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THE EFFECT OF SMALL ADDITIONS OF SILICON, IRON, AND ALUMINUM ON THE ROOM-TEMPERATURE TENSILE PROPERTIES OF HIGH-PURITY URANIUM

R. L. Ludwig

November 1983

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Printed in the United States of America. Available from
National Technical Information Service
U.S. Department of Commerce
5285 Port Royal Road, Springfield, Virginia 22161
NTIS price codes—Printed Copy: A03 Microfiche A01

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Date of Issue: **November 14, 1983**
Distribution Category: **UC-25**

Y--2286

DE84 004979

THE EFFECT OF SMALL ADDITIONS OF SILICON, IRON, AND ALUMINUM ON THE ROOM-TEMPERATURE TENSILE PROPERTIES OF HIGH-PURITY URANIUM

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Prepared for the U.S. Department of Energy
under Contract No. W-7405-eng-26

ABSTRACT

Eleven binary and ternary alloys of uranium and very low concentrations of iron, silicon, and aluminum were prepared and tested for room-temperature tensile properties after various heat treatments. A yield strength approximately double that of high-purity derby uranium was obtained from a uranium-400 ppm silicon-200 ppm iron (U-400 ppm Si-200 ppm Fe) alloy after beta solution treatment and alpha aging. Higher silicon plus iron alloy contents resulted in increased yield strength, but showed an unacceptable loss of ductility.

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SUMMARY

Experience gained in the 1950s and 1960s during the development of metallic uranium reactor fuel elements indicated that a considerable degree of elevated temperature strength was imparted by small additions of iron, silicon, and aluminum. With other possible applications in mind, work was initiated to determine if the room-temperature properties would be similarly improved. Eleven alloys were prepared in the form of 12-mm-thick wrought plate, and tensile bar blanks were beta (730°C) and gamma (800°C) solution annealed in vacuum and water-quenched. Others were similarly treated and alpha aged (350°-450°C) while a third group was alpha annealed (450°-630°C). In all cases, high-purity derby uranium control specimens were treated simultaneously.

A yield strength approximately double that of high-purity uranium was attained in a uranium-400 ppm silicon-200 ppm iron (U-400 ppm Si-200 ppm Fe) alloy after beta solution treatment and alpha aging. There was some concurrent loss of ductility after aging that also occurred in the high-purity uranium; however, the uranium had not shown the increase in yield strength. Alloys that were higher in silicon and iron had higher yield strengths but showed an unacceptably low level of ductility. The grain size was shown to be refined, and the grain-boundary smoothness was affected by the presence of alloying elements in the beta- and gamma-quenched alloys.

Aluminum additions (to the 250-ppm-maximum amount alloyed) appeared to have no effect on tensile properties. Alpha annealing the wrought alloys overaged them such that they were somewhat strengthened by dispersion hardening, but were slightly stronger and nearly as ductile as similarly treated high-purity uranium. The presence of the alloying elements was shown to reduce the rate of recrystallization and grain growth in alpha-annealed materials.

The detrimental effect of hydrogen on ductility and tensile strength was demonstrated in the course of the testing. The results indicated that the hydrogen level should be no more than 0.05 ppm to eliminate its influence on properties.

INTRODUCTION

For certain applications, the tensile properties of relatively high-purity (derby) uranium are marginal or too low. Experience with less-pure uranium has shown that low concentrations of silicon and iron impurities that are normally present in uranium will strengthen uranium at room temperature by supersaturated solid solution or dispersion hardening. At elevated temperatures, aluminum reportedly imparts a substantial degree of strengthening by precipitation hardening treatments. Work performed in the 1950s and 1960s using these three elements (iron, silicon, and aluminum) as alloying additions was mainly in connection with heat-treatment studies of metallic uranium reactor fuel elements. The work was directed toward grain size reduction, elevated temperature strengthening, and randomization of the crystallographic texture that resulted from fuel rod processing. To assist in the performance of the current studies at the Oak Ridge Y-12 Plant,[a] a survey of the early literature was completed.[b]

Iron, silicon, and aluminum are only slightly soluble in alpha uranium and show increased solubilities in the beta and gamma phases, which offers the potential for solid solution and precipitation strengthening. The presence of these elements in uranium also tends to affect phase transformation temperatures slightly, makes phase transformations and recrystallization sluggish, and reduces the rate of grain growth. The extent to which these elements are influential is a function of their concentration, the beta- or gamma-solution treatment conditions, and alpha aging temperature and time.

During the course of this study, all of the alloys listed in Table 1 were prepared and properties were determined after a variety of heat treatments. The objectives of this report are to describe the work done to obtain the room-temperature tensile properties of this group of alloys and to illustrate the effects of alloying and heat treatment on microstructure.

Table 1

NOMINAL COMPOSITIONS OF ALLOYS INVESTIGATED

Alloy Designation	Alloy Composition (ppm by weight)
A	Uranium-400 ppm silicon
B	Uranium-150 ppm silicon-100 ppm iron
C	Uranium-200 ppm silicon-300 ppm iron
D	Uranium-400 ppm silicon-200 ppm iron
E or Derby	High-purity uranium
F	Uranium-300 ppm iron
G	Uranium-250 ppm aluminum
H	Uranium-200 ppm aluminum-300 ppm iron
I	Uranium-400 ppm silicon-200 ppm aluminum
K	Uranium-800 ppm silicon-400 ppm iron
L	Uranium-600 ppm iron
N	Uranium-600 ppm silicon-300 ppm iron

The number of alloys investigated and the variety of heat treatments used mandated that each condition could

(a) Operated by the Union Carbide Corporation, Nuclear Division, for the Department of Energy.

(b) R. L. Ludwig, ***Low Alloy Additions of Iron, Silicon, and Aluminum to Uranium - A Literature Survey***, Union Carbide Corporation, Nuclear Division, Oak Ridge Y-12 Plant, Oak Ridge, Tenn., Y-2213 (1980).

be characterized by only a few tests. In this sense, each item of data should be viewed as possessing a standard deviation such that its utility for future applications should be more firmly established by further testing.

THE EFFECT OF SMALL ADDITIONS OF SILICON, IRON, AND ALUMINUM ON THE ROOM-TEMPERATURE TENSILE PROPERTIES OF HIGH-PURITY URANIUM

EXPERIMENTAL WORK

Alloy Preparation

Derby uranium rolled to 13-mm-thick plate and sheared into small squares was used as charge material for the alloy castings. Three representative analyses of this material appear in Table 2. The atomic absorption method was used to analyze for iron and high-aluminum contents because of its superior accuracy (compared to emission spectroscopy). At the time of this study, no method superior to emission spectroscopy was available for silicon analyses. Reagent grade silicon and high-purity iron and aluminum metal were used as alloy additions. The very low alloy levels minimized the effect of any impurities present in these materials.

Table 2

REPRESENTATIVE EMISSION SPECTROGRAPHIC AND CARBON ANALYSES OF DERBY URANIUM PLATE USED IN ALLOY CASTINGS

Element	Concentration (ppm) Sample 1	Sample 2	Sample 3	Notes
Al	12	8	10	
B	< 0.1	0.1	0.1	
Be	< 0.1	< 0.1	< 0.1	
C	67	68	64	Generally 40 to 55 ppm
Ca	12	< 10	< 10	
Cd	< 0.1	< 0.1	< 0.1	
Co	< 1	< 1	< 1	
Cr	< 2	< 2	< 2	
Cu	5	4	4	
Fe	20	20	25	By spectrographic method
Fe	46	47	46	By atomic absorption method
Mg	2	2	6	
Mn	8	8	< 10	
Mo	< 10	< 10	< 10	
Nb	< 10	< 10	< 10	
Ni	5	6	4	
Pb	8	6	8	
Si	35	35	25	
Ti	< 4	< 4	< 4	
V	< 1	< 1	< 1	

Alloy castings were prepared that weighed 18 kg and measured 127 X 178 X 38 mm. Small, master-alloy-button arc melts incorporating the alloy additions and a portion of the uranium provided the means of adding the alloy. The buttons were then vacuum-induction melted (VIM) with the balance of the uranium in a coated graphite crucible, and the melt was poured into a coated graphite mold. The castings were homogenized in a vacuum at about 1.3×10^{-3} Pa for 4 h at 900°C, and the top and bottom of each casting were sampled for chemistry. One to four castings were made of each composition. The chemical analyses of the alloy castings are given in Table 3. The analytical samples were obtained from top and bottom locations in each ingot, and the results indicated an homogeneous distribution of the alloying elements and impurities.

The ingots were hot cross-rolled to a 50% reduction in thickness at 625°C, annealed at 625°C, and warm cross-rolled at 300°C to about a 40% reduction to yield plates ~ 220 X 300 X 12 mm. Two different heating environments were used for the ingots (as shown in Table 4). The salt bath was replaced by the argon-air atmosphere to reduce hydrogen pickup from the salt bath by the uranium. The air contamination in the argon caused considerable oxidation of the ingots but not enough to be a problem. The wrought plates were not annealed at the completion of warm rolling. Tensile bar blanks ~ 110 X 12 X 13 mm were sawed from each plate to be used in the heat-treatment experiments.

Heat Treatment

Properties were measured after alpha annealing, beta-solution treatment at 730°C and water quench, gamma-solution treatment at 800°C and water quench, and after aging subsequent to the beta and gamma treatments. Because of the damaging effects of small concentrations of hydrogen, the heat treatments were to be carried out in vacuum or inert gas to remove most of the dissolved hydrogen. The majority of the heat treatments were performed in a vertical-tube vacuum furnace, heated with external electrical resistance elements, and arranged such that the furnace could be backfilled with argon, the bottom swung open, and the charge dropped into a barrel of agitated water. As many as 12 to 16 bars could be heat-treated in a single run.

It was originally intended that, since yield strength was of primary interest, hardness measurements might be a satisfactory screening method for establishing optimum times at temperature to achieve the highest strength level for a given alloy. Some tests were run in which Rockwell A hardness measurements were used for this purpose, but the hardness test was insufficiently discriminating to be useful, probably because of the relatively small differences in yield strength that were of interest. The inability of this method to indicate differences in ductility was also a decisive factor in its being abandoned.

Binary Alloys

Four binary alloys were investigated: uranium-400 ppm silicon (U-400 Si), Alloy A; uranium-300 ppm iron (U-300 Fe), Alloy F; uranium-600 ppm iron (U-600 Fe), Alloy L; and uranium-250 ppm aluminum (U-250 Al), Alloy G. The properties of these alloys, as beta- and gamma-quenched, appear in Table 5 compared with similarly treated high-purity derby uranium. Tensile bar blanks were also subjected to various alpha annealing temperatures, and properties were determined after vacuum annealing for 4 h at 450° and 550°C and for 0.5 and 1 h at 630°C. These data are presented in Table 6 and, based on the derby uranium properties, indicate that, to a considerable extent, the strengthening obtained is a function of the wrought and recrystallized structure and is only slightly enhanced by the alloying additions.

