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THE EFFECT OF ONE-DIMENSIONAL PLASTIC STRAIN WAVES IN α-PHASE TITANIUM-ALUMINUM ALLOYS

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Charles J. Bruggemann
(Ph. D. Thesis)

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motion with the onset of twinning occurring at strain levels that increased with alloying. The rate of increase of twin volume with additional plastic strain also increased at higher strain levels. The volume of twins in constrained compression was greater at the same strains than that in tension. The volume of twins in high strain rate one dimensional strain compression were much greater and equivalent to that produced during constrained compression to between 3 and 6 times greater plastic strain. Twinning on \{10\bar{1}2\} dominated deformation structures at low strain rate, followed by \{11\bar{2}2\} with traces of \{11\bar{2}1\}. Kinking was evident after about 1.5% plastic strain. Twin densities were highly nonuniform at low strain rate but more uniform at high strain rate, both overall and within each grain. Twins at high strain rate were narrower and more numerous, indicating that more twin nuclei were activated. Twins on \{10\bar{1}2\} at low strain rate were broad, but at high strain rate \{10\bar{1}2\} twins were as narrow as \{11\bar{2}2\} twins. Twinning on \{11\bar{2}2\} and \{11\bar{2}4\} dominated at high strain rate, followed closely by \{10\bar{1}2\} and \{11\bar{2}1\} twinning. In low strain rate constrained compression, abundant kinking accompanied twinning, however in high strain rate, one dimensional strain kinking was absent.
I. INTRODUCTION

The variation of the mechanical behavior of materials as a function of strain rate has long been a subject of theoretical and practical interest. The dependence of the flow stress on changes in strain rate has been utilized as a means of characterizing the mechanisms responsible for strain hardening and identifying the rate controlling mechanisms during plastic flow of metals. It has also been useful in studying the effect of radiation damage. The rate of decay of elastic stress wave amplitudes with propagation distance has also been used to test theories of dislocation dynamics and dislocation multiplication.

There is additional motivation for interest in the strain rate sensitivity of material properties in many fields. In metal forming variable strain rates are encountered, typically lying in the range of 1 to 100 sec\(^{-1}\). Fracture mechanics must treat both the plastic flow of material in the zone immediately surrounding an advancing crack tip and the rate dependence of the ductile-brittle transition phenomenon. In machining, chip formation proceeds with strain rates lying in the range of \(10^3\) to \(10^5\) sec\(^{-1}\). In applications, such as energy absorbing media and devices, explosive cladding techniques for bonding dissimilar materials, and the developing field of high energy rate forming, high strain rate information on materials is important.

The nature of the lattice defect structure after plastic deformation can also be a function of strain rate. The phenomenon
of shock hardening in many metals and alloys is a result of the high rate of generation of lattice line and point defects and the increased likelihood of stress- and strain-induced phase transformations such as the formation of the $\varepsilon$ phase in iron\textsuperscript{13} or twin production in titanium and other materials.\textsuperscript{14} The increased flow stress exhibited during very high strain rate deformation is not entirely retained during subsequent quasi-static deformation and if the deformation mechanisms active at high strain rate are the same as those active at low strain rate, as is the case for aluminum and some of its alloys\textsuperscript{15} which deform entirely by slip, the subsequent flow stress will be only slightly increased by the prior high strain rate deformation. If however a different microstructure is developed during high strain rate deformation, then appreciable increase of the subsequent quasi-static flow stress is found. Examples of this are when phase transformations or twin morphology and density are a function of strain rate and state of stress.\textsuperscript{16}

Titanium and its $\alpha$-phase aluminum alloys are of interest in this respect because of the large variety of deformation modes that have been found active. Slip occurs in the $\{11\bar{2}0\}$ direction on the (0001), $\{10\bar{1}0\}$ and $\{10\bar{1}1\}$ planes as well as in the $\langle11\bar{2}3\rangle$ direction on the $\{10\bar{1}1\}$\textsuperscript{17} planes. Twinning also occurs on the six habit planes $\{10\bar{1}1\}$,\textsuperscript{18} $\{10\bar{1}2\}$, $\{11\bar{2}1\}$, $\{11\bar{2}2\}$, $\{11\bar{2}3\}$, $\{11\bar{2}4\}$,\textsuperscript{19} and kinking is found,\textsuperscript{20} as would be expected. The volume of twinned material and the operation of the various slip and twinning modes are sensitive to the temperature and strain rate at which deformation occurs.\textsuperscript{18,21-23}
Titanium and $\alpha$-phase titanium-aluminum alloys have been shown to twin profusely and under shock loading$^{24}$ and impulsive loading$^{25}$ conditions twin formation contributes more to the total deformation than at lower strain rate.

A variety of deformation modes are potentially active in $\alpha$-phase titanium alloys. Those modes which are activated, and their effect on the microstructure and the flow stress are a function of strain rate. In addition, changes in deformation modes active at high and low strain rate should produce corresponding changes in the subsequent quasi-static flow stress for specimens prestrained by the same amount at both rates. Therefore, to describe the variation in loading path and microstructure change in $\alpha$-phase titanium-aluminum alloys, tests should be performed in compression at strain rates near $10^6$ sec$^{-1}$ as well as compression and tension tests performed at strain rates near $10^{-3}$ sec$^{-1}$

The difference in loading path arises from the strain rate sensitivity of the titanium and titanium-aluminum alloys. In this study, low strain rate loading can be done by conventional tension testing whereas high strain rate loading is more easily done by the plate impact technique to produce one-dimensional strain, plane wave loading.$^{26}$ The plate impact configuration is attractive because the geometry is simple and the loading paths can be deduced at high strain rate. This technique does not require corrections for lateral inertia effects and no averaging of material behavior through the specimen thickness is required, as is the case in split Hopkinson bar experiments.$^{27}$
In plate impact experiments performed at very high stress levels where the planar plastic wave develops into a shock wave, plastic deformation occurs during both loading and unloading. To insure that negligible plastic deformation occurred on unloading, the stress levels in the present investigation were confined to a few times the yield stress of the materials. Thus changes in microstructure developed during high rate loading would not be obscured by the unloading plastic strain, and dynamic compression heating of the specimens would not be significant.

One consequence of dynamic one-dimensional strain loading is that deformation occurs under the simultaneous influence of the strain rate sensitive flow stress and a large geometrically imposed hydrostatic compressive stress. To demonstrate the influence of large superimposed hydrostatic compressive stresses on the deformation mechanisms at low and high strain rate, additional experiments were devised to test the effect of hydrostatic compressive stresses during deformation.

Three types of loading procedure were devised to provide data on high and low strain rate deformation and to duplicate at low strain rate the superimposed hydrostatic compressive stresses inherent in the high strain rate technique. Low strain rate, simple tension tests were performed at a strain rate near $10^{-3}$ sec$^{-1}$ to provide reference values of the stress-strain behavior for three α-phase titanium-aluminum alloys. The microstructures developed during deformation by this method were also reference conditions.
The high strain rate deformation was produced by the flying plate technique. Here a thin disc of titanium alloy is arranged parallel to a target disc of the same alloy and caused to impact the target disc. Compressive stress waves propagate through the target disc to the target rear surface where a transducer measures the stress increases associated with the elastic wave, which arrives first, and the later arriving plastic wave.

The flying plate technique produces deformation in a state of one-dimensional strain and consequently a large hydrostatic compressive stress component is present. This level of hydrostatic compressive stress was reproduced in a series of low strain rate tests by compressing thin rectangular specimens. In this way, the deformed microstructures produced by low and high strain rate deformation under comparable states of hydrostatic compressive stress could be compared.

The development of deformed microstructures in the α-phase titanium-aluminum alloys during the three types of loading is interesting because these alloys exhibit a variety of deformation modes. Slip has been found in the $(11\overline{2}0)$ directions on the $(0001)$, $(10\overline{1}0)$ and $(10\overline{1}1)$ planes and in the $(11\overline{2}3)$ directions on the $(10\overline{1}1)$ planes. Twinning also occurs on the following six habit planes: $(10\overline{1}1)$, $(10\overline{1}2)$, $(11\overline{2}1)$, $(11\overline{2}2)$, $(11\overline{2}3)$, $(11\overline{2}4)$. Not all deformation modes act concurrently, and the particular mode operating should be influenced by the type of loading. The operating mode of deformation at any given strain rate should be discernable from a comparison of
the stress strain behavior experimentally measured for the three types of loading.
II. EXPERIMENTAL DEVICES AND PREPARATIONS

A. Materials

The titanium alloys used in this investigation were obtained as special melts from Titanium Metals Corporation of America. The alloys were produced by the addition of 99.99% Al to Ti-50A melts. The resulting ingots were hot forged at temperatures just above the α-β transus and hot rolled to 0.20 in. × 4 in. × RL sheet at temperatures just below the α-β transus.

The sheet, as received in the above condition, was sandblasted and pickled to remove contaminated surface material, then degreased and given a grain growth anneal in a high purity helium atmosphere contained within double sealed Senpack stainless steel heat treating envelopes. The hot working temperatures and annealing conditions are given in Table A. Grain growth curves for the materials are shown in Fig. 1. The compositions of the materials are given in Table B. The resulting grain sizes, amount of retained β phase, and lattice parameters are given in Table C. Grain sizes and the amount of retained β phase were determined by quantitative metallography. Lattice parameters were determined using the iterative extrapolation method suggested by H. Lipson and A. J. C. Wilson28 with the aid of crystallographic functions provided by T. B. Massalski and H. W. King.29

Micrographs of the material in the hot rolled and annealed conditions respectively are shown in Fig. 2a and b. The decrease in worked structure resulting from the anneal is evident from the two micrographs.

* A Titanium Metals Corp. of America designation for nominal 99.2% wt/o Ti, having an annealed room temperature 0.2% yield strength of 40,000 psi.
B. Apparatus for High Strain Rate, Plane Wave Loading

Samples of the materials investigated were subjected to plane wave loading at high strain rates by impacting two flat discs of identical material. The requirement of a well controlled impact velocity as well as very little deviation from simultaneous impact over the entire contacting surfaces of the discs were satisfied by the use of a light compressed-gas gun located at the Lawrence Livermore Laboratory. The gun was used to propel a projectile holding one of the specimen discs against a target consisting of the other disc along with instrumentation for measuring the stress waves generated in the target disc, the velocity of impact, and the angle between discs at the time of impact.

1. Gun

The gun shown in Fig. 3 consisted of a 15 foot long barrel with a nominal inner diameter of 3.5 inch. A target chamber was located at the muzzle end of the gun, shown at the right in the figure, and contained the tilting target mount used to align the target to the barrel. Projectile and target parts were recovered in a ten foot long catcher situated directly behind the target chamber and filled with shredded nylon parachute cloth.

The large circular part in the center of the figure is the compressed helium reservoir. The volume of reservoir could be varied so that for a given projectile mass, the volume and pressure (4500 psi maximum) of propelling gas could be adjusted to obtained the desired projectile velocity. Before firing, the projectile was attached to the breech block, shown protruding from the gas reservoir, which
prevented pressure leakage during firing by transmitting reaction forces through the heavy recoil bar shown to a heavy I beam mounted in the foundation. Air shocks in front of the advancing projectile were eliminated by evacuating the gun barrel and target chamber before firing.

2. Projectile

The projectile, shown in Fig. 4, was in the form of a body of circular symmetry to reduce bending forces on the projectile during accelerations. It was made of aluminum alloy to provide low mass. The rear of the projectile was in the form of a cup with 1/4 in. thick walls about 3 in. deep. The outer surface of the cup, with associated O-ring and chevron seals, acted as a valve sealing off the reservoir of pressurized He gas used to propel the projectile. The rear riding surface in the form of a teflon ring was situated on the front edge of the cup.

A steel rod was attached between the inner base of the cup and the breach block of the gas gun. This prevented forward motion of the projectile due to the pressure difference of 1 atmosphere in the cup and a vacuum of about 100 microns in the gun barrel ahead of the projectile.

The projectile was launched by bleeding pressure from the reservoir into the cup. At ~300 psi the steel rod broke, allowing the projectile to move forward and open the large ports leading to the He reservoir previously established at the firing pressure.

The front of the projectile was either in the form of a pair of flat discs as shown in Fig. 4 or a thin walled cup open at the
on the impact side of the target. The mirror surface was produced by continued lapping of one side in a 1900 grit slurry on a paper tool followed by polishing with Linde A polishing compound in a solution composed of 10 ml. 50% NaOH, 75 ml. H$_2$O$_2$ and 540 ml. H$_2$O on a Buehler Microcloth.

A mirror surface was achieved within about five minutes by initially polishing with a generous application of the slurry followed by allowing the Microcloth to run almost dry before removing the target. Immediate washing of the polished surface was necessary to prevent clouding. The abrasive papers, paper tool and polishing cloth were attached to glass porthole covers that had been lapped flat. To limit rounding of the edges of the polished surface, lapping and polishing with the slurries was done as rapidly as possible, and with the procedure described the rounding was limited to about 1/8 inch from the edge.

