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TABLE I. Dimensions and physical characteristics of the specimens.

Sample	Direct current through sample (mA)	Length (mm)	Width (mm)	Thickness (microns)	Hall coefficient (cm^3/C)	Resistivity (Ω-cm)	Hall mobility (cm^2/V sec)
InSb (flash evaporated)	3	7.5	1.2	3.0	220	3.7×10^{-2}	6000
InAs (flash evaporated)	5	7.5	1.1	0.3	5	5×10^{-3}	1000
InSb (three-temperature method)	10	6.5	2.4	1.5	145	1.3×10^{-2}	11 200

effective surface-to-volume ratio affects the level of $1/f$ noise.[10] Hence, the presence of $1/f$ noise in evaporated films of InSb and InAs may be related to their large effective surface-to-volume ratios which far exceed those obtained in specimens made of bulk material. On the other hand, comparison of the noise levels in Fig. 1 with the data in Table I indicates that the effective surface-to-volume ratio cannot completely account for the noise. The generation of $1/f$ noise is known to depend on many other factors such as surface conditions (i.e., inverted layers) or grain boundaries in the material. It is of interest to note, however, that the many possible causes of the noise generation notwithstanding the noise spectra in Fig. 1 do follow exactly the $1/f$ pattern.

The specimens of InSb and InAs fabricated by flash evaporation were prepared by K. K. W. Heid of Beckman Instruments, Inc., while the sample of InSb made by utilizing the three-temperature method was obtained from the Naval Ordnance Laboratory Corona, courtesy of H. H. Wieder.

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Residual Strains in Shock-Loaded Aluminum

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In a previous paper[1] it was reported that apparent large reverse flow stresses were observed upon unloading of aluminum samples which had been shock loaded to stress levels of 100 and 175 kbars. The purpose of this communication is to report some recent measurements of residual strains in aluminum samples which have been exposed to similar loading–unloading histories. These measurements agree with the values predicted by quasi-static, rate-independent, elastic–plastic theory,[2] and thus indicate that the observed large flow stresses are transient phenomena. Some preliminary results of this work have been presented elsewhere.[3]

The zero stress point on the uniaxial strain unloading curve of a shock-loaded aluminum alloy (composition: 93.75% Al, 4.25% Cu, 0.50% Mg, 0.75% Mn, 0.75% Si; tensile yield strength (0.2% offset): 3.95 kbars; Young's modulus: 735 kbars; Vickers hardness: 134) was determined by measuring the residual strain after shock loading. The experiments are similar to those of Hartman.[4] The essential difference is that in the present work explosive loading was used in order to obtain residual shock pressures of 110 and 240 kbars, whereas Hartman's work used a gas gun to obtain pressures below 40 kbars.

The experimental arrangement for producing a 240-kbar shock wave consists of a "mouse-trap" plane wave generator, which causes a 1.5-mm aluminum plate to impact simultaneously on a 3-mm-layer of "Metabel" (supplier: Imperial Chemical Industries) sheet explosive covering a 1.5-mm aluminum flyer plate. The flyer plate is accelerated across a 6-mm gap and strikes the target plate arrangement, which consists of a stack of three plates in intimate contact.

The 110-kbar shock wave is produced in the target by removing the flyer plate and sheet explosive so that the mouse-trap plate strikes the target arrangement directly.

The first target plate is 1.5 mm thick and the other two plates are 7 mm thick. The diameter of the plates is 30 mm. The last two plates both have cylindrical samples 10 mm in diameter pressed snugly into holes bored in their centers. The plane shock wave generated by the impact passes through the three plates and reflects from the free surface. The two cylindrical pellets are accelerated through windows cut in wooden plates and through layers of styrofoam and foam rubber before being stopped in a water tank. They are recovered from the water tank after the shot. The wooden plates serve to retard and deflect the aluminum holder plates.

