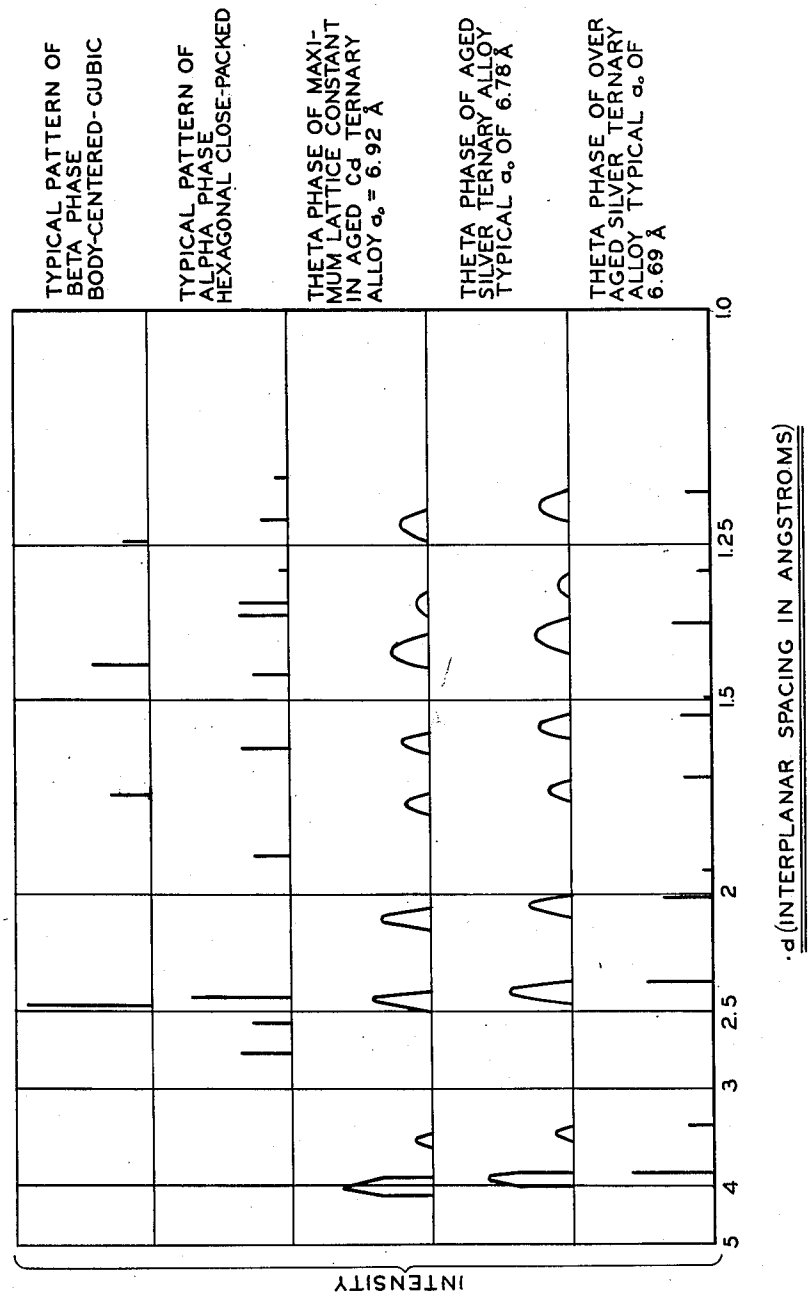
Dec. 16, 1952 A. H. HESSE 2,622,049

METHOD OF PRODUCING AGE-HARDENED MAGNESIUM-BASE ALLOY

Filed May 10, 1950 2 SHEETS—SHEET 1

FIG. 1




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METHOD OF PRODUCING AGE-HARDENED MAGNESIUM-BASE ALLOY

Filed May 10, 1950 2 SHEETS—SHEET 2

FIG. 2

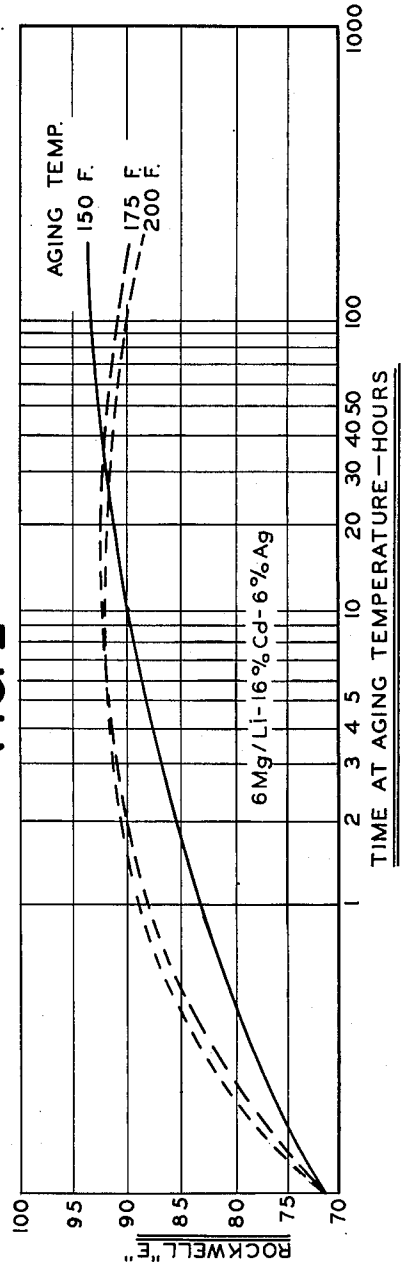
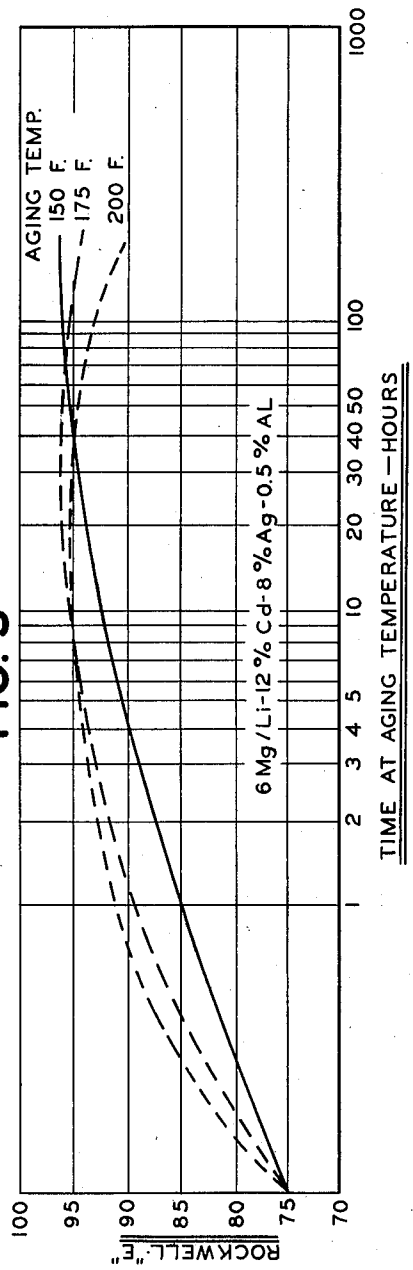



FIG. 3




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2,622,049

# UNITED STATES PATENT OFFICE

2,622,049

## METHOD OF PRODUCING AGE-HARDENED MAGNESIUM-BASE ALLOY

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Application May 10, 1950, Serial No. 161,247

2 Claims. (Cl. 148—21.9)

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This invention relates to the age-hardening of magnesium-base alloys comprising magnesium and lithium in the ratio to one another of 5:1 to 10:1. The invention is based upon the discovery of magnesium-base alloy compositions consisting of lithium and one or more other metals resulting in the formation of a new crystal phase which I shall hereinafter call "theta" phase, and the significant discovery that such alloys having the theta phase are susceptible to age-hardening.

Magnesium-base binary alloys containing lithium are composed either of the magnesium-rich terminal solid solution alpha phase, or the lithium-rich terminal solid solution beta phase, or both, depending upon the composition of the alloy. The addition of a sufficient quantity of one or more other metals of the group consisting of zinc, cadmium, silver and aluminum to the magnesium-lithium binary alloy produces the theta phase alloys. Magnesium-base alloys comprising magnesium and lithium in the ratio to one another of 5:1 to 10:1, including one or more of the metals consisting of 4–10% zinc, 4–24% cadmium, 0–12% silver and 4–12% aluminum have the theta phase and are age-hardenable according to the invention. The invention provides a heat-treatment method of producing age-hardened magnesium-base alloys having this theta phase resulting in a precipitation of at least a part of the theta phase in the age-hardened alloy.

