Heat Treatment of Duralumin

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INTRODUCTION

The remarkable phenomena exhibited by the aluminum alloy known as duralumin were discovered during the years 1903–1911 by A. Wilm[1,2] and have been described by him and by others.[3,4,5,6]

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[1] A. Wilm: *Metallurgie* (1911) **8**, 225.

[2] A. Wilm: *Metallurgie* (1911) **8**, 650.

[3] L. M. Cohn: *Verh. Z. Beforderung des Gewerbefleisses* (1910) **89**, 643.

[4] L. M. Cohn: *Elektrotechnik u. Maschinenbau* (1914) **31**, 430.

[5] L. M. Cohn: *Elektrotechnik u. Maschinenbau* (1912) **30**, 809, 829.

[6] P. D. Merica: Aluminum and its light alloys, *Circular* 76, U. S. Bureau of Standards, 1918; also *Chem. and Met. Eng.* (1918) **19**, 135, 200, 329, 587, 635, 729, 780.

The unusual feature of this alloy is the fact, as was shown by Wilm, that it can be hardened quite appreciably by quenching from temperatures below its melting point followed by aging at ordinary temperatures, which consists merely of allowing the material to stand at these temperatures. The hardness is not produced by the quenching alone but increases during the period of aging, which may be from one to three days. Cohn[3,5] gives data showing the increase of hardness of duralumin during aging, after quenching in water from about 450° C. Upon annealing, the alloy so hardened by aging is softened exactly as is hardened steel.

The composition of this alloy usually varies within the following limits: Copper, 3 to 4.5 per cent.; magnesium, 0.4 to 1.0 per cent.; manganese, 0 to 0.7 per cent.; aluminum, balance; iron (as impurities), 0.4 to 1 per cent.; silicon, 0.3 to 0.6 per cent. Its density is about 2.85. It is used only in the forged or rolled condition.

This alloy has been produced for some years commercially and is in demand for the fabrication of parts for which both lightness and strength are required, such as for aircraft. Its tensile strength will average 50,000 to 60,000 lb. per sq. in. (3515 to 4218 kg. per sq. cm.) after appropriate heat treatment, such as that described by Wilm.

With the purpose of ascertaining whether the heat treatment described by him actually developed the best mechanical properties possible for duralumin, the authors undertook a study of the effect of variation in heat-treatment conditions, *i.e.*, quenching temperature, aging temperature, etc., upon these properties and, in connection with another investigation,[7] a study of the effect of chemical composition upon them.

Mr. E. Blough had already called the attention of one of the authors to the fact that the amount of hardening produced by heat treatment was influenced quite markedly by the temperature from which the material was quenched, a most interesting fact that was not brought out by Wilm's published investigations, which mentioned merely the effect of aging after quenching from one temperature, in the neighborhood of 450° C. An explanation was sought for the mechanism of hardening during aging of this alloy, and additional data were obtained bearing upon this phase of the matter.

The experiments here described were carried out partly in the laboratories of the Bureau of Standards and partly in coöperation with The Aluminum Company of America in its laboratories at New Kensington. The alloys used were prepared at the New Kensington plant of this company and the authors wish to express their appreciation of the

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[7] P. D. Merica, R. G. Waltenberg and A. N. Finn: The Tensile Properties and Resistance to Corrosion of Rolled Light Alloys of Aluminum and Magnesium with Copper, with Nickel and with Manganese. *Tech. Paper* No. 132; U. S. Bureau of Standards, 1919. Also *Bull.* A. I. M. E.

assistance and coöperation which has been given throughout by this company through Mr. E. Blough, chief chemist. Mr. H. H. Beatty of Mr. Blough's staff was active in assisting this work.

Composition and Preparation of Alloys

In Table 1 are given the chemical compositions of the alloys of the aluminum-copper-magnesium series that were used in these experiments. The ingots, 12 by 24 by 3½ in. (30 by 60 by 8.89 cm.) were rolled hot, at about 410° C., to ¼ in. (6.35 mm.) thick and thereupon cold rolled to 0.081 in. (2.05 mm.) (12 B. & S. gage), annealed at about 425° C., rolled cold to 0.051 in. (1.29 mm.) (16 B. & S. gage), annealed again and cold rolled to 0.032 in. (0.81 mm.) (20 B. & S. gage). The casting and rolling were done at the New Kensington plant of the U. S. Aluminum Co.

Table 1.—*Composition of Alloys**

Number	Aluminum	Magnesium	Copper	Iron	Silicon
C-1	97.27	1.16	0.72	0.56	0.29
C-2	96.69	2.37	0.04	0.62	0.28
C-3	97.15	None	2.15	0.36	0.34
C-4	96.65	2.84	0.04	0.27	0.20
C-5	96.11	None	3.19	0.40	0.30
C-6	96.72	2.03	0.72	0.30	0.23
C-7	96.62	1.00	1.80	0.35	0.23
C-8	96.68	1.07	1.67	0.33	0.23
C-9	95.98	3.50	0.08	0.26	0.18
C-10	95.83	2.95	0.74	0.27	0.21
C-11	95.51	1.26	2.58	0.41	0.22
C-12	95.74	0.46	3.18	0.34	0.24
A-1-12	95.48	0.64	3.22	0.39	0.27
E-3	96.80	1.06	1.56	0.32	0.26
N-34	94.36	1.08	3.74	0.52	0.30
E-4	94.47	1.06	3.68	0.50	0.29

* Aluminum by difference.

Heat Treatment and Aging

Tensile tests and scleroscope measurements were made upon specimens taken from the sheets as rolled, from the rolled sheets, annealed, and from the rolled sheets after heat treatment consisting of heating to various temperatures in a gas or electric furnace, quenching in water, and aging at room or other temperatures for different periods. The results of these tests are given in Table 2. All of the alloys, except those containing no copper (Nos. C-2, C-4, and C-9), show an increase of hardness of the heat-treated specimens over that of the annealed samples.

TABLE 2.—*Tensile Properties and Scleroscope Hardness of Rolled,*

No.	Tensile Properties and												
	As Rolled			After Annealing at 422° C.			After Heat Treatment Consisting						
							Quenched from 478° C.					Quenched	
							Aging, Days					Aging, Days	
	Scleroscope Hardness Magnifying Hammer	Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.	Scleroscope Hardness Magnifying Hammer	Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.	At 110°	At 20°	Scleroscope Hardness Magnifying Hammer	Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.	At 110°	At 20°
C-1	42	49,000	2.0	15.5	33,000			11		36,870	18.5		11
		48,400	2.5		33,100	15.0		11	20	37,080	17.0		11
		48,600	2.5		32,700	14.0		11		36,260	14.0	3	8
		49,600	2.5									3	8
C-2	19	25,800	4.0	7.5	16,600	35.0		11		16,830	35.0		11
		23,600	3.0		15,900	35.0		11	8	16,510	38.5		11
		23,600	3.5		16,100	33.0		11		16,510	38.5	3	8
												3	8
C-3	35	34,900	2.5	7.0	21,600	31.0		11		28,020	15.5		11
		35,700			21,800	33.0		11	11	25,810	25.0		11
		34,000	1.5		22,000	33.5		11		25,440	25.0	3	8
												3	8
C-4	37	38,400		10.5	29,200	18.0		11		29,300	20.0		11
		38,600	1.5		29,200	18.0		11	14	28,910	20.0		11
		37,200	1.5		29,400	21.0		11		30,280	22.0	3	8
												3	8
C-5	34	35,900		8.0	23,000	30.0		11		33,220	15.0		11
		37,500	2.5		22,400	28.5		11	15	32,580	18.5		11
		37,700	2.0		22,800	32.5		11		31,930	16.0	3	8
												3	8
C-6	38	35,300	1.0	13.0	30,500			11		31,050	20.0		11
		38,500	0.5		29,900	18.5		11	14	33,790	19.0		11
		38,100	1.5		30,800	16.0		11		31,640	17.5	3	8
												3	8
C-7	44	44,200	2.0	17.0	35,300	26.0		11		42,350	21.0		11
		45,500	2.0		34,600	25.5		11	23.5	42,530	21.0		11
		45,300			34,800	25.0		11		42,350	21.0	3	8
												3	8
C-8	38	38,100	1.5	12.5	28,500	18.5		13		46,400	19.5		11
		38,100			29,100	18.5		13	26	47,030			11
							7	6		48,900	20.0	3	8
							7	6	28.5	47,650	22.0	3	8
C-9	38	41,200	1.5	12.0	31,600	17.5		11		31,790	21.5		11
		43,200	1.5		31,200	17.5		11	15	30,450	18.0		11
		41,200	1.5		30,500			11		30,070	14.0	3	8
												3	8
C-10	45	44,800	1.5	12.0	30,600	19.0		11		38,030	26.0		11
		44,600	1.5		30,200	19.0		11	14	37,630	25.0		11
		47,500	1.5		30,200	17.0		11		38,430	22.5	3	8
												3	8
C-11	50	56,700	2.0	15.5	34,900	20.5		13		50,450			11
		52,900	1.5		36,000	24.0		13	28	48,950	22.0		11
		58,400	2.0				7	6		51,740	22.0	3	8
							7	6	31	50,880	22.0	3	8
C-12	31	38,900	5.0	7.5	23,100	24.0		11		38,330	14.0		11
		38,600	5.0		23,000	24.0		11	19–23	38,730	13.5		11
								11		35,910	12.5	3	8
												3	8

Annealed, and Heat-treated Aluminum-copper-magnesium Alloys

Scleroscope Hardness

of Quenching in Water and Aging

from 510° C.			Quenched from 520° C.					Quenched from 525° C.				
			Aging, Days					Aging, Days				
Scleroscope Hardness Magnifying Hammer	Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.	At 20°	At 110°	Scleroscope Hardness Magnifying Hammer	Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.	At 20°	At 110°	Scleroscope Hardness Magnifying Hammer	Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.
	38,030	17.0		11		36,870	18.5		11		37,220	14.0
17	37,220	16.5		11	18.5	36,910	19.0		11	16	38,130	17.0
	48,120	16.0						3	8		50,140	17.0
27	47,210	18.5						3	8	27	49,930	17.5
	16,670	34.0		11		13,970	37.0		11		16,350	
8	16,670	33.0		11	8	14,290	35.0		11	8	16,870	35.0
	16,510	28.0						3	8		16,980	31.0
8	16,510	33.0						3	8	8	17,340	30.0
	26,350	19.0		11		24,700	18.5		11		28,550	23.0
13	27,690	11.5		11	13	25,440	21.0		11	13	28,640	
	29,420	20.0						3	8		26,800	17.5
11	27,790	19.5						3	8	10.5	26,980	20.0
	30,060	23.0		11		27,540	22.0		11		29,600	20.5
11	29,700	16.5		11	12	27,150	20.0		11	12	29,100	22.0
	31,590	19.0						3	8		34,850	17.0
14	31,350	20.0						3	8	17	35,350	16.5
	31,960	15.5		11		29,150	13.0		11		29,540	19.0
14	30,500	14.0		11	14	29,580	16.0		11	10.5	32,490	16.5
	30,910	19.0						3	8		28,170	23.0
14	33,970							3	8	12	27,960	26.0
	33,370	17.0		11		36,140	21.5		11		34,980	17.5
15	33,950	23.5		11	17	33,990	19.5		11	15	35,470	18.0
	43,190	18.5						3	8		47,580	20.5
26	43,560	18.0						3	8	26	45,840	16.5
	45,650	18.5		11		42,530	22.5		11		46,850	20.5
24	45,740	19.5		11	25	42,160	22.5		11	16	46,760	19.5
	53,970	20.0						3	8		54,170	21.0
35	52,250							3	8	35	56,110	19.0
	44,130	24.5		11		45,160	26.0		11		46,520	23.0
22	44,910	23.0		11	25.5	44,720	25.0		11	25	46,030	25.5
	49,680	19.5		11		44,720	21.0	3	8		53,440	21.0
32	51,530	17.0		11		44,070	22.0	3	8	31	53,000	19.0
	29,120	21.0		11		29,500	22.5		11	33	33,790	21.0
13	29,500	22.0		11	13	30,070	19.5		11	13	32,510	17.5
	30,270	23.0						3	8		34,800	17.5
14	30,270	22.0						3	8	18	34,890	18.0
	37,430	24.5		11		37,430	23		11		35,160	22.5
14	37,630	21.5		11	14.5	38,230	22		11	15	35,430	22.0
	47,690	21.5						3	8		44,260	19.9
26	47,690	22.5						3	8	26	46,450	20.0
	51,520	21.0										
29.5	50,870	24.0										
	54,740	23.0										
34	55,590	20.0										
	42,370	14.5							11		42,660	14.0
25–28	39,340	16.5							11	22	36,500	19.9
	49,230	26.5										
26	49,830	25.5						3	8	26	50,890	23.0

The increase of hardness in those alloys containing copper but no magnesium is smaller than that in those containing both, but is quite definite. This is shown in Table 3.