Table 3

CHEMICAL ANALYSIS OF ALLOY CASTINGS
AT TOP AND BOTTOM LOCATIONS

Alloy Casting ID and Sample Location	Nominal Composition (ppm)	Element Concentration (ppm)									
		C	Al	Cr	Cu	Fe	Mn	Mo	Ni	Pb	Si
1A Top	U-400 Si	40	6	< 2	6	57	6	< 10	5	6	500
Bottom	U-400 Si	51	6	< 2	6	86	6	< 10	4	8	500
2A Top	U-400 Si	36	5	< 2	3	54	4	< 10	2	< 5	427
Bottom	U-400 Si	43	4	< 2	3	55	6	< 10	2	< 5	401
3A Top	U-400 Si	60	3	< 2	2	56	2	< 10	1	< 5	374
Bottom	U-400 Si	56	5	< 2	3	45	8	< 10	3	< 5	406
1B Top	U-150 Si-100 Fe	48	9	3	3	126	4	< 10	3	< 5	190
Bottom	U-150 Si-100 Fe	47	9	< 2	3	131	< 1	< 10	2	< 5	193
2B Top	U-150 Si-100 Fe	48	7	< 2	3	119	4	< 10	4	< 5	189
Bottom	U-150 Si-100 Fe	46	8	< 2	3	126	3	< 10	4	5	185
3B Top	U-150 Si-100 Fe	49	7	< 2	3	126	7	< 10	3	< 5	174
Bottom	U-150 Si-100 Fe	45	5	< 2	2	128	2	< 10	1	< 5	177
1C Top	U-200 Si-300 Fe	54	7	< 2	2	324	4	< 10	6	< 5	180
Bottom	U-200 Si-300 Fe	46	5	< 2	2	320	1	< 10	1	< 5	214
2C Top	U-200 Si-300 Fe	41	6	< 2	3	318	7	< 10	3	< 5	221
Bottom	U-200 Si-300 Fe	47	5	< 2	3	320	3	< 10	2	< 5	214
3C Top	U-200 Si-300 Fe	72	7	< 2	4	322	6	< 10	5	< 5	219
Bottom	U-200 Si-300 Fe	75	7	< 2	4	315	12	< 10	2	8	240
1D Top	U-400 Si-200 Fe	38	5	< 2	3	234	10	< 10	3	5	421
Bottom	U-400 Si-200 Fe	62	5	< 2	2	228	7	< 10	5	5	413
2D Top	U-400 Si-200 Fe	40	6	< 2	3	242	6	< 10	12	7	402
Bottom	U-400 Si-200 Fe	54	7	< 2	3	239	8	< 10	2	10	410
3D Top	U-400 Si-200 Fe	41	7	2	4	246	13	< 10	9	10	374
Bottom	U-400 Si-200 Fe	37	7	< 2	4	247	4	< 10	6	10	378
4D Top	U-400 Si-200 Fe	50	8	< 2	2	167	9	< 10	5	6	391
Bottom	U-400 Si-200 Fe	63	10	< 2	2	163	7	< 10	6	5	415
1E Top	Pure uranium	38	6	< 2	4	81	6	< 10	4	8	50
Bottom	Pure uranium	42	7	< 2	3	61	8	< 10	6	11	49
2E Top	Pure uranium	35	9	< 2	3	64	7	< 10	3	8	52
Bottom	Pure uranium	47	6	2	3	59	9	< 10	3	11	47
3E Top	Pure uranium	28	8	< 2	3	64	4	< 10	5	8	52
Bottom	Pure uranium	38	6	< 2	2	65	3	< 10	4	12	41
1F Top	U-300 Fe	40	< 20	< 2	3	271	6	< 10	5	< 5	20
Bottom	U-300 Fe	54	< 20	< 2	2	270	6	< 10	6	< 5	20

Table 3 (continued)

Alloy Casting ID and Sample Location	Nominal Composition (ppm)	Element Concentration (ppm)									
		C	Al	Cr	Cu	Fe	Mn	Mo	Ni	Pb	Si
2F Top	U-300 Fe	43	< 20	< 2	6	266	5	< 10	6	6	40
Bottom	U-300 Fe	34	< 20	< 2	4	256	6	< 10	6	6	40
3F Top	U-300 Fe	40	< 20	< 2	4	268	5	< 10	6	6	40
Bottom	U-300 Fe	46	< 20	< 2	8	278	4	< 10	6	6	40
1G Top	U-250 Al	49	235	< 2	1	35	4	< 10	2	< 5	30
Bottom	U-250 Al	53	237	< 2	1	33	4	< 10	2	< 5	30
2G Top	U-250 Al	53	242	< 2	6	71	5	< 10	6	6	55
Bottom	U-250 Al	56	242	< 2	6	61	5	< 10	5	6	40
3G Top	U-250 Al	44	234	< 2	4	37	4	< 10	6	8	40
Bottom	U-250 Al	42	234	< 2	4	40	5	< 10	10	6	40
1H Top	U-200 Al-300 Fe	42	196	< 2	-	270	5	< 10	6	6	60
Bottom	U-200 Al-300 Fe	50	192	< 2	-	269	5	< 10	4	6	60
2H Top	U-200 Al-300 Fe	44	192	< 2	-	271	5	< 10	6	6	70
Bottom	U-200 Al-300 Fe	44	188	< 2	-	284	5	< 10	6	6	65
3H Top	U-200 Al-300 Fe	36	186	< 2	-	271	5	< 10	4	10	60
Bottom	U-200 Al-300 Fe	36	194	< 2	-	277	6	< 10	6	10	65
1I Top	U-400 Si-200 Al	35	187	< 2	6	38	5	< 10	5	8	400
Bottom	U-400 Si-200 Al	36	194	< 2	6	52	6	< 10	8	10	400
2I Top	U-400 Si-200 Al	56	184	< 2	8	30	5	< 10	6	8	400
Bottom	U-400 Si-200 Al	55	186	< 2	6	30	5	< 10	30	10	400
3I Top	U-400 Si-200 Al	56	179	< 2	7	32	5	< 10	40	7	400
Bottom	U-400 Si-200 Al	53	184	< 2	6	30	5	< 10	8	8	400
1K Top	U-800 Si-400 Fe	54	6	< 2	2	425	3	< 10	< 1	< 5	1000
Bottom	U-800 Si-400 Fe	54	6	< 2	3	414	3	< 10	< 1	< 5	1000
2K Top	U-800 Si-400 Fe	153	40	< 2	15	480	15	25	15	8	1000
Bottom	U-800 Si-400 Fe	156	35	< 2	15	480	15	15	15	8	1000
1L Top	U-600 Fe	46	13	< 2	3	574	8	< 10	7	< 5	52
Bottom	U-600 Fe	61	12	< 2	3	573	8	< 10	4	6	47
1N Top	U-600 Si-300 Fe	216	35	< 2	18	400	20	40	20	6	1000
Bottom	U-600 Si-300 Fe	246	225	< 2	15	390	25	15	25	8	1000

Table 4

HEATING ENVIRONMENTS USED IN PROCESSING CAST INGOTS

Alloy	Hot Rolling Heating Medium (625°C)	Warm Rolling Medium (300°C)	Annealing Medium (625°C)
A	Salt bath	Argon-Air(1)	Salt bath
B	Salt bath	Argon-Air	Salt bath
C	Salt bath	Argon-Air	Salt bath
D	Salt bath	Argon-Air	Salt bath
E	Salt bath	Argon-Air	Salt bath
F	Argon-Air	Argon-Air	Argon-Air
G	Argon-Air	Argon-Air	Argon-Air
H	Argon-Air	Argon-Air	Argon-Air
I	Argon-Air	Argon-Air	Argon-Air
K	Argon-Air	Argon-Air	Argon-Air
L	Argon-Air	Argon-Air	Argon-Air
N	Argon-Air	Argon-Air	Argon-Air

(1) The furnace atmosphere was intended to be argon, but leaks resulted in a considerable dilution with air.

The binary alloys were also subjected to various aging treatments after being beta- and gamma-solution treated in vacuum and water-quenched. Some of the aging was done in evacuated Pyrex ampoules, but the majority was in a dynamic vacuum; this difference will be discussed later in the report. Tensile properties were obtained and appear in Table 7. In general, aging was ineffective except possibly for the U-Fe alloys, Alloys F and L, for which yield strengths on the order of 414 MPa (60 ksi) were obtained on aging following a gamma-solution heat treatment and water quench. There was also an apparent tendency for aging treatments to generally lower the ductility. In contrast to the alpha-annealed properties shown in Table 6, the properties of the alloyed materials shown in Table 7 are clearly superior to those of unalloyed uranium.

Uranium-Silicon-Iron Ternary Alloys

Note from Table 1 that five uranium-silicon-iron ternary alloys were investigated. The analyses of the various alloy castings, Alloys B, C, D, K, and N, are given in Table 3. Aside from the silicon and iron compositional differences, Alloy 2K differs from Alloy K in that its carbon content and general impurity level are higher than is normal for derby uranium. Alloy N was also prepared using lower purity uranium. Table 8 summarizes the tensile properties of these alloys as quenched from the beta- (730°C) and gamma- (800°C) phase regions. The table reveals several tendencies: (1) a minimum solution heat-treatment time for tensile bar blanks of 2 h, but preferably 4 h, at temperature to achieve high ductilities; (2) a general increase in yield strength with increasing alloy content; and (3) a drop in ductility at high-alloy contents that is more severe for gamma- than beta-quenched material, particularly for the alloys made with less-pure uranium.

The U-Si-Fe ternary alloys in the form of warm-rolled plate were also subjected to a variety of alpha annealing treatments for various times at temperatures of 450°, 550°, and 630°C. Some of the specimens had been encapsulated in evacuated quartz ampoules prior to heat

Table 5

TENSILE PROPERTIES OF BETA- AND GAMMA-QUENCHED ALLOYS
NOMINALLY URANIUM-300 IRON, URANIUM-600 IRON,
URANIUM-250 ALUMINUM, URANIUM-400 SILICON

Alloy	Nominal Composition (ppm)	Heat Treatment	Tensile Strength		Yield Strength (0.2%)		Elongation in 25.4 mm	Reduction in Area	Notes
			(MPa)	(ksi)	(MPa)	(ksi)	(%)	(%)	
F	U-300 Fe	730°, vacuum, 4 h, W-Q[1]	872	126.5	308	44.6	35	34	Average of 4 tests
		800°, vacuum, 4 h, W-Q	880	127.6	325	47.2	33	36	Average of 3 tests
L	U-600 Fe	730°, vacuum, 4 h, W-Q	956	138.6	325	47.2	31	32	Average of 2 tests
		800°, vacuum, 4 h, W-Q	952	138.0	334	48.5	27	26	Average of 3 tests
G	U-250 Al	730°, vacuum, 4 h, W-Q	778	112.9	218	31.6	25	21	Average of 3 tests
		800°, vacuum, 4 h, W-Q	618	89.6	218	31.6	10	12	Average of 2 tests
A	U-400 Si	730°, vacuum, 2 h, W-Q	857	124.3	304	44.1	31	28	Average of 2 tests
		800°, vacuum, 2 h, W-Q	842	122.1	338	49.1	18	18	Average of 2 tests
E	Derby U	730°, vacuum, 4 h, W-Q	749	108.7	201	29.1	33	37	Average of 3 tests
		800°, vacuum, 4 h, W-Q	674	97.8	213	30.9	20	21	Average of 3 tests

[1] Water-quenched.

Table 6

TENSILE PROPERTIES OF BINARY ALLOYS AND DERBY URANIUM ALPHA ANNEALED IN VACUUM FOR DIFFERENT TIME-TEMPERATURE CONDITIONS (EACH RESULT IS AVERAGE OF THREE TESTS, UNLESS NOTED)

Alloy and Nominal Composition (ppm)		Temperature (°C)	Time (h)	Tensile Strength		Yield Strength (0.2%)		Elongation in 25.4 mm (%)	Reduction in Area (%)
				(MPa)	(ksi)	(MPa)	(ksi)		
A	U-400 Si	As-rolled		985	142.8	543	78.8	18.3	19.6
		550	4	874	126.8	307	44.5	22.8	21.6
		630	1	774	112.3	279	40.5	7.0	7.6
F	U-300 Fe	As-rolled		967	140.3	534	77.4	25.0	34.8
		450	4	869	130.0	367	53.2	32.3	56.8
		550	4	829	120.2	319	46.2	40.8	63.7
		630	0.5	824	119.5	318	46.1	29.7	29.3
L	U-600 Fe	450	4	930	134.9	396	57.5	24.0	33.0
		550	4	867	125.8	316	45.9	39.0	49.0
		630	0.5	823	119.3	358	50.4	19.0	21.0
G	U-250 Al	As-rolled		989	143.4	559	81.1	27.0(1)	30.5(1)
		450	4	944	136.9	386	56.0	27.3	50.1
		550	4	843	122.3	296	42.9	34.2	41.3
		630	0.5	809	117.4	325	47.1	26.0	21.1
Derby U		As-rolled		907	131.6	521	75.5	17.7	17.6
		450	4	877	127.2	305	44.3	30.3	40.3
		550	4	792	114.8	317	46.0	41.8	64.1
		630	0.5	697	101.0	300	43.6	13.7	15.4

(1) Average of two tests.