The quartz gauges were applied in two ways. The first was by epoxying the gauge to the target with Hysol epoxy, and then installing the brass shielding cup and electrical connection to the gauge electrodes, and finally potting the assembly in Hysol epoxy. The back side of a target assembly made in this way is shown in Fig. 6. Measurement of the target and gauge thickness before and after assembly showed the epoxy joint to be less than .0001 inch thick.

The second form of gauge application was a pre-connected and potted package in a brass shielding cup. In this case the front face of the gauge and the brass cup were given electrical connections by an
evaporated layer of gold. Both methods of assembly gave comparable results.

The target disc with gauge assembly was located in the center of a circular aluminum plate, Figs. 5 and 6, which contained a circle of holes concentric with the target disc for BaTiO₃ crystal pins which were used for projectile velocity and tilt measurements. The contact surfaces of two sets of velocity pins were located 3, 2, and 1 mm above the target surface. Four pins sensed the flatness of impact and two pins triggered the oscilloscopes. The oscilloscopes that recorded the velocity and tilt pins were triggered by a pair of bare wire ground pins set at 5 mm above the target and a pair of charged trigger pins of piano wire at -150 volt set at 4 mm above the target. The BaTiO₂ pins were measured for offset from the target surface to ± 0.0004 inch.

The pins were set on a diameter of 2.7 inch and thus were struck by the face plate around the flying plate. The offset of the face plate surface from the flying plate surface was accounted for in setting the pins. The diameter of the circle on which the pins lay was chosen to prevent the heads of screws used to attach the face plate to the projectile from contacting the pins.

4. Target Alignment

Two alignment procedures were used. In one the projectile was made for a slip fit at the muzzle end of the gun. This diameter was about 0.003 inch less than the diameter of the breech end of the gun. With the projectile protruding to about the location of the target, an optical flat was placed against the projectile face. By adjusting
the target tilting ring the optical flat was brought parallel to the flying plate face. Without moving the optical flat the projectile was removed and an auto-collimator was adjusted to be normal to the optical flat. The optical flat was then removed and the target assembly set in place and tilted into correct orientation by means of observations through the fixed auto collimator.

This procedure resulted in loose fitting riding surfaces on the projectile when placed in the firing position. This was considered undesirable as a possible cause of projectile vibration and flexure in transit to the target.

The second alignment procedure allowed for a projectile slip fit in the breach position but required the flying plate face to be perpendicular to the projectile riding surface axis to within one minute of arc. The teflon riding surface interference at the muzzle end was presumed to vaporize off before reaching the muzzle. In this procedure a perpendicular to six inches of the muzzle was established and recorded on the auto collimator. These settings were then used to bring the reflective target surface into alignment.

5. Quartz Gauges

The stress-time profiles at the back surface of the alloy targets were determined with x-cut quartz gauges. The gauges in the form of 1 inch diameter discs about .2 inch and .25 inches thick were gold plate on the opposing flat surfaces and had a concentric circular groove about .1 mm wide on the surface not contacting the alloy target, forming a guard ring as suggested by Graham, et al. The insulating groove that separated the inner electrode from the guard
ring was checked for current leakage several times during the construction of the target assembly, at an applied potential of 200 volts. This voltage was about 10 times the potential experienced during the actual stress-time profile measurement. The current from the inner electrode was picked off across a 50Ω matching resistor flash soldered to the electrode and fed to the oscilloscopes via a 50Ω cable. The guard ring current was either grounded directly or picked off across sufficient resistance to balance the voltages of two electrode areas during signal emission.

The recorded gauge signal is not directly proportional to the stress on the gauge. To obtain the stress time profile at the front of the gauge it is necessary to introduce two experimentally determined corrections for the gauge reading. The first is a 12% variation between 0 and 25 kbar of the current coefficient \( k \) in the expression \( \sigma = \left( \frac{i}{k} \right) \left( \frac{1}{A U_s} \right) \) where \( i \) is the current output, \( l \) is the gauge thickness, \( A \) is the inner electrode area, and \( U_s \) is the shock velocity of the gauge. \( U_s \) varies also but only about 0.2% between 0 and 25 kbar. The variation of \( k \) with stress has been determined by Graham et al.\(^{35}\) and has been fitted by a linear expression \( k = (2.007 + 0.00113\Delta \sigma) \times 10^{-8} \) Coulomb/cm\(^2\) kbar by Lysle et al.\(^{36}\)

The other gauge reading correction compensates for the linear increase in current output with time for a constant stress input below about 30 kbar as measured also by Graham et al.\(^{35}\) In addition, Graham and Ingram\(^{37}\) indicated that anomalous signals may arise for stress pulses of greater than 11.5 kbar which have shorter durations.
than the gauge transit time. The stress time profile in the measurements made in this investigation have a stress drop, due to release waves from the back of the flying plate, that may represent similar conditions to the stress drop at the back of a short duration pulse. Thus an anomalous current increase during the relaxation portion of the stress-time profile may raise the current output above that due only to stress differences between the front and rear of the gauge. No such anomalous increase is seen in Fig. 1 of Ref. 34 in which the relaxation of the maximum stress on arrival of the 5 kbar elastic precursor that had reflected from the back side of the impacting plate can be seen at the back of the stress time record. Thus, when no nonlinear increase in output is observed, the record should be considered true, subject to corrections for the variation in k with σ and a linear current increase. Above 30 kbar and in the case of anomalous effects the gauge should at least be useful as a measure of the magnitude of stress jumps. In all cases of measurements made within the quartz gauge transit time, the timing of events at the specimen gauge interface should be unaffected.

6. Effect of Tilted Impact on Quartz Gauge Measurements

The record obtained from the quartz gauge is a history of the stress state at the target-quartz interface. However it is also the sum of stress histories at all points on the interface. Thus, if the input stress wave arrives at the target-quartz interface with some angle between the plane of the wave and the plane of the interface, the entire interface will not have the same stress. The gauge output will then be a non-linear averaging of the input wave.
profile over a time interval that is equal to the time necessary for
the leading edge of the wave to sweep across the gauge area.

To see this, consider an incoming wave in the shape of a stress
impulse, Fig. 7, traveling at a velocity \( V \) at an angle \( \theta \) with respect
to the sample gauge interface, Fig. 8. Here \( D \) is the diameter of
the center electrode of the quartz gauge projected onto the gauge-
sample interface. The time for the wave to traverse the gauge area
is \( V\Delta t = D \tan \theta \) or \( \Delta t = D(\tan \theta)/V \). By inspecting the condition at the
launch of the stress wave at the impact surface, Fig. 9, \((\tan \theta)/V\)
may be determined. Here the flying plate impacts at an angle of \( \alpha \)
and velocity \( v \), completing the impact in a time \( \Delta t \). The stress wave
is emitted at a normal velocity, \( V = V(\text{elastic}) \cos \theta \) and is shown
in the position it attains at a time \( \Delta t \) after the flying plate first
makes contact with the target. From Fig. 9, \( v\Delta t = d \tan \alpha \), \( V\Delta t = d \tan \theta \),

\[
\Delta t = \frac{d \tan \alpha}{v} = \frac{d \tan \theta}{V} \quad \text{or} \quad \frac{\tan \alpha}{V} = \frac{\tan \theta}{V} \quad [1]
\]

Thus the sweep time across the gauge is \( \Delta t = d \tan \alpha/v \) all of which are
measurable.

The gauge output \( i = (6kU_s/l)A \) for a tilted impulse wave input
is proportional to \( dA/dt = (dA/dx)(dx/dt) \).
For a circular gauge area

\[ \frac{\mathrm{d}A}{\mathrm{d}x} = 2y = 2 \sqrt{r^2 - x^2}, \quad x = -r + \frac{vt}{\tan \alpha}, \quad \frac{\mathrm{d}x}{\mathrm{d}t} = \frac{v}{\tan \alpha} \quad [2] \]

thus

\[ \frac{\mathrm{d}A}{\mathrm{d}t} = 2 \left( r^2 - \left( -r + \frac{vt}{\tan \alpha} \right)^2 \right)^{1/2} \frac{v}{\tan \alpha} = 2 \left( \frac{v}{\tan \alpha} \right)^2 \sqrt{\frac{\mathrm{d} \tan \alpha}{v}} t - t^2 = \omega(t) \quad [3] \]

for

\[ 0 < t < \frac{\tan \alpha}{v} = \Delta t. \]

The impulse response of the gauge, \( \omega(t) \) is shown in Fig. 10.

The output \( c(t) \) of a transducer is equal to the convolution of the input \( r(t) \) with the impulse response \( \omega(t) \) of the system. 38

\[ c(t) = \int_0^t r(t) \omega(t - \tau) \, \mathrm{d}\tau \quad [4] \]

In the form of a sum

\[ c(t) = \sum_{n=0}^{N} r(n\Delta t) \omega(t - n\Delta t) \Delta t \quad [5] \]

An example of the response of a gauge of smear time \( \Delta t = 10T \) to a ramp input of duration \( 20T \), Fig. 11(a), is shown in Fig. 11(b) and has a duration of \( 30T \). If the ramp duration is \( T \) then the output duration "T" is

\[ "T" = T + \Delta t \quad [6] \]
The slope of the linear portion of the output can be determined by looking at the expression \( \frac{dc(t)}{dt} = \frac{c(t) - c(t+\Delta t)}{\Delta t} \)

\[
= \frac{1}{\Delta t} \sum_{n=0}^{N} r(n\Delta t) \omega(t-n\Delta t) \Delta t - r((n-1)\Delta t) \omega(t-(n-1)\Delta t) \Delta t
\]

\[
= \frac{1}{\Delta t} \sum_{n=0}^{N} \left( r(n\Delta t) - r((n-1)\Delta t) \right) \omega(t-n\Delta t) \Delta t. \tag{7}
\]

For an input wave having a strain history in the shape of a ramp of constant strain rate \( \dot{\varepsilon} \) to some final strain,

\[
\dot{\varepsilon} \Delta t = r(n\Delta t) - r((n-1)\Delta t) \tag{8}
\]

then

\[
\frac{dc(t)}{dt} = \sum_{n=0}^{N} \dot{\varepsilon} \Delta t \omega(t-n\Delta t) \Rightarrow \dot{\varepsilon} \int_{0}^{t} \omega(t-\tau) d\tau \tag{9}
\]

but since \( \omega(t) \) is symmetrical about \( \Delta t/2 \),

\[
\frac{dc(t)}{dt} = \dot{\varepsilon} \int_{0}^{t} \omega(\tau) d\tau = \dot{\varepsilon} A \tag{10}
\]

for the time in which the linear part of the ramp overlaps the gauge rise time \( \Delta t \), Fig. 12. In Fig. 11 it can be seen that the lower curved foot of the output lasts a time interval equal to the smear time and the upper curved foot lasts the same time interval. The duration of the linear slope \( \dot{\varepsilon} A \) is then equal to \( "\tau" - 2\Delta t \). The ramp time can also be seen to be \( "\tau" - \Delta t = T. \)
In general then the output corresponding to an input $r(t)$ is its convolution with $\omega(t)$. One course for finding the input wave shape would be to perform the convolution on many different input wave shapes and compare with the recorded output to obtain the best match and call that the input. Alternatively the input is the inverse Laplace transform of the ratio $L(\text{output})/L(\text{impulse response})$ a rather cumbersome calculation. If all inputs are assumed to be ramps, the simplest determination of the slope of the ramp ($\dot{\varepsilon}$) is $\dot{\varepsilon}/T$ where $\varepsilon$ is the peak strain in the wave and $T = \tau - \Delta t$ as determined from experimental parameters of gauge dimensions, $v$, $a$, etc. The $\varepsilon$ for the tests were estimated by assuming a ramp input.

7. **Data Acquisition**

The signals produced by the quartz gauges were recorded on a series of Tecktronix 585 oscilloscopes. The current output of the gauge was grounded through a 50Ω load resistor. The resulting voltage signal across the load was transmitted to the oscilloscope through 50Ω coaxial cable which was terminated by a 50Ω resistor after the last oscilloscope. The transmission time from the gauge to the oscilloscope was less than the duration of the signal and terminating resistors were carefully chosen to minimize signal reflections from the cable end.

High speed Polaroid film was used to record both the oscilloscope display of the quartz gauge as well as accurate voltage and timing calibration signals applied to the oscilloscope at the same settings.

