THE TENSILE STRENGTHS OF THE COPPER ALUMINUM ALLOYS

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This work was begun in order to establish the relation between the constitution of the aluminum bronzes and their tensile strengths. About twenty years ago, Tetmajer¹ published data on this subject. His work was handicapped seriously by an incomplete knowledge of the internal constitution of these alloys and by the impure aluminum obtainable at that time. Since then we have been able to obtain very pure aluminum and we now know about the constitution of these alloys.

Aside from the results of Tetmajer we have only incomplete data on this subject. Others have made a greater or less number of individual tests without any very definite knowledge of the internal structure.

The Equilibrium Diagram

In Fig. 1 the equilibrium diagram for the copper aluminum series is shown. The coordinates are temperature and percentage composition. The lines in the diagram mark the occurrence of some physical or chemical change. The upper full line or liquidus represents the temperatures above which the alloys are completely melted. The line just below the liquidus or the solidus line represents the temperatures below which the alloys are completely solid. Between these lines melt and crystals exist together. The lines below the solidus represent the temperatures at which some chemical change takes place.

In the diagram the field designated as α represents the concentration and the temperature over which pure α -crystals are stable under equilibrium conditions. The crystals are composed of a solid solution of aluminum in copper. With

¹ Tetmajer: Mitteilungen der Materialprüfungsanstalt IX Heft. Methoden und Resultate des Untersuchung des Aluminums und seiner Abkömmlinge.

more than 91 percent copper there is but one form of crystallization.

β , γ , δ , ϵ and η also are other series of solid solutions, each having a definite crystal form. The η -series is formed by the solution of copper in aluminum. The limits of this field are very narrow when determined from chill-cast alloys. The effect of annealing is very marked over this range. The only compound, CuAl_2 , occurs at 53.7 percent copper.

The fields which are designated as $\alpha + \beta$, $\beta + \gamma$, etc., represent the concentrations and temperatures over which two

Fig. 1

crystalline formations exist together. A field of this sort occurs between any two successive phases.

There are some chemical changes in this series of alloys which are caused by temperature changes. For example, when an 89 percent melt begins to solidify, β -crystals separate from the mother liquor. These continue to separate as the temperature drops. When the dotted line or solidus is reached the melt freezes completely and the mass is composed of pure β -crystals. As the temperature drops still further the β -crystals begin to break down into α . Then over a considerable

temperature α - and β -crystals exist together. At 566° the β -crystals break down completely to form α - and γ -crystals. The β -phase does not appear below 566° under equilibrium conditions. If an 86 percent copper alloy be followed as it cools, changes occur very much as they do in an 89 percent alloy excepting that the β -crystals break down partially to form γ -crystals at the higher temperature. At 566° the β -crystals change over completely to the α - and γ -form. In general all changes in the solid alloys take place most rapidly at the higher temperatures. At ordinary temperatures the reaction rate is practically negligible.

Because of this slow rate at which internal changes take place at ordinary temperatures, an alloy may be heated to a given temperature, and this temperature maintained until equilibrium is established, and then if cooled very rapidly those conditions will remain fixed at ordinary temperatures. If an 88 percent alloy be heated to 700° it will, under equilibrium conditions, be composed of pure β -crystals. If quenched quickly in water from this temperature the β -crystals do not have time to change over into α - and γ -crystals, and at ordinary temperature the alloy retains the same crystalline structure it had at 700° . If allowed to cool more slowly, this alloy undergoes a structural change at 566° . If one wishes to determine the strength of a pure β -alloy an 88 percent alloy, for example, may be annealed at any temperature falling within the β -field until equilibrium is reached and then quenched rapidly. This method is of very general application.

The alloys with a copper content falling between 25 percent and 86 percent are too brittle to be of any technical value. This report is concerned only with the α , $\alpha + \beta$, β , $\beta + \gamma$, $\alpha + \gamma$, η and $\eta + \text{CuAl}_2$ alloys.

The complete diagram is discussed in a separate paper.¹

Artificial Acheson graphite was used in making the moulds for the test pieces. This material was first made use of for this purpose by Shepherd and Upton² in their investi-

¹ Journ. Phys. Chem., **11**, 425 (1907).

² Ibid., **9**, 441 (1905).

gation of the bronzes. The material is almost ideal for the purpose; it is easily worked with ordinary wood-working tools and, being usually homogeneous, takes a very smooth finish.

In making the moulds most of the cutting was done with carpenter's gouges, but the grooves for the test section were finished by means of a sharp screw thread of the desired diameter. The centrifugal sprew devised by Shepherd and Upton was made use of. To obtain this the sprew cavity was cut so as to give a corkscrew shape to the sprew. Thus when the molten metal first enters the mould it is given a whirling motion. This prevents splashing which tends to entrap air. The "airtrap" in the sprew was also used. Vents leading from this trap at the grip end and a large riser at the opposite end of the mould were also provided. Very little local necking occurs in the alloys tested, and standard proportions in the test pieces were not necessary. The diameter of the test section was $3/8$ " and the test length 5". The grip ends were made $3/4$ " in diameter and about 2" long.

Very little trouble was experienced because of occluded gases or trapped air. Some of the pieces when broken showed small amounts of slag, but this was no fault of the moulds. The only difficulty which could be attributed to the moulds was the breaking of the pieces near the ends. This was found to be caused by a too sharp angle in the test piece where grip end and test section meet. This difficulty disappeared when the angle in the mould was sufficiently rounded off. If the halves of the mould were clamped too tightly, shrinking stresses were often found to develop. Best results were obtained when the halves of the mould were clamped together at the sprew end, the riser ends being held together by means of bricks placed on the floor against the sides.

Oxidation was prevented by addition of powdered charcoal and by passing illuminating gas directly into the crucible. This was found to be very effective in excluding oxygen from the melt. By observing these precautions the composition of the melt was not materially affected by oxidation and the amount of slag was kept at a minimum.

A certain amount of slag and oxide is inevitable. After a test piece had been cast the remaining metal was poured into a small ingot mould and the crucible then jarred on the floor. In this manner all the slag was removed from the crucible after each melt. There was a tendency for the slag to cling to the crucible and the metal was almost always entirely free from slag or oxide as it entered the moulds.

