18 Mo (W-30 at.% Re-30 at.% Mo) alloy. Its density is 17.0 g/cm^3 , its modulus of elasticity is between 49 and 55×10^6 psi, it has a ductile-brittle transition below room temperature, and it has a Vickers hardness in the recrystallized condition of 380. From this, one would guess its yield strength at about 130-135,000 psi. Milner et al.⁽³⁸⁾ produced this same alloy by direct conversion of powders. The powders were dry-blended together with 3% of methyl cellulose binder and mixed with water and a glycerin plasticizer to the consistency of a clay body. This body was extruded into 0.300" diameter rods which were prefired in wet hydrogen at 1200°C and vacuum sintered at 2200°C for 5 hours. The product had a density of 16.3 g/cm^3 (95.9% theoretical). The as-sintered prop-

Fig. 9. The influence of tungsten and/or molybdenum additions on the ductile-brittle transition temperature of tantalum.

erties at room temperature were: (1) ultimate strength = 165,000 psi, (2) 0.2% offset yield strength = 115,000 psi, (3) elongation = 10%. This alloy had appreciable ductility at -70°C and could be given a drawing reduction of 35% at room temperature without prior thermal or mechanical treatment. Because of the improved low-temperature ductility, this alloy appears better than the W-25 wt % Re binary. However, it does have a lower density, a lower modulus of elasticity, and would be more expensive since it requires a greater percentage of rhenium.

IV. TANTALUM AND TANTALUM BASE ALLOYS

Tantalum is a relatively weak metal of atomic weight 180.95, melts at 2996°C and has a density of 16.6 g/cm^3 . This metal is ductile to very low temperatures, and this ductility persists even when the metal contains fairly large amounts (several hundred ppm) of interstitial impurities. It is embrittled by hydrogen and nitrogen and must therefore be protected at elevated temperatures. Its room temperature tensile strength^(39,40) is 29,000 psi, its yield strength is 26,000 psi and it has an elongation of 36%. The presence of oxygen, nitrogen, or carbon in fairly large amounts can increase its tensile strength to 73,000 psi with the elongation dropping to about 25%.

Tungsten and molybdenum, separately and together, have been added to tantalum to improve its high-temperature strength. The properties of some of these and many other alloys are described in the literature.⁽⁴¹⁻⁶³⁾

For obtaining high density, one would like to add as much tungsten as possible. However, the addition of either tungsten or molybdenum or a combination of both raises the ductile-brittle transition temperature. Fig. 9 shows this effect, (taken from the work of Bartlett et al.⁽⁵⁰⁾) in weight percent alloying addition and with temperature in degrees centigrade. On this basis, tungsten is a much preferred alloying element. Additions of about 18 wt % tungsten raise the ductile-brittle transition temperature to room temperature; therefore, increased low-temperature strength will require the addition of other alloying elements which will retain low-temperature ductility (e.g., rhodium etc.) or the alloy will require some treatment to increase its strength such as cold or warm mechanical working or perhaps an explosive forming operation.

First, let us consider some mechanical treatment to improve the strength of an alloy that is otherwise acceptable. For the purpose of this discussion, we will use the alloy Ta-10 W, not because it is the best alloy choice but because it is a commonly used commercial alloy which has been extensively studied. After recrystallization or annealing at high temperature, it has an ultimate strength of 85-105,000 psi, a yield strength of 75-95,000 psi, an elongation of 20 to 35% with good low-temperature ductility. The tensile and yield strengths as a function of temperature are shown in Fig. 10. The question is, can this alloy be cold worked or warm worked sufficiently to achieve a yield strength of about 140,000 psi? For the W-25 wt % Re alloy, the problem was to cold or warm work it sufficiently to increase the grain boundary surface

Fig. 10. The ultimate strength and yield strength for the Ta-10 wt % W alloy at elevated temperature.

