On the Change of Internal Structures of Some Light Aluminium Alloys due to External Tension

By Keizo Niimi and Masao Seo

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Abstract

With various specimens of pure aluminium and its alloys subjected to different treatments, the process of the change of their crystalline configurations and crystal structures due to external tension, was investigated mainly by means of X-rays. Supplementary to the X-ray investigation, some mechanical properties and the microscopic structures of these specimens were also examined. From the experimental results thus obtained with specimens of considerable tensile strength and elongation percentage such as those of Cu-Al alloys containing 4% of Cu, this process of change of their internal structures, was found to be essentially the same as had already been observed by other investigators[1] with specimens of some pure metals, i. e., the micro-crystals in the specimens of Cu-Al alloys were confirmed to break and then to rearrange themselves in a fibrous-like manner having the normals to one of the (111) planes of their face-centred cubic lattice parallel to a definite common direction. But, when either the tensile strength or the elongation percentage of the specimens was small, as was the case with pure aluminium and the so-called "ESD" alloy, such a process of the internal structure due to external tension could be only partly observed.

Introduction

To obtain some fundamental information in connection with the mechanical properties of duralumin and other light aluminium alloys of industrial importance, the writers were induced to examine their internal structures in various stages of the breaking process due to external tension, mainly utilizing X-rays. By way of comparison, a similar X-ray examination was also performed with some specimens of pure aluminium and Cu-Al alloys containing 4% of Cu.

The experimental results thus obtained, show us that the changes of the crystal structures and the crystalline configurations of pure aluminium and some of its alloys due to external tension, were not necessarily the same as was suggested by the previous researches performed by numerous investigators[2] with other kind of metals and alloys sub-

1. G. Clark and M. Beckwith; Trans. Amer. Soc. Metals, **25**, 1207, (1937).
2. R. Jacqusson; Compt. rend., **205**, 331 (1937).
H. Möller and M. Hepmel; Mitt. Kaiser-Wilhelm-Institut für Eisenforschung, 20, 15 (1938).
C. Barett; Metals and Alloys, **8**, 13 (1937).

jected to various mechanical treatments. Furthermore, it was found that in a few respects, the conclusions drawn from the previous researches must be revised, even with regard to the internal structure of the poly-crystalline specimens of pure aluminium subjected to the external tension.[1] The change of the internal structures of pure aluminium and its alloys throughout the breaking process, was seen in the present experiment to be not always the same as was previously observed, but to vary its course according to the tensile strength and the elongation percentage of the specimens.

The experimental results, which led us to the above considerations, will briefly be described below.

Experimental Parts

1. *The Mechanical Properties and Microscopic Structures of the Specimens*:—Besides various specimens of pure aluminium and Cu-Al alloys containing 4% of Cu prepared by different procedures, those of several duralumin-type light alloys newly devised by the writers, were examined in the present experiment. Supplementary to these specimens, the so-called "ESD" alloy (a modified super-duralumin-type light alloy of extremely great strength) prepared at the Research Laboratory of the Sumitomo Metal Industries, Ltd., was also used as a specimen. The composition of each of the specimens above mentioned, is represented in the second column of Table I: While in the third and fourth columns of the same table, the procedures of obtaining these specimens and heat-treatments given to them before they were used, are tabulated respectively.

The properties of some of the specimens of alloys here tabulated had remained almost unknown, except with regard to those of "ESD" alloy and Cu-Al alloys containing 4% of Cu. So, the writers attempted to set up the present experiment, by making clear some mechanical properties of the specimens together with the outlines of their microscopic structures, for the purposes of reference to our further considerations.

To measure the tensile strength and elongation percentage, a test piece of each specimen 25 mm. in gauge length, 8 mm. in width and about 0.7 mm. in thickness, was used except in the case of the "ESD" alloy: In the latter case, the thickness of the test piece was altered to 2 mm. The observed values deduced from the experimental results,

1. G. Clark and M. Beckwith; ibid.

when the specimens were subjected to the standard fracture, i.e., the fracture taking place within the range of $\frac{1}{4}$ of the gauge length from the middle point of the specimen, are tabulated in the fifth and sixth columns of Table I. Here it must be postscribed that the hardness of some of the specimens was also studied, in addition to the aforesaid examinations of mechanical properties.