The crucibles were made from cylinders of Acheson graphite about $4\frac{1}{2}$ " in diameter. They were turned out by chucking in a lathe and cutting with ordinary machine tools. These crucibles were entirely satisfactory in every respect except durability. No practical means were found for preventing the flames in the furnace from striking the sides of the crucible. When casting the alloys rich in copper, only 15 to 20 pourings could be made from one crucible. With the aluminum-rich alloys many more pourings could be made before the walls burned through.

An ordinary gas furnace equipped with air blast was used for the source of heat. In this a temperature of about 1500° could be obtained. After the furnace became hot the bronzes with the higher melting points could be made quickly. We made it a practice to keep the melt in the furnace the shortest practicable length of time. The time factor has much to do as regards the amount of slag and oxide formed.

The aluminum used in making these alloys was about 99.97 percent. It was presented to us by the Pittsburgh Reduction Company to whom we express our thanks. The copper was the best electrolytic product. In making the aluminum-rich alloys it was found practicable to melt the aluminum and then to add the copper. This method was satisfactory when the surface of the metal aluminum was free from slag and the surface of the copper was bright. Otherwise the copper did not always dissolve readily. In making the copper-rich alloys the copper was always melted first and the aluminum added to the melt. Thorough mixture of the two metals was obtained by stirring with a graphite rod.

It was early noted that when aluminum is added to molten

copper the temperature rises from 150° to 250° . This was first supposed to be caused by the reduction of copper oxide but this is not the reason. If a second amount of aluminum is added a similar rise in temperature is observed. The phenomena must be due to heat of solution. A like phenomenon occurs when copper is dissolved in molten tin.

All of the test-pieces were repeatedly melted. The composition was checked from time to time by analysis and in nearly every case the variation was less than one-tenth of one percent. The number of meltings did not affect the quality of the copper-rich test-pieces, but after the aluminum-rich alloys were melted several times the average percent of good pieces began to drop.

It was not found necessary to heat the moulds before pouring. In most cases good pieces could be obtained by pouring into a cold mould. It was usually necessary to have the first melt somewhat hotter than the succeeding ones.

The temperature of pouring was kept as low as possible as a matter of economy of fuel, crucibles, and time. It is difficult to say definitely at what temperature any given alloy should be poured, for that is a matter requiring experience and judgment. The best copper-rich test-pieces were made by pouring at such a temperature that the metal would about half fill the riser. If poured colder than this the piece would be likely to be bad at the end. If poured much hotter the lower part of the mould would fill first, then the upper part would fill and a seam, noticeable on the outside of the piece but not apparent on the inside, would be formed. This fault was not a serious one but did not add to the appearance of the piece. This phenomenon did not appear in the aluminum-rich alloys at all. The castings obtained were as nearly perfect as could be desired and the surface in most cases was as smooth as if machined. The part of the casting corresponding to the joint in the mould showed a slight fin in most cases. This was not great enough to cause any inconvenience.

The treatment given the pieces recorded as chill-cast was as uniform as possible. The pieces were poured from about the same temperature head above their melting point. They

were then allowed to cool sufficiently to allow handling and then quenched in water. The annealings below 500° were carried on in a specially constructed furnace. This was made of brick and large enough to hold about thirty pieces. It was constructed to give an evenly distributed heat. The heat was supplied from two large Bunsen burners, the hot gas passing along the bottom and sides and then over the top of the enclosed muffle in which the pieces were packed.

The temperature was read by means of a pyrometer and could be kept sufficiently constant for our purpose. For the annealings above 565° a coke-fired assay furnace was used. The time required at these temperatures was short and the assay furnace was very satisfactory. The temperature was observed as in the first furnace. The test-pieces were packed in a fire-clay muffle in powdered graphite. The mouth of the muffle was closed with clay or asbestos paper. The aluminum-rich alloys showed no evidence of oxidation due to annealing. There was a marked tendency for the copper-rich pieces to oxidize, and a stream of illuminating gas was passed into the muffle. At the worst, the film of oxide formed on the surface of the bronze pieces was too thin to cause any trouble. In fact the film could be easily removed with the nail or a knife point.

The aluminum-rich alloys were given but one heat treatment; namely, annealing at 400° . The copper-rich pieces were given two heat treatments; 1st, annealing at 500° for 2 days and 2d, annealing above 565° for 30 minutes. In the latter treatment the pieces were quenched from about 750° . The pieces annealed at or below 500° were not quenched but allowed to cool in the muffle.

A few of the bronzes were cast and allowed to cool slowly in the air, protected from draughts.

Testing

Practically all of the tests were made on a 10,000 pound Olsen machine. This machine was well adapted to our purposes because of its small size and accuracy. The cross section of the test-pieces was small and the actual breaking load was

only in a few cases greater than 10,000 pounds. Because of this nearly all of the pieces broke within the range of the machine.

The mechanism of this machine was worked by hand. This was found to be advantageous in a number of ways.

The method of procedure in testing was briefly as follows:

The diameters of the test-pieces were carefully measured with micrometer calipers reading directly in thousandths of an inch. The smallest diameter was always the one recorded. The pieces were not perfectly circular but varied slightly in diameter at any one section. Three or four diameters were taken and the average of these was recorded. In some cases the fin, formed by the imperfect fitting of the halves of the mould, was filed away. This fin is not large enough, however, to affect the strength of the piece appreciably, so that the filing was generally omitted.

The records of extension were taken by means of a 6-inch drafting compass fitted with fine steel points. It would have been desirable to use an extensometer for this purpose, but none short enough was obtainable. The method finally adopted was to make a fine prick punch mark at the end of the piece uppermost in the machine, and mark the lower end by means of the point of the compass. Making these two marks exactly 5" apart results in a considerable saving of time of computation. It was easy to take the extensions quite accurately in this way. After the beam had been balanced with the desired load, the compass was set to coincide with both marks, and the distance thus measured was read on a steel scale divided in hundredths of an inch; with the aid of a magnifying glass, extensions were thus easily read in thousandths.

A few of the stronger pieces were broken in a 100,000 pound Riehle machine, which has a noticeable throwback. In all the tests run on this machine, the mechanism was run at low speed, and extension readings were taken with the machine running, as the load was raised. By this means, the effect of throwback is eliminated.