TABLE 3.—*Effect of Addition of Copper and Magnesium*

Number of Alloy	Increase of Tensile Strength of Heat-treated Alloy (510° C.) Over Annealed Alloy, Per Cent.	Number of Alloy	Increase of Tensile Strength of Heat-treated Alloy (510° C.) Over Annealed Alloy, Per Cent.
Alloys containing no copper		Alloys containing both copper and magnesium	
C-2	+ 2	C-1	60
C-4	+ 3	C-11	110
C-9	− 4	C-12	56
Alloys containing copper but no magnesium			
C-3	+30		
C-5	+36		

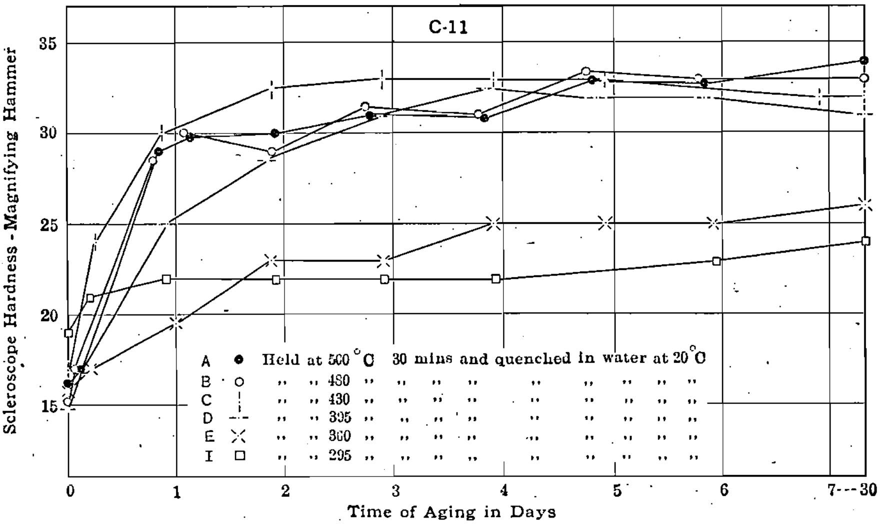

FIG. 1.—EFFECT OF QUENCHING TEMPERATURE ON SCLEROSCOPE HARDNESS DURING AGING OF QUENCHED SPECIMENS OF C-11.

The best mechanical properties are produced by quenching from the higher temperatures (500° to 525° C.). This is further shown in Table 4, which gives further data on two alloys, C-8 and C-11, and will be shown more clearly below. Not only does the hardness increase after heat treatment, but so also does the ductility, as evidenced by the elongation in the tensile test. This is shown in Tables 2 and 4.

Effect of Quenching Temperature

In Fig. 1 are shown the scleroscope hardness values of C-11 quenched in water (20° C.) from different temperatures and aged at room temperature for periods of time from a few hours to 30 days. The form of these aging curves is similar to that shown by Cohn; *i.e.*, the hardness increases after quenching, at first rapidly and then more slowly. It is further evident that the maximum hardness attained increases with the temperature up to approximately 520° C.

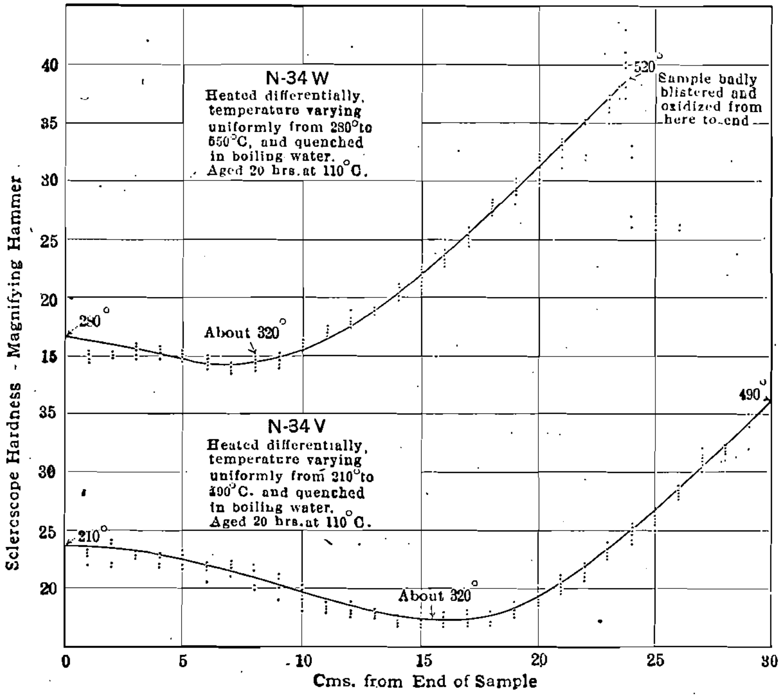

FIG. 2.—EFFECT OF QUENCHING TEMPERATURE ON SCLEROSCOPE HARDNESS. (ALLOY N-34.)

The effect of quenching temperature is also shown very nicely in an experiment of which the results are shown in Fig. 2. Two strips of 0.087-in. (2.20 mm.) sheet of alloy N-34 were used. The strip was placed in the furnace for heating in such a manner that a nearly linear temperature gradient existed between the two ends, as shown by thermocouples placed along the strip. Upon attaining the desired range of temperatures the strip was quenched in boiling water and aged 20 hr. at 110° C. The scleroscope hardness was then determined along the axis of the strip and is shown in Fig. 2 as a function of the distance from one end of the sample. The distance may be regarded as a rough temperature scale; the outside temperature limits have been determined and marked on the curve. One strip was quenched when the two ends were at 520°

TABLE 4.—*Effect of Quenching Temperature on Tensile Properties*

Temperature of Quenching, Degrees C.	Alloy C-8					Alloy C-11				
	Aging, Days		Mechanical Properties			Aging, Days		Mechanical Properties		
	At 20°	At 110°	Scleroscope Hardness	Ultimate Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.	At 20°	At 110°	Scleroscope Hardness	Ultimate Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.
371	13	..	..	27,260	16.0	13	..	..	35,900	19.0
	13		11	26,220	18.0	13	..	18	36,550	17.0
	7	6	..	29,130	18.5	7	6	..	35,020	21.0
	7	6	12	29,130	20.0	7	6	17	35,020	21.0
422	13	..	..	39,540	12.0	13	..	..	43,790	15.5
	13	..	23	40,160	13.5	13	..	25	43,360	18.5
	7	6	..	41,200	17.0	7	6	..	44,010	23.5
	7	6	23	41,200	22.0	7	6	24	43,580	24.0
478	13	..	..	46,400	19.5	13	..	..	50,450	
	13	..	26	47,030	....	13	..	28	48,950	22.0
	7	6	..	48,900	20.0	7	6	..	51,740	22.0
	7	6	28.5	47,650	22.0	7	6	31	50,880	22.0
500	11	..	..	47,230	18.5	11	..	..	52,590	20.5
	11	..	27	47,860	20.5	11	..	28	52,590	21.0
510	11	..	..	44,130	24.5	11	..	..	51,520	21.0
	11	..	22	44,910	23.0	11	..	29.5	50,870	24.0
	8	3	..	49,680	19.5	8	3	..	54,740	23.0
	8	3	32	51,530	17.0	8	3	34	55,590	20.0
520	11	..	..	45,160	26.0					
	11	..	25.5	44,720	25.0					
	11	..	..	44,720	21.0					
	11	..	..	44,070	22.0					
525	11	..	..	46,520	23.0					
	11	..	25	46,030	25.5					
	8	3	..	53,440	21.0					
	8	3	31	53,000	19.0					
533	13	..	..	34,960	5.0	13	..	..	48,370	9.5
	13	..	30-35	40,570	9.0	13	..	33	47,060	10.0

and 280° C., respectively; the other, when the ends were at 490° and 210° C., respectively. Beginning at about 300° C., the effect of increased quenching temperature, other factors remaining alike, is to increase the hardness after aging until a temperature of about 520° is reached. Beyond that temperature, the hardness again decreases; the material becomes covered with a dark gray oxide coating and generally also with

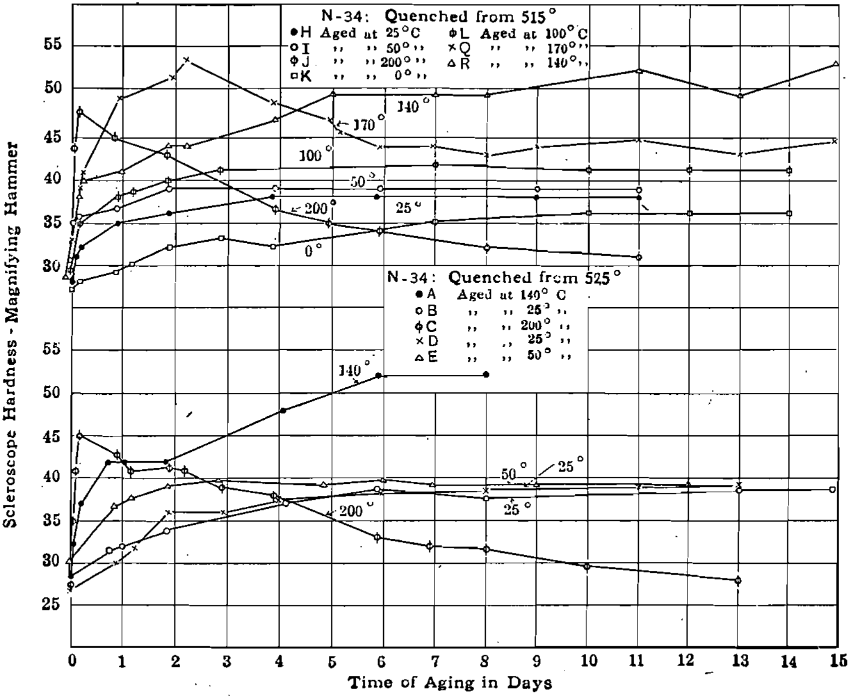

FIG. 3.—EFFECT OF AGING AT DIFFERENT TEMPERATURES ON SCLEROSCOPE HARDNESS OF SAMPLES QUENCHED FROM 515° C. AND 525° C. (ALLOY N-34.)

