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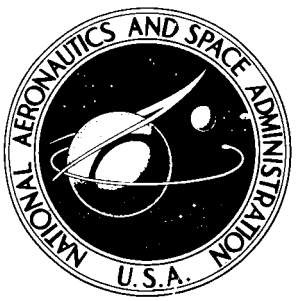

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MECHANICAL PROPERTIES OF A TUNGSTEN – 23.4-PERCENT-RHENIUM – 0.27-PERCENT-HAFNIUM – CARBON ALLOY

by William D. Klopp and Walter R. Witzke

Lewis Research Center
Cleveland, Ohio 44135

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SUMMARY

The objective of this study was to determine if particle strengthening by a hafnium carbide precipitate could be obtained in a ductile tungsten - high-rhenium alloy. A tungsten - 23.4-rhenium - 0.27-hafnium - carbon alloy (composition in at. %) was consolidated by arc melting, fabricated to sheet and rod, and evaluated by tensile, creep, and bend testing. Carbon content of test specimens ranged from 0.12 to 1.12 atomic percent. The alloy exhibited good high-temperature strength properties. For example, at 3000° F (1649° C), the hafnium carbide strengthened alloy had a short-time tensile strength of 62.7 ksi (432 MN/m^2), more than double the strength of 28.1 ksi (194 MN/m^2) observed earlier for tungsten - 24-rhenium. The strength advantage decreases at higher temperatures because of rapid hafnium carbide particle coarsening and grain-boundary sliding.

The alloy exhibited good tensile and bend ductility at room temperature.

It is concluded that the tungsten - high-rhenium alloy with a fine hafnium carbide precipitate has an attractive combination of low-temperature ductility and high strength to temperatures of about 3000° F (1649° C).

INTRODUCTION

Recent studies on both tungsten (refs. 1 to 3) and molybdenum (refs. 4 and 5) have shown that hafnium carbide (HfC) is a very effective precipitate strengthener at elevated temperatures. Tensile strength improvements up to eightfold have been reported at about 0.6 of the melting point for both metals. This strengthening effect is obtainable not only in matrices of unalloyed tungsten and molybdenum but also in an alloy of tungsten and rhenium (W-4Re). (All compositions are in atomic percent.) It has been further shown that the low-temperature ductility, an area of concern with tungsten and molybdenum, is not adversely affected by precipitate strengthening with HfC (ref. 3).

In view of the excellent strengthening ability of HfC, it was considered desirable to determine the effects of HfC on the mechanical properties of the ductile W-26Re alloy. This alloy has been shown to be ductile in slow bending at -150° F (-101° C) in the worked condition, but has low creep strength at elevated temperatures (ref. 6). It would obviously be advantageous to combine the high-temperature strengthening from a fine HfC precipitate with the substantially improved low-temperature ductility effected by alloying with 26 percent Re.

The present study was undertaken to evaluate briefly the potential for combining HfC strengthening of the tungsten with high-rhenium ductilizing. An ingot of the desired alloy was arc-melted, fabricated, and evaluated. Tensile and creep tests were conducted at 2500° to 3500° F (1371° to 1927° C) to determine high-temperature strength properties, while low-temperature ductility was determined primarily by bend tests. These evaluations were supplemented by optical and electron transmission metallography.

EXPERIMENTAL PROCEDURE

Fabrication

Starting materials for this study consisted of commercially pure -325 mesh tungsten and hafnium powders, -200 mesh rhenium powder, and spectroscopically pure -325 mesh carbon (graphite) powder. The powders were blended to the charge composition indicated in table I, hydrostatically pressed into a 1.125-inch- (2.86-cm-) diameter electrode, sintered, and vacuum arc-melted into a 2.5-inch- (6.4-cm-) diameter ingot. The ingot was machined into a 2.25-inch- (5.7-cm-) diameter billet, encapsulated in a -.375-inch- (0.95-cm-) thick sintered molybdenum can, and extruded at 4000° F (2204° C) to an 8:1 reduction ratio.

A portion of the 1-inch- (2.54-cm-) diameter clad extrusion was swaged at 3200° F (1760° C) to 0.29-inch- (0.74-cm-) diameter rod, a reduction of 93 percent. The cladding was removed during specimen machining. The remainder of the extrusion was hot-rolled at 3200° to 3000° F (1760° to 1649° C) to 0.046-inch- (0.12-cm-) thick strip, the cladding being chemically removed at a thickness of about 0.2 inch (0.51 cm).

Analysis of the swaged alloy is shown in table I and indicates that significant reductions in alloying element contents occurred during processing, primarily during melting. Carbon analyses of the front and tail of the portion of the extrusion used for mechanical property evaluations were 1.21 and 1.07 atomic percent, respectively.

Evaluation

Tensile and creep tests were performed on specimens having a 1-inch (2.5-cm) gage length and a reduced diameter of 0.12 inch (0.3 cm). Tensile testing was conducted at a pressure below 10^{-5} torr (10^{-3} N/m^2) in a water-cooled vacuum chamber equipped with a resistance-heated tantalum sleeve heater. Specimens were strained at a constant crosshead speed of 0.05 inch per minute (2.1×10^{-5} m/sec). Test temperatures were measured by W/W-26Re thermocouples attached to the reduced section of the specimen. Step-load creep tests were conducted at pressures below 10^{-5} torr (10^{-3} N/m^2) in a conventional water-cooled beam-loaded machine equipped with a tungsten mesh heater. Strain measurements were made from loading rod movement. For tests at 3500° F (1927° C), both tensile and creep specimens were wrapped in tungsten foil to minimize compositional changes during testing.

Transmission microscopy observations were made on disks cut transversely from the reduced sections of specimens after creep and tensile testing. These disks were thinned for observations by electropolishing in a hydrogen peroxide - sodium hydroxide solution.

Longitudinal sheet specimens were bend tested following grinding and electropolishing to remove about 0.005 inch (0.012 cm) of material per side. Final specimen dimensions were 1 by 0.3 by 0.03 inch (2.5 by 0.8 by 0.08 cm). The bend ductile-brittle transition temperature is defined here as the lowest temperature at which a sheet specimen could be bent through 90° over a bend radius four times the specimen thickness without failure. Three-point loading was used with a ram speed of 1 inch per minute (2.5 cm/min).

Annealing at temperatures up to 3200° F (1759° C) was conducted in an induction-heated hydrogen-atmosphere furnace. Annealing treatments requiring temperatures from 3400° to 4900° F (1871° to 2704° C) were conducted in a resistance-heated vacuum furnace at a pressure of 10^{-5} torr (10^{-3} N/m^2). Tensile and creep specimens annealed at 3600° and 4900° F (1982° and 2704° C) were wrapped in tungsten foil and placed in a loosely covered tungsten crucible containing coarse HfC powder to minimize carbon losses during annealing. Even with these precautions, substantial reductions in carbon content were observed. These carbon losses are attributed to reaction with trace amounts of oxygen in the vacuum annealing environment.