In all the tests on both machines, care was taken to have the beam floating at the time of fracture.

After the pieces were broken the halves were removed from the machine and carefully observed. The character of the fracture was recorded in every case. The final extension was obtained by measuring between the marks with the broken ends held together. Care was taken to correct for the opening left between the ends.

Results

The results of the experiments are presented in the accompanying tables, and are shown graphically on the curve sheets. While it might have been desirable to present all the extension data exactly as taken, it would also have been very cumbrous. Extension data are given for representative pieces of every composition tested.

Table I gives records of all the pieces broken, which gave tests which were considered not seriously affected by slag, air holes, or shrinkage strains. Strengths per square inch have been computed from the diameter of the smallest section, measured before testing. Ductilities are computed from the final extension, corrected by adding the initial extension as shown by the stress-strain diagrams, as described later.

Table II gives complete records of the "best" pieces. Since the highest strength obtainable by experiment is the true strength of any material, these records are representative of the various alloys.

Two corresponding sets of curves have been plotted:

Fig. 2 shows the variation of strength and ductility with composition, for the best pieces of the aluminum-rich alloys. Fig. 3 shows the same for the bronzes. The points corresponding to pure copper are taken from the data obtained by Shepherd and Upton in 1904.

Figs. 4-7 are the stress-strain diagrams. The coördinates of such a diagram are, as standardized, load per square inch, as ordinates, and extension from initial length, per inch of length as abscissae. In the present case, since all the pieces were of the same length, actual extensions were used as abscissae.

They are all plotted in accordance with standard practice; that is, zero extension is plotted for initial load, and for the other loads; extensions reckoned from the length at initial load

Fig. 2

Fig. 3

are plotted. The curve is then drawn, and the intercept on the axis of abscissae noted. This intercept is the extension at the initial load, and the curve is then shifted a distance equal to

this intercept, to the right so as to pass through the origin. In all cases the initial load was so small as to give practically no extension.

In all the curves, tables and diagrams, A refers to chill-cast pieces; B to pieces annealed at the lower temperatures; C to pieces annealed above 565° for 30 minutes and water-quenched; D to pieces allowed to cool in the air.

Referring to Fig. 2, both the chill-cast and annealed pieces

Fig. 4

Fig. 5

show a very rapid increase in strength as we go from pure aluminum to the 5 percent alloy. For the chill-cast pieces this increase continues more gradually until we reach the maximum at 20 percent Cu. From here there is a very marked drop to the 25 percent alloy. This may be assigned largely to a peculiarity of the very brittle alloys which we are approaching. When the metal enters the mould, it freezes at the surface, thus forming a skin. When shrinkage takes place,

the skin already formed, being very stiff, cannot yield and a void is formed at the center. It seems almost impossible to prevent this. The annealed pieces show almost a constant strength throughout the range from 5 percent copper to 20 percent copper. A considerable rise in strength is noticed at 25 percent copper, due to the fact that the piece showed no

Fig. 6

Fig. 7

signs of the void at the center, which was so noticeable in the other pieces of this composition.

The curves for ductility show a steady decrease as the copper content increases, rapid at first and then more gradual. This shows very clearly the stiffening effect of the CuAl_2 crystals.

Both the strength and ductility curves show the effect of

annealing quite markedly. For every alloy tested, barring the 25 percent, the annealing reduces strength and increases ductility. As noted before, the 25 percent copper alloys did not show shrinkage voids, so that this apparent exception to the rule is the result of the absence of these voids and not of structural change. Temperature changes do not affect or change the structure in any way except to increase the size of the crystals. The best *dependable* alloy is about 10 percent copper. This corresponds very closely to the edge of the η -field or to the appearance of CuAl_2 crystals as shown in Fig. 1. While some good tests were obtained on alloys with more than 10 percent copper the pieces were by no means uniform.

In the strength curves (Fig. 3) there is very marked increase in strength due to small amounts of aluminum. This is due in all likelihood to two reasons; first, the aluminum removes all traces of copper oxide and produces a sound casting, and second, to the formation of an alloy. From this composition, 99.6 percent copper, the curves for all heat treatments are very nearly coincident and show a gradual rise in strength until the copper content is 92 percent. As the aluminum content is further increased, the curves rise more rapidly and continue to a maximum with 90 percent copper. The curve begins its rapid rise while in the α -field but very close to the edge where β -crystals first appear, the 91 percent copper alloy is largely made up of α -crystals with traces of the β -phase appearing between. Chill-cast alloys with a composition lying in the α - + β -region (see Fig. 1) contain only these two phases as do the same alloys when annealed above 566° . When these same alloys are air-cooled the β breaks down, partly at least, to form the α - and γ -phases. Under these conditions α -, β - and γ -phases are probably present. The maximum strength was found in an 89.9 percent copper chill-cast piece which broke under a load of 108,200 pounds per square inch. The strength drops slightly again at 89 percent copper and still more at 88 percent, where pure β -crystals are still present. These crystals, as shown by the fractures, are very large, even when cooled rapidly,

The annealing has a very noticeable effect on the strength of the bronzes after we pass the 92 percent alloy. At 92 percent, the strength of the pieces annealed above 566° falls considerably below that of chill-cast pieces and those annealed at 500° , which last run very close together. At 91 percent, the strength of the pieces annealed above 566° runs up nearly to that of the chill-cast, while that of the pieces annealed at 500° is quite a little lower. At 90 percent, the chill-cast pieces have a strength intermediate between the strengths of the annealed pieces, the difference being about 10,000 pounds per square inch with those annealed above 566° highest at 98,000 pounds per square inch.

At 89 percent and 88 percent, the pieces annealed at 500° show about 8,000 pounds per square inch less strength than the chill-cast pieces. For these compositions, the pieces annealed above 565° showed up rather more poorly. This is due to the fact that at the temperature of quenching (about 800°), these alloys soften and tend to bend. Also above 600° the crystals grow very rapidly, and this causes the strength of the alloys to be greatly reduced. The 86 percent alloys were the most brittle with which we had to deal; so much so that if allowed to fall to the floor they were almost certain to break. The strengths recorded mean very little, for the pieces nearly always failed by breaking in the grips.

