Morphological and Crystallographic Characteristics of Incoherent Octahedral FCC Co Precipitates in a Cu Matrix

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A Cu-3.0 mass%Co alloy was heat-treated by a method named “short-cut aging”. Incoherent fcc Co precipitates of about 200 nm in size formed in an fcc Cu matrix were observed with conventional and ultra-high-voltage transmission electron microscopes. Although the Co precipitates had shapes similar to a regular octahedron composed of (111) of the Cu matrix, small deviations from the regular octahedron were found. Most Co/Cu interfaces deviated from {111} of Cu and the average deviation angle δ was about 7 deg. Dislocations in the Cu matrix intersecting the Co precipitates were observed for those having deviated interfaces. Deviations of the orientation relationship from an exact cube-on-cube relationship were also found. Although the deviation angle θ of the orientation relationship was much smaller than δ, there was a linear relationship between θ and δ. The morphological and crystallographic characteristics of the octahedral Co precipitates are discussed by considering dislocations in the Cu matrix intersecting the Co precipitates. [doi:10.2320/matertrans.M2011335]

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1. Introduction

From a scientific point of view, Cu–Co alloys are known as model alloys to study precipitation of metal particles in a metal matrix.1–21 Although Co itself has an hcp structure at ambient temperature and under normal pressure,23) Co precipitates in a Cu matrix have an fcc structure and the orientation relationship between them is reported to be of a cube-on-cube type.2–4,6,8–20) The lattice constant of fcc Co is a_{Co} = 0.3548 nm,23) which is 1.9% smaller than that for Cu, a_{Cu} = 0.3615 nm.24)

At the initial stage of aging, spherical Co precipitates with a size equivalent to a diameter of about 10 nm and coherent interfaces between the fcc Co precipitates and the Cu matrix have been observed.2–4,9–13,15,17–21) Keeping the coherent Co/Cu interfaces, a shape change from a sphere to a cube occurs when the precipitates become about 40 nm.9–13,15,17–20) However, when the Co precipitates become larger than about 100 nm, the shape of the Co precipitates changes to that similar to a regular octahedron composed of {111} of the Cu matrix.4,6,12,13,15,17–20) When this shape change occurs, the state of the interfaces between the fcc Co precipitates and the Cu matrix changes from coherent to incoherent. The origin of the [111] octahedral shape has been explained by the lowest energy density of the incoherent [111] Co/Cu interface.8,12,18–20,25) However, some deviations of the shape from the regular [111] octahedron seem to exist for the Co precipitates observed in previous studies.12,15)

The equilibrium shapes of small particles controlled by the particle/matrix interface energy are known in many alloy systems. However, as far as we know, the deviations of the incoherent fcc Co precipitate shape from the expected equilibrium-shape have not been investigated. In the present study, the morphological and crystallographic characteristics of the incoherent Co precipitates are studied.

2. Experimental Procedure

A Cu–3 mass%Co alloy was prepared from 99.99% Cu and 99.99% Co. The alloy was cold rolled to a plate with the thickness of 2 mm. Using the specimens cut from the plate, a special heat-treatment named “short-cut aging” was applied in the present study. In usual precipitation treatment to obtain precipitates in a solid metal matrix, quenching is carried out after a solid-solution treatment and before an aging treatment as shown in broken lines in Fig. 1. However, in the short-cut aging, the aging treatment after the solid-solution treatment without quenching is made as shown by solid line in Fig. 1. The short-cut aging was applied in the present study in order to efficiently obtain larger fcc Co precipitates uniformly dispersed in the Cu matrix with shorter aging-time. The conditions of the present short-cut aging were (i) the solid-solution treatment at 1323 K for 3 h, (ii) the furnace cooling from 1323 to 1023 K for 1 h, and (iii) the aging treatment at 1023 K for 3 h. After the short-cut aging, the Co precipitates

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with a size of about 200 nm were obtained in the Cu matrix. In contrast, after the usual precipitation treatment with the conditions of solution-treatment at 1323 K for 3 h, quenching into water and aging at 1023 K for 3 h, the average size of the incoherent Co precipitates was about 50 nm. In the case of the usual precipitation treatment aged at 1023 K, the aging time much longer than 3 h is needed to obtain the Co precipitates with the size of 200 nm.

