Effects of Changes in Strain Path on Work Hardening in Cubic Metals

M. ZANDRAHIMI, S. PLATIAS, D. PRICE, D. BARRETT, P.S. BATE, W.T. ROBERTS, and D.V. WILSON

Effects of abrupt changes in strain path on work hardening in stretching 1200 aluminum, OFHC copper, 70-30 brass, a low-carbon ferritic steel, and 310 and 304 austenitic steels have been investigated. Tests were made with first-stage extension in uniaxial and equibiaxial tension. Second-stage stretching was in uniaxial tension. It is shown that all of the above materials have some susceptibility to transient reductions in work-hardening rate, $d\sigma/d\varepsilon$, after changes in strain path. Some of the material and process variables which can influence the form and magnitude of the transients are identified. Within the group of materials tested, stacking fault energy (SFE) appears to be the most generally influential material variable. The magnitudes of prestrain required to give significant reductions in $d\sigma/d\varepsilon \cdot 1/\sigma$ after strain path changes are higher in the low SFE alloys. Changes in strain path are accompanied by a reduction in the effective extension at the first onset of strain localization, when the reduction in $d\sigma/d\varepsilon$ is sufficient to cause $d\sigma/d\varepsilon \cdot 1/\sigma$ to fall below unity. Within the ranges of prestrain explored, the maximum reductions in $d\sigma/d\varepsilon \cdot 1/\sigma$ were in the range of 0.5 to 1.0. Thus, none of the changes in strain path investigated caused a reduction in the limit of effective uniform elongation, until the prestrain was sufficient to reduce $d\sigma/d\varepsilon \cdot 1/\sigma$ in monotonic deformation to less than a value in the range of 1.5 to 2.0. For these reasons, the possible reductions in effective uniform elongation were much more severe in the high SFE materials than those in the low SFE alloys.

I. INTRODUCTION

WHEN the strain path is changed during plastic deformation of a metal, its flow strength and work-hardening rate diverge from those which are characteristic of monotonic deformation. The responses of different metals and alloys to changes in the mode of deformation have been investigated in two-stage mechanical tests involving an abrupt change in strain path. Early investigations of this kind, notably by Ghosh and Backofen[1] and Hutchinson *et al.*,[2] demonstrated the practical importance of changes in work hardening caused by changes in strain path in relation to stretch formability in sheet metal shaping operations. Salient features of investigations made in the period up to 1982 are summarized by Wagoner and Laukonis[3] in their paper on the tensile behavior of an aluminum-killed (AK) steel after prestraining in plane strain. It was concluded from this early work that the changes in work hardening following a change in strain path are predominantly transient and that both positive and negative changes in hardening rate can occur in two-stage tests made on different materials. It was also concluded that the transient changes in plastic response can often be characterized as being one of two types,[3-6] *viz.*,

(1) type 1: lowered initial flow stress accompanied by an increased hardening rate and
(2) type 2: increased initial flow stress accompanied by a reduced hardening rate.

Using finite element modeling, Chung and Wagoner[6] have shown how particular forms of each of these types of transient can influence uniform elongation and total elongation in uniaxial tensile tests.

The prominent transients developed in ferritic low-carbon steels, commercial purity (CP) aluminum, and some low-strength aluminum alloys are usually type 2.[2-5] Transmission electron microscope (TEM) investigations of the microstructural behavior of aluminum[7] and low-carbon steels[8] have shown that large changes in strain path can destabilize the dislocation cell structures established in prior deformation. In the early stages of the second deformation mode, the original cell structures are disrupted and appear to be partially dissolved, and then, in continued deformation, new dislocation cell structures are built up. Relative to work-hardening rates developed in monotonic deformation, the first phase of such structural changes appears to coincide with a reduction in hardening rate and the generation of a new cell structure with recovery in hardening toward that characteristic of monotonic deformation.

The 70-30 brass is widely quoted as being a material which exemplifies type 1 behavior, although some precipitation and dispersion-hardened alloys provide more striking examples.[5] Wagoner[9] has recognized that the lowering of flow stress and the increase in work-hardening rate, which is developed in 70-30 brass within about the first 0.04 of plastic strain after a large change in strain

S. PLATIAS, formerly Postgraduate Student, School of Metallurgy and Materials, University of Birmingham, is Technical Superintendent, Hellenic Ferroalloys S.A., Athens, Greece. P.S. BATE, formerly Research Fellow, School of Metallurgy and Materials, University of Birmingham, is with INCO Engineering Products Ltd., Birmingham B16 OAJ, United Kingdom. M. ZANDRAHIMI, Postgraduate Student, D. PRICE, Senior Technician, D. BARRETT, Senior Technician, W.T. ROBERTS, Reader, and D.V. WILSON, Emeritus Professor, are with the School of Metallurgy and Materials, University of Birmingham, P.O. Box 363, Birmingham B15 2TT, United Kingdom.

Manuscript submitted October 3, 1988.

path, are controlled by the internal stresses which are generated during the prestrain. However, the measurements of Hutchinson *et al.*[2] and Ranta-Eskola[10] on 70-30 brass prestrained in uniaxial tension indicate that after some changes in strain path, the hardening rate can be reduced within an interval of second-stage strains beyond about 0.04. This effect is most easily recognized in tests at 45 deg to the prestrain direction.[10] Evidently, transients of types 1 and 2 are not necessarily mutually exclusive.

When the change in hardening rate is initially positive but decays to ~0 within an interval of second-stage strains which is very small compared with ε_u (the limit of uniform elongation in monotonic stretching), then, after prestrains which are appreciably less than ε_u, it is unlikely that such a transient will have a significant influence on $\bar{\varepsilon}_u$, the effective limit of uniform elongation in two-stage stretching.[6] In contrast, a transient reduction in hardening rate may reduce $\bar{\varepsilon}_u$ significantly, if its amplitude is sufficient to initiate strain localization at an early stage in the second deformation mode. For example, with an AK steel having a prestrain in excess of ~0.08 in uniaxial tension in tests at 45 or 90 deg to the prestrain direction, diffuse strain localization may be initiated after second-stage strains of ~0.02. However, when the prestrain is not too severe, the recovery in hardening rate at higher second-stage strains can be sufficient to reestablish stable elongation, so that the development of localization toward necking failure is interrupted by an interval of stable extension. For the case of a low-carbon AK steel, factors influencing this kind of behavior have been investigated and analyzed by Laukonis.[11]

Quantitative differences in transient changes in work-hardening behavior, which can be significant in relation to the effects of strain path changes on stretch-forming limits, are difficult to evaluate from published stress/strain relationships. More precise knowledge of the influences of process and material variables on the amplitudes and strain dependencies of the transients in work-hardening rates is required. The main objective of the present research was to provide this kind of information for a wide range of the sheet metals and alloys which are used in press forming. However, the phase of the investigation which is described in this paper was confined to cubic metals and alloys, which are expected to behave, essentially, as single-phase polycrystals.

II. EXPERIMENTAL METHODS

A. *Material Characteristics*

Commercial quality sheets of a batch-annealed low-carbon AK steel, three different samples of AA 1200 aluminum, an oxygen-free high-conductivity (OFHC) copper, 310 and 304 austenitic steels, and a 70-30 brass were investigated. Their chemical compositions and the abbreviated designations used in the text are given in Table I. These materials were selected to provide a wide range of stacking fault energies, SFE's, and corresponding work-hardening characteristics. The ferritic steel and CP aluminums represent high SFE materials. The SFE of aluminum in m J m $^{-2}$ is believed to be about 200. In contrast, those of the 310 steel, the 304 steel, and 70-30 brass are probably about 45, 20, and 10, respectively. Within this wide range, the SFE of copper, having a value of 70 to 80 m J m^{-2}, can be regarded as intermediate.

Sheet thicknesses, grain sizes, and mechanical properties measured in uniaxial tension in the conditions used in the tests are summarized in Table II. A1, A3, the copper, and the 70-30 brass were received in the half-hard (H4) condition. A1 was annealed at 350 °C for 15 minutes, A3 at 350 °C and 400 °C for 15 minutes, the copper at 400 °C for one hour, and the 70-30 brass at 550 °C for one hour and at 690 °C for 15 minutes. Aluminum A2 was cold rolled from 6- to 1.0-mm thickness and then annealed at 350 °C for 15 minutes. The 310 and 304 austenitic steels were received and used in the mill-annealed condition. Optical metallography showed that the grains in A3 were elongated in the rolling direction, whereas those in A1 and A2 were close to being equiaxed. A TEM examination showed that after annealing at 350 °C, A3 had a predominantly recovered structure with well-developed subgrain boundaries. A1 and A2 had recrystallized microstructures.