Table 7

TENSILE PROPERTIES OF DILUTE BINARY ALLOYS AFTER BETA AND GAMMA SOLUTION HEAT TREATMENTS AND ALPHA AGING AT VARIOUS TEMPERATURES AND TIMES (AVERAGE OF TWO TESTS, EXCEPT AS NOTED)

Alloy	Nominal Composition (ppm)	Solution Treatment Conditions	Aging Treatment Conditions	Tensile Strength (MPa)	Tensile Strength (ksi)	Yield Strength (0.2%) (MPa)	Yield Strength (0.2%) (ksi)	Elongation in 25.4 mm (%)	Reduction in Area (%)	Remarks
A	U-400 Si	800°C, vacuum	350°C, 1 h	776	112.5	302	43.8	12	11	Aged in Pyrex ampoule(1)
		2 h, W-Q	350°C, 30 h	696	101.0	316	45.9	8	6	Aged in Pyrex ampoule(1)
		730°C, vacuum	400°C, 18 h	799	115.9	327	47.4	15	11	Aged in Pyrex ampoule(1)
		2 h, W-Q	400°C, 48 h	780	113.1	314	45.6	14	12	Aged in Pyrex ampoule(1)
F	U-300 Fe	730°C, vacuum	350°C, 2h	788	114.3	372	54.0	11	8	Aged in dynamic vacuum
		4 h, W-Q	350°C, 72 h	874	126.7	379	54.9	14	10	Aged in dynamic vacuum
			450°C, 1 h	807	117.0	379	54.9	11	9	Aged in dynamic vacuum
			450°C, 8 h	847	122.9	359	52.0	17	13	Aged in dynamic vacuum
			450°C, 24 h	848	123.0	348	50.5	35	37	Aged in dynamic vacuum
		800°C, vacuum	350°C, 2 h	916	132.9	410	59.5	21	15	Aged in dynamic vacuum
		4 h, W-Q	350°C, 72 h	963	139.6	440	63.8	20	11	Aged in dynamic vacuum
			350°C, 100 h	839	121.7	347	50.3	13	12	Aged in dynamic vacuum(1)
			450°C, 1 h	934	135.5	373	54.1	24	20	Aged in dynamic vacuum
			450°C, 8 h	865	125.5	360	52.2	16	14	Aged in dynamic vacuum(1)
			450°C, 24 h	852	123.5	305	44.3	28	26	Aged in dynamic vacuum
L	U-600 Fe	730°C, vacuum	400°C, 2 h	907	131.5	361	52.3	15	14	Aged in dynamic vacuum
		4 h, W-Q	400°C, 6 h	931	135.0	390	56.6	13	11	Aged in dynamic vacuum
			400°C, 18 h	947	137.3	375	54.4	13	11	Aged in dynamic vacuum
			400°C, 54 h	816	118.3	345	50.0	12	12	Aged in dynamic vacuum
		800°C, vacuum	400°C, 2 h	716	103.9	359	52.0	6	6	Aged in dynamic vacuum
		4 h, W-Q	400°C, 6 h	923	133.9	432	62.7	10	10	Aged in dynamic vacuum
			400°C, 18 h	867	125.8	370	53.7	11	12	Aged in dynamic vacuum
			400°C, 54 h	852	123.5	383	55.5	12	10	Aged in dynamic vacuum
G	U-250 Al	730°C, vacuum	350°C, 2 h	711	103.1	241	34.9	15	12	Aged in dynamic vacuum
		4 h, W-Q	350°C, 72 h	694	100.7	249	36.1	14	12	Aged in dynamic vacuum
			450°C, 1 h	716	103.8	233	33.8	15	13	Aged in dynamic vacuum(1)
			450°C, 8 h	735	106.6	239	34.7	18	16	Aged in dynamic vacuum
			450°C, 24 h	741	107.4	244	35.4	17	12	Aged in dynamic vacuum(1)
E	Derby U	730°C, vacuum	400°C, 6 h	623	90.4	194	28.1	15	14	Aged in dynamic vacuum(2)
		4 h, W-Q								
		800°C, vacuum	400°C, 2 h	563	81.6	233	33.8	8	8	Aged in dynamic vacuum(2)
		4 h, W-Q	400°C, 18 h	696	101.0	229	33.2	15	21	Aged in dynamic vacuum(2)

(1) Average of 3 tests.

(2) Single test.

Table 8

TENSILE PROPERTIES OF URANIUM-SILICON-IRON TERNARY ALLOYS AFTER BETA (730°C) AND GAMMA (800°C) SOLUTION HEAT TREATMENTS AND WATER QUENCH

Alloy			Tensile Strength		Yield Strength (0.2%)		Elongation in 25.4 mm	Reduction in Area	
Letter ID	Nominal Composition (ppm)	Heat Treatment	(MPa)	(ksi)	(MPa)	(ksi)	(%)	(%)	Remarks
B	U-150 Si-100 Fe	0.75 h, 800°C, vacuum, W-Q	798	115.8	335	48.6	15	13	Average of 3 tests
		2 h, 800°C, vacuum, W-Q	867	125.8	354	51.3	27	26	Average of 2 tests
		1 h, 730°C, vacuum, W-Q	712	103.3	330	47.8	11	9	Average of 2 tests
		2 h, 730°C, vacuum, W-Q	838	121.6	336	48.7	27	27	Average of 3 tests
C	U-200 Si-300 Fe	2 h, 800°C, vacuum, W-Q	930	134.9	376	54.5	29	29	Average of 5 tests
		1 h, 730°C, vacuum, W-Q	794	115.2	358	52.0	12	10	Average of 3 tests
		2 h, 730°C, vacuum, W-Q	940	136.3	386	56.0	29	25	Average of 2 tests
D	U-400 Si-200 Fe	0.25 h, 800°C, vacuum, W-Q	849	123.1	366	53.1	13	11	Average of 2 tests
		0.5 h, 800°C, vacuum, W-Q	847	122.9	362	52.5	13	11	Average of 2 tests
		1 h, 800°C, vacuum, W-Q	960	139.2	356	51.6	31	32	Average of 1 test
		2 h, 800°C, vacuum, W-Q	947	137.3	372	53.9	33	36	Average of 7 tests
		4 h, 800°C, vacuum, W-Q	952	138.1	362	52.5	31	29	Average of 5 tests
		1 h, 730°C, vacuum, W-Q	836	121.2	390	56.5	10	9	Average of 3 tests
		2 h, 730°C, vacuum, W-Q	913	132.4	370	53.7	24	22	Average of 2 tests
		4 h, 730°C, vacuum, W-Q	941	136.5	360	52.2	34	37	Average of 3 tests
K	U-800 Si-400 Fe	4 h, 730°C, vacuum, W-Q	991	143.7	432	62.7	19	15	Average of 6 tests
		4 h, 800°C, vacuum, W-Q	937	135.8	483	70.0	13	9	Average of 3 tests
2K	U-800 Si-400 Fe	4 h, 730°C, vacuum, W-Q	975	141.4	453	65.7	13	11	Average of 6 tests
	(impure)	4 h, 800°C, vacuum, W-Q	780	113.1	483	70.0	4	3	Average of 4 tests
N	U-600 Si-300 Fe	4 h, 730°C, vacuum, W-Q	1012	146.8	381	55.3	23	22	Average of 6 tests
	(impure)	4 h, 800°C, vacuum, W-Q	827	119.9	361	52.3	6	5	Average of 6 tests

treatment, while others were annealed in a dynamic, high-vacuum atmosphere. Table 9 contains a compilation of the properties. Since Alloys B, C, D, and E were processed by salt bath heating, the tensile strengths, elongations, and reductions of area show low values typical of uranium for annealing times too short to eliminate the embrittling effects of hydrogen. However, the yield strengths were relatively unaffected and reflect the effects of the annealing conditions. Compared with the alpha-annealed binary alloys in Table 6, the more highly alloyed materials in Table 9 show yield strengths that are more obviously higher than that of derby uranium. A number of examples appear in the table which show that extended annealing times cause dramatic increases in ductility, with little or no effect on yield strength. These increases occurred regardless of whether the annealing took place in a quartz ampoule or a dynamic vacuum.

Because annealing at high-alpha temperatures is essentially an overaging process with respect to the alloying additions, there appears to be a limit of about a 100 MPa increase in yield strength over derby uranium due to these reported levels of alloying in alpha-annealed material. This is probably the consequence of dispersion hardening by intermetallic compound particles formed from uranium and the alloying additions in the course of casting and working the ingot as well as during the alpha anneal.

Aging of the beta- and gamma-quenched U-Si-Fe ternary alloys was also performed in an attempt to enhance their properties. The tensile properties obtained for various heat treatments appear in Table 10. Compared with the as-quenched data in Table 8, Alloys D and N appeared to be most amenable to strengthening by aging. Both the lower and higher alloy content materials, Alloys B and K, did not benefit greatly from the aging treatments. The optimum improvement in properties was observed for Alloy D after a beta vacuum solution treatment and water quench followed by an aging treatment of 18 h at 400°C in vacuum. The yield strength obtained was 456 MPa (66.2 ksi). An attempt was made to duplicate this value; but, as shown in the table, the average of six test specimens gave a yield strength of only 393 MPa (57.0 ksi). However, these tensile bars were from Alloy 4D (Table 3) which contained only about two-thirds as much iron as the other three Alloy D castings. These six bars showed generally higher elongations than any of the other aged alloys, which may also relate to the alloy content.

Alloy N could be aged to a considerable increase in yield strength, but with a substantial loss of ductility. As was experienced for Alloy D, aging at 400°C after a beta-solution treatment and water quench produced optimum results for Alloy N. The yield strengths of the unalloyed-derby-uranium tensile bars that were heat-treated with these alloy bars are not given in the table, but they all showed yield strengths of about 200 to 240 MPa (29 to 35 ksi) and relatively low elongations.

The desirable high yield strengths shown in Table 10 are probably of little significance because the associated ductilities are too low to make the alloys in these conditions of heat treatment of any practical use.

Table 9

TENSILE PROPERTIES OF ALPHA-ANNEALED URANIUM-SILICON-IRON TERNARY ALLOYS
(ALL RESULTS ARE AVERAGE OF THREE TESTS, EXCEPT AS NOTED)

Alloy Letter ID	Alloy Nominal Composition (ppm)	Heat Treatment	Tensile Strength (MPa)	Tensile Strength (ksi)	Yield Strength (0.2%) (MPa)	Yield Strength (0.2%) (ksi)	Elongation in 25.4 mm (%)	Reduction in Area (%)	Remarks
B	U-150 Si-100 Fe	As-rolled	872	126.4	496	72.0	16	13	
		550°C, 0.25 h	783	113.6	340	49.3	12	11	Annealed in quartz ampoule
		550°C, 1 h	725	105.2	361	52.3	8	9	Annealed in quartz ampoule
		630°C, 0.25 h	703	101.9	276	40.1	7	8	Annealed in quartz ampoule
		630°C, 0.5 h	703	102.0	281	40.8	7	6	Annealed in quartz ampoule
C	U-200 Si-300 Fe	As-rolled	920	133.4	551	79.9	8	9	
		550°C, 0.25 h	796	115.5	348	50.5	7	8	Annealed in quartz ampoule
		550°C, 4 h	863	125.2	348	50.4	30	40	Annealed in quartz ampoule
		630°C, 0.25 h	756	109.7	312	45.2	8	9	Annealed in quartz ampoule
		630°C, 2 h	827	119.9	300	43.5	21	16	Annealed in quartz ampoule
D	U-400 Si-200 Fe	As-rolled	959	139.1	547	79.3	11	13	
		450°C, 4 h	983	142.6	452	65.5	23	54	Annealed in dynamic vacuum
		550°C, 0.25 h	856	124.1	352	51.1	10	12	Annealed in quartz ampoule
		550°C, 4 h	888	128.8	352	51.0	35	42	Annealed in dynamic vacuum
		550°C, 8 h	873	126.6	355	51.5	32	44	Annealed in quartz ampoule
		630°C, 0.25 h	780	113.1	302	43.8	8	7	Annealed in quartz ampoule
		630°C, 0.5 h	873	126.6	334	48.4	31	39	Annealed in dynamic vacuum
		630°C, 4 h	861	124.9	323	46.9	34	44	Annealed in quartz ampoule
K	U-800 Si-400 Fe	450°C, 4 h	1047	151.9	511	74.1	14	20	Annealed in dynamic vacuum
		550°C, 4 h	956	138.7	380	55.1	31	42	Annealed in dynamic vacuum
		630°C, 0.5 h	938	136.0	410	59.4	19	17	Annealed in dynamic vacuum
2K	U-800 Si-400 Fe (impure)	As-rolled	1053	152.7	459	66.6	20	19	
		550°C, 4 h	1021	148.1	342	49.6	22	27	Annealed in dynamic vacuum; average of 6 tests
		550°C, 7 h	983	142.6	333	48.3	24	25	Annealed in dynamic vacuum
N	U-600 Si-300 Fe (impure)	As-rolled	1030	149.4	443	64.3	21	32	Average of 2 tests
		550°C, 4 h	959	139.1	346	50.2	23	26	Annealed in dynamic vacuum; average of 6 tests
		550°C, 7 h	938	136.1	344	49.9	24	31	Annealed in dynamic vacuum
E Derby U		As-rolled	854	123.9	514	74.5	14	15	Annealed in quartz ampoule
		550°C, 0.25 h	714	103.5	279	40.3	13	11	Annealed in quartz ampoule
		550°C, 0.5 h	703	101.9	294	42.7	12	10	Annealed in quartz ampoule
		630°C, 0.5 h	679	98.5	274	39.8	11	12	Annealed in quartz ampoule