The very short duration signals from the BaTiO$_3$ velocity and tilt pins were transmitted via coaxial cables to oscilloscopes and recorded via either sheet film or Polaroid film. Figure 13 is an example of one
of these oscilloscope records. The five small spikes on each line are timing marks, .5μ sec. apart. The third whole line down shows two pair of spikes. The first pair is an externally generated fiducial mark used to correlate the records on the four oscilloscopes used. The projectile tilt at impact was obtained from the last pair of spikes on the four oscilloscope records. The second spike of all pairs is an echo provided to enable time measurements to be made between time marks if the signal arrived at the front or back of a line. The second, third and fourth pair of spikes represent the arrival of the projectile at the Ba TiO₃ velocity pins located 3, 2 and 1 mm above the target and were used to calculate projectile velocity.

C. **Apparatus for Low Strain Rate Testing:**

**Compression Test**

Specimens of the titanium and titanium-aluminum alloys were subjected to compressive strains at low strain rate for comparison of the deformed structures with those produced during high strain rate, plane wave loading. The deformation produced in the plane wave loading tests occurred under the influence of stress fields containing a large component of hydrostatic pressure. To duplicate this large hydrostatic pressure during low strain rate compression, samples were made with width to height ratios varying from 7.2 to 9.1 depending on the value of hydrostatic pressure desired at the center of the sample.
The average hydrostatic pressure prevailing during the deformational part of the loading wave was determined for each alloy subjected to high strain rate loading. That is, the hydrostatic pressure at maximum strain and at yield strain were found from the $\sigma_x$ and $\varepsilon_x$ behavior of the material determined in one dimensional strain. The low strain rate compression specimens were designed to yield under a stress field having a hydrostatic component that was the average of the two hydrostatic pressures mentioned above. The average hydrostatic pressure, $H$ and the ratio of the average hydrostatic pressure to the yield stress $\sigma_y$ in simple tension are given in Table D.

Thomsen, Yang, and Kobayashi\textsuperscript{39} give the following expressions for the normal stress $\sigma_y$ and the hydrostatic stress $H$ in terms of the effective stress $\bar{\sigma}$ at the center of a slab of height $h$, length $l_f$ and width $W > l_f$, Fig. 14,

\[
\frac{-\sigma_y}{\bar{\sigma}} = \frac{2x_0}{h\sqrt{3}} + \frac{1}{\mu\sqrt{3}} \quad \text{and} \quad \frac{H}{\bar{\sigma}} = \frac{\sigma_y}{\bar{\sigma}} + \frac{1}{\sqrt{3}} \quad [11]
\]

where

\[
x_0 = \frac{l_f}{2} - \frac{h}{2\mu} \ln \frac{1}{2\mu} \quad [12]
\]

is the distance from the center of the specimen over which sticking between the sample and compression plattens occurs. The average coefficient of friction $\mu$ between the alloys and the tungsten carbide platten was experimentally determined to be $\mu=.26$.

The equations above can be reduced to the expression

\[
-\frac{H}{\bar{\sigma}} = \frac{1}{\sqrt{3}} \left[ \frac{1-\mu}{\mu} \frac{l_f}{h} - \frac{1}{\mu} \ln \frac{1}{2\mu} \right] \quad [13]
\]
and for given $H/\bar{\sigma}$ the ratio $l_r/h$ of length to height for the specimen may be determined. The value of $\bar{\sigma}$ was taken as the yield stress $\sigma_y$ in simple tension in this calculation.

As a check on the applicability of these expressions, one of the specimens was compressed to residual strain of 0.2%. The average stress on the sample to cause this plastic strain was compared with the average stress predicted by the analysis used to derive the above equations. The measured value was 30% higher than the predicted value, due most likely to early yield along the perimeter of the sample and pillowing as discussed in Section III B.

The specimen dimensions given in Table E were determined from the required value of $l_r/h$ as well as the maximum load capacity of the compression machine and the size of the tungsten carbide compression plattens used. The compression tests were done on a Material Testing System (MTS) 300 KIP capacity universal testing system.

The samples were fly cut and lapped flat to within .0001 inch. They were compressed at a strain rate of $\dot{\varepsilon} = 10^{-3}$ sec$^{-1}$ between two flat ground tungsten carbide plattens. The angle between compression platten surfaces was about 5 minutes of arc. Strain values were obtained by measuring the thickness of the center of the specimen before and after compression to within .0001". The strained specimens were then sectioned for metallographic observation of the center of the specimen.

**D. Tension Tests**

The tensile properties of the titanium and titanium-aluminum alloys investigated were determined from flat tensile specimens cut along the
rolling direction of the alloy sheet. The dimensions of the .125 in. thick tensile specimens are shown in Fig. 15. Tension tests were made at a strain rate of $0.7 \times 10^{-3}$ sec$^{-1}$ on an Instron testing machine. The properties determined in these tests are given in a later section.

E. Ultrasonic Velocity Measurements

The adiabatic sound velocities required in the analysis of the plane wave loading experiments were measured at the Lawrence Livermore Laboratory. The samples on which the measurements were made were the target discs used in the plane wave loading experiments and thus had both faces parallel to less than 15 seconds of arc. The same alloy discs were used to determine the densities by the water immersion method. The longitudinal sound velocity was measured at a frequency of 10.0 MHz using a water column and the shear wave velocity was measured at a frequency of 5.0 MHz using quartz rods coupled to the samples by a viscous resin.

The bulk wave velocity $c_B$ is related to the measured quantities by the expression:

$$c_B^2 = c_L^2 - \frac{4}{3} c_S^2$$  \[14\]

where $c_L$ is the longitudinal wave velocity and $c_S$ is the shear wave velocity. The wave velocities for the alloys measured in the thickness direction are given in Table F along with the densities $\rho_o$. 
F. Metallography

Sections of the alloys were prepared for optical metallography by a polish and anodize technique. Samples were mounted with an electrical contact attached. After grinding through 240, 320, 400 and 600 grit abrasive papers the samples were polished directly on microcloth with 0.05μ alumina abrasive in an attack polishing solution. The solution contained approximately 75 ml distilled H₂O, 7 ml of 10 wt/o NaOH solution, 12 ml H₂O₂ (31.5%) and 10 ml Gamma Micropolish (0.05μ alumina). After polishing for about one hour the surfaces were flat and mirror like, however metallographic examinations showed a speckled surface artifact, most likely colonies of thick oxide. This could be removed by electropolishing at 25 volts for 3 seconds at room temperature, with a stainless steel cathode on a Disa Electropol apparatus using Disa solution A-3. The composition of this solution is 60 ml Perchloric Acid, 600 ml Methanol and 360 ml Butylcellosolve (2-n-butoxyethanol). The short electropolish however also tended to pit the polished surface by rapid attack on the retained β grain, as well as to leave a slightly rough surface as the result of differences in polishing times between those areas initially free of the artifact and those areas briefly protected by the artifact.

These surface irregularities were substantially diminished by returning to the attack polish solution for about 15 minutes. The artifact would form again in the solution but much more slowly. This artifact was less noticable in the alloys with increasing aluminum content.
The anisotropy of the HCP lattice results in a metallographically useful effect. When observed using polarized light, individual grains were revealed by color differences between grains. This eliminated the necessity of using a grain boundary etch. The dependence of color on orientation of the surface of the grain could be enhanced by anodizing. The anodizing solutions used in the investigation was given by Pickleshimer and is composed of 240 ml ethanol, 140 ml H₂O, 20 ml phosphoric acid, 40 ml lactic acid, 80 ml glycerine, and 8 gm citric acid. At room temperature and 18 volts the anodized layer was completed within about 10 secs. The layer thickness could be varied by changing the voltage setting.

The optical reflectivity of the small grains of retained \( \beta \) phase, a bcc structure, showed a different behavior from that of the \( \alpha \) phase when viewed under polarized light in that the \( \beta \) phase grains did not change color when the sample was rotated.
III. THEORETICAL DEVELOPMENT AND EXPERIMENTAL RESULTS

A. Plane Wave Loading

The position of stress waves in the flying plate and target assembly are shown at two times after the flying plate impacts the target disc in Fig. 16. In the figure the flying plate and a target assembly are shown in cross section. The plane $x=0$ is the plane of contact of the flying plate and target. The left side of the $y=0$ line shows the position of stress waves $A$ and $B$ produced by the impact shortly after the time of impact. The small arrows indicate the direction of motion of the waves. The right side of the $y=0$ line shows the position of stress waves after the elastic part $A_e^r$ of the stress wave $A$ has reflected from the target-gauge interface.

In this investigation stresses greater than the elastic limit were generated in the materials. Under these conditions the stress waves separate into a wave of magnitude equal to the elastic limit and running at the elastic velocity followed by a wave that brings the materials to the maximum stress and running at a velocity that depends on the peak pressure of the wave.\(^3\) For peak pressure a few times greater than the elastic limit the velocity of the second wave is close to the bulk sound velocity. Waves $A$ and $B$ then separate into $A_e$ and $B_e$, the elastic waves, and $A_p$ and $B_p$ the slower plastic waves.

The position and direction of these separated waves shortly after the elastic wave $A_e$ has reached the target-quartz gauge interface is shown the right side of Fig. 16. Here $A_e^t$ is the magnitude of the
wave that is transmitted into the quartz, $A_e^t$ with magnitude of the wave that is reflected from the quartz target interface, $A_p$ is the slower moving plastic part of wave A that has not yet reached the position of $A_e^t$, and $B_e^t$ and $B_p^t$ are the separated parts of wave B that have reflected from the back surface of the flying plate.

The quartz gauge output is a function of the difference between the normal stress acting on the surface in contact with the target and the normal stress acting on the back surface of the quartz.\(^{35}\) For times less than the time required for the first wave $A_e^t$ to traverse the thickness of the gauge, the gauge output is a function only of the stress history of the gauge-target interface. (The potting material that surrounds the gauge and the circuit leads attached to the "active" center electrode at the back of the gauge lead to an undetermined stress condition at the back of the gauge for times greater than the transition times of wave $A_e^t$.)

An example of records obtained for the stress history at the target-quartz interface resulting from the impact of the flying plate on the target is shown in Fig. 17. The flying plate and target material in this case was Ti 10 atomic percent Al. The duration of the trace was 1 \(\mu\)sec.

The straight horizontal line is a base line from which the vertical deflection of the wave profile was measured. The stress history with time increasing from left to right shows a stress build up on the left corresponding to the arrival of the front running elastic wave.
A gradually increasing stress follows until the arrival of the plastic wave which brings the stress to its maximum value. The long segment at the top of the wave has a slightly positive slope and is a characteristic of the gauge when a constant stress level of sufficient magnitude exists at the target-quartz interface. At the right side of the figure, the curving decrease in gauge output is the result of decrease in normal stress at the front surface of the gauge caused by the arrival of rarefaction waves $B_r^e$ and $B_r^p$ that had reflected from the back of the flying plate. The signal terminates when wave $A_t^e$ arrives at the back surface of the gauge.

At the front of the wave, between the top of the elastic wave and the middle of the deformational wave a disturbance of the signal is seen. This disturbance was observed on all wave profiles recorded. Several different methods of mounting the gauge on the target and installing the associated circuit leads, coaxial shielding and grounding were used in an attempt to eliminate the disturbance. The disturbance occurred at differing current levels in each experiment as well as different times after the beginning of the quartz output. Close examination of the disturbance on expanded records obtained from other oscilloscopes showed the fundamental frequency of the disturbance to be inconsistent. The one characteristic which was consistent from one experiment to the next was that the disturbance was associated with the deformational part of the wave profile. The disturbance decreased in magnitude with specimens of increasing aluminum content. Later metallographic examination of the target
materials showed that the amount of twinning also decreased with increased aluminum content. A check was made on the supposition that the target material was the cause of the disturbance by performing an impact experiment with an aluminum alloy target. No disturbance was observed in this case. These observations lead to the interpretation that the disturbance was caused either by i) twinning in the target material at the target quartz interface which effects the instantaneous state of average stress on the quartz front surface or by ii) elastic wave generation during twin formation within the target.

The oscilloscope records obtained from these experiments were quite adequate, however, for the determination of the stress wave profiles in the titanium and titanium aluminum alloys. The stress history at the target-gauge interface was determined from the expressions

\[ \sigma = \frac{V}{R} \cdot \frac{\ell}{A U_S} \]

(see section on Quartz Gauges) and \( V = i \cdot R \) where \( V \) is the voltage measured on the record obtained from the quartz gauge and \( R \) is the resistance through which the short circuit current generated by the gauge was passed. This determination should be valid for initial stress jumps up to 45 kbar and for all stress levels below 25 kbar.\(^{37}\)

The impedance method\(^{26}\) was used to translate stress history at the target quartz interface to stress histories within the target materials. The impedance method is based on the second of the Rankine Hugoniot relations\(^{41}\) which state conservation of mass
and conservation of momentum

\[ \Delta \sigma = \rho_1 U_s \Delta u_p \]  

across a wave having time independent pressure profile. Here \( \rho_1 \) and \( \rho_2 \) are the material density in front of and behind the wave, \( U_s \) is the velocity of the wave with respect to the material ahead of it, \( u \) is the velocity of material behind the wave, \( \Delta \sigma \) is the stress increase or decrease across the wave and \( \Delta u_p \) is the change in particle velocity through the wave.