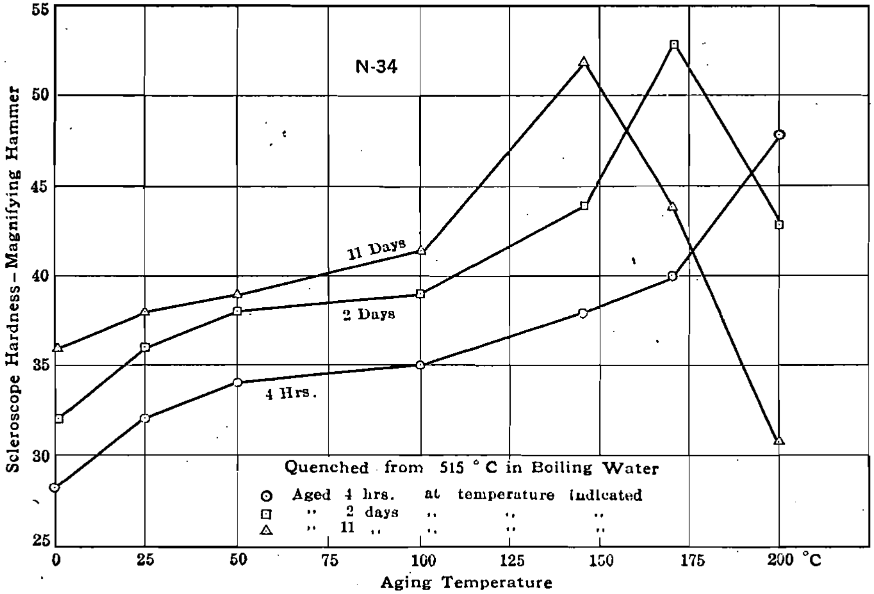

FIG. 4.—EFFECT OF TEMPERATURE OF AGING ON SCLEROSCOPE HARDNESS. (ALLOY N-34.)

blisters, marking the temperature of eutectic melting. The effect of heating to temperatures around 300° C. is chiefly to anneal the specimen and to give lower values of the hardness (minimum on the curve) than is given by heating at lower temperatures.

Effect of Aging Temperature

In Table 5 are given results of tests showing the effect of temperature of quenching bath and of aging carried out in the bath. The samples used were strips of A-1-12 quenched from 520° C. The increase of strength with time of aging is evident.

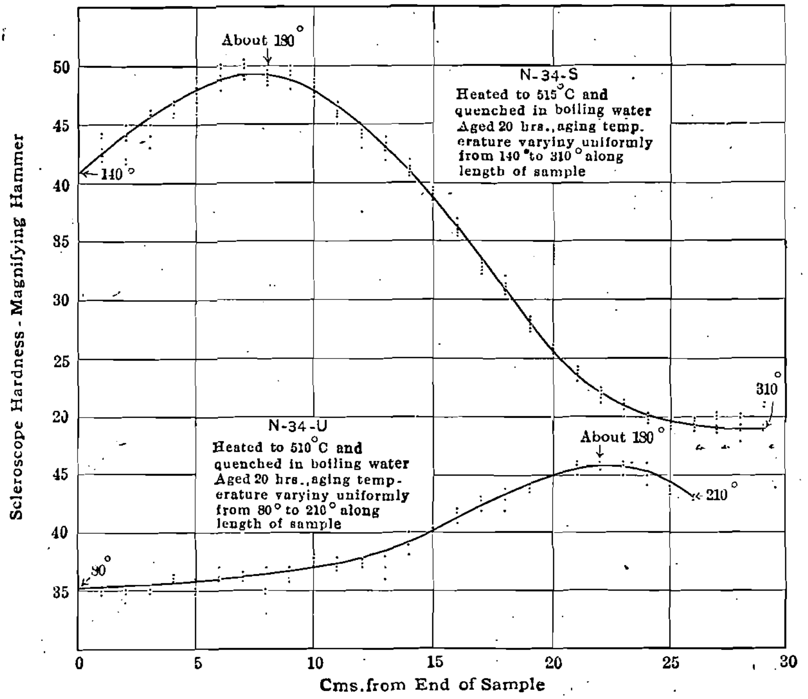

FIG. 5.—EFFECT OF TEMPERATURE OF AGING ON SCLEROSCOPE HARDNESS. (ALLOY N-34.)

A more complete picture of the phenomenon of hardening by aging at different temperatures is obtained from Figs. 3, 4, and 5, based on data obtained on specimens of N-34. The scleroscope values of Fig. 3 were obtained upon samples quenched in boiling water from two temperatures, 515° and 525° C., and aged at different temperatures. The same figures are replotted in Fig. 4 in different form.

It is noted that the rate of hardening increases as the temperature of aging increases, that the maximum hardness is obtained by aging

TABLE 5.—*Effect of Temperature of Quenching Bath and Time of Aging in Quenching Bath*

(Specimen of Alloy A-1-12 Quenched from 520° C.)

Hours Aged in Quenching Bath Before Testing	Quenched to 20° C. Aged at 20° C.			Quenched to 100° C. Aged at 110° C.			Quenched to 150° C. Aged at 150° C.			Quenched to 200° C. Aged at 200° C.			Quenched to 230° C. Aged at 250° C.		
	Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.	Scleroscope Hardness	Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.	Scleroscope Hardness	Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.	Scleroscope Hardness	Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.	Scleroscope Hardness	Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.	Scleroscope Hardness
0	35,200	28.0	13.5	41,800	27.0	20.0	43,250	23.5	23	42,950	20.5	23	41,950	23.5	24
0	35,100	17.5	13.0	41,250	22.0	21.0	43,550	23.0	23	43,600	19.5	23	42,500	26.0	24.5
½	35,100	24.0	16.0				51,000	22.0	31	50,250	14.0	41	43,800		35
½							50,700	25.5	31	52,000	12.0	41	45,600	10.0	35
1	38,900	24.0	18.0	47,600	24.0	27.0	48,500	18.0	31	49,250	12.0	41–42	44,500		37
1	36,200	20.0	17.0	46,100	17.5	27	48,600	18.0	31	52,450	12.0	41–43	45,450	9.5	36
1½	36,000	20.0	20.0	47,500	21.5	25									
1½	38,200	22.5	19.0	49,500	24.0	26									
2	37,100	20.0	19.0	48,000	21.5	27	50,550	23.5	34	52,950	11.0	42	47,600	9.0	37
2	37,500	20.0	21.0	48,100	24.0	27	52,000	23.5	34	51,600	11.0	42	47,700	9.0	38
3							48,500	18.0	31	51,950	10.0	45–6			
3							49,450	18.0	31	54,250	10.0	44–5			
19	47,525	25.0	27.0	51,500	21.5	30									
19	47,450	18.0	28.0	52,250	26.0	32									
48	48,150	18.5	30.0	51,600	22.5	30									
48	49,050	22.0	30.0	52,750	20.5	32									
96	50,050	22.5	31.0												
96	50,300	22.5	31.0												

TABLE 6.—*Effect of Quenching in Baths at Different Temperatures Followed by Aging at Room Temperature (20° C.)*

(Specimen of Alloy A-1-12, Quenched to Room Temperature from 520° C.)

Aged 0 Hr. at Quenching Temperature. Hours Aged at 20° C.	Quenched to 100° C. Aged at 20° C.			Quenched to 150° C. Aged at 20° C.			Quenched to 200° C. Aged at 20° C.			Quenched to 230° C. Aged at 20° C.		
	Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.	Scleroscope Hardness	Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.	Scleroscope Hardness	Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.	Scleroscope Hardness	Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.	Scleroscope Hardness
0	41,800	27.0	20.0	43,250	23.5	23	42,950	20.5	23	41,950	23.5	24
0	41,250	22.0	21.0	43,550	23.0	23	43,600	19.5	23	42,500	26.0	24.5
½							49,300	15.5	39–40	45,500	21.5	25
½							50,450	14.0	39	44,900	26.0	25
1	45,750	21.5	25	44,500	26.0	25.5	49,750	15.0	37–39	44,100	22.5	25–28
1	45,025	22.0	24	44,050		25.0	48,550	15.0	36–37	45,000	22.0	26
1½	44,900	19.0	24									
1½	44,450	23.0	24									
2	44,550	23.0	24	43,850		24.0	50,100	16.0	37	44,850		25
2				44,600	27.0	23.5	49,000	11.5	37	43,700	23.0	22–25
3							49,100	16.5	37			
3							50,950	16.0	37–39			
19	44,700	24.5	25				48,850	16.0	37			
19	47,100	24.5	27				48,950	12.5	38			
48	46,950	25.5	26				49,250	16.0				
48	47,000	22.0	26				48,950	14.0				
96	47,000	22.0	26	51,500	20.0	31				51,600	25.0	30
96	48,250	22.0	26	52,500		30				50,400	24.0	30

at temperatures above 100° C., and that at aging temperatures above 140° C. the hardness eventually drops after reaching its maximum.

Fig. 5 shows the results of an experiment similar to that of Fig. 2. The strips were quenched from 515° C. in boiling water and aged for 20 hr. thereafter in a furnace giving a temperature gradient from one end to the other of the sample. For a time of aging of 20 hr. the hardness first increases with the temperature to a maximum at about 175° C. and then decreases. Above this temperature annealing sets in.

Effect of Temperature of Quenching Bath

Table 6 shows the effect of temperature of the quenching bath upon samples of A-1-12 quenched from 520° C. The tensile strength of the alloys, as well as the elongation, increases with the time of aging. There is no marked effect of the temperature of the quenching bath indicated in these results; those samples quenched to 150° C. gave practically the same results as those quenched to 230° C., although there is a slight improvement in the tensile properties of those quenched to 150° C. over those quenched to 100° C.

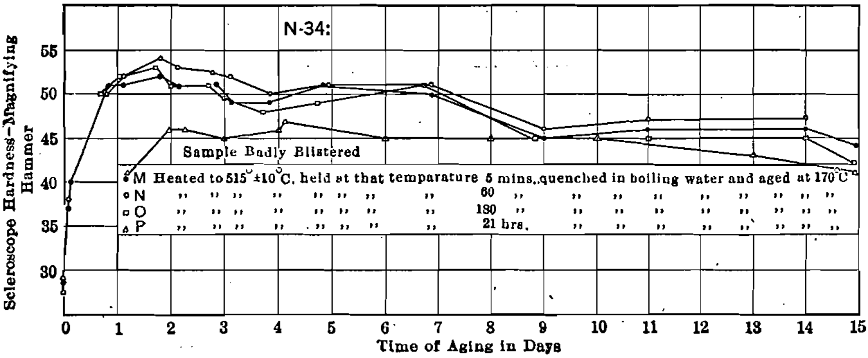

FIG. 6.—EFFECT ON SCLEROSCOPE HARDNESS OF HOLDING SPECIMENS FOR SOME TIME AT QUENCHING TEMPERATURE BEFORE QUENCHING.

In Table 7 are shown results of tests to determine the effect of aging at room temperature after aging at the temperature of the quenching bath. It will be noted that there is only a slight increase in the strength of the alloy produced by aging at 20° after aging at the temperature of the quenching bath.

Effect of Prior Heating at Quenching Temperature

Fig. 6 shows the results of hardness measurements on samples held at the quenching temperature for varying periods of time, quenched and aged at 170° C. The samples held from 5 to 180 min. at the quenching temperature give values of the hardness differing by less than the probable error of measurement. The low values found on the sample held 21 hr. are due probably to the blistering which was noticed in the sample.

Effect of Preheating to 515° C. before Quenching from Lower Temperatures

In Figs. 7 and 8 are shown the results of experiments to determine whether preheating to a temperature higher than the quenching one before quenching gave a hardness different from that obtained by heating merely to the quenching temperature.