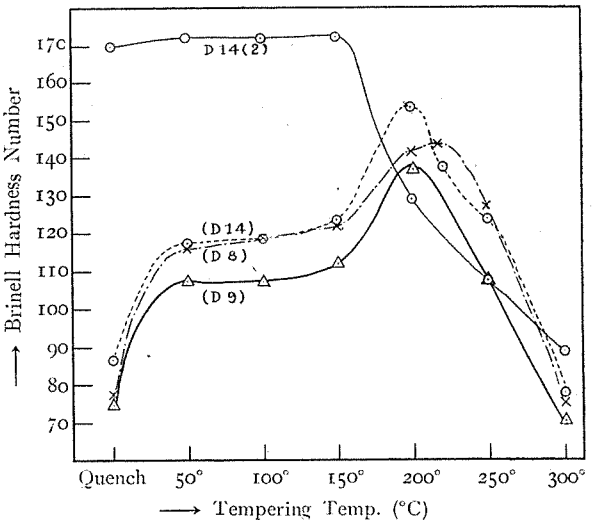

Fig. 1 Temper-Hardening Curves of some Duralumin-type Light Alloys (Tempered for 1 hr.)

Table I

No. of Specimen	Composition of the Specimen	Procedures of Preparing the Specimens	Heat-treatments given to the Specimens before use.	Tensile Strength (kg/mm²)	Elongation (%)
A1	Al containing 0.26% of Si, 0.12% of Fe and 0.01% of Cu as impurities	after chill-casting, annealed at 510°C for 1 hr. and then quenched into ice-water, before being cold-rolled by 93%	after annealing at 510°C for 1 hr. and quenching, tempered at 100°C for 1 hr	7.8	26.7
A2			tempered at 200°C	7.7	30.0
A3			tempered at 300°C	7.7	26.7
ACR	4% Cu-Al			40.9	5.0
ACQ			quenched after annealing at 510°C for 1 hr.	25.6	27.5
AC1			tempered at 100°C	28.4	16.3
AC2			tempered at 200°C	24.4	25.0
AC2.5			tempered at 250°C	26.4	22.5
AC3			tempered at 300°C	24.4	15.0
D8	3.5% Cu-Al containing 0.5% of Mg, Mn, Si, Ag, Ni and Cr, respectively	after chill-casting and annealing at 510°C for 1hr. hot-	tempered at 200°C	45.0	12.0

D9	3.5% Cu-Al containing 0.5% of Mg, Mn, Si, Ag, Ni and Mo, respectively	rolled at about 400°C and then cold-rolled by 50%	ditto	39.0	12.0
D9(2)	ditto	after chill-casting, annealed and hot-rolled as above stated, annealed again at 510°C for 1 hr. and then quenched into ice-water before being cold-rolled by 50%	annealed at 150°C for 1 hr.	41.4	14.0
D14(2)	4.5% Cu-Al containing 0.5% of Mg, Mn, Si, Ag, Ni and Cr, respectively		ditto	47.5	12.0
ESD	n. l.	n. l.	n. l.	52.7	9.0

As typical representations of the temper-hardening curves consequent upon the measurements of hardness in the present experiment with various specimens, those referring to some duralumin-type light alloys are given in Fig. 1.[1] It was noticed that the hardness represented by these curves as shown in Fig. 1, is comformable to the corresponding tensile strength of various specimens observed in the present experiment.

In Fig. 2, Plate I, some microscopic structures of the rolled surfaces of some specimens taken before the examination of their tensile strength, are compared with those observed after fracture due to tensile stress. To obtain Fig. 2, Plate I, the rolled surface of each specimen was etched by means of the Keller's method, after polishing. As a consequence of this comparison, it becomes clear that the crystal grains in all the specimens of Cu-Al alloys containing 4% of Cu, were remarkably lengthened by the external tension, as shown in Fig. 2(a) taken with Specimen AC1. The same process of change of the microscopic structures caused by tension, could also be detected, though not so conspicuously as in the cases of Cu-Al alloys, with the specimens of pure aluminium used in the present experiment. But, such an elongation of the crystal grains as above stated, was hardly observable in Figs. 2(b) and 2(c), Plate I, which were obtained with Specimen D8 of the duralumin-type light alloy and the "ESD" alloy respectively. By the way, it is noticed that the microscopic structures taken by etching the rolled surface of Specimen AC2 of the Cu-Al alloy, with a dilute solution of sodium hydroxide (containing 10% of NaOH), not only indicated

1. The temper-hardening curves in Fig. 1, excepting the uppermost one designated by D14 (2) without the bracket, were obtained with specimens of the same composition as Specimens D8, D9 and D14 (2) respectively left after chill-casting, instead of those subjected to the procedures tabulated in Table I. Nevertheless, they are reproduced only for the sake of reference.

the state of precipitation of the compound $CuAl_2$, but disclose a number of slip bands given rise to by the local contraction due to the tensile stress, as shown in Fig. 3, Plate I: While, the aforesaid slip bands could hardly be detected in the microscopic structures of the duralumin-type light alloys and "ESD" alloy.

