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ON THE NATURE OF DISPERSED PARTICLE
STRENGTHENING AT LOW TEMPERATURES

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STRENGTHENING AT LOW TEMPERATURES

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ABSTRACT

Polycrystalline aluminum-copper alloys containing 4 and 5% copper were given heat treatments designed to develop dispersions of Al_2Cu precipitate particles of various sizes and distributions. Tensile tests were performed on the dispersion hardened alloys at 4.2, 77, 90 and 114°K to determine the strain rate and temperature dependence of the flow stress. The low temperature tensile data were analyzed and evaluated in terms of the strain hardening theory of thermally activated intersection of dislocations.

The intersection analysis of the tensile data showed that the dispersed particles significantly increase the average effective forest dislocation density produced for a given amount of strain. The increased dislocation density was found to be in qualitative agreement with selected electron-microscope observations of the deformed structure of the alloys.

Estimates of contributions to the flow stress arising from effects coincident with the intersection of dislocations and those associated with the build-up of long range stress fields were made.

The results of this investigation indicate that the Fisher, Hart and Pry mechanism of strain hardening in dispersion hardened alloys is not the major source of strengthening in the aluminum-copper system. The major strengthening contribution of the dispersed particles during plastic deformation is shown to be due to an increase in the dislocation density. The increase in dislocation density appears to be associated with relatively high density tangles of dislocations observed to be localized near particles and groups of particles.

INTRODUCTION

Although investigations of mechanical properties of pure metals have firmly established the role of dislocations in many fundamental processes of deformation and strain hardening, relatively little fundamental experimental information is available concerning the nature of the particle-dislocation interactions which contribute to the strengthening of dispersion hardened alloys during plastic deformation. Previous investigations carried out on dispersion hardened alloys have been concerned principally with correlating the strengthening with observable parameters such as particle size and degree of dispersion. Several theories have been advanced, notably by Mott and Nabarro,¹ Gensamer,² Orowan³ and Fisher, Hart and Pry⁴ to explain the results of these investigations in terms of simple models of particle-dislocation interaction.

The Fisher, Hart and Pry model has generally been considered an acceptable explanation of non-deformed dispersed particle contributions to work hardening due to its simplicity and reasonable agreement with experimental results of mechanical tests on alloys containing a uniform dispersion of essentially spherically shaped particle.⁵ This theory explains the contribution of precipitate particles to strain hardening by assuming that the accumulation of dislocation loops left around precipitate particles lying on active slip planes due to the operation of the Orowan mechanism gives rise to a back stress which must be overcome by dislocations moving on the plane. This back stress is assumed to be a significant contribution to work hardening. The results of the Fisher, Hart and Pry theory show that after a given

amount of strain, the shear stress for further deformation, τ_h , should be given by

$$\tau_h = 3NGB f^{3/2}/r + \tau'_h$$

where N is the number of dislocation loops around each particle, f is the volume fraction of precipitate, r is the mean planar particle radius and τ'_h is the increment of flow stress due to all other sources of work hardening.

Recent electron-microscopy observations of deformed dispersion hardened alloys suggest that the nature of the particle-dislocation interaction is considerably more complex than indicated by the Fisher, Hart and Pry model. Thomas and Nutting⁶ using replica techniques have shown that during deformation of overaged Al-Cu alloys, slip is confined to regions between precipitate particles and the particles are not generally deformed. These observations also revealed that many slip systems are operating in the early stages of deformation of these alloys. Nicholson, Thomas and Nutting⁷ have made transmission electron-microscope observations of the deformed structure in thin foils of overaged Al-Cu alloys. This investigation showed that considerable cross-slip occurs in the alloys during the early stages of plastic strain. A prominent feature in these observations are complex tangles of dislocations of relatively high density tending to be localized at precipitate particles. Phillips⁸ has described the same features in his observations of thin foils of dispersion hardened Al-Cu alloys. While the complexity of detail so far observed in the electron-microscopy prohibits positive identification of the fundamental mechanisms by which particles strengthen the matrix, these observations do indicate the need for a more comprehensive theory of dispersion hardening.

Strain hardening of pure face-centered cubic metals has been a

subject that has received much attention by investigators attempting to relate observed mechanical properties of metals to their structure and imperfections. There is general agreement that a significant contribution to strain hardening in F. C. C. metals results from the intersection of glide dislocations with forest dislocations threading their glide plane. Mitra, Osborne and Dorn⁹ have demonstrated that the self-strengthening of pure aluminum and copper during low temperature deformation arises principally from effects associated with the intersection of dislocations. Their investigations were found to be in agreement with an analytical theory of strain hardening which is essentially an extension of theories originally proposed by Seeger¹⁰ and Basinski.¹¹ This analysis permits an estimation of the force-displacement diagram for the thermally aided intersection of dislocations and enables an estimation of the separate contributions to the flow stress resulting from effects coincident with the intersection of dislocations and from the build up of long range stress fields.

Assuming that the rate controlling mechanism of strain hardening in dispersion hardening F. C. C. metals deformed at low temperatures is the thermally activated intersection of dislocations, an analysis of low temperature mechanical tests in terms of intersection theory would disclose the effects of dispersed particles on contributions to the flow stress resulting from the build-up of long range stress fields and contributions associated with the intersection of dislocations. Knowledge of these effects are of significant aid in getting an insight into the nature of the fundamental mechanism by which dispersed particles strengthen the matrix.