Crystallographic textures of the annealed sheets were determined using standard X-ray methods. The "as-received" low-carbon ferritic steel had a predominantly

Table I. Compositions of Materials in Weight Percent

Material	Designation	Fe	Ni	Cr	Mn	C	Si	Al	S	P
Low-carbon ferritic AK steel	AK steel	remainder	0.05	0.02	0.35	0.076	0.007	0.05	0.024	0.009
310 Austenitic steel	310	50.9	19.9	26.6	1.2	0.05	0.54	0.02	—	0.025
304 Austenitic steel	304	68.8	8.8	19.5	1.4	0.04	0.38	0.01	0.007	0.025

CP Aluminum

		Al	Fe	Si	Cu	Mn	Zn	Ti	Mg
1 AA 1200	A1	remainder	0.40	0.20	0.003	0.010	0.02	0.001	0.001
2 AA 1200	A2	remainder	0.29	0.08	0.001	0.005	0.01	0.012	0.001
3 AA 1200	A3	remainder	0.37	0.04	0.001	0.004	0.01	0.011	0.001

Copper-Based Materials

		Cu	Fe	Ni	Pb	Sn	Si	Zn
OFHC copper	Copper	remainder	0.01	<0.01	<0.005	<0.01	<0.005	<0.01
70 Cu-30 Zn brass	70-30	69.5	0.02	0.012	0.005	0.005	0.003	remainder

Table II. Sheet Thicknesses, Grain Sizes, and Tensile Properties

Material	Sheet Thickness (mm)	Grain Size (μm)	UTS (MPa)			Uniform Elongation (Pct)			Total Elongation (Pct)		
			0 Deg	45 Deg	90 Deg	0 Deg	45 Deg	90 Deg	0 Deg	45 Deg	90 Deg
AK steel	0.88	24	274	286	273	24	23	26	50	46	42
A1 (350)	0.91	105	72	77	68	30	33	29	40	49	45
A2 (350)	1.0	22	91	88	91	30	28	30	43	44	43
A3 (350)	0.89	80	80	84	79	13	24	15	36	45	35
A3 (400)	0.89	99	75	81	73	20	25	19	36	45	38
Copper	1.2	73	232	209	215	41	42	41	51	56	50
70-30 (550)	1.2	30	356	330	349	44	45	45	—	—	—
70-30 (690)	1.2	125	309	280	303	63	65	67	70	79	80
304	0.62	15	665	655	652	75	72	76	84	86	86
310	0.87	14	651	629	624	41	40	40	51	51	50

Grain sizes are described in terms of average grain boundary intercept lengths in a plane section. Mechanical properties were measured in uniaxial tension at 0, 45, and 90 deg to the rolling direction. Elongations were measured on a 25 mm gage length.

{111} ⟨*uvw*⟩ texture typical of a commercially processed and batch-annealed AK steel. However, it was shown that an extension of ~15 pct in the rolling direction (RD) in uniaxial tension gave a significant reinforcement of the {111} ⟨110⟩ components in this texture. The textures of annealed sheets of aluminum A1 and A3 had strong {100} ⟨001⟩ components, whereas that of A2 had a more prominent "retained rolling" component. In their annealed conditions, both the copper and 70-30 brass had very weak textures, and planar anisotropy of their tensile properties was slight (Table II). The textures of the 310 and 304 austenitic steels in the annealed conditions were also weak, but they showed a trace of the retained rolling component. However, after prestrains of ~20 pct in uniaxial tension, the texture was more strongly developed.

Tests were made for transformation to α' martensite in the austenitic steels using a Sucksmith magnetic balance. These tests were made on small samples taken from testpieces, which had received various extensions in monotonic and two-stage stretching. The 310 alloy gave no evidence of transformation to α' martensite within the range of strains applied in the mechanical tests. The proportions of austenite transformed to α' in the 304 alloy were estimated using the relationship with magnetic saturation intensities given by Hoselitz.[12] In monotonic stretching of the 304 alloy, under uniaxial tension in the RD, transformation of γ to α' increased almost linearly with increasing strain in the range of 0.1 to 0.5. However, in second-stage stretching at 45 and 90 deg to the prestrain direction, the rate of transformation was reduced. For example, after prestraining to ~0.24 in the RD, second-stage stretching to a total strain of ~0.45 gave ~23 pct transformation in a 0 deg (monotonic) test, ~20 pct transformation in a 90 deg test, and ~16 pct transformation in a 45 deg test.

B. *Mechanical Testing*

For prestraining in uniaxial tension, a large tensile testpiece was used with a parallel length of 350 mm and a width of 112 mm. Unless otherwise stated, extension was in the rolling direction at a constant crosshead speed, which gave an initial strain rate of $\sim 1.5 \times 10^{-4}$ s^{-1}. Prestraining of flat sheets in equibiaxial tension was carried out at a similar effective strain rate over a cylindrical recessed flat-faced punch 152 mm in diameter. The tensile testpieces used in the second-stage deformations had a parallel length of 40 mm and a width of 8.5 mm. Several such testpieces were machined from the uniaxially prestrained sheets at 0, 45, and 90 deg to the direction of maximum extension and from the equibiaxially stretched sheets at 0, 45, and 90 deg to the RD. Similar testpieces were cut from sheets which had not been prestrained. The prestrained sheets were stored at ~15 °C before and after machining, and care was taken to keep the testpieces cool during the cutting and machining operations, which took about three hours. The testpieces were marked with a 25-mm gage length for measurements of total elongation and were extended to fracture using an initial strain rate of 2×10^{-4} s^{-1}. An Instron strain gage extensometer with a 10-mm gage length was used to follow the extension, and the load-extension relationship was recorded directly. A BBC model B microcomputer was interfaced with the tensile testing machine. From the load-extension measurements, the computer's software was designed to give both numerical and graphical outputs of true stress, σ, true strain, ε, and the relationship of $d\sigma/d\varepsilon$ with ε.

The input signals were converted and stored in the form of true stress and true strain values. For further processing of the data, the selected total strain range was divided into 50 intervals. These small strain intervals were either all of the same magnitude, or, more usually, the first 25 were equally spaced within the first third of the total strain range and the second 25 were equally spaced within the last two-thirds of this strain range. The average slope of the σ/ε relationship was derived within each of these small strain intervals using a regression coefficient calculated from the distribution of σ and ε values collected over the time intervals corresponding to the individual strain intervals. Finally, an optimal smoothing routine was applied before plotting the derived $d\sigma/d\varepsilon$ relationships. These $d\sigma/d\varepsilon$ relationships were calculated from total strains, but the plots illustrated in the next section are not concerned with the initial steeply rising parts of the σ/ε relationships. For the strain intervals considered, the changes in elastic components were usually negligibly small compared with the changes in plastic components. However, the parts of the σ/ε curves corresponding to the onset of general yielding after a change in strain path were often "rounded." To give an indication of the effects of this in the different tests, small crosses are plotted on the $d\sigma/d\varepsilon$ curves at the stage of deformation (when it occurred within the range of plotted results), where ~10 pct of the total test strain was still elastic. For applied strains in excess of ~0.015, the derived $d\sigma/d\varepsilon$ relationships were shown to agree with values calculated directly from precise load/extension measurements within the limits of accuracy of the latter. With the exception of results with 70-30 brass, which were affected by discontinuous plastic flow, reproducibility of the forms of the transient reductions in work-hardening rate was good. This is illustrated by representative examples of duplicate tests in Figure 2.

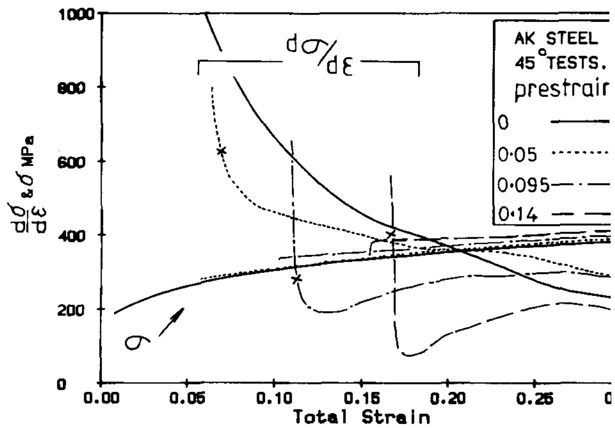
III. RESULTS

Representative results are shown graphically in this section in the form of plots of the relationships among the total applied effective strain and the flow stress, σ, and the work-hardening rate, $d\sigma/d\varepsilon$. These results were derived from the second-stage uniaxial tests following the procedures described in the previous section. Most of the results refer to tests in which prestraining was in uniaxial tension in the RD. In the case of equibiaxial prestraining, the effective prestrain was calculated using the von Mises relationship without allowance for the effects of plastic anisotropy. Each diagram refers to a set of tests made in a particular direction in the sheet plane. In the case of uniaxially prestrained sheets, the tests were either at 45 or 90 deg to the direction of extension in the prestrain. In each diagram, a curve showing behavior in monotonic deformation (zero prestrain) is included for comparison. The $d\sigma/d\varepsilon$ relationships and σ/ε curves are reproduced on the same scale, so that the strains at which $d\sigma/d\varepsilon$ falls to the current value of σ may be identified from the diagrams. Diffuse strain localization is assumed to start at this stage of the test. When $d\sigma/d\varepsilon < \sigma$, ε values measured beyond this stage refer to the average strain measured over the 10 mm gage length sampled by the extensometer.