Thus having been confirmed as above with regard to the mechanical properties and microscopic structures of various specimens, we may conclude that the magnitude of the deformation of crystal grains due to external tension, is influenced at least by the tensile strength and the elongation percentage. The crystal grains seem, within the scope of our investigation, to be lengthened the more remarkable in the specimens of larger tensile strength and elongation percentage such as those of the Cu-Al alloys. But, when one of these factors is comparatively small, the stretching of the crystal grains tends to be less detectable. Especially in the specimens of extremely small elongation percentage such as those of the duralumin-type light alloys or the "ESD" alloy, which were confirmed as undergoing hardly the local contraction by the tensile stress, the crystal grains remain almost unaltered. The microscopic structures of the fractured portions of the specimens, corresponding to the aforesaid cases of the deformable and undeformable crystal grains, are represented in Figs, 4(a) and 4(b), Plate I, respectively.

2. *Crystalline Configurations of the Specimens*:—Utilizing the heterogeneous X-rays emitted from the molybdenum anticathode, the changes of the crystalline configuration in the specimens due to external tension, were examined by the Laue method. To begin with this X-ray examination, the following two series of the changes of the crystalline configuration in a few specimens, were preliminarily investigated: The one corresponds to the alteration of the crystalline configuration at the central portion of the specimen, observed from time to time until the normal fracture takes place, which is caused by the successive addition of the external tension; While the other refers to the difference in the crystalline configurations of various portions of the specimen after the aforesaid destruction, which can be attributed to the magnitude of infliction of the tensile stress varying with the distance from the middle point of the test piece. By comparing the diffraction patterns obtained, it was confirmed that these two series of changes are always of essentially the same process. So the writers advanced their examination with regard to the crystalline configurations of various portions

of each specimen after destruction, instead of the internal structure of its central portion subjected to diverse magnitudes of the tensile stress before this fracture.

(i) *Pure Aluminium*:—By illuminating Specimen A2 of pure aluminium with the X-ray beam, before the infliction of external tension, we always obtained the diffraction pattern consisting of a set of the Debye-Hull rings each formed of an assemblage of many intense spots, as was reproduced in Fig. 5, Plate II. Essentially the same diffraction pattern as Fig. 5, was also given rise to with any one of the specimens of pure aluminium. This shows us that all the specimens of pure aluminium prepared by the procedures tabulated in Table I, are equally of polycrystalline aggregates made up by micro-crystals of considerable dimension (about 10^{-2} cm. in diameter).

Next, the writers extended their investigation to the effects of tensile stress upon the crystalline configuration of the specimens of pure aluminium. For this purpose, the diffraction patterns were taken with the following two portions of each specimen: The one is the portion situated at a distance of about $\frac{1}{4}$ of the gauge length from the middle point of the test piece (named *e* Portion for the sake of convenience), while the other being the central portion, in which the fracture usually took place (named *f* Portion), Figs. 6 and 7, Plate II, are the diffraction patterns corresponding to the aforesaid two portions of Specimen A2. To take these diffraction patterns, the direction parallel to the tensile stress in the specimens, was always placed vertically. As a consequence of this X-ray examination, it was confirmed on the one hand, that not only the diffraction pattern reproduced in Fig. 6, Plate II, but all the patterns obtained with *e* Portion of the specimens of pure aluminium, were also of a set of comparatively continuous Debye-Hull rings superimposed by an irregular assemblage of many radiating bands; while, on the other hand, the patterns composed of a number of radiating bands, melting down in the rather regular distribution as shown in Fig. 7, Plate II, were usually obtained with *f* Portion of these specimens.

The arguments which have hitherto been advanced dealing with the variation of the diffraction patterns, led us to conclude that the micro-crystals in the specimens of pure aluminium tend, on account of the external tension, to break into small pieces (10^{-4} cm. in diameter) and then to rearrange themselves in a fibrous way having one of their

crystallographic direction parallel to a definite common direction. This is in no way different from the experimental results previously given by Tanaka[1], and Clark and Beckwith.[2] But it is noticed that a few exceptions to the process of the change of crystalline configuration above stated, were occasionaly observed in the present experiment: e. g., by comparing Fig. 8, Plate II, taken with *f* Portion of Specimen A1, with those corresponding to *e* Portion of the same specimen (not reproduced as they were essentially the same as Fig. 6, Plate II), the equalization of the intensity of Debye-Hull rings being observed instead of the melting down of radiating bands, the further destruction of the micro-crystals can be conceived to occur in place of their rearrangement as above stated.

