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# University of California

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# DISLOCATION ARRANGEMENTS AND STACKING FAULT ENERGY OF α{Cu-Sn} ALLOYS

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# DISLOCATION ARRANGEMENTS AND STACKING FAULT ENERGY OF α(Cu-Sn) ALLOYS

It is now well known that in fcc metals a decrease in stacking fault energy produces a change in the dislocation arrangements from complex tangles to coplanar groupings (1). In copper alloys this change-over occurs where the stacking fault energy is ~5 ergs/cm$^2$ or less, corresponding to e/a ratios of about 1.2 (depending on the valency of the solute).

Recent transmission electron microscopic studies on alpha brasses (2) and stainless steels (3) have shown that the stacking fault energy (SFE), as estimated from node measurements, and consequently, the dislocation distribution, not only varies with increasing solute content but also with the thermal history of the alloy (4). The SFE of quenched alpha brasses falls continuously with increasing % Zn in agreement with earlier experiments (5) but rises for alloys in excess of 25 at % Zn which have been furnace-cooled or quenched and aged. The apparent increase in SFE has been interpreted to be due to short range ordering, since this increase is always accompanied by the appearance of many dislocation pairs (superdislocations) of like sign (6). This phenomenon was also found in certain Fe-Cr-Ni alloys (3).

We have now investigated α Cu-Sn alloys using similar experiments to those done on brass. Three sets of alloys varying in content from 1-10 wt % Sn were annealed at 700°C for two hours and at 600°C for an additional hour. Two sets were water quenched to room temperature while the third was allowed to furnace cool over a twenty-four hour period. All three sets were deformed

5% and 10% in tension, then electropolished just prior to observation in the Siemens Elmiskop electron microscope. The alloys were prepared from high purity copper and tin by Bram Metallurgical Corporation.

In the range 1-6 wt % Sn, tangled dislocations exhibiting distinct cellular structure were observed in both quenched and furnace-cooled specimens. At higher % Sn the cellular structure tended to give way to coplanar dislocation arrays with the transition occurring at a somewhat lower % Sn in the furnace-cooled alloys. As the $\alpha/(\alpha+\beta)$ boundary was approached the coplanar distribution in both quenched and furnaced-cooled alloys consisted of long rows of dislocation pairs and contrast experiments showed these pairs to be dipoles (6). The change in dislocation arrangement with tin content is shown in Fig. 1. No superdislocations were observed. Thus there is no evidence for order in Cu-Sn alloys, confirming x-ray observations by Cohen et.al. (private communication).

Measurements of the cell size and the yield stress are shown in Fig. 2, and as found for the brasses (2) the yield stress increases with alloying corresponding to a decrease in SFE. Yield drops, followed by ~1% extensions at constant stress, were consistently observed for slowly cooled specimens, but rarely in alloys that had been quenched (into water from vacuum) or quenched and aged (1 hr up to 200°C). It should also be noted that an anomaly occurs at 6% Sn. The reasons for these phenomena are not know at this time but from the evidence it is unlikely that ordering is responsible.

The above observations indicate that the stacking fault energy decreases with increasing tin content. Attempts were made to estimate the SFE from observations of dislocation nodes (5). However extended nodes were observed only in the Cu-10% Sn furnaced-cooled alloy. This is not really surprising because the $\alpha$ solid solution is still relatively dilute at the phase boundary when compared to, say, brasses. The average radius of curvature

of extended nodes was 1900Å after allowance was made for projections. The SFE was calculated from the equation,

$$\gamma = \frac{Gb^2}{2R}$$

R = radius of curvature of the extended node

b = Burgers vector of the partial dislocation

G = shear modulus of the alloy

Assuming a modulus of $G = 0.248 \times 10^{12}$ dynes/cm$^2$, the SFE of Cu-10% Sn is calculated to be 1.50 ergs/cm$^2$. To correspond with previous data on copper alloys this result is expressed by the quantity $10^3\gamma/Gb$ and is plotted in Fig. 3. The result is in agreement with the observation that the SFE for group 1B solvents decreases as the valency of the solute element increases (5).

This work was done under the auspices of the United States Atomic Energy Commission.

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## FIGURE CAPTIONS

Fig. 1 (a) Cu-2% Sn furnace cooled from 600°C. Deformed 10% at a strain rate of $2 \times 10^{-3}$. Cell structure.

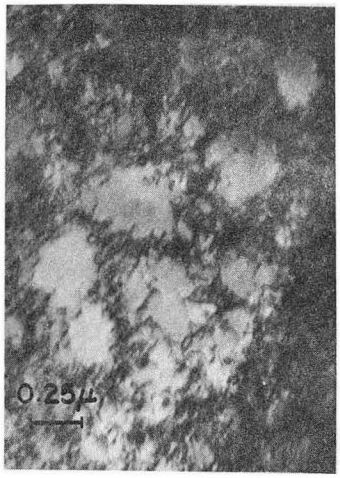
(b) Cu-6% Sn as (a) showing tendency for tangles to lie parallel to slip traces.