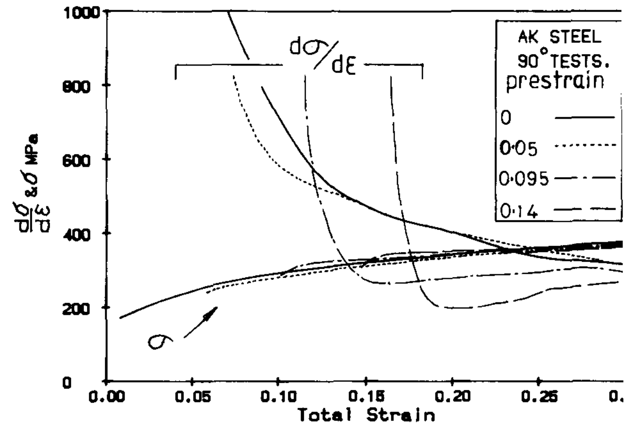
A. *Low-Carbon Ferritic AK Steel*

Figures 1(a) and (b) refer to tests on the AK steel made at 45 and 90 deg to the prestrain direction after prestraining in uniaxial tension in the RD. Relative to the flow strengths attained at equivalent strains in monotonic deformation, with prestrains in excess of ~0.1, the changes in strain path were accompanied by substantial increases in the initial flow stress. Such increases were more prominent in the 45 deg tests. The $d\sigma/d\varepsilon$ relationships show that after the changes in strain path, transient reductions in work hardening were clearly developed after 0.05 prestrain, but with this prestrain, the recovery in work hardening was sufficient to prevent any reduction in the residual uniform elongation, ε_u. The maximum amplitude of the transient increased with increasing prestrain. Prestrains of ~0.1 or more were sufficient to cause severe reductions in the residual ε_u, as defined by the first departure from stable elongation.

Tests were made on the AK steel after prestraining in



(a)



(b)

Fig. 1—σ and $d\sigma/d\varepsilon$ as a function of total effective strain for / steel prestrained in uniaxial tension in the RD. Tested after (*a*) a deg change in strain path and (*b*) a 90 deg change in strain path.

plane strain and equibiaxial stretching. As with the te: using prestraining in uniaxial tension, the results of the tests were consistent with those described by Hutchins‹ and Davis.[13] After prestraining in equibiaxial stretchi: and testing in uniaxial tension at 0, 45, and 90 deg the RD, the increases in the initial flow stress were si ilar to or slightly higher than those developed in the 45 d tests made after uniaxial prestraining. On the whole, aft equibiaxial stretching, the amplitudes of the transient r ductions in work hardening were slightly more seve than those developed in the 45 deg tests (Figure 1(a and distinctly more severe than those developed in tl 90 deg tests (Figure 1(b)) after prestraining in uniaxi tension. To investigate some effects of planar anisotro in the AK steels, tests were also made after prestraini in uniaxial tension at 45 deg to the RD. In these test the differences in second-stage tests made at 45 and 90 d to the prestrain direction (90 and 45 deg to the RD, r spectively) were small. More detailed reference to tl effects of differences in the mode of prestraining on tl behavior of the AK steel is made in the next section.

B. *Commercial Purity Aluminum*

Results for tests on the three batches of AA 1200 sheet A1, A2, and A3, with prestraining in uniaxial tensic

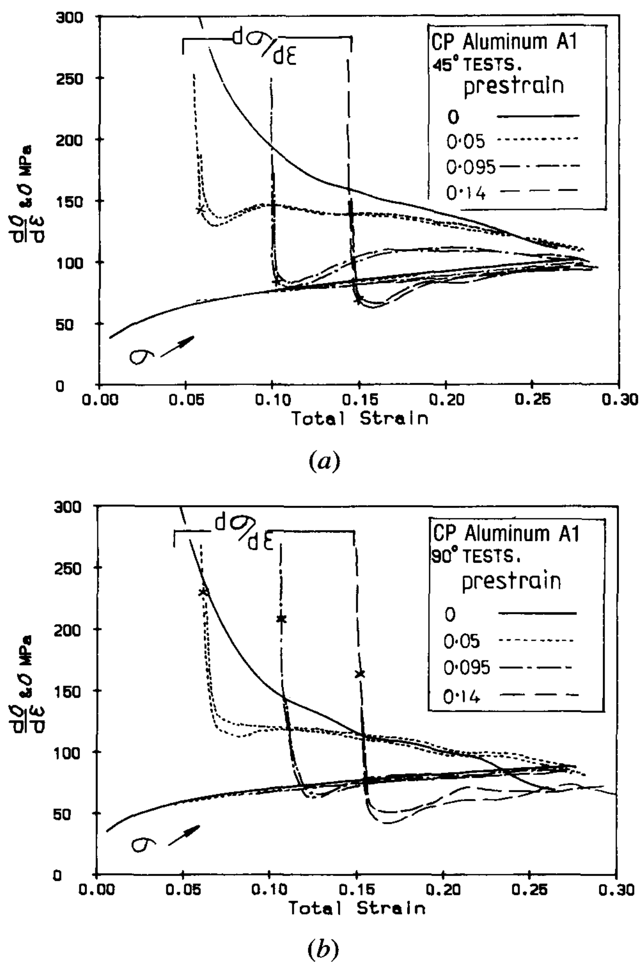


Fig. 2 — σ and $d\sigma/d\varepsilon$ as a function of total effective strain for CP aluminum A1 prestrained in uniaxial tension in the RD and then tested after (*a*) a 45 deg change in strain path and (*b*) a 90 deg change in strain path.

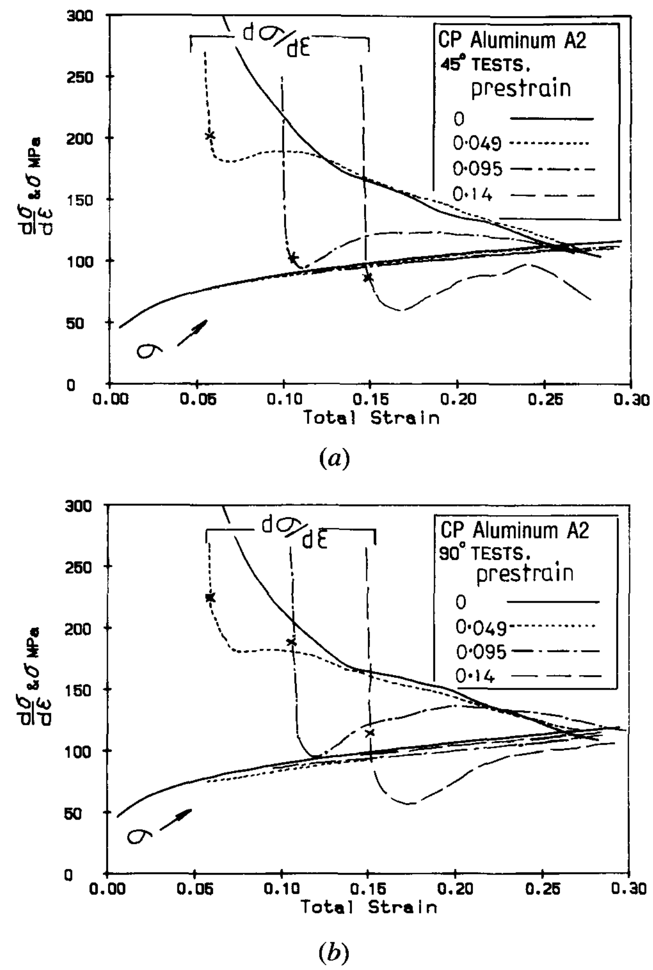


Fig. 3 — Plastic behavior of aluminum A2 after prestraining in uniaxial tension. Tested after (*a*) a 45 deg change in strain path and (*b*) a 90 deg change in strain path.

are given in Figures 2, 3, and 4. The increases in initial flow stress were small compared with those developed in the AK steel. However, judged as a proportion of the work-hardening rate at an equivalent total strain in monotonic extension, the amplitudes of the transients in work hardening in A1 and A2 were rather similar to those developed in the AK steel after comparable uniaxial prestrains.