RESULTS AND DISCUSSION

Metallography and Annealing Response

The cast structure, shown in figure 1(a), contains a considerable amount of inter-

granular second-phase (white) and shrinkage voids (dark). Since the rhenium content of 23.4 percent is less than the solubility limit of about 28 percent, the second phase is assumed to be a mixture of carbides, probably including W_2C, (W, Hf)C, and HfC. The cast diamond pyramid hardness was 488, slightly higher than the hardness of 417 observed for an electron-beam-melted ingot of W-24Re (ref. 6).

The annealing response of this W-Re-Hf-C alloy after swaging was evaluated by heating for 1 hour at temperatures ranging from 2400° to 4600° F (1316° to 2538° C). Representative microstructures after annealing are shown in figures 1(b), (c), and (d). Hardness and grain size data are presented in table II and figures 2 and 3.

Some recovery occurs in the temperature range 2400° to 2600° F (1316° to 1427° C), as evidenced by a slight drop in hardness from that of the as-swaged material. The structure after annealing at 2400° F (1316° C) is still highly worked, as shown in figure 1(b). Visual observations indicated that recrystallization begins between 2600° and 2800° F (1427° and 1538° C) and is complete after annealing at 3200° F (1760° C). Microstructurally the recrystallization temperature is estimated at 3100° F (1704° C). In comparison, an electron-beam-melted alloy of W-24Re was fully recrystallized after annealing for 1 hour at 2900° F (1593° C) (ref. 6). The higher recrystallization temperature of the W-Re-Hf-C alloy suggests reduced grain-boundary mobility, such as might be caused by the presence of fine carbide particles.

Very little grain growth occurred during annealing at temperatures up to 4000° F (2204° C), as indicated in figure 3. This behavior indicates that carbide particles are fine enough to pin grain boundaries at least up to 4000° F (2204° C). At 4200° to 4600° F (2316° to 2538° C), "secondary recrystallization" resulted in a mixture of large grains and small grains, as shown in figure 1(d). In contrast, grain boundaries in W-Hf-C and W-4Re-Hf-C alloys were immobile until heated to the carbide solutioning temperature, approximately 4600° F (2538° C) (refs. 2 and 3). Thus, carbide particles in the present alloy are less effective in stabilizing a worked structure and preventing grain growth than those in a tungsten or W-4Re matrix. Although the decreased effectiveness of the particles in the high-rhenium alloy appears to be related to the higher rhenium content, it is not clear whether this is due to an increase in carbon solubility, an increase in diffusivities leading to higher coarsening rates, or other factors.

Microprobe traverses of the specimen annealed at 4400° F (2427° C) showed that the second phase particles were substantially higher in carbon and hafnium but lower in rhenium than the matrix, indicating that these are hafnium-rich carbide particles.

Tensile properties. - Tensile properties of W-23.4Re-0.27Hf-C are given in table III and are compared with those of several other tungsten materials in figure 4. Note that the actual carbon contents of the specimens after testing are reported, since carbon loss during testing was observed.

At room temperature, the as-swaged alloy exhibits a very high tensile strength,

289 ksi (1990 MN/m^2), and an elongation of 8 percent. These values compare with 236 ksi (1627 MN/m^2) and 11 percent, respectively, exhibited by sheet of W-24Re (ref. 6) and indicate that the W-23.4Re-Hf-C alloy combines the relatively good room-temperature ductility characteristic of W-24Re with an unusually high tensile strength.

At 2000° and 2500° F (1093° and 1371° C), the swaged W-23.4Re-0.27Hf-C alloy retains its high strength and is significantly stronger than the swaged W-4Re-0.41Hf-C alloy, as shown in figure 4. The strength decreases fairly rapidly at higher temperatures, however, and at 3500° F (1927° C), the W-23.4Re-Hf-C alloy has only a modest strength advantage over a binary W-24Re alloy.

Annealing at 3600° F (1982° C) reduced the tensile strengths at 2500° and 3000° F (1371° and 1649° C) as compared to the swaged alloy. After solution annealing at 4900° F (2704° C) and aging at 2500° F (1371° C), the W-23.4Re-Hf-C alloy had a strength at 3500° F (1927° C) of 48 ksi (331 MN/m^2), substantially higher than in the swaged or 3600° F (1982° C) annealed condition and approaching that of the W-4Re-Hf-C alloy.

Electron transmission micrographs of several specimens after tensile testing are shown in figure 5. The structure of the as-swaged specimen tested at room temperature, shown in figure 5(a), contains a visible subgrain network and a high density of dislocations which appear to be pinned by fine second-phase particles, presumably HfC. This structure is very similar to that of swaged W-Hf-C and W-4Re-Hf-C alloys described previously (refs. 2 and 3). The particle diameters could not be accurately measured in this specimen because of the high dislocation density but appear to range from about 100 to 500 Å (100×10^{-10} to 500×10^{-10} m).

The swaged specimen tested at 3500° F (1927° C) shows a distinctly different structure from that of the room-temperature tensile specimen. As shown in figure 5(b), no subgrain boundaries are visible, the dislocation density is very low, and the HfC particles are much larger, with a median diameter of 700 Å (700×10^{-10} m). Assuming an exposure time of 1 hour at 3500° F (1927° C), this particle size indicates a growth rate of the order of 10^{-20} cubic centimeter per second, a rate much more rapid than the average particle growth rate of 3×10^{-22} cubic centimeter per second observed earlier for W-Hf-C and W-4Re-Hf-C alloys (ref. 3). The increased size of the HfC particles in this W-23.4Re matrix as compared to the W and W-4Re matrices decreases their ability to pin dislocations and may be the cause of the low strength at 3500° F (1927° C) observed for the W-23.4Re-Hf-C alloy.

An interesting portion of the structure of the solution-annealed and aged specimen after tensile testing at 3500° F (1927° C) is shown in figure 5(c). The Widmanstatten-type pattern of particle alinement is similar to that observed by Perkins and Lytton (ref. 7) in a molybdenum-columbium-titanium-zirconium-carbon alloy and is attributed to precipitation during the 2500° F (1371° C) aging treatment. A meaningful particle-size

measurement could not be obtained because of the elongated particle shape and the presence of numerous fine particles, less than 100 Å (100×10^{-10} m) diameter. However, the platelet geometry of the carbide in this solution-annealed and aged specimen probably contributed to its higher strength relative to the as-swaged specimen.