Table 10

TENSILE PROPERTIES OF URANIUM-SILICON-IRON TERNARY ALLOYS AGED AFTER BETA- OR GAMMA-SOLUTION HEAT TREATMENT AND WATER QUENCHING (AVERAGE OF THREE TESTS, UNLESS NOTED)

Letter ID	Nominal Composition (ppm)	Solution Treatment Conditions	Aging Treatment Conditions	Tensile Strength (MPa)	Tensile Strength (ksi)	Yield Strength (0.2%) (MPa)	Yield Strength (0.2%) (ksi)	Elongation in 25.4 mm (%)	Reduction in Area (%)	Remarks
B	U-150 Si-100 Fe	800°C, 2 h	350°C, 1 h	809	117.4	315	45.7	16	12	Aged in Pyrex ampoules
		vacuum, W-Q	350°C, 8 h	714	103.5	303	44.0	10	9	Aged in Pyrex ampoules
		730°C, 2 h	400°C, 2 h	704	102.1	332	48.2	9	9	Aged in Pyrex ampoules
		vacuum, W-Q	400°C, 48 h	738	107.0	341	49.4	11	9	Aged in Pyrex ampoules
C	U-200 Si-300 Fe	800°C, 2 h	350°C, 1 h	794	115.1	336	48.7	11	8	Aged in Pyrex ampoules
		vacuum, W-Q	350°C, 24 h	783	113.5	394	57.2	6	7	Aged in Pyrex ampoules
		730°C, 2 h	400°C, 6 h	769	111.5	396	57.5	8	8	Aged in Pyrex ampoules
		vacuum, W-Q	400°C, 48 h	769	111.6	402	58.3	7	6	Aged in Pyrex ampoules
D	U-400 Si-200 Fe	800°C, 2 h	350°C, 1 h	895	129.8	359	52.0	19	18	Aged in Pyrex ampoules
		vacuum, W-Q	350°C, 24 h	813	117.9	366	53.1	11	6	Aged in Pyrex ampoules
		730°C, 2 h vacuum, W-Q	400°C, 18 h	837	121.4	456	66.2	9	7	Aged in Pyrex ampoules, high yield strength
			400°C, 48 h	840	121.8	442	64.1	9	8	Aged in Pyrex ampoules
		730°C, 4 h vacuum, W-Q	400°C, 18 h	976	141.5	393	57.0	21	16	Average of six tests; aged in dynamic vacuum; repeat of high yield strength
K	U-800 Si-400 Fe	730°C, 4 h vacuum, W-Q	400°C, 2 h	801	116.1	417	60.5	5	5	Average of two tests; aged in dynamic vacuum
			400°C, 6 h	-	-	-	-	-	-	Average of two tests; aged in dynamic vacuum; broke before yield
			400°C, 18 h	904	131.1	394	57.1	8	6	Average of two tests; aged in dynamic vacuum
			400°C, 54 h	897	130.1	435	63.1	9	6	Average of two tests; aged in dynamic vacuum
		800°C, 4 h vacuum, W-Q	400°C, 2 h	938	136.1	487	70.6	6	5	Average of two tests; aged in dynamic vacuum
			400°C, 6 h	812	117.8	475	68.9	4	0	Single test; aged in dynamic vacuum
			400°C, 18 h	854	123.9	467	67.8	5	5	Average of two tests; aged in dynamic vacuum
			400°C, 54 h	820	118.9	439	63.6	4	4	Average of two tests; aged in dynamic vacuum

Table 10 (continued)

Letter ID	Nominal Composition (ppm)	Solution Treatment Conditions	Aging Treatment Conditions	Tensile Strength (MPa)	Tensile Strength (ksi)	Yield Strength (0.2%) (MPa)	Yield Strength (0.2%) (ksi)	Elongation in 25.4 mm (%)	Reduction in Area (%)	Remarks
2K	U-800 Si-400 Fe (impure)	730°C, 4 h vacuum, W-Q	400°C, 2 h	891	129.2	524	76.0	4	4	Aged in dynamic vacuum
			400°C, 18 h	897	130.1	463	67.1	4	5	Average of two tests; aged in dynamic vacuum
			500°C, 2 h	913	132.4	398	57.7	8	7	Average of two tests; aged in dynamic vacuum
N	U-600 Si-300 Fe (impure)	730°C, 4 h vacuum, W-Q	400°C, 2 h	909	131.9	476	69.0	6	4	Aged in dynamic vacuum
			400°C, 18 h	914	132.5	429	62.2	6	5	Average of two tests; aged in dynamic vacuum
			500°C, 2 h	891	129.2	394	57.2	9	8	Average of two tests; aged in dynamic vacuum

Other Ternary Alloys

Two other alloys were investigated in addition to the U-Si-Fe ternary alloys described. These were Alloy H, uranium-200 ppm aluminum-300 ppm iron, and Alloy I, uranium-400 ppm silicon-200 ppm aluminum. The tensile properties of these two alloys in the beta- and gamma-solution-annealed and water-quenched conditions and as alpha annealed appear in Table 11. Similarly treated derby uranium specimens are included for comparison.

The presence of aluminum in the two alloys did little or nothing to improve the properties over those of the corresponding binary alloys, Alloys F and A (Tables 5 and 6). However, the aluminum apparently did result in the deterioration of the ductility of Alloy I after the 800°C solution anneal and water quench; however, only a single test is reported for these conditions. As has been observed previously, the presence of the alloying additions improves the properties of the alpha-annealed alloys only slightly over those of the alpha-annealed wrought derby uranium plate. However, it can be seen from the 450°C annealing data that the rate of softening is decreased by the presence of the alloying elements.

Table 12 presents the results of aging the U-200 Al-300 Fe and U-400 Si-200 Al alloys after beta- and gamma-solution heat treatment. Comparing the yield strengths with the values from the unaged solution-treated alloys in Table 11 shows that almost all aging treatments increased the yield strength of Alloy H, but to a degree no greater than that observed for Alloy F, the U-300 Fe binary alloy (Table 7). Similarly, the data for Alloy I show no significant variations from those of the binary Alloy A, U-400 Si (Table 7). These results tend to reinforce the observation based on Alloy G, the U-250 ppm Al binary alloy, that a small concentration of aluminum has essentially no effect upon the room-temperature tensile properties of uranium. The elongations for Alloy H tended to be lowered by aging, as did those of the derby uranium control material; but those of Alloy I remained reasonably high.

Hydrogen Effects

Certain processing methods result in the diffusion of hydrogen into uranium, and even very low concentrations of hydrogen adversely affect the tensile strength and ductility of uranium, but not its yield strength. Hydrogen analyses were made of a number of specimens from many of the alloys in an attempt to establish a feeling for the relationship among hydrogen content, properties, and processing.

Tables 13 and 14 contain the obtained analytical data. Alloys A, B, C, D, and some derby uranium had been heated for processing in a salt bath. The analyses of these materials after various processing steps are shown in Table 13 and range from 0.2- to 0.4-ppm hydrogen for the as-processed alloys. In Tables 8 and 10, the properties of some of these alloys are presented as a function of heat treating times and environments. Improved ductilities are seen to accompany increasing times at the 730° and 800°C solution-annealing temperatures, a consequence that is undoubtedly the result of hydrogen diffusion from the specimen bar. This reduction is shown in Table 13 for vacuum solution treatment times of one hour.

Table 11

TENSILE PROPERTIES OF URANIUM-ALUMINUM-IRON AND URANIUM-SILICON-ALUMINUM TERNARY ALLOYS AFTER BETA- AND GAMMA-SOLUTION HEAT TREATMENT AND AFTER ALPHA ANNEALING (RESULTS ON DERBY URANIUM INCLUDED FOR COMPARISON—— AVERAGE OF THREE TESTS, UNLESS NOTED)

Alloy			Tensile Strength		Yield Strength (0.2%)		Elongation in 25.4 mm	Reduction in Area	
Letter ID	Nominal Composition (ppm)	Heat Treatment	(MPa)	(ksi)	(MPa)	(ksi)	(%)	(%)	Remarks
		Beta- and Gamma-Solution Heat-Treated and Water-Quenched							
H	U-200 Al-300 Fe	800°C, 4 h, vacuum, W-Q	915	132.7	327	47.4	32	34	Average of 5 tests
		730°C, 4 h, vacuum, W-Q	899	130.4	323	46.8	35	39	Average of 5 tests
I	U-400 Si-200 Al	800°C, 4 h, vacuum, W-Q	813	117.9	307	44.5	13	12	Single Test
		730°C, 4 h, vacuum, W-Q	866	125.6	294	42.6	34	38	
Derby	U-30 Fe-25	800°C, 4 h, vacuum, W-Q	674	97.8	213	30.9	20	21	
U	Si-< 20 Al	730°C, 4 h, vacuum, W-Q	749	108.7	201	29.1	33	37	
		As-Rolled and Alpha Annealed							
H	U-200 Al-300 Fe	As-rolled	986	143.0	507	73.6	22	24	
		450°C, 0.25 h, vacuum	973	141.1	416	60.4	8	7	
		4 h	972	141.0	407	59.1	23	39	
		550°C, 0.25 h, vacuum	885	128.3	294	42.7	21	20	
		4 h	842	122.1	338	49.0	38	58	
		630°C, 0.5 h, vacuum	826	119.8	303	44.0	22	19	
I	U-400 Si-200 Al	As-rolled	1031	149.5	493	71.5	27	41	
		450°C, 0.25 h, vacuum	1013	146.9	469	68.0	26	37	
		4 h	987	143.1	419	60.7	28	43	
		550°C, 0.25 h, vacuum	955	138.5	341	49.5	28	39	
		4 h	888	128.8	359	52.1	34	63	
		630°C, 0.5 h, vacuum	868	125.9	341	49.4	35	54	
Derby	U-30 Fe-25	As-rolled	907	131.6	521	75.5	18	18	
U	Si-< 20 Al	450°C, 0.25 h, vacuum	897	130.1	378	54.8	20	15	Average of 2 tests
		4 h	877	127.2	305	44.3	30	40	
		550°C, 0.25 h, vacuum	786	114.0	285	41.4	29	25	
		4 h	792	114.8	317	46.0	42	64	
		630°C, 0.5 h, vacuum	714	103.5	326	47.3	17	18	

Table 12

TENSILE PROPERTIES OF URANIUM-ALUMINUM-IRON AND URANIUM-SILICON-ALUMINUM TERNARY ALLOYS AGED AFTER BETA- AND GAMMA-SOLUTION HEAT TREATMENT (RESULTS ON DERBY URANIUM INCLUDED FOR COMPARISON—AVERAGE OF TWO TESTS, UNLESS OTHERWISE NOTED)

Letter ID	Nominal Composition (ppm)	Solution Treatment Conditions	Aging Treatment Conditions	Tensile Strength (MPa)	Tensile Strength (ksi)	Yield Strength (0.2%) (MPa)	Yield Strength (0.2%) (ksi)	Elongation in 25.4 mm (%)	Reduction in Area (%)	Remarks
H	U-200 Al-300 Fe	730°C, 4 h vacuum, W-Q	350°C, 8 h vacuum	850	123.3	326	47.3	14	10	Single Test
			350°C, 24 h	798	115.8	345	50.0	11	9	
			350°C, 72 h	963	139.6	392	56.8	18	13	
			450°C, 3 h vacuum	859	124.6	388	56.2	18	12	
			450°C, 8 h	880	127.6	397	57.6	12	10	
			450°C, 24 h	794	115.1	381	55.3	10	7	
		800°C, 4 h vacuum, W-Q	350°C, 8 h vacuum	804	116.6	336	48.8	11	10	
			350°C, 24 h	808	117.2	341	49.5	11	10	
			350°C, 72 h	978	141.8	372	54.0	18	12	
			450°C, 3 h	913	132.4	385	55.9	14	9	
			450°C, 8 h	917	133.0	392	56.8	16	9	
			450°C, 24 h	908	131.7	400	58.0	13	10	
I	U-400 Si-200 Al	730°C, 4 h vacuum, W-Q	350°C, 8h vacuum	861	124.9	297	43.1	23	17	
			350°C, 24 h	872	126.4	328	47.5	28	23	
			350°C, 72 h	869	126.0	336	48.7	27	16	
			450°C, 3 h vacuum	865	125.4	291	42.2	26	22	
			450°C, 8 h	864	125.3	303	43.9	23	20	
			450°C, 24 h	864	125.3	315	45.7	24	19	
Derby	U-30 Fe-25 Si-20 Al	730°C, 4 h vacuum, W-Q	350°C, 2 h vacuum	723	104.9	214	31.1	22	24	
			450°C, 24 h vacuum	703	102.0	217	31.5	17	13	
		800°C, 4 h vacuum	350°C, 72 h vacuum	656	95.2	201	29.1	15	15	
			450°C, 24 h vacuum	551	79.9	229	33.2	8	8	