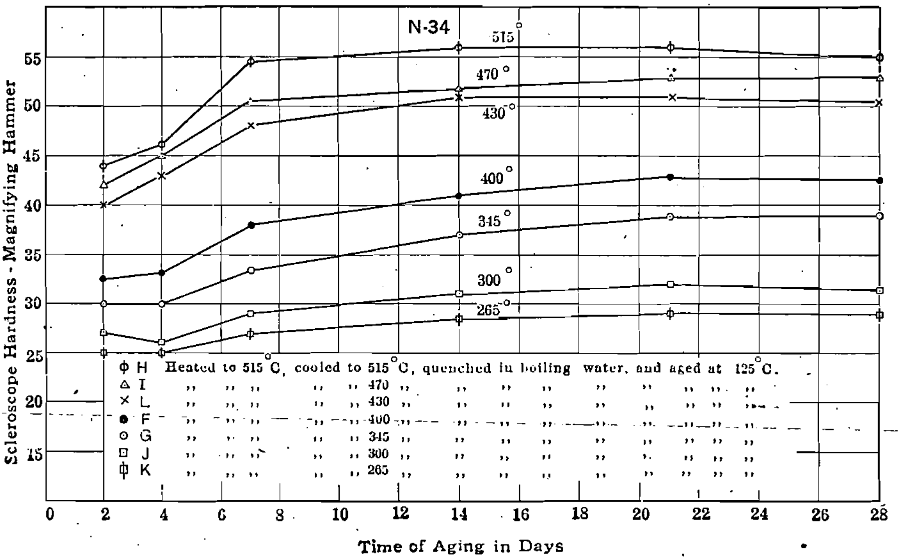

FIG. 7.—EFFECT ON SCLEROSCOPE HARDNESS OF PREHEATING SPECIMENS HIGHER THAN QUENCHING TEMPERATURE, COOLING TO LATTER, THEN QUENCHING AND AGING.

Although owing to a slight difference in the aging conditions, the comparison is not quite definite, it is obvious (1) that the hardness obtained by heating to 515° C., cooling to a temperature t (when $t < 515°$ C.), and quenching, is always greater than that obtained by quenching from $t°$; and (2) that whether the specimen is preheated or not to a higher temperature before quenching from some lower temperature, the hardness obtained increases with higher quenching temperatures.

MISCELLANEOUS TESTS

In Table 8 are shown the results of a number of tests of alloy N-34, including determinations of the proportional limit of several heat-treated samples. Inasmuch as so many of the tests made during this investigation were measurements of scleroscope hardness, a comparison was made between the hardness and the tensile proportional limit of some specimens of N-34 during aging at two temperatures. The results are shown in Fig. 9, the curves in each show quite close parallelism.

Experiments were conducted on specimens of A-1-12 to ascertain whether the hardening during aging could be hastened by vibration. The results of several such tests in which the vibration was produced by a bell clapper indicated that there was no difference in the rate of hardening between vibrated and quiet specimens.

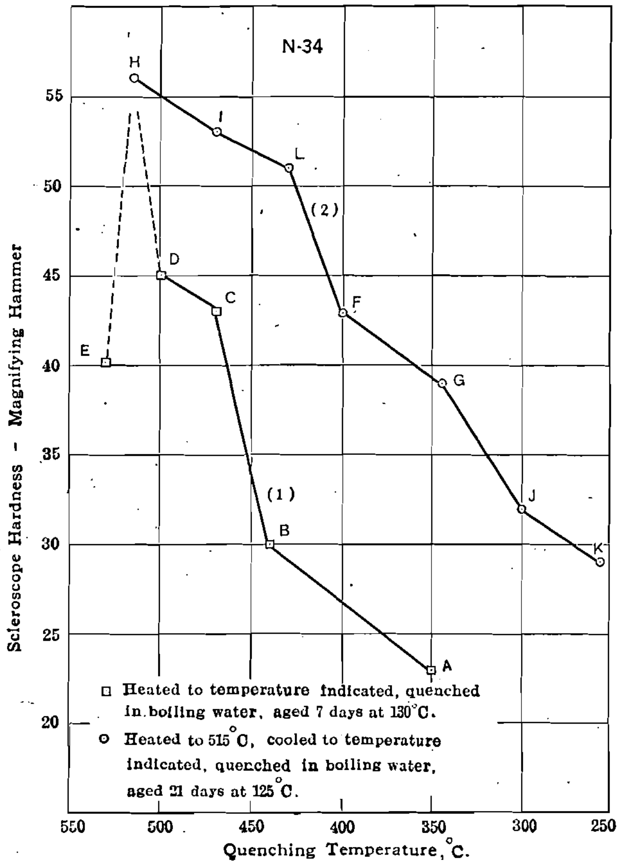

FIG. 8.—COMPARISON OF SCLEROSCOPE HARDNESS OF SPECIMENS OF ALLOY N-34. (1) HEATED TO QUENCHING TEMPERATURE, QUENCHED, AND AGED 7 DAYS AT 130° C., AND (2) HEATED TO 515° C., COOLED TO QUENCHING TEMPERATURE, QUENCHED, AND AGED 21 DAYS AT 125° C.

Density and Dilation

The density was determined of samples of N-34 in different conditions; Table 9 gives the results of these tests. In some cases one dimension of the specimen was determined also, and its changes recorded in the

TABLE 7.—*Effect of Aging at Room Temperature (20° C.) after Aging at Temperature of Bath*

(Specimen of Alloy A-1-12 Quenched from 520° C.)

Hours Aged at 20° C.	Quenched to 100° C. Aged at 20° C.			Quenched to 150° C. Aged at 20° C.			Quenched to 200° C. Aged at 20° C.			Quenched to 230° C. Aged at 20° C.		
	Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.	Scleroscope Hardness	Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.	Scleroscope Hardness	Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.	Scleroscope Hardness	Tensile Strength, Lb. per Sq. In.	Elongation in 2 In., Per Cent.	Scleroscope Hardness
Aged ½ Hr. at Quenching Temperature.												
½										45,150	10.0	35
½										46,250	10.0	37
1				49,100	26.0	31				42,850	12.0	32
1				50,500	23.0	31				42,600	10.5	35
2				49,500	25.5	28				45,750	9.0	36
2				49,600	26.0	28–30				46,100	10.5	36
3												
3												
96				*50,800		31	53,200	10.0	40	48,500	10.0	34
96				*52,000	25.5	31	51,700	12.0	41	46,450	8.5	34
Aged 1 hr. at quenching temperature.												
										(Aged ½ hr. at 20° C.)		
2				49,250	25.5	31				(47,050	9.5	38–41)
2				50,000	25.0	31				(47,300	9.5	36–39)
96				*52,500		31	50,600	9.0	43	48,600		37
96				*52,500		31	49,400	7.0	41	48,550		39
Aged 2 hr. at quenching temperature.												
										(Aged ½ hr. at 20° C.)		
2				51,450	24.5	33				(47,600	9.0	37)
2				48,500	23.0	33				(47,100	9.0	37)
96	48,900	24.5	28.0	*54,600	24.0	32	52,150		44	50,400	8.5	38
96	47,450		27	*53,850	22.5	32	53,550	11.0	42	50,750		41
Aged 3 hr. at quenching temperature.												
2				52,450	22.0	34						
2				49,900	23.0	34						
96							54,950		43			
96				*56,250	19.5	32.5	54,550	10.0	44			
Aged 19 hr. at quenching temperature.												
96	51,000	24.5										
96	51,550											
Aged 48 hr. at quenching temperature.												
96	52,000	20.5										
96	51,600	23.5										

* These specimens were aged at 20° C. for 120 hr.

TABLE 8.—*Effect of Varying Aging Temperature and Time of Aging on Tensile Properties of Aluminum Alloy Sheet.*[1] *Alloy N-34.*

No.	Thickness of Sheet, Inch	Heat Treatment			Scleroscope Hardness Magnifying Hammer	Ultimate Strength, Lb. per Sq. In.	Proportional Limit, Lb. per Sq. In.	Elonga- in 2 In., Per Cent.	Remarks
		Quenched from C°	Aged						
			at ° C.	Days					
14	0.034	515		0	26		17,000	5	Broke at extensometer contact.
15	0.034	515		0	26	38,200	17,000	6	Broke at extensometer contact.
24	0.034	515	105	2	39	52,900	27,000	7	Broke at extensometer contact.
25	0.034	515	105	2	39	54,900	29,000	7	Broke at extensometer contact.
26	0.034	515	105	4	39	55,500	27,000	9	Broke at extensometer contact.
27	0.034	515	105	4	39	55,200	26,000	8.5	Broke at extensometer contact.
28	0.034	515	105	7	38	63,200		23	No extension measurements.
29	0.034	515	105	7	38	56,200	29,000	9	Broke at extensometer contact.
30	0.034	515	105	9	42	61,200	30,000	16	Extensometer attached to flat surfaces. Broke at gage point.
31	0.034	515	105	9	41	60,600	29,000	14.5	Extensometer attached to flat surfaces. Broke at gage point.
32	0.034	515	105	14	40	62,200	31,000	21	Extensometer attached to flat surfaces.
33	0.034	515	105	14	41	61,600	33,000	23	Extensometer attached to flat surfaces.
16	0.034	515	125	2	41–46	49,500	27,000	2.5	Broke at extensometer contact.
17	0.034	515	125	2	42–50	54,500	32,000	3	Broke at extensometer contact.
18	0.034	515	125	4	45–50	58,200	30,000	3	Broke at extensometer contact.
19	0.034	515	125	4	45–50	58,500	33,000	4	Broke at extensometer contact.
20	0.034	515	125	7	44–47	60,000	39,000	4.5	Broke at extensometer contact.
21	0.034	515	125	7	42–48	58,000	37,000	4	Broke at extensometer contact.
34	0.087	515	125	7	46–50	61,200	32,000	18	Extensomotor attached to flat surface.
35	0.087	515	125	7	47–52	62,000	31,000	18	Extensometer attached to flat surface.
22	0.034	515	125	14	47–52	64,900	37,000	11	Extensometer attached to flat surface.
23	0.034	515	125	14	47–54	61,500	35,000	3	Extensometer attached to flat surface. Fractured at blister.
2	0.034	515	150	2	50	51,110		16	No extension measurements.
3	0.034	515	150	2	49	50,940		16.5	No extension measurements.
4	0.034	515	150	4	50	59,800		11.5	No extension measurements.
5	0.034	515	150	4	50	61,500	41,000	6	Broke at extensometer contact.
6	0.034	515	150	6	50	63,800		6	No extensometer measurements.
7	0.034	515	150	6	50	62,200	43,000	6.5	Broke at extensometer contact.
1	0.087	515	170	2	50	63,900		10	No extension measurements.
8	0.034	515	170	2	51.5	51,420		10.5	No extension measurements.
9	0.034	515	170	2	51.5	51,850		10	No extension measurements.
10	0.034	515	170	2	51.5	50,760		9.5	No extension measurements.
11	0.034	515	170	4	44	58,200		9.5	No extension measurements.
12	0.034	515	170	4	44	55,500		5	No extension measurements.
13	0.034	515	170	4		57,300	34,000	6	Broke at extensomoter contact.

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[1] Where two values of hardness are given, the lower one shows the hardness of the end near the door of the furnace in which the sample was heated for quenching and the other value is the hardness of the opposite end, the difference in hardness being the result of a temperature gradient in the furnace. All specimens in this condition broke at the soft end and hence their tensile properties are hardly as high as can be expected of the material.