The variation of ductility in this series of alloys is very remarkable. As with the strength, the ductility increases very rapidly at first as aluminum is added, and then more gradually, though still quite rapidly until the 95 percent bronze is reached. It probably reaches a maximum value between 95 percent and 92 percent, at which latter point the "best" piece, chill-cast showed a slightly greater value than at 95 percent. One 95 percent specimen, annealed above 566° , gave 60.8 percent ductility. In passing from 92 percent to 91 percent the ductility drops from 57.5 percent to 16 percent and again to 1.2 percent at 90 percent. This shows clearly the character of the β -crystals which are exceedingly stiff. The ductility from here remains, as far as our observations continued, very nearly constant.

The effect of annealing on the ductility is not very marked. Chill-cast pieces and those annealed above 566° run very closely together from 99.6 percent to 95 percent copper, while the pieces annealed at 500° show a slightly smaller value. At 92 percent, the chill-cast pieces are the best as regards ductility, the annealed pieces showing somewhat smaller values, both close together. At 91 percent copper, the ductility curves are dropping so rapidly that it would be safe to say only that the ductilities lie in the neighborhood of 20 percent or about $1/3$ the value at 92 percent. At 90 percent chill-cast pieces and those annealed above 566° C were very low in ductility, while those annealed at 500° C showed considerably more, 6–8 percent.

A few air-cooled pieces were made—90 percent and 91 percent. These show a rather marked increase in ductility over that of any other specimens of the same composition (see Tables). This gain in ductility does not reduce the strength in the 90 percent alloys, but reduces it slightly in the 91 percent.

From the diagram, Fig. 3, it is evident that the ductility of the β - and γ -crystals is practically zero while the α -crystals have a higher ductility the greater the aluminum content. The ductility appears to decrease at 92 percent before any of the β - or γ -phases has appeared. It is quite possible, however, that the falling off of the ductility with increasing aluminum content is more rapid than is shown by the diagram and that a sample with a copper content of 91.5 percent might have shown as high a ductility as that with 92 percent. No explanation can be offered at present for the fact that annealing appears to decrease the ductility. The opposite effect was observed with the copper-tin bronzes and is quite marked in the aluminum-rich bronzes. At the aluminum end of the series, the ductility of the η -crystals decreases with increasing copper content and becomes very low for CuAl_2 .

The tensile strength of the β bronzes is higher than that of the γ bronzes, while that of the α bronzes increases with decreasing copper content. The maximum strength does not

come at the point where the alloy ceases to be homogeneous. The presence of the stiffer β - or γ -crystals appears to add to the strength of the alloy up to a certain point. A similar phenomenon was observed with the copper-tin bronzes. The essential feature is, however, the amount of the α -crystals present. For this reason the effects due to annealing above and below the inversion point are not so striking as with the copper-tin bronzes. Owing to the greater intrinsic strength of the β -alloys, the maximum strength is greater for the α - and β -alloys than for the α - and γ -alloys. At the aluminum end of the series, the maximum strength occurs in the field for γ and CuAl_2 instead of at the boundary curve.

The character of fractures in the aluminum-rich series varies from the fibrous of pure aluminum to semivitreous in the 20 percent and 25 percent copper alloys. It becomes semi-granular in the 5 percent, quite granular at 10 percent, crystalline at 15 percent. Pure aluminum and the 5 percent alloys generally show a cupped or irregular fracture. The 10 percent and 15 percent alloys are usually quite irregular but never cupped. The 20 percent and 25 percent pieces break almost square across, the 25 percent alloy breaking very much like glass.

The bronzes show a cupped, diagonal or irregular, fibrous fracture, from pure copper to the 92 percent bronze. At 91 percent, the fracture is semi-granular and almost straight across. In the 90 percent, two sorts of fractures are noticeable. All the pieces show a crystalline fracture, those annealed at 500° showing a fairly smooth break, while those chill-cast or annealed above 565° were very irregular and showed large β -crystals. The air-cooled pieces were similar to those annealed at 500° , except that the fracture was somewhat more irregular. The 88 percent and 89 percent pieces show a very irregular, coarsely crystalline fracture and that of the 86 percent bronze is quite vitreous.

Alloys containing 5 percent copper or less show a slight local necking. This is also true of bronzes containing 92 percent copper or more,

The color of the aluminum-rich alloys is uniformly white. The alloys containing 10 percent copper or more show a beautifully smooth surface when cast. Below this amount of copper content, the surface of the castings, while smooth, was not nearly so good. Aside from this difference in smoothness, the appearance of all the alloys of this series is the same.

The bronzes show a greater variation in respect to color. The 99.6 percent bronze is very nearly the color of pure copper but slightly lighter. The 97.5 percent copper alloy has a rich gold color with a pinkish tint. The other copper-rich bronzes have a golden color which becomes lighter as the aluminum content increases. With less than 84 percent copper the yellow color disappears.

Little needs to be said about the stress-strain diagrams. These in themselves are graphical descriptions of the behavior of the various alloys. However, for any one alloy the diagrams are of similar form regardless of the heat treatment. The yield point or elastic limit can not be easily located. This, of course, is true of any unworked metal. In a few cases after the load had been run well up it was reduced to zero and then again applied and the extension readings taken. This gives the elastic lines shown in some of the diagrams. From these lines it would be possible to compute the modulus of elasticity.

In the copper-rich alloys which have a high ductility the breaking load is really much greater than the data imply. The reduction in cross-section is considerable, due to the elongation.

In the tables, the following abbreviations have been used:

Both tables, with respect to heat treatment,

A—chill-cast,

B—for aluminum-rich alloys, annealed at 400°, for aluminum bronzes, annealed 2 days at 500°,

C—annealed above 566°, 30 minutes, and quenched in water,

D—air-cooled.

Table I, with respect to character of fracture,

G—good,

B—bad,

f.—fibrous,

gr.—granular,

cr.—crystalline,

v.—vitreous,

ox.—oxide,

irr.—irregular,

diag.—diagonal,

tr.—trace.

Table II,

P = load in pounds,

f = load per sq. in., pounds,

L = length between marks,

γ = extension.

