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ALLOYS OF ALUMINUM, THORIUM AND
URANIUM

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Gene E. Bobeck and H. A. Wilhelm

ABSTRACT

The aluminum-rich corner of the ternary alloy system aluminum-thorium-uranium has been investigated by thermal and microscopic analyses to determine an area of low-melting alloys that could be used as a liquid metal fuel in uranium-233 breeder reactors. A phase diagram of this aluminum-rich corner of the system has been proposed.

The data indicate that the binary eutectic valleys from the aluminum-uranium and aluminum-thorium systems meet at a point in the neighborhood of the composition of 76 weight per cent Al, 18 weight per cent Th and 6 weight per cent U to give a ternary eutectic which solidifies at 630°C. Although microstructures indicate two different primary phases as required, a third binary valley leading to this ternary eutectic has not been established.

Results of the investigation show that there is a rather wide range of compositions of uranium and thorium in aluminum, varying from 13 weight per cent uranium at zero weight per cent thorium to zero weight per cent uranium at 25 weight per cent thorium, that can be maintained in a one-phase liquid solution at temperatures even lower than 640°C. If a temperature of 650°C. is the minimum, then an even wider variation in uranium and thorium compositions can be tolerated for a single liquid-phase alloy.

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I. INTRODUCTION

Alloys of aluminum, thorium and uranium have been investigated to find a low-melting alloy containing sufficient uranium and thorium to be of possible use as a liquid metal fuel in a nuclear reactor. Although most of the research in the reactor field concerning fuels has been devoted to the development of solids as fuel elements, liquid metal alloys have received consideration as the fuel in certain reactor designs. The feasibility of using a liquid metal fuel has been discussed in some detail in a report on developments on this type of reactor at Brookhaven (1).

The components of the ternary alloy under investigation might be considered as serving separate and distinct purposes in a liquid nuclear fuel. First, the thorium could serve as the fertile material which would, through nuclear conversions, yield the atomic fuel uranium-233. The uranium, either as uranium-235 or uranium-233, could serve as the fissionable isotope component of the alloy. The aluminum, having a low melting point, could serve as the solvent for the high melting thorium and uranium to give a low melting liquid alloy fuel. The composition of the alloy then might be adjusted to a reactor design so as to give a regenerating liquid fuel in which essentially thorium would be added to make up for the spent fissionable atoms. There are many other details aside from preparation of the alloy that would have to be worked out to arrive at a practical operating reactor, but the composition limits and characteristics of the ternary alloy are of prime importance before any further definite steps can be proposed for such a reactor design.

To find a low melting alloy of aluminum, thorium and uranium, the liquidus surface of the system in the region of interest must be determined. Two methods of investigating the liquidus surface are thermal analysis and metallographic examination of ternary alloys of various compositions. Both methods were used in this study.

Aluminum was selected to be alloyed with uranium and thorium for several reasons. First, its melting point is relatively low. Second, the aluminum-uranium system is well-known and the aluminum-thorium system has been investigated. Both binary systems contain eutectics. Third, the neutron capture cross-section of aluminum is quite low, so that neutron conservation would not be seriously affected should such an alloy be used as a nuclear fuel. Fourth, the electronegativities and atomic radii of aluminum, thorium, and uranium are similar. Fifth, the crystal structures of aluminum and α -thorium are face centered cubic; the crystal structure of γ -uranium is orthorhombic. Since the crystal structure of α -uranium differs from that of aluminum and α -thorium, little ternary solid solubility is to be expected.

The binary system, aluminum-thorium, has been investigated by Saller and Rough, and their diagram is given in Fig. 1 (2). They report a eutectic consisting of aluminum and ThAl_3 at 3.7 atom per cent (25 weight per cent) thorium and at a temperature of 632°C . Some solid solubility of thorium in aluminum is indicated. Four intermetallic compounds have been characterized having the formulae ThAl_3 , ThAl_2 , Th_3Al_2 and Th_2Al (3).