The second equation can be represented graphically as \( \sigma \) vs \( u_p \) in which states reached by the passage of such waves are connected by lines of slope \( \rho U_s \), the acoustic impedance. For waves moving in a positive \( x \) direction the slope is positive, for waves moving in a negative \( x \) direction the slope is negative.

The method of translating pressures in the quartz gauge to pressures in the target is shown in Fig. 18. The quartz impedance line was obtained from Van Thiel et al.\(^4\) In the notation of in Fig. 16, line segment \( A_e \) represents the stresses resulting from a change in material velocity across an elastic wave moving in a positive \( x \) direction in the target. Line \( A_e^r \) represents a reflected elastic wave moving in the negative \( x \) direction in the target and line \( A_e^t \) represents a wave transmitted into the quartz gauge. Point 1
is the stress level in the quartz caused by the elastic wave in the target. The stress and material velocity of Point 1 must be common to both sides of the target-gauge interface for equilibrium and continued contact in compression. Because the quartz impedance is less than the titanium elastic impedance the wave \( A_e^R \) reflected back into the target after the arrival of the elastic part \( A_e \) of the stress profile is a release wave. Wave \( A_e^R \) does not exceed the elastic limit and thus travels with the elastic velocity and has the same slope in magnitude as wave \( A_e \) in the \( \sigma_x - u_p \) diagram. (The coefficient \( \rho \) in the impedance \( \rho U_s \) in actually slightly larger due to elastic compression at the elastic limit, however for this material the correction is about 1%). The stress level at the Hugoniot Elastic Limit * (HEL) of the alloys was thus obtained from the quartz stress history by finding the intersection Point 2 Fig. 18 of two lines having the elastic impedance of the target one drawn from the origin of the \( \sigma_x, u_p \) diagram and one drawn with negative slope from Point 1 on the quartz lines.

The translation of the quartz stress history to the target stress history for stresses above the HEL can be made in similar fashion provided the rarefaction wave remains elastic and the impedance for the target material is known above the elastic limit. If the assumption of elastic perfectly plastic behavior is made for the target material, and the maximum stress in the target is less than a few times the elastic limit, the impedance of the plastic wave in \( \rho_{HEL} U_s = \rho_{HEL} c_B \), where \( c_B \) is the bulk sound velocity of the material, obtained from ultrasonic velocity measurements.

* The Hugoniot Elastic Limit is the elastic limit displayed by a material undergoing one dimensional strain in a plane wave loading condition.
A method which overcomes any uncertainty in the accuracy of the quartz gauge record at maximum stress utilizes impedance matching between the flying plate and the target. Having determined the elastic limit and impedance for the target and flying plate material, one can draw lines on the $\sigma x$, $u_p$ diagram Fig. 19 for the stress-particle velocity dependence of the target initially at rest with $u_p = 0$ and subjected to compression waves moving in the positive $x$ direction, and of the flying plate with material velocity initially equal to the projectile velocity and subjected to compression waves moving in the negative $x$ directions. Because the stress and material velocity at the flying plate target interface must be equal in compression, the maximum stress produced by the impact is given by the interaction of the two sets of lines. The flying plate and target materials were of the same alloy and taken from neighboring positions on the annealed sheet, thus their stress-particle velocity dependence is identical and the diagram is symmetrical about the vertical line $u_p = 1/2 u_{\text{projectile}}$. The projectile velocity was obtained quite accurately (less than 1% error) and so the maximum stress reached in the experiments were taken as the intercept of the plastic impedance lines with the line $u_p = 1/2 u_{\text{projectile}}$. The values of elastic and plastic impedance, target thickness, projectile velocity and angle of tilt at impact for the experiments are given in Table G.

The change in density of the target associated with the elastic and plastic compressions were determined from the Rankine-Hugoniot equation for conservation of mass. The compression strain at each
change in density was found from the expression $\epsilon = 1 - \rho_0 / \rho_1$ for compressive strains taken as positive. Strain rates associated with each wave at the back side of the target were calculated by assuming the waves to be truncated ramps of constant strain rate (see discussion in Section II. B6 and Fig. 11a). Values for $\sigma_x$ and $\epsilon_x$ at the Hugoniot elastic limit and at the maximum stress as well as the average strain rate associated with each wave using the two wave loading assumption are given in Table H.

An improvement in the description of the loading path for these experiments is possible if details of the shape of the plastic loading wave are considered. When the plastic loading wave is represented by a single wave moving at the bulk velocity, $c_B$, no account is made for increases of the flow stress offset from the hydrostatic compression curve due to work hardening or decreases of this flow stress offset due to decreasing strain rate as the maximum strain is approached. Flow stress increases due to work hardening appear as stress increases in the plastic loading wave which run at velocities greater than the bulk sound velocity. The tailing off of the flow stress due to decreasing strain rate results in a increasing effective compressibility and thus a decreasing velocity of propagation near the top of the plastic stress wave.

To investigate whether these effects were discernable, the wave profiles recorded were represented by a number of segments having differing slopes. The angle of tilt at impact could lead to misinterpretation with this method because the sweep time of the wave across the active area of the gauge gives rise to rounding of the responses
of the gauge to the input wave. Care was taken that the segments used to represent the wave profile were of a greater time duration than the tilt related rise time of the gauge.

Figure 20 shows the segmentation of one of the wave profiles recorded. The time of arrival of the elastic precursor after impact was taken to be the transit time of a wave traveling at velocity \( c_L \) through the target. The times of arrival of subsequent waves were taken at the half height of their stress increment. The velocities of these subsequent waves were not inversely proportional to the transit time, however, because the position of the target quartz gauge interface varies after the arrival of the first wave.

The motion of the interface can be accounted for by constructing a position time diagram Fig. 21 (described below) for the target in laboratory coordinates. The material velocities required in this construction were obtained from a \( \sigma_x - u_p \) diagram, Fig. 22 (also described below) based on a postulated three wave representation of the loading path. Here the elastic wave is followed by two plastic waves. Several iterations of a set of equations equivalent to the position-time diagram were required to obtain agreement between plastic wave velocities and arrival times.

Figure 21 shows the state of stress \( \sigma_x \) and material velocity \( u_p \) within the target after the impact of the flying plate. The numbers designating each state refer to points in Fig. 22. Thus state 0 represents the stationary part of the target which has not been reached by the elastic precursor wave. The lines separating states in Fig. 21,
represent the locus of position of the loading and reflected unloading waves in the target. The lines connecting states in Fig. 22 have slopes equal to the impedance \( \rho U_s \) of the materials for each change in states. The lines 0-1-2-3 describe the loading path of the target. Lines 0-4-8-11-12 describe the loading path of the x-cut quartz gauge. Lines 13-14-15-3 describe the loading path in the flying plate.

The sequence of events in the target are as follows: the elastic wave followed by the two plastic waves compress the target material to states, 1, 2 and 3 respectively. At \( t_1 \) the elastic wave meets the gauge interface, and is reflected as unloading wave 1-4 with the titanium elastic impedance and transmitted as loading wave 0-4 with the quartz impedance in order to maintain the target and gauge in contact at the same stress level and particle velocity. Loading wave 1-2 interacts with unloading wave 1-4 and re-compresses the material in state 4. State 4 however is at a stress below the elastic limit and consequently elastic wave 4-5 develops which is followed by plastic wave 5-6. Wave 5-6 meets another reflected unloading wave and breaks up into elastic and plastic loading waves 8-9. At \( t_2 \) the plastic part of this wave eventually reaches the gauge interface, which has been moving at the progressively increasing material velocities of state 4, 8 etc.

The final loading wave 2-3 moving at the lowest velocity interacts with the reflected unloading waves before meeting the gauge interface at time \( t_3 \) and bringing the target and gauge to state 12. The motion of the gauge interface is evident in Fig. 21. The set of equations
used to calculate wave velocities were iterated until the calculated arrival time differed from the measured arrival times by 0.5% or less. Wave velocities determined without consideration of material velocities were up to 3% in error. The results of these calculations are given in Table I where \( \sigma_x \) is the stress at the top of each wave moving at the wave velocity shown. The impedance associated with each wave is also shown as well as the average strain rate calculated assuming a ramp wave. It was also possible to represent the records for Ti-10 at.% Al by a four wave structure having three plastic wave segments and these results are also shown in Table I.

The differences between the behavior of the materials during impact loading in one dimensional strain and the conventional low strain rate tensile curves can be conveniently shown by reducing the one dimensional strain loading paths to one dimensional stress conditions. Fowles\(^{14}\) has given the expressions

\[
\sigma_x = P + \frac{2}{3} \sigma_S \quad [18]
\]

\[
\varepsilon_x = \frac{3}{2} \varepsilon_S - \frac{\sigma_S}{6K} \quad [19]
\]

for converting stress-strain behavior in one dimensional stress to stress-strain behavior in one dimensional strain. Here \( \sigma_x \) and \( \varepsilon_x \) are the stress and strain in one dimensional strain, \( \sigma_S \) and \( \varepsilon_S \) are the stress and strain in one dimensional stress, \( K \) is the adiabatic bulk modulus, taken here to be \( K = \rho c_B^2 \) where \( \rho \) is the density and \( c_B \) is the bulk sound speed, and \( P = K \varepsilon_x \) is the hydrostatic pressure for a one dimensional strain of \( \varepsilon_x \). These expressions can be rearranged to
convert from one dimensional strain to one dimensional stress in the form

\[ \sigma_s = \frac{3}{2} (\sigma_x - K\varepsilon_x) \]  \hspace{1cm} [20]

\[ \varepsilon_s = \frac{1}{2} \varepsilon_x + \frac{\sigma_x}{6K} \] \hspace{1cm} [21]

The loading paths of the materials converted to one dimensional stress are shown in Figs. 23, 24, and 25 as \( \sigma_s, \varepsilon_s \). The isothermal quasi-static stress-strain curves for the materials are also shown in the figures as 1-D stress. The full and dashed curves distinguish between 2, 3 and in Fig. 25, 4 wave loading path representations.

B. Planar Compression Loading

Thin rectangular specimens of the titanium and titanium-aluminum alloys were compressed at a strain rate of \( \dot{\varepsilon} = 10^{-3} \) to investigate the deformational behavior of these materials at low strain rate subject to large lateral constraining forces. The specimens were designed to produce a hydrostatic pressure component of the stress field at the center of the specimen which was equal to the average hydrostatic stress experienced by the materials subjected to high strain rate, plane wave loading during the passage of the plastic strain wave. The method used to determine the proper dimensions of these specimens was given in Section IIC.

Information on the stress experienced by the specimens was obtained from load-displacement curves recorded during the tests. Examples of the loading behavior of the materials are given in Fig. 26 which shows the average engineering stress on the faces of the specimens.
versus the loading ram displacement. The residual plastic strain at
the center of the samples used for these curves was about 6%. The
average stress at this amount of plastic strain was about five times
the flow stress at the same plastic strain in simple tension. The
local normal stress near the unsupported perimeter of the sample must
be near the flow stress in simple tension due to lack of constraint.
Thus the material at the center must experience normal stresses even
greater than five times the flow stress in simple tension.

For a thin flat compression specimen the process of yielding
and plastic flow is a complex one. As opposed to testing in simple
tension, yielding does not occur for all material experiencing the same
normal stress. Material closer to the center of the sample will not
yield at a normal stress slightly greater than the yield stress when
the lateral constraints give rise to sufficient lateral stress to
reduce the shear stress on this material below the shear stress
required for yield.

This behavior is evident on the average stress-displacement
curves of Fig. 26. The initial slope of the curves corresponds to
elastic behavior on all parts of the loading system, that is, the
compression machine, the compression plattens and the specimen. The
location of first yielding of the specimen is at the edges and as the
load on the specimen increases the volume of yielded material increases.
The boundary between elastic and yielded material, Fig. 27, moves
toward the center of the specimen as the load increases. Thus the
material has yielded at increasing levels of the hydrostatic pressure
on going from the perimeter of the sample to the center.
When the perimeter of the specimen yields, it contributes a diminishing amount to the increase in load with increase in displacement and the average stress displacement curve exhibits a slight knee. On continued loading the center of the specimen yields and another slight knee appears. These knees between all elastic, mixed elastic and plastic, and all plastic strains should be quite evident in an elastic-perfectly plastic material where the change in $\sigma$-$\epsilon$ behavior is abrupt. In fact the knees were slightly more evident for the high aluminum alloy, which exhibited the least difference between ultimate tensile stress and yield stress in simple tension, than the other two materials, which displayed somewhat greater work hardening.