TABLE I

Aluminum-rich alloys

0 percent Cu				
Heat treatment	Diam.	Ultimate strength	Percentage extension	Fracture
A.....	0.382	13750	15.4	G—f. cup, necked.
	0.398	13320	17.0	"
B, 4 days...	0.381	13130	18.0	"
	0.375	13400	18.0	"

5 percent Cu				
A.....	0.383	23750	7.5	B—gr. cr. cup tr. slag
	0.399	23780	9.0	B—gr. cr. irr. tr. slag
	0.394	25750	6.4	G—gr. irr
B, 4 days...	0.376	21050	7.0	"
B, 6 days...	0.386	23620	8.6	G—f. cup
	0.376	23200	8.4	G—f. diag.

Table I—(Continued)

10 percent Cu				
Heat treatment	Diam.	Ultimate strength	Percentage extension	Fracture
A.....	0.376	29150	1.7	G—gr. st
	0.376	28600	1.8	B—tr. ox.
	0.392	27500	1.2	B—tr. ox.
	0.380	26450	2.0	G—gr.
B, 4 days...	0.375	21650	2.78	G—gr. cr. irr.
B, 3 days...	0.374	23300	2.8	"
	0.380	20650	2.6	"
	0.377	21750	2.86	"
	0.375	22150	3.36	"
15 percent Cu				
A.....	0.377	29300	1.4	G—gr. cr. irr.
	0.380	29200	0.9	"
	0.379	29700	0.9	"
	0.374	29550	0.84	"
B, 6 days...	0.379	22550	1.8	B—gr. tr. slag
	0.378	22500	2.0	G—gr.
	0.388	23600	2.0	"
	0.377	23000	1.9	"
20 percent Cu				
A.....	0.376	32700	0.15	G—cr. semi-v.
	0.373	31050	0.4	"
	0.375	26950	0.5	"
(Test length, 6 inches)				
B, 4 days...	0.379	22800	1.1	"
	0.379	23050	1.2	"
25 percent Cu				
A.....	0.377	17850	0.14	G—v. void
	0.381	18950	0.4	"
	0.381	20000	..	"
	0.382	22700	0.18	"
B, 4 days...	0.379	25650	0.6	G—v. solid
	0.378	13900	..	G—v. void

Table I—(Continued)

Aluminum bronzes				
99.6 percent Cu				
Heat treatment	Diam.	Ultimate strength	Percentage extension	Fracture
A.....	0.381	28550	33.6	G—f. neck
	0.382	28800	32.8	"
	0.379	29450	33.8	"
B.....	0.375	28900	29.2	"
	0.376	28450	32.6	"
	0.374	28450	35.0	"
	0.375	25600	20.8	G—semi-cr.
	0.376	28100	32.0	G—f. cr. semi-cup
C.....	0.376	28100	33.0	B—f. small air-hole
	0.374	28600	33.0	G—f. cup neck
	0.378	28900	32.0	" "
	0.376	27800	32.0	B—small air-hole
97.5 percent Cu				
A.....	0.378	35600	40.0	G—f. diag. neck
	0.378	35450	37.2	G—f. cup neck
	0.378	35800	38.0	"
B.....	0.371	32500	23.0	"
	0.376	35100	36.6	"
	0.380	34100	27.2	G—cr.
C.....	0.378	35450	37.0	G—f. cup neck
	0.376	35300	38.0	"
	0.377	36000	39.2	"
95 percent Cu				
A.....	0.381	46850	56.6	G—f. semi-cup neck
	0.377	46600	59.7	"
	0.375	45250	47.0	B— " tr. slag
B.....	0.378	43000	57.2	" "
	0.374	46250	46.4	G—f. cr. cup neck
	0.377	47300	49.6	"
	0.375	45250	44.4	B— " small blow-hole
	0.378	45750	54.4	B— " tr. slag
C.....	0.377	47200	58.4	G—f. diag. neck
	0.374	47000	60.8	G—f. cup neck
	0.377	47650	53.8	"
	0.375	47000	54.4	"

Table I—(Continued)

92 percent Cu				
Heat treatment	Diam.	Ultimate strength	Percentage extension	Fracture
A.....	0.386	66200	57.6	B—f. cup tr. slag
	0.378	65400	43.6	G—f. cup neck
	0.372	63000	47.0	B— " tr. slag
B.....	0.373	56500	42.0	B— " tr. slag
	0.371	64500	53.0	G—f. diag. neck
	0.368	61900	48.4	G—f. cup neck
C.....	0.372	57900	43.4	B— " tr. slag
	0.376	45100	30.0	B—f. cr. tr. slag
	0.372	50500	38.1	G—f. irr.
	0.369	56350	52.0	G—f. cup neck
	0.370	58600	51.6	"
91 percent Cu				
A.....	0.372	84200	26.0	B—gr. st. air-hole
	0.369	81800	16.0	G—gr. st.
	0.372	86200	16.2	"
	0.374	76100	25.4	B " tr. slag
B.....	0.370	74100	24.0	G—gr. st.
	0.372	73400	23.4	"
	0.376	67250	17.2	G—cr. gr.
	0.374	69700	17.1	"
C.....	0.374	81000	24.2	B— " air hole
	0.375	74700	23.6	B— " tr. slag
	0.370	81800	30.0	G—f. cr.
	0.369	83200	22.6	"
D.....	0.372	74250	42.8	"
	0.373	74700	42.4	"
90 percent Cu				
A.....	0.371	88000	1.0	G—cr. st.
	0.370	83700	1.2	"
	0.372	79200	6.4	"
B.....	0.368	78700	5.6	"
	0.365	72000	8.0	"
	0.374	73000	16.4	"
	0.369	70100	7.2	"
C.....	0.374	97500	1.2	G—cr. very irr.
	0.369	98000	1.3	"
C.....	0.363	98400	1.3	G—cr. very irr.
D.....	0.367	85900	18.4	G—cr. irr.
	0.361	86900	23.6	"
	0.369	86800	20.4	"
	0.369	86250	18.8	"

Table I—(Continued)