The binary system aluminum-uranium shown in Fig. 2 has been determined by Gordon and Kaufmann (4). They report negligible solid solubility of uranium in aluminum. Three intermetallic compounds occur in the system; these compounds are represented by the formulae UAl_4 , UAl_3 , and UAl_2 . Borie reports UAl_4 as the formula of the intermetallic compound richest in aluminum (5). A eutectic occurs in the system of aluminum and UAl_4 at 1.7 atom per cent (13 weight per cent) uranium and at a temperature of 640°C .

The uranium-thorium system has been determined by O. N. Carlson (6), and his diagram is given in Fig. 3. A eutectic occurs at 2.5-3.0 weight per cent thorium and at a temperature of 1086°C . A liquid immiscibility gap exists over the range from 10 up to 72 weight per cent thorium with the base of the gap at 1375°C . and the top of the loop at 1650°C . A monotectic occurs at 72 weight per cent thorium and 1375°C . There is negligible solid solubility of one metal in the other.

From a consideration of the three binary alloy systems, it appeared that low-melting three-component alloys would be most likely to occur in the aluminum-rich areas of the ternary system. The present investigation was then limited to studies of those ternary alloys that involved the region of the two aluminum-rich binary eutectics. Determination of the shape of the three-component diagram and determination of the alloy composition and melting temperature of any ternary eutectic in this aluminum-rich region were the primary objectives of this study. As a result of this work a diagram showing a possible ternary eutectic is proposed for the aluminum-rich alloy region.

II. INVESTIGATION

The search for a suitable alloy that could be used as a liquid fuel in a nuclear reactor has led to the investigation of aluminum-rich alloys of aluminum, thorium and uranium. To know the temperatures at which these ternary alloys begin to solidify, it is necessary to determine the ternary liquidus surface. The liquidus surface represents the minimum operating temperatures of the various alloy compositions in a nuclear reactor.

Fig. 1. Aluminum-thorium phase diagram.

Fig. 2. Aluminum-uranium phase diagram.

Fig. 3. Uranium-thorium phase diagram.

The aluminum used in this investigation was 99.99 per cent pure Alcoa high purity aluminum ingot, which contained approximately 0.001 per cent silicon, 0.001 per cent iron, and 0.005 per cent zinc. The thorium metal was from a batch of Ames production thorium. Analytical data on thorium billet metal showed each of the impurities iron, nitrogen, aluminum, beryllium, silicon, and total rare earths to be generally less than 100 ppm, while the carbon and oxygen were often as high as 400 and 1200 ppm, respectively. The uranium metal was cut from a slug of reactor-grade normal uranium.

Alloys were prepared by vacuum induction heating with the equipment which is shown in Fig. 4. The vacuum system consisted of an oil diffusion pump backed by a mechanical pump; a six kilowatt Ajax mercury-arc converter supplied the power to the induction coil.

After the alloys were prepared they were then removed to another vacuum induction heating set-up for the thermal analysis experiments. The apparatus used to obtain the thermal data is shown in Fig. 5. This set-up consisted of a high vacuum system, a three kilowatt Ajax mercury-arc converter, and a Brown x-x type recorder. The arrangement of the parts of the vacuum system is shown in the block diagram in Fig. 6.

Details of the heating chamber are shown in Fig. 7. The prepared sample was contained in a magnesium oxide crucible which was mounted in insulation that had been cut from insulating fire brick. The support for the insulation was a steel plate fitted with three legs which stood on the steel vacuum base. The thermocouple wires to the heating chamber entered through a rubber stopper in a tee connection to the vacuum base. A silica tube open at one end was placed in an inverted position over the heating assembly and sealed to the vacuum base by means of a neoprene gasket. The sample was heated by means of self-induction from a coil around the outside of the silica envelope.

The vacuum system components were a Welch Duo-Seal mechanical vacuum pump of 300 liters per minute capacity which served as a fore pump to a National Research Corporation oil diffusion pump (model H-2-P).