The sheets of A1 and A3 had significant planar anisotropy, with work hardening better sustained in tests at 45 deg to the RD. This anisotropy was evidently associated with the relatively strong cube components in the textures of the A1 and A3 sheets. A2, which had a more balanced sheet texture, developed very little planar anisotropy of work hardening in monotonic tests. With the A1 sheets, the amplitudes of the work-hardening transients were larger in the 45 deg tests than those in comparable 90 deg tests, but despite this difference, the minimum prestrain required to give a severe reduction in $\bar{\varepsilon}_u$ was slightly higher in the 45 deg tests than that in the 90 deg tests. This behavior is associated with the superior basic work-hardening rate in stretching A1 at 45 deg to the RD. With A2, the amplitudes of the transients developed in 45 and 90 deg tests and the critical prestrains required to cause a severe reduction in $\bar{\varepsilon}_u$ were closely similar. Figure 5 illustrates the behavior of A1 sheets after equibiaxial stretching. The most striking feature of these results, with respect to the magnitudes of the transients and the influence of the planar anisotropy of work hardening on the stability of tensile extension, is their similarity to the results given by tests made at 45 and 90 deg to the RD after uniaxial prestraining in the RD (Figure 2).

After annealing at 350 °C, the initial flow strengths of the recovery-annealed A3 were higher, and particularly in tests at 0 and 90 deg to the RD, the work-hardening rates and ε_u values were considerably lower than those given by the recrystallized A1 and A2 sheets (Figure 4). However, the amplitudes of the transient reductions in work hardening developed in two-stage tests on A3 were much smaller than those in corresponding tests on A1 and A2. Nevertheless, reductions in residual ε_u in the two-stage tests on A3 annealed at 350 °C were quite severe. In the case of the 90° tests, a prestrain of about 0.05 was sufficient to reduce the residual ε_u to ~0.04. Annealing at 400 °C for 15 minutes gave a useful improvement in the work-hardening characteristics of the A3 sheets, while the amplitudes of the work-hardening transients in two-stage tests remained relatively small.

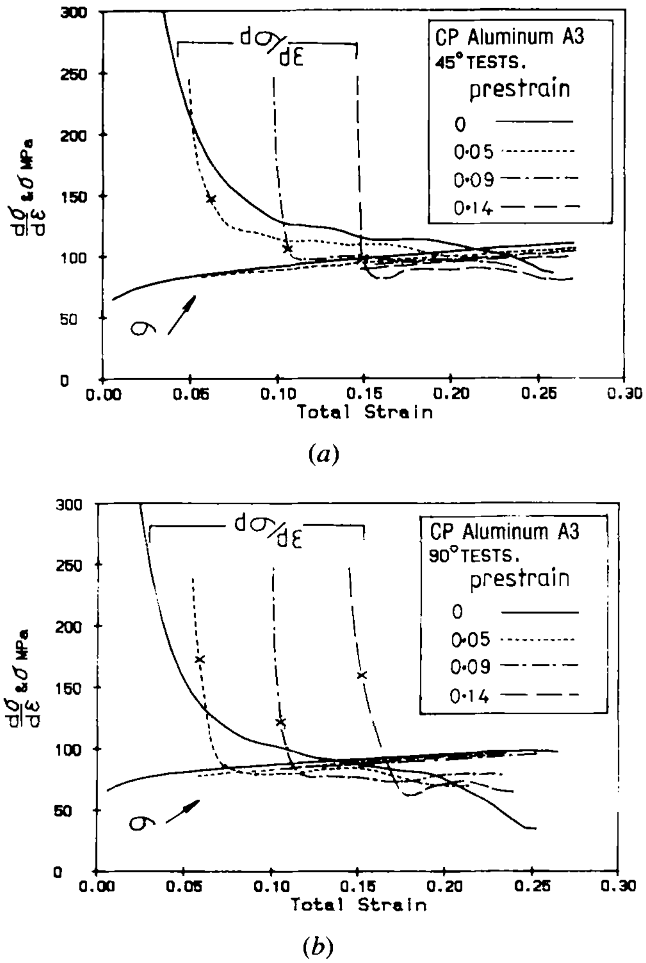


Fig. 4—Plastic behavior of aluminum A3 after prestraining in uniaxial tension. Tested after (*a*) a 45 deg change in strain path and (*b*) a 90 deg change in strain path.

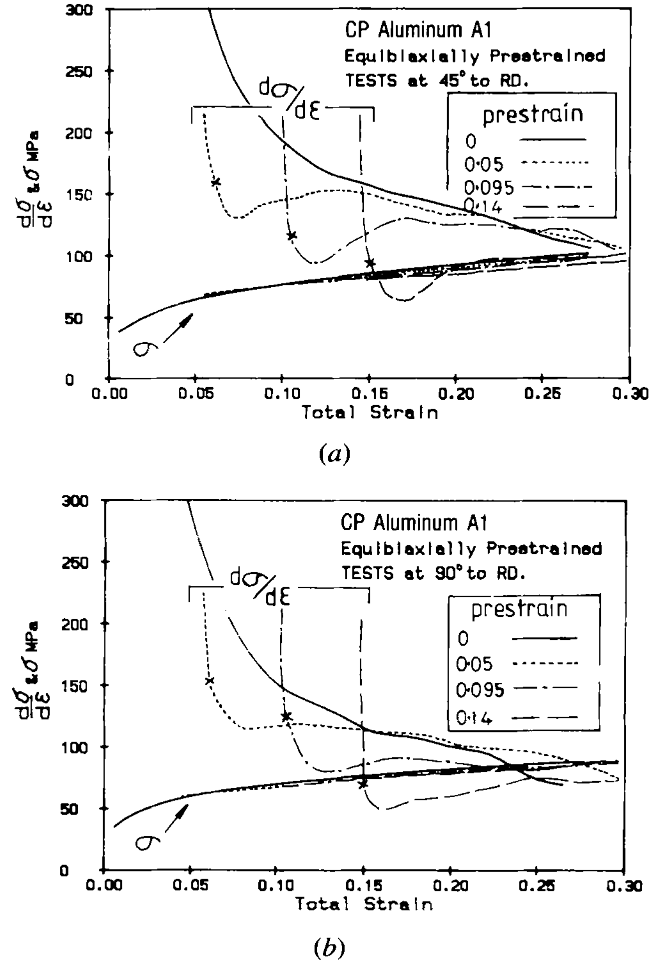


Fig. 5—Plastic behavior of aluminum A1 after prestraining in equibiaxial tension. Tested in uniaxial tension at (*a*) 45 deg to the RD and (*b*) 90 deg to the RD.

C. *Copper*

Results for tests on the OFHC copper annealed at 400 °C and prestrained in uniaxial tension are given in Figure 6. Planar anisotropy of work hardening was slight (Table II), and the transient reductions in work hardening developed in two-stage tests, with the second extension at 45 and 90 deg to the prestrain, were not widely different. Prestrains in excess of ~0.18 were required to give significant reductions in the residual ε_u. Although the work-hardening transients were well developed in two-stage tests after a prestrain of ~0.09, the minimum hardening rates in second-stage stretching after this prestrain were well above those required to cause tensile instability. In continued second-stage deformation after prestrains of less than ~0.15, the hardening rates eventually recovered to values slightly in excess of those at corresponding strains in monotonic deformation, so that total $\bar{\varepsilon}_u$ was slightly enhanced by the change in strain path. This behavior after small prestrains was clearly established with the copper. Similar effects of the same kind occurred in the tests on other materials, but in most cases, the improvements in total $\bar{\varepsilon}_u$ were too small to emerge clearly from the experimental scatter.