Thin foils for transmission electron microscopy (TEM) were prepared from the aged specimens by mechanical and electrolytic polishing. Observations were carried out with a conventional TEM, JEOL JEM-2011 and an ultra-high-voltage transmission electron microscope (HVEM), HITACHI H-1250, operated at 200 and 1000 kV, respectively. The Co precipitates were observed from the three directions, [001], [011] and [111] of Cu. The Burgers vectors of dislocations in the Cu–Co alloy were determined by finding the reflection in which the dislocation becomes invisible. 26,27 We also observed the moiré fringes formed between the Co precipitates and the Cu matrix. To exclude surface effects of thin foils, stereoscopic observations were made to select Co precipitates embedded well into the foils. 28,29 The HVEM makes it possible to observe the Co precipitates in the foils with a thickness of about 1 µm. Figure 2 is an example of the result of the stereoscopic observation showing locations of fcc Co precipitates in a cross-section of Cu foil for ultra-high-voltage transmission electron microscope (HVEM) observation. The Co precipitates shown by closed circles embedded well into the thick foil were used to investigate the shapes.

3. Results

3.1 Shape of Co precipitates

Figure 3 shows a bright-field (BF) image of the fcc Co precipitates in Cu taken from the direction near [011]. The selected area diffraction pattern (SADP) of this image is also shown. The dark rhombuses in Fig. 3 are the [011] projection of the Co precipitates. It has been reported that the Co precipitates have an fcc structure and a cube-on-cube orientation relationship with Cu. 2,4,6,9,23 Dislocations in the Cu matrix intersecting the Co precipitates are also seen in Fig. 3.

A three-directional observation of the Co precipitates was carried out. Figure 4 shows the variation of the projected images of the Co precipitates when observed along near (a) [001], (b) [011] or (c) [111]. The mark “A” is added in Figs. 4(a) to 4(c) to show the identical precipitate observed from the different directions. When we observe the [111] regular octahedron from the three directions [001], [011] and [111], the corresponding projected shapes are square, rhombus, and hexagon, as shown in Figs. 4(d) to 4(f), respectively. The images of the Co precipitates in Figs. 4(a) to 4(c) are similar to these geometrical shapes. This means that the Co precipitates have a three-dimensional shape similar to the [111] regular octahedron. However, close inspection of Figs. 3 and 4 show that the projected shapes of Co precipitates vary from precipitates to precipitates and are not completely the same as those shown in Figs. 4(d) to 4(f).

Figure 5(a) shows the BF image of a Co precipitate observed from the exact [011] direction. In this on-axial image, we can see superposition of the moiré fringes in the Co precipitate formed by various reflections. A rhombus drawn by solid lines in Fig. 5(b) shows the Co/Cu interface of the Co precipitate shown in Fig. 5(a). On the other hand, another rhombus drawn by broken lines in Fig. 5(b) is a projection of the [111] regular octahedron viewed from [011]. We can certainly find the deviation of the Co precipitate shape from the on-axial image. The deviation of the Co precipitate shape can be discussed with the BF images taken under two-beam conditions, as far as the tilt angles of specimens from the exact zone-axes are small. As will be shown later, moiré fringes observed under the two-beam conditions are utilized to discuss crystallographic characteristics of the Co precipitates in the Cu matrix.

We evaluated the degree of deviation of the Co precipitate shape from the ideal {111} regular octahedron using the images of the Co precipitates taken from the direction near [011]. Figure 6(a) is a close-up view of a Co precipitate when the observation direction is near [011]. A rhombus drawn by broken lines in Fig. 6(b) is a projection of the [111] octahedron viewed from [011]. The four sides of this rhombus are traces of four {111} planes of the {111}
octahedron, which are perpendicular to the sheet. On the other hand, another rhombus drawn by solid lines in Fig. 6(b) shows the Co/Cu interface of the Co precipitate shown in Fig. 6(a). The deviation of the shapes of Co precipitates from the \{111\} octahedron can be shown by the angle \( \delta \) in Fig. 6(b). The variation of \( \delta \) is shown in Fig. 6(c), where the vertical axis is the number of the observed Co/Cu interfaces.