Table 13

HYDROGEN CONTENT OF ALLOYS A, B, C, D, AND DERBY URANIUM
AS A FUNCTION OF PROCESSING PARAMETERS

		Bulk Hydrogen Content After Given Heat Treatment (ppm)								
							Alpha Annealed in Ampoule			
Alloy	Nominal Composition, ppm	As-Processed	800°C 1 h Vacuum, Water-Quench	730°C 1 h Vacuum, Water-Quench	800°, 0.75 h, Vacuum, Water-Quench 350°, 81 h in Ampoule	730°, 1 h, Vacuum, Water-Quench 400°, 48 h in Ampoule	550°, 0.25 h	550°, 4 h	630°, 0.25 h	630°, 4 h
A	U-400 Si	0.44	0.14	0.05	0.26	0.47				
B	U-150 Si-100 Fe	0.44	0.06	0.12						
C	U-200 Si-300 Fe	0.26	0.04	0.08			0.29	0.10	0.25	0.06
D	U-400 Si-200 Fe	0.22	0.10	0.11	0.18	0.41	0.26	0.20	0.28	0.07
Derby	U-25 Si-30 Fe	0.17	0.08	0.13	0.17	0.66				

Table 14

HYDROGEN CONTENT AND TENSILE ELONGATION OF ALLOYS F, G, H, I, AND DERBY URANIUM AS A FUNCTION OF PROCESSING PARAMETERS

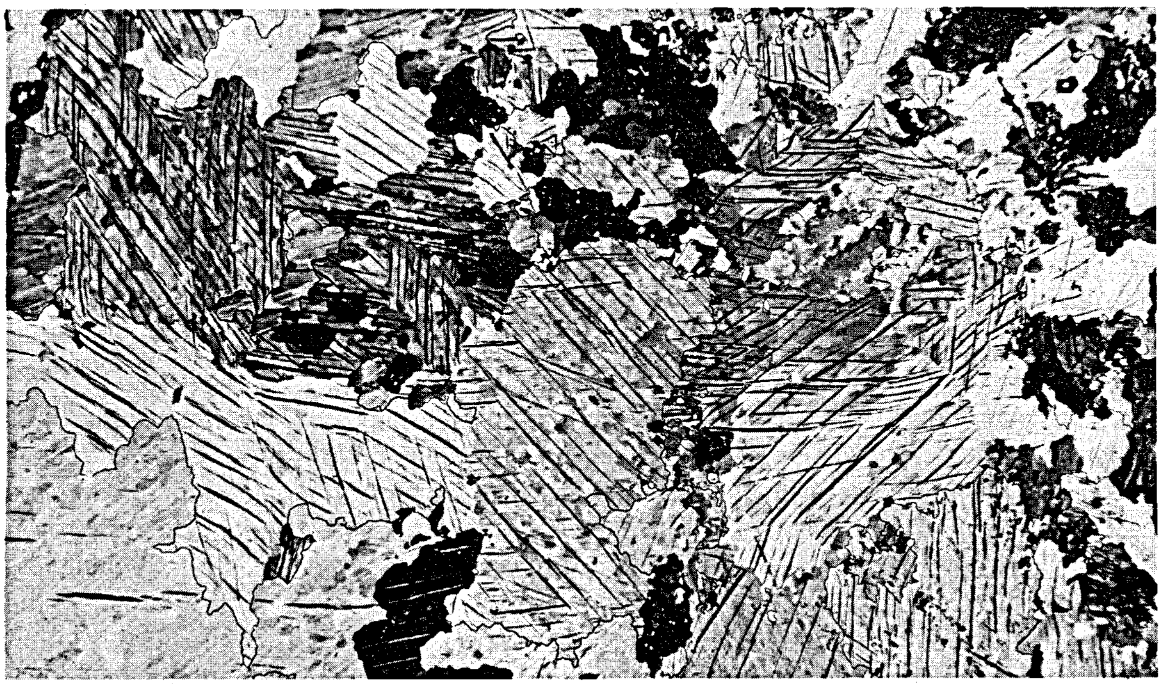
		Bulk Hydrogen Content and Tensile Elongation After Given Heat Treatment																					
	Nominal Composition	As-Processed		800°C, 4 h Vacuum, Water-Quench		730°C, 4 h Vacuum, Water-Quench		800°C, 4 h Vacuum, Water-Quench 350°C, 72 h in Vacuum		800°C, 4 h Vacuum, Water-Quench 350°C, 24 h in Vacuum		730°C, 4 h Vacuum, Water-Quench 350°C, 2 h in Vacuum		730°C, 4 h Vacuum, Water-Quench 350°C, 72 h in Vacuum		730°C, 4 h Vacuum, Water-Quench 450°C, 24 h in Vacuum		450°C, 0.25 h in Vacuum		450°C, 24 h in Vacuum		550°C, 4 h in Vacuum	
Alloy	(ppm)	H (ppm)	El. (%)	H (ppm)	El. (%)	H (ppm)	El. (%)	H (ppm)	El. (%)	H (ppm)	El. (%)	H (ppm)	El. (%)	H (ppm)	El. (%)	H (ppm)	El. (%)	H (ppm)	El. (%)	H (ppm)	El. (%)	H (ppm)	El. (%)
F	U-300 Fe	0.06	25	0.02	33	0.03	37	0.03	16					0.02	15	0.01	37					0.02	41
F	U-300 Fe					0.06	12																
G	U-250 Al	0.05	24	0.04	8	0.03	28			0.02	34			0.02	37							0.02	37
H	U-200 Al-300 Fe	0.02	22			0.01	34	0.02	18					0.02	18	0.02	6	0.04	6			0.02	41
H	U-200 Al-300 Fe					0.01	34																
I	U-400 Si-200 Al	0.03	16	0.01	13	0.03	34					0.01	15	0.02	37					0.03	23	0.02	37
Derby	U-25 Si-30 Fe	0.03	16	0.03	27	0.03	42	0.02	8					0.03	16	0.02	20					0.04	43

Hydrogen analyses are also given in Table 13 for six samples that were aged while encapsulated in quartz or Pyrex ampoules. The analyses all show higher hydrogen concentrations than those of the same alloys vacuum water-quenched from 730° and 800°C, suggesting a pickup of hydrogen (inside the ampoule) during aging. From Table 10, however, specimens aged in ampoules appear to have no lower elongations than those aged in a dynamic vacuum, suggesting that embrittling effects of the aging treatment itself are overriding any such effect due to hydrogen. Also, samples alpha-annealed in ampoules are shown to decrease in hydrogen content with time (Table 13), an effect apparently just the opposite of what occurred in the lower temperature aging treatments.

In Table 14, hydrogen specimens for analysis were from broken tensile bars of Alloys F, G, H, I, and derby uranium, all of which had been heated in argon for processing. These alloys also had been solution heat-treated for extended times of four hours in vacuum and were held in a dynamic vacuum environment for aging and alpha annealing treatments. Hydrogen concentrations were very low, suggesting that if the level is maintained below 0.05 ppm, loss of ductility can probably be attributed to another cause. As an example, the aluminum-containing, iron-free alloys (G and I) show low ductilities after an 800°C solution treatment, but higher ductilities after long aging treatments, all with approximately equal hydrogen contents.

Effect of Alloy Additions on Beta- and Gamma-Quenched Microstructures - With the alloying additions being sufficiently low to be completely in solution in the beta and gamma phases, it is possible that their presence might affect the processes governing the formation of the alpha and beta phases. Such effects might be evidenced in the microstructure as a change in the size and shape of the grains or possibly as a martensitic structure. Upon optically examining the microstructures of the various alloys quenched from the beta (730°C) and gamma (800°C) phases, differences from the usual pure-uranium microstructures have been observed.

Figure 1 is typical of the large, irregular grain size produced upon beta- and gamma-solution treatment and water quenching of high-purity uranium. Figures 2 and 3 of Alloy G indicate that aluminum has no apparent grain refining effect, but, in Figure 4, there is a suggestion that the silicon in Alloy A will produce such an effect. Figures 5 and 6 reveal that 300 ppm of iron will result in grain refinement upon both beta and gamma quenching, a situation which is again noted for Alloy L, U-600 ppm iron, in Figures 7 and 8. However, in these latter two figures, there is evidence of a lessening of grain boundary irregularity, particularly in the gamma-quenched alloy (Figure 8). The microstructure in Figure 9 (Alloy K, U-800 ppm Si-400 ppm Fe) shows about the same grain refinement as Alloy A (Figure 4), but there is a substantially increased grain boundary raggedness as compared with not only the lower silicon Alloy A, but particularly with the iron-containing alloys and derby uranium. Alloy 2K, impure U-800 ppm Si-400 ppm Fe, showed a beta-quenched structure identical with that of Alloy K; but in similarly treated Alloy N, impure U-600 ppm Si-300 ppm Fe, the grain boundaries were less ragged, resembling Alloy A (Figure 4) more than Alloy K (Figure 9).



MS-79-0398-20

Figure 1. DERBY URANIUM VACUUM SOLUTION HEAT-TREATED FOR ONE HOUR AT 730°C AND WATER-QUENCHED. (Polarized light; 100X)

MS-82-0684-8

Figure 2. ALLOY G, URANIUM-250 PPM ALUMINUM, VACUUM SOLUTION HEAT-TREATED FOR FOUR HOURS AT 730°C AND WATER-QUENCHED. (Polarized light; 100X)

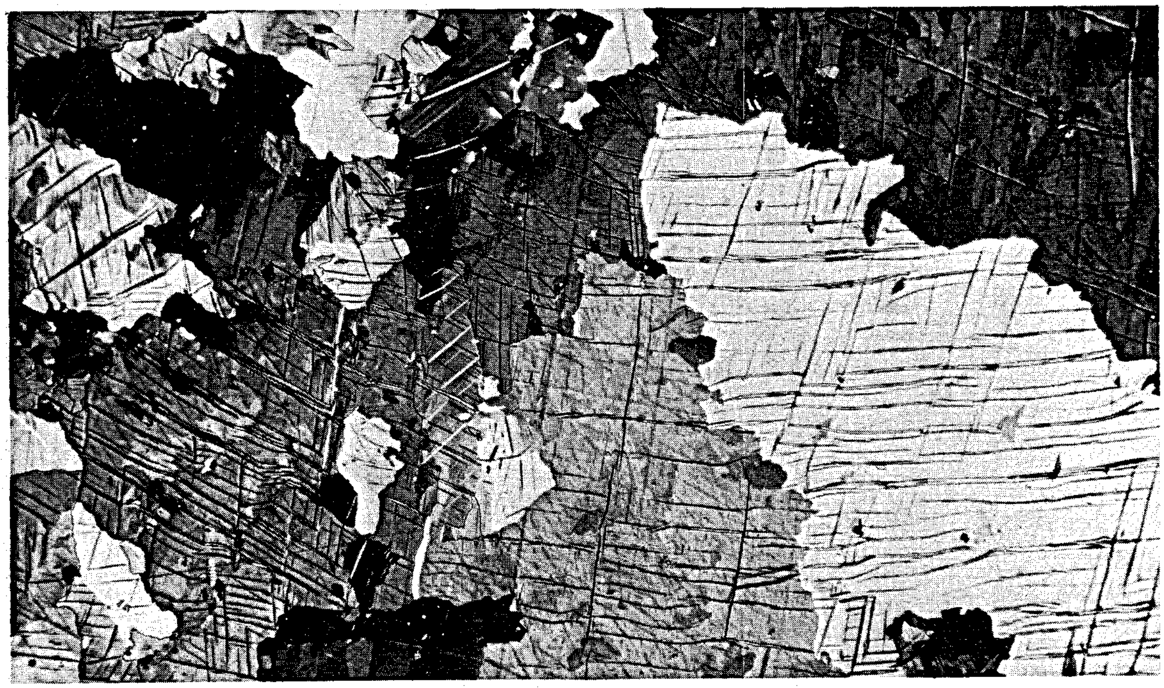


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Figure 3. ALLOY G, URANIUM-250 PPM ALUMINUM, VACUUM SOLUTION HEAT-TREATED FOR FOUR HOURS AT 800°C AND WATER-QUENCHED. (Polarized light; 100X)