89 percent Cu				
Heat treatment	Diam.	Ultimate strength	Percentage extension	Fracture
A.....	0.374	83500	1.8	B—coarsely cr. air hole
	0.375	79600	2.0	G—coarsely cr.
	0.373	86900	1.8	"
B.....	0.368	79800	1.2	"
	0.380	59800	0.6	"
	0.371	57700	0.6	"
	0.373	65700	..	"
C.....	0.372	55200	0.24	G—"
88 percent Cu				
A.....	0.371	78600	1.4	G—very coarsely cr.
	0.374	56200	1.2	"
	0.372	41450	0.8	B— " bent
B.....	0.372	71350	0.4	G— " "
	0.371	64600	0.5	" "
	0.371	60100	0.5	" "
	0.373	64000	0.3	" "
C.....	0.371	23100	1.2	B—very large crystals, bent
	0.368	37600	1.2	These pieces showed
	0.372	29500	1.4	tracks developed in heat treatment.
86 percent Cu				
B.....	0.372	17300	..	Vitreous
C.....	0.373	28800	..	Vitreous
All the other pieces of this composition broke in the grips.				
91.6 percent Cu				
C.....	0.369	85500	27.5	G—gr. f. cr.
89.8 percent Cu				
A	0.370	108200	1.25	G—cr. irr.

TABLE II

Aluminum-rich Alloys

Aluminum—A, $d = 0.382$

Aluminum—B, $d = 0.375$

P	f	L	λ	P	f	L	λ
200	1750	4.987	0	200	1810	5.003	0
400	3500	4.991	0.004	400	3620	5.008	0.005
600	5250	5.016	0.029	600	5430	5.037	0.034
800	7000	5.061	0.074	800	7240	5.084	0.081
1000	8750	5.118	0.131	1000	9050	5.132	0.129
1300	11350	5.266	0.279	1200	10860	5.231	0.228
1460	12750	5.457	0.470	1400	12670	5.440	0.437
1580	13750	5.759	0.772	1480	13400	5.90	0.897

5 percent Cu—A, $d = 0.394$

5 percent Cu—B, $d = 0.386$

300	2460	5.000	0	300	2560	5.000	0
600	4830	5.008	0.008	600	5130	5.006	0.006
900	7390	5.011	0.011	900	7700	5.011	0.011
1200	9850	5.014	0.014	1200	11350	5.038	0.038
1500	12300	5.031	0.031	1400	12000	5.061	0.061
1800	14800	5.059	0.059	1600	13700	5.082	0.082
2000	16400	5.084	0.084	1800	15400	5.111	0.111
2200	18100	5.112	0.112	2000	17100	5.152	0.152
2500	20500	5.168	0.168	2200	18800	5.200	0.200
2800	23000	5.239	0.239	2400	20500	5.255	0.255
3000	24600	5.289	0.289	2600	22200	5.320	0.320
3120	25750	5.320	0.320	2760	23600	5.430	0.430

10 percent Cu—A, $d = 0.376$

10 percent Cu—B, $d = 0.374$

200	1800	5.000	0	300	2730	5.002	0
600	5400	5.004	0.004	600	5460	5.006	0.004
1000	9000	5.010	0.010	900	8190	5.012	0.010
1200	10800	5.011	0.011	1200	10920	5.021	0.019
1400	12600	5.013	0.013	1500	13650	5.039	0.037
1600	14400	5.020	0.020	1800	16400	5.054	0.052
1800	16200	5.028	0.028	2000	18200	5.070	0.068
2000	18000	5.036	0.036	2200	20000	5.090	0.088
2200	19800	5.049	0.049	2400	21850	5.117	0.115
2400	21600	5.060	0.060	2500	22700	5.129	0.127
3240	29150	5.120	0.120	2560	23300	5.140	0.138

Table II—(Continued)

15 percent Cu—A, $d = 0.379$				15 percent Cu—B, $d = 0.388$			
P	f	L	λ	P	f	L	λ
300	2660	5.002	0	300	2540	5.002	0
600	5320	5.005	0.003	900	7620	5.009	0.007
900	7980	5.007	0.005	1200	10160	5.020	0.018
1500	13300	5.012	0.010	1800	15240	5.040	0.038
2000	17700	5.018	0.016	2000	17100	5.050	0.048
2500	22080	5.022	0.020	2200	18650	5.059	0.057
2700	23900	5.025	0.023	2400	20300	5.071	0.069
3000	26600	5.030	0.028	2600	22050	5.086	0.084
3350	29700	5.045	0.043	2800	23730	5.100	0.098
20 percent Cu—A, $d = 0.376$				20 percent Cu—B, $d = 0.377$			
P	f	L	λ	P	f	L	λ
500	4500	5.002	0	500	4460	5.010	0
1000	9000	5.006	0.004	1000	8900	5.014	0.004
1500	13500	5.009	0.007	1500	13400	5.022	0.012
2000	18000	5.012	0.010	2000	17850	5.042	0.032
2500	22500	5.016	0.014	2400	21400	5.059	0.049
3640	32700	5.023	0.021	2580	23050	5.060	0.050
25 percent Cu—A, $d = 0.381$				25 percent Cu—B, $d = 0.379$			
P	f	L	λ	P	f	L	λ
500	4400	5.010	0	500	4400	5.002	0
1000	8750	5.016	0.006	1000	8850	5.005	0.003
1500	13140	5.020	0.010	1500	13250	5.011	0.009
1800	16000	5.022	0.012	1800	15950	5.016	0.014
2000	17500	5.025	0.015	2000	17700	5.020	0.018
2160	18950	5.028	0.018	2300	20350	5.023	0.021
				2500	22100	5.029	0.027
				2800	24600	5.032	0.030
				2900	25600	5.033	0.031
Aluminum Bronzes							
99.6 percent Cu—A, $d = 0.379$				99.6 percent Cu—C, $d = 0.378$			
P	f	L	λ	P	f	L	λ
500	4420	5.020	0	500	4460	5.018	0
1000	8840	5.130	0.110	1000	8920	5.110	0.092
1500	13280	5.230	0.210	1500	13380	5.230	0.212
2000	17680	5.410	0.390	2000	17840	5.400	0.382
2500	22100	5.620	0.600	2400	21400	5.562	0.534
2800	24800	5.790	0.770	2600	22300	5.662	0.644
3000	26500	5.870	0.850	2800	25000	5.850	0.832
3100	27400	6.050	1.030	3000	26800	5.972	0.954
3200	28400	6.190	1.110	3100	27700	6.060	1.042
3300	29200	6.480	1.460	3200	28600	6.260	1.242
3320	29400	6.690	1.670	3250	29000	6.600	1.582