### 3.2 Dislocations in Cu matrix

As shown in Figs. 3, 4(a)–4(c), 5(a) and 6(a), dislocations in the Cu matrix intersecting the Co precipitates were observed. The Burgers vectors of them were determined by the contrast experiments finding the reflection in which the dislocation becomes invisible.26,27) Figure 7 is an example of the contrast experiment showing a Co precipitate and the dislocations taken from the direction near [011] with various reflection vectors (g vectors). Table 1 shows the results of the contrast experiment for the dislocations a1 and a2 in Fig. 7. Although the Burgers vectors of the dislocations a1 and a2 were those for perfect dislocations in common, that of a1 at the left side of the Co precipitate was not the same as that of a2 at the right side. Dislocations on the interface had various Burgers vectors.

### 3.3 Parallel moiré fringes

As shown in Figs. 3, 5(a), 6(a) and 7, moiré fringes were observed in the Co precipitates. When the Co precipitates keep exactly the cube-on-cube relationship with the Cu matrix and the lattice planes of Co and Cu are parallel, the
moiré pattern should be of the parallel type caused by the difference between the lattice constants of Co and Cu. Figure 8 shows an example of the parallel moiré fringes in a Co precipitate formed by {200} of Co and Cu with $g = 200$. A characteristic of the parallel moiré fringes is that the direction normal to the moiré fringes is parallel to $g$. When {200} of Co and Cu form the parallel moiré fringes, the spacing of the parallel moiré fringes $D_{\text{pm}}$ is given by

$$D_{\text{pm}} = d_{\text{Cu200}}/d_{\text{Co200}},$$

where $d_{\text{Co200}}$ and $d_{\text{Cu200}}$ are the {200} lattice plane spacings in the Co precipitate and Cu matrix, respectively. Assuming that $d_{\text{Co200}}$ and $d_{\text{Cu200}}$ are simply given from $a_{\text{Co}}$ and $a_{\text{Cu}}$ without taking into account the effects of elastic deformation, we obtain $D_{\text{pm}} = 9.4 \text{ nm}$. On the other hand, the measured spacing of the parallel moiré fringes $D_{\text{exp}}$ observed in the Co precipitate in Fig. 8 was 9.6 nm. The measured value $D_{\text{exp}}$ is very close to the calculated value $D_{\text{pm}}$. This again indicates that the Co precipitates are incoherent with the Cu matrix and elastic distortion, if any, in the precipitates is very small.

### 3.4 Rotational moiré fringes

In addition to the parallel moiré fringes, moiré fringes including both parallel and rotational contributions were also observed in the Co precipitates. The rotational contribution is caused by relative rotation of the lattice planes between Co and Cu. When the relative rotation angle $\theta$ of the lattice planes between Co and Cu is much less than unity, the effect of the rotational contribution on the spacing of moiré fringes is negligibly small. However, the rotational contribution significantly changes the direction of the moiré fringes. Figure 9 shows an example of the rotational moiré fringes in a Co precipitate observed with $g = 111$. It is noted that the direction perpendicular to the moiré fringes is deviated from the direction of $g$. When this deviation angle is $\varphi$, the relative rotation angle $\theta$ of {111} of Co and Cu is given by

$$\cos \theta = \left( d_{\text{Co111}}/d_{\text{Cu111}} \right) \sin^2 \varphi + \cos \varphi \sqrt{1 - \left( d_{\text{Co111}}/d_{\text{Cu111}} \right)^2 \sin^2 \varphi},$$

where $d_{\text{Co111}}$ and $d_{\text{Cu111}}$ are the {111} lattice plane spacings in the Co precipitate and Cu matrix, respectively. The angle $\varphi$ shown in Fig. 9 is about $12.5^\circ$. Using the values of $\varphi = 12.5^\circ$, $d_{\text{Co111}}$, and $d_{\text{Cu111}}$ given from the above values of $a_{\text{Co}}$ and $a_{\text{Cu}}$, we obtain $\theta = 0.24^\circ$. The angle $\theta$ is understood to be the deviation of the orientation relationship between the Co precipitate and Cu matrix from the cube-on-cube relationship.
The deviation angles $\theta$ and $\delta$ were measured for other identical Co precipitates. Figure 10 shows the relationship between $\theta$ and $\delta$. Although the values of $\theta$ are generally much smaller than those of $\delta$, a linear relationship is found between them. This result will be discussed later.