An interesting aspect of the high-temperature deformation of the W-23.4Re-Hf-C alloy is the unusually high tensile elongations exhibited by the as-swaged and 3600° F (1982° C) annealed specimens at 3500° F (1927° C) (92 and 105 percent, respectively), shown in figure 6. The microstructure near the fracture point of the annealed and tensile-tested specimen is fine-grained and essentially equiaxed, as shown in figure 7. These observations suggest that the presence of hafnium and carbon does not change the tendency toward superplasticity previously observed for binary W-Re alloys (ref. 8). The absence of high elongation in the large-grained specimens annealed at 4900° F (2704° C) is also consistent with superplasticity, which is attributed to grain boundary sliding in fine-grained materials (ref. 9).

Creep properties. - Five step-load creep tests were performed on W-23.4Re-0.27Hf-C in the swaged, recrystallized, and solution-annealed conditions. The results are presented in table IV and figure 8.

The creep strengths in the swaged condition are substantially higher at 2500° and 3000° F (1371° and 1649° C) than at 3500° F (1927° C), as would be expected. However, at 3500° F (1927° C), the swaged and solution-annealed and aged materials showed little strength advantage over the 3600° F (1982° C) annealed material, suggesting that the HfC dispersion is a relatively ineffective strengthener during prolonged exposure at this temperature.

The creep strength of swaged W-23.4Re-Hf-C is compared to those for several other high-temperature materials in figure 9. At 2500° F (1371° C), the W-23.4Re-Hf-C alloy is weaker in creep than W-4Re-Hf-C, but about three times stronger than the tantalum alloys Ta-10W, T-111, and T-222. The strength advantage over tantalum alloys is reduced to about twofold at 3000° F (1649° C), while at 3500° F (1927° C) W-23.4Re-Hf-C has about the same creep strength as arc-melted W-26Re.

A transmission micrograph of the structure of the swaged specimen after creep testing at 3500° F (1927° C) is shown in figure 10. The structure is similar to that of the tensile specimen shown in figure 5(b), with fewer dislocations and larger HfC particles. The median particle diameter of 960 Å (960×10^{-10} m) indicates a particle growth rate of about 6×10^{-21} cubic centimeter per second, a rate about twentyfold faster than that for W-Hf-C and W-4Re-Hf-C alloys.

Low-temperature ductility. - A limited evaluation was conducted to determine the ductile-brittle bend transition temperature. The results of this evaluation, given in table V, show that the W-23.4Re-0.27Hf-C alloy is ductile (4t bend) to -25° and 30° F (-31° and -1° C) in the as-rolled and 3600° F (1982° C) annealed conditions, respectively.

This high-rhenium alloy is ductile to lower temperatures than the W-4Re-Hf-C alloys, but has a higher DBTT than W-26Re in the as-rolled condition.

The W-23.4Re-Hf-C alloy thus appears to be a good compromise alloy for use up to about 3000° F (1649° C) because it has better low-temperature ductility than W-4Re-Hf-C and better high-temperature strength than W-26Re. This Hf-C strengthened W - high-Re alloy may be useful in such applications as high-strength fibers for fiber-strengthened composites.

Strengthening Mechanisms

The mechanisms of carbide particle strengthening in the W - high-Re matrix appear similar to those postulated for W-Hf-C and W-4Re-Hf-C, at least for temperatures to 3000° F (1649° C). In the worked condition, indirect particle strengthening likely predominates, with carbide particles stabilizing the very fine grain size which in turn limits the distance that mobile dislocations can move before being immobilized at grain boundaries. Direct particle strengthening, that is, immobilization of dislocations by direct interaction with a particle, probably also occurs in the worked material but to a much lesser extent than indirect strengthening.

In contrast, the solution annealed and aged alloy derives its strength advantage over the matrix almost entirely from direct particle-dislocation interactions, since the grain size is so large that boundaries are infrequent barriers to mobile dislocations.

However, at 3500° F (1927° C) and probably also at higher temperatures, worked W-23.4Re-0.27Hf-C has only a slight strength advantage over the solid solution alloy W-26Re. This decrease in strength appears related to two factors: (1) very rapid Hf-C particle coarsening rates at elevated temperatures, and (2) a strong tendency towards deformation by grain-boundary sliding in fine-grained structures (superplasticity).

Although coarsening rates could be estimated for only three specimens at two temperatures, 3500° and 3600° F (1927° and 1982° C), these rates were consistent in that they were 20 to 30 times faster than rates for W-Hf-C and W-4Re-Hf-C at the same temperatures. These more rapid coarsening rates in the high-Re quaternary alloy indicate strength retention for correspondingly shorter time periods, since the ability of particles to strengthen by pinning mobile dislocations is inversely related to the particle size. The tensile strength value of 46.4 ksi at 3500° F (1927° C) for a solution-annealed and aged specimen does indicate that short-time strengthening is possible under optimum conditions, but the absence of strengthening during creep confirms the rapid loss of strength as the particles coarsen.

In addition to strength loss from particle coarsening, the W-23.4Re-Hf-C alloy is weakened at temperatures above 3000° F (1649° C) by grain-boundary sliding, partic-

ularly in the fine-grained condition. This behavior is in contrast to W-HfC and W-4Re-Hf-C alloys, where a fine-grained structure, such as a stabilized warm-worked structure, is quite strengthening. Grain-boundary sliding apparently does not become the prime deformation mode in these materials until temperatures of the order of 0.9 T_m are reached (ref. 10), that is, temperatures much higher than those to which HfC particles are stable.

However, as alloying levels are increased, grain-boundary sliding becomes prominent at lower temperatures. Under these conditions, a fine-grained structure, such as one stabilized by second-phase HfC particles, will be weaker than a coarse-grained structure. The observation of high tensile elongations in several of the specimens in the present study is interpreted as a tendency towards superplasticity and grain-boundary sliding.

Thus, the stabilization of a fine-grained structure, which is strengthening in W-Hf-C and W-4Re-Hf-C alloys, is probably weakening in the present W-23.4Re-Hf-C alloy at 3500° F (1927° C).

CONCLUSIONS

The following conclusions are drawn from a study of mechanical properties of a tungsten-rhenium-hafnium-carbon alloy (W-23.4Re-0.27Hf-C):

1. HfC precipitate strengthening is effective in a W - high-Re matrix in both tensile and creep at temperatures up to about 3000° F (1649° C). At this temperature the alloy has a tensile strength of 62.7 ksi (432 MN/m^2) and a creep strength of 12 ksi (83 MN/m^2) for a creep rate of 10^{-6} per second.