D. *Austenitic Steels*

The austenitic steels were prestrained in uniaxial tension with extension in the RD. Results for the stable 310 steel are shown in Figure 7. Although the sheets had a crystallographic texture which was further developed in stretching, there was no significant difference between the work-hardening rates and limiting ε_u values measured in monotonic stretching at 45 and 90 deg to the RD. However, in the two-stage tests, work-hardening behaviors after the 45 deg changes in strain path were distinctly different from those after the 90 deg changes. In both forms of the test, significant reductions in the work-hardening rate developed in the two-stage tests when the prestrain exceeded ~0.1, but the amplitudes of these changes, and the consequent reductions in the residual ε_u after the higher prestrains, were larger in the 45 deg tests than those in the 90 deg tests. In comparison with the behaviors of aluminum and the AK steel, considerably larger prestrains were required in the austenitic steels in order to develop work-hardening transients of significant magnitude, but when such transients were developed, the rate of recovery in work hardening in continued second-stage straining was more gradual in the fcc steels

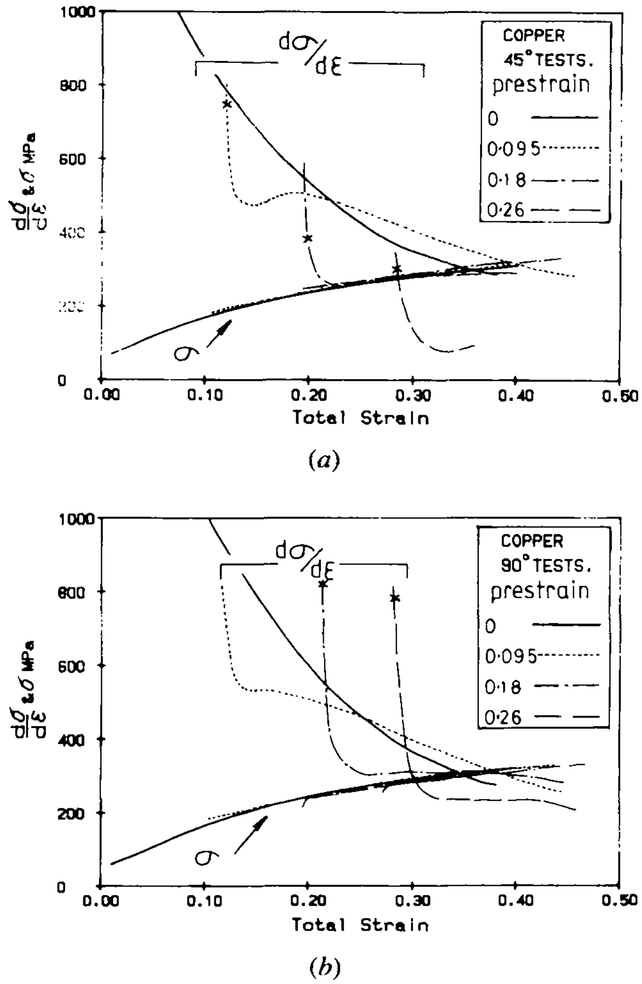


Fig. 6—OFHC copper prestrained in uniaxial tension. Tested after (*a*) a 45 deg change in strain path and (*b*) a 90 deg change in strain path.

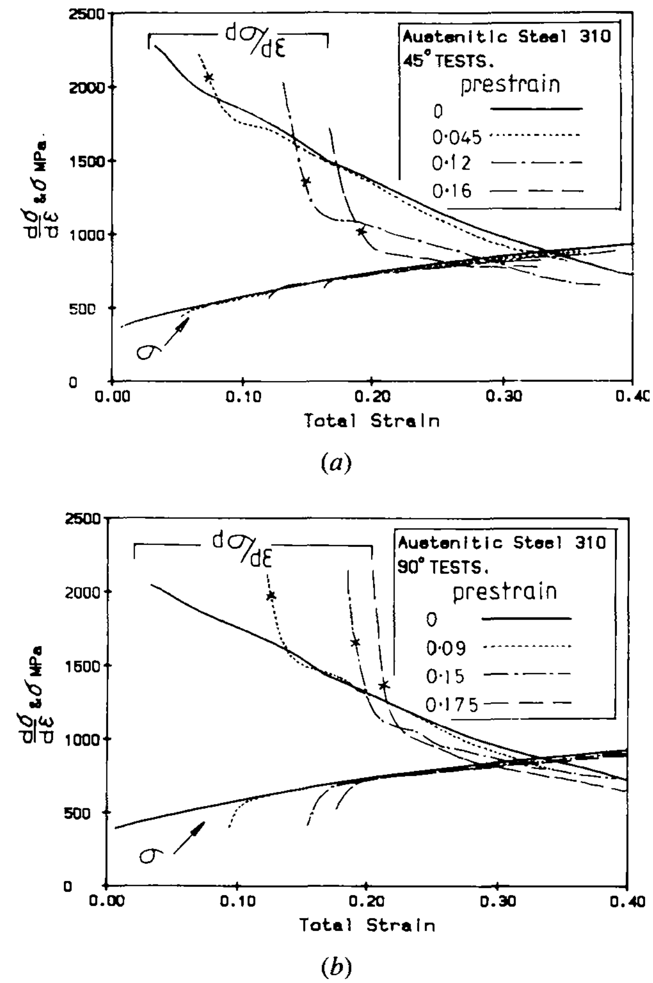


Fig. 7—310 austenitic steel prestrained in uniaxial tension. Tested after (*a*) a 45 deg change in strain path and (*b*) a 90 deg change in strain path.

than in the high SFE materials. Also, in tests made after prestraining in uniaxial tension in the RD, the magnitudes of the transients were more sensitive to the directional relationship between the prestrain and the test strain in the austenitic steels than they were in the materials of higher SFE.

Figure 8 shows results for the 304 austenitic steel in tests with uniaxial prestraining in the RD. With this alloy, the transformation of γ to α' during room temperature deformation caused work hardening to be well sustained, and ε_u in monotonic stretching was in the neighborhood of 0.56. Prestrains in excess of ~0.15 were required to give significant reductions in work hardening in the two-stage tests, but after relatively high prestrains of ~0.3, the reductions in hardening were quite severe, particularly in the 45 deg tests, in which a prestrain of ~0.31 reduced the total $\bar{\varepsilon}_u$ to ~0.36. As with the 310 alloy, recovery of the hardening rate with continued second-stage straining was quite gradual. However, with the 304 alloy, after the changes in strain path, reductions in the rate of transformation to α' (noted in Section II–A) may have contributed significantly to the observed reductions in work-hardening rate.

E. *70-30 Brass*

The first set of tests on the 70-30 brass were made after annealing at 550 °C, which gave an average grain size of ~30 μm. After prestraining in uniaxial tension in the RD with prestrains in the range of 0.05 to 0.18, second-stage extensions at 45 and 90 deg to the RD caused significant changes in the σ/ε relationships relative to those developed in monotonic stretching. In the 90 deg tests, the change mainly involved a lowering of the σ/ε relationship, whereas in the 45 deg tests, there was a more obvious lowering of the work-hardening rate over the first part of the σ/ε relationship (Figure 9(a)). Similar changes of these kinds are illustrated in Figure 10. When the prestrain was limited to 0.18, these changes in plastic behavior were not accompanied by a reduction in $\bar{\varepsilon}_u$. On the contrary, after moderate prestrains, the changes in strain path were generally accompanied by small increases in total $\bar{\varepsilon}_u$. Tests were made on the finer grained 70-30 brass after equibiaxial prestraining within a range of effective strains up to 0.2. σ/ε and $d\sigma/d\varepsilon$ relationships measured in the second-stage uniaxial tests on the equibiaxially stretched sheets at 90 deg to the RD

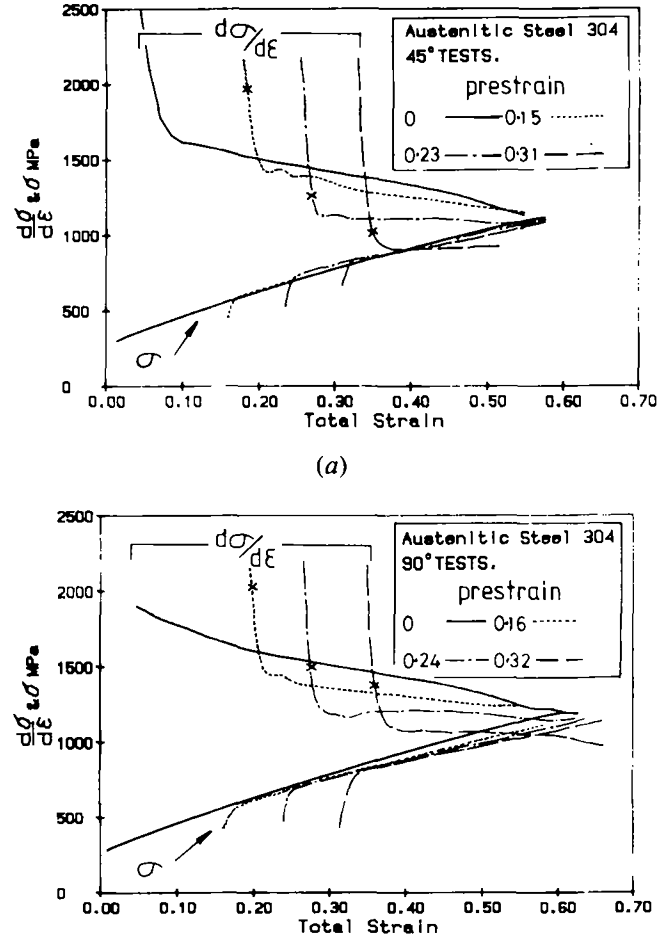


Fig. 8—304 austenitic steel prestrained in uniaxial tension. Tested after (*a*) a 45 deg change in strain path and (*b*) a 90 deg change in strain path.