The present investigation, devoted to the low temperature deformation of dispersion hardened aluminum-copper alloys was initiated

to ascertain if strain hardening of these alloys could be interpreted in terms of the intersection model and to establish through the intersection analysis the nature of the strengthening contributions of the dispersed particles.

EXPERIMENTAL PROCEDURES

Preparation of Specimens:

The aluminum-copper system was selected to develop dispersions of Al_2Cu precipitate particles for this investigation. This system has received considerable study and it is known that good control of auxiliary variables can be achieved while developing a uniform dispersion of precipitate particles. Also, the results of the analysis of the low temperature mechanical tests could be compared with previous work on pure aluminum and aluminum-magnesium solid solution alloys. In addition, to the degree of dispersion, the following factors were considered to be important in interpreting the properties of the alloys.

1. Grain size of the alpha phase.
2. Composition of the alpha phase.
3. Uniformity of dispersed particles, particularly the absence or only nominal presence of grain boundary precipitates.
4. A single type of incoherent equilibrium precipitate.

Alloys containing 4 and 5% copper by weight as shown in Table I were prepared from high purity aluminum-copper alloys cast into an ingot, homogenized, hot and cold rolled to 12-inch wide sheets, 0.100 inches thick and finally annealed.

The heat treatments required to obtain the desired features noted above had been previously determined.¹² The heat treatments selected for the test specimens were as follows:

A. The as-received alloys were solution heat treated and homogenized for 72 hours at 540°C and water quenched.

B. The alloys were then cold worked 30% preliminary to a recrystallization treatment at 540°C to provide a constant final grain size of the alpha phase. Since additional grain growth occurred during subsequent precipitation treatments, the time at the recrystallization and grain growth temperature had to be controlled for the various alloys and dispersion so that the final alpha solid solution grain size would be identical in all specimens.

C. The recrystallization treatments were followed by appropriate aging treatments. The aging treatments were selected to avoid excessive grain boundary precipitation and crystallographically oriented precipitates.

D. Finally, the alloys were given a preliminary aging treatment after which they were furnace cooled to 305°C and aged for two days. Specimens were then machined and an additional two days of aging at 305°C was practiced to remove the effects of machining and to establish the composition of copper in the alpha solid solution at the limit of solubility of copper in aluminum at 305°C. The actual heat treatments which were used are shown in Table II.

Table I
Chemical Composition of Alloys*

Nom. % by wt. Cu	Percent Weight Alloying Component						
	Chemical Analysis		Spectrographic Analysis				
	Cu	Fe	Si	Mg	Ca	Na	Mn
4%	4.03	.002	.002	.0004	<.0006	<.0006	.001
5%	5.240	.001	.001	.000	.000	.000	.000

* All alloys and analyses were furnished courtesy of the Aluminum Company of America Research Laboratories.

The resulting solid solutions were established by previous studies to contain about 0.19 atomic percent copper.¹² Furthermore, subsequent room temperature aging or strain aging during testing were observed to be insignificant. The specimens were etched in a solution of 6 HNO₃, 9 HCl, 2HF and 15 H₂O parts by volume to establish identical surface conditions and to determine the average alpha grain size. The grain size determinations were made by the intercept method. The results of these observations are recorded in Table II.

The average number of particles observed per unit area of polished surface, (N), were established for each group of specimens by metallographic inspection of a number of specimens of each of the degrees of dispersion. Calculations of the mean planar particle radius and mean planar interparticle spacing were made assuming the particles to be spherical and uniformly distributed throughout the matrix. Neither of these assumptions is perfectly valid in these alloys. Assuming also that each particle of radius r on the plane of polish has a circle of influence of radius R, then $r^2/R^2 = f$ where f is the volume fraction of particles in the matrix. The number of particles per unit area of matrix is $N = 1/\pi R^2$ and the mean planar particle spacing, D, is taken approximately equal to 2R. The volume fraction of precipitate is obtained from the aluminum-copper phase diagram together with values of the density of the alpha solid solution and the particles. Taking a density of 4.35 for CuAl₂, a density of 2.73 for the alpha solid solution and the solid solution to contain 0.194 atomic percent copper the values of f , r and D obtained are shown in Table II.

Mechanical Tests:

The low temperature tensile tests required for the analysis of strain hardening in terms of the intersection model were made on

Table II
Heat Treatments for Dispersions of CuAl_2

Description of % Cu dispersions	Recrystallization		Preliminary Aging Treatments				Grain size grains/mm.	N x 10^{-4}	f	r $\text{cm} \times 10^{-5}$	D $\text{cm} \times 10^{-4}$
	Time min.	Temp $^{\circ}\text{C}$	Time hrs	Temp $^{\circ}\text{C}$	Time hrs	Temp $^{\circ}\text{C}$					
4-M	2 1/2	540 W. Q.	1/2	440	A. C. & 3	350	2.3	1631	.0453	2.98	2.8
4-C	2 3/4	540 W. Q.	24	465			2.4	98	.0453	10.8	11.4
5-F	2 1/2	540 W. Q.	.025	525	W. Q. & 1	400	2.3	7620	.0595	1.56	

Symbols: C = coarse, M = medium, F = fine, W. Q. = water quench, A. C. = air-cooled, N = number of particles per unit area of plane of polish, r = mean planar particle radius, D = mean planar distance between particles, f = volume fraction of precipitate.

the Instron testing machine. Recorded stresses were estimated to be accurate to $\pm 3 \times 10^4$ dynes/cm² and strains to $\pm 10^{-4}$.