One specimen of the Ti-10 at.% Al alloy was loaded up to near the second knee on the curve, where central yielding should begin. Thickness measurements before and after loading indicated a residual plastic strain at the center of $0\% \pm 1\%$ and at an edge of $0.84 \% \pm 0.1\%$.

The strain gradient across the sample results in a slightly pillow-shaped sample after loading. The supposition might be made that the "pillowing" is substantially due to a cupping of the compression platten resulting from the high level of normal stress near the center of the sample. Although it is true that the plattens cannot remain planar under the unevenly distributed load, the ratio of compressibilities of the tungsten carbide and titanium alloys indicate that the tungsten carbide will strain about one-third as much as the elastic strain in the alloys and about one sixteenth as much as the
plastic strain in the alloys. It seems then that the pillow shape of the specimens is due in part to the cupping of the compression plattens and in part to the fact that elastic release at the center of the sample during unloading is from a much higher stress level than the elastic release at the edges. Consequently even if the sample and plattens were designed so that the platten-sample interface were flat at the peak load, the center of the sample would be thicker than the edges when the load was released.

With the 0.100 inch thick specimens used, the contribution of plastic strain in the samples to the total compression ram displacement was quite small. For example, a plastic strain of 5.5% at the center of the commercially pure titanium specimen, lowest curve of Fig. 26, represented only 9% of the ram displacement. A plastic strain of 1% contributed less than 3% to the ram displacement. It was initially felt that compressing the specimens to specific values of residual strain would be quite difficult. This however was not the case. Specific values of residual strain could be obtained by compressing a specimen to the maximum load capability of the compression system and measuring the resulting residual strain. The ram displacement corresponding to this plastic strain was subtracted from the total ram displacement at maximum load. This determined point A on Fig. 26. A line was drawn through point A tangent to the loading curve. The distance between the tangent line and the loading curve at any load level was then equivalent to the decrease in thickness of the center of the specimen due to plastic strain. Loading of subsequent specimen
could be carried on until the desired displacement between the curve and this tangent line was reached. This method was particularly helpful in obtaining specimens strained to value near 0.5% where the contribution of specimen plastic strain to the ram displacement was about 0.15%. The tangent line for determining plastic strains is shown with the loading curve for each material in Fig. 26, as well as extensions of the initial all elastic behavior of the specimen and compression system.

The magnitude of the hydrostatic pressure at maximum strain for each specimen can be obtained from the expression used to design the specimen. Once again, at the center of the specimen \( H = \sigma_y + \bar{\sigma}/\sqrt{3} \). For the derivation of these expressions \( \bar{\sigma} \) is required to be constant from the edge of the specimen to the center. Then at the edge of the specimen \( \sigma = 3/2 \sigma_f \) where \( \sigma_f \) is the flow stress at that particular edge strain. Then \( H = \sigma_y + \sigma_f/2 \). Combining this with the expression for the normal stress \( \sigma_f \) at the center of the specimen results in the expression:

\[
H = \sigma_f \left[ \frac{\sigma_f}{h} + \frac{1-\mu}{\mu} - \frac{1}{\mu} \ln \frac{1}{2\mu} \right]
\]  

The results of calculations for the compression specimens are plotted versus residual plastic strain at the center of the specimen in Fig. 28. The values for \( \sigma_f \) required in the calculations were obtained from tensile test results at the corresponding edge strain. The curves for hydrostatic stress versus plastic strain in Fig. 28
thus display a decreasing slope at higher strains that corresponds to decreasing work hardening.

At the left of the figure, the three dashed lines are the increase in hydrostatic stress with plastic strain for the material in the one dimensional strain tests. It is quite evident that specimens having values of $\lambda_f/h$ in excess of ten would be needed to provide sufficient constraint to duplicate the one dimensional strain behavior during plastic strain. The purpose of these tests was to study the deformational behavior of the materials subjected to the same range of hydrostatic pressure as prevailed during deformation in the plane wave loading tests. Clearly this was the case.

C. Microstructural Changes Produced by Low Strain Rate Deformation

Specimens of the titanium and titanium-aluminum alloys that had been compressed in the manner described above were sectioned through the center to reveal the resulting deformation structure. Optical micrographs of the material at the center of the specimens, where deformation occurred at the maximum level of hydrostatic pressure, are given in Fig. 29, 30 and 31, (a)-(f). Figures 29, 30 and 31 show compositions Ti, Ti-5 at.% Al and Ti-10 at.% Al, respectively with the amount of residual plastic strain increasing from (a) to (f). The differences in gray tone from grain to grain is the result of anodizing the polished surfaces and observation by means of polarized light metallography. Very thin twins were quite easily distinguished from polishing scratches using this technique because the gray tone of twins
and grains varied as the specimen was rotated while the appearance of scratches did not vary.

The amount of twinning is seen to increase with increasing strain although to a lesser extent with increasing aluminum content. For example by comparing Fig. 29(e), 30(d) and 31(c) corresponding to Ti, Ti-5 at.% Al and Ti-10 at.% Al all at a plastic strain of about 2.9%, the amount of twinning is seen to diminish considerably with increasing aluminum content. This comparison is for deformation under quite different levels of hydrostatic constraint $\bar{H}/\sigma_y$, Table D. To eliminate this variable, similar compression tests were made with specimens all having the same ratio $e_2/h=5.0$ for which $\bar{H}/\sigma_y=2.67$. As shown in Fig. 32 where (a) is Ti, (b) is Ti-5 at.% Al and (c) is Ti-10 at.% Al, for 2% plastic strain the same reduction of twinning with increase of aluminum content in the $\alpha$-phase exists.

Twin volume fractions were determined for the materials and strain conditions described. The determinations were made by point-count quantitative metallography on micrographs of the same areas shown in Fig. 29-31 but taken at less than half the magnification to average the somewhat nonuniform distribution of twinning notable in these micrographs. The results are given in Fig. 33 for the high hydrostatic pressure compression tests and in Fig. 34 for the moderate hydrostatic pressure compression tests.

Several features are noticiable in Fig. 33. There is evidence of initial plastic strain by pure slip motion before the onset of twinning. The extent of this initial "incubation strain" increases
with addition of aluminum in the α-phase titanium-aluminum alloys. Also the rate of increase in twinning with increasing strain becomes greater after a certain level of strain is reached. That is, at small plastic strain no twinning occurs. Then twinning begins at a rate which increases at a higher level of strain. The level of strain at which the twinning rate changes also increases with the addition of aluminum. However, the initial twinning rate decreases with the addition of aluminum. The results shown in Fig. 34 for specimens having the same ratio of hydrostatic stress to yield stress tend to support the above observations and in addition to show that the higher twinning rate also decreases with the addition of aluminum. The values of "incubation strain", initial rate of twinning, transition strain and final rate of twinning for the two series of compression tests described in Figs. 33 and 34 are given in Table J.

As an indication of how the contribution of twinning to plastic strain under the influence of large hydrostatic stress constraints differs from the contribution of twinning to plastic strain with no hydrostatic stress constraint, several tensile specimens were made with their long axis in the longitudinal direction of the sheet stock. These were elongated to plastic strains of 2, 3 and 5.5%, sectioned and polished for twin volume measurements. The results of these measurements are shown in Fig. 35. Although the number of strain levels chosen does not allow the same detailed description of twinning behavior as was possible in the compression tests, it was still possible to say that at equivalent strains, the amount of twinning in all
three materials was much less in tensile elongation than in compression under the influence of large hydrostatic stress constraints. The same decrease in twins volume with increasing aluminum content also prevailed.

There was a marked tendency for those regions of the specimens that did not deform by twinning to develop kinks. Kinking was particularly evident in the highly constrained compression specimens. The appearance of kinking with the metallographic techniques used to observe the microstructures was that of diffused variations in the gray tone within a grain. The kinks had neither the characteristic sharp boundaries nor the great difference in gray tone associated with the large changes from matrix crystallographic orientation typical of twins. Figure 36a and b are included to show that the variation in gray tone within a grain that reveals kinking was definitely not caused by surface rumpling. In Fig. 36a the large grain containing a \(\{1122\}\) twin with several secondary twins also displays a wide range of gray tones which might be thought to be shadowing resulting from slight oblique illumination of a rumpled surface. The same area was photographed in Fig. 36b using a Nomarsky interference-contrast objective which shows with good sensitivity surface irregularities such as polishing scratches, small dimples and pits due to preferential removal of small retained \(\beta\)-phase particles, and differences in elevation of grains of various orientations. Figure 36b shows quite clearly that the heavily kinked grain of Fig. 36a is flat and thus the variation of gray tone within the grain cannot be the result of surface undulations.
Kinking in compression specimens of the three material composition can be clearly seen in Fig. 29(f), 30(f) and 31(f) which show the three materials compressed to about 5.7% residual plastic strain.

Kinking was more pronounced with increasing aluminum content and was also noticeable at smaller strains in the higher aluminum alloys. The same level of kinking was judged to occur at 3% strain in the commercially pure titanium, 2.9% strain in the titanium-five atomic percent aluminum alloy and 2.7% strain in the titanium-ten atomic percent aluminum alloy. Kinking was first noticeable at strains of the order of 1.5%.

Some specific observations can be made on the deformation structures developed in these materials under highly constrained thickness compression and under simple longitudinal extension. Titanium has been observed to twin on \{10\bar{1}2\}, \{11\bar{2}1\} and \{11\bar{2}2\} planes at room temperature\(^{21,22}\) and in addition on \{11\bar{2}3\} and \{11\bar{2}4\} at low temperature\(^{23}\) and on \{10\bar{1}1\} planes at high temperature.\(^{18}\) These twins can be fairly well distinguished on the basis of morphology and by single surface trace analysis augmented by a polarized light metallography technique.\(^{46}\) The c/a ratio of the materials was in the range from 1.588 to 1.595. Twins of the type \{10\bar{1}2\} and \{11\bar{2}1\} were expected for grains oriented with c axis in tension and those of the type \{11\bar{2}2\} and \{11\bar{2}4\} expected for c axis in compression. At room temperature the \{10\bar{1}2\} and \{11\bar{2}2\} twins are most likely to be confused, however, noting the direction of basal plane trace in any grain observed (the orientation of this trace could be determined within 10° using the technique
described in Ref. 46 with a quartz first order red plate in the optical path of the metallograph) and the amount of polarization effect the grain displayed, the type of twin favored by the grain orientation with respect to the stress state could be found.

The deformation structures for the three materials compressed under high hydrostatic stress constraint to about 5.5% strain are as follows. The commercially pure titanium specimens twinned primarily on the \{10\bar{1}2\} planes, with \{11\bar{2}2\} and \{11\bar{2}1\} twinning occurring with decreasing frequency. The volume fraction of \{11\bar{2}2\} twins was considerably greater than \{11\bar{2}1\} twins which were occasionally seen to propagate across grain boundaries with an expected shift in direction. The \{10\bar{1}2\} twins were quite broad and a large amount of coalescence between neighboring twins of this type occurred, in some cases converting most of the parent grain to a new orientation. The grain A in Fig. 37a is an example of this showing two habits of light gray \{10\bar{1}2\} twins in a dark gray grain. Twin intersections of all types described by Rosi \cite{47} were observed. Of particular interest is grain B of Fig. 37b in which two \{11\bar{2}2\} twins intersect by forming constrictions while the narrow twin on the right, probably type \{11\bar{2}1\}, intersects a \{11\bar{2}2\} twin with no effect other than a change in direction within the intersected twin. Secondary twinning was very common as is evident in the three parts of Fig. 37, and occasionally a zig-zag form of secondary twinning occurred as shown at C in Fig. 37c. The secondary twins were most likely \{10\bar{1}2\} in the broad \{11\bar{2}2\} twin. The small particles of retained
\( \beta \)-phase had little effect in retarding the advance of the twin boundaries, in marked contrast to their effect on grain boundaries that resulted in the somewhat irregular shape of the grain boundaries in this material.

The titanium-five atomic percent aluminum alloy differed in deformation structure in that there was a greater tendency for grains twinning on \( \{10\bar{1}2\} \) planes to convert completely to the new orientation as seen at A in Fig. 38a. Those grains oriented for twinning on \( \{11\bar{2}2\} \) displayed a marked preference for deformation by the formation of kinks as seen at B in Fig. 38a and in Fig. 38b. Where \( \{11\bar{2}2\} \) twinning occurred it was generally heavily re-twinned as at C in Fig. 38a and at A in Fig. 38c. Figure 38c also shows an example of \( \{11\bar{2}1\} \) twinning, which was very rare, at B. As a result of the tendency for kinking in preference to \( \{11\bar{2}2\} \) twinning, those grains that had twinned on \( \{10\bar{1}2\} \) planes displayed no secondary twinning but rather continued to deform by kink formation. There was also considerable evidence of lattice rotation and kinking at grain boundary nodes.

The deformation structure of the titanium-ten atomic percent aluminum alloy was similar to that of the five atomic percent alloy is displaying very little \( \{11\bar{2}1\} \) twinning. There were no indications of secondary twinning and few twins traversed one another. Of those grains twinning on \( \{10\bar{1}2\} \) planes, none were seen as completely twinned as in the five atomic percent alloy. These grains, as at A in Fig. 39a
displayed somewhat more lattice bending following appreciable twinning. In this figure the twins sloping upward to the left are seen to change to a darker gray. The matrix material between them at the upper left hand boundary is lighter than the major part of the parent grain. If the twins remained the same color along their length the lightness of the matrix material might be due to local accommodation kinking, however the simultaneous variation of shading in the twins and matrix is strongly indicative of bulk lattice bending in the entire grain. Bands of heavy shear developed as at B in Fig. 39a. Figure 39b shows another example of simultaneous \{10\overline{1}2\} twinning and kinking. Occasionally primary kinking was insufficient to allow continued deformation under polycrystalline constraints and secondary kinking, as described by Rosi,\(^2\) of a feathery, duplex nature developed. An example is given at A in Fig. 39c.

The deformation structures produced by tensile strain to about 5.5\% plastic strain were considerably cleaner for each alloy than those produced by the same plastic strain in constrained compression. Observations were made on sections taken through the thickness of the original sheet and parallel to the tensile axis as this orientation displays the same thickness contraction and lateral extension as was the case for section prepared from the compression specimens. The commercially pure titanium twinned again primarily on \{10\overline{1}2\} planes, and these twins occasionally coalesced or contained secondary twins although there were few twins per grain. Fewer \{11\overline{2}2\} twins appeared, but several were in zig-zag configurations, while \{11\overline{2}1\} twinning was almost absent. There was also no indication of kink formation. The titanium-five atomic percent alloy
retained the same ranking of twin occurrence. The \{10\bar{1}2\} twins were narrower than in the unalloyed titanium and showed little coalescence as well as no secondary twinning. Twin intersections were infrequent and kinking was also absent in this alloy. The titanium-ten atomic percent alloy showed very little twinning and then only on the \{10\bar{1}2\} planes with no coalescence or secondary twinning. In this alloy a small amount of early stage kink formation was noted.

A comparison between the simple tension and constrained compression tests on the basis of applied shear stress is more revealing. The maximum imposed shear stress for simple tension is \(2\tau_{xy} = \sigma\). For the case of constrained compression, Thomsen, Yang and Kobayashi\(^2\) give the expression for plane strain compression of a slab,

\[
\frac{P_{\text{avg}}}{\sigma} = \frac{2}{\sqrt{3}} \frac{h}{\mu l_f} \left\{ \frac{[1 + (2\mu x_0/h)]^2 + 1}{4\mu} - 1 \right\} \tag{23}
\]

where \(P_{\text{avg}}\) is the average normal stress on the top and bottom surfaces of the specimen and the other symbols are given in Section II D. Rearranging and making use of the expression

\[
2\tau_{xy} = \sigma_y - \sigma_x = 2/\sqrt{3} \, \sigma \tag{24}
\]

and including the effect of compression strain \(\varepsilon\) on the dimensions of the specimen:

\[
2\tau_{xy} = (P_{\text{avg}})_0 \, \mu^2 \left( \frac{l_f}{h} \right)_0 \left( \frac{1}{(1-\varepsilon)} \right) \frac{\frac{1}{[1-\ln(1/2\mu + \mu(l_f/h)_0 \, \frac{1}{(1-\varepsilon)^2})]^2 + 1-4\mu}}{[25]}
\]

where \((P_{\text{avg}})_0\) and \((l_f/h)_0\) are measurements based on the unstrained dimensions of the specimen.
The measured twin density for the tension and compression specimens are shown as a function of the applied shear stress as found from the above expression in Fig. 40. The commercially pure titanium specimens show little effect of hydrostatic pressure on the development of twins although the compression data are slightly above the tension data. The two alloys are however sensitive to the hydrostatic component of the applied stress. The offset of the compression and tension data points at the same level of twin volume cannot be accounted for by the level of accuracy of the equations used for the compression data although the compression data is probably shown at high shear stress than actually prevailed. A more exact comparison is difficult due to preferred orientation of the sheet stock. The rate of increase in volume of twinned material with increase in shear stress appears to diminish with the addition of aluminum but does appear to have similar rates for tension and compression for each of the alloys.

D. Microstructural Changes Produced by High Strain Rate Deformation

The target discs used in the light gas gun experiments for the measurement of stress-time histories in plane wave loading were recovered for detailed microstructural observations. An inherent requirement for the observed microstructures to be characteristic of the loading and unloading history of the impact process is that the target discs should not be deformed during the deceleration from velocities on the order of 300 ft/sec. The target assembly and projectile were decelerated in a ten foot long catcher, shown at the far right hand side of Fig. 3, that was filled with shredded parachute cloth. The flying plate and target discs were recovered in very good condition. The target discs were not perfectly
flat due to a shallow circular depression on the back surface where the quartz gauge had been situated and a corresponding level raised area on the impact surface. Aside from these distortions resulting from the target assembly configuration, there was no evidence of distortions caused by the recovery technique. An example of a recovered flying plate and target is shown in Fig. 41. The thin disc is the flying plate and the thick disc is the target. The discs were recovered nested in the damaged heavy wall brass cup, shown on the right, and held in place by the remains of the front of the projectile, shown on the left. The other flying plates and targets were recovered in similar condition. The targets were cast in an epoxy to prevent damage due to clamping and sectioned for metallographic observation.

1. **High Strain Rate Deformation in One-Dimensional Strain**

   Observations were made of the target cross sections at a region where the deformation had occurred under conditions of one-dimensional strain during loading by the impact generated elastic and plastic compression waves as well as unloading by the reflections of these waves from both the flying plate rear surface and the target rear surface. This region was at the mid-thickness of the targets and about 4 mm toward the center from the location of the edge of the quartz gauge. A quantitative survey of twin densities throughout the target indicated that the effect of the difference in acoustic impedance between the quartz gauge and the potting epoxy at the rear surface of the target had dissipated by the time this perturbation in the state of one-dimensional strain had reached the region investigated.
In the state of one-dimensional strain, the unloading process is not always completely elastic. This can be seen by considering the idealized case shown in Fig. 42 which describes the loading and unloading paths for the normal stress \( \sigma_x \), one of the two equal tangential stresses \( \sigma_y \), and the hydrostatic pressure \( p = \frac{1}{3} (\sigma_x + 2\sigma_y) \). The initial loading is elastic along path OA and OB until the difference between the normal and lateral stress is equal to the flow stress \( \sigma_f \). Loading continues along paths AC and BD with the difference between \( \sigma_x \) and \( \sigma_y \) equal to the flow stress for each state of plastic strain. The initial unloading paths are elastic for \( \sigma_x \) along CEF and for \( \sigma_y \) along DE-. The stress difference between \( \sigma_x \) and \( \sigma_y \) decreases to zero at point E and then increases again until the flow condition is reached at point F. From point F to point G unloading is accompanied by plastic strain. The positions of point F depends upon the maximum stress level, point C, and the unloading characteristics of the material. In the absence of a Bauschinger effect, point F will lie at negative values of \( \sigma_x \) for \( \sigma_x \) at point C less than twice \( \sigma_x \) at point A and will lie at positive values of \( \sigma_x \) for \( \sigma_x \) at point C greater than twice \( \sigma_x \) at point A. Thus plastic strain can occur on unloading if the maximum normal stress is greater than twice the Hugoniot Elastic Limit, the normal stress at which yielding occurs during loading. If a Bauschinger effect is present, point F lies closer to point E and plastic strain may occur during unloading even though the maximum normal stress was not greater than twice the Hugoniot Elastic Limit. The plastic strains on loading and unloading were calculated to characterize the deformation history of the regions observed. Because the maximum normal stress in the materials subjected to one-dimensional
strain loading was close to twice the Hugoniot Elastic Limit, the usual expression \( \varepsilon_p = \frac{4}{3} \ln \frac{V}{V_0} \) where \( V_0 \) is the initial specific volume and \( V \) is the specific volume at maximum compression was not used. This expression neglects strength effects, which constituted a large part of the material behavior in these tests. Instead the plastic strain on loading was taken as \( \varepsilon_p = \frac{2}{3} (\varepsilon_C - \varepsilon_A) \) of Fig. 42 and the plastic strain on unloading as \( \varepsilon_p = \frac{2}{3}(\varepsilon_F - \varepsilon_G) \) where point \( G \) is the final position reached in the unloading process.

The magnitude of the Baushinger effect in these materials was not determined and so two values of unloading plastic strain were calculated, one for the assumption of no Baushinger effect and one in which the Baushinger effect was estimated by taking the flow stress on unloading to be equal to the flow stress in simple tension at low strain rate. The results of these calculations are given in Table K. The reason for the large values of unloading plastic strain for the titanium and titanium five atomic percent aluminum experiments as compared with the loading plastic strain is that in these two cases the rear surface of the flying plate was in contact with a vacuum and the final state at point \( G \) of Fig. 42 was -5 and -6 kbar respectively.

The deformation structures produced by high strain rate, one-dimensional compression were quite different from those produced by low strain rate compression under hydrostatic constraint. The volume of twinned material was much greater in the high strain rate case and equivalent to the twinned volume found at low strain rate after between three and six times greater plastic strain. It is also interesting to note that the volume of twinned material near the impact surface was not signifi-
cantly greater than that further through the target thickness. In the case of the commerically pure titanium target, shown in Fig. 43, the twinned volume was about 4.5%, which is an order of magnitude greater than the twinned volume at the same level of plastic strain in low strain rate compression. The bright spots on these micrographs are the result of β-phase pitting during electropolishing. The microstructure at the same volume of twinned material produced by slow compression is shown in Fig. 29(e). It is evident from Fig. 43(a) that the high strain rate deformation produced a much more uniform distribution of twins which were also very thin. There were few grains that had not twinned and the twinning was fairly uniform within a grain, occurring on several habits, as is more readily seen in the Fig. 43(b) and (c). There was no coalescence of twins and also little secondary twinning however a large amount of twin propagation across grain boundaries was noted. The \{10\overline{1}2\} twins were generally thinner than the \{11\overline{2}2\} twins and twins of type \{11\overline{2}2\}, \{11\overline{2}4\} and \{\overline{1}1\overline{2}1\} were abundant. Although at this level of twinned volume, kinking was noted in slow compression, none was evident in the high strain rate case.

The deformation structure produced in the titanium-five atomic percent aluminum alloy is shown in Fig. 44. The twinned volume was 2.5% and was close to that shown for slow compression in Fig. 30(f). This volume fraction of twin was about five times that produced by slow compression to the same level of plastic strain. The deformation structure was again composed of fine twins more uniformly distributed and on several habits per grain. Few grains were not twinned. The ranking of twins occurrence was again primarily \{11\overline{2}2\} and \{11\overline{2}4\} about equally
followed by \{1121\} and last \{10\overline{1}2\}. No kinking or coalescence was noted, however, there was substantial secondary twinning in \{11\overline{2}2\} twins, as can be seen near the center of Fig. 44b. In Fig. 44c the dark grain just above center has twinned almost completely on \{1121\} while the darker grain just below it has twinned primarily on \{10\overline{1}2\} with a small amount of \{1121\} twinning. Twinning on \{1121\} was often found alone but \{10\overline{1}2\} twinning was generally accompanied by \{1121\} twins.

The titanium-ten atomic percent aluminum alloy, shown in Fig. 45, contained about 1.3% twinned volume. For slow compression to the same strain no twinning occurred. An example of the deformation structure at the same volume of twinned material produced in slow compression was given in Fig. 31(f). A number of grains gave no appearance of twinning but they also showed no evidence of kink formation. Twinning on \{11\overline{2}2\} and \{11\overline{2}4\} again predominated but in this material \{10\overline{1}2\} twinning was more evident than \{1121\} twins and individual twins tended to tranverse the grains completely.

Comparing again on the basis of applied shear stress, the twin volume measured in the recovered target is shown in Fig. 40 as the horizontal bar for each alloy. The range of shear stress values indicated by the bars correspond to the maximum and minimum stress values shown for the dynamic strain curves, \(\sigma_s - \varepsilon_s\) of Figs. 23, 24 and 25 after the indicated yield stress. The high aluminum alloy required a greater applied shear stress to develop the same twin volume than did the unalloyed
titanium. The dynamically strained specimen appeared to be in the same neighborhood of the constrained compression samples with the unalloyed titanium showing the greatest deviation.

2. High Strain Rate Deformation Under the Influence of Edge Effects

The deformational behavior of the materials was also investigated at the corner formed by the impact surface of the target and the unrestrained side. At this location the strain was not one-dimensional due to the influence of the lateral free surface. The reduction of normal stress due to lateral free surface effects has been more fully described by L. D. Bertholf and C. H. Karnes.⁴⁸ The lateral stress at the free surface must remain equal to zero, thus the loading path along the lateral free surface must be closer to the high strain rate, one-dimensional stress curves of Figs. 23, 24, and 25. The maximum normal stress in the waves propagating through the target near the lateral free surface was then less than half that of waves propagating under one-dimensional strain conditions, \( \sigma_x, \epsilon_x \). The material velocity increases associated with waves propagating under the two different conditions were unequal with material away from the lateral free surface attaining greater velocities than that neighboring the free surface. A fan of heavy shear developed at approximately 45° to the lateral free surface within which large scale shear stresses were produced as the velocity differences were equilibrated.

The resulting deformation structures for the three materials are shown in Fig. 46 where (a) is the commercially pure titanium, (b) is the five atomic-percent aluminum alloy and (c) is the ten atomic-percent aluminum alloy. The loading conditions were comparable in that the peak
stress in one-dimensional strain was about twice the Hugoniot Elastic Limit in each case. The volume of twinned material again decreased with increasing aluminum content and was 20%, 12%, and 10% respectively. The deformation structures were comparable to those produced in one-dimensional strain in that the occurrence of twinning was fairly evenly distributed. Twinning on \{10\overline{1}2\}, \{11\overline{2}1\}, \{11\overline{2}2\} and \{11\overline{2}4\} was noted in comparable quantity however \{10\overline{1}2\} and \{11\overline{2}2\} twins were thicker than in the one dimensional strain cases and thus contributed more to the volume of twinned material. Coalescence of twins was evident and in some grains where only \{10\overline{1}2\} twinning occurred the grains were almost completely transformed to twinned material. Twinning occurred on several habits in each grain and many of the \{11\overline{2}2\} twins contained secondary twins. The thicker twins had wavy boundaries and the thinner twins of type \{11\overline{2}1\} and \{11\overline{2}4\} had become distorted from their initially straight shapes. Kinking was evident in the three materials, increasing with aluminum content. Almost every grain had twinned in the commercially pure titanium sample but some grains deformed solely by kinking in the high aluminum alloy. Thus it appeared that at high strain rate and on the border of the region undergoing one dimensional strain the deformation structure took on a character that was intermediate between that of one-dimensional strain compression and of low strain rate highly constrained compression.
IV. DISCUSSION

A. Dynamic Yielding in One-Dimensional Strain

The stress levels at which the three alpha-phase titanium-aluminum alloys yielded in one-dimensional strain loading were given in Table H as the Hugoniot Elastic Limit. These values were obtained by impedance calculations based on quartz gauge records made on targets between 4.0 and 4.3 mm thick. The values of the dynamic flow stress in one-dimensional strain were converted to dynamic flow stress in one-dimensional stress by the use of Eqs. [20] and [21] and shown in Figs. 23, 24 and 25 for the three material compositions. Comparison of the dynamic and quasi-static one dimensional stress curves in these figures indicate an appreciable strain rate sensitivity of flow stress.

The dynamic yield stresses are about twice the quasi-static yield stresses. The actual increase in the yield stress under dynamic conditions is shown by the solid line in Fig. 47. Here the strain rate sensitivity is represented as the ratio of the dynamic yield stress to the quasi-static yield stress as a function of quasi-static yield stress. The ratios exhibit a decrease in strain rate sensitivity with increasing base strength level of the material. A similar decrease in strain rate sensitivity with alloying to increase base strength level has been cited for aluminum alloys.49 The strain rates during dynamic testing, shown also in Table H, were well above \( \dot{\epsilon} = 10^3 \) at which viscous damping of dislocations is considered to be rate controlling in aluminum.50

As further evidence for ascertaining the behavior of the titanium alloys in dynamic yielding, quasi-static tensile yield stresses were obtained for the alloys at 77°K. The ratios of yield stress at
low temperature to that at room temperature were again near two, with specific values shown by the dashed line in Fig. 47. A general strain rate and temperature dependence of flow stress has been given in which the flow stress at any strain rate reaches a maximum at 0°K for all but viscous damping controlled deformation. If the strain rate imposed produced flow stresses in excess of the 0°K, non-viscous damping values, then the deformation must have occurred by a viscous damping mechanism. The implication is that for the dynamic yield stress of the titanium aluminum alloys investigated to have been the result of viscous damped dislocation motion, the values of yield stress displayed should have been greater than the yield stress at 0°K and definitely greater than the yield stresses at 77°K. That this was not the case can be seen in Fig. 47 where the ratios of dynamic room temperature and quasi-static 77°K yield stresses to the quasi-static room temperature yield stresses were in the same range and actually crossed within the composition range of the alloys.

The yield stress at 0°K for flow mechanisms that are not viscously damped can be estimated for each alloy by assuming a linear increase in yield stress with decrease in temperature at the constant quasi-static strain rate of ε=10^{-3}. The extrapolated values obtained for σ_Y at 0°K for the three alloys in increasing order of aluminum content are respectively 7.5, 10.1 and 14.6 kbar. In terms of the ratio of σ_Y at 0°K to σ_Y at 298°K and ε=10^{-3} these are respectively 2.67, 2.38 and 2.15. As can be noted from Fig. 47, the yield stress ratios for dynamic yield are less than the estimated yield stress ratios at 0°K. The rate controlling mechanism for dynamic yield in these materials was then evidently not viscous damping but must be a thermally activated mechanism.
The specific rate controlling mechanism active in titanium and its aluminum alloys is currently under investigation. As discussed by Tung and Sommer, both a Fleisher type dispersed barrier hardening (probably thermal overcoming of effective oxygen interstitials) or a Pierls-Nabarro double kink nucleation process have been suggested as the rate controlling mechanism for deformation at and below room temperature.

The decrease in the strain rate and temperature sensitivity of the yield stress ratios with increasing aluminum content can be described as due to the effect of aluminum on the athermal part of the flow stress. Evans has observed that the addition of aluminum to titanium causes an increase in only the athermal part of the flow stress. The combined flow stress at high aluminum content would then be composed to a larger extent of the strain-rate and temperature insensitive athermal flow stress than at low aluminum content. Then if the thermal component of the flow stress were unaffected by aluminum addition, the observed decrease in flow stress (yield stress) would be expected. The decrease in strain rate and temperature sensitivity of the combined flow stress shown in Fig. 47 may then not be solely the result of the effect of aluminum additions on the thermal component of the flow stress. The actual behavior of the thermal component of the flow stress with aluminum addition could not be determined because the data was insufficient to extract the athermal component, which apparently changes below 650°K, from the combined flow stress.

In addition, a number of slip systems may have been active during dynamic yielding. The systems observed in alpha titanium aluminum
alloys are the \(\langle 11\bar{2}0\rangle\) direction on \((0001)\), \(\{10\bar{1}0\}\), and \(\{10\bar{1}1\}\) planes as well as \(\langle 11\bar{2}3\rangle\) on \(\{10\bar{1}1\}\). The \(\langle 11\bar{2}3\rangle\) \(\{10\bar{1}1\}\) system was of secondary importance and was less notable with increasing aluminum content. The \(\langle 11\bar{2}0\rangle\) \((0001)\) system also played a minor role. The \(\langle 11\bar{2}0\rangle\) \(\{10\bar{1}0\}\) was active at yielding with slip on \(\{10\bar{1}1\}\) occurring at higher strains. Orava et al\textsuperscript{2} found in addition that \(\langle 11\bar{2}0\rangle\) \(\{10\bar{1}0\}\) was the dominant slip system at 77°K. If as is generally true for metals, the high strain rate behavior of titanium is similar to the low temperature behavior, then slip on \(\langle 11\bar{2}0\rangle\) \(\{10\bar{1}0\}\) is indicated as the unit slip dislocation system responsible for dynamic yielding.

Yielding may not be due entirely or predominately to unit slip dislocation motion, however. Rosi and Perkins\textsuperscript{53} found that at 77°K in coarse grained specimens deformation was primarily the result of twinning. The supposed similarity between high strain rate and low temperature deformation would then point to the possibility of twin formation as a cause for dynamic yielding. Rose et al.\textsuperscript{23} have shown the following twin systems to operate at 77°K: \(\{10\bar{1}2\}\), \(\{11\bar{2}1\}\), \(\{11\bar{2}2\}\), \(\{11\bar{2}3\}\), and \(\{11\bar{2}4\}\) with \(\{11\bar{2}4\}\) having a dominant role.

Twin dominated deformation has been cited for many materials at high strain rate\textsuperscript{54}. Iron in various forms has been the subject of a number of investigations. C. S. Smith\textsuperscript{55} suggested that twinning in iron subjected to shock compression occurred solely at the base of the slower moving deformational wave that follows the elastic precursor. This idea is not consistent with the observation\textsuperscript{56} that the density of twins increases in iron as the maximum stress level due to shock loading increases. The plastic strain during loading increases with increasing maximum stress and the above observation implies that twin formation and growth occurs through
the entire plastic compression wave. Other investigators\textsuperscript{57,58} have shown that in fcc metals twinning will not occur in shock compression until a minimum peak stress level is reached which increases with increasing stacking fault energy. Thus the initial deformation of these materials from dynamic yielding through the first part of the plastic compression wave was accomplished by dislocation motion, with twin formation occurring after appreciable cell formation. The requirement for the development of a proper substructure before twinning occurred is consistent with the low strain rate, constrained compression work of this paper because the strain at which twinning occurred increased with aluminum content. The addition of aluminum to $\alpha$-titanium has been shown by Blackburn and Williams\textsuperscript{17} to reduce the tendency for cell formation giving rise to almost planar dislocation arrays in a ten atomic percent aluminum alloy at the same strain which produced tangles and cells in commercially pure $\alpha$-titanium.

The existence of an "incubation strain" preceding twin formation as was evident in the low strain rate constrained compression tests may indicate that under dynamic compression conditions the yield and initial deformation occur by unit slip dislocation motion only, accompanied later on in the plastic compression wave by twin formation. The evidence for the dominant deformation mode at low temperature (77°K) that could be considered by similarity to dominate at high strain rate is conflicting. Rosi and Perkins\textsuperscript{53} indicated twinning to dominate for their coarse grained specimens at 77°K while Orava et. al.\textsuperscript{54} found no twinning at that temperature for strains up to 10%. The result was explained by noting that a fine grain size and a high impurity level tend to inhibit twinning in
titanium. The grain size of their specimens was 4 to 5 μ, about one-tenth that of this investigation, and their material also had a higher interstitial content than the material of this investigation. Thus the results of Orava et al. should not be pertinent to this investigation.

A recent report by Odinokova 25 for titanium and a titanium 7 atomic % aluminum alloy with 150 to 200μ grain size and unspecified impurity level indicated that twinning increased with increasing strain rate in the unalloyed titanium. Deformation was almost entirely due to twin formation at 77°K. The alloy deformed solely by slip at low and moderate strain rate and at 77°K. The alloy only twinned during impulsive loading. The grain size in the above work was about three times that of the present work, rendering the two comparable, except for the unknown comparison of impurity levels. The fact that the high aluminum alloy of the present work, which exhibited the largest incubation strain for slow constrained compression, showed substantial twinning at room temperature dynamic compression to a plastic strain at which no twinning was observed in slow compression suggests that during dynamic yielding twinning may have accompanied or even preceded significant unit slip dislocation motion.

It is not unusual to suggest that dynamic yielding was caused by twin formation. Rohde 59 observed in iron that the effect of temperature on the dynamic yield stress and the rate of decay of the yield stress with propagation distance could not be described by motion of unit slip dislocations. He found that mechanical twinning was responsible for the yield stress behavior.
A yield drop which might be associated with a burst of twin formation was not observed on the stress histories measured in the present investigation, however a yield drop could have been masked by the averaging effect of the quartz gauges on the input stress waves. This effect occurs when the gauge is impinged upon by stress waves that are not coplanar with the front surface of the gauge, as was the case in this investigation due to tilted impacts. The results of this investigation are not conclusive in revealing at what stage during dynamic plastic flow twin formation began. Additional experiments would be required in a lower range of impact velocities in excess of the velocity required to exceed the Hugoniot Elastic Limit, thereby varying the level of maximum plastic strain while maintaining, to a first approximation, the maximum applied shear stress.

B. Twinning

The twinning behavior of the three alloys investigated can be understood if the requirement for twin formation involves both a critical local shear stress and the development of a proper microstructure, such as twin nuclei, by dislocation interaction. At high strain rate and low temperature the lattice stress and hence the energy of the lattice is higher at any level of plastic strain than in the low strain rate-room temperature case. The amount of dislocation activity required to create twin nuclei would be correspondingly less. Nuclei may be available in the low strain rate-room temperature case at approximately the same level of plastic strain as at high strain rate, however the lattice may be required to undergo larger plastic strain to develop the local shear stresses required for twin formation. The slow compression tests at two levels of
The hydrostatic constraint demonstrate the lattice energy requirement for twinning. The specimens of all three alloys showed larger twin volumes at the same residual plastic strain in the specimens having the greater constraint and thus greater hydrostatic stress component and larger lattice strain energy. Twinning was more favored in the unalloyed titanium because the dislocation substructure is more complex than in the high aluminum alloy. The planar nature of plastic strain in the high strain rate tests also promoted a more uniform twin distribution because a more homogeneous substructure was developed.

The relative abundance of the twins of type \{10\overline{1}2\}, \{11\overline{2}1\}, \{11\overline{2}2\}, \{11\overline{2}4\} in the three types of loading can also be explained. The amount of any twin type is sensitive to the texture of the sheet from which the specimens were taken and the orientation of the tensile and compression stresses with respect to the sheet texture. Rolled titanium sheet generally develops a texture with basal plane normals inclined about 25° from the perpendicular to the plane of the sheet, in the rolling directions. Thus the dynamic and slow compression specimen had the compression axis relatively parallel to the c-axis texture while the tensile specimens had the tensile axis relatively perpendicular to the c-axis texture. Twins on \{10\overline{1}2\} and \{11\overline{2}1\} produce extensions along the c-axis and twins on \{11\overline{2}2\} and \{11\overline{2}4\} produce contractions along the c-axis. The net geometrical effect was then to promote \{11\overline{2}2\} and \{11\overline{2}4\} twins in both the tension and compression specimen.

For the low strain rate tension and compression tests, the \{10\overline{1}2\} mode was dominant. This mode has a twinning shear of 0.156, the lowest of the four twin modes observed. The \{11\overline{2}1\} mode, with the
greatest twinning shear of 0.624 was little noted in the tension specimen but slightly more notable in the compression specimens. This may be due to higher lattice strain energy in the constrained compression tests, as can be seen from Fig. 40. In those grains properly oriented for \{11\overline{2}2\} twinning, slip was preferred in the tension case, while slip and twinning were observed in the compression specimens. In compression the slip developed into kinks, probably as a result of the constrained flow. Kinking was more evident at higher aluminum composition. This may be a result of the interference created by the substitutional aluminum atoms with the complicated atomic shuffles \(^{19}\) required in the formation of the \{11\overline{2}2\} twins in the double lattice HCP structure.

In the high strain rate compression specimens, the \{10\overline{1}2\} mode was less abundant than the \{11\overline{2}1\} mode. The \{11\overline{2}1\} mode has a twinning shear between three and four times that of the \{10\overline{1}2\} mode, but requires no shuffles. Thus in the case of high strain rate, plane wave loading, the \{11\overline{2}1\} mode produces a larger shift in the lattice per layer of twin and consequently a more rapid rate of energy release. The great abundance of \{11\overline{2}2\} and \{11\overline{2}4\} twins was favored by the texture of the specimens and even though the twinning shears for the two modes are approximately equal, 0.236 and 0.219 respectively, the complexity of the shuffles for the \{11\overline{2}2\} twin is not as great as that for the \{11\overline{2}4\} twin. This may explain why the \{11\overline{2}2\} twin was more in evidence. The greater tendency to form the \{11\overline{2}2\} and \{11\overline{2}4\} twins at high strain rate is probably associated with the increase in lattice energy due to the large strain rate sensitivity of the flow stress and the requirement for rapid deformation, made possible by the large twinning shear as compared with the low strain rate compression tests.
Kinking was not evident for high strain rate, plane wave loading because the dynamic straining allowed little time for the cooperative motion of large groups of like dislocations on parallel planes as is required for the development of kinks.

C. Dynamic Loading Path

The stress waves produced in a solid by the impact of a flat plate can be conveniently described by a front running elastic wave bringing the material to the Hugoniot Elastic Limit followed by a second, slower deformational wave, moving at or above the bulk sound velocity and bringing the material to the maximum state of stress. This description treats the material as an elastic-perfectly plastic medium with no work hardening and limits the strain rate effects only to the height of the elastic wave. This two-wave description was one of the ways used in the present investigation to present the dynamic stress-strain curves of the three materials tested. The dynamic one-dimensional strain loading paths were converted to dynamic one dimensional stress paths for convenient comparison with the quasi-static, one dimensional stress loading paths.

It was also possible to describe the dynamic strain loading paths on the stress-strain diagrams of Figs. 23, 24 and 25 by three and in one case four wave loading paths. The three wave description displayed an increasing flow stress after the elastic limit had been reached, which is characteristic of work hardening. The slope of the work hardening, second wave also decreased with the addition of aluminum to the un-alloyed titanium, Figs. 23 and 24, behaving in a similar way to the
work hardening portion of the quasi-static, simple tension curves for these materials.

It is commonly observed in metals that the rate of work hardening decreases with increasing strain rate. Thus it is surprising that the slope of the second wave for the un-alloyed titanium is larger in the dynamic loading curve than the slope of the quasi-static loading curve over the same strain interval. This effect can be explained by noting the four wave loading path of Fig. 25 for the high aluminum alloy. In this case the comparatively low tilt of the flying plate at impact with the target allowed a more detailed treatment of the loading path. Here the greatest amount of work hardening is seen to occur just after the elastic limit was exceeded. The third wave has a very small slope, almost behaving as a perfectly plastic medium. It is suggested that were it possible to represent the loading paths of the other two materials by four waves, similar curves would have been found. Nonetheless the treatment of the loading paths as three or four wave structures shows that the flow stress of the material increased with strain after the elastic limit was exceeded. In reality the loading paths followed by the materials was a smooth curve with a knee at the Hugoniot Elastic Limit and the three and four wave paths are piecewise linear approximations to the correct curves.

The last segment of the loading paths shows a decreasing flow stress in increasing strain. Near the top of the deformational wave the strain rate decreases continuously but somewhat abruptly to zero because the experiments were performed at peak stresses below those necessary to develop true shock waves. Thus if the material were strain-rate sensitive,
as is evident from the increase in elastic limit in dynamic strain as compared with quasi-static strain, then a drop off of flow stress with decreasing strain rate would be expected. The drop off exhibited by the three materials was rather large considering the fact the average strain rate in the last segment of these curves, Table I, is greater than half of the average strain rate in the second to last segment. However as the maximum strain was approached, the strain rate was several orders of magnitude below that at the middle of the entire plastic wave, and close to zero. The final flow stress point was developed under very low strain rate conditions and this is the reason for its comparatively low value. The effect of compressive heating on the dynamic flow stress should not be large since the largest temperature increase, which would be expected for the high aluminum alloy, can be estimated to be about 25°C.41

Thus the strain rate sensitivity of the materials caused the loading paths in dynamic strain to fall off as the maximum strain was reached, and curve down toward the quasi-static loading curves for the materials. This effect has also been seen in aluminum45,63 where the stress drop off fell to the same level on the quasi-static simple tension loading curves. If the three titanium-aluminum alloys of this investigation had actually undergone a drop off large enough to bring the final dynamic one-dimensional stress point down to the quasi-static loading curve, this would mean that the final flow stress measured experimentally was in error by about 65%. This error is too large to be expected and the offset between the quasi-static loading path and the final point on the dynamic one-dimensional stress loading path must be real. Evidently
then, the nature of the residual defect structure introduced by dynamic strain are sufficiently different from that introduced by quasi-static strain to account for the difference in material flow stress.
V. CONCLUSIONS

The experimental results of this study show that the flow stress of α-titanium alloys having an aluminum content between zero and 10 atomic % is not dependent on viscous damping mechanisms at strain rates up to $10^5$-$10^6$ sec$^{-1}$. The rate controlling process is probably a Pierls-Nabarro double kink nucleation or the thermal overcoming of a Fleisher-type dispersed barrier.

The yield stress of the alloys at 298°K is very strain rate sensitive. At a strain rate of $\sim 10^6$ the yield stress is almost twice the value at a strain rate of $10^{-3}$, and is comparable to the yield stress at 77°K at a strain rate of $10^{-3}$. The strain rate sensitivity decreases with aluminum additions.

In high strain rate, one dimensional strain loading, the strain rate sensitivity of the materials gives rise to positive deviations from the loading path predicted from the low strain rate tensile behavior of the materials. Near the end point on the loading path, the decreasing strain rate results in a drop off of this deviation in the direction of the low strain rate behavior. The defect structure produced during the intervening period of high strain rate deformation is sufficiently different from that produced by low strain rate deformation to the same strain that the flow stress of the material strained at high strain rate remains above that of the material strained at low strain rate.

Twinning in the α-titanium alloys is markedly sensitive to aluminum alloying additions as well as the conditions under which strain to any level occurs. In high and low strain rate compression and in low strain rate tension, the volume of twinned material decreases with increasing
aluminum content. This is presumably due to the refinement of the dislocation substructure from cells and tangles to more planar arrays of dislocations with the addition of aluminum. The rate of development of critical twin nuclei is evidently less for the cleaner dislocation substructure that accompanies the addition of aluminum. The volume of twinned material is sensitive to the lattice strain energy at any level of strain. The effective shear stress and the lattice strain energy at the same level of strain increased progressively from quasi-static simple tension to slow compression with a moderate hydrostatic stress component to slow compression with a high hydrostatic pressure component to dynamic, one-dimensional strain compression. The volume of twinned material at the same level of plastic strain also increased in the same sequential order.

Twins were distributed more uniformly at high strain rate than at low strain rate. The relative abundance of the four twinning modes observed also varied with the testing conditions. At low strain rate in tension and compression the \{10\(\overline{1}\)2\} mode was most abundant and varied in appearance with alloy composition. The \{11\(\overline{2}\)2\} and \{11\(\overline{2}\)1\} modes occurred less frequently and in decreasing order. Both modes decreased with aluminum addition and were less evident in tension than compression. In high strain rate compression the \{11\(\overline{2}\)2\} mode was slightly more evident than the \{11\(\overline{2}\)4\} mode and both were more evident than \{11\(\overline{2}\)1\} and \{10\(\overline{1}\)2\}. The relative order of the last two modes was dependent on alloy composition. These results can be understood by considering the magnitude of the individual twinning shears and the relative complexities of the atomic shuffles during twin formation.
Table A. Hot working and annealing conditions.

<table>
<thead>
<tr>
<th></th>
<th>Forging temperature</th>
<th>Rolling temperature</th>
<th>Annealing temperature</th>
<th>time</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti</td>
<td>980°C</td>
<td>735°C</td>
<td>630°C</td>
<td>16 hr</td>
</tr>
<tr>
<td>Ti-5%Al</td>
<td>1095°C</td>
<td>845°C</td>
<td>760°C</td>
<td>16 hr</td>
</tr>
<tr>
<td>Ti-10%Al</td>
<td>1190°C</td>
<td>900°C</td>
<td>780°C</td>
<td>16 hr</td>
</tr>
</tbody>
</table>
Table B. Material composition, atomic percent.

<table>
<thead>
<tr>
<th></th>
<th>Al</th>
<th>Fe</th>
<th>O</th>
<th>N</th>
<th>H</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti</td>
<td></td>
<td>.112</td>
<td>.254</td>
<td>.017</td>
<td>.081</td>
</tr>
<tr>
<td>Ti-5%Al</td>
<td>4.7</td>
<td>.103</td>
<td>.287</td>
<td>.017</td>
<td>.072</td>
</tr>
<tr>
<td>Ti-10%Al</td>
<td>10.1</td>
<td>.105</td>
<td>.251</td>
<td>.020</td>
<td>.057</td>
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</table>


Table C. Microstructure and crystal lattice data.

<table>
<thead>
<tr>
<th></th>
<th>Grain size in</th>
<th>Vol. % retained</th>
<th>Lattice constants, Å</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>mills ± 20%, (μ)</td>
<td></td>
<td>a</td>
</tr>
<tr>
<td>Ti</td>
<td>1.50</td>
<td>38</td>
<td>1.16</td>
</tr>
<tr>
<td>Ti-5 at.%Al</td>
<td>2.02</td>
<td>52</td>
<td>.22</td>
</tr>
<tr>
<td>Ti-10 at.%Al</td>
<td>1.90</td>
<td>48</td>
<td>.13</td>
</tr>
</tbody>
</table>
Table D. Average hydrostatic stress $\bar{H}$ prevailing during the deformational portion of the planar impact loading waves, yield stress $\sigma_y$ in simple tension and ratio of length to height $l_f/h$ for specimens used in slow compression to achieve the same level of hydrostatic stress $\bar{H}/\sigma_y$ as occurred in dynamic compressions.

<table>
<thead>
<tr>
<th></th>
<th>$\bar{H}$</th>
<th>$\sigma_y$</th>
<th>$\bar{H}/\sigma_y$</th>
<th>