(c) Cu-9% Sn quenched from 600°C 5% strain showing coplanar rows of dislocations.

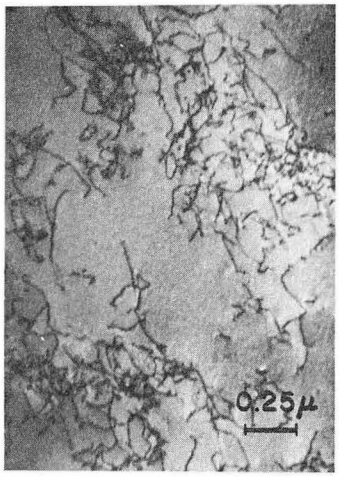
(d) Cu-10% Sn as (a) showing long coplanar dipole groups.

Fig. 2 Data on yield stresses and cell diameters for quenched and furnace cooled Cu-Sn alloys.

Fig. 3 Variation of stacking fault energy with alloying of copper. The data of 1-4 is taken from Swann (ref. 1) and 5 from Thomas (ref. 2).




(a)

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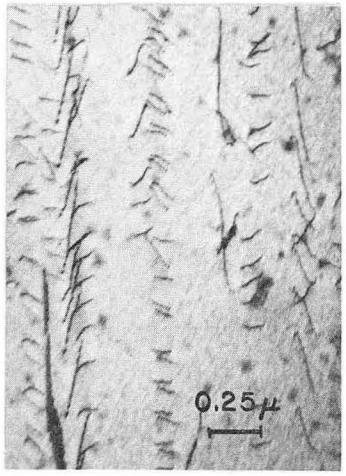
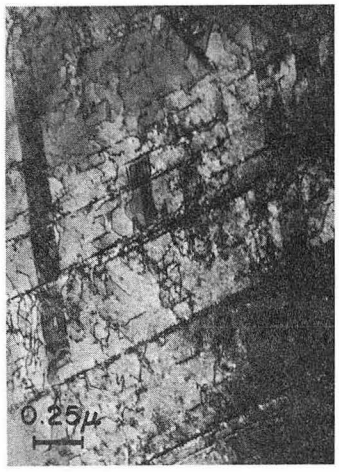



(c)




(d)