Here, it must be remarked again, that the specimens of pure aluminium were confirmed to be of the smallest tensile strength, despite their largest elongation percentage, among the specimens used in the present experiment.

(ii) *Cu-Al Alloys*:—Essentially the same experiment as in the case of pure aluminium, were carried out with the specimens of Cu-Al alloys containing 4% of Cu. Some of the diffraction patterns thus obtained, utilizing Specimens AC1, AC2, AC3, were reproduced in Figs. 9~19, Plate II. By the way, Fig. 20, Plate II and Figs. 21~23, Plate III represent the patterns taken with Specimen ACR and ACQ,[3] from which the specimens above mentioned were prepared by the mechanical and heat-treatments tabulated in the fourth column of Table I.

As can be seen from Figs. 9, 12 and 17, Plate II together with Figs. 22~23, Plate III, all the diffraction patterns given rise to by the specimens of the Cu-Al alloy immediately before the infliction of the tensile stress, were observed to be essentially the same as those taken with the raw material of these specimens as above stated: i. e., Figs. 9, 12 and 17, Plate II consist mainly of a set of the discontinuous Debye-Hull rings as in Figs. 22~23, Plate III. Nevertheless a few Laue spots of considerable size were sometimes intermingled with these rings when the tempering temperature was high.[4] This shows us not

1. K. Tanaka. J. Inst. Metals, Japan, **2**, 562 (1938).
2. G. Clark and M. Beckwith. ibid.
3. To obtain the diffraction patterns with Specimen ACQ, the reduction percentage of the specimen due to cold-rolling was not necessarily confined to 93% as shown in the third column of Table I, but varied from 50% to 93%.
4. As may be seen from Figs. 22 and 23, Plate III, the difference*

only that the procedure of tempering tabulated in the fourth column of Table I, does not remarkably affect the crystalline configuration of the Cu-Al alloys, but also that all the specimens of the Cu-Al alloys excepting Specimen AC3 used in the present experiment are mostly, if not all, made up of polycrystalline aggregates similar to the untreated specimens of pure aluminium previously examined.

Some diffraction patterns given rise to by the various portions of Specimens AC1, AC2 and AC3 lying between *e* Portion and the opposite end of the test piece, are reproduced in Figs. 10, 13, 14 and 18, Plate II. It can be conceived from these patterns that each one of the intense spots forming the aforesaid Debye-Hull rings would tend to reduce its area the more markedly the more the specimens were elongated by the external tension: At last, when the effect of this elongation increased so largely as in *e* Portion of the fractured test piece of the Cu-Al alloys, the discontinuity of these rings becomes almost indetectable.

After this equalisation of the intensity of Debye-Hull rings was completed, the further elongation of the test piece due to external tension was confirmed gradually to change the rings mentioned above into an irregular assemblage of many radiating bands, as may be seen in Fig. 15, Plate II. Finally, these radiating bands were found to melt down in a regular distribution as was observed in Figs. 11, 16 and 19, Plate II, which was taken with *f* Portion of Specimen AC1 AC2 and AC3.

By a calculation, these diffraction patterns, Figs. 11, 16 and 19, were supposed to be given rise to by the aggregation of micro-crystals which arranged themselves in a fibrous way, each having the normal to one of its atomic planes (111) parallel to the direction of the tensile stress. The above supposition, deduced from the diffraction phenomena due to the normal incidence of the X-ray beam to the direction of the tensile stress in the specimens, was also found to be in good accordance with the diffraction patterns produced by the oblique incidence of the beam to the aforesaid direction: e. g., Fig. 24, Plate III, which was taken when the direction of the tensile stress in Specimen AC1 was tilted 45° from its vertical position towards the incident beam,

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* in the procedures of preparing these specimens was found not appreciably to alter their crystalline configuration; the slight difference between the diffraction patterns in Figs. 9, 12 and 17, Plate II, seems to be due to the difference in the temperature of tempering the specimens.

agrees well with the diffraction patterns expected to appear in the present configuration. The arguments which have hitherto been advanced with regard to the polycrystalline specimens of Cu-Al alloys containing 4% of Cu, led us to conclude that the crystalline configuration of these specimens (excepting those of the extremely small elongation percentage as Specimen ACR), which were noted to be of considerable tensile strength and elongation percentage as shown in Table I, always takes the most regular process of the change due to the external tension, in spite of the variation of the temper-hardening procedures previously given to them: i. e., the micro-crystals in the specimens of the primary solid solution of Cu-Al alloys were confirmed to form a number of fibrous arrangements after their fragmentation, and then to rearrange themselves in a fibrous-like manner each having the normal to one of the (111) planes of the face-centred cubic lattice parallel to the direction of the tensile stress.