The mechanical testing consisted of determining the change in flow stress with rapid changes in the strain rate while the specimen was being strained in tension at a given temperature. The strain rate changes were made by means of a quick gear shift which changed the rate of head motion on the Instron from a minimum of 0.002"/min. to a maximum of 0.02"/min.

Tests were also conducted to determine the change in flow stress with changes in specimen temperature while straining at a constant rate of .02"/min. These tests were made by straining the specimen a desired amount, reducing the load to approximately 15% of the deformation load and quickly changing temperature controlled baths. The specimen was then allowed to come to equilibrium with the new bath temperature for ten minutes and the tensile straining was resumed at this temperature. This procedure was found to minimize creep and recovery during the interval of time the specimen was coming to equilibrium at the new bath temperature.

Application of the theory of thermally activated intersection of dislocations to the strain hardening of F.C.C. metals requires that the strain rate and temperature dependence of the flow stress be known for various states of deformation. The state of deformation of a work hardened F.C.C. metal is defined by a certain average density and distribution of dislocations that exhibit a characteristic flow stress at a given temperature and strain rate. Thus, two identical specimens strained at the same temperature and strain rate exhibiting identical stress-strain curves are assumed to contain the same average density and distribution of dislocations at a given point on the stress-strain curve

In order to experimentally establish the temperature and strain rate dependence of the flow stress at a given state of strain in all of the specimens of a given degree of dispersion, the specimens were pre-strained at 77°K and .02 "/min. to a previously selected stress level at which point they were assumed to be in the same average structural state of strain. The temperature was then changed as previously described and the straining resumed. Immediately after the specimen began to deform plastically at the new temperature, strain rate changes were made to determine the strain rate dependence of the flow stress at this temperature. A portion of the stress-strain curve depicting the results of the above procedure is given in Fig. 1. Calculations were made of the parameter $\beta' = \Delta \ln \dot{\epsilon} / \Delta \sigma$ at each point where strain rate changes were made. The β' values were then plotted as a function of the flow stress and the resulting curve extrapolated to the stress corresponding to the previously selected strain state. This procedure was repeated at test temperatures of 114, 90, 77 and 4.2°K for pre-strained states corresponding to tensile strains 0.90, 1.49, 2.08, 2.66 and 3.24%. The results of these tests are shown in Figs. 2a to 2l. The τ and β values plotted in these figures are the von Mises values where $\tau = \sigma/2$ and $\beta = 2\beta'$

Some questions existed as to the possibility of specimen recovery upon changing from the pre-strain temperature of 77°K to the higher temperatures of 90 and 114°K. This problem was resolved by alternating the bath temperatures so as to compare the Cottrell-Stokes ratios¹³ for changes from higher to lower temperatures with those obtained for changes from lower to higher temperatures over identical stress regions. The stress ratios were found to be identical within experimental error indicating that the flow stress changes were completely reversible and

FIGURE I. PORTION OF STRESS-STRAIN CURVE FOR A SPECIMEN PRE-STRAINED AT 77°K 2.66 % AND TESTED ALTERNATIVELY AT TWO STRAIN RATES AT 90°K.

FIGURE 2-A. PLOT OF $2\beta'$ vs $\sigma/2$ FOR SPECIMENS 4-C STRAINED AT 4.2 °K.
 $\dot{\epsilon} = 10^{-4}/\text{SEC.}$

FIGURE 2-B. PLOT OF $2\beta'$ vs $\sigma/2$ FOR
SPECIMEN 4-C STRAINED AT 77°K.
 $\dot{\epsilon} = 10^{-4}/\text{SEC.}$

FIGURE 2-C. PLOT OF $2\beta'$ vs $\sigma/2$ FOR SPECIMENS 4-C STRAINED AT 90°K .
 $\dot{\epsilon} = 10^{-4}\text{SEC}$.

FIGURE 2-D. PLOT OF $2\beta'$ vs $\sigma/2$ FOR SPECIMENS 4-G STRAINED AT 114°K. $\dot{\epsilon} = 10^{-4}/\text{SEC.}$

FIGURE 2-E. PLOT OF $2\beta'$ vs $\sigma/2$ FOR SPECIMENS 4-M STRAINED AT 4.2°K . $\dot{\epsilon} = 10^{-4}/\text{SEC}$.

FIGURE 2-F. PLOT OF $2\beta'$ vs $\sigma_{1/2}$ FOR SPECIMEN 4-M
STRAINED AT 77°K. $\dot{\epsilon} = 10^{-4}/\text{SEC}$.

FIGURE 2-G. PLOT OF $2\beta'$ vs $\sigma/2$ FOR SPECIMENS 4-M STRAINED AT 90°K . $\dot{\epsilon} = 10^{-4}/\text{SEC}$.