The dimensions of the aluminum plates are chosen so that the pellets are loaded and unloaded in the direction of the cylinder axis before relief waves from the edges arrive and destroy the one-dimensional nature of the strain. The history of the sample in the middle plate is that it is first shock loaded, and then relieved by means of the rarefaction fan from the rear of the flyer plate. This sample does not experience tension. The pellet in the last plate, on the other hand, undergoes tension due to the intersection of relief waves from the front and the rear of the target arrangement, and this pellet was always found upon recovery to have spalled.

The thickness of the sample from the middle plate is measured before and after each shot. The shock pressures induced in the samples are obtained from auxiliary experiments in which electronic pin switches are used to measure the free-surface velocity of the middle plate in the absence of the last plate. Known Hugoniot equation-of-state data[5] are then used to obtain the pressure and density initially present behind the shock wave. The analysis of the residual strain measurement is identical to that described by Hartman.[4] The strain rates present upon unloading depend on the unknown time history of the unloading stress strain curve. However, the rates can be estimated from rate-independent theory to be of the order of 3×10^{5} sec^{-1}.

The measured residual strains as a function of shock stress are plotted in Fig. 1. Hartman's experimental data and the theoretical curves based on rate-independent elastic–plastic theory are also shown for comparison. The theoretical curve in Fig. 1 was con-

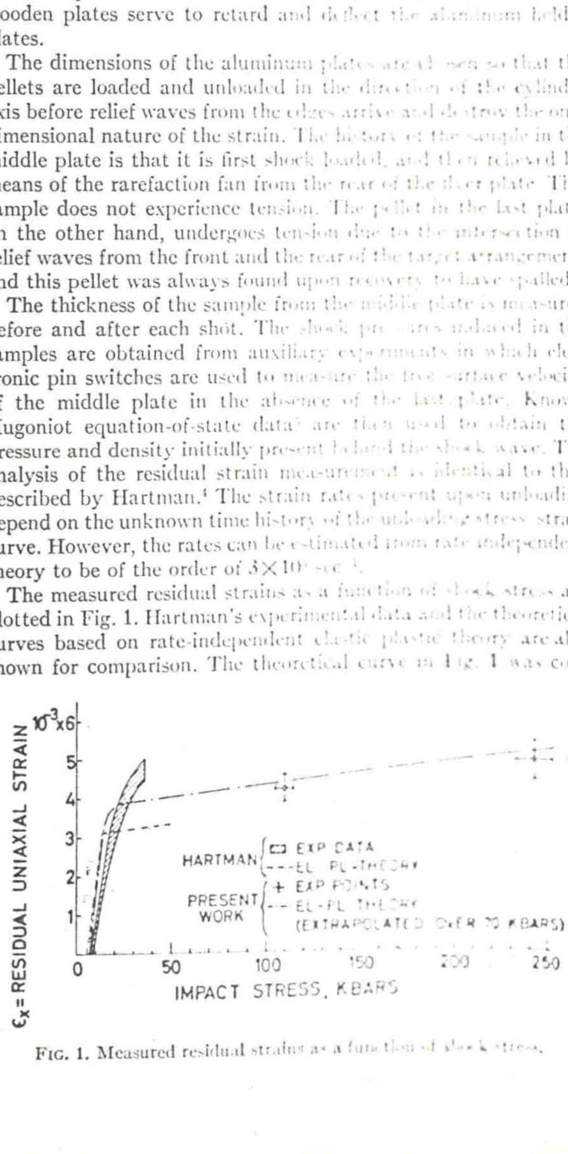

FIG. 1. Measured residual strains as a function of shock stress.

structed in the following way. First, static tensile tests were carried out on specimens cut from the same bar as the pellets. The uniaxial stress curve thus obtained was then translated into a uniaxial strain curve by the method discussed by Fowles.[2] The unloading curves were then constructed from elastic–plastic theory which takes into account the work hardening during the compressive cycle. The residual strain is the intercept of the unloading curve with the zero stress axis. The above analysis process is described in detail by Hartman.[4] It was necessary to linearly extrapolate the theoretical curve in Fig. 1 beyond shock pressures of 70 kbars, since the static tensile data did not extend into this strain region.