Fig. 11. The ultimate strength and yield strength of the Ta-10 wt % W alloy after reduction by rolling at either 1095°C or at 1370°C.

area and in this way achieve low-temperature ductility. In this case, we have low-temperature ductility but must cold or warm work this alloy to achieve satisfactory strength. Fig. 10 shows that the strength decreases rapidly as the temperature is increased; it is only 1/3 as strong at 1100°C as it is at room temperature and has a yield strength of well under 20,000 psi at 1400°C. From these data, there appears to be no reason why extrusion by dynapacking or some other process could not be used successfully. But would the alloy be strengthened by working at 1100°C or at 1400°C? Fig. 11 shows the room temperature strengths of the Ta-10 W alloy after rolling various amounts at 1095°C and at 1370°C.⁽⁴⁴⁾ Obviously, warm working below the recrystallization temperature does strengthen the alloy, the degree of strengthening decreasing with increasing working temperature. The recrystallization temperature of this alloy is roughly 1600°C, and mechanical work at this temperature or above would produce a recrystallized structure with no improvement in room temperature properties.

The data presented in the previous section indicate a method for achieving high strength and good low-temperature ductility. But a yield of 140,000 psi might be achieved more readily. Therefore, one should consider alloys having a higher initial yield strength before deformation. The Ta-9 Hf-5 Re alloy has a yield strength of 116,000 psi,⁽⁶⁰⁾ the Ta-5 W-5 Re one of 108,000 psi,⁽⁶⁰⁾ the Ta-9.6 W-2.4 Hf-.01 C one of 100,000 psi,⁽⁴⁷⁾ the Ta-10 W-

1 Zr one of 130,000 psi. ⁽⁴⁷⁾ Many other alloys have been investigated but are either inferior strengthwise at room temperature or have only been tested at elevated temperatures.

The Ta-10 W alloy can be cold worked to 90% reduction at room temperature and in this condition has a tensile strength of 200,000 psi. At an ultimate of 180,000, its yield strength is 164,000 psi. ⁽⁴¹⁾ This shows that the alloy hardens readily. This suggests the feasibility of using an explosive forming operation where some of the hardening could be accomplished by passage of the shock wave.

There are no Ta-alloys listed in the literature which have the desired strength in the recrystallized condition. The strongest is the one mentioned earlier, the Ta-10 W-1 Zr alloy which has an ultimate strength of 137,000 psi, a yield of 130,000 psi and 16% elongation at room temperature. A number of alloys were surveyed ⁽⁵³⁾ that probably possessed adequate strength but were not evaluated as to tensile properties and low-temperature ductility. These single phase alloys (even after a homogenization treatment at 2000°C) will be listed along with their measured hardnesses so that an idea of their probable strengths can be obtained. 1. Ta-8 W-2 Hf = 255 DPH (for this commercial alloy, the ultimate strength is 90,000 psi and its yield is 80,000 psi) 2. Ta-8 Re = 355 DPH 3. Ta-16 Re = 520 DPH 4. Ta-16 W-4 Re = 375 DPH 5. Ta-8 W-8 Re = 420 DPH 6. Ta-16 W-8 Re = 455 DPH 7. Ta-4 W-16 Re = 525 DPH 8. Ta-8 W-16 Re = 540 DPH 9. Ta-16 W-8 Hf = 465 DPH 10. Ta-8 W-16 Hf = 390 DPH. The ultimate strength of tantalum can be increased to 160,000 psi by the addition of 10% vanadium. ⁽⁴²⁾ These results indicate that the desired strength can be achieved by alloying tantalum with other metals, but experimental verification will be required to ensure that the alloys possess the necessary low-temperature ductility.

V. COMPOSITES

Composite materials are defined as a class of materials containing two or more phases. As such, they can hardly be classed as new to the field of materials science. All carbon steels which contain pearlite, cementite, or carbides, and all precipitation or dispersion hardening

alloys are multi-phase, and therefore are, by definition composites. But the broad approach of the last ten years or so is new. In this new approach, each component is chosen for one particular property and combined with another which has some other desirable property. For example, tungsten with good wear and abrasion resistance was combined with silver which has good electrical conductivity to produce long life, low resistance electrical contacts. Thus, the composite materials approach reduces the problem to one of specifying the desired properties in the finished product, choosing the necessary components and structures that will give these properties, and then finding some manufacturing method whereby these properties can be combined.