### 3.5 Disturbance of moiré fringes

Figure 11 is a set of BF images of an identical Co precipitate taken from the direction near [011] with various $g$. In Fig. 11(a), white or black vertical lines disturbing the moiré fringes are observed in the Co precipitate. The same disturbances of the moiré fringes appear in Figs. 11(b) and 11(c) taken with $g$ different from that in Fig. 11(a). The strain field of dislocations on the Co/Cu interface causes the disturbances of the moiré fringes. The lines added in Fig. 11(d) indicate the extra dislocations on the interface determined from Figs. 11(a) to 11(c). The spacings of the dislocations on the Co/Cu interface were about 30–50 nm.

### 4. Discussion

#### 4.1 Dislocations on the Co/Cu interface

As described in 3.3, the moiré fringes in Fig. 8 show that the Co precipitate in this figure is incoherent with the Cu
matrix. The incoherent state should be achieved by the generation of misfit dislocations on the Co/Cu interfaces. However, different from micrographs shown in Figs. 9 and 11, very few dislocations are observed on and around the Co precipitates in Fig. 8. Here we discuss a reason why the misfit dislocations are not seen in Fig. 8.

The spacing of the misfit dislocations on the Co/Cu interfaces necessary for the incoherent state is evaluated from the difference between \(a_{\text{Co}}\) and \(a_{\text{Cu}}\). The spacing \(l_{200}\) of the misfit dislocations along (200) is written using \(d_{\text{Cu}200}\) and \(d_{\text{Co}200}\) as

\[
l_{200} = nd_{\text{Cu}200} = (n + 1)d_{\text{Co}200},
\]

where \(n\) is a positive integer satisfying

\[
 n \approx d_{\text{Cu}200}/(d_{\text{Cu}200} - d_{\text{Co}200}).
\]

From eqs. (3) and (4), we have

\[
l_{200} \approx d_{\text{Cu}200}d_{\text{Co}200}/(d_{\text{Cu}200} - d_{\text{Co}200}) \approx 9.4 \text{ nm}.
\]

It is noted that the spacing \(l_{200}\) of the misfit dislocations given by eq. (5) is identical to the spacing of the parallel moiré fringes \(D_{\text{pm}}\) given by eq. (1). Because of the moiré fringes that have the same spacing as that of the misfit dislocations, images of the misfit dislocations themselves are not observed in BF images of TEM. A symmetrical dislocation-network consisting of edge dislocations have been considered as a reasonable configuration of the misfit dislocations on the coherent Co/Cu interfaces.\(^6\)

In Fig. 11, the dislocations on the Co/Cu interfaces with the spacings of about 30–50 nm are shown. These are not the misfit dislocations to accommodate the difference between \(a_{\text{Co}}\) and \(a_{\text{Cu}}\) but are the extra dislocations related to dislocations in the Cu matrix. The dislocations disturbing the moiré fringes were observed when the dislocations in the Cu matrix are intersecting the Co precipitates. Although quantitative measurements of densities of the dislocations in the matrix and the extra dislocations on the interface were not made in the present study, the disturbance of the moiré fringes was less when the density of the matrix dislocations intersecting the Co precipitates were less. When glide motion of dislocations in materials containing precipitates is possible at high temperatures, dislocations connecting precipitates are often found in the materials. Once such dislocation structure is formed, departure of the dislocations from the precipitates is hard to occur. We think the extra dislocations on the interface are portions of the dislocations connecting the precipitates.