TABLE 9.—*Density and Dilatation of Duralumin*

Sample	Treatment	Density	Length, Inches
N34 D-1	Quenched; not aged	2.762	
N34 D-1	Same as above, after aging at 150° C.	2.762	
N34 D-2	Annealed, after rolling, at 515° C.	2.759	
N34 D3	As rolled, 0.033 in. thick	2.754	12.014
N34 D3-a	Same after annealing at 500°	2.742	12.024
N34 D3-a	Annealed at 530° C.	.....	12.0477
N34 D4	As rolled, 0.088 in. thick	2.750	11.982
N34 D4-a	Same after annealing at 500°	2.747	11.982
N34 D4-a	Annealed at 530° C.	.....	11.9973
N34 D5	As rolled, 0.25 in. thick	2.764	11.9954
N34 D5-a	Same after annealing at 500°	2.762	11.9963
N34 D5-a	Annealed at 530° C.	.....	12.0019

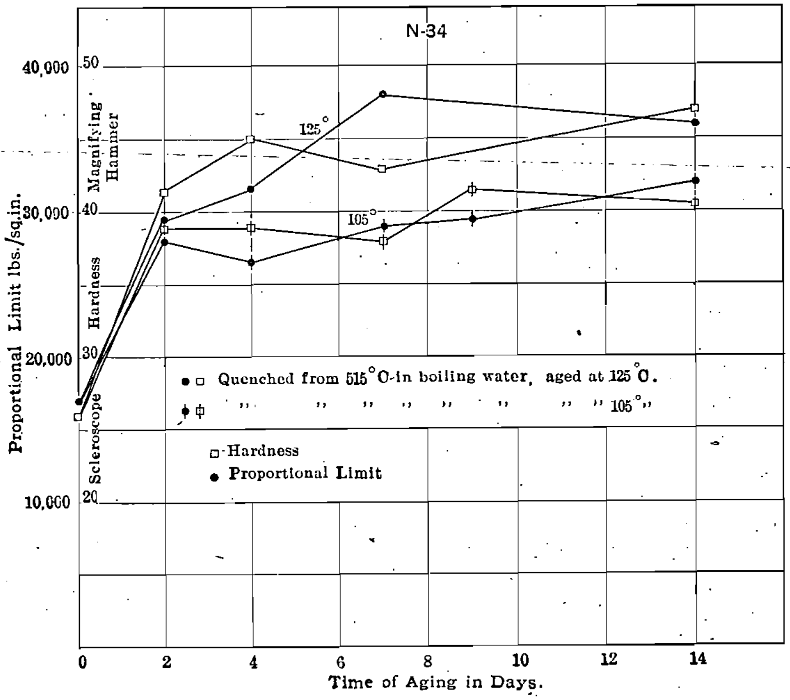

FIG. 9.—COMPARISON DURING AGING OF SCLEROSCOPE HARDNESS AND TENSILE PROPORTIONAL LIMIT. (ALLOY N-34.)

fourth column of the table. The changes in density are quite small as the material undergoes heat treatment or annealing, except when the temperature exceeds from 520 to 530° C., the temperature of eutectic melting, when a marked increase in length is noted.

The linear expansion, up to 520° C., was determined on two bars of N-34, one as rolled, the other after heat treatment consisting of quenching from 520° C. and aging two days at 120° C. The expansion curves are given in Fig. 10, and show irregularities in the neighborhood of 300°.

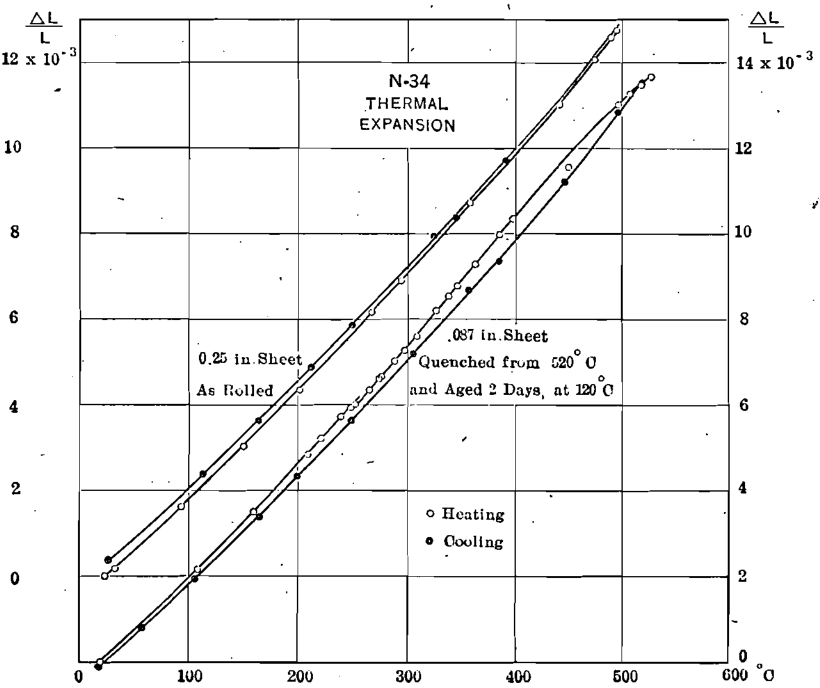

Fig. 10.—Linear expansion of N-34, 0 to 500° C.

Electrical Resistivity

Electrical resistivity measurements were made in vacuo by the method described by Burgess and Kellberg[8] on 0.25-mm. wire drawn from a cylinder cut out from ¼-in. (6.35-mm.) sheet of C-11 over the temperature range 0 to 530° C. It was necessary, however, to bring both of the aluminum-alloy leads out of the thermometer as it was impossible to weld them to platinum. The data obtained from the first run is plotted as resistance of aluminum alloy against temperature in Fig. 11.

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[8] *Sci. Paper* No. 236, U. S. Bureau of Standards (1914).

The change in direction of the resistivity curve at about 300° is quite evident and indicates a change in the constitution of the alloy. It is evident both on heating and cooling, although a change in resistivity at room temperature has taken place resulting from the annealing produced during the series of measurements.

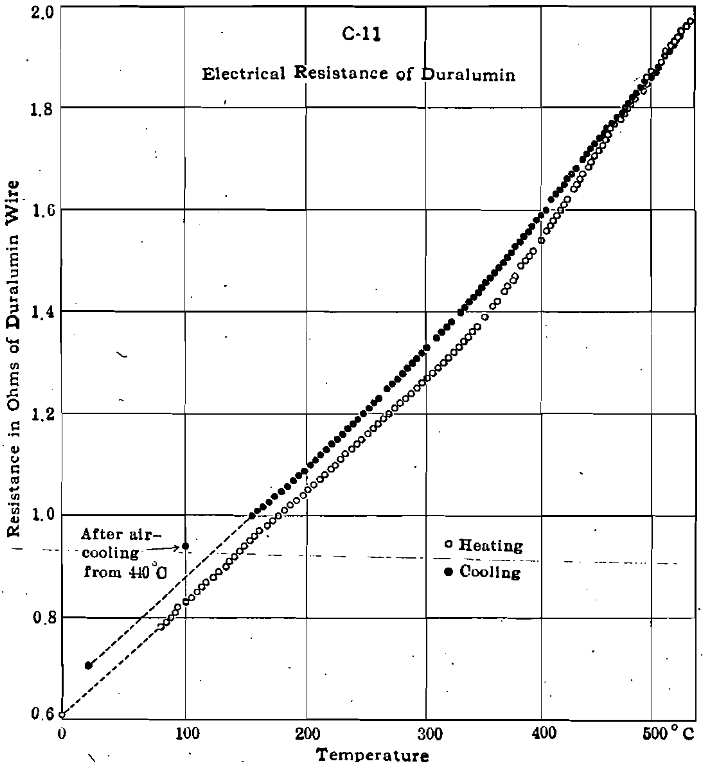

Fig. 11.—Electrical resistivity of C-11, 0 to 520° C.

Following this run the material was heated to 440° C. in its tube and cooled in air. The cooling was fairly rapid as the outside diameter of the quartz tube was only 8 mm. The tube was then put in a steam bath and resistance measurements taken as follows:

Time in Steam Bath, Hours	Pt Resistance	Al Resistance	Time in Steam Bath, Hours	Pt Resistance	Al Resistance
0	1.7302	0.9035	4½	1.7300	0.9063
½	1.7302	0.9047	6	1.7301	0.9069
1	1.7302	0.9051	7	1.7301	0.9068
2	1.7297	0.9054	11	1.7298	0.9069
3	1.7301	0.9060			

The specific resistance of this alloy was determined on a wire drawn to 0.254 mm. diameter and annealed at 400° C. It was found to be 3.35 microhm-cm.

Mechanism of Hardening During Aging after Rapid Cooling

Apparently no attempt has been made to develop an explanation for the changes in the physical, particularly mechanical, properties of this alloy during aging after rapid cooling. The changes that take place are quite marked and definite and must correspond to some quite as definite changes in the structure and constitution of the alloy, or at least to profound molecular changes. If we are not able to show that actual phase changes take place during aging, we must ascribe these changes in physical properties to alterations in the atomic or molecular structure. All evidence the authors have been able to find or to accumulate seems to indicate that the hardening during aging is actually accompanied by a phase change within the alloy. In so far as it can be said then that this phase change causes the hardening; for the reason that it accompanies it, this phase change may be regarded as its active cause.

Elsewhere,[9] the authors have determined the solubility at different temperatures in aluminum of $CuAl_2$ and of Mg_4Al_3, the aluminum-rich compounds of the copper-aluminum and magnesium-aluminum binary alloy series respectively. The solubility-temperature curves of these compounds are reproduced in Figs. 12 and 13; the solubility of both compounds diminishes rapidly with lowered temperature.

Upon slowly cooling an alloy containing 3 per cent. of copper from 500° C., the $CuAl_2$ precipitates from solid solution to maintain equilibrium along the line *be*. The alloy so obtained is soft and does not harden upon aging. Rapid cooling of the same alloy from 500° C. by quenching, partly or wholly suppresses this precipitation of $CuAl_2$. If the alloy is held at a low temperature, such as at that of liquid air ($-180°$ C.), no further hardening takes place upon aging. The alloy is not in equilibrium but the rate of nuclear formation of $CuAl_2$ is so small that no $CuAl_2$ precipitates to bring about equilibrium. If, however, the temperature of the alloy is raised to 100° C. or even to ordinary room temperature, according to the theory which the authors propose, the mobility of the molecules becomes sufficiently great that precipitation of the $CuAl_2$ takes place in the form of very fine particles of colloidal dispersion. To this precipitation is due the hardening during aging of duralumin. The evidence

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[9] P. D. Merica, R. G. Waltenberg, and J. R. Freeman, Jr.: The Constitution and Metallography of Aluminum and its Light Alloys with Copper and with Magnesium, *Sci. Paper* No. 337, U. S. Bureau of Standards (1919). Also *Bull.* A. I. M. E.

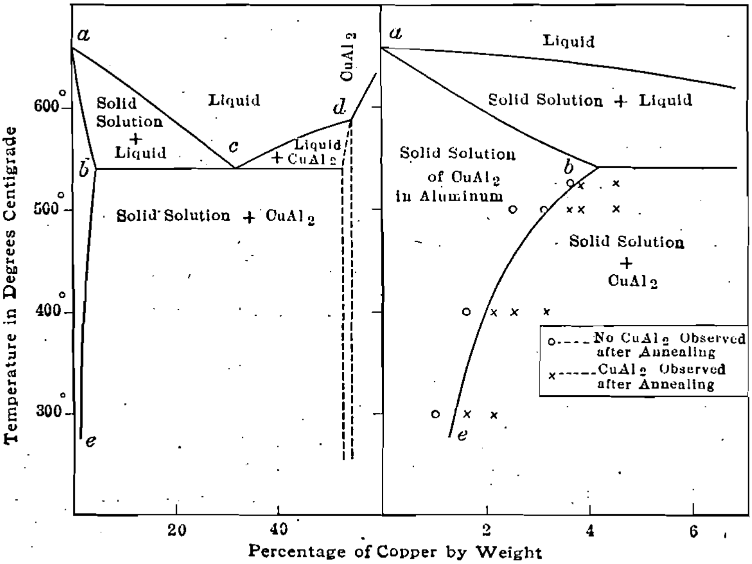

FIG. 12.—PORTION OF EQUILIBRIUM DIAGRAM OF COPPER-ALUMINUM ALLOY SERIES SHOWING SOLUBILITY CURVE *be* OF $CuAl_2$ IN ALUMINUM.