To illustrate further, let us design a material that would satisfy the requirements specified in this report, i.e., high density, high-strength, and low-temperature ductility. The high-density metals are tungsten, tantalum, uranium, rhenium, iridium, platinum, and gold. None of these in the pure state except possibly tungsten has adequate strength. However, it does not have low-temperature ductility. To overcome this difficulty, the tungsten can be embedded in a ductile matrix. Since the volume of the matrix is limited to about 10 or 15%, the density, strength, and other properties are not very important, and such a composite could meet the necessary requirements. Composites have also been made which use high-density uranium as the matrix ⁽⁶⁴⁾ with reinforcements of tungsten, molybdenum, tantalum, or niobium. Unfortunately, the mechanical properties for these composites have not been reported.

Considering tungsten composites, there are several general comments which are applicable. To achieve the best density, sintering should take place in the liquid phase. The molds must be evacuated and the application of some pressure is desirable to eliminate air pockets and reduce porosity. However, when sintering in the liquid phase, the temperatures obviously are fairly high. Therefore, the time at temperature must be kept as short as possible to keep grain growth to a minimum. ⁽⁶⁵⁾ If sintering is done in the solid phase, then rolling or some other mechanical deformation operation becomes neces-

sary to achieve a satisfactory density.

The ability to predict the elastic and plastic properties of two-phase composite materials from a knowledge of the properties of each individual constituent is of theoretical as well as practical interest. Two factors are considered in determining the upper and lower boundaries of the elastic moduli of the composite. First, there is the strong material weakened by the introduction of the weak second phase, and second, the weak second phase being strengthened by the addition of the stronger phase. According to Krock⁽⁶⁵⁾ a theory based on minimum polarization was derived by Haskin and Schtrikman.⁽⁶⁷⁾ This theory, based on both the shear modulus and bulk modulus of the individual components and their concentrations, appeared reasonable but the difference between the lower bound and upper bound was only about 1%. The experimental data generally did not fall within the two boundaries. A second theory based on minimum energy was put forth by Paul.⁽⁶⁸⁾ The upper bound, or linear mixture law, represents the situation where, under load, the phases strain an equivalent amount, and is represented by the equation

$$E_{UB} = X_A E_A + X_B E_B \quad (1)$$

where E_{UB} is the elastic modulus at the upper bound for the composite, X_A and X_B are the concentrations of the two materials, and E_A and E_B are the elastic moduli of the individual materials A and B. The lower bound represents the case where, under load, the two materials are equally stressed, and is represented by the equation

$$E_{LB} = E_A E_B / (X_A E_B + X_B E_A) \quad (2)$$

Here E_{LB} is the lower bound of the modulus. From these equations, the difference between the bounds for any concentration is much larger, and at the point of maximum separation, may reach 20%. The elastic moduli determined experimentally for a number of composites fall between the upper and lower bounds specified by the above equations.^(69, 70, 71)

Much has been said about predicting the elastic modulus of a composite material. However, of even greater importance is predicting the strength of the composite. The equation for predicting the upper bound of the modulus,

with proper adjustment of the variables, predicts quite accurately the stress relationship.^(72, 73, 74) The equation is:

$$\sigma_c = V_f \sigma_f + V_m \sigma_m \quad (3)$$

where σ_c is the stress of the composite, V_f and V_m the volume fraction of the fiber and matrix respectively, and σ_f and σ_m their stresses. Examples of experimental data compared with calculated results are shown in Fig. 12 for .005 inch diameter tungsten wire reinforcement in a 90:10 brass matrix taken from the work of Schoene and Scala.⁽⁷⁵⁾

For tungsten fiber composites, continuous wire filaments, wire segments, wire fragments, wire mesh screens and tungsten powder have all been used. The final properties, however, depend quite markedly on the properties of the tungsten filler used, as specified by equation (3) in the last paragraph. When using a tungsten filler, there are a number of factors that should be considered. If a powder is chosen, what procedure does one use to remove the surface oxide? If a low-density skeletal body is formed by

Fig. 12. Stress-strain curves for 0.005" diameter tungsten wire reinforcement in a 90 Cu-10 Zn brass matrix.