Here, it must be noticed that the above statement is deduced from the specimens subjected to the "standard fracture". The breaking process was observed to change its course even with the specimens belonging to this category, when the fracture took place outside the range of $\frac{1}{4}$ of the gauge length from their middle point. In such a case, it was confirmed that the diffraction patterns given rise to by *f* Portion of each specimen indicate the fibrous nature not so remerkably as these corresponding to *e* Portion of the same specimen, but to be essentially the same as Fig. 25, which was produced from the fractured portion of Specimen AC1 broken by bending the test piece around the direction perpendicular to its axis. Thus, we may conclude that the non-standard fracture of the specimens under examination, was due not only to the longitudinal tension, but also to the bending force acting transversely to the test piece.

(iii) *Duralumin-type Alloys*:—By repeating the similar X-ray examination with regard to the duralumin-type alloys, it was found that all the specimens of these alloys not subjected to tensile stress give rise to a diffraction pattern consisting of a set of discontinuous Debye-Hull rings, as shown in Fig. 26, Plate III. This shows us that these specimens are nothing but an irregular aggregation of micro-crystals of the considerable dimension of about 10^{-2} cm. in diameter, as was the case with pure aluminium and some Cu-Al alloys containing 4% of Cu. Furthermore, as can be seen in Figs. 27 and 28, Plate III

taken with *e* and *f* Portions of Specimen D8, the further fragmentation of each micro-crystal in the specimens, succeeded by the formation of a fibrous structure due to the tension, was also found to take place, though not so conspicuously as with the specimens of duralumin-type alloys examined, excepting Specimens D9(2) and D14(2) prepared by an exceedingly different process (see Fig. 29, Plate III). Thus, duralumin-type alloys may be said to partake usually the normal course of breaking process due to external tension, just as previously found with Cu-Al alloys containing 4% of Cu. Here, it must be postscribed that each micro-crystal forming the fibrous structure in *f* Portion of the test piece of duralumin-type alloys, was confirmed not necessarily to arrange itself having the normal to its (111) plane parallel to a definite common direction, which coincides with the direction of the tensile stress; but, this common axis of the micro-crystals was sometimes observed to take the direction normal to their (110) planes.

(iv) "*ESD*" *Alloy*:—Next, the writers extended their X-ray examination to the specimen of the so-called "ESD" alloy, which was remarked to be of especially large tensile strength and of small elongation percentage. All the diffraction patterns obtained with various portions of this specimen, consist mainly of several intense Debye-Hull rings interspersed by an irregular assemblage of many diffuse radiating bands, as shown in Figs. 30 and 32, Plate III. The aforesaid radiating bands in these patterns were observed to become the less appreciable the more the specimen was elongated. Furthermore, the difference in the crystalline configurations between the distructed portions corresponding to the "bending" and "standard fracture" previously observed, could not remarkably be detected with the specimen of "ESD" alloy (see Fig. 31). Thus, we may conclude that the light alloys of small elongation percentage such as "ESD" alloy, would take a course of breaking process entirely different to that previously observed, even the infliction of tensile stress gives rise to the normal fracture.

Crystal Structure

It is generally known that when various specimens of metals and alloys have undergone tensile stress so as to be deformed, the spectral lines of X-rays consequent upon diffraction, would usually be displaced or be widened owing mainly to the strain of crystal lattice, though this effect of strain may sometimes be accompanied by other effects such as those connected with the shattering of micro-crystals. To

confirm whether such a change in X-ray spectra may also be observable or not with the specimens used in the present experiment, the writers tried to examine the spectral lines produced from various portions of these specimens. In this examination, the photographs were taken, adopting the Debye-Scherrer method, by the aid of a cylindrical camera of 7.03 cm. in diameter.

Some of the spectra thus obtained utilizing K_α radiation of copper, are reproduced in Fig. 33, Plate III. As can be seen in these figures, it was confirmed that all the spectral lines appear just at the positions consistent with Vegard's additive law. But, the displacement of each line together with its widening concomitant to the deformation of the specimen, was hardly detected. Consequently, the specimens of pure aluminium and its alloys examined, may be said to retain their proper crystal lattice almost (if not entirely) undistorted even by tensile stress large enough to elongate them to be fractured. Moreover, it can be seen from the absence of widening of each spectral line that the further fragmentation of micro-crystals in these specimens due to the external tension, never goes over the dimension of the order 10^{-4} cm. in diameter.