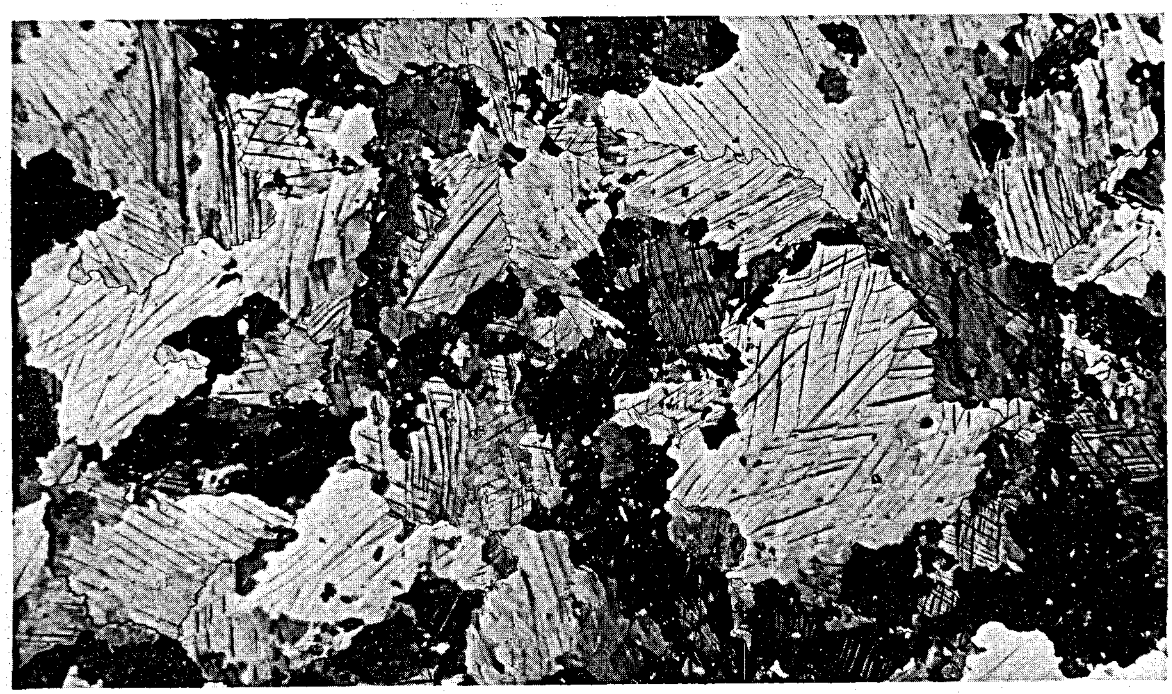

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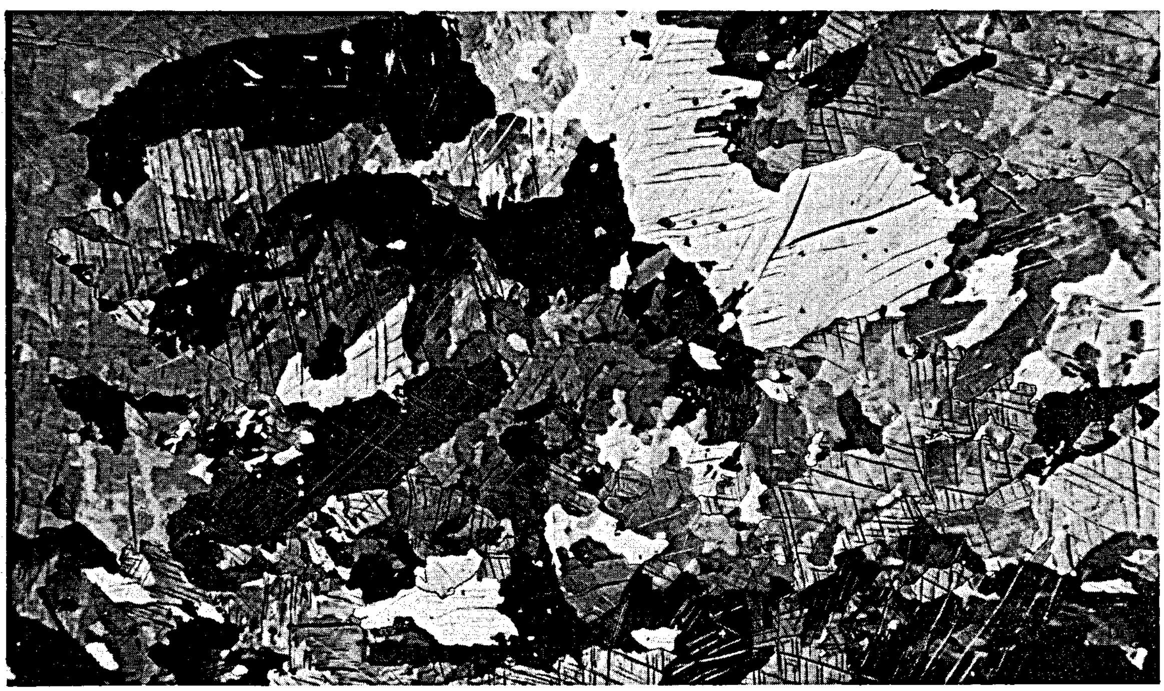
Figure 4. ALLOY A, URANIUM-400 PPM SILICON, VACUUM SOLUTION HEAT-TREATED FOR ONE HOUR AT 730°C AND WATER-QUENCHED. (Polarized light; 100X)

MS-82-0684-5
Figure 5. ALLOY F, URANIUM-300 PPM IRON, VACUUM SOLUTION HEAT-TREATED FOR FOUR HOURS AT 800°C AND WATER-QUENCHED. (Polarized light; 100X)



MS-82-0684-6
Figure 6. ALLOY F, URANIUM-300 PPM IRON, VACUUM SOLUTION HEAT-TREATED FOR FOUR HOURS AT 730°C AND WATER-QUENCHED. (Polarized light; 100X)

MS-82-0684-1

Figure 7. ALLOY L, URANIUM-600 PPM IRON, VACUUM SOLUTION HEAT-TREATED FOR FOUR HOURS AT 730°C AND WATER-QUENCHED. (Polarized light; 100X)

MS-81-0432-4

Figure 8. ALLOY L, URANIUM-600 PPM IRON, VACUUM SOLUTION HEAT-TREATED FOR FOUR HOURS AT 800°C AND WATER-QUENCHED. (Polarized light; 100X)

Effect of Alloy Additions on Alpha-Annealed Microstructure - The microstructure of wrought, alpha-annealed, normal purity uranium is dependent upon the amount of hot and warm working and annealing temperature and time. Varying these factors will influence the rate of recrystallization, grain size, and rate of grain growth; and the structure ultimately produced is one consisting of fine to somewhat coarse equiaxed grains, which is normal for a pure metal. Also, depending upon working-annealing schedules, vestiges of the orientation of the original coarse grains may be evidenced by both clusters of fine or coarse grains and nonuniform recrystallized grain contrast.

Even the small concentrations of alloying elements added to form these alloys are substantially insoluble in the alpha phase and are present at some stage of incoherency as fine particles of various intermetallic compounds. As such, they could act to vary the kinetics of the recrystallization and grain growth processes. Because all of the alloys tested behaved similarly and were subject to the same variety of heat treatment s, a relatively few photomicrographs will serve to illustrate the effects which were observed.

Figures 10-23 were chosen to show selected effects of annealing on the microstructure of wrought uranium with alloying additions. For consistency in comparisons, samples annealed at a single temperature (550°C) have been used for all illustrations. Several conclusions may be drawn from the features in the photomicrographs:

1. Low levels of iron and aluminum, separately or combined, result in an annealing response that is little different (or not different at all) from that of derby uranium. This is illustrated by Figure 14 for derby uranium, Figure 15 for the 300-ppm Fe alloy, Figure 16 for the 250-ppm Al alloy, and Figure 17 for the 200-ppm Al-300 ppm Fe alloy. The only apparent difference is that derby uranium, Figure 14, may show a somewhat larger grain size.

2. Higher iron concentrations retard recrystallization; e.g., the 600-ppm Fe alloy appears to have just become completely recrystallized after 4 h at 550°C in Figure 23.

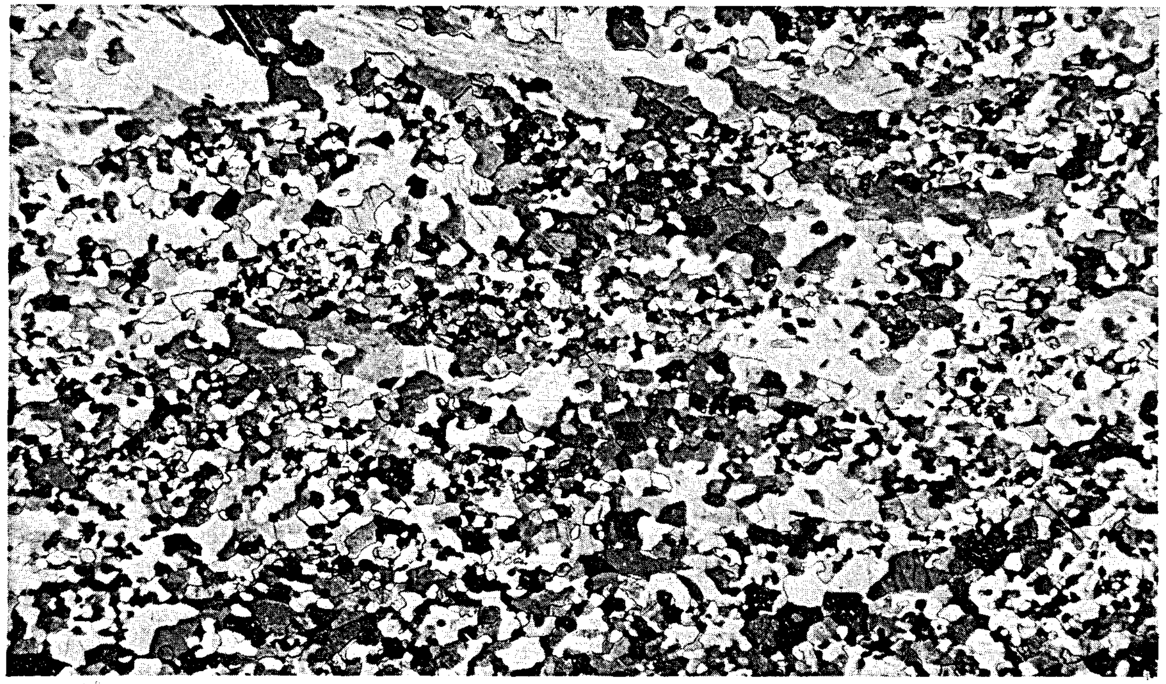
3. Of the three alloying additions, silicon has by far the greatest influence on recrystallization. At low levels, as in the 150 ppm Si-100 ppm Fe alloy in Figure 11, there appears to be little or no effect, but as the concentration increases, recrystallization and grain growth are retarded. In Figure 10, 400-ppm Si has produced these effects. Similarly, in Figures 12 and 13, silicon with iron additions has caused more severe retardation. Figure 18 also reveals the effectiveness of silicon in the presence of aluminum; in Figure 17, somewhat less iron with the same amount of aluminum showed essentially complete recrystallization. Figures 19-22 illustrate a very significant increase in the resistance to recrystallization caused by increased amounts of silicon and iron. Both the impure 600 ppm Si-300 ppm Fe and impure 800 ppm Si-400 ppm Fe alloys showed only incipient recrystallization after 4 h at 500°C, with partial recrystallization in the former (but not in the latter) taking place after seven hours.