Fig. 13. Yield strength versus the volume content of either tungsten fiber or continuous tungsten powder skeleton in a copper matrix. From Ref. 76.

low-temperature sintering, then one has to rely on sufficient hedgehog porosity for the infiltrant liquid metal to fill essentially all of the pores, otherwise a low-density product results. Also, how does one go about improving the mechanical properties of the powder, and what effect does particle size have on the properties, amount of infiltrant needed, etc.? Perhaps the powders could be given a low-temperature reducing treatment to clean the surfaces, and then either rolled at low temperature or shock loaded to increase their strength. Another possibility of course is getting prealloyed powder where the powder particles are solid solution hardened by small amounts of some alloying element.

Now, one questions whether there is any particular advantage in using either dispersed particle strengthening, continuous skeleton composites, or fiber-reinforced composites. The differences were studied by Krook.⁽⁷⁶⁾ To make a valid comparison, the properties of the tungsten

wire had to be adjusted so as to be comparable to the properties of the dispersed tungsten powder and to the tungsten used in the continuous skeleton. He concluded that for a given concentration of reinforcement, the stiffness of filament-reinforced composite materials of tungsten and copper is greater by about 20% than either dispersed particle or continuous skeleton composite materials. This superiority decreases only slightly with increasing composition of the reinforcing phase. Therefore, the greatest improvement in absolute strengthening of the fiber-reinforced materials is due to the superior strength of the fiber itself. This strengthening advantage can be appreciable, as can be seen from the strength increase obtained by drawing tungsten into very fine diameter wires where its tensile strength can exceed 400,000 psi. A comparison of yield strength versus volume concentration reinforcement for tungsten wire and continuous tungsten skeleton taken from Ref. 76 is shown in Fig. 13.

At this point, the use of wire or filament reinforcement looks promising. Cold drawn tungsten wire can be produced with tensile strengths exceeding 400,000 psi, finishing operations are at low temperatures which removes the surface oxides, the ductile-brittle transition temperature has been lowered to about -150°C , and the wires are quite flexible, especially in small diameters. A serious drawback here is that the infiltration must be done at some elevated temperature, and there are at least three serious problems which must be contended with. These are:

- (1) the infiltrant might react with the wire, perhaps forming a brittle intermetallic on the wire surface, (2) the interstitial impurities might diffuse to the grain boundaries in the tungsten wire, thereby rendering it brittle, and (3) the wire might recrystallize thereby becoming both brittle and weak. The effect of these on the mechanical properties of the tungsten wire have already been discussed at considerable length in the section on tungsten and tungsten alloys. There are a number of things that must now be considered. For example, the infiltrant or matrix material should have a reasonably low melting point so that diffusion of interstitial impurities in the tungsten wire would be slow, but at the same time, its surface tension should be as low as possible so that small pores are readily filled

Also, the use of a thoriated tungsten wire having a much higher recrystallization temperature might be desirable. Very little has been done in evaluating these possibilities.

From a practical standpoint, the use of wire might complicate the production problems. It's difficult to powder blend a twisted wad of wire with some infiltrant powder. To alleviate this problem, the wires could be chopped into short segments. The shorter the wire segments, the greater the possibility for good blending. The question now is, how short can these wire segments be before they change their characteristics and behave like powders? This problem has received some attention,⁽⁷⁵⁾ and the off-hand answer is an interesting one. It depends on the temperature. A length of wire at room temperature may be a length of powder at 1000°C. An equation has been worked out based on results for the Inconel matrix-tungsten wire system. The equation is:

$$l = \frac{\sigma d}{\tau 2} \quad (4)$$

where l is the minimum length required, σ is the stress in the wire, τ is the mean shear stress in the matrix near the wire-matrix interface, and d is the wire diameter.

Thus, the aspect ratio, i.e., the shape of the fiber is very important. As the temperature increases, τ decreases and therefore l must increase. Using 0.003" diameter tungsten wire and plugging Ellison and Harris'⁽⁷⁵⁾ values into the equation, a minimum length of 0.008" is necessary at room temperature, and this length for 0.003" diameter wire increases to 0.060" when the temperature is raised to 1000°C. For 0.010" wire, the minimum length for room temperature strength would be about 0.027".