### 4.2 Deviations of the Co precipitate shapes from the \([111]\) regular octahedron

Although the Co precipitates observed in the present study had shapes similar to a regular octahedron composed of \([111]\) of the Cu matrix, deviations of the Co precipitate shape from the regular octahedron were found. We have evaluated two kinds of the deviations. One is the deviation of the orientation of the Co/Cu interfaces from \([111]\) of Cu and the other is the deviation of the orientation relationship between the Co precipitates and the Cu matrix from the cube-on-cube relationship. Here, we consider the origins of the deviations by using the operations of the extra dislocations observed on the Co/Cu interfaces.

#### 4.2.1 Deviation from cube-on-cube relationship

Firstly, we discuss the deviation of the orientation relationship between the Co precipitates and the Cu matrix from the cube-on-cube relationship. As far as the symmetrical network of misfit dislocations surrounds the Co precipitates,\(^5\) the cube-on-cube relationship is maintained for the incoherent Co precipitates. However, the extra dislocations observed on the Co/Cu interfaces change the situation. In the following, we make the two-dimensional analysis to discuss the operations of the extra dislocations considering a Co/Cu interface of an octahedral Co precipitate. This is a simplified analysis to treat the deviation of the three-dimensional Co precipitates.

Figure 12 shows the schematic illustration of the \([011]\) view of the incoherent Co octahedral precipitates. Figure 12(a) shows the Co precipitate with the misfit dislocations and without the extra dislocations, which corresponds to that in Fig. 8. On the other hand, Fig. 12(b) shows the Co precipitate with both of the misfit and extra dislocations, which corresponds to that in Fig. 11. Here, we consider a case that the extra dislocations on the interface are dislocations with strong edge components. The close-up view of the interface with the extra dislocations is also shown in Fig. 12(c), where these were edge dislocations on \([111]\) of Cu. The magnitude of the Burgers vector of Cu is \(b = 0.256 \text{ nm}\)\(^{24}\) and the spacing between the extra dislocations on the Co/Cu interface is defined as \(l_e\).

The extra dislocations cause the deviation from the cube-on-cube relationship. Applying the simple low-angle tilt-boundary model to the dislocations array on the Co/Cu interface shown in Fig. 12(c), the angle \(\theta\) between \((111)\) of Co and \((111)\) of Cu is given by\(^{22}\)
where $l_e \gg b$ is assumed. As shown in Fig. 10, the magnitude of the value of $l_e$ observed in the present study was $l_e \approx 50$ nm. The angle $\theta$ given by this value of $l_e$ is $\theta \approx 0.3^\circ$. This value of $\theta$ is the same magnitude as the experimentally observed angles shown as the horizontal axis in Fig. 10. The deviation from the cube-on-cube relationship is reasonably explained by the array of extra dislocations on the Co/Cu interface.

The extra dislocations on the interfaces are considered to be parts of the dislocations in the Cu matrix intersecting the Co precipitates. Since various matrix dislocations with different Burgers vectors may interact with the Co precipitates, the above results $\theta \approx 0.3^\circ$ should be the upper bound of the deviation angle. Most of the experimental results of $\theta$ were actually less than $0.3^\circ$. It is interesting to note that, if we think the dislocations shown in Fig. 12(b) are shear dislocation loops surrounding a plastically non-deformable precipitate, we also reach a relationship similar to eq. (6).

This relationship can be obtained from the results shown in the paper by Sato et al. where they have observed and analyzed the rotation of $\alpha$-Fe particles in the Cu matrix caused by plastic deformation of the Cu-Fe dispersion hardened alloy.\(^{28,31}\)