FIGURE 2-H. PLOT OF $2\beta'$ vs $\sigma/2$ FOR SPECIMENS 4-M STRAINED AT 114°K . $\dot{\epsilon} = 10^{-4}/\text{SEC}$.

FIGURE 2 - I. PLOT OF $2\beta'$ vs $\sigma/2$ FOR SPECIMENS 5-F STRAINED AT 4.2°K. $\dot{\epsilon} = 10^{-4}/\text{SEC.}$

FIGURE 2-J. PLOT OF $2\beta'$ vs $\sigma/2$ FOR SPECIMEN 5-F
STRAINED AT 77°K. $\dot{\epsilon} = 10^{-4}/\text{SEC}$.

FIGURE 2-K. PLOT OF $2\beta'$ vs $\sigma/2$ FOR SPECIMENS 5-F
STRAINED AT 90°K . $\dot{\epsilon} = 10^{-4}/\text{SEC}$.

FIGURE 2-L. PLOT OF $2\beta'$ vs $\sigma/2$ FOR SPECIMENS 5-F STRAINED AT 114°K .
 $\dot{\epsilon} = 10^{-4}/\text{SEC}$.

specimen recovery was insignificant.

Additional Observations:

A few selected transmission-electron microscope observations of the deformed structure of the dispersion alloys were made to establish the nature of the dislocation distribution around the Al_2Cu particles and to check previous observations described in the literature.^{6, 7, 8}

The specimens observed were prepared by heat treating thin sheets of the aluminum-copper alloy approximately 100 microns thick in the same manner as the tensile specimens and straining them about 5% at 77°K.

Thin films were then obtained by electropolishing the deformed sheets in an electrolyte solution of 20% perchloric acid (60%) and 80% absolute alcohol at an e. m. f. of 20 volts and current density of 0.1 amps/cm².

During thinning the electrolyte was suspended over a dewar of liquid nitrogen to maintain a low polishing temperature.

RESULTS AND DISCUSSION

The intersection theory of work hardening is based on the assumption that the thermally activated intersection of forest dislocations by glide dislocations is the rate controlling process of deformation. The analysis presented here is that developed by Osborne, Mitra and Dorn.⁹

As shown by Seeger¹⁰ the strain rate arising from thermally activated intersection of dislocations is given by

$$\dot{\gamma} = N L^2 b v_0 e^{-U/kT} \quad (1)$$

where $\dot{\gamma}$ = shear strain rate,
 N = number of points of contact per unit volume between
 dislocations moving on principal glide planes and
 forest dislocations,
 L = the mean spacing between forest dislocations,
 ν_0 = the Debye frequency,
 b = the Burgers vector,
 k = Boltzmann's constant,
 T = the absolute temperature,
 U = the activation energy for intersection.

In general, the activation energy for intersection will depend upon the type of intersection and fault created. One would expect screw-screw, edge-edge, and screw-edge intersections to contribute to the deformation; therefore, the activation energy in Eqn. (1) should be thought of as a smeared average of energy supplied by a thermal fluctuation to effect intersection.

Basinski¹¹ has postulated a unique force-activation distance curve for intersection between dislocations shown in Fig. 3. The thermal activation energy is related to this curve by the equation

$$U = \int_F^{F_m} X dF \quad (2)$$

where F_m = the maximum force required for intersection,
 F = the average applied force,
 X = the average distance dislocations move during thermal
 activation.

The average applied force on the dislocation in the lower limit of Eqn. (2) is given by

FIGURE 3. SCHEMATIC REPRESENTATION OF THE AVERAGE F - x CURVE FOR INTERSECTION.

$$F = (\tau - \tau^*) L b \quad (3)$$

where τ = the applied shear stress,

τ^* = the average back stress acting on the dislocations.

The back stress, τ^* , exerted on the glide dislocations arises from contributions which are independent of the temperature except through the shear modulus. The sources of these contributions will be considered in more detail later.

The temperature dependent contribution to the flow stress is associated with the constriction of the extended dislocations and the formation of the jogs occurring during intersection. This contribution is retained in the force-distance curve. If the specimen was deformed at absolute zero the work done to effect intersection would be contributed solely by the mechanically applied force given by:

$$F = F_m = (\tau_m - \tau^*) L b \quad (4)$$

During deformation at a temperature above absolute zero some of the energy required to effect intersection is contributed by a thermal fluctuation as shown by the shaded area under the $F-X$ curve in Fig. 3. The maximum mechanical force required to effect intersection is now

$$F = F_L = (\tau_L - \tau^*) L b \quad (5)$$

The very heavy upper curve in Fig. 3 refers to the F_0-X curve for intersection at absolute zero. At any other temperature, T , the ordinates are reduced by G_T/G_0 as shown by the light solid curve. The ratio G_T/G_0 corrects for the change in force through the temperature

dependence of the shear modulus.