There seems to be some question as to whether the matrix must wet the wire surface, and perhaps alloy slightly with it in order to obtain a good composite. The results with tungsten wire using a copper matrix indicate that alloying is unnecessary. Fig. 14 reproduces some of the results of McDanels et al.⁽⁷³⁾ There seems to be a straight line correlation between the properties of the matrix and those of tungsten. The results plotted here are for 0.003" diameter tungsten wire which had a tensile strength of 331,000 psi. Additions to almost 70 volume percent were used. They also used wires of 0.005" and 0.007" diameter with similar results. Naturally, the

Fig. 14. The influence of the volume concentration of 0.003" diameter tungsten wire on the tensile strength, yield strength, and elastic modulus when embedded in a copper matrix. From McDanels et al.⁽⁷³⁾ Experimental data available to about 70% fiber with the experimental points generating the straight line drawn.

strengths of the larger diameter wires are somewhat lower. Some effort has been expended using silver as the matrix material^(77,78) with both tungsten powder and tungsten wire. However, the amounts of tungsten used and the conditions imposed does not enable one to make a good evaluation.

The use of nickel and nickel-base alloys as matrix materials has been studied more extensively than other types and P. R. Mallory and Co., Inc., markets a number of these alloy composites commercially. They market an alloy called "Gyromet 1100" which has a density of 17.25 g/cm³, a tensile strength of 150,000 psi, a yield strength of 140,000 psi, a proportional limit of 80,000 psi, an elongation of 2% and a modulus of elasticity of 48×10^6 psi.⁽⁷⁹⁾ They developed a new alloy⁽⁸⁰⁾ by adding 0.5% ruthenium

to Gyromet 1100 with some slight adjustment of the other components. This "Anviloy" alloy has a density of 17.25 g/cm^3 , a tensile strength of 160,000 psi, the yield strength is not given, a proportional limit of 133,000 psi and an elongation of 10%. These alloys can be cold worked and can be welded using a special welding rod without serious changes in properties. Parts are formed by hydrostatic pressing of the powders (in cylindrical or tubular shapes to 7" diameter) thereby giving essentially isotropic property behavior. The present cost for these alloys is nominally \$10 to \$15 a pound. The properties of these materials are such that they merit further evaluation. In fact, they seem to meet the specification requirements which were the original objectives of this search.

From the results obtained in earlier studies of uranium-molybdenum alloys, the 8 wt % Mo-U alloy and variations of it have quite good mechanical properties. Its tensile strength is around 122,000 psi with just a slightly lower yield. During tensile testing, it obeys Hooke's law nicely giving an essentially straight line plot of stress versus strain until the yield strength is reached. It is quite ductile (10% or more elongation) and has good impact strength. We propose to improve the nearly adequate strength of the 8 wt % Mo-U alloy by the addition of tungsten fibers. Fig. 15 shows the calculated tensile strength, yield strength and elastic modulus curves derived from the law of mixtures for adding 0.007" diameter tungsten wire filaments to the uranium-alloy matrix. It is interesting to note that the addition of only 30 volume percent tungsten wire should give a composite having a tensile strength of 173,000 psi, a yield strength of 166,000 psi and an elastic modulus of 28×10^6 psi.

The fabrication technique used most often for composites involves mixing solid powders of the materials followed by a bonding step. But oxidation of uranium alloy powders would surely prevent good bonding. We propose, therefore, to add finely chopped solid tungsten wire to the molten uranium alloy. Placing tungsten wire in a mold might present some problems on distribution of the reinforcement due to the buoyancy effects. To alleviate this problem, the density of the fiber could be adjusted to match that of the liquid phase matrix. We carried out a

Fig. 15. The theoretical curve showing the expected influence of the volume content of 0.007" diameter tungsten wire in an 8 wt % Mo-U alloy matrix on the tensile strength, yield strength and elastic modulus.