MS-81-0432-6

Figure 9. ALLOY K, URANIUM-800 PPM SILICON-400 PPM IRON, VACUUM SOLUTION HEAT-TREATED FOR FOUR HOURS AT 730°C AND WATER-QUENCHED. (Polarized light; 100X)

MS-79-0475-43

Figure 10. ALLOY A, URANIUM-400 PPM SILICON, ANNEALED FOR ONE HOUR AT 550°C. (Incompletely recrystallized; fine grain size.) (Polarized light; 100X)



MS-79-0475-13

Figure 11. ALLOY B, URANIUM-150 PPM SILICON-100 PPM IRON, ANNEALED FOR ONE HOUR AT 550°C. (Almost completely recrystallized.) (Polarized light; 100X)

MS-79-0475-21

Figure 12. ALLOY C, URANIUM-200 PPM SILICON-300 PPM IRON, ANNEALED FOR ONE HOUR AT 550°C. (Incompletely recrystallized.) (Polarized light; 100X)

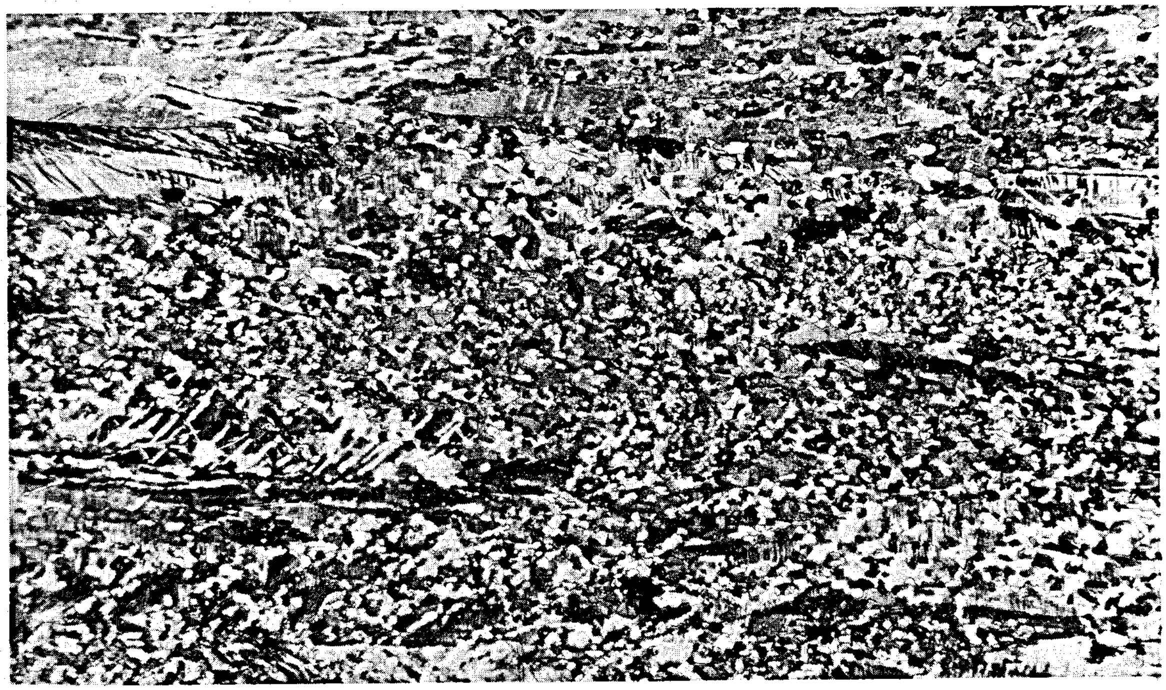


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Figure 13. ALLOY D, URANIUM-400 PPM SILICON-200 IRON, ANNEALED FOR ONE HOUR AT 550°C. (Incompletely recrystallized.) (Polarized light; 100X)

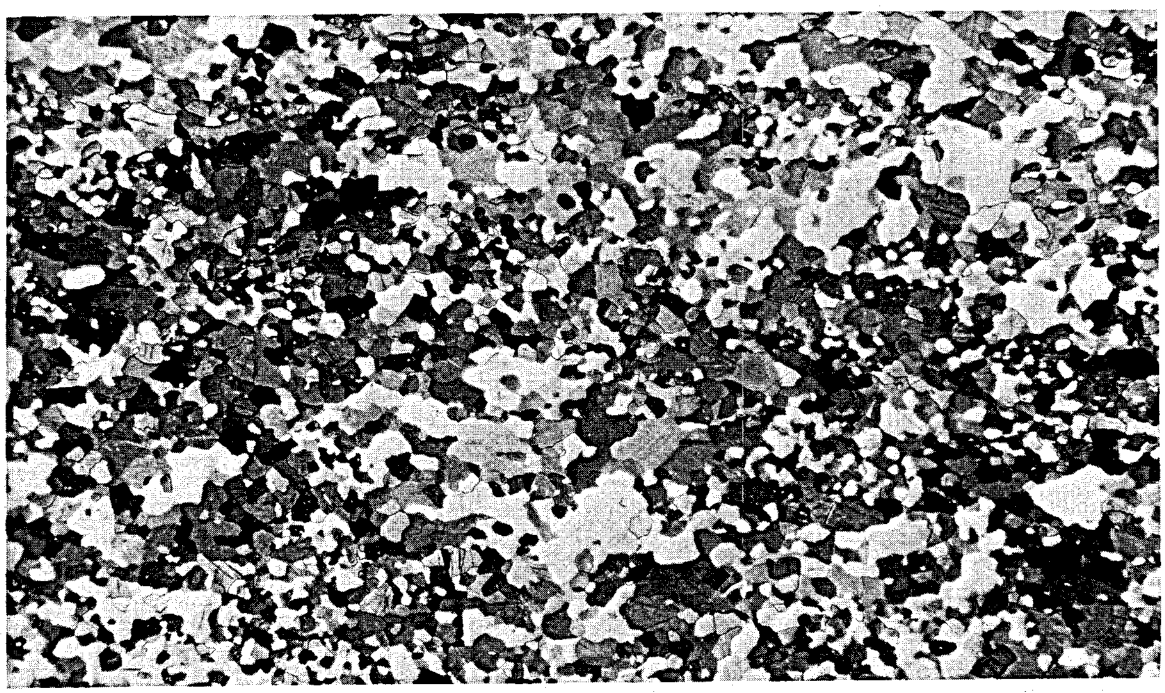

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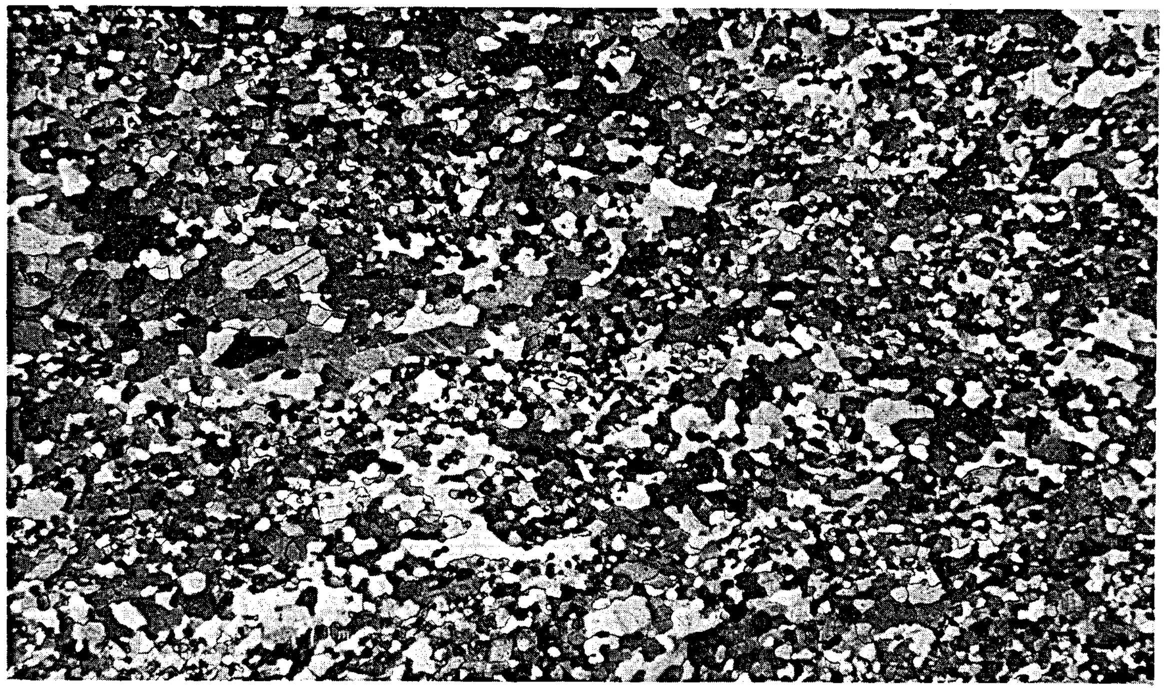
Figure 14. ALLOY E, DERBY URANIUM, ANNEALED FOR ONE HOUR AT 550°C. (Completely recrystallized; some grain growth.) (Polarized light; 100X)



MS-80-0724-2

Figure 15. ALLOY F, URANIUM-300 PPM IRON, ANNEALED FOR ONE HOUR AT 550°C. (Essentially completely recrystallized.) (Polarized light; 100X)

MS-80-0724-5

Figure 16. ALLOY G, URANIUM-250 PPM ALUMINUM, ANNEALED FOR ONE HOUR AT 550°C. (Completely recrystallized.) (Polarized light; 100X)

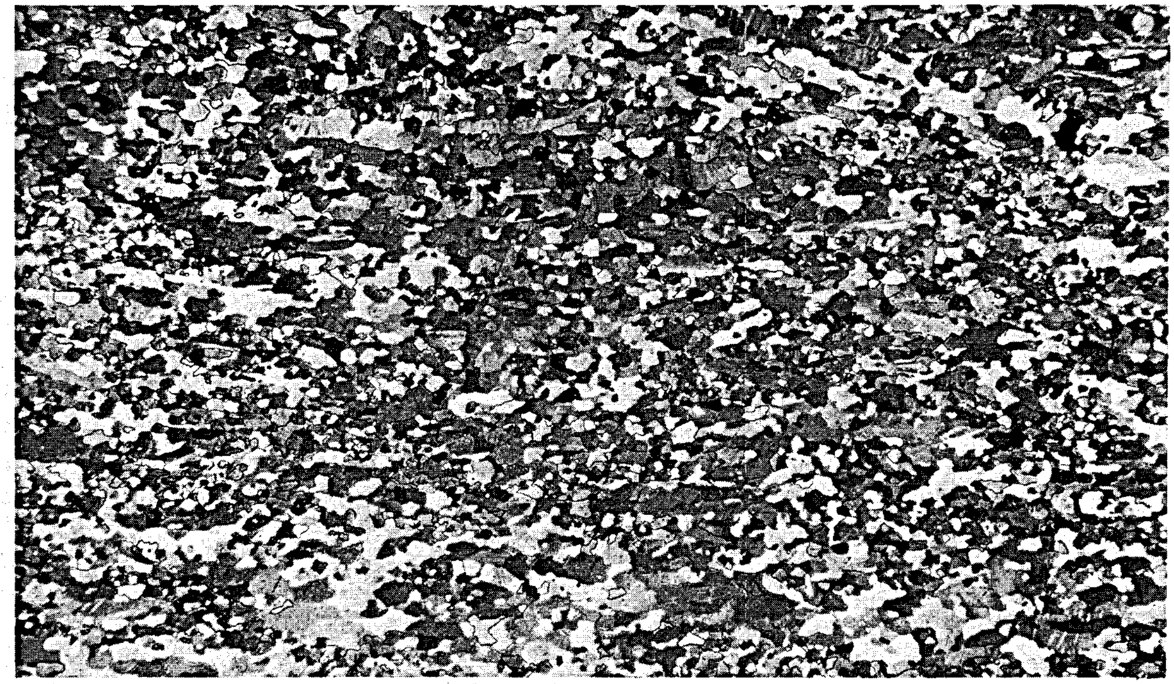


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Figure 17. ALLOY H, URANIUM-200 PPM ALUMINUM-300 PPM IRON, ANNEALED FOR ONE HOUR AT 550°C. (Completely recrystallized.) (Polarized light; 100X)

MS-80-0724-11

Figure 18. ALLOY I, URANIUM-400 PPM SILICON-200 PPM ALUMINUM, ANNEALED FOR ONE HOUR AT 550°C. (Incompletely recrystallized.) (Polarized light; 100X)