Combining Eqns. (1), (2) and (3) it follows that

$$\beta = \left(\frac{\partial \ln \dot{\gamma}}{\partial \tau} \right)_T = - \frac{1}{kT} \left(\frac{\partial U}{\partial \tau} \right) = \frac{1}{kT} \left(\frac{\partial U}{\partial F} \right) \left(\frac{\partial F}{\partial \tau} \right)$$

$$\beta = \frac{x}{kT} \left(1 - \frac{\partial \tau^*}{\partial \tau} \right) Lb \quad (6)$$

Osborne, Mitra and Dorn⁹ have shown that τ^* is not sensitive to τ but depends primarily on the strain hardened state, thus

$$\beta kT = 2\beta' kT = LbX \quad (7)$$

Due to its units (cm^3) βkT is called the activation volume. The activation volumes calculated for the three dispersion alloys is shown in Table III.

Eqn. (7) permits an estimate of L for various strain states to be made by utilizing Dorn and Mitra's⁹ assumption that at 4.2°K the thermal energy is small and the activation distance, x, is approximately equal to the Burgers vector b, hence at 4.2°K

$$L = 2\beta' kT / b^2 \quad (8)$$

The calculated values of L, using $b = 2.86 \times 10^{-8}$ cm, are shown in column (6) in Table III.

Considering the previous discussion and the approximations made in determining the average values of L it is doubtful that these values are reliable measures of the average spacing between the forest dislocations in the matrix. However, they do give a reasonable measure of the

Table III
Specimens 4-M

	T $^{\circ}K$	τ $\text{dynes/cm}^2 \cdot 10^8$	$2\beta'$ $\text{cm}^2/\text{dynes} \cdot 10^7$	$2\beta'kT$ $\text{cm}^3 \cdot 10^{21}$	χ $\text{cm} \cdot 10^3$	L $\text{cm} \cdot 10^6$	$\tau L b G_0/G_T$ $\text{dynes} \cdot 10^5$
$\epsilon = .010$	114	2.69	5.90	9.30	11.0	2.94	2.38
	90	2.78	5.25	6.52	7.75	2.94	2.44
	77	2.82	5.00	5.31	6.3	2.94	2.46
	4.2	3.36			2.86	2.94	2.83
$\epsilon = .015$	114	3.68	4.25	6.70	10.6	2.20	2.45
	90	3.79	3.85	4.80	7.6	2.20	2.50
	77	3.85	3.75	3.98	6.3	2.20	2.52
	4.2	4.59			2.86	2.20	2.90
$\epsilon = .021$	114	4.46	3.65	5.75	10.7	1.89	2.53
	90	4.60	3.30	4.10	7.6	1.89	2.58
	77	4.66	3.20	3.40	6.3	1.89	2.60
	4.2	5.55			2.86	1.89	2.99
$\epsilon = .027$	114	5.17	3.10	4.87	11.0	1.54	2.40
	90	5.33	2.80	3.48	7.9	1.54	2.45
	77	5.41	2.75	2.78	6.3	1.54	2.47
	4.2	6.45	21.6	1.26	2.86	1.54	2.85
$\epsilon = .034$	114	5.80	2.70	4.25	10.9	1.36	2.36
	90	5.99	2.45	3.05	7.85	1.36	2.42
	77	6.08	2.40	2.44	6.3	1.36	2.44
	4.2	7.25	19.2	1.11	2.86	1.36	2.81

Values of $2\beta'kT$, χ , L , and $\tau L b G_0/G_T$ for five pre-strain states in each dispersion alloy.

Table III
Specimens 4-C

	T $^{\circ}K$	τ $\text{dynes/cm} \cdot 10^{-9}$	$2\beta'$ $\text{cm}^2/\text{dyne} \cdot 10^7$	$2\beta'kT$ $\text{cm}^3/10^{21}$	X $\text{cm} \cdot 10^8$	L $\text{cm} \cdot 10^6$	$\tau L b^{G_0/G_T}$ $\text{dynes} \cdot 10^5$
$\epsilon = .010$	114	2.26	5.45	8.55	10.6	2.82	1.92
	90	2.34	5.00	6.20	7.7	2.82	1.97
	77	2.40	4.80	5.10	6.3	2.82	2.00
	4.2	2.91			2.86	2.82	2.36
$\epsilon = .015$	114	2.92	4.30	6.80	10.6	2.25	1.99
	90	3.07	4.05	5.02	7.8	2.25	2.04
	77	3.14	3.82	4.05	6.3	2.25	2.07
	4.2	3.78			2.86	2.25	2.43
$\epsilon = .021$	114	3.52	3.80	5.96	10.9	1.91	2.03
	90	3.67	3.52	4.36	8.0	1.91	2.08
	77	3.75	3.25	3.46	6.3	1.91	2.11
	4.2	4.54			2.86	1.91	2.48
$\epsilon = .027$	114	4.01	3.45	5.42	11.1	1.72	2.08
	90	4.16	3.15	3.90	7.9	1.72	2.13
	77	4.26	2.90	3.08	6.3	1.72	2.16
	4.2	5.14	24.4	1.41	2.86	1.72	2.53
$\epsilon = .034$	114	4.51	3.08	4.85	11.0	1.54	2.09
	90	4.70	2.75	3.41	7.8	1.54	2.15
	77	4.80	2.62	2.78	6.3	1.54	2.18
	4.2	5.79	21.8	1.26	2.86	1.54	2.54

Values of $2\beta'kT$, X , L and $\tau L b^{G_0/G_T}$ for five pre-strain states in each dispersion alloy.