rather primitive experiment to determine the feasibility of this. To match the density of the U-8 wt % Mo alloy, which we assumed to be 18.0 g/cm^3 , we chose an 86 wt % W-14 wt % Mo alloy. An arc-melted button of this composition was crushed into powder in a mortar and pestle. The powder was placed in the bottom of the arc-melting mold and the U-8 wt % Mo alloy matrix arc-melted over the powder giving a composite of 28 vol % (W-Mo) alloy filler. The distribution of the powder in the matrix was fairly uniform. A few photomicrographs are shown in Fig. 16 in which the light areas are of the W-Mo particles. Two important features are demonstrated. First, the matrix alloy is sufficiently fluid and of low enough surface tension that small cracks in the powder particles are filled, and second, the angular shape of the powder indicates that no chemical reaction took place. The hardness of the

Fig. 16. Microstructures of 86 wt % W-14 wt % Mo powder particles embedded in an 8 wt % Mo-U alloy matrix.

matrix was DPH 270 (the same as the matrix alloy itself) while that of the powder was DFH 365. The hardness results were not unexpected, however, since tungsten is completely insoluble in the U-8 wt % Mo alloy. This was established by first making an alloy of 16 to 1 molybdenum to tungsten ratio, and later it was melted with uranium. The tungsten coalesced into discrete particles on solidification. Furthermore, the melting point of the 86 wt % W-14 wt % Mo alloy is 3120°C while that of the U-8 wt % Mo alloy is 1240°C. With this difference in melting temperature, we should expect that the leaching of the molybdenum from the powder to be extremely slow at temperatures even several hundred degrees above the melting point of the matrix alloy. Our recommendation is that thoriated tungsten-molybdenum alloy be drawn into small diameter wire (to improve its mechanical properties and ductility), chopped into segments, and added to the molten uranium-alloy.

The same technique could be used with the 2 wt % Mo-U alloy. In this case, the yield strength of the matrix metal would be much lower, only about 65,000 psi, and so the volume addition of the W-Mo alloy would have to be about 40 volume % according to the law of mixtures. The density of a composite could be 18.1 g/cm³ if the combinations are chosen to equate the liquid 2 wt % Mo-U and the solid W-Mo densities at the melting temperature of the matrix. (The composition of the solid particles should be 80 wt % W-20 wt % Mo). An experimental melt using alloy powder of fairly large size could then be made and the distribution in the melt analyzed. The proper adjustment in alloy composition for wire reinforcement could then be made.

Similar composites could be made using other uranium-base alloys as the matrix material. Also, the fiber addition could be a tungsten-tantalum alloy, with perhaps other additives to improve the fiber strength. Such combinations are almost endless, and will not be discussed further here.

VI. PLATINUM AND PLATINUM-BASE ALLOYS

Platinum, with an atomic number of 78, is the end member of transition elements in the periodic table. It is face-centered-cubic, has a density of 21.45 g/cm³, a

Fig. 17. The tensile strengths of platinum alloys containing either tungsten, nickel, or ruthenium.

modulus of elasticity of about 22×10^6 psi, and melts at 1769°C. Unalloyed platinum is very ductile to subzero temperatures, and can be cold worked to quite large reductions. However, in the unalloyed condition, it does not have sufficient strength to meet the requirements specified in this proposal. (81)

The properties of platinum and platinum alloys were surveyed by Vines and Wise (82) and more recently by the Metals Handbook. (83) If the cost of the platinum-base alloys can be ignored, then a number of the alloys appear interesting. The strength versus percent alloying addition for ruthenium, nickel, and tungsten are shown in Fig. 17. These alloys are heat-treatable to some extent, are work hardenable, are capable of achieving high strengths, have high densities, and are reasonably ductile. If these alloys cannot be considered because of the cost of the platinum and the limited amount of platinum available, then they merit consideration as matrix alloys for composites along with the gold-platinum alloys, and others.