It is seen that the measured residual strains are close to the values one would expect from rate-independent theory. This indicates that when aluminum unloads from shock strain levels, the initially large reverse flow stresses reported in Ref. 1 later relax to the rate-independent values. The complete unloading curves might, therefore, be expected to look qualitatively like the loading curves reported by Barker *et al.*,[6] i.e., there should exist an initial elastic stress deviation which relaxes toward the static curve as the strain rate decreases.

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Fault Planes in Steam-Oxidized Silicon

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Stacking faults have been observed by Thomas[1] in silicon after annealing mechanically polished single-crystal slices, using transmission electron microscopy. The origin of these faults has been attributed to surface damage from mechanical polishing. Queisser and van Loom[2] have reported the observation of line patterns (visible with an optical microscope) on {100} surfaces of silicon wafers mechanically polished prior to steam-oxidation and then suitably etched. The authors interpret the line as the intersections of stacking faults with the surface. It has been indicated that the presence of water vapor during the oxidation is essential to the growth of the faults.

In this letter, we report the observation of two-dimensional faults in (111) silicon in which the surface damage was removed before steam-oxidation. Silicon slices (25-mil thick) were cut parallel to (111) from a pulled single crystal of high electrical resistivity. They were lapped and chemically polished to a considerable depth (6 mils on each side) to remove the surface damage. Chemical etching showed less than 5×10^2 etch pits/cm^2. Some of these slices were vacuum-annealed at 1000°C for 45 min. Neither Dash etch nor the electron microscope revealed significant defects. The annealed wafers were then steam-oxidized at 1150°C for 45 min. The oxide layer grown was about $0.7\ \mu$ thick. After removal of the oxide and after application of Dash etch for about 5 min, etch markings became visible, as shown in Fig. 1. Lines bounded on both sides by etch pits are oriented to maintain 60° angles between each other and delineate the ⟨110⟩ directions, and are of rather uniform length. Often, the lines are arranged in long rows that can be followed over several hundred microns. The rows are often parallel to each other and do not seem to lie in particular low-indexed crystallographic directions. The lines have a limited depth; they disappear after being subjected to the Dash etch for $\sim$8 min, removing $\sim 1\frac{1}{2}\ \mu$ of silicon. On wafers of 12-mil thickness (which showed etch markings on both sides), no correspondence of row directions or pit densities on the two sides was detected.

Fig. 1. Dash-etched line faults (taken at ×650, shown here ×455).

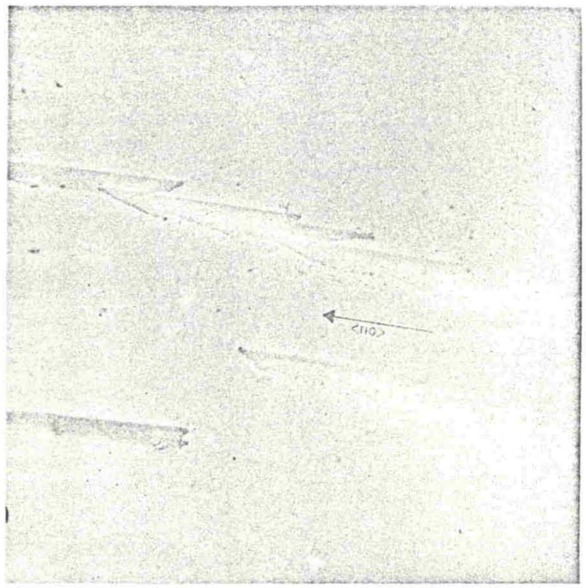

Fig. 2. Electron micrograph of stacking faults; note impurities bounding the faults; taken at ×19 000, shown here ×7600.

Fig. 3. Electron micrograph of ends of stacking faults; taken at ×52 000, shown here ×20 800.