In the accompanying drawings:

Fig. 1 illustrates representative X-ray diffraction patterns of phases observed in magnesium-base alloys.

Fig. 2 is an aging curve of an alloy comprising 6 Mg/Li, 16% Cd and 6% Ag, and

Fig. 3 is an aging curve of an alloy comprising 6 Mg/Li, 12% Cd, 8% Ag, and 0.5% Al.

The theta phase has been identified by means of X-ray diffraction (X R D) patterns as a face-centered-cubic structure having a lattice constant ranging from 6.5 Å. to 6.92 Å. in the alloys thus far investigated. In other alloys of this type, this lattice constant for theta phase may be slightly greater than 6.92 Å. or somewhat less than 6.5 Å. As indicated in Fig. 1, pattern 1 is a typical pattern of the beta phase or body-centered cubic lattice structure, pattern 2 is typical of the hexagonal close-packed magnesium structure known as alpha phase, pattern 3 is exemplary of a diffuse pattern of a theta phase structure of maximum lattice constant, pattern 4 illustrates a diffuse theta phase pattern having a relatively smaller lattice constant, and pattern 5 shows the sharp, well-defined pattern of the theta phase typically observed in over-aged alloys. Pattern 3 was obtained from a cadmium ternary alloy, whereas patterns 4 and 5 were obtained from a silver ternary alloy.

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In Fig. 1, the diffraction lines making up the various patterns are plotted with the visually estimated relative intensity shown on the vertical axis and the interplanar spacing shown in angstrom units on the horizontal axis (an Angstrom, abbreviated Å., is equal to $10^{-8}$ centimeter). In magnesium-base alloys which contain the theta phase, the alpha phase and beta phase are also normally present to some extent and the X-ray diffraction photogram of the sample examined will be a mixture of the X-ray diffraction patterns of all the phases present; however, the patterns of these phases may be all separated from one another by using a device such as the graph in Fig. 1.

Patterns 3 and 4 are diffuse theta-phase patterns, as represented by the broad bands indicated for each diffraction maximum, which are obtained upon the examination of alloys which have been age-hardened. The diffraction lines in pattern 4 are each slightly to the right of the corresponding lines shown in pattern 3, since the lattice constant of the alloy which produced pattern 4 is somewhat less than that in the alloy which produced pattern 3. A still further shift of the diffraction lines to the right is shown in pattern 5 which is a pattern for the same alloy as that of pattern 4, but after overaging at a higher temperature. The sharpness of pattern 5 is typical of theta-phase patterns obtained in examination of over-aged alloys, and it is because of the sharp diffraction patterns so obtained that it is usually preferable to examine the alloys in the over-aged condition to determine the presence of theta phase.

One of the outstanding characteristics that identifies the theta phase and differentiates it from other face-centered-cubic structures is the high intensity of the third diffraction line from the left shown in patterns 3, 4, and 5 (recognized by persons familiar with the art as having Miller indices 220), as compared with the intensity of the first diffraction line from the left (Miller indices 111) in patterns 3, 4 and 5. In the usual pattern of a metal or terminal solid solution having a face-centered-cubic structure, the third line (Miller indices 220) has only about one-third the intensity of the first line. For theta phase, however, the intensity of these two lines is almost equal. Furthermore, a face-centered-

cubic metal or terminal solid solution would have a much smaller lattice constant. A further distinctive feature of the theta-phase diffraction pattern is the presence of a line at Miller indices of 200, shown as the second line from the left in patterns 3, 4 and 5. The simultaneous existence of these two features helps to readily identify the theta phase. As will be apparent to those skilled in the art, the diffraction patterns represented in Fig. 1 are merely typical, and slight departures therefrom in intensities and lattice spacings will be found in various types of theta-phase patterns.

In the following Table 1, the Miller indices, the visually estimated intensities, the same relative intensities placed on a numerical scale, and the interplanar spacings of the lines found in a typical theta-phase pattern are tabulated. The X-ray tube had an iron target or anode and the sample, photographed in a Debye camera, comprised filings filed under 200-proof ethyl alcohol and dried. The alloy employed contained 4% zinc, 6% cadmium, 6% silver, the balance being magnesium and lithium in the weight ratio to one another of 7:1. Patterns of theta phase can also be observed in diffraction patterns obtained on an X-ray diffraction spectrometer, either from block samples or from filings.

TABLE 1

*Table of X-ray diffraction lines observed for theta phase*

| Miller indices (HKL) | Visually estimated intensity of line | Relative intensity of line on numerical scale | Interplanar spacing in angstrom units when lattice constant equals 6.75 Å. |
|---|---|---|---|
| 111 | Strong | 1.0 | 3.89 |
| 200 | Faint | .3 | 3.37 |
| 220 | Medium-strong | .8 | 2.39 |
| 311 | Medium | .6 | 2.04 |
| 222 | Very, very faint | .05 | 1.95 |
| 400 | Faint | .3 | 1.69 |
| 331 | Faint to medium-faint | .4 | 1.55 |
| 420 | Very, very faint | .05 | 1.51 |
| 422 | Medium-faint | .45 | 1.38 |
| 333, 511 | Very faint | .15 | 1.30 |
| 440 | Faint | .3 | 1.19 |

The points of distinction for identifying the theta phase brought out above in connection with the patterns in Fig. 1 are further emphasized by the above table. The line having Miller indices of 111 had a relative intensity of 1.0, whereas the line having Miller indices of 220 had a relative intensity of 0.8. The line having Miller indices of 200 had an intensity of 0.3.

I have found that none of the magnesium-lithium binary alloys form the theta phase, but that a large number of magnesium-base alloys containing other added metals do form theta phase and respond to age-hardening treatment, as shown in Table 3.

In carrying out this process, the usual pretreatments such as hot working, cold working, extruding, etc. may be employed, but are not requisite to the age-hardening treatment. In general, it is desirable to solution heat treat the alloy, that is, to put all or a major part of the theta phase into solution. The alloy may be hot worked or subjected to any other treatment which aids in dissolving the theta phase. The solution of theta phase involves two inter-related factors, namely, time and temperature. Below certain temperatures, theta phase will not go into solution regardless of time and this solution temperature varies from one alloy to another. As the temperature is increased, the time required for complete solution is decreased. Usually, temperatures of from 400 to 600° F. are satisfactory and holding periods of from a few minutes to 16 hours may be employed. The following Table 2 shows the effect of the solution treating temperature on an alloy containing 66.9% magnesium, 11.1% lithium, 16% cadmium, and 6% silver.

After all of the theta phase has gone into solution, the alloy is quenched with sufficient rapidity to prevent precipitation of the theta phase. This is normally accomplished by a water quench, although any other quenching means which effects a sufficiently rapid quench can be employed.

TABLE 2

| Solution treating temperature, ° F. | Effect on precipitate | A. S. T. M. grain size, number | Maximum Rockwell E hardness reached on aging at 150° F. |
|---|---|---|---|
| 500 | Dissolved | 3½ | 93 |
| 425 | do | 5 | 94 |
| 400 | do | 6 | 94 |
| 350 | Coarse, undissolved particles | 6-7 | 89 |

The above table shows that below 400° F. theta phase was not completely dissolved, but solution treatment at 400° F. produced an alloy containing a finer grain size than the same alloy solution treated at 500° F. Grain size appears as a direct function of the solution treating temperature.

The solution heat treated alloy must then be reheated at a temperature substantially below the solution temperature in order to bring about the precipitation of the theta phase. For most of these magnesium-base alloys, a temperature of about 150° F. produced the best age-hardening conditions, although certain alloys age-hardened as high as 300° F. and others age-hardened at 70° F. As is shown in Table 3, however, alloys which hardened only at 70° F. required such a long period of aging that the process would be commercially impracticable. On the other hand, alloys which age-hardened at 300° F. were generally quite unstable. Aging times of about 100 hours have been found preferable. Referring now to Table 3, it is evident that for both heat Nos. 25 and 32 higher aging temperatures hardened the alloys more rapidly than the lower aging temperatures. In both cases, however, aging at 150° F. produced as high or higher hardnesses than aging at 175 and 200° F., so that the ultimate hardness obtained was not affected materially by the aging temperature. This is probably due to the fact that the maximum hardness is attained before all of the theta phase precipitates, and the precipitation of additional theta phase merely over-ages the alloy.

In order to enable those skilled in the art to better practice the present invention, the following detailed examples are set forth:

EXAMPLE 1

An ingot of heat No. 25 with a composition of 66.9% Mg, 11.1% Li (6 Mg/Li), 16% Cd, and 6% Ag was extruded at 450° F. from a diameter of 2 inches to a strip of ¼ inch thick by ⅞ inch width. The strip was given one hot-rolling pass at 30% reduction at 550° F., air cooled to room temperature and then reduced by cold rolling 15%. The cold-rolled strip was solution treated by holding for 16 hours at 500°

F. To protect it from oxidation during solution treating the strip was coated with a silicone varnish. After solution treating, the strip was water quenched and aged in air at temperatures of 70° F., 150° F., and 300° F. As shown in Table 3, it required 2000 hours at 70° F. for this alloy to reach maximum hardness, whereas only 85 hours were required to reach the same maximum hardness at 150° F. At 300° F. this alloy aged rapidly (1.5 hours) but only increased from 74 Rockwell E to 76 Rockwell E. An additional fact, not shown in Table 3, is that it required only 48 hours at 150° F. to obtain a hardness of 91 Rockwell E, as compared with the maximum hardness of 92 Rockwell E obtained at the end of 85 hours.

EXAMPLE 2

An ingot of heat No. 32 with a composition of 66.9% Mg, 11.1% Li (6 Mg/Li), 16% Cd, and 6% Ag was forged at 550° F. from about 4 inches square to a cross-section of 4 inches by 1⅝ inches. The forged ingot was machined to clean up at 1.462 inches thickness and then hot rolled at 550° F. to 0.668 inch thickness in the cross forging direction. It was then turned and hot rolled at 550° F. in the forging direction to 0.250 inch thick. The temperature was then dropped to 400° F. and rolling continued in the forging direction to a thickness of 0.065 inch. Tensile specimens from this sheet were coated with a silicone varnish and solution treated by holding at 400° F. for 16 hours. The solution-treated hardness was 73 Rockwell E. On aging at 150° F. for 135 hours the hardness was increased to 93.5 Rockwell E. In the aged state, the tensile yield strength was 47,800 p. s. i., the ultimate tensile strength was 49,200 p. s. i., and elongation was 1.8%.

This application is a continuation-in-part of my copending application Serial No. 792,232, filed December 17, 1947 (now abandoned).

TABLE 3

| Heat No. | Mg/Li | Intended composition, percent | | | Hot rolling temperature, ° F. | Cold rolling reduction, percent | Solution treating temperature, ° F. | Aging temperature, ° F. | Hardness as solution treated | Maximum aged hardness | Time to reach maximum aged hardness, hours | Time at maximum aged hardness, hours | Duration of test, hours | Aged hardness at end of test |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| | | Cd | Ag | Other | | | | | | | | | | |
| 1 | 6 | 4 | 6 | 0.5Al | 550 | 15 | 500 | 70 | 63 | 87 | 2,500 | (1) | 2,500 | 87. |
| | | | | | | | | 150 | 62 | 76 | 130 | 350 | 2,500 | 70. |
| | | | | | | | | 300 | 62 | 62 | | | | Did not age harden. |
| 2 | 6 | 4 | 6 | 1Al | 550 | 15 | 500 | 70 | 67 | 91 | 2,500 | (1) | 2,500 | 91. |
| | | | | | | | | 150 | 68 | 90 | 19 | 90 | 1,000 | 78. |
| | | | | | | | | 300 | 68 | 68 | | | | Did not age harden. |
| 3 | 6 | 4 | 6 | 6Al | 550 | 15 | 500 | 70 | 86 | 99 | 900 | 1,600 | 2,500 | 99. |
| | | | | | | | | 150 | 85 | 94 | 1.5 | 24 | 1,000 | 78. |
| | | | | | | | | 300 | 85 | 85 | | | | Did not age harden. |
| 4 | 6 | 4 | 6 | 12Al | 550 | 2 | 500 | 70 | 93 | 97 | 30 | 420 | 2,500 | 93. |
| | | | | | | | | 150 | 91 | 94 | .5 | 2.5 | 1,000 | 78. |
| | | | | | | | | 300 | 91 | 91 | | | | Did not age harden. |
| 5 | 6 | 4 | 6 | 1Tl | 550 | 15 | 600 | 70 | 57 | 77 | 2,000 | (1) | 2,000 | 77. |
| | | | | | | | | 150 | 56 | 68 | 500 | 280 | 3,000 | 63.5. |
| | | | | | | | | 300 | 57 | 57 | | | | Did not age harden. |
| 6 | 6 | 4 | 6 | 8Tl | 550 | 15 | 600 | 70 | 75 | 87 | 2,000 | (1) | 2,000 | 87. |
| | | | | | | | | 150 | 81 | 89 | 120 | 140 | 3,000 | 78. |
| | | | | | | | | 300 | 75 | 75 | | | | Did not age harden. |
| 7 | 6 | 4 | 6 | 4Si | 550 | 15 | 600 | 70 | 70 | 86.5 | 2,000 | (1) | 2,000 | 86.5. |
| | | | | | | | | 150 | 69 | 69 | | | | Did not age harden. |
| | | | | | | | | 300 | 69 | 69 | | | | Do. |
| 8 | 6 | 4 | 6 | 1.2In | 550 | 15 | 600 | 70 | 56 | 80.5 | 2,000 | (1) | 2,000 | 80.5. |
| | | | | | | | | 150 | 56 | 65 | 2,000 | 1,000 | 3,000 | 65. |
| | | | | | | | | 300 | 56 | 56 | | | | Did not age harden. |
| 9 | 6 | 4 | 6 | 2Sn | 550 | 15 | 600 | 70 | 61 | 61 | | | | Do. |
| | | | | | | | | 150 | 60 | 64 | 1,500 | 1,500 | 3,000 | 64. |
| | | | | | | | | 300 | 60 | 60 | | | | Did not age harden. |
| 10 | 6 | 4 | 6 | 8Sn | 550 | 15 | 600 | 70 | 70 | 88.5 | 2,000 | (1) | 2,000 | 88.5. |
| | | | | | | | | 150 | 70 | 70 | | | | Did not age harden. |
| | | | | | | | | 300 | 70 | 70 | | | | Do. |
| 11 | 6 | 4 | 6 | 8Bi | 550 | 15 | 600 | 70 | 61 | 72 | 2,000 | (1) | 2,000 | 72. |
| | | | | | | | | 150 | 60 | 65 | 1,500 | 1,500 | 3,000 | 65. |
| | | | | | | | | 300 | 60 | 60 | | | | Did not age harden. |
| 12 | 6 | 4 | 6 | 2Pb | 550 | 15 | 600 | 70 | 59 | 59 | | | | Do. |
| | | | | | | | | 150 | 59 | 63 | 2,000 | 1,000 | 3,000 | 63. |
| | | | | | | | | 300 | 59 | 59 | | | | Did not age harden. |
| 13 | 6 | 4 | 6 | 8Pb | 550 | 15 | 600 | 70 | 67 | 83 | 2,000 | (1) | 2,000 | 83. |
| | | | | | | | | 150 | 65 | 68 | 1,500 | 1,500 | 3,000 | 68. |
| | | | | | | | | 300 | 65 | 65 | | | | Did not age harden. |
| 14 | 6 | 4 | 6 | 2Sb | 550 | 15 | 600 | 70 | 58 | 58 | | | | Do. |
| | | | | | | | | 150 | 58 | 67 | 2,500 | 500 | 3,000 | 67. |
| | | | | | | | | 300 | 58 | 58 | | | | Did not age harden. |
| 15 | 6 | 4 | 6 | 2Bi | 550 | 15 | 600 | 70 | 57 | 57 | | | | Do. |
| | | | | | | | | 150 | 57 | 62 | 2,000 | 1,000 | 3,000 | 62. |
| | | | | | | | | 300 | 57 | 57 | | | | Did not age harden. |
| 16 | 6 | 4 | 6 | 1.5AlMn | 550 | 15 | 600 | 70 | 60 | 87 | 2,000 | (1) | 2,000 | 87. |
| | | | | | | | | 150 | 61 | 82 | 72 | 170 | 2,500 | 71. |
| | | | | | | | | 300 | 61 | 61 | | | | Did not age harden. |
| 17 | 6 | 4 | 6 | 1.5BiTl | 550 | 15 | 600 | 70 | 57 | 77.5 | 2,000 | (1) | 120 | 59. |
| | | | | | | | | 150 | 57 | 61 | 500 | 2,000 | 2,500 | 61. |
| | | | | | | | | 300 | 57 | 57 | | | | Did not age harden. |
| 18 | 6 | 4 | 6 | 1.5CaTl | 550 | 15 | 600 | 70 | 57 | 76.5 | 2,000 | (1) | 2,000 | 76.5. |
| | | | | | | | | 150 | 56 | 68 | 140 | 2,360 | 2,500 | 68. |
| | | | | | | | | 300 | 56 | 60 | 1 | 50 | 1,000 | 56. |
| 19 | 6 | 4 | 6 | 1CuZn | 550 | 15 | 600 | 70 | 62 | 74 | 2,000 | (1) | 2,000 | 74. |
| | | | | | | | | 150 | 62 | 65 | 1,500 | 1,000 | 2,500 | 65. |
| | | | | | | | | 300 | 62 | 62 | | | | Did not age harden. |
| 20 | 6 | 4 | 6 | 1InNa | 550 | 15 | 600 | 70 | 58 | 78.5 | 2,000 | (1) | 2,000 | 78.5. |
| | | | | | | | | 150 | 57 | 65 | 1,500 | 1,000 | 2,500 | 65. |
| | | | | | | | | 300 | 57 | 57 | | | | Did not age harden. |
| 21 | 6 | 4 | 6 | 2SbTl | 550 | 15 | 600 | 70 | 61 | 82 | 2,000 | (1) | 2,000 | 82. |
| | | | | | | | | 150 | 60 | 79 | 140 | 860 | 2,500 | 74. |
| | | | | | | | | 300 | 60 | 60 | | | | Did not age harden. |
| 22 | 6 | 4 | 6 | 2SrTl | 550 | 15 | 600 | 70 | 57 | 79 | 2,000 | (1) | 2,000 | 79. |
| | | | | | | | | 150 | 56 | 72 | 170 | 330 | 2,500 | 68. |
| | | | | | | | | 300 | 57 | 57 | | | | Did not age harden. |

(See footnotes at end of table.)

TABLE 3—Continued

| Heat No. | Mg/Li | Intended composition, percent | | | Hot rolling temperature, ° F. | Cold rolling reduction, percent | Solution treating temperature, ° F. | Aging temperature, ° F. | Hardness as solution treated | Maximum aged hardness | Time to reach maximum aged hardness, hours | Time at maximum aged hardness, hours | Duration of test, hours | Aged hardness at end of test |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| | | Cd | Ag | Other | | | | | | | | | | |
| 23 | 6 | 8 | 6 | | 550 | 15 | 600 | 70 | 61 | 83 | 1,500 | 1,000 | 2,500 | 83. |
| | | | | | | | | 150 | 60 | 69 | 1,000 | 1,500 | 2,500 | 69. |
| | | | | | | | | 300 | 60 | 60 | | | | Did not age harden. |
| 24 | 6 | 12 | 6 | | 550 | 15 | 600 | 70 | 67 | 88 | 2,000 | 500 | 2,500 | 88. |
| | | | | | | | | 150 | 66 | 88 | 84 | 280 | 2,500 | 80.5. |
| | | | | | | | | 300 | 66 | 68 | 9 | 28 | 1,000 | 64. |
| 25 | 6 | 16 | 6 | | 550 | 15 | 600 | 70 | 75 | 92 | 2,000 | 500 | 2,500 | 92. |
| | | | | | | | | 150 | 74 | 92 | 85 | 500 | 2,500 | 85.5. |
| | | | | | | | | 300 | 74 | 76 | 1.5 | 11 | 1,000 | 69. |
| 26 | 6 | 4 | 12 | | 550 | 15 | 600 | 70 | 71 | 87.5 | 2,000 | 500 | 2,500 | 87.5. |
| | | | | | | | | 150 | 70 | 91 | 90 | 510 | 2,500 | 80.5. |
| | | | | | | | | 300 | 70 | 70 | | | | Did not age harden. |
| 27 | 6 | 4 | 6 | 4 InNa | 550 | 8.5 | [4] 600 | 70 | 64 | 85 | 2,000 | 500 | 2,500 | 85. |
| | | | | | | | | 150 | 63 | 83 | 28 | 320 | 2,500 | 73. |
| | | | | | | | | 300 | 63 | 65 | 4 | 25 | 1,000 | 60. |
| 28 | 6 | 12 | 2 | 1Al | 550 | 15 | 600 | 70 | 76 | 93 | 2,500 | [1] | 2,500 | 93. |
| | | | | | | | | 150 | 75 | 89 | 12 | 135 | 2,500 | 76.5. |
| | | | | | | | | 300 | 75 | 80 | .5 | 1.5 | 1,000 | 63. |
| 29 | 6 | 16 | 2 | 1Al | 550 | 15 | 660 | 70 | 80 | 95 | 1,500 | 1,000 | 2,500 | 95. |
| | | | | | | | | 150 | 79 | 93 | 15 | 130 | 2,500 | 81. |
| | | | | | | | | 300 | 79 | 88 | .5 | 1.5 | 1,000 | 70. |
| 30 | 6 | 12 | 2 | 0.5Al | 550 | 15 | 600 | 70 | 61 | 87.5 | 2,500 | [1] | 2,500 | 87.5. |
| | | | | | | | | 150 | 60 | 81 | 16 | 250 | 2,500 | 72. |
| | | | | | | | | 300 | 60 | 60 | | | | Did not age harden. |
| 31 | 6 | 12 | 4 | 0.5Al | 550 | 15 | 600 | 70 | 69 | 90 | 2,500 | [1] | 2,500 | 90. |
| | | | | | | | | 150 | 67 | 89 | 40 | 220 | 2,500 | 79. |
| | | | | | | | | 300 | 67 | 67 | | | | Did not age harden. |
| 32 | 6 | 12 | 8 | 0.5Al | 550 | 7.5 | 600 | 70 | 76 | 93.5 | 2,500 | [1] | 2,500 | 93.5. |
| | | | | | | | | 150 | 75 | 95 | 100 | 900 | 2,500 | 87.5. |
| | | | | | | | | 300 | 75 | 91 | .5 | 2 | 1,000 | 73. |
| 33 | 6 | 6 | 6 | 4Zn | 600 | 15 | 600 | 70 | 77 | 96.5 | 2,000 | 500 | 2,500 | 96.5. |
| | | | | | | | | 150 | 75 | 96 | 30 | 115 | 2,500 | 82.5. |
| | | | | | | | | 300 | 75 | 77 | 3 | 25 | 1,000 | 71. |
| 34 | 6 | | 4 | 4Zn-4Ca | 550 | 15 | [4] 500 | 70 | 68 | 92 | 1,500 | 1,000 | 2,500 | 92. |
| | | | | | | | | 150 | 69 | 72 | 85 | 155 | 1,000 | 65. |
| | | | | | | | | 300 | 69 | 69 | 0 | 1 | 1,000 | 59. |
| 35 | 6 | | 4 | 4Zn | 550 | 15 | 500 | 70 | 66 | 91 | 900 | 2,600 | 3,500 | 91. |
| | | | | | | | | 150 | 65 | 70 | 82 | 180 | 1,000 | 66. |
| | | | | | | | | 300 | 65 | 67 | 1 | 2.5 | 1,000 | 58. |

[1] The hardness of this alloy will probably remain at this level indefinitely.
[2] A second phase which could not be put into solution at 500° F. could be dissolved at 600° F.
[4] A second phase could not be put in solution at either 500° F. or 600° F.
NOTE.—All hardnesses in this table are reported on the Rockwell "E" scale.

I claim:

1. The method of producing an age-hardened magnesium-base alloy consisting of magnesium and lithium in the ratio to one another of 5:1 to 10:1 and at least one metal of the group consisting of 4–10% zinc, 4–24% cadmium, 0–12% silver and 4–12% aluminum and having a precipitated theta phase which comprises heating the alloy to dissolve at least a major portion of the theta phase, quenching the alloy to prevent precipitation of the theta phase, and thereafter heating the quenched alloy to a temperature above 150° F. but below 300° F. to precipitate at least a part of the theta phase to age-harden the alloy, said theta phase being identified by X-ray diffraction pattern as a face-centered cubic structure having a lattice constant ranging from 6.5 Å. units to 6.92 Å. units.

2. The method of producing an age-hardened magnesium-base alloy consisting of magnesium and lithium in the ratio to one another of 5:1 to 10:1 and at least one metal of the group consisting of 4–10% zinc, 4–24% cadmium, 0–12% silver and 4–12% aluminum and having a precipitated theta phase which comprises heating the alloy to a temperature of from 400° F. to 600° F. to dissolve at least a major portion of the theta phase, quenching the alloy by rapid cooling to prevent precipitation of the theta phase, and heating the quenched alloy to a temperature above 150° F. but below 300° F. to precipitate at least a part of the theta phase to age-harden the alloy, said theta phase being identified by X-ray diffraction pattern as a face-centered cubic structure having a lattice constant ranging from 6.5 Å. units to 6.92 Å. units.

ALFRED H. HESSE.

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