Table III
Specimens 5-F

	T $^{\circ}K$	τ $\text{dynes} \cdot 10^{-8} / \text{cm}^2$	$2\beta'$ $\text{cm}^2 / \text{dyn} \cdot 10^7$	$2\beta'kT$ $\text{cm}^3 \cdot 10^{21}$	χ $\text{cm} \cdot 10^8$	L $\text{cm} \cdot 10^6$	$\tau L b G_0 / \epsilon_T$ $\text{dynes} \cdot 10^5$
$\epsilon = .010$	114	3.24	5.40	8.50	10.5	2.82	2.76
	90	3.33	4.95	6.15	7.7	2.82	2.80
	77	3.38	4.80	5.10	6.3	2.82	2.82
	4.2	3.98			2.86	2.82	3.22
$\epsilon = .015$	114	4.29	3.95	6.21	10.7	2.03	2.61
	90	4.42	3.60	4.47	7.7	2.03	2.66
	77	4.48	3.45	3.66	6.3	2.03	2.68
	4.2	5.28			2.86	2.03	3.06
$\epsilon = .021$	114	5.19	3.20	5.04	10.9	1.62	2.53
	90	5.36	2.85	3.54	7.65	1.62	2.58
	77	5.44	2.75	2.92	6.3	1.62	2.60
	4.2	6.41			2.86	1.62	2.97
$\epsilon = .027$	114	5.76	2.85	4.49	10.8	1.44	2.50
	90	5.94	2.60	3.22	7.8	1.44	2.55
	77	6.03	2.45	2.60	6.3	1.44	2.57
	4.2	7.11	20.3	1.18	2.86	1.44	2.94
$\epsilon = .034$	114	6.62	2.40	3.78	10.9	1.21	2.40
	90	6.85	2.20	2.73	7.9	1.21	2.46
	77	6.94	2.05	2.18	6.3	1.21	2.48
	4.2	8.20	17.1	0.99	2.86	1.21	2.89

Values of $2\beta'kT$, χ , L and $\tau L b G_0 / \epsilon_T$ for five pre-strain states in each dispersion alloy.

effective forest density changes with strain in the three dispersion alloys.

With these values of L for each strain state, Eqn. (7) can be used to determine the activation distance (x) corresponding to the other test temperatures shown in column (5) of Table III.

The force acting on a dislocation depends on the temperature through the shear modulus, therefore, the force-distance curves will differ for deformation at different temperatures. This difference is corrected for by reducing all test data to equivalent values at absolute zero through the relation

$$F_0 = F_T (G_0/G_T) = (\tau - \tau^*) (G_0/G_T) Lb \quad (9)$$

While the temperature dependence of the shear modulus for the alloys is not known, the corrections should be quite small over the temperature range of the tests. Also, the ratios should not be very different from those of pure aluminum; therefore, we shall use the shear moduli ratios of pure aluminum for this correction.⁹

Fig. 4 shows plots of $\tau Lb (G_0/G_T)$ versus x for the three dispersions. Since the curves for a given dispersion differ within experimental error only by constant differences in elevation, it follows that

$\tau^* Lb (G_0/G_T)$ between these curves is constant regardless of the value of τ , confirming the previous statement that τ^* is not sensitive to τ but depends primarily on the strain hardened state. These curves are essentially force-distance curves except that they are displaced along the ordinate axis by an amount equal to the force exerted by the average back stress corresponding to the particular strain state.

The plot also shows that the $F_0 - x$ curves for the three dispersions are essentially identical within experimental error over the region of

FIGURE 4. PLOT OF $\tau L b G_0/G_T$ vs X FOR THE 3 DISPERSION ALLOYS

strain studied indicating that the same unique $F_0 - X$ curve exists during deformation of the three dispersions. These results are in agreement with the intersection theory postulate that a unique $F_0 - X$ curve exists for intersection controlled deformation and show that the rate controlling mechanism of deformation is not altered by variations in the degree of dispersions of precipitate particles.

Having shown that the rate controlling mechanism of deformation in the alloys might be interpreted in terms of intersection theory, the analysis can be extended to estimate the strengthening contributions arising from effects associated with the intersection of dislocations and those due to the build-up of long range stress fields.

The temperature dependent contribution to the flow stress during deformation at a given temperature, τ_T^* , can be estimated from the $\tau L b (G_0/G_T)$ versus x curve in Fig. 4. During deformation at higher temperatures corresponding to larger values of x where the curves become asymptotic to the x -axis, the energy required for constriction and jogging is contributed by thermal fluctuations and the mechanical contribution is essentially zero. Thus, an estimate of the mechanical force required for constriction of the extending dislocations during deformation at a lower temperature can be obtained from the difference in ordinate values of the curve at the x corresponding to the temperature of deformation and at values of x where the curve becomes asymptotic to the x -axis. The temperature dependent force at 77°K is estimated to be 0.2×10^{-5} dynes.