2. HfC precipitate strengthening is not effective during creep at 3500° F (1927° C), as indicated by creep properties similar to those of the solid solution alloy W-26Re. This lack of strengthening is apparently due to rapid particle coarsening at this temperature and/or large amounts of deformation by grain-boundary sliding.

3. The alloy maintains a low ductile-brittle bend (4t) transition temperature after annealing at 3600° F (1982° C), namely, 30° F (-1° C) compared with 350° F (177° C) for arc-melted W-26Re.

4. The W-high-Re-Hf-C system has an attractive combination of low-temperature ductility and high strength to temperatures of about 3000° F (1649° C). This type of alloy might be useful in applications such as high-strength fibers in fiber-strengthened composites.

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TABLE I. - COMPOSITION OF CHARGE AND OF SWAGED ALLOY

[Vickers hardness of swaged alloy, 673.]

Element	Composition, at. %	
	Charge	Melted, extruded, and swaged[a]
Rhenium	26	23.4
Hafnium	.4	.27
Carbon	2.4	1.14
Oxygen	----	[b].0057
Nitrogen	----	[b].0022

[a]Rhenium and hafnium were determined by wet chemical analyses, carbon by combustion, oxygen by vacuum fusion, and nitrogen by micro - Kjeldahl.

[b]Oxygen and nitrogen contents equal to 5 and 1.7 wt. ppm, respectively.

TABLE II. - ANNEALING RESPONSE OF W-23.4Re-Hf-C

Annealing temperature		Fraction recrystallized	Average grain diameter, cm	Vickers hardness
°F	°C			
As swaged			------	673
2400	1316	0	------	627
2600	1427	0	------	642
2800	1538	.3	------	579
3000	1649	.8	0.0015	519
3200	1760	1.0	.0011	512
3400	1871	↓	.0012	525
3600	1982	↓	.0011	493
3800	2093	↓	.0013	493
4000	2204	↓	.0014	493
4200	2316	↓	[a].019	478
4400	2427	↓	[a].018	473
4600	2538	↓	[a].021	468

[a]Mixed grain sizes.

TABLE III. - TENSILE PROPERTIES OF W-23.4Re-Hf-C

Test temperature		Yield strength		Ultimate strength		Elongation, percent	Reduction in area, percent	Carbon content after test, at. %	Median particle diameter after test, Å (10^{-10} m)
^{o}F	^{o}C	ksi	MN/m^2	ksi	MN/m^2				
As swaged									
75	24	268	1850	289	1990	8	14	1.12	---
2000	1093	149	1030	172	1190	16	57	1.04	---
2500	1371	138	951	145	1000	23	65	1.00	---
3000	1649	56.9	392	62.7	432	39	84	.78	---
3500	1927	15.5	107	16.7	115	92	87	.81	700
Annealed 1 hr at 3600^{o} F (1982^{o} C)									
2500	1371	59.3	409	67.9	468	60	91	0.99	---
3000	1649	32.9	227	34.7	239	73	>95	.56	---
3500	1927	17.3	119	17.8	123	105	>95	.64	840
Annealed 10 min at 4900^{o} F (2704^{o} C) and 1 hr at 2500^{o} F (1371^{o} C)									
3500	1927	46.4	320	48.0	331	9	13	0.34	---

TABLE IV. - CREEP PROPERTIES OF W-23.4Re-Hf-C

Annealing conditions			Test temperature		Stress		Minimum creep rate, sec^{-1}	Carbon content after test, at. %	Median particle diameter after test, Å (10^{-10} m)
Time, hr	Temperature		°F	°C	ksi	MN/m^2			
	°F	°C							
As swaged			2500	1371	40	276	0.099×10^{-6}	1.01	---
					43	296	.28		
					46	317	.58		
					50	345	1.1		
					54	372	2.6		
As swaged			3000	1649	10	69	0.33×10^{-6}	0.67	---
					11	76	.52		
					12	83	.95		
					13	90	1.3		
					14	97	2.1		
					15	103	3.3		
As swaged			3500	1927	4.2	29	2.7×10^{-6}	0.83	960
					4.8	33	4.4		
					5.8	40	7.0		
					6.9	48	18		
					8.3	57	44		
1	3600	1982	3500	1927	2	14	0.37×10^{-6}	0.25	---
					2.5	17	.64		
					3	21	.91		
					4	28	1.7		
0.17	4900[a]	2704	3500	1927	4.5	31	0.62×10^{-6}	0.12	---
					5.5	35	5.9		
					6.5	45	15		
					8	55	65		

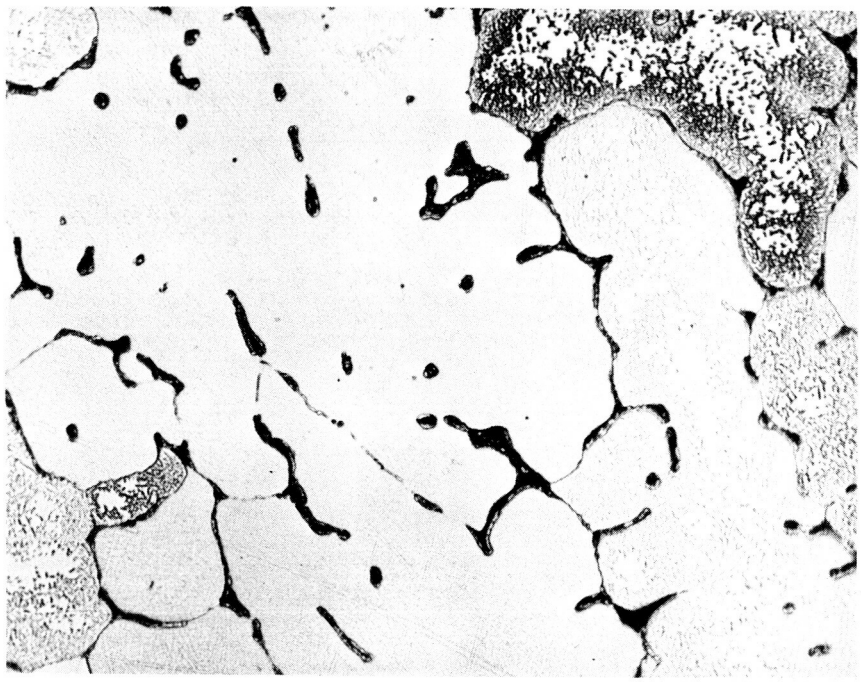
[a]Plus 1 hr at 2500° F (1371° C).