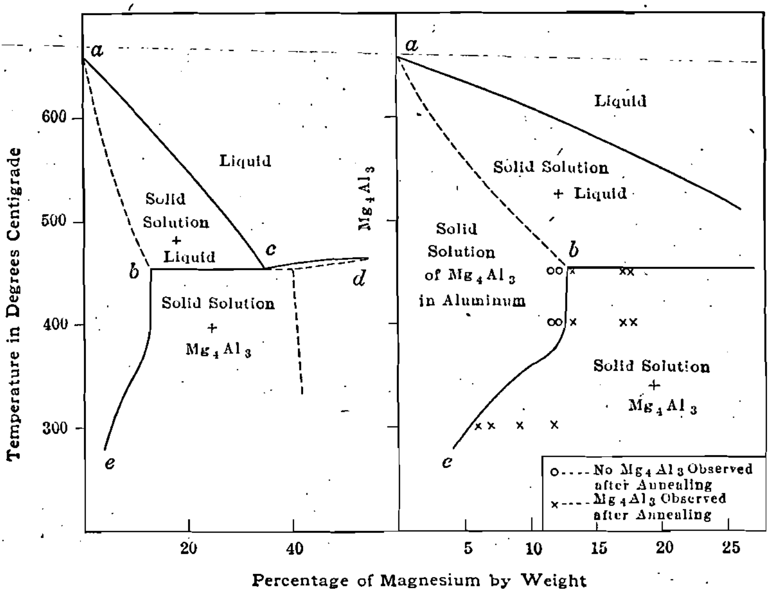

FIG. 13.—PORTION OF EQUILIBRIUM DIAGRAM OF MAGNESIUM-ALUMINUM ALLOY SERIES SHOWING SOLUBILITY CURVE *be* OF Mg_4Al_3 IN ALUMINUM.

in favor of this theory is largely of an indirect nature; the only direct confirmation of its truth is furnished by the results of thermal analysis.

Upon heating a specimen of duralumin that has been quenched from 500° C., but not aged, an evolution of heat occurs at from 250 to 275° C. This is shown in inverse-rate heating curves of three compositions, C-8, C-11, and N-34 in Figs. 14, 15, and 16. No thermal change takes place upon cooling the same specimen, provided it has not been heated beyond 520° C. Upon reheating the same slowly cooled specimen, no evolution of heat is found corresponding to that upon the first heating. Without doubt, therefore, a chemical reaction takes place at 250 to 275° C. upon heating the quenched sample with evolution of heat; *i.e.*, indicating

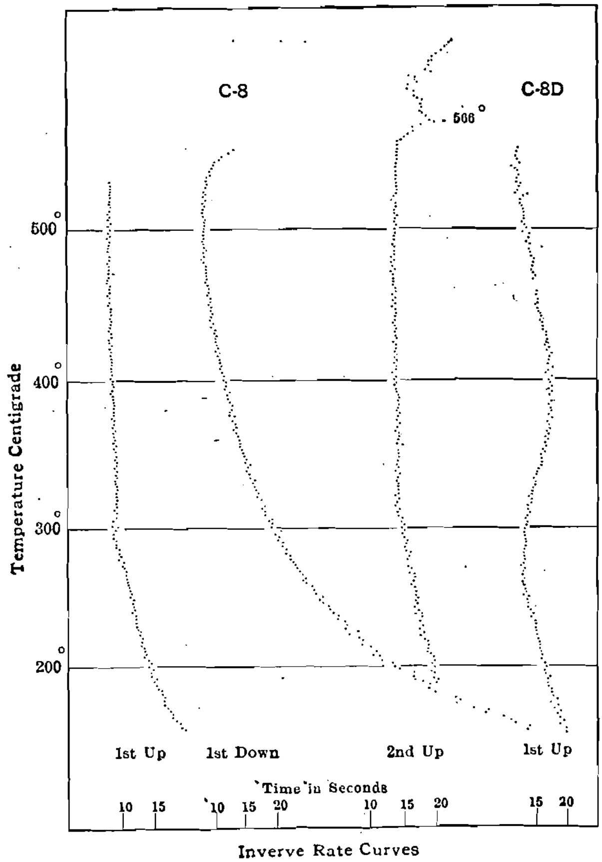

FIG. 14.—HEATING AND COOLING CURVES OF C-8. FIRST RUN UP SHOWING ARREST AT 300° C. WAS TAKEN 3 HR. AFTER QUENCHING. C-8-D IS A CURVE OBTAINED ON A QUENCHED SAMPLE AFTER AGING 18 MO. AT 20° C.

the formation of stable from unstable phases, not a transformation of stable to other stable phases, the two systems being in equilibrium during the transformation. Such a transformation must take place with heat absorption upon raising the temperature.

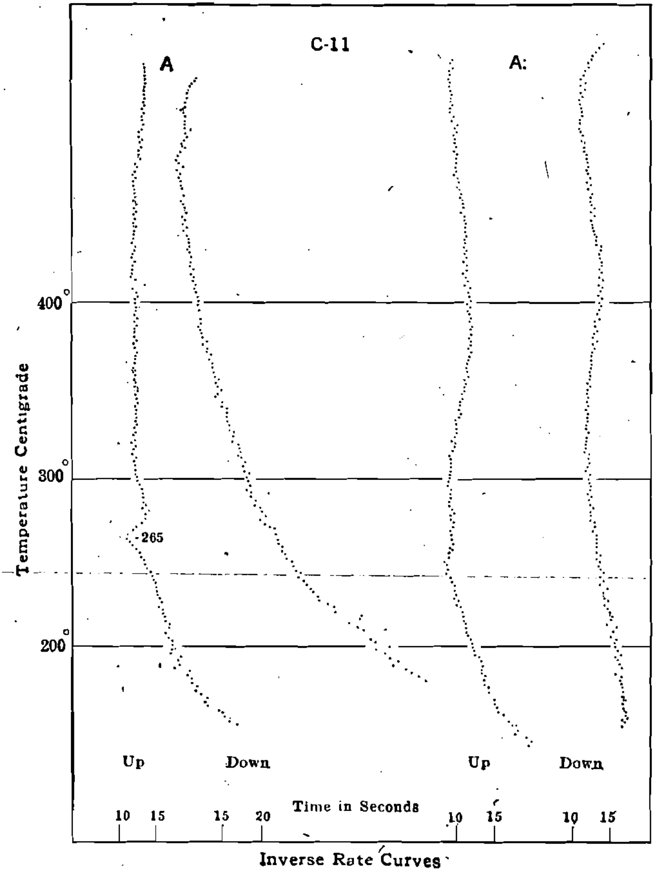

FIG. 15.—HEATING AND COOLING CURVES OF C-11. FIRST RUN UP ON C-11-A SHOWS AN ARREST AT 265° C. IN A SAMPLE THAT HAD BEEN QUENCHED AND THEN AGED 12 DAYS AT 20° C. SECOND RUN UP, MARKED C-11-A, SHOWS PRACTICALLY NO ARREST IN A SAMPLE THAT HAD BEEN QUENCHED IN BOILING WATER AND AGED 10 DAYS AT 120° C.

A specimen that has been quenched and aged at from 100° to 150° C. to secure maximum hardness shows little or no evolution of heat upon heating, see curves in Figs. 14, 15, and 16. Whatever may be the chemical reaction that is indicated on the first heating curve of a quenched specimen, it has taken place during the aging of the specimen, during which the hardening also occurs; stable phases have formed and the subsequent heating curve shows no arrest corresponding to that of the quenched specimen.

This chemical reaction can hardly be other than the precipitation of $CuAl_2$ from its super-saturated solution in aluminum, although direct visual evidence bearing on this question is also lacking. In describing the attempt made to recognize microscopically the phase change during aging just predicated, a digression must be made in order to discuss the general features of the microstructure of duralumin, which has apparently never been done before.

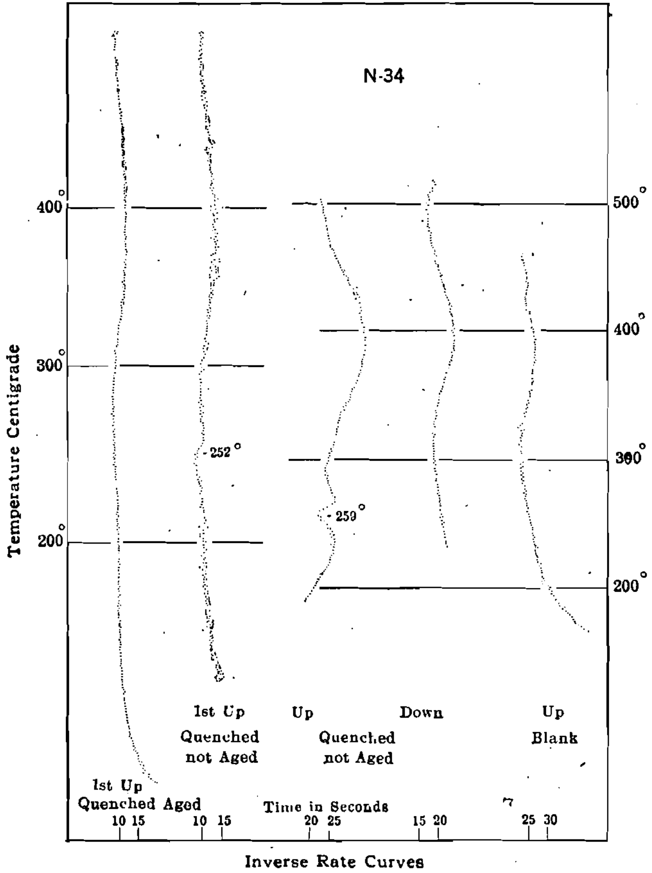

FIG. 16.—HEATING AND COOLING CURVES OF N-34 SHOWING INVERSE ARREST IN QUENCHED SAMPLES THAT HAD NOT BEEN AGED BUT NO ARREST IN SAMPLES THAT HAD BEEN AGED AFTER QUENCHING.

Structure of Duralumin

The microstructure of duralumin may be developed either by etching in a relatively concentrated solution of sodium hydroxide, NaOH, a dilute solution of hydrofluoric acid, HF, or in a dilute solution of NaOH. The grain structure of the alloy is best developed by the two former solutions; 10 per cent. NaOH and 5 per cent. HF are generally used for this purpose. For the identification and study of the different microscopic constituents of the alloy, a 0.1 per cent. solution of NaOH has, however, shown itself much superior to the former ones, and this solution has been used in most of the authors' investigations.