B. Methods of Procedure

1. Alloy preparation

Alloys were prepared by co-melting the metals in a magnesium oxide crucible in the vacuum induction equipment shown in Fig. 4. For convenience in making up the alloys, the aluminum ingot was first recast in the form of a one-inch diameter rod. A section of the rod was placed in a crucible with lathe turnings of thorium and uranium which had been cleaned in acid to remove oxide film. The total weight of aluminum, thorium and uranium in each case was sufficient to give 100 grams of alloy. The crucible with the alloying constituents was put in a graphite

Fig. 4. Alloy preparation equipment.

Fig. 5. Thermal analysis equipment.

Fig. 6. Arrangement of the parts of the vacuum system used for thermal analysis.

Fig. 7. Details of heating chamber.

heater. This heater was supported in granular insulation contained in a magnesium oxide pot. The insulation material was graphite chips from lathe cuttings. These chips were placed between the outer wall of the heater and the inner wall of the pot. This assembly was placed on a firebrick base in the silica tube shown in Fig. 4.

The tube was evacuated to about 10^{-3} millimeters of mercury before power was applied to the induction coil around the outside of the tube. Heat generated in the graphite heater radiated to the inner crucible containing the alloying constituents. Each charge was held at a temperature of 1000°C . for 30 minutes to insure complete alloying of the constituents. The alloys were allowed to cool to room temperature in the vacuum. Two samples as prepared in this manner are shown in Fig. 8.

2. Thermal analysis

After an alloy was prepared and still contained in its magnesium oxide crucible, a one-eighth inch hole was drilled through its center from top to bottom. A magnesium oxide thermocouple shield was inserted in this hole and a thermocouple placed inside the shield. The specimen was set in place in the thermal analysis chamber shown in Fig. 7.

The chamber was evacuated to about 10^{-4} millimeters of mercury. Power from the three-kilowatt mercury-arc converter was applied to the induction coil, and the sample was heated self-inductively. The cold junction of the thermocouple was maintained at 0°C . by an ice-water bath. The progress of the heating was followed by means of the potentiometer recorder. When the sample was melted, as indicated by an increase in the slope of the heating curve on the recorder chart, the power to the induction coil was reduced, and the specimen was allowed to cool at a rate of about two degrees per minute. The cooling curve was recorded on the potentiometer chart.

3. Metallographic examination

After thermal analysis was completed on a specimen it was taken from the thermal analysis chamber and the crucible broken away. The cylindrical specimen was cut in two along its central axis by means of a silicon carbide cut-off wheel. A full-length slab was then cut from the middle of one of the halves for microscopic examination. One face of this slab was ground flat on a kerosene-lubricated belt sander, after which it was ground on a rotating disc machine having a series of four silicon carbide grinding papers of successively decreasing grit sizes. Water was used as a lubricant and as a coolant on the papers of this machine to prevent the surface of the specimen from smearing. The polishing was carried out on billiard-cloth covered wheels using first a liquid soap suspension of 400-grit and finally of 600-grit silicon carbide.

Fig. 8. Two samples prepared in equipment shown in Fig. 4.

Etching was not required to bring out microstructures of most of the alloys. Due to the differences of the phases in hardness and possibly in resistance to corrosion by the water used in polishing, the microstructures of the samples were often readily observable even before final polishing. However, in some cases the alloys were treated with a dilute hydrofluoric acid etch to bring out their microstructures to best advantage.

Since large sections were examined microscopically, some voids due to shrinkage on solidification and some segregation that had occurred on precipitation from the liquid were detectable in areas of the specimens. In each case an effort was made to select the area for the photomicrograph so as to represent an unaltered alloy.

4. Chemical analysis

Each alloy was quantitatively analyzed for aluminum, thorium and uranium. A wedge-shaped sample cut lengthwise from the other half of the sectioned alloy was selected for analysis.

The sample was brought into acid solution and the uranium reduced to uranium(IV) by passing this solution through a lead reductor. The uranium(IV) was then reoxidized to uranium(VI) with an excess of iron(III). The iron(II) produced was oxidized to iron(III) by titrating with a standard solution of cerium(IV). The volume of standard cerium(IV) solution needed to react with the iron(II) was a measure of the amount of uranium in the alloy.