TABLE V. - DUCTILE-BRITTLE BEND TRANSITION TEMPERATURES FOR W-23.4Re-Hf-C, W-4Re-Hf-C, AND W-26Re

Material	Ductile-brittle bend transition-temperature				Reference
	As rolled		Annealed 1 hr at 3600° F (1982° C)		
	°F	°C	°F	°C	
W-23.4Re-Hf-C	-25	-31	30	-1	Present work
W-4Re-Hf-C[a]	200	93	[b]375	191	3
W-26Re	-150	-101	350	177	6

[a]Median values for 12 alloys.

[b]Annealed 1 hr at 3800° F (2093° C).

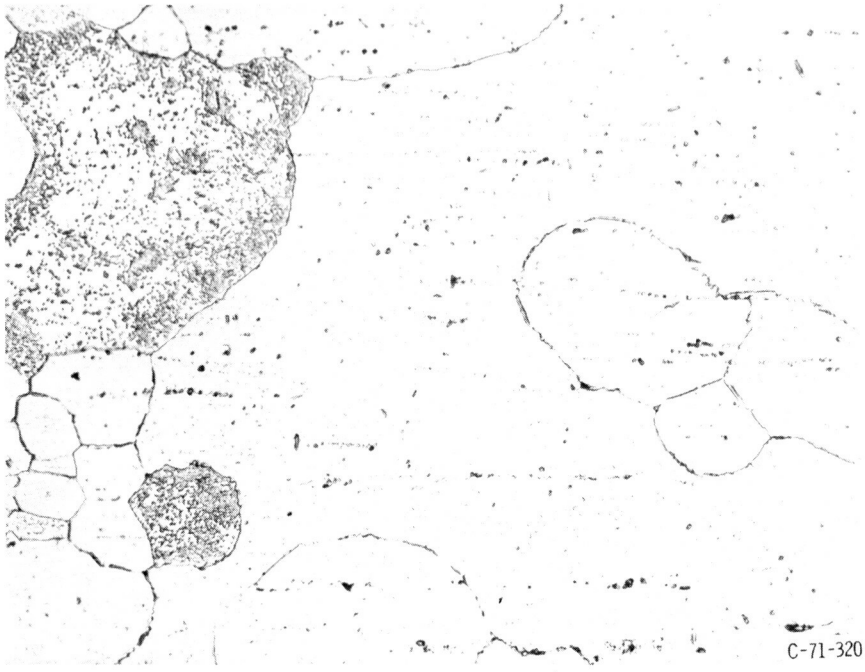


(a) As cast.

(b) Swaged to 93 percent reduction at 3200° F (1760° C) and annealed 1 hour at 2400° F (1316° C).

Figure 1. - Microstructure of W-23.4 Re-0.27 Hf-C. X250.

(c) Swaged to 93 percent reduction at 3200° F (1760° C) and annealed 1 hour at 3600° F (1982° C).



(d) Swaged to 93 percent reduction at 3200° F (1760° C) and annealed 1 hour at 4400° F (2427° C).

Figure 1. - Concluded.

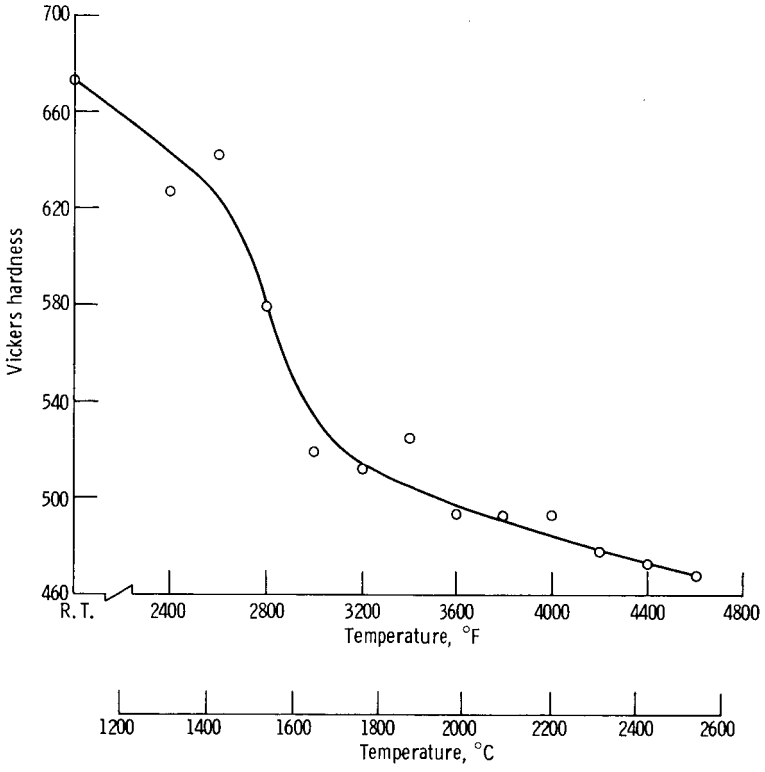


Figure 2. - Hardness after annealing for 1 hour at indicated temperature.

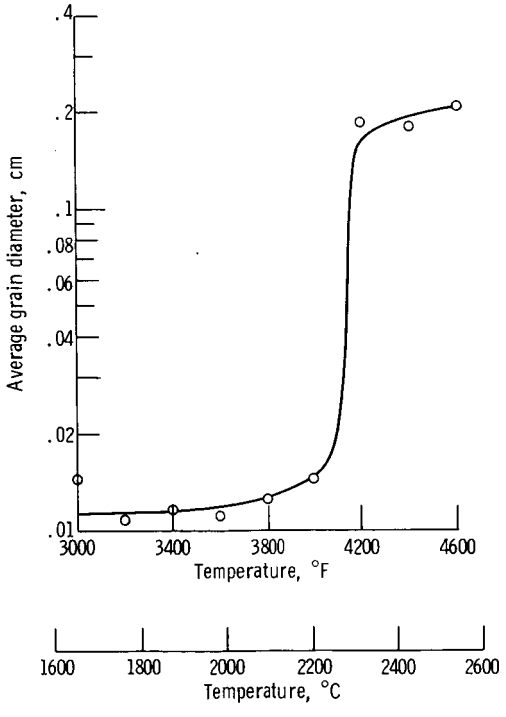


Figure 3. - Average grain diameter after annealing for 1 hour at indicated temperature.