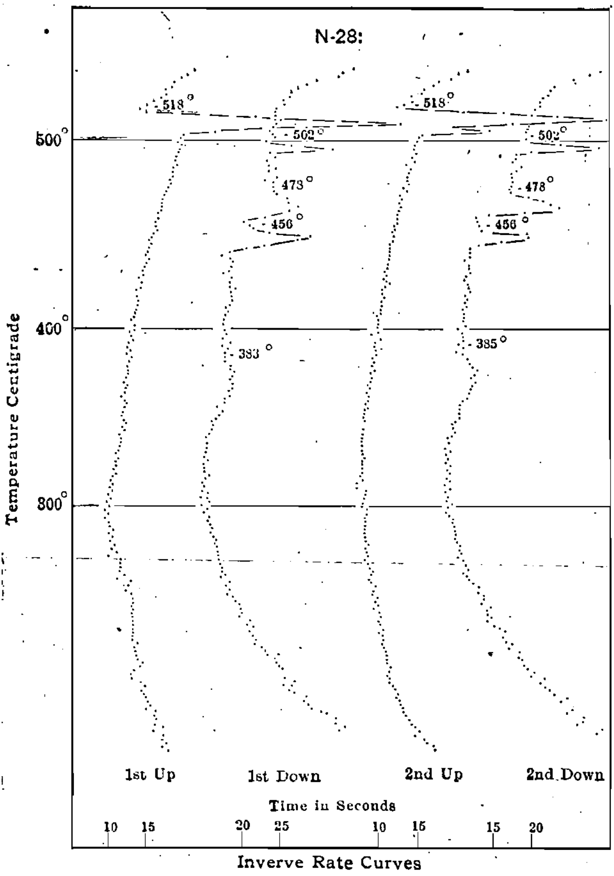

FIG. 17.—HEATING AND COOLING CURVE OF N-28. (CU 4.98%, MG 2.41%.)

Duralumin after rolling shows a structure similar to that in Fig. 18, which is quite typical. Fig. 19 shows the same alloy at a higher magnification. Grains of aluminum (in which are dissolved Si, $CuAl_2$, and Mg_4Al_3) are surrounded by strings of islands of eutectic ($CuAl_2$—aluminum, $FeAl_3$—aluminum and possibly others), which are white in Fig. 18. Upon examination under a higher power, the eutectic is seen to consist of two constituents; one a brownish color, the other white. These two constituents are evident in Fig. 19. In another article by the authors,[10] these two constituents have been identified as $FeAl_3$ (brown) and $CuAl_2$ (white), respectively. Quite often, but not always, the $FeAl_3$ surrounds the $CuAl_2$, as is shown in the figure.

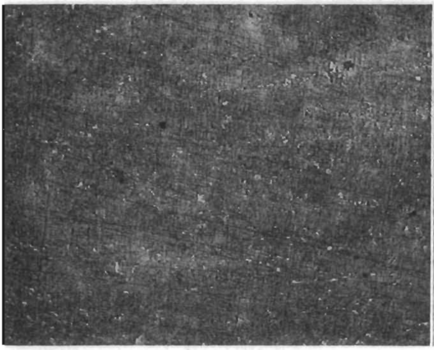

Fig. 18.

Fig. 19.

Fig. 18.—Rolled duralumin N-34. Etched with 0.1 per cent. NaOH. × 65.

Fig. 19.—Rolled duralumin, N-34(E4). Etched with 0.1 per cent. NaOH. × 650.

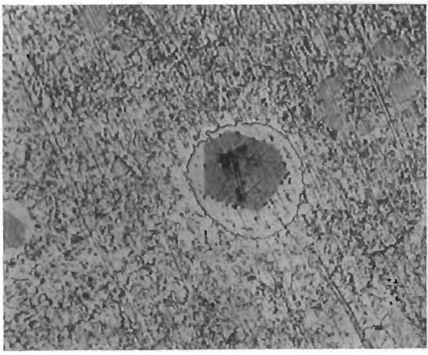

Fig. 20.

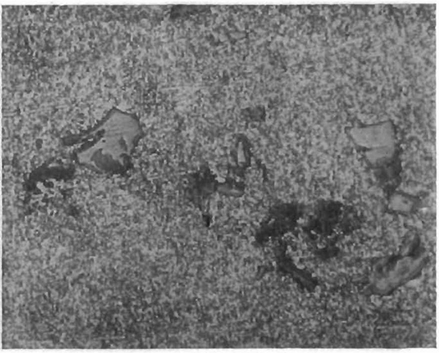

Fig. 21.

Fig. 20.—Specimen of N-28 containing Cu and Mg; showing island of Mg_2Si (dark) within one of $CuAl_2$ (white). × 650.

Fig. 21.—Sample of duralumin, E3; showing $FeAl_3$ and $CuAl_2$ eutectic, and fine particles throughout groundmass. × 650.

Besides these two constituents, a third, of pronounced bluish color, is visible. This is readily distinguished under the microscope, not always so readily in a photograph; it is seen within an island of $CuAl_2$ in Fig.

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[10] Reference 9.

20. In the same article, the authors have expressed the opinion that this is Mg_2Si; it occurs only in alloys containing magnesium.

Upon still closer observation, the grains of aluminum solid solution are seen to contain minute particles of a constituent, shown in Figs. 21 to 25, which are so small that it is impossible to identify them with certainty. Inasmuch as they occur also in aluminum itself, they must consist, in part at least, of the compound X (of iron, silicon, and probably also, aluminum,[11]) and possibly $FeAl_3$; probably $CuAl_2$ is also present

FIG. 22.

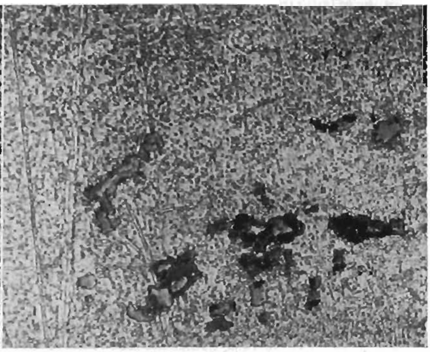

FIG. 23.

FIG. 22.—SAMPLE OF E3, SAME AS FIG. 21. × 1285.

FIG. 23.—SAMPLE OF DURALUMIN E3-F, AFTER ANNEALING 20 HR. AT 500° C., QUENCHING AND AGING AT ROOM TEMPERATURE. × 650.

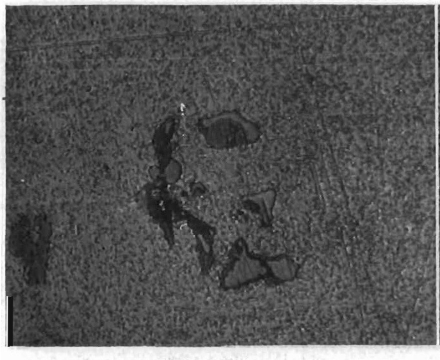

FIG. 24.

FIG. 25.

FIG. 24.—SAME MATERIAL AS IN FIG. 23. × 650.

FIG. 25.—SAMPLE OF DURALUMIN, E4-F, AFTER ANNEALING 20 HR. AT 500° C., QUENCHING AND AGING AT ROOM TEMPERATURE. × 650.

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[11] Reference 9.

in this form. All of this generation of particles have undoubtedly separated during cooling from a solid solution in aluminum at higher temperatures.

The visible structure of duralumin changes but slightly on heat treatment. Rolled duralumin consists of elongated grains. Upon heating such material to 500° C., recrystallization of the aluminum (solid solution) grains first occurs, and the fine grains so formed increase in size. This growth is naturally interrupted by quenching. Immediately after quenching, therefore, the grains may be either larger or smaller than the original ones, depending on the period of heating at 500° C. and the rate of heating to that temperature. During subsequent aging, the grains do not change in size. Heating to 500° C. also results in the solution of some or all of the $CuAl_2$ eutectic grains seen in the rolled material, to correspond to equilibrium. The $FeAl_3$ does not dissolve.

If there occurs during the aging of duralumin after quenching a gradual precipitation of $CuAl_2$ particles to correspond to its diminished solubility at the lower temperatures, one would expect to be able to observe some difference between the microstructure of the quenched unaged specimen and that of the quenched one after thorough aging. The particles of $CuAl_2$ may quite well be too small to be resolvable microscopically but the presence of a large number of such colloidal particles might [illegible] expected to accelerate the etching of the specimen; at least troost[illegible] etches much more readily than martensite or sorbite, and it is considered quite generally to consist of a colloidal solution of Fe_3C in alpha iron. Samples of N-34, some of which had been heated at 500° C. quenched in water and immediately etched, and some of which had been subsequently aged at 130° C. after identical treatment to develop maximum hardness, were carefully compared in their appearance after etching in the same solution (0.1 per cent. NaOH) and for the same periods of time. No difference was observed in the structure nor in the general shades of the etched surfaces of these two groups of specimens. The authors have to date, therefore, no direct structural evidence of the precipitation of $CuAl_2$ during aging of duralumin.

A difference in the rate of etching of quenched, unaged, and of quenched and aged may quite possibly be obscured by the presence of other constituents in fine dispersion, present in both cases. It was noted above that there are always present a number of fine particles of the X constituents. A structural study of duralumin made with pure aluminum, free from iron and silicon, might yield more positive results.

It is interesting to note that although the velocity of nuclear formation of $CuAl_2$ at temperatures from 20° to 400° C. seems to be quite normal, judging by thermal analysis, the velocity of crystallization or of coalescence of the nuclei is evidently quite remarkably small. Thus it was found[12] that there was no visible precipitation of $CuAl_2$ in an alloy containing 3 per cent. of copper upon annealing at 300° C. for 20 hr., after obtaining all of the $CuAl_2$ in solution by annealing at 500° C. Only by very slow cooling from 500° to 20° C. could a visible precipitate of $CuAl_2$ be produced. Slow velocities of crystallization seem to be characteristic both of $CuAl_2$ and of aluminum.

Although it cannot be directly proved that the thermal arrest at about 250° C. noticed upon heating a quenched unaged specimen of duralumin is due to the precipitation of $CuAl_2$, no evidence directly contradicts this assumption, which is in entire accord with our knowledge of the equilibrium within the alloy, and this arrest cannot be assigned to any other phase change.

It has been shown by many previous investigations, and confirmed by the authors, that aluminum undergoes no transformation in the solid state between ordinary temperatures and its melting point. No other phase changes could occur in the main mass of duralumin, the grains of solid solution, therefore, except those of solution or precipitation of $FeAl_3$, of the X compound, of $CuAl_2$, of Mg_4Al_3, or of Mg_2Si within the grains. Aluminum, which contains the same amounts of $FeAl_3$ and of the X compound as does duralumin, is not altered by heat treatment as is [illegible]lumin, nor does it show a reverse heat effect upon heating as does [illegible]tter. This heat effect must, therefore, be due to the precipitation [illegible]er of $CuAl_2$, Mg_4Al_3, or Mg_2Si. But the alloys containing only magnesium, in amounts up to 3 per cent., also do not harden upon aging. There remains only the precipitation of $CuAl_2$ with which to explain this heat effect.

The theory outlined above of the mechanism of the hardening of duralumin during aging most readily explains the interesting fact discovered by Mr. Blough, and confirmed by the authors, that the amount of hardening during aging increases as the temperature of quenching increases. At higher quenching temperatures, more and more $CuAl_2$ is dissolved in solid solution. After quenching the $CuAl_2$ is in excess of its solubility; the higher the quenching temperature the greater is the excess, and this is precipitated during aging. The hardening is in proportion to the amount of the highly dispersed $CuAl_2$ formed.

If this theory is accepted for the moment, it is interesting to consider the effect of degree of dispersion upon hardness in the case of a solid solution, in this case of $CuAl_2$ in aluminum. Duralumin immediately after quenching is generally softer than it is in the annealed condition. Thus alloy C-11, in the form of sheet, gave the following values of hardness:

	Annealed at 300°	Quenched, but not Aged	Quenched and Aged 8 Days
Scleroscope hardness, magnifying hammer.....	17	16	35

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[12] Reference 9.

This is probably due to the fact that a specimen, as ordinarily cooled after annealing, still contains some dissolved $CuAl_2$ in excess of its solubility; the material hardens slightly during cooling. Specimens cooled extremely slowly give a scleroscope hardness of from 7 to 10, much lower than that of the quenched, unaged ones.