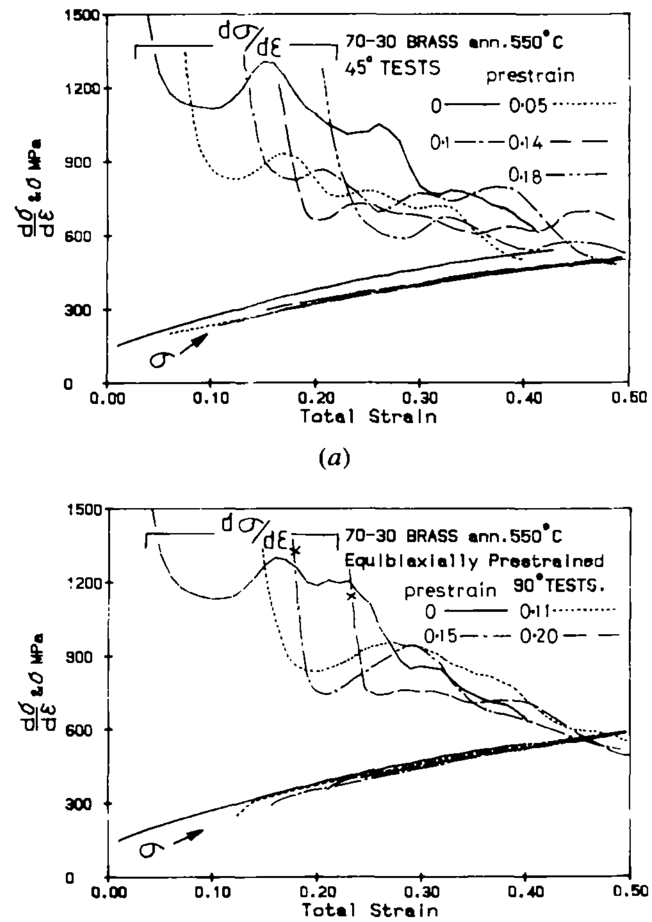


Fig. 9—Plastic behavior of fine-grained 70-30 brass (*a*) in tests at 45 deg to the RD after prestraining at 0 deg to the RD in uniaxial tension and (*b*) tested at 90 deg to the RD after prestraining in equibiaxial tension.

are shown in Figure 9(b). Tests made at 0 and 45 deg to the RD were closely similar to those shown in Figure 9(b). The changes in the σ/ε relationship caused by the changes in strain path were smaller than those developed in the 45 and 90 deg tests made on uniaxially stretched sheets. Comparison of the behavior after uniaxial and equibiaxial prestraining of the 70-30 brass, which had a weak crystallographic texture in the "as-annealed" condition, suggests that the sheet textures influencing directionality in the responses to changes in strain path were developed during prestraining.

Plastic deformation of 70-30 brass was complicated by discontinuous flow of type A, which caused long wavelength fluctuations in the $d\sigma/d\varepsilon$ relationships which were not much reduced by the curve-smoothing procedure. This severely obscured interpretation of the overall trends in the $d\sigma/d\varepsilon$ relationships (Figure 9). The amplitudes of the fluctuations in flow stress caused by dynamic strain aging in the 70-30 brass were reduced by annealing at 690 °C to give a grain size of ~125 μm. Figure 10 summarizes results for 90 and 45 deg tests made on this coarse-grained brass after uniaxial prestraining. The prestrains, in the range of ~0.2 to 0.4, were all higher than those used in tests on the finer grained sheets. Qualitatively, the changes in the σ/ε relationships in the two directions of the test were similar to those observed in the corresponding tests in the finer grained condition, but the reduction in the amplitude of jerky flow allowed a clearer definition of the overall trends in the $d\sigma/d\varepsilon$ relationships. These confirmed that 70-30 brass is susceptible to significant transient reductions in work hardening when the second stage extension is at 45 deg to the direction of the (uniaxial) prestrain but not so susceptible when the second stage extension is at 90 deg to the prestrain. In the 45 deg tests, after prestrains up to ~0.3, $d\sigma/d\varepsilon$ recovered sufficiently in continued extension to prevent significant reductions in the residual ε_u. Such reductions occurred only when the prestrain approached 0.4.

IV. DISCUSSION

The change in initial flow stress, relative to that developed in continued monotonic deformation which occurs at the onset of macroscopic plastic deformation after a change in strain path, will be termed $\Delta\sigma_i$. $\Delta\sigma_i$ can be positive or negative, depending on the material and the change in strain path. As noted in Section I, early discussions of the effects of changes in strain path have

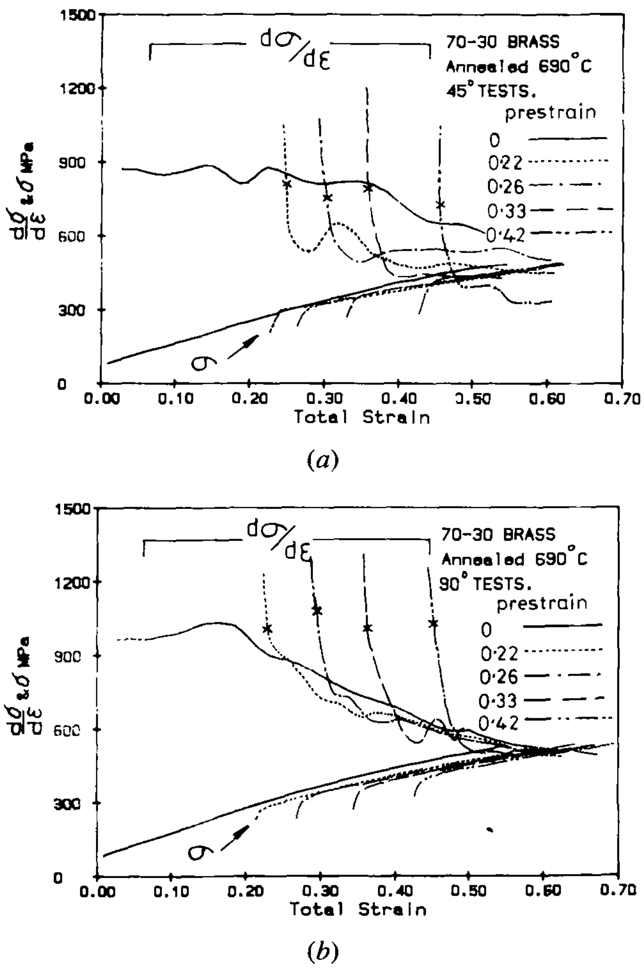


Fig. 10—Coarse-grained 70-30 brass prestrained in uniaxial tension. Tested after (*a*) a 45 deg change in strain path and (*b*) a 90 deg change in strain path.

often emphasized correlations between $\Delta\sigma_i$ and the subsequent changes in work-hardening rate. However, as Wagoner and Laukonis[3] and Hutchinson and Davis[13] have recognized, for materials which develop transient reductions in work hardening, there is no simple general relationship between the magnitude of $\Delta\sigma_i$ and the amplitude of the transient. Moreover, present results have confirmed that the occurrence of a negative $\Delta\sigma_i$ does not preclude the development of a transient reduction in $d\sigma/d\varepsilon$ within an interval of strains beyond the earliest stages of plastic deformation in the second mode. Clearly, the mechanisms involved in positive and negative changes in hardening rates are different. They will be considered separately here.

A. *Changes in Initial Flow Stress*

Large positive values of $\Delta\sigma_i$ (*e.g.*, >10 pct) were observed only in tests on the ferritic low-carbon steel. Relatively small positive values (*e.g.*, up to 3 pct) occurred in CP aluminum, but with the fcc alloys of lower SFE, $\Delta\sigma_i$ was usually negative.

A comparison of $\Delta\sigma_i$ values in tests on the AK steel at 45 and 90 deg to the RD, after a prestrain of 0.14 in uniaxial tension in the RD, was made after allowing for the effect of the greater initial "rounding" of the plastic part of the σ/ε curves for the 90 deg tests. This indicated that the increase in flow stress was ~14 pct in the 45 deg tests and ~7 pct in the 90 deg tests, in agreement with similar measurements made by Raphanel et al.[14] However, our supplementary tests made after uniaxial prestraining at 45 deg to the RD gave a different result. After 0.15 prestrain, $\Delta\sigma_i$ values measured at 45 and 90 deg to the prestrain direction corresponded to increases of ~8 and 7 pct, respectively. After equibiaxial stretching to an effective prestrain of 0.15 before second-stage testing in uniaxial tension, $\Delta\sigma_i$ values were relatively large. In tests at 0, 45, and 90 deg to the RD, the increases were approximately 16, 14.5, and 18 pct, respectively. Considered together, the results of these different forms of the test imply that the $\Delta\sigma_i$ values developed in a commercially processed AK steel depend on both the directional characteristics of the change in strain path and the plastic anisotropy of the steel.