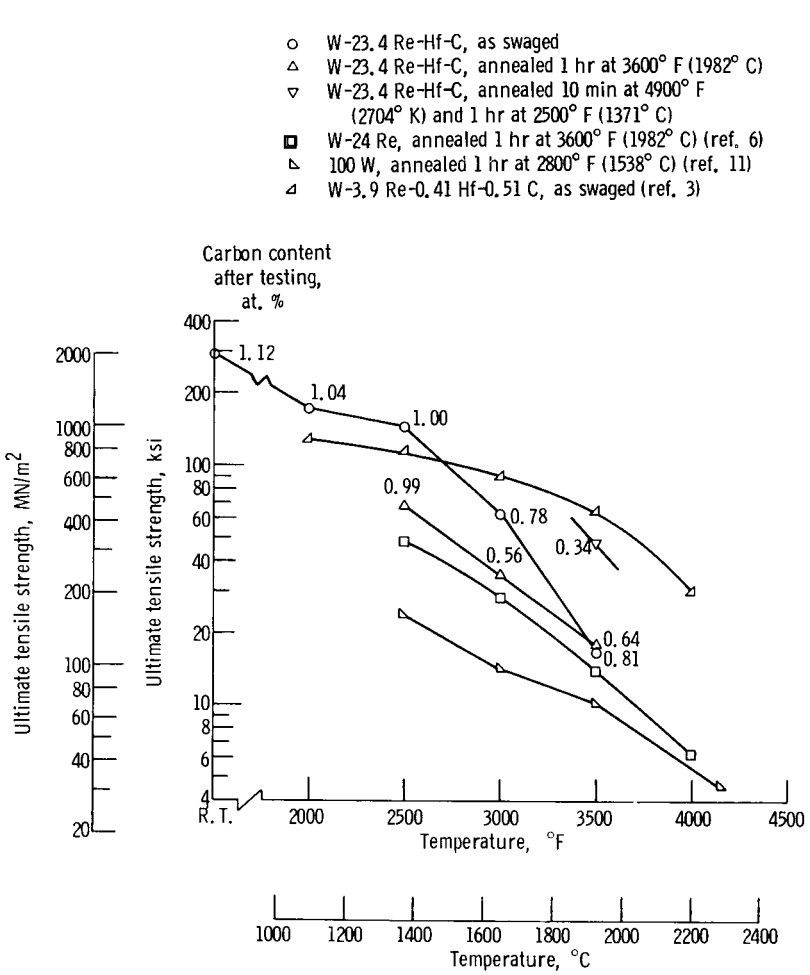
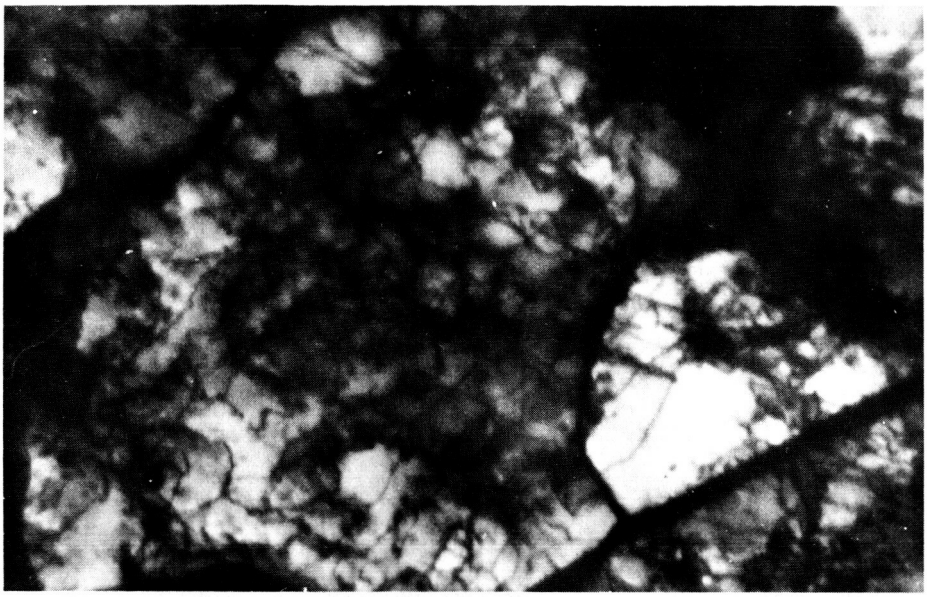
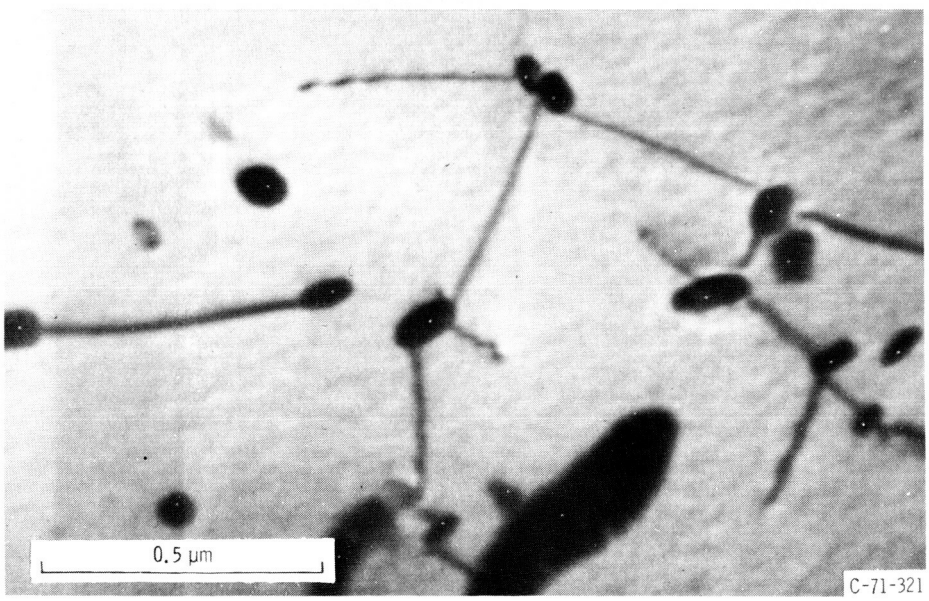


Figure 4. - Tensile strength of W-23.4 Re-Hf-C and other tungsten alloys at elevated temperatures.