Conclusion

The arguments which have hitherto been advanced with regard to the various specimens of pure aluminium and its alloys, led us to the following conclusions :—

(1) The inner structure of specimens of considerable tensile strength and elongation percentage such as those of some Cu-Al alloys containing 4% of Cu, take the " normal process " of change due to external tension, as has already been observed by other investigators with various specimens. In the " normal process ", the micro-crystals of the specimens break and then rearrange themselves in a fibrous-like manner, each having a certain crystallographic direction parallel to the direction of the tensile stress ; this common crystallographic direction of the micro-crystals is confirmed in the present case as coinciding usually with the normal to one of the (111) planes of their face-centred cubic lattice.

(2) When either the tensile strength or the elongation percentage of the specimens is small, as in the case of pure aluminium and the so-called " ESD " alloy, the aforesaid " normal process " of change of the internal structure due to external tension does not entirely occur but only partly.

(3) The crystalline configuration at the broken end of each specimen, on which has already been inflicted a "non-standard" fracture, differs from that corresponding to its "standard fracture". On the contrary, the aforesaid portion of the specimen is seen to be a crystalline configuration rather similar to that of the broken end of the same specimen, produced by bending the test piece around the direction perpendicular to its axis.

(4) The further fragmentation of micro-crystals less than of the order 10^{-4} cm. in diameter, does not take place due to the external tension in the specimens examined, before these specimens are fractured.

In conclusion, the writers wish to express their sincere thanks to Ass. Prof. Hideki Hirata for his guidance and to Prof. Denzo Uno and Ass. Prof. Wakae Nakayama for their efficacious suggestions and the interest they have taken during the progress of this investigation. Thanks are also due to Dr. Tomojirô Tanabe who kindly supplied the specimen of "ESD" alloy used in the present experiment.

Institute of Metallography,
Kyoto Imperial University.

Plate I

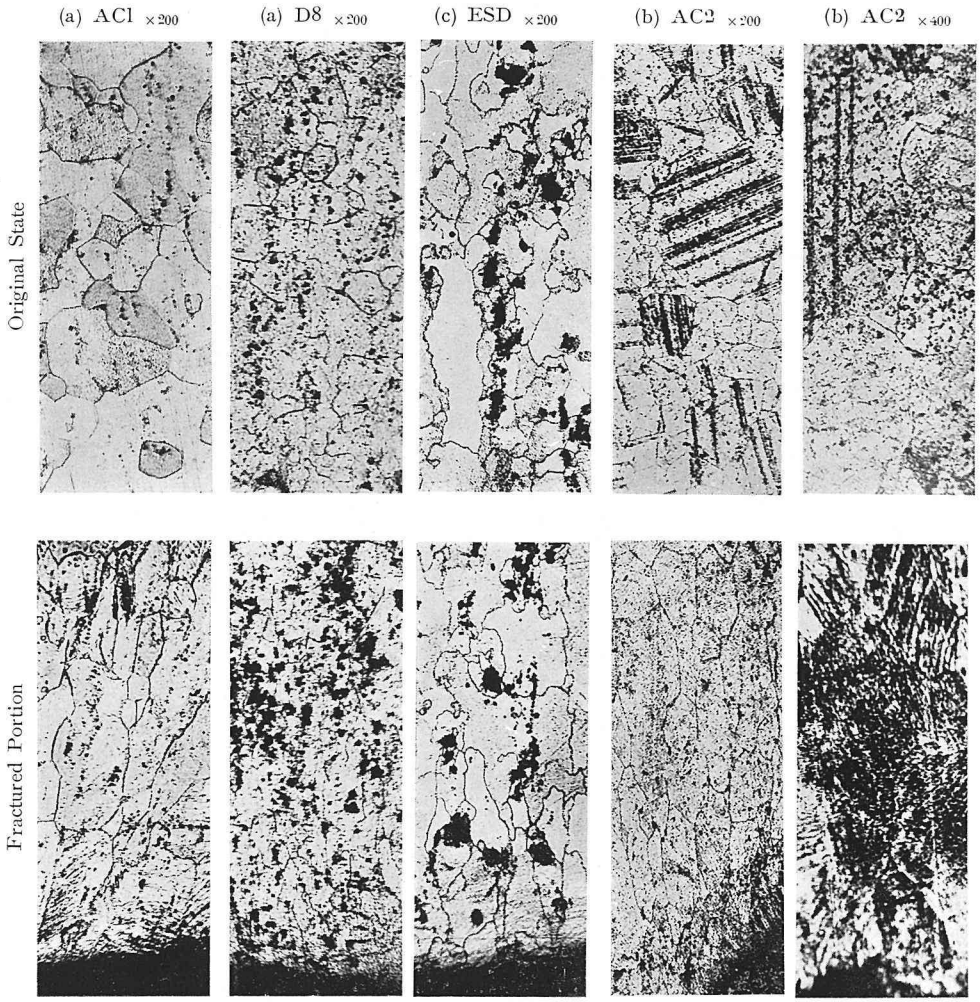
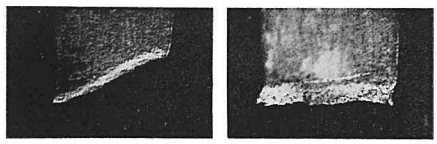