Upon aging a quenched specimen at 200° C., for example, the hardness first increases to a maximum and afterward decreases. During that aging there has been first a formation of fine nuclei of $CuAl_2$ followed by coalescence of these particles into ones of larger size. There is, therefore, a certain average size of particle of $CuAl_2$, for which the hardness of the material is a maximum; atomic dispersion of the solute, $CuAl_2$, is not the dispersion that produces the maximum hardness, but some intermediate one between it and that at which the particles become visible by ordinary means.

It is interesting to observe that the properties of other light alloys of aluminum are influenced by heat treatment and aging. Thus Rosenhain and Archbutt[13] have found that the tensile strength of sand-cast aluminum-zinc alloys increases upon aging. In another article[14] by two of the authors, it has been shown that whereas alloys of aluminum-magnesium, aluminum-manganese, aluminum-manganese-magnesium, and aluminum-nickel do not harden upon quenching and aging; those of aluminum-magnesium-nickel do. The solubility of zinc in aluminum decreases from 40 per cent. at the eutectic temperature to about 25 per cent. at 256° C. and is probably much less at still lower temperatures. As in the case of the copper-aluminum alloys, decreasing solubility at lower temperatures of the constituent, $CuAl_2$ or zinc, is accompanied by the possibility of hardening by quenching and aging.

Inasmuch as the aluminum-nickel-magnesium alloys also harden by aging, we may expect an appreciable solubility of $NiAl_3$ in solid aluminum at higher temperatures. The solubility of $MnAl_3$ is undoubtedly quite low.

Analogy Between Hardening of Duralumin and That of Steel

The hardening of duralumin upon the basis of this hypothesis presents an interesting analogy with that of steel. The hardening of steel is due to the partial or entire suppression of the eutectoid transformation; most recent thought regards it as due more directly to the suppression of the cementite precipitation (as pearlite), the transformation of γ and α iron having taken place at least in part. The partial suppression there-

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[13] Report to the Alloys Research Committee, *Proc.* Inst. Mech. Eng. (1912) 319.

[14] P. D. Merica, A. G. Waltenberg and A. N. Finn: *Op. cit.*

fore of the precipitation of a compound from a solid solution is common both to rapidly cooled steel and to duralumin.

A sample of steel that has been hardened but not tempered shows an evolution of heat upon heating[15] through its tempering range exactly as does duralumin. This is due to the precipitation of Fe_3C in finely divided form in the case of steel exactly as it seems to be due to that of $CuAl_2$ in duralumin.

During the tempering or aging of steel at from 100° to 300° C., the hardness usually decreases immediately; *i.e.*, the maximum hardness of steel is obtained by quenching alone whereas that of duralumin is produced by tempering after aging. In the case of some high-carbon steels (from 0.9 to 1.7 per cent. C), however, the hardness increases during tempering after quenching exactly as in the case of duralumin.[16] The maximum hardness in hardened steel increases with the carbon content, as it does in duralumin with the copper content. It has been found that tool steel containing tungsten undergoes an increase of hardness during tempering at from 400° to 650° C. after quenching from 1350° C.[17]

Eutectic Structure and Influence of Magnesium

There is one fact that is not readily explained by the authors' hypothesis. Although alloys containing only magnesium and no copper do not harden and alloys containing only copper with no magnesium do harden, those containing both copper and magnesium undergo a much greater hardening than do those with copper alone. Magnesium, therefore, exerts no effect by itself in this direction and is not essential to the hardening power, but it materially increases the effect of the copper. The hypothesis developed above does not indicate any reason for this effect.

The authors are of the opinion that the influence of the magnesium is of a secondary nature. Thus, it seems probable that some magnesium unites with the silicon present to form Mg_2Si, the blue constituent always found in alloys containing magnesium. The removal of the silicon in this manner may be the direct cause of the resultant increase of hardening effect. This would agree with the observed fact that with the usual silicon content, 0.5 per cent. magnesium is enough fully to develop the

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[15] H. Scott: Effect of Rate of Temperature Change on Transformations in Alloy Steel, *Sci. Paper* No. 335, U. S. Bureau of Standards (1919); also *Bull.* A. I. M. E. (February, 1919).

[16] E. Maurer: Härten und Anlassen von Eisen und Stahl, *Metallurgie* (1909) **6,** 33.

[17] Edwards and Kikkawa: *Jnl.* Iron and Steel Inst. (1915) **92,** 6.

partly latent hardening power of the copper-aluminum alloys; the addition of more magnesium produces a somewhat harder alloy in all conditions but does not materially increase the hardening effect. This is shown by the following comparison:

ALLOY	COPPER	MAGNESIUM	TENSILE STRENGTH ANNEALED, POUNDS PER SQ. IN.	TENSILE STRENGTH HARDENED, POUNDS PER SQ. IN.	INCREASE IN TENSILE STRENGTH UPON HARDENING, PER CENT.
C-11	2.6	1.3	35,000	56,000	60
C-12	3.2	0.5	23,000	49,000	110

Consideration of the test results of Table 2 shows that magnesium hardens the aluminum matrix considerably even in the annealed condition. It is probable that the alteration of this matrix affects markedly the dispersion of the precipitation of $CuAl_2$ during aging and, consequently, the mechanical properties obtained.

There is another feature of the structure of duralumin that is of great importance and in which may be found some part of the explanation for the effect of magnesium. This is the manner in which the $FeAl_3$ and the $CuAl_2$ eutectics crystallize. There are several possible binary eutectics in duralumin, namely the following:

EUTECTIC	EUTECTIC TEMPERATURE, DEGREES CENTIGRADE
$FeAl_3$ + aluminum solid solution	640–650
Si (cryst) + aluminum solid solution	570–580
X compound + aluminum solid solution	610
$CuAl_2$ + aluminum solid solution	520–540
Mg_4Al_3	450
Mg_2Si	440

The amounts, by volume, of the eutectics with $FeAl_3$ and with $CuAl_2$ in ordinary duralumin are fairly large and about equal, that with Mg_2Si somewhat less, that with X and with Mg_4Al_3, usually, almost nil. The approximate temperatures of eutectic solidification are given above; they represent in all cases the temperatures observed in the presence of both the $FeAl_3$ and the X eutectic. The presence of $CuAl_2$ or Mg_2Si lowers the eutectic temperatures of the other binary eutectics. Thus in the presence of Mg_2Si, the eutectic temperature of $CuAl_2$-aluminum is reduced from 540 to 520–530° C., and this is always obtained as a thermal arrest in heating or cooling duralumin.

The order of solidification of these binary eutectics in aluminum-rich alloys is a matter of the greatest importance. Fig. 26 shows the probable form of the equilibrium at the aluminum end of the ternary system, Al-

Cu-Fe. An alloy containing about 0.5 per cent. Fe and 3 per cent. Cu (at *g* in the figure) would follow the line *gf*-*fc* upon solidification. A solid solution of aluminum with $CuAl_2$ ($FeAl_3$ is almost insoluble in aluminum) first crystallizes and the composition of the liquid changes along the curve *gf* with lowering of temperature. At *f*, and along *fc*, the binary eutectic $FeAl_3$-aluminum solid solution crystallizes. The liquid remaining at *f* is contained in the interstices between the solid grains of aluminum solid solution, and the $FeAl_3$ crystallizes upon these grains at the boundary between solid and liquid. At *c*, the binary eutectic $CuAl_3$-aluminum solid solution also crystallizes with the remainder of the first eutectic. The resultant structure is shown in Figs. 19, 20, 21, 22, 23, and 24. The $FeAl_3$ often entirely surrounds and isolates the $CuAl_2$ crystals.

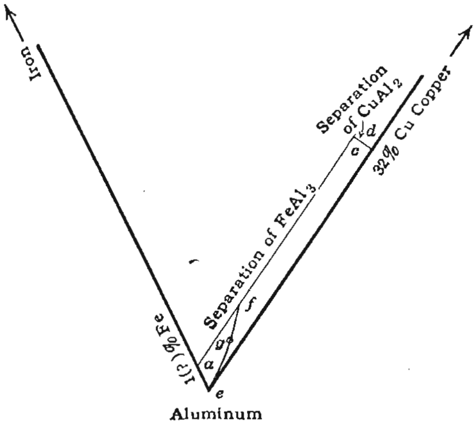

FIG. 26.—SUGGESTED FORM OF LIQUIDUS SURFACES OF TERNARY SYSTEM ALUMINUM-IRON-COPPER NEAR ALUMINUM END.

When a specimen having such a structure is heated to 500° for quenching, much of the $CuAl_2$ may be separated from the aluminum by this layer of insoluble $FeAl_3$ and is effectually prevented from dissolving. Thus E3-F, containing only 1.56 per cent. Cu, heated 20 hr. at 500° C. and quenched still contains free $CuAl_2$, although its solubility at that temperature was about 3 per cent. Its structure is shown in Fig. 23. The undissolved $CuAl_2$ (light) is surrounded by $FeAl_3$ (dark). (The other light islands are Mg_2Si, which are distinguishable under the microscope as of bluish color, but photograph light). This enclosure of the compound of one binary eutectic by that of another seems to be characteristic of light aluminum alloys.

Fig. 20 shows an island of $CuAl_2$ inclosing one of Mg_2Si. Such a structure explains probably the confusing heating and cooling thermal curves often obtained with copper-aluminum-magnesium alloys. In Fig. 16 were shown several normal heating and cooling curves for N-34 containing both copper and magnesium. The inverse heat effect in the quenched alloy at about 360° C. and the eutectic arrest at 510° C. in all, are both visible. In Fig. 17 are shown the heating and cooling curve of N-28 containing: Cu, 4.98 per cent.; Mg, 2.41 per cent. On the up curve, the usual 520° C. arrest is noticed; on cooling, however, instead of one, three arrests are noticed: at 502° C., at 478° C., and at 456° C. This cycle will repeat itself indefinitely, not only in this alloy but in others containing copper and magnesium, particularly when of rather high copper and magnesium content.

The structure of Fig. 20 was obtained in N-28 after the thermal analysis was completed and is characteristic; practically all of the Mg_2Si is surrounded completely by $CuAl_2$. Upon cooling, $CuAl_2$ separates at the first arrest (500° C.); at the second and third, Mg_2Si, and possibly some traces of Mg_4Al_3. These crystallize inside of the $CuAl_2$; the aluminum particles of the respective eutectics coalesce with the aluminum grains. Upon reheating this alloy, the surface of contact between Mg_2Si and aluminum is so slight that the melting of the eutectic, which should normally occur at the two lower cooling arrests, proceeds too slowly to give an arrest; and not until the protecting sheath of $CuAl_2$ melts as eutectic at the higher (520° C.) arrest does the Mg_2Si melt also. Those thermal arrests obtained around 500° C. are related to the formation of the various eutectics and do not have anything to do with the hardening of duralumin.

Conclusions Relative to Manufacture and Heat Treatment of Duralumin

It has been shown that when duralumin is rapidly cooled by quenching from temperatures between 250° and 520° C., and aged thereupon at temperatures from 0° to 200° C., the hardness and, at least at lower aging temperatures, the ductility increases. The actual values of hardness and ductility thus obtained depend on the quenching temperatures; they increase with that temperature up to about 520° C., corresponding to the increase of $CuAl_2$ in solid solution. At this temperature, any free $CuAl_2$ melts as a eutectic and the material is spoiled; this eutectic temperature, therefore, marks the upper limit of the useful quenching temperature range.

In order to develop the best mechanical properties by heat treatment, a quenching temperature should be used as near this as is possible without running risk of burning the metal by the melting of this eutectic. In practice, it should be possible to quench from temperatures between 510° and 515° C. The period of time at which sheet material should profitably be held at the quenching temperature lies between 10 and 20 min. Heavier sections, such as bars, might require more time at this