TABLE I

Alloy Composition	Water Quenched from 950°C		Held for 1 hour at 950°C and Furnace Cooled		
	Hardness	Structure	Hardness	Structure	Ductility
U-8 Mo	270	γ	271	γ	(4)
U-8 Mo-2 Pt	265	--	272	γ	(1)
U-8 Mo-2 Cr	--	--	292	γ	(1)
U-8 Mo-3 Re	Free Rhenium present				
U-8 Mo-2 Re	265	γ (6)	276	γ	(1)
U-8 Mo-0.5 W	270	Free W came out of solution			(1)
U-8 Mo-0.25 Si	268	γ (6)	273	γ (6)	(1)
U-8 Mo-1.5 Si	330	--	346	$\gamma + ?$	(2)
U-8 Mo-1.5 Zr	289	γ (7)	276	γ (6)	(4)
U-8 Mo-2 Zr	282	γ	348	$\gamma + \alpha$ (trace)	(3)
U-6 Mo-2 Zr	270	γ	268	γ	(1)
U-8 Mo-0.5 V	268	γ (6)	280	γ	(1)
U-8 Mo-1 V	264	γ (5)	360	$\gamma + \alpha$ (5, 7)	(4)
U-8 Mo-2 V	196	γ	385	$\gamma + \alpha$	(2)
U-6 Mo-1.5 V	163	γ (7)	399	$\gamma + \alpha$ (5)	(3)
U-8 Mo-0.5 V-1 Zr	279	γ	272	γ (6)	(1)
U-8 Mo-0.25 Si-0.5 V-1 Zr	277	γ	288	γ	(1)

(1) Not measured.

(2) Brittle.

(3) Some ductility.

(4) Good ductility.

(5) Some primary V solid solution detected.

(6) Some compositional inhomogeneity was apparent.

(7) Exhibited pronounced structural distortion.

VII. ALLOYING ADDITIONS TO AN 8 WT % Mo-U

ALLOY BASE

We suggested earlier that the 8 wt % Mo-U alloy be used as a matrix in a composite. However, since its strength properties are almost adequate, a suitable alloying addition might enhance its strength sufficiently. We arc-melted a number of ternary and higher alloys which were subsequently either water quenched or furnace cooled after holding for one hour at 950°C. The hardness results, phases present, and some ductility data are presented in Table I.

Alloying additions which can be ignored either because they form grain boundary precipitates (no ductility) or do not strengthen the alloy are tungsten, platinum, rhenium, chromium, nickel, cobalt, iron, and manganese. Small additions of titanium have been used successfully

by others. The elements silicon, zirconium, and vanadium, either singly or in combination offer hope since they do increase the hardness of the base alloy significantly. Silicon additions increase the hardness, but additions of 0.5% or more result in the formation of a brittle intermetallic. However, an addition of 0.25% (the minimum amount was not determined) decreased the grain size appreciably, and may be an important alloying addition solely on this basis.

A zirconium addition of 1.5 wt % does not harden the alloy, while a 2 wt % addition does. This is probably a result of stress induced by the partial transformation to the α -phase. Reducing the molybdenum content changes the effectiveness of the zirconium addition but this effect has not been evaluated. Vanadium additions behave in an unusual manner. In the quenched alloys, additions to 1.0

wt % V have no effect, but the addition of 1.5 wt % V to the 6 wt % Mo-U alloy or 2 wt % V to the 8 wt % Mo-U alloy causes a decrease in the hardness. On furnace cooling, additions of 1 wt % V or more increase the hardness. This ternary is very unusual, and should be studied in detail. This system, along with the U-8 wt % Mo-(0-3) wt % Zr are influenced markedly by heat treatment. Furthermore, some of these alloys have some ductility as determined by room temperature rolling of the arc-cast buttons.

This survey shows that the 8 wt % Mo-U alloy can be strengthened by alloying with silicon, zirconium, vanadium, or a combination of these. However, heat treatment will be necessary to produce an alloy having the proper combination of high strength and adequate ductility. For successful heat treatment, the isothermal transformation diagrams and quenching and aging curves will be required along with x-ray phase identification and microstructural studies to understand the correlation between properties and structure.

We recommend that these alloys be included with those already being studied by CMB-6 as part of a long-range program.

VIII. ITEMS WHICH MERIT CONSIDERATION

1. The commercial composite "Anviloy" with the ruthenium addition seems to possess good strength, high density, good ductility, is available in large sizes, and is weldable. The properties of this composite should be evaluated for the particular use intended. It is a commercial product of reasonably low cost (\$10 to \$15/pound), and seems to meet the strength and ductility requirements.