The relatively large positive values of $\Delta\sigma_i$ developed in noncoaxial stretching tests on ferritic low-carbon steels have been referred to as a form of latent hardening.[13,14] For the case of prestraining and testing an AK steel in uniaxial tension, Raphanel *et al.*[14] have shown that predictions of the effects of differences in the direction of the second-stage test on $\Delta\sigma_i$ can be based on the Taylor model of polycrystalline plasticity, if it is assumed that the critical resolved shear stress (CRSS) of the slip systems which remain active after the change in strain path are unchanged, but the CRSS of new slip systems, not previously active, are increased by about 10 pct. Thus, the latent hardening concept appears useful in describing the behavior of annealed low-carbon ferritic steels. In the case of fcc metals and alloys, there is no evident relationship between the $\Delta\sigma_i$ values observed in two-stage stretching and measurements of latent hardening made in single crystals. However, there is strong evidence that differences in the $\Delta\sigma_i$ values developed in two-stage stretching tests on polycrystalline fcc metals and alloys can be dominated by the effects of internal stresses generated in the prestrain.

B. *Effects of Directional Internal Stresses*

During plastic deformation of a mechanically inhomogeneous aggregate, a system of internal stresses is generated, which has directional components reflecting the imposed shape change. On unloading after plastic deformation, the disparities between the local stresses generated in the harder and softer regions give rise to a system of residual stresses. In the direction of maximum strain, the long-range directional stresses in the harder regions are of the same sign as the previously applied stress, and those in the softer regions are of the opposite sign.[15–18] When plastic deformation is resumed under the same system of applied stresses as was used in the first deformation, these directional internal stresses promote a rather homogeneous onset of macroscopic plastic deformation in the second stage of straining. In contrast, if the stress system applied in the second stage of straining is sufficiently different from that in the prestrain, then the residual stresses in the softer regions can combine with the new applied stress to cause plastic deformation to start in the softer regions at a relatively low applied stress.[19]

In plastic deformation after a change in strain path, the directional components of the original residual internal stress system are changed progressively until a reoriented system of internal stresses, which is stable in the second mode of deformation, is built up. This process of reorientation of long-range internal stresses makes a positive contribution to the work-hardening rate in the early stages of plastic deformation after a change in strain path which does not operate in monotonic deformation through an equivalent interval of effective strains. The extent to which the initial flow stress is reduced and the initial rate of hardening is increased after a change in strain path depends on the directional characteristics of the change and the magnitude of the directional component of the original internal stress system. In two-phase alloys containing significant volume fractions of hard second-phase particles, the directional internal stresses generated during moderate prestrains are usually much larger than those developed in single-phase cubic metals and alloys, and the enhancements in work hardening after a large change in strain path are correspondingly larger in dispersion-hardened alloys than those developed in a similar deformation sequence applied to single-phase materials.[16,20-22] However, among single-phase fcc metals and alloys, differences in the ease of cross-slip associated with differences in SFE can cause significant differences in the magnitude of directional internal stresses generated in a prestrain. Evidence of the effects of such differences has been recognized in Bauschinger tests.

In the present cross-tensile tests on 70-30 brass and the two austenitic steels, after a 90 deg change in strain path following moderate prestrains in uniaxial tension, $d\sigma/d\varepsilon$ remained higher than that at equivalent strains in continued monotonic deformation over the first 0.035 to 0.045 of plastic strain in the second mode of deformation (Figures 7 through 10). The corresponding interval of enhanced work hardening in the CP aluminum was only ~0.01 (Figures 2 through 5). However, it is expected that the rate of decline in $d\sigma/d\varepsilon$ in the aluminum was enhanced by relatively rapid development of the work-softening reaction. With respect to the effect of the directional characteristics of the change in deformation mode, it is notable that after uniaxial prestraining, the initial "rounding" of the part of the σ/ε curves corresponding to the start of macroscopic plastic deformation was almost invariably greater in the 90 deg tests than that in the 45 deg tests.

The maximum positive contribution to $d\sigma/d\varepsilon$ from internal stress reorientation occurs at the start of plastic deformation after the change in strain path, but in single-phase cubic metals, the positive contribution from this cause remains significant over a very limited range of second-stage strains. In a material which is susceptible to the work-softening reaction, on the other hand, the negative contribution to $d\sigma/d\varepsilon$ remains significant over a wider range of second strains. Because of this difference in the strain dependencies of the positive and negative transients, it is expected that in single-phase materials which are susceptible to both, the negative change in $d\sigma/d\varepsilon$ will usually become dominant within a range of strains beyond the earliest stages of plastic deformation after the change in strain path.

C. *Susceptibility to Premature Strain Localization*

In all the tests described in this paper, second-stage stretching was in uniaxial tension. It is assumed that diffuse strain localization would be initiated in uniaxial tension when $d\sigma/d\varepsilon \cdot 1/\sigma$ falls below unity, and that the total effective strain at which this occurs can be identified as the limit of effective uniform elongation, $\bar{\varepsilon}_u$. This criterion allows direct comparison with estimates of $\bar{\varepsilon}_u$ made in previous investigations, which are based on the maximum load criterion, and it is useful in making comparisons of the influences of material variables on susceptibility to premature tensile instability after changes in strain path. However, when the reduction in $d\sigma/d\varepsilon$ is transient, the evolution of diffuse localization toward failure may be interrupted. This is considered in the next section.

Figure 11 illustrates the effects of the magnitude of prestrain in uniaxial tension on $d\sigma/d\varepsilon \cdot 1/\sigma$, measured at equivalent second-stage strains in monotonic deformation and after a 45 deg change in strain path, for aluminum A2, copper, and the 304 austenitic steel. In this example, $d\sigma/d\varepsilon \cdot 1/\sigma$ values were measured at the test strain, which gave the *maximum* difference in hardening rates. Thus, the difference in heights of the two curves referring to a particular material is a measure of Δ_{max} $(d\sigma/d\varepsilon \cdot 1/\sigma)$, the maximum amplitude of the reduction in $d\sigma/d\varepsilon \cdot 1/\sigma$, which was developed after the indicated values of prestrain. With copper and the 304 alloy, considerably higher prestrains were required to give Δ_{max} $(d\sigma/d\varepsilon \cdot 1/\sigma)$ values of a magnitude comparable with those developed in the aluminum after prestrains in the range of 0.05 to 0.10, but toward the upper limits of the different ranges of prestrain considered, Δ_{max} $(d\sigma/d\varepsilon \cdot 1/\sigma)$ values in the range of ~0.5 to 1.0 were attained in all

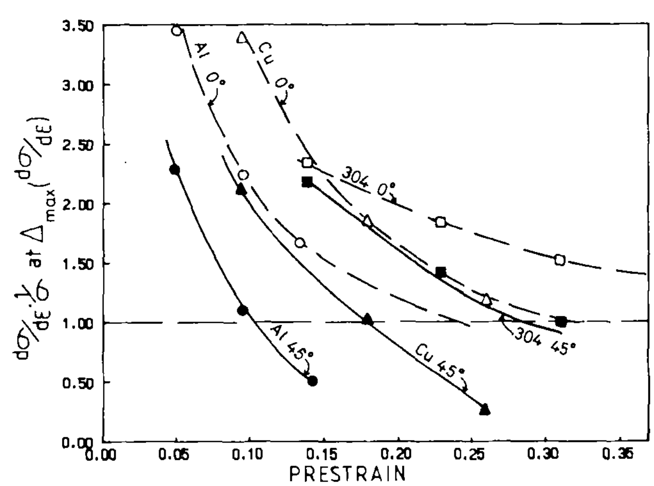


Fig. 11 — Effect of magnitude of prestrain on $d\sigma/d\varepsilon \cdot 1/\sigma$ values measured in continued monotonic deformation (chain lines) and after a 45 deg change in strain path (continuous curves) for aluminum A2, copper, and 304 austenitic steel. The tests were at 45 deg to the RD. $d\sigma/d\varepsilon \cdot 1/\sigma$ values were measured at the second-stage strain which gave the *maximum* difference in work-hardening rates.

three materials. When Δ_{max} $(d\sigma/d\varepsilon \cdot 1/\sigma)$ is confined within this range, it follows that the prestrain must be sufficient to reduce $d\sigma/d\varepsilon \cdot 1/\sigma$ in monotonic deformation to below a value in the range of ~1.5 to 2.0, before the amplitude of the transient can be sufficient to cause the initiation of localization to occur at an early stage of stretching after the change in strain path. This condition was satisfied in the lower SFE alloys only after relatively large prestrains.