Table II—(Continued)

99.6 percent Cu—B, $d = 0.375$							
P		f		L		λ	
500		4525		5.010		0	
1000		9050		5.058		0.048	
1500		13575		5.178		0.168	
1800		16300		5.278		0.268	
2000		18100		5.330		0.320	
2300		20800		5.550		0.530	
2500		22600		5.586		0.576	
2700		24400		5.700		0.690	
2900		26300		5.906		0.896	
3000		27200		6.025		1.015	
3100		28000		6.143		1.133	
3190		28900		6.460		1.450	

97.5 percent Cu—A, $d = 0.378$				97.5 percent Cu—B, $d = 0.376$			
1500	13400	5.000	0	1000	9000	5.04	0
2000	17850	5.150	0.150	1500	13500	5.120	0.080
2500	22300	5.312	0.312	2000	18000	5.300	0.260
3000	26800	5.493	0.493	2500	22270	5.470	0.430
3500	31200	5.770	0.770	3000	27000	5.660	0.620
3800	33900	6.018	1.018	3500	31500	5.990	0.950
4000	35700	6.370	1.370	3750	34800	6.250	1.210
4020	35900	6.900	1.900	3890	35000	6.830	1.790

97.5 percent Cu—C, $d = 0.377$							
P		f		L		λ	
1500		13450		5.120		0	
3000		26880		5.650		0.530	
3500		31400		5.710		0.590	
3750		33600		6.130		1.010	
4000		35850		6.520		1.400	
* *		*		*		* *	
1000		8930		6.629		1.503	
2000		17360		6.637		1.515	
3000		26790		6.640		1.520	
4000		35720		6.702		1.582	
4020		35900		6.960		1.840	

Table II—(Continued)

95 percent Cu—A, $d = 0.381$				95 percent Cu—B, $d = 0.377$			
P	f	L	λ	P	f	L	λ
1000	8770	5.012	0	1000	8960	5.011	0
1300	11400	5.042	0.030	1500	13440	5.060	0.049
1600	14030	5.130	0.118	2000	17920	5.290	0.279
2000	17540	5.282	0.270	2500	22400	5.500	0.489
2500	21900	5.495	0.483	3000	26900	5.695	0.684
3000	26300	5.700	0.688	3500	31400	5.908	0.897
3500	30700	5.908	0.896	4000	35800	6.198	1.187
4000	35100	6.190	1.178	4500	40350	6.525	1.514
45 0	39500	6.537	1.525	5000	44800	7.050	2.039
5000	43800	7.070	2.058	5200	46600	7.338	2.327
*	*	*	*	5280	47300	7.480	2.469
0	0	7.030	2.018	95 percent Cu—C, $d = 0.377$			
1000	8770	7.040	2.028	1000	8960	5.010	0
2000	17540	7.050	2.038	2000	17920	5.270	0.260
3000	26300	7.060	2.048	2500	22400	5.470	0.460
4000	35100	7.070	2.058	3000	26900	5.670	0.660
5000	43800	7.082	2.070	3500	31400	5.890	0.880
5200	45500	7.360	2.348	4000	35800	6.190	1.180
5350	46900	7.830	2.818	4400	39400	6.470	1.460
				4800	43000	6.820	1.810
				5000	44800	7.020	2.010
				5200	46600	7.370	2.360
				5300	47500	7.690	2.680
				5320	47650	7.750	2.740
92 percent Cu—A, $d = 0.386$				92 percent Cu—A, $d = 0.371$			
1000	8550	5.010	0	1000	9250	5.007	0
2000	17100	5.020	0.010	2000	18500	5.016	0.009
3000	25650	5.100	0.090	2500	23050	5.072	0.065
3500	29900	5.190	0.180	3000	27750	5.191	0.184
4000	34200	5.320	0.310	3500	32400	5.340	0.333
4500	38400	5.450	0.440	4500	41600	5.715	0.780
5000	42700	5.575	0.565	5500	50800	6.160	1.160
5500	47000	5.710	0.700	6000	55500	6.430	1.423
6000	51300	5.870	0.860	6500	60100	6.886	1.879
6500	55550	6.075	1.065	6800	62900	7.230	2.223
7000	59800	6.310	1.300	6980	64500	7.650	2.643
7500	64050	6.640	1.630				
7750	66200	6.880	1.870				

Table II—(Continued)

92 percent Cu—C, $d = 0.370$			
P	f	L	λ
1000	9300	5.008	0
2000	18600	5.048	0.040
3000	27900	5.400	0.392
3500	32500	5.630	0.622
4000	37200	5.878	0.870
4500	41800	6.160	1.152
5000	46500	6.310	1.302
5500	51150	6.620	1.612
5750	53500	6.820	1.812
6000	55800	7.007	1.999
6250	58100	7.420	2.412
6300	58500	7.580	2.572

91 percent Cu—A, $d = 0.372$				91 percent Cu—A, $d = 0.370$			
1000	9210	5.000	0	1000	9300	5.004	0
2000	18420	5.003	0.003	2000	18600	5.008	0.004
3000	27630	5.015	0.015	3000	27900	5.020	0.016
4000	36840	5.040	0.040	3500	32550	5.040	0.036
5000	46050	5.100	0.100	4000	37200	5.080	0.076
6000	55260	5.200	0.200	4500	41800	5.150	0.146
6500	59800	5.265	0.265	5000	46500	5.225	0.221
7000	64470	5.350	0.350	5500	51100	5.312	0.308
7500	69000	5.420	0.420	6000	55800	5.470	0.466
8000	73500	5.520	0.520	6500	60500	5.612	0.608
8500	78200	5.620	0.620	7000	65100	5.790	0.786
9000	82800	5.740	0.720	7500	69750	6.015	1.009
9378	86400	5.810	0.810	7960	74000	6.200	1.196

Table II—(Continued)