During deformation of the alloys, the average back stress on the glide dislocations arises from the sum of contributions at two sources; the interaction back stress and the long range back stress.

$$\tau^* = \tau_i^* + \tau_l^* \quad (10)$$

The interaction back stress, τ_i^* , results from the elastic interaction between dislocations intersecting at various angles of inclination and having various angles between their Burgers vectors. An estimate of the average value of the interaction back stress can be obtained from experiments on single crystals of aluminum. Near the beginning of Stage I of the stress-strain curve of aluminum single crystals the dislocation density is quite low and the long range back stress arising from the stress fields of dislocations in the matrix is assumed to be negligible. When a single crystal of aluminum is strained at a low strain rate and at a temperature corresponding to a point on the $\tau L b (G_0/G_T)$ versus $\dot{\epsilon}$ curve, where it is asymptotic to the x-axis, the flow stress will be almost entirely due to the back stress on glide dislocations resulting from short range elastic interaction. Mitra, Osborne and Dorn⁹ have obtained values of $\tau_i^* L b$ for single crystals of aluminum of about 0.76×10^{-5} dynes. This value is thought to be a little high due to the assumptions stated above.

Hirsch¹⁴ and Basinski¹¹ have shown that the temperature independent contribution to the flow stress due to short range elastic interactions between intersecting dislocations is given by

$$\tau_i^* = \alpha G b / L \quad (11)$$

Theoretical discussions of interactions between intersecting dislocations have also been presented by Friedel¹⁵ and Saada.¹⁶ These investigators suggest that a major source of strengthening from dislocation intersection arises from the recombination of the dislocations over a length

depending on the angle of inclination at which the dislocations cross.

The results of their calculations give the back stress due to recombination as

$$\tau_r^* = \alpha' Gb/L \quad (12)$$

The factor α' for recombination is calculated from an average inclination angle between the intersecting dislocations and is approximately equal to 0.2.

The value of α in Eqn. (10) or (11) determined from aluminum single crystal experiments gave $\alpha \approx 0.035$. Similar experiments for copper gave $\alpha \approx 0.046$. While it is difficult to establish definitely the interactions that accompany the thermally activated intersection of dislocations, the discrepancy between the experimental values of α and those calculated from the recombination theory indicates that recombination is not a significant factor in the strengthening of these metals. The short range interaction contribution to the flow stress will be taken as

$$\tau_i^* = 0.035 Gb/L \quad (13)$$

The long range back stress, τ_l^* , may arise from the stress fields associated with the distribution of dislocations in the matrix or in overcoming attractive stresses between dislocations in entanglements. While the back stress resulting from the interaction of the long range stress fields is complex and may vary locally throughout the matrix, an estimate of the average long range back stress can be obtained from the intersection analysis as follows:

The force acting on a dislocation is given by

$$F = (\tau - \tau^*) Lb$$

hence $\tau = F/Lb + \tau_i^* + \tau_L^*$

$$= FX/\beta kT + \alpha Gb^2X/\beta kT + \tau_L^*$$

$$\therefore \log \tau = \log \left[(FX + \alpha Gb^2X/\beta kT) + \tau_L^* \right] \quad (14)$$

Thus, if τ_L^* increases linearly with $1/L$, a log-log plot of the above equation could result in a 45° line. A non-linear increase in τ_L^* would be indicated by a deviation from the 45° line. Fig. 5 is a plot of $\log \tau$ versus $\log \beta kT$ for the three dispersion alloys strained at 77°K . The deviation from the 45° line indicates a non-linear increase in the long range back stress with strain. An estimate of the magnitude of the long range back stress at a given strain state is taken as the difference between the datum values and the 45° line resulting from the calculated constriction and short range interaction contributions to the flow stress.

The change in the effective forest dislocation density with strain for the three dispersion alloys strained at 77°K is shown in Fig. 6. These plots indicate that the degree of dispersion has a significant effect on the average forest dislocation density produced for a given amount of strain. Comparison of the relative magnitudes of these forest densities with those previously determined for polycrystalline pure aluminum specimens (having the same grain size) reveal that the presence of precipitate particles significantly increases the effective forest dislocation density over that which would be obtained for the same amount of strain in a particle free matrix deformed at the same temperature and strain rate.

FIGURE 5. PLOT OF LOG τ vs LOG βkT FOR THE 3 DISPERSION ALLOYS STRAINED AT 77°K.
 $\dot{\epsilon} = 10^{-4}/\text{SEC.}$

FIGURE 6. VARIATION OF $1/L$ WITH STRAIN FOR THE 3 DISPERSION ALLOYS AND FOR PURE Al POLYCRYSTALLINE SPECIMENS STRAINED AT 77°K . $\dot{\epsilon} = 10^{-4}/\text{SEC}$.

Transmission electron-microscope observations of the deformed structure in thin foils of these dispersion alloys revealed complex tangles of dislocations of relatively high density tending to be localized at particles and groups of particles. Observation of loops around particles was uncommon; however, there was some tendency for the dislocations to bend around the particles. These observations are in agreement with those described in previous electron-microscope investigations of Al-Cu alloys.^{6,7,8} The calculated values of the average spacing between forest dislocations is in qualitative agreement with the spacing between dislocations observed in the tangles in the thin foils. Thus, it appears that the effective densities of forest dislocations are associated with the relatively high density tangles of dislocations localized near the precipitate particles.