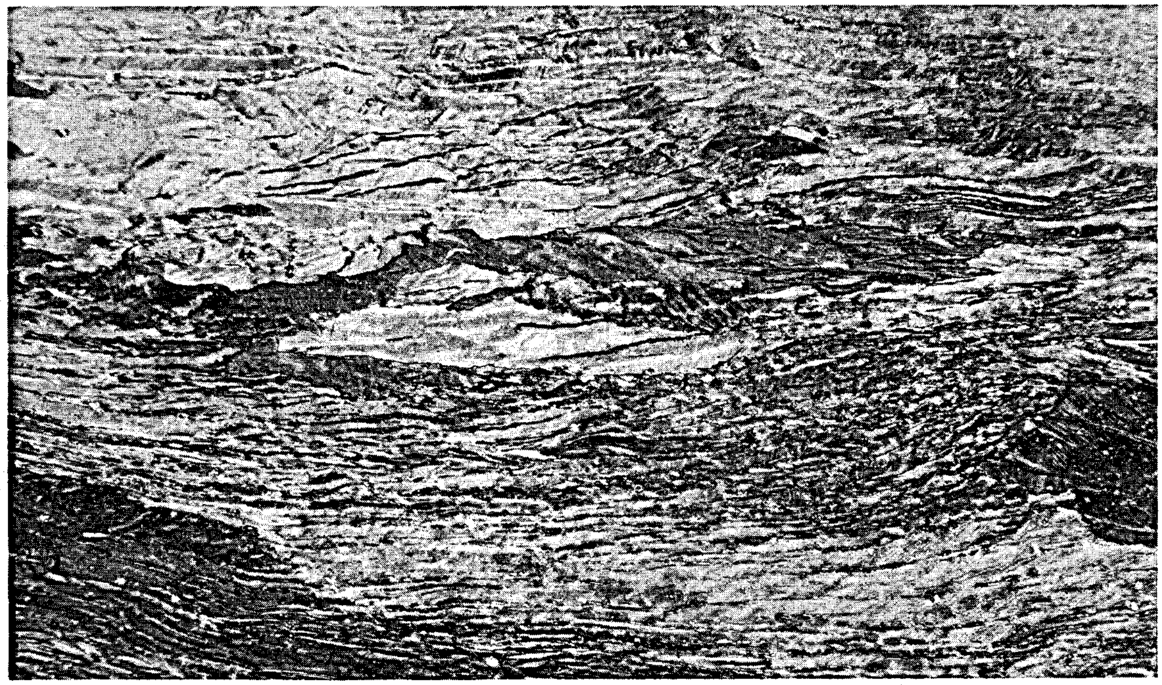

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Figure 19. ALLOY 2K, URANIUM- (IMPURE) 800 PPM SILICON-400 PPM IRON, ANNEALED FOR FOUR HOURS AT 550°C. (Incipient recrystallization.) (Polarized light; 100X)

MS-82-0219-6

Figure 20. ALLOY 2K, URANIUM- (IMPURE) 800 PPM SILICON-400 PPM IRON, ANNEALED FOR SEVEN HOURS AT 550°C. (Incipient recrystallization.) (Polarized light; 100X)

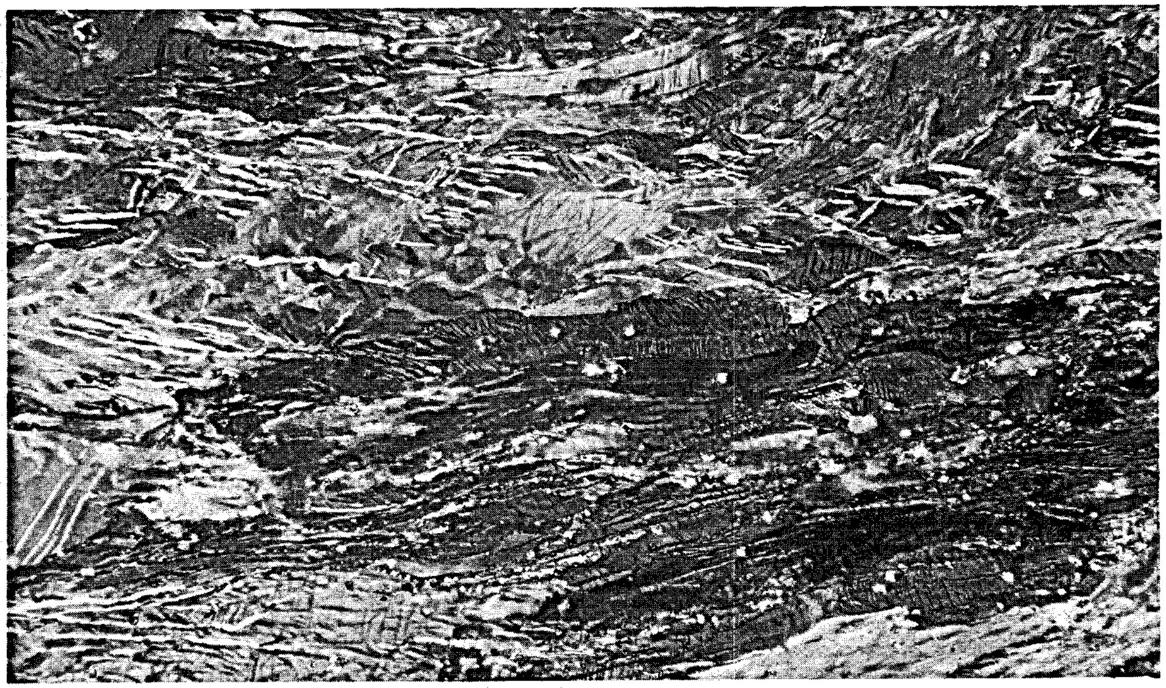

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Figure 21. ALLOY N, URANIUM- (IMPURE) 600 PPM SILICON-300 PPM IRON, ANNEALED FOR FOUR HOURS AT 550°C. (Incipient recrystallization.) (Polarized light; 100X)

MS-82-0219-2

Figure 22. ALLOY N, URANIUM- (IMPURE) 600 PPM SILICON-300 PPM IRON, ANNEALED FOR SEVEN HOURS AT 550°C. (Partial recrystallization.) (Polarized light; 100X)

MS-82-0684-4

Figure 23. ALLOY L, URANIUM-600 PPM IRON, ANNEALED FOR FOUR HOURS AT 550°C. (Completely recrystallized.) (Polarized light; 100X)

DISCUSSION

The alloy analyses shown in Table 3 reveal that there was no difficulty with obtaining a uniform distribution of the alloy additions in the castings. It is believed that a direct addition of the elemental alloying additions to the casting charge might be nearly as effective as using prealloyed arc-melted buttons, particularly for the iron and aluminum; but this was not verified.

There were several observations made in which 800°C gamma-solutioned alloys that were annealed long enough to reduce hydrogen to a low level showed considerably lower tensile elongations than the same alloy solution heat-treated at 730°C in the beta phase. This was noted particularly for the high-silicon alloys, derby uranium, and to some extent, some aluminum-containing alloys. The general difference noted for derby uranium and the aluminum binary alloy is represented by the data in Table 5, and, for the derby uranium, is also supported by a considerable amount of additional data not included here. In the microstructures examined, there appeared to be no visual difference between the beta- and gamma-quenched structures of either derby uranium or the binary U-250 ppm Al, Alloy G.

The reason for the differences in elongation remains obscure, but appears not to be aluminum or grain-size related. Smaller amounts of iron, as in Alloys B, C, D, F, and H, do not appear to promote this problem, as the beta- and gamma-quenched elongations are about equal. A possible reason for this in the high-silicon alloys (even those containing iron) may involve the fact that in the uranium-silicon binary system, the beta-to-gamma transformation involves a peritectoid reaction such that the solubility of silicon in the low-temperature gamma region is less than in the high-beta regime. The available phase diagram[b] shows a solubility of silicon in uranium at 800°C of somewhat less than 900 ppm. If this were slightly in error on the high side, some of the gamma-quenched, higher silicon alloys might contain some embrittling U_3Si intermetallic compound. Possible additional evidence for the partial solubility of silicon may lie in the extremely ragged grain boundaries for the high-silicon Alloy K illustrated in Figure 9. The ragged boundaries may be the consequence of pinning by fine, undissolved intermetallic particles.

Another possible explanation might lie in the quench rate that is required to retain the alloying elements in supersaturated solid solution. There is presently no knowledge that the rate required has or has not been achieved. If a high rate is required, it is possible that it has not always been achieved; consequently, some embrittling intermetallic particles may have been precipitated in the quenching process from 800°C in the gamma phase where the alloy composition is closer to the two-phase $\gamma + U_3Si$ boundary than it is to the $\beta + U_3Si$ boundary at 730°C in the beta phase.

Generally, the aging treatments can be said to have had only a marginal effect upon properties. There were pronounced increases in yield strength for a few alloys, while there was essentially none for others. Without exception, however, aging was accompanied by a decrease in ductility that was greater for higher alloy additions. These decreases in ductility were apparent even when the associated increases in yield strength were small. In general, 400°C appeared to be an optimum aging temperature in that moderate times and temperatures (6 to 24 h) were generally required to produce maximum strength increases. Alloys F and L with 300-ppm and 600-ppm Fe and the Si-Fe Alloys D and N, U-400 ppm Si-200 ppm Fe and U-600 ppm Si-300 ppm Fe, appeared to respond best to aging, but the accompanying loss in ductility was probably excessive with the latter.

The behavior of hydrogen in samples heat-treated in Pyrex or quartz ampoules differed depending upon the temperatures of the heat treatment. In the low-temperature (350° to 400°C) aging treatments for long times, the hydrogen level of the samples was observed to increase with time (Table 13), but the hydrogen level decreased with time at the 550°C alpha-annealing temperature. Although the ampoules were intended to have been evacuated, it is known that, in some, the vacuum was either partial or nonexistent; there could, therefore, have been moisture, which is a source of hydrogen, within the ampoule. The interrelationships among moisture, oxide on the metal, temperature, and time all apparently contributed to the variation in hydrogen content.

The essentially complete ineffectiveness of aluminum in increasing the room-temperature yield stress of uranium was not expected because aluminum is reportedly quite effective in increasing the high-temperature strength of uranium. However, the concentrations that have been reported to be effective at high temperatures were considerably in excess of the 250-ppm aluminum used for the current alloy tests. Alloy G, U-250 ppm Al, showed neither solid-solution strengthening nor an increase in strength upon aging. These results indicate, therefore, that for the given heat treatments, this level of aluminum will not influence the room-temperature tensile properties of uranium and that a considerably increased concentration might be necessary if an effect is to be seen.

The greatest influence of the alloy elements was observed for either beta- or gamma-solution annealing or beta- or gamma-solution annealing plus alpha aging treatments. As mentioned previously, there was not a large variation in properties among the alpha-annealed alloys. In derby uranium, the fine-grain alpha-annealed structure tended to have a yield strength greater than that of the beta- and gamma-quenched material, but in most of the alloys the alpha processing steps and the alpha annealing were essentially overaging treatments which lowered the strength to below that of the quenched alloys. The difference in yield strength between the alloys and derby uranium was therefore reduced, but the strength of the alloys generally remained higher than that of the derby uranium because of the dispersion strengthening resulting from the overaging.

CONCLUSIONS

1. The room-temperature tensile yield strength of high-purity derby uranium may be approximately doubled in 12-mm-thick plate by small additions of silicon and iron, with the alloy being water-quenched from the beta or gamma phase and alpha aged. The alloy elements and levels most desirable are about 400-ppm Si and 200-ppm Fe. For lower alloy levels, the strengths attained are decreased; and for the higher alloys, the strength may increase but the ductility drops rapidly.

2. Higher alloy levels, at least to 800-ppm Si and 400-ppm Fe, can be tolerated in wrought, alpha-annealed materials. However, there is only a limited yield strength increase over derby uranium, with little reduction in ductility because of the precipitation and subsequent growth of the alloy as uranium intermetallic compound particles. Any increase in strength of the alpha-annealed alloy over that of high-purity uranium is probably the consequence of these dispersed particles.

3. Aluminum additions at the 200- to 250-ppm level had essentially no effect on room-temperature tensile properties either as binary additions or in ternary combinations with iron or silicon.

4. The concentration of hydrogen in uranium and alpha alloys of the type discussed here should be 0.05 ppm or less to assure that it will not cause embrittlement with consequent reduction of tensile strength and ductility.

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