Fig. 2 Microscopic Structures of the Specimens

Fig. 3 Microscopic Structures of the Specimens



(a) AC2 Fig. 4 (b) ESD

Plate II

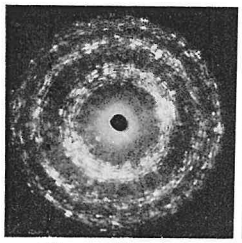


Fig. 5 Specimen A2 (Original State)

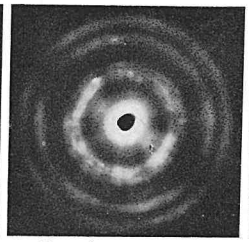


Fig. 6 Specimen A2 (*e* Portion)

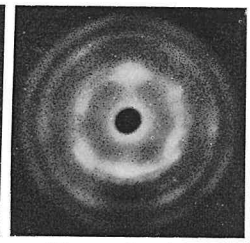


Fig. 7 Specimen A2 (*f* Portion)

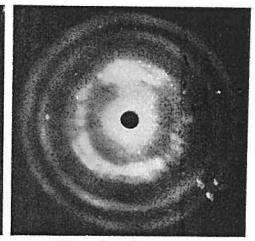


Fig. 8 Specimen A1 (*f* Portion)

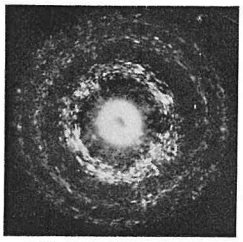


Fig. 9 Specimen AC1 (Original State)

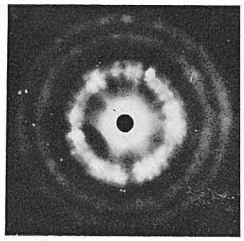


Fig. 10 Specimen AC1 (*e* Portion)

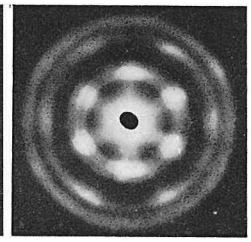


Fig. 11 Specimen AC1 (*f* Portion)

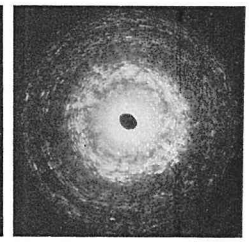


Fig. 12 Specimen AC2 (Original State)

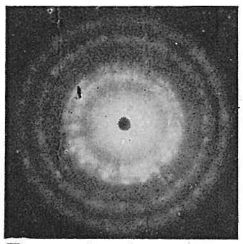


Fig. 13 Specimen AC2 (Intermediate between the End and *e* Portion)

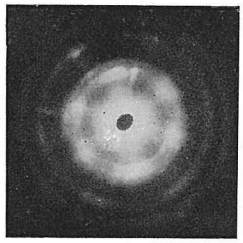


Fig. 14 Specimen AC2 (*e* Portion)

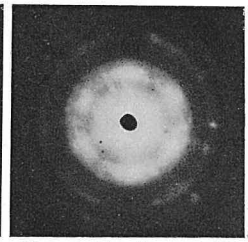


Fig. 15 Specimen AC2 (Intermediate between *e* and *f* Portions)

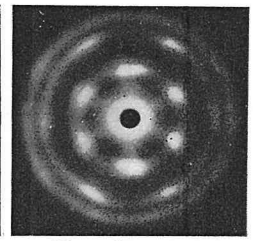


Fig. 16 Specimen AC2 (*f* Portion)

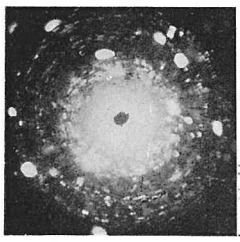


Fig. 17 Specimen AC3 (Original State)

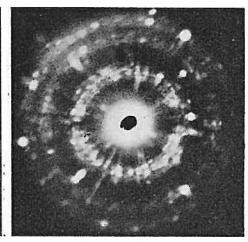


Fig. 18 Specimen AC3 (*e* Portion)