The total equivalent of aluminum and thorium in a solution of the alloy was determined first by titrating with the di-sodium salt of Enta (ethylenediaminetetraacetic acid). Thorium was determined separately by precipitating thorium oxalate from a solution of the alloy. The thorium oxalate was then heated to the oxide and weighed. This amount of thorium determined gravimetrically was subtracted from the Enta determination of aluminum and thorium to give aluminum.

The compositions of the alloys as determined by chemical means quite often were short of 100 per cent, but always fell close to the compositions intended in their preparation.

III. RESULTS AND INTERPRETATION

On the basis of thermal and metallographic data and a knowledge of the two binary systems, aluminum-uranium and aluminum-thorium, a tentative phase diagram of the liquidus surface of the aluminum-rich corner of the

aluminum-thorium-uranium alloy system has been proposed. This diagram is represented in Fig. 9 as a polythermal projection of the liquidus surface of the region under investigation. The three areas labeled "A", "B" and "C" in Fig. 9 are primary phase fields. These primary phase fields are separated by eutectic curves or binary valleys. The binary valleys are drawn with dotted lines since there are still some questions as to their exact locations or actual existence. Although the evidence is not conclusive, a ternary eutectic point is assumed for these alloys, and it is represented on this diagram by the point E_4 . A primary phase field is that volume of a temperature-composition prism (diagram) for a three-component system in which one solid is in equilibrium with the melt. A binary valley is the line of intersection of two adjacent primary phase fields along which line the two solids of the adjacent primary phase fields are in equilibrium with the melt. The binary valleys in Fig. 9 are represented by the lines E_1E_4 , E_2E_4 , and E_3E_4 . The exact locations of the lines E_1E_4 and E_2E_4 , and the existence of the line E_3E_4 are still doubtful. At the ternary eutectic point E_4 , the three solid phases and the melt are in equilibrium.

The liquidus surface of most of the primary phase field labeled "A" in Fig. 9 has been well-established by the thermal and microscopic data on the alloys whose compositions lie in this field. The microstructures of most of these alloys show a dendritic pattern of aluminum that has solidified from the melt as the primary phase during slow cooling. Figure 10 is a typical microstructure from this phase field. Liquidus surfaces for the supposed two primary phase fields labeled "B" and "C" in Fig. 9 have not been established by the thermal and microscopic data on the alloys whose compositions lie in these two areas. The microstructures of the alloys in field "B" show that an intermetallic compound has solidified from the melt as the primary phase during slow cooling. Figure 11 is a typical microstructure from phase field "B". Similarly, the microstructures of the alloys in field "C" show that another intermetallic compound has solidified from those melts as the primary phase during slow cooling. Figure 12 is a typical microstructure from phase field "C". The identities of these two intermetallic compounds have not been determined. They are thought to be the binary intermetallic compounds UAl_4 and $ThAl_3$, rejected respectively as the primary phases from the melts of areas "B" and "C".

A. Thermal Analysis

The thermal data on all of the ternary alloys investigated in this study are given in Table 1. No evidence was found in the cooling curve data to support the existence of solid solubility over the range covered. The existence of minor solid solubility cannot be ruled out, however, since no special effort was made to determine this point. These thermal data and other data have been considered in the construction of a prism that represents the temperature-composition relationships for the liquidus

Fig. 9. Proposed phase diagram of aluminum-thorium-uranium system (tentative).

Fig. 10

Fig. 11

Fig. 12

Fig. 10. Typical microstructure from phase field "A", 85% Al, 7.5% Th, 7.5% U, 50X.

Fig. 11. Typical microstructure from phase field "B", 80.0% Al, 2.5% Th, 17.5% U, 50X.

Fig. 12. Typical microstructure from phase field "C", 72.5% Al, 22.5% Th, 5.0% U, 50X.