2. The alloy 8 wt % Mo-U could be used as a matrix for a composite. About 20 to 40 volume % tungsten or tungsten alloy wire reinforcement could be added. Such a composite should be relatively simple to produce, have high strength, reasonable elastic modulus, and good ductility.

3. Experimental studies designed to produce a composite of tungsten with a matrix of copper, silver, gold, tin, alloys of these, and perhaps other metals should be initiated to produce a product having the strength and density to

satisfy the requirements stated earlier. Composites are the materials of the future, not only for this particular use but for many others, and basic studies in this entire area should have a high priority.

4. Several of the commercially available tantalum alloys should be studied and mechanical working procedures tried both at room temperature and at elevated temperatures to achieve the desired strength. At the same time, shock loading and explosive forming operations should be carried out. These materials are inherently ductile and need only be strengthened.

5. The low-temperature properties, particularly, and the mechanical properties in general of the W-35 wt % Re-18 wt % Mo alloy should be studied. This alloy is available commercially, has good mechanical properties and supposedly good ductility at low temperatures (this has not been established beyond doubt). Literature studies indicate that the rhenium content cannot be reduced so it is expected that the price would remain high, probably around \$500/pound of alloy.

6. Other alloying additions to tantalum should be studied, and additions in greater amounts than have been evaluated so far. The low-temperature ductility makes it an ideal base material, and it should be considered seriously simply on that basis.

7. If cost is of no concern, then the use of pure rhenium or very low alloys of rhenium should be considered. These cannot be formed warm, but can be deformed to about 30 % reduction in cross-sectional area at room temperature - perhaps mechanically or perhaps by explosive forming. Deformations of about 10% will give adequate strength, and low-temperature ductility is not a problem.

8. Again, if cost is of no concern, then platinum base alloys could be studied. The available data are limited, but additions of tungsten, nickel, ruthenium, and molybdenum increase the strength rapidly. Platinum is basically ductile and shows promise in this respect.

9. The additions of rhenium to tungsten produces a more ductile product. However, effective additions are from 25 to 28 wt %, and this barely reduces the ductile-brittle

transition temperature to room temperature. The alloy is expensive, difficult to fabricate because of its high strength at elevated temperatures, and is generally difficult to produce by powder metallurgy methods while at the same time retaining good ductility. We are pessimistic about the future success of this alloy where low temperature ductility is required.

10. The factors responsible for embrittling tungsten should be studied. Many years have already been spent in this study, and if any degree of success is to be hoped for, then good quantitative measurements for interstitial impurities will be necessary. Also, the use of Auger spectroscopy and good transmission electron microscopy will be necessary.

11. The effect of additions of silicon, vanadium, zirconium, or combinations of these to the 8 wt % Mo-U alloy should be studied. Heat treatment may be a prerequisite to achieving the desired combination of properties, thereby making it necessary to study the isothermal transformation behavior and the effect of aging after quenching.

IX. RECOMMENDATIONS

1. The evaluation of the commercial high-density, high-strength composites. These materials have adequate strength, reasonably high density (17.25 g/cm^3), and good ductility at room temperature. These properties should be evaluated not only at room temperature but also at about -40°C . The density of this type composite could probably be increased by replacing the small amount of the nickel-rich matrix phase by a platinum-rich one without serious cost increase. This approach should be taken if the increased density justifies it.

2. A study should be initiated on producing high-density composites. These should probably contain tungsten wire reinforcement with platinum, gold, copper, silver, or other metals or alloys as the matrix. The matrix materials should have good ductility, high density, and a very low surface tension in the liquid state.

3. The ternary alloys of Mo-W-Re should be evaluated to determine if they have adequate strength at room temperature, and adequate low-temperature ductility.

4. The fiber reinforcement of the 8 wt % Mo-U alloy with tungsten wire, or preferably with a controlled density wire from a tungsten alloy, should be explored. We do not have facilities for producing uranium alloy castings, so we would recommend that such a program be discussed with CMB-6. They have the casting facilities, are pioneers in producing uranium alloys, and possess the technological ability to ensure success in such a program.

5. Alloying additions of silicon, vanadium, zirconium, or combinations of these should be evaluated as strengtheners of the base 8 wt % Mo-U alloy.

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