The transient reduction in $d\sigma/d\varepsilon$ can be viewed as a relative softening effect of limited amplitude and duration which is superimposed on the work-hardening relationship characteristic of monotonic deformation. The differences between the behaviors of the three batches of CP aluminum illustrated in Figures 2 through 4 can be considered from this viewpoint. Although there were differences in the amplitudes of the transient reductions in $d\sigma/d\varepsilon$ developed in the three batches, their relative susceptibilities to premature strain localization in tests made after changes in strain path were strongly influenced by differences in work-hardening behavior which were evident in monotonic stretching.

Present results tend to confirm that the most important effects of changes in strain path can be explained on the assumption that the changes in the σ/ε relationship are transient. However, strictly speaking, this is an approximation. Results with the austenitic steels and 70-30 brass have shown that large changes in strain path can cause significant "permanent" softening (Figures 7 through 10). This observation reintroduces the question of the extent to which dislocation annihilation, as opposed to rearrangement alone, may contribute to relative softening when the direction of deformation is changed.

For a more comprehensive evaluation of the effects of the changes in $d\sigma/d\varepsilon$ on stretching limits, it is necessary to take account of their strain dependence.

D. *Influence of the Form of Transient Reductions in Work Hardening*

After a change in strain path, in second-stage deformation beyond that which gives the maximum reduction in relative hardening rate, $d\sigma/d\varepsilon$ recovers toward that developed at equivalent strains in monotonic deformation, but the rate of recovery with increasing strain can vary widely in different materials. Present results indicate that this rate of recovery is more rapid in a low-carbon ferritic steel and CP aluminum than in alloys of low SFE. Figure 12 illustrates an example of this difference. In the case of the CP aluminum, when the prestrain exceeds ~0.1, Δ_{max} $(d\sigma/d\varepsilon \cdot 1/\sigma)$ can be sufficient to initiate diffuse strain localization after a second-stage strain of 0.02 to 0.03, but owing to the relatively rapid recovery in $d\sigma/d\varepsilon$, the interval of strains in which stretching remains unstable may be quite narrow (Figures 2 through 4). In this event, the localization may not develop beyond its earliest stages before stable extension is restored, and the growth of the inhomogeneity toward localized necking is interrupted. For the case of an AK steel, the consequences of this kind of behavior are illustrated by the investigation of Laukonis.[11] Within a range of prestrains beyond that which causes the first significant reduction in $\bar{\varepsilon}_u$ (as defined here) after a change in strain path, the residual postuniform elongation is increased. Figure 13 shows that the behavior of CP aluminum is essentially similar in this respect. In such cases, the initial limit of uniform elongation is not necessarily a reliable guide to performance in practical stretch-forming

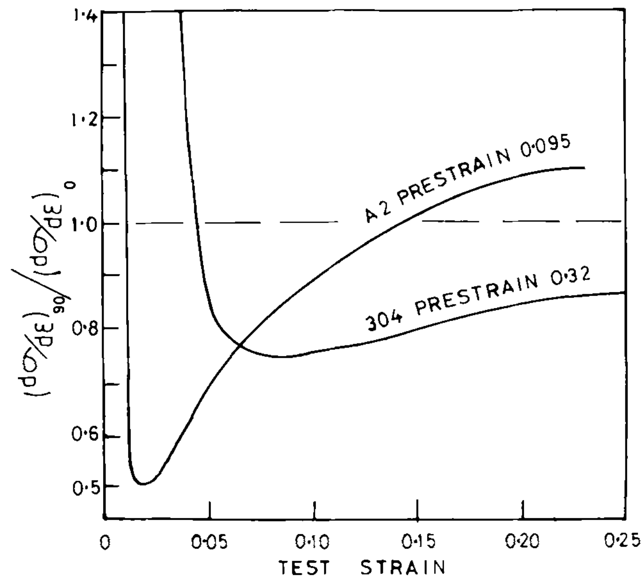


Fig. 12—Forms of work-hardening transients developed in aluminum A2 and 304 austenitic steel after 90 deg changes in strain path, following uniaxial prestrains in the RD of 0.095 and 0.32, respectively. Relative changes in hardening rates are represented in terms of the ratio $(d\sigma/d\varepsilon)_{90}/(d\sigma/d\varepsilon)_0$, where the subscripts 90 and 0 signify the directional relationship between the first- and second-stage extensions.

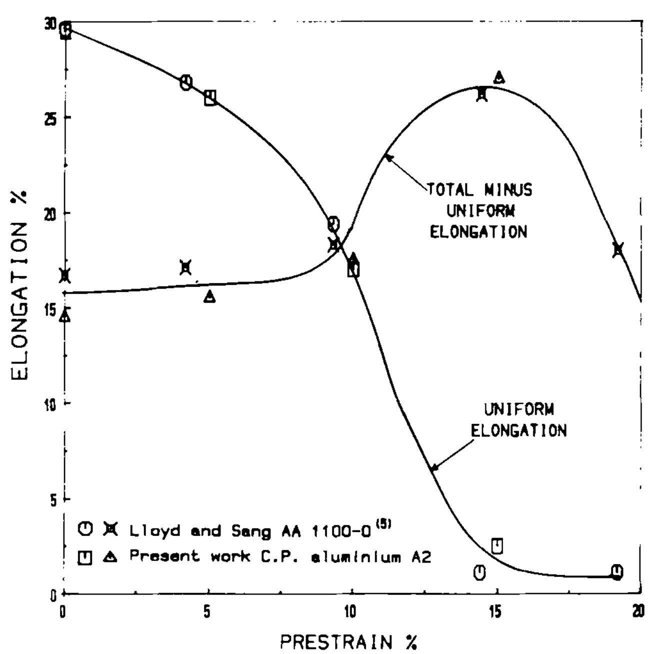


Fig. 13—Effects of prestrain in uniaxial tension on residual uniform elongation and postuniform elongation measured in tests at 90 deg to the prestrain direction for CP aluminum. Present results for aluminum A2 are combined with Lloyd and Sang's measurements on AA 1100.[5]

operations, in which some diffuse localization is acceptable.

V. SUMMARY AND CONCLUSIONS

The following conclusions are based on observations of the work-hardening behaviors of a low-carbon ferritic steel, CP aluminum, copper, 70-30 brass, and 310 and 304 austenitic steels in stretching with an abrupt change in strain path.

1. In the earliest stages of plastic deformation after a change in strain path, the work-hardening rate is often increased relative to that developed at equivalent strains in monotonic deformation. This effect is associated with reorientation of directional internal stresses which are inherited from the prestrain. Among single-phase cubic metals subjected to a given deformation sequence, such increases in the initial hardening rate are developed more strongly in alloys of low SFE.
2. Changes in strain path often cause transient reductions in hardening rate within an interval of strains beyond the earlier stages of plastic deformation in the second mode of deformation. These reductions in hardening rate are believed to be associated with partial dissolution of the dislocation substructures generated in the prestrain. Fcc alloys of low SFE are less susceptible to such reductions in hardening rate than are CP aluminum and low-carbon ferritic steels.
3. Comparisons of different materials in similar two-stage stretching tests indicate that the changes in $d\sigma/d\varepsilon$ can be influenced by numerous material variables other than SFE. These are structural features which influence overall work-hardening behavior, such as grain structure and crystallographic texture, and factors which mainly affect flow stress in the early stages of plastic deformation, such as apparent latent hardening and residual internal stresses.
4. Changes in strain path cause reductions in the limit of uniform elongation in uniaxial tests when the reduction in $d\sigma/d\varepsilon$ is sufficient to cause $d\sigma/d\varepsilon \cdot 1/\sigma$ to fall below unity. Within the range of conditions explored in this investigation, this occurred only when the prestrain was sufficient to reduce $d\sigma/d\varepsilon \cdot 1/\sigma$ in monotonic stretching to below a value in the range of 1.5 to 2.0. In the alloys of low SFE, this condition was satisfied only after prestrains in the range of 0.15 to 0.40, whereas with the ferritic steel and CP aluminum, prestrains in the neighborhood of 0.10 were sufficient.
5. With aluminum and ferritic low-carbon steels, when the second-stage extension is continued beyond that giving the maximum reduction in $d\sigma/d\varepsilon$, the work-hardening rate recovers quite rapidly toward that characteristic of monotonic deformation. With this form of the transient and when the initial reduction in $d\sigma/d\varepsilon$ is not too severe, development of the initial strain inhomogeneity toward localized necking may be interrupted by an interval of stable elongation.

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