Table 1. Thermal data

Nominal composition (wt.%)			Temperatures of breaks (°C.)		
Al	Th	U	1st	2nd	3rd
90.0	2.5	7.5	649	640	626
90.0	8.75	1.25	650	634	
85.0	5.0	10.0	636	632	622
85.0	7.5	7.5	635	626	622
85.0	2.5	12.5	639	639	622
85.0	0.0	15.0	645	643	
85.0	3.75	11.25	641	639	621
85.0	6.25	8.75	643	639	630
85.0	10.0	5.0	644	631	630
85.0	12.5	2.5	642		630
85.0	15.0	0.0	642	632	
85.0	13.75	1.25	642	632	
85.0	8.75	6.25	643	634	630
80.0	5.0	15.0	648	647	636
80.0	7.5	12.5	633	629	624
80.0	2.5	17.5	643	642	626
80.0	3.7	16.3	643	641	627
80.0	6.3	13.7	641	640	630
80.0	10.0	10.0	636	635	630
80.0	12.5	7.5	636	633	631
80.0	15.0	5.0	636	633	
80.0	20.0	0.0	633	631	
77.5	16.3	6.2	630		
75.0	10.0	15.0	632	628	625
75.0	12.5	12.5	634	631	629
75.0	7.5	17.5	639	637	630
75.0	15.0	10.0	633	631	
75.0	20.0	5.0	634	633	
75.0	17.5	7.5	631	629	
72.5	17.5	10.0	632	631	629
72.5	22.5	5.0	631	630	
72.0	16.0	12.0	633	632	
71.3	18.7	10.0	631	631	630
70.0	15.0	15.0	631	630	
70.0	12.5	17.5	635	633	630

Table 1. (continued)

Nominal composition (wt.%)			Temperatures of breaks (°C.)		
Al	Th	U	1st	2nd	3rd
70.0	20.0	10.0	630	629	629
70.0	25.0	5.0	634	632	
70.0	17.5	12.5	633	630	628
70.0	22.5	7.5	630	629	
67.5	27.5	5.0	631	631	
67.5	22.5	10.0	634	631	
66.3	17.5	16.2	632	631	
65.0	20.0	15.0	630	629	628
65.0	25.0	10.0	630	629	627
65.0	30.0	5.0	632	630	
60.0	30.0	10.0	632	630	629
60.0	27.5	12.5	632	631	629
60.0	25.0	15.0	632	631	628
60.0	22.5	17.5	629	628	624
60.0	20.0	20.0	629	627	

of the ternary system. This prism, made of plaster, is shown in the photograph in Fig. 13. It shows the three primary phase fields "A", "B" and "C". As previously mentioned, most of the liquidus surface of the phase field "A" has been well-established. The liquidus surfaces of phase fields "B" and "C", however, are assumed on the basis of rather meager data to exist as shown. They are represented on the model in Fig. 13 as having relatively steep contours. If reference is made to Figs. 1 and 2, the binary aluminum-uranium and aluminum-thorium phase diagrams, it will be noted that the liquidus lines rise steeply in temperature on the uranium-rich side and on the thorium-rich side of their respective binary eutectics. On the basis of these well-established liquidus lines of these two binary diagrams, it has been assumed that the liquidus areas "B" and "C" of the ternary system of Fig. 13 also have rather steep contours as shown.

It will be noted by comparing the thermal data in Table 1 with the temperatures represented by the surfaces "B" and "C" on the model in Fig. 13 that these surfaces have not been constructed according to the data in the table. It has been assumed that the heat effects caused by the rejection of a solid phase from these melts at the liquidus surfaces for these compositions were too small to be detected during the thermal analysis. The compositions of the primary phases that are assumed to be separating at these surfaces are so far removed from the compositions of the alloys being investigated that only minor amounts of the primary phases should actually be ejected in cooling these alloys to the eutectic valleys.

Fig. 13. Model showing temperature-composition relationships of aluminum-thorium-uranium system.

The model which represents the proposed diagram for the aluminum-rich corner of the three-component system and which is shown in the photograph in Fig. 13 has actually been constructed of two plaster sections that fit together. In Fig. 14 a photograph of the model is shown in which one of these sections has been removed. That portion of the model which is shown in Fig. 14 represents the temperature-composition relationships of the areas "A", "B" and "C" according to the thermal and microscopic data taken on all of the alloys investigated.