### 4.2.2 Deviation of Co/Cu interfaces from (111) of Cu

Next, we discuss the deviation of the orientation of the Co/Cu interfaces from (111) of Cu. As well as the case treated in the preceding section, we consider the extra dislocations when these are dislocations with strong edge components. Figures 13(a) to 13(c) are the schematic illustrations to consider this deviation. Figure 13(a) shows the incoherent Co/Cu interfaces without the extra dislocations. The stable orientation of the incoherent interface is (111) and $\mathbf{n}_m$ is the normal unit vector of the interface. Figure 13(b) shows the extra dislocations approaching to the interface, where they are edge dislocations on (111) with $b = 1/2[110]$. The original interface, (111), may not be stable for the array of the extra dislocations. For example, the array on (111) shown by a broken line may be stable for the edge dislocations. $\mathbf{n}_e$ on the [011]-[100] plane in Fig. 13(b) is the unit vector perpendicular to [111]. The generation of the extra dislocations on the interface should change the stable orientation of the interface as shown by a thick solid line in Fig. 13(c). The angle $\delta$ shown in Fig. 13(c) is the deviation angle of the Co/Cu interfaces from (111) of Cu. Effects of the misfit and extra dislocations on the final orientation of the interface probably depend on the densities of the misfit and extra dislocations.

Using the above ideas and assumptions, we perform a semi-quantitative analysis to consider the present experimental results assuming that (110) is the stable interface for the extra dislocations. When the density of the misfit dislocations on a Co/Cu interface is $\rho_m$ and that of the extra dislocations is $\rho_e$, using $\mathbf{n}_m$ and $\mathbf{n}_e$, the normal vector $\mathbf{n}$ of the interface with the extra dislocations may be given by

$$\mathbf{n} = \frac{\rho_m \mathbf{n}_m + \rho_e \mathbf{n}_e}{\rho_m + \rho_e}. \quad (7)$$

Then, we have

$$\cos \delta = \frac{\langle \mathbf{n}, \mathbf{n}_m \rangle}{\|\mathbf{n}\| \|\mathbf{n}_m\|}, \quad (8)$$

where $\langle \mathbf{n}, \mathbf{n}_m \rangle$ is the inner product, $\|\mathbf{n}\|$ and $\|\mathbf{n}_m\|$ the norms of $\mathbf{n}$ and $\mathbf{n}_m$. From eqs. (7) and (8), when $\rho_e/\rho_m \ll 1$, we have

$$\delta \approx (1/3)\rho_e/\rho_m. \quad (9)$$

Derivation of this equation is shown in Appendix. Using the dislocation spacings $l_m$ for the misfit dislocations and $l_e$ for the extra dislocations, the ratio of the dislocation densities is evaluated to be $\rho_e/\rho_m \approx l_m/l_e$. Then, eq. (9) is rewritten as

$$\delta \approx (1/3)l_m/l_e. \quad (10)$$

The deviation angles $\theta$ given by eq. (6) and $\delta$ given by eq. (10) are the results of our present analysis. We can compare these results with the experimental results shown in Fig. 10. From eqs. (6) and (10), we have the relationship between $\theta$ and $\delta$ given by

$$\delta \approx (l_m/3b)\theta. \quad (11)$$

The above relationship reproduces the experimentally obtained linear relationship between $\theta$ and $\delta$. Moreover, since $l_m$ is evaluated to be $l_m \approx 11.5$ nm by eqs. (4) and (5), $l_m/3b$ with $b = 0.256$ nm\(^2\) is $l_m/3b \approx 15. \quad (12)$

The linear relationship with this slope, $l_m/3b \approx 15$, is shown by a line in Fig. 10. Although the present analysis is
simplified two-dimensional consideration, the deviations of the Co precipitate shapes from the \{111\} regular octahedron were reasonably explained.

### 4.3 Equilibrium shape and interfacial structure

As we have discussed, it is reasonable to consider that the extra dislocations on the Co/Cu interface are the reasons of the deviations of the Co precipitate shapes from the \{111\} regular octahedron. However, it is needed to discuss why the deviations of the Co precipitate shapes from the \{111\} regular octahedron were observed even after the aging at 1023 K for 3 h.

During the aging treatment at high temperatures, the precipitates should change to the equilibrium shape by diffusion as far as the free energy monotonically decreases with the shape change under a certain driving force. This is the situation where the equilibrium spherical shape is achieved for the amorphous SiO$_2$ particles in Cu after the internal oxidation of the Cu–Si alloy.\(^{33}\) In contrast to the amorphous structure of the SiO$_2$ particles in Cu, the Co precipitates are crystalline. When the dislocations in Cu matrix intersect at the Co/Cu interface, it is probable that special microstructures such as terraces and ledges are formed on the interface. According to Gabrish,\(^{34}\) the ledges on the interface act as an energy barrier of diffusion. Thus, shape changes to an equilibrium state are retarded or blocked. Although we have not observed details of microstructure of the Co/Cu interface in the present study, this may be a reason why the Co precipitates kept the non-equilibrium shapes in Cu matrix even after the aging at high temperatures.