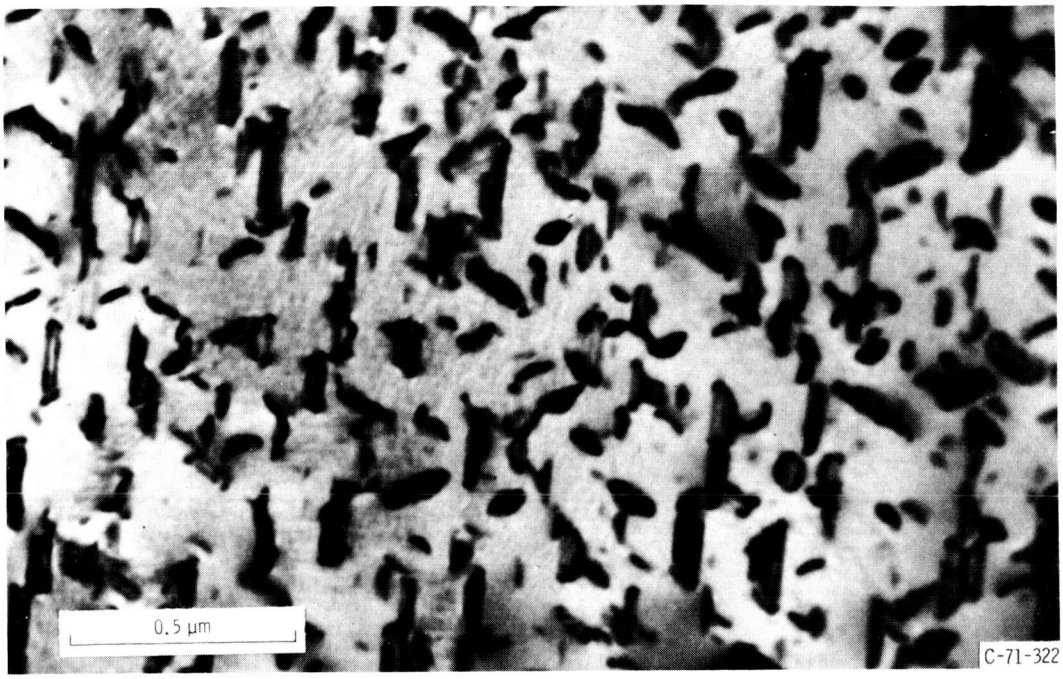


(a) As-swaged specimen tensile tested at room temperature.



(b) As-swaged specimen tensile tested at 3500° F (1927° C). Median particle diameter, 700 Å (700×10^{-10} m).

Figure 5. - Transmission micrographs of as-swaged and heat-treated specimens after tensile testing.



(c) Specimen annealed for 10 minutes at 4900° F (2704° C) and 1 hour at 2500° F (1371° C) and tensile tested at 3500° F (1927° C).

Figure 5. - Concluded.

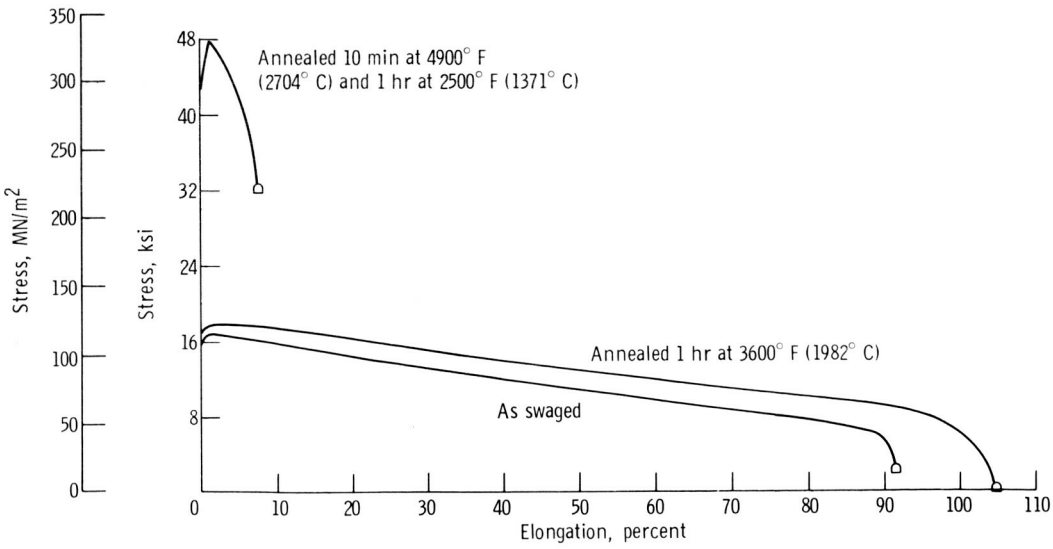


Figure 6. - Engineering stress-strain tensile curves for W-23.4 Re-Hf-C at 3500° F (1927° C) after indicated heat treatments.

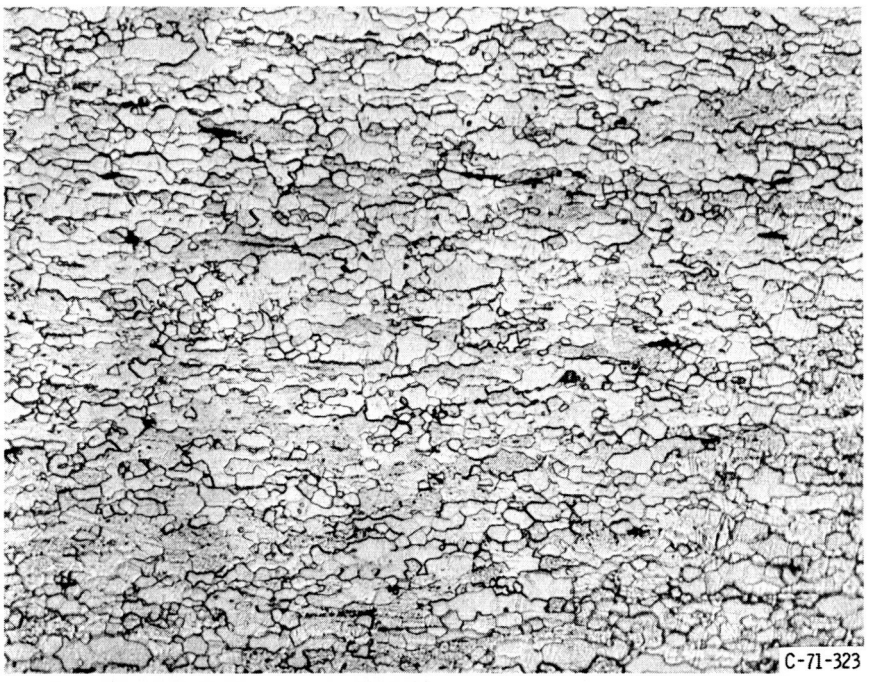


Figure 7. - Microstructure of fracture zone of specimen annealed at 3600° F (1982° C) after tensile testing at 3500° F (1927° C). X250.