ZN-4646

Fig. 1

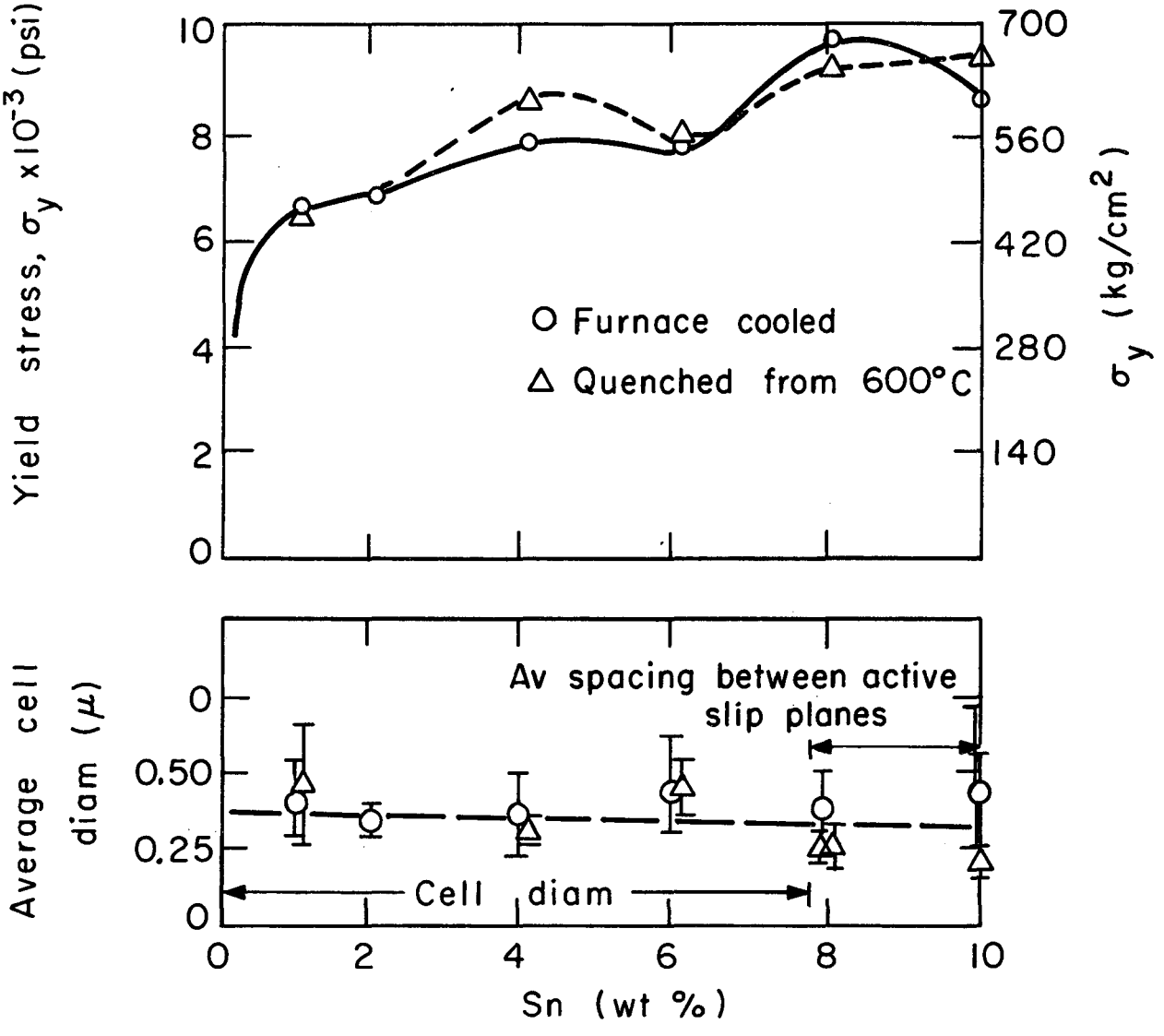

Yield stress, $\sigma_y$ x $10^{-3}$ (psi)
$\sigma_y$ (kg/cm$^2$)
O Furnace cooled
Δ Quenched from 600°C
Av spacing between active slip planes
Cell diam
Average cell diam (μ)
Sn (wt %)
MUB-4272-A

Fig. 2

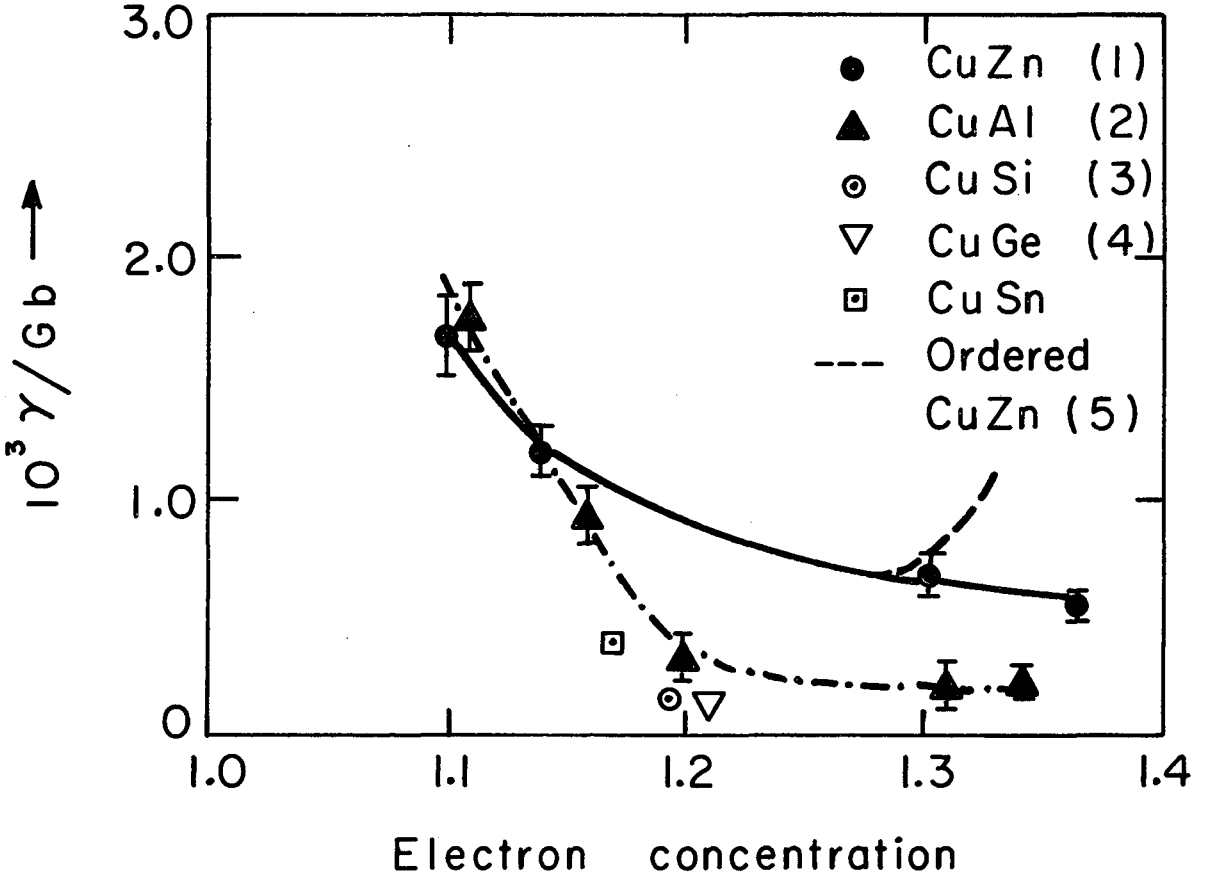



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Fig. 3