### 5. Conclusions

A Cu–3.0 mass%Co alloy was heat-treated by a method named “short-cut aging”. Incoherent Co precipitates formed in the Cu matrix were observed with conventional TEM and HVEM. The following conclusions have been obtained.

1. Although the Co precipitates had shapes similar to a regular octahedron composed of \{111\} of the Cu matrix, deviations of the Co precipitate shapes from the regular octahedron were found.
2. Most of the Co/Cu interfaces deviated from \{111\} of Cu and an average of the deviation angle \(\delta\) was about 7 deg.
3. Deviations of the orientation relationship from an exact cube-on-cube relationship were also found. The deviation angle \(\theta\) of the orientation relationship was much smaller than \(\delta\).
4. A linear relationship was found between \(\theta\) and \(\delta\). This relationship was explained by considering operations of dislocations on the Co/Cu interfaces.

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### REFERENCES

2. V. A. Philips and J. D. Livingston: Phil. Mag. 7 (1962) 969–980.
Appendix

Using the $x_1$-$x_2$ coordinate system where $x_1//[011]_{\text{Cu}}$ and $x_2//[100]_{\text{Cu}}$, the components of the unit vectors $\mathbf{n}_m$ and $\mathbf{n}_e$ shown in Figs. 13(a) and 13(b) are written as

$$\mathbf{n}_m = (\cos \alpha, -\sin \alpha) \quad \text{and} \quad \mathbf{n}_e = (\cos \beta, -\sin \beta) \quad (A1)$$

where $\alpha = 35.3^\circ$ and $\beta = 54.7^\circ$. Using eq. (A1), the components of $\mathbf{n}$ given by eq. (7) are written as

$$\mathbf{n} = \left( \frac{\rho_m \cos \alpha + \rho_e \cos \beta}{\rho_m + \rho_e}, \frac{-\rho_m \sin \alpha - \rho_e \sin \beta}{\rho_m + \rho_e} \right) \quad (A2)$$

From eqs. (A1) and (A2), we have

$$(\mathbf{n}, \mathbf{n}_m) = \frac{\rho_m}{\rho_m + \rho_e} \left[ 1 + (\rho_e/\rho_m) \cos(\alpha - \beta) \right] \quad (A3)$$

and

$$\|\mathbf{n}\| = \frac{\rho_m}{\rho_m + \rho_e} \sqrt{1 + 2(\rho_e/\rho_m) \cos(\alpha - \beta) + (\rho_e/\rho_m)^2} \quad (A4)$$

When $\rho_e/\rho_m \ll 1$, using eqs. (A3) and (A4), eq. (8) is rewritten as

$$\cos \delta \approx 1 - \frac{(\rho_e/\rho_m)^2}{2} \left[ 1 - \cos^2(\alpha - \beta) \right]. \quad (A5)$$

Hence $\delta$ satisfying $\delta \ll 1$ is written as a function of $(\rho_e/\rho_m)$ as

$$\delta = \sqrt{1 - \cos^2(\alpha - \beta)(\rho_e/\rho_m)} \approx (1/3)(\rho_e/\rho_m). \quad (A6)$$

The numerical coefficient $(1/3)$ in front of $(\rho_e/\rho_m)$ is the value obtained by the assumption of the (110) stable plane for the extra dislocations shown in Fig. 13(b). Other stable planes may be appropriate for other types of extra dislocations. This affects the value of $\beta$ for $\mathbf{n}_e$ in eq. (A1). However, even if we use other values of $\beta$, we finally have $\delta \approx \rho_e/\rho_m$. In this analysis, it is essential that the deviation angle $\delta$ is roughly proportional to $\rho_e/\rho_m$. 