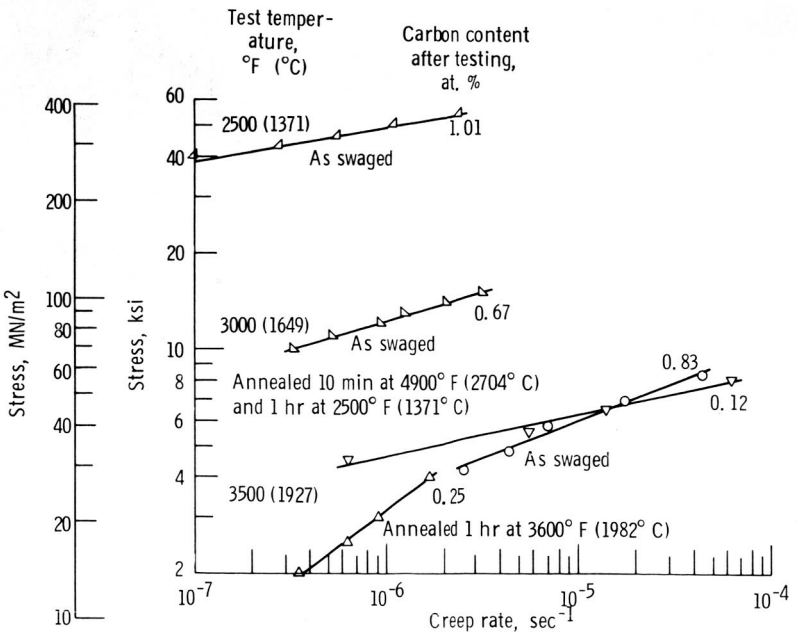


Figure 8. - Creep behavior of W-23.4 Re-Hf-C at 2500° to 3500° F (1371° to 1927° C). Carbon content of each specimen after step-load testing is indicated.

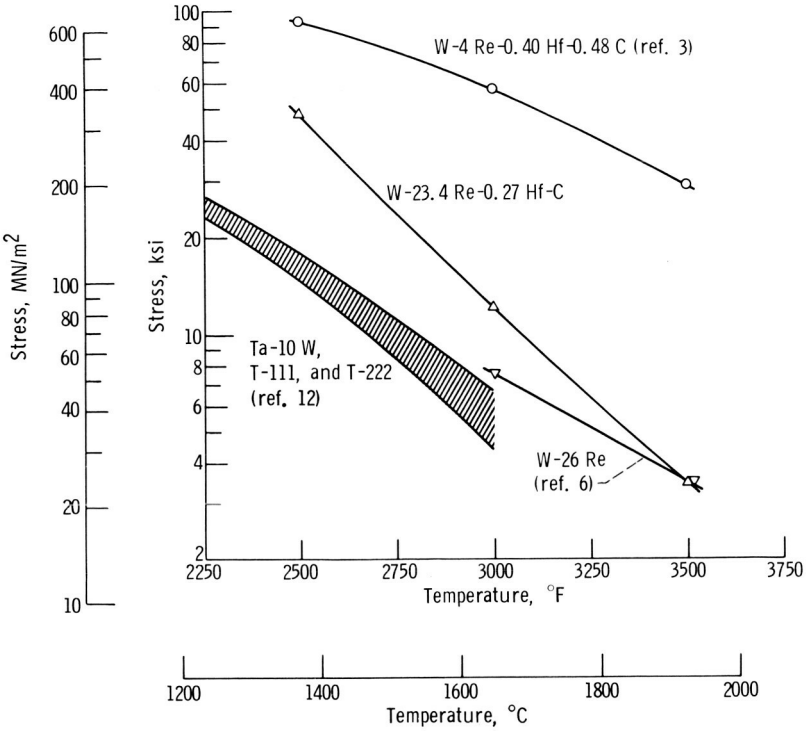


Figure 9. - Creep strength at creep rate of 10^{-6} per second for swaged W-23.4 Re-0.27 Hf-C compared to swaged W-4 Re-0.40 Hf-0.48 C and high-strength tantalum alloys at elevated temperatures.

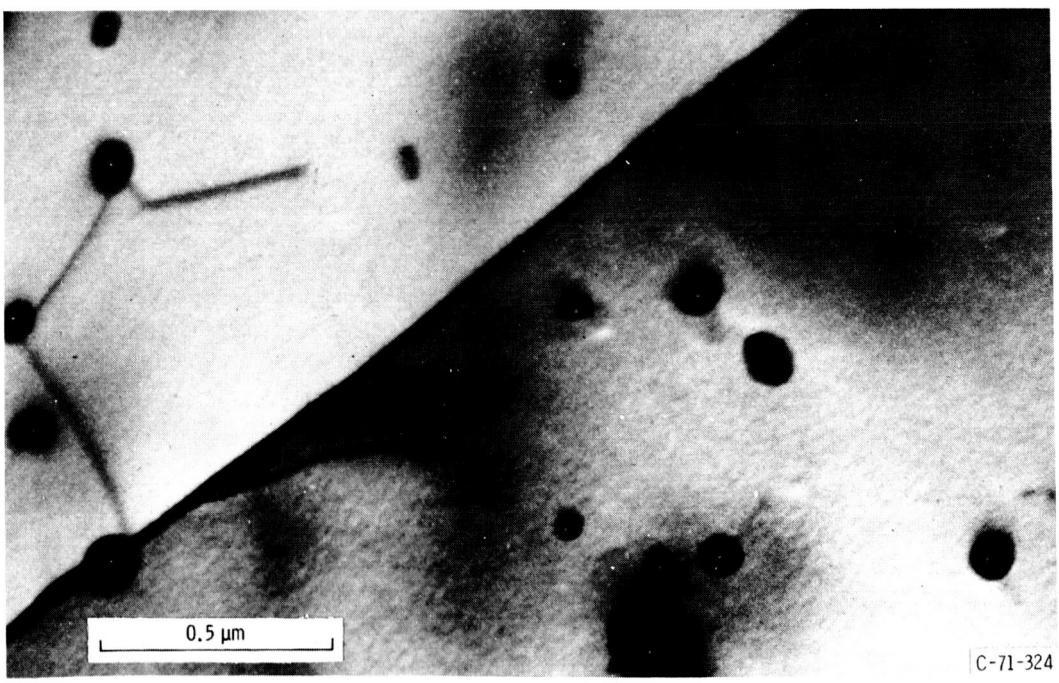


Figure 10. - Transmission micrograph of as-swaged specimen after creep testing at 3500° F (1927° C). Median particle diameter, 960 Å (960×10^{-10} m).

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15. Supplementary Notes

16. Abstract

An alloy of tungsten - 23.4-rhenium - 0.27-hafnium - carbon (composition in at. %) was arc-melted, fabricated, and evaluated to determine if particle strengthening by hafnium carbide could be achieved in a ductile, tungsten - high-rhenium matrix. The alloy was found to exhibit a good combination of tensile and creep strength to 3000° F (1649° C) with a ductile-brittle transition temperature near room temperature.

17. Key Words (Suggested by Author(s))	18. Distribution Statement
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