91 percent Cu—C, $d = 0.369$				91 percent Cu—D, $d = 0.372$			
P	f	L	λ	P	f	L	λ
2000	18700	5.010	0	1000	9220	5.005	0
3000	28050	5.030	0.020	2000	18400	5.010	0.005
4000	37400	5.065	0.050	3000	27650	5.100	0.095
4500	42060	5.110	0.100	4000	36900	5.198	0.193
5000	46750	5.165	0.155	5000	46100	5.410	0.405
5500	51450	5.275	0.265	6000	55300	5.680	0.675
6000	56100	5.330	0.320	7000	64500	6.080	1.075
*	*	*	*	8000	73500	6.755	1.750
1000	9350	5.457	0.429	8180	74700	7.120	2.115
2000	18700	5.439	0.447				
3000	28050	5.464	0.454				
4000	37400	5.470	0.460				
5000	46750	5.474	0.464				
6000	56100	5.480	0.470				
7000	65450	5.506	0.496				
8000	74800	5.730	0.720				
8500	79500	5.955	0.945				
8900	83200	6.130	1.120				
90 percent Cu—A, $d = 0.371$				90 percent Cu—B, $d = 0.368$			
1000	9250	5.000	0	1000	9410	5.000	0
2000	18500	5.001	0.001	2000	18820	5.001	0.001
3000	27800	5.010	0.010	3000	28230	5.003	0.003
4000	37000	5.015	0.015	4000	37640	5.010	0.010
5000	46200	5.018	0.018	5000	47000	5.023	0.023
6000	55500	5.020	0.020	6000	56460	5.070	0.070
7000	64700	5.030	0.030	6500	61200	5.130	0.130
8000	74000	5.040	0.040	7500	70500	5.210	0.210
9000	83200	5.050	0.050	8000	75280	5.260	0.260
9500	88000	5.060	0.060	8350	78500	5.280	0.280
90 percent Cu—D, $d = 0.361$				90 percent Cu—C, $d = 0.363$			
2000	19560	5.010	0	1000	9680	5.001	0
3000	29340	5.020	0.010	3000	19360	5.009	0.008
4000	39120	5.080	0.070	4000	38700	5.012	0.011
5000	48900	5.220	0.210	5000	49400	5.016	0.015
6000	58680	5.400	0.390	6000	58700	5.025	0.024
7000	68400	5.600	0.590	7000	67700	5.031	0.030
8500	83100	6.060	1.050	8000	77400	5.037	0.036
8900	87000	6.180	1.170	9000	87000	5.043	0.042
				10000	96800	5.062	0.061
				10200	98560	5.066	0.065

Table II—(Continued)

89 percent Cu—A, $d = 0.373$				89 percent Cu—B, $d = 0.368$			
P	f	L	λ	P	f	L	λ
2000	18600	5.005	0	2000	18800	5.004	0.0
4000	37200	5.010	0.005	4000	37600	5.010	0.006
6000	55800	5.018	0.013	6000	56400	5.020	0.016
7000	64100	5.030	0.025	8000	75200	5.050	0.046
8000	74400	5.038	0.038	8500	80000	5.060	0.056
9000	82500	5.050	0.045				
9500	86900	5.090	0.085				

89 percent Cu—C, $d = 0.372$

P	f	L	λ
2000	18400	5.000	0
4000	36800	5.010	0.010
6000	55200	5.012	0.012

88 percent Cu—A, $d = 0.371$ 88 percent Cu—B, $d = 0.372$

2000	18500	5.005	0	2000	18400	5.002	0
3000	27750	5.010	0.005	3000	27650	5.003	0.001
4000	37000	5.017	0.012	4000	36800	5.008	0.006
5000	46250	5.020	0.015	5000	46100	5.010	0.008
6000	55500	5.030	0.025	6000	55300	5.013	0.011
7000	64750	5.040	0.035	7000	64500	5.018	0.016
8000	74000	5.070	0.065	7770	64800	5.020	0.018
8500	78600	5.070	0.065				

88 percent Cu—C, $d = 0.368$

P	f	L	λ
2000	18800	5.028	0
3000	28200	5.040	0.012
4000	37600	5.060	0.032

Conclusions

In summing up the results of our investigations, we may draw the following conclusions:

1. In the aluminum-rich series the maximum dependable

strength occurs in the neighborhood of the 10 per cent copper alloy and its value is about 28,000 pounds per square inch. The alloy is, however, of low ductility. 20 percent copper seems to be the greatest copper content for this series which will give a useful alloy.

2. Throughout the aluminum-rich series, annealing for from 3 to 6 days at 400° reduces tensile strength and increases ductility.

3. In the bronzes with more than 92 percent copper, annealing seems to have no marked effect on either strength or ductility.

4. In the bronzes, between 92 percent and 95 percent copper, tensile strengths varying from 48,000 to 66,000 pounds per square inch with ductilities of from 50 percent to 60 percent may be expected.

5. From 89-91 percent copper, annealing above 565° and quenching in water gives an increased strength, but very low ductility.

6. Bronzes with less than 90 percent copper are brittle and likely to be unreliable; of coarse crystalline structure, and soft at temperatures considerably below the melting point.

7. Air-cooling bronzes of 90-91 percent copper give an increased ductility with but slightly reduced strength.

8. The ductility of the chill-cast bronzes decreases very rapidly in the 91 percent copper content, the ductility in the 92 percent bronze being 50 percent to 60 percent and at 90 percent copper the ductility is only between 6 percent and 8 percent.

9. With 90 percent copper content a strength of 100,000 pounds per square inch may be expected from pieces chill-cast or annealed above 566°. The strength is considerably reduced by annealing below 566°.

10. The β -alloys may be expected to give the highest strengths.

11. Small amounts of the β -phase in contact with α cause a very marked decrease in ductility and a very marked increase in strength.

12. Small amounts of the γ -phase in contact with α increases the strength and decreases the ductility of these alloys.

13. The general effect of annealing is to decrease the ductility of the copper-rich alloy and to increase the ductility of the aluminum-rich alloys.

14. The most dependable high-strength aluminum-rich alloy is found near the edge of the η -field or with a 10 percent copper content.

15. The maximum tensile strength does not occur at a boundary curve.

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Cornell University