Considering now the areas "B" and "C" in the model of Fig. 13 and their bordering liquidus lines in the two binary aluminum diagrams of Figs. 1 and 2, it is to be noted that the liquidus surfaces of these two areas as represented in this model appear to be consistent with these known binary systems. If we consider the binary diagrams and those portions of the model in Fig. 14 that represent its areas "B" and "C", it is quite improbable that those areas could be interpreted as properly representing liquidus surfaces. These surfaces in Fig. 14, therefore, have no significance in representing the true phase boundaries. The breaks in the cooling curves from which these surfaces were constructed are therefore considered as not representing primary solidification and are interpreted otherwise.

Since it is assumed that the actual point of primary phase separation was not detectable in the thermal analyses of the alloys under consideration, the first detectable breaks in the cooling curves were quite likely associated with secondary solidification. If, then, this liquidus break has not been observed in the cooling curves of the alloys whose compositions lie in phase fields "B" and "C", the breaks that have been observed in the cooling curves may be due to solidification along a binary valley and at a ternary eutectic point. The minor breaks in the cooling curves of these alloys (see Table 1), that is, those breaks that were just barely detectable and which differed in temperature from the major breaks by only one or two degrees, are then considered to be extraneous and due to small fluctuations in the cooling rate and not to real thermal arrests.

A typical cooling curve from phase field "B" is shown in Fig. 15. It exhibits two major breaks; the first one is assumed to be due to secondary solidification along a binary valley, and the second one to ternary eutectic solidification. This cooling curve is also typical in shape for the cooling curves obtained from alloys having compositions in phase field "C".

Figure 16 depicts a cooling curve of an alloy whose composition lies in phase field "A". This cooling curve exhibits three distinct breaks. The first of these breaks is assigned to primary solidification of aluminum at the liquidus surface. The second break is considered to be due to secondary solidification of two solids along a binary eutectic valley. The third break on cooling, then, is assumed to correspond to a ternary eutectic solidification.

Fig. 14. Model showing temperature-composition relationships of aluminum-thorium-uranium system according to thermal and microscopic data.

Fig. 15

Fig. 16

Fig. 15. Typical cooling curve from phase field "B".

Fig. 16. Typical cooling curve from phase field "A".

In some cases the cooling curve data from alloys having compositions in phase field "A" exhibited only two resolvable breaks. The compositions of the particular alloys in field "A" that gave only two breaks during thermal analysis were near pure aluminum or close to the proposed binary eutectic valleys. The first of the two breaks for compositions near pure aluminum is large and is due to primary solidification at the liquidus surface; the second break is due to eutectic solidification. It is to be noted that the drop in temperature along a binary eutectic valley bounding area "A" is very gradual. This should make differentiation of binary and ternary eutectic breaks in some cooling curves somewhat difficult. Furthermore, it can be assumed that the amounts of the residual melts that solidify at the ternary eutectic point in the cases considered are so small that their heat of solidification effects are insufficient to be resolved from any binary eutectic valley effect; the small binary and ternary eutectic effects then blend into one overall break in these cooling curves.

The cooling curves of alloys whose compositions lie in phase field "A" may also exhibit two breaks if the crystallization path as discussed by (7) leads directly to the ternary eutectic point. In this case the first break is assumed to be due to the material crystallizing from the melt at the liquidus surface; the second break is due to ternary eutectic solidification. In case the composition of the alloy is near or on a binary eutectic valley, then two breaks may be detected if the composition of the alloy is far enough removed from the ternary eutectic point. In any of the alloys having a composition near the ternary eutectic, it was difficult or impossible to resolve the cooling curve into more than two breaks.

One cooling curve that exhibited a single break was obtained for a particular alloy. The composition of this alloy lay in the area in which the phase fields "A", "B" and "C" were expected to meet at a point. From other data and the cooling curve of this alloy, the approximate composition of a ternary eutectic for this system has been assigned. The composition of this ternary eutectic alloy is 76 weight per cent aluminum, 18 weight per cent thorium and 6 weight per cent uranium. The temperature at which this ternary eutectic occurs has been determined as 630°C.