Figs. 7a-7c show the contributions to the flow stress as a function of strain for the three dispersion alloys strained in tension at 77°K and at a strain rate of 10^{-4} /sec. The long range back stress, τ_l^* , increases rapidly in the early stages of deformation and levels off to nearly a constant value at higher strains. The interaction back stress, τ_i^* , increases in proportion to the increase in forest density with strain as does the temperature dependent contribution to the flow stress.

The results of this investigation indicate that the Fisher, Hart and Pry mechanism of dispersed particle hardening is not a major source of strengthening of the Al-Cu dispersion hardened alloys. While it may be possible that this mechanism does make some contribution to strain hardening the observed distribution of dislocations around the precipitate particles indicates that in general the particle dislocation interactions are much more complex than the simple mechanism of dislocation loops forming around the particles. Previous investigations which tend to confirm the Fisher, Hart and Pry theory of particle strengthening were based on the assumption that the dispersed particles

FIGURE 7-A. VARIATION OF CONTRIBUTIONS TO THE FLOW STRESS WITH STRAIN FOR THE DISPERSION ALLOY 4-C AND PURE Al POLY-CRYSTALLINE STRAINED AT 77°K. $\dot{\epsilon} = 10^{-4}$ SEC.

FIGURE 7-B. VARIATION OF CONTRIBUTIONS TO THE FLOW STRESS WITH STRAIN FOR THE DISPERSION ALLOY 4-M STRAINED AT 77°K.
 $\dot{\epsilon} = 10^{-4}$ /SEC.

FIGURE 7-C. VARIATION OF CONTRIBUTIONS TO THE FLOW STRESS WITH STRAIN FOR THE DISPERSION ALLOY 5-F STRAINED AT 77°K.
 $\dot{\epsilon} = 10^{-4}$ SEC.

make no significant changes in the dislocation density in the matrix.⁵ These investigations assume the strengthening contribution from the matrix by some mechanism such as the intersection of dislocations is independent of the degree of dispersion of the particles and on this basis the experimental results were shown to be in reasonable agreement with the Fisher, Hart and Pry theory. The results of this investigation show that the dispersed particles significantly alter the density and distribution of dislocations in the matrix and that this is a major source of strengthening the alloys.

The initial large increase in the long range back stress relative to the increase in the back stress associated with the intersection of dislocations leads one to expect that the Cottrell-Stokes ratios should initially be relatively small and increase during the first few percent of strain. At larger strains the long range back stress levels off to nearly a constant value and since the back stress associated with the intersection of dislocations increases in proportion to the dislocation density one would expect the Cottrell-Stokes ratios to be nearly constant. Measurements of the Cottrell-Stokes ratios at larger strains were found to be constant within experimental error. During the first few percent of strain there did appear to be a tendency for the Cottrell-Stokes ratios to be smaller and increase with strain; however, these observations are not positive since the flow stress values for changes in temperature were extrapolated from recorded stress-strain curves and the accuracy of these extrapolated values is not considered good enough to establish definitely a trend of small changes in the Cottrell-Stokes ratios.

The results of the intersection analysis together with the electron-microscope observations of the deformed structure of the dispersion

hardened alloys suggest that the major strengthening effect of the dispersed particles is due to an increase in the effective forest dislocation density in the matrix. The increase in forest density appears to be associated with the relatively high density tangles of dislocations localized near particles. While the fundamental particle-dislocation interactions which result in these tangles are obscure there appears to be only two likely mechanisms of developing the observed distributions of dislocations. The dislocation tangles may result from a complex process of bending and cross-slip of dislocations that meet particles lying in their glide plane as they move through the matrix. A mechanism of this type has been proposed by Hirsch⁶ to explain the slip characteristics and dislocation distribution observed by Thomas and Nutting⁶ in overaged Al-Cu alloys. A second possible mechanism responsible for the increase in the forest dislocation density is the generation of dislocations at the particle-matrix interface. This is not an unreasonable mechanism since the incoherent particle-matrix interface could contain local regions of high stress concentration which act as sources of dislocations under an applied stress; however, the available evidence seems to favor the cross-slip mechanism of increasing the dislocation density. Determination of the principal mechanism responsible for the increase in the dislocation density in the dispersion hardened alloys is a problem for the electron-microscopist.

CONCLUSIONS

Assuming that the rate controlling mechanism of deformation of the dispersion hardened alloys is the thermally activated intersection of dislocations, the following conclusions may be reached:

1. The presence of Al_2Cu precipitate particles in the solid solution matrix significantly increases the effective forest dislocation density over that which would be produced for the same amount of deformation under identical experimental conditions in a particle free matrix. The rate of increase in dislocation density with strain increases with finer dispersions of precipitate particles.

2. The major strengthening contribution of the precipitate particles arises from effects coincident with the increased dislocation density in the matrix. The large increase in effective forest dislocation density during deformation is associated with the formation of relatively high density dislocation tangles localized near particles and groups of particles.

3. The Fisher, Hart and Pry mechanism of particle contribution to strain hardening does not appear to be a significant source of strengthening of the alloys. While the build-up of long range back stress in the alloys is in accord with the predictions of the Fisher, Hart and Pry theory, the back stress more likely results from the observed dislocation tangles than from loops formed around particles which were observed to be relatively uncommon.

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