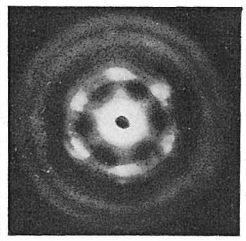


Fig. 19 Specimen AC3 (*f* Portions)

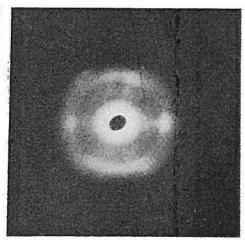


Fig. 20 Specimen ACR (Original State)

Plate III

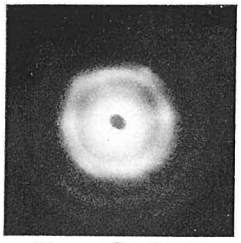


Fig. 21 Specimen ACR (*f* Portion)

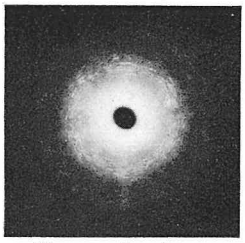


Fig. 22 Specimen ACQ (Original State)

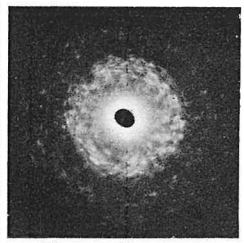


Fig. 23 Specimen 4%Cu-Al quenched (50% Cold-rolled)

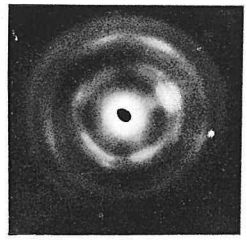


Fig. 24 Specimen ACl (*f* Portion, $\alpha=45°$)

Fig. 25 Specimen ACl (Portion fractured by bending)

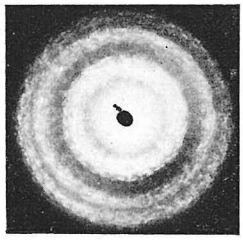


Fig. 26 Specimen D8 (Original State)

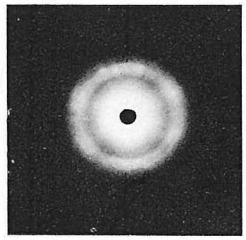


Fig. 27 Specimen D8 (*e* Portion)

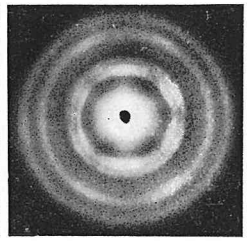


Fig. 28 Specimen D8 (*f* Portion)

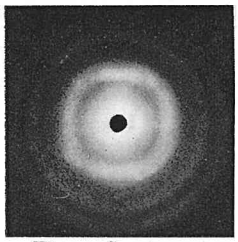


Fig 29 Specimen D9 (2) (*f* Portion)

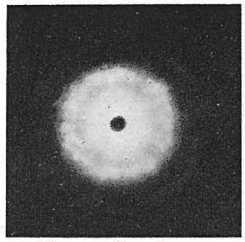


Fig. 30 Specimen ESD (Original State)

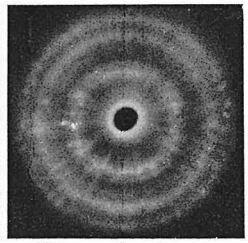


Fig. 31 Specimen ESD (Portion fractured by bending)

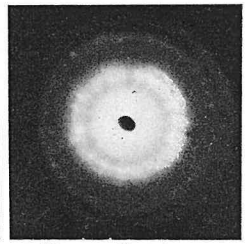


Fig. 32 Specimen ESD (*f* Portion, $\alpha=0°$)

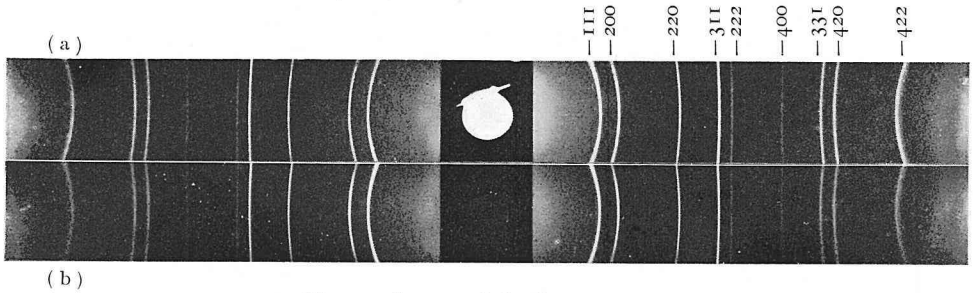


Fig. 33 Spectra of the Specimen ACl (a: Original State; b: *f* Portion)