B. Microscopic Examination

The microscopic examination of all of the ternary alloys investigated in this study has not clearly located the boundaries of the three primary phase fields "A", "B" and "C" of Figs. 9 and 13. However, the microstructures of the alloys from each of these three fields exhibit certain characteristic features which have been used to estimate the locations of two of these binary valleys and to propose the existence of a third. These features are in those parts of the microstructures assigned to the primary phases, and they differ in appearance for the alloys from the three phase fields.

The alloys whose compositions lie in the phase field "A" exhibit a characteristic pattern of dendritic aluminum. This dendritic aluminum can be seen in Figs. 10, 17 and 18 as the light areas. The finely divided material that can be seen dispersed in this aluminum matrix is interpreted as having crystallized from the residual melt along both a binary valley and at the ternary eutectic point. The material crystallizing from these residual melts along their binary valleys will be aluminum and either the aluminum-uranium intermetallic compound or the aluminum-thorium intermetallic compound, depending on which binary valley the composition of the residual melt falls. At the ternary eutectic point of the diagram as proposed, the two intermetallic compounds and aluminum should crystallize from the residual melt.

The alloys whose compositions lie in phase field "B" reject the aluminum-uranium intermetallic compound from the melt during cooling, which forms crystals shaped like parallelograms. Figures 11 and 19 show the characteristic shape of these crystals. The finely divided material and light areas of aluminum that can be seen in those photomicrographs are due to eutectic transformations in which aluminum and the aluminum-uranium intermetallic compound formed along a binary valley, and to ternary crystallization at the ternary eutectic point. It is assumed that the slow rate of cooling employed in obtaining alloys for these microstructures has permitted considerable agglomeration of the aluminum after separation as eutectic.

The alloys whose compositions lie in phase field "C" reject the aluminum-thorium intermetallic compound as the primary phase from the melt during cooling. The crystals of this compound quite often exhibit hexagonal cross-sections in the microstructures, as can be seen in Figs. 12, 20 and 21. The presence of the finely divided material in these photomicrographs can be explained as it was in the previous two cases. The matrix is essentially aluminum containing the two intermetallic compounds in a finer state of division. The primary aluminum-thorium compound is rejected from the residual melt along a binary valley with either aluminum or the aluminum-uranium compound, depending on which binary valley the composition of the residual melt falls. At the ternary eutectic point, the two intermetallic compounds and aluminum solidify from the residual melt.

The photomicrographs that were selected for presentation here show structures considered to be characteristic of all of the alloys of their respective phase fields. These photomicrographs were more easily interpreted than some of the ones not shown. All of the alloys whose microstructures were examined in this study have undergone considerable annealing during the thermal analysis heating and cooling cycles. This annealing tends to agglomerate the phases, especially the aluminum phase, in the alloys, thus altering many of the features that would have shown in their microstructures with normal cooling rates. The microstructures of two alloys having compositions near the composition assumed for the ternary eutectic are shown in Figs. 22 and 23. These microstructures exhibit to a fair degree typical eutectic structures in spite of the annealing effect.

Fig. 17

Fig. 18

Fig. 19

Fig. 20

Fig. 17. Typical microstructure from phase field "A", 85% Al, 3.7% Th, 11.3% U, 50X.

Fig. 18. Typical microstructure from field "A", 85.0% Al, 5.0% Th, 10.0% U, 50X.

Fig. 19. Typical microstructure from field "B", 75.0% Al, 7.5% Th, 17.5% U, 50X.

Fig. 20. Typical microstructure from field "C", 60.0% Al, 30.0% Th, 10.0% U, 50X.

Fig. 21

Fig. 22

Fig. 23

Fig. 21. Typical microstructure from field "C", 65.0% Al, 30.0% Th, 5.0% U, 50X.

Fig. 22. Microstructure near the ternary eutectic, 77.5% Al, 16.2% Th, 6.3% U, 50X.

Fig. 23. Microstructure near the ternary eutectic, 72.5% Al, 17.5% Th, 10.0% U, 50X.

IV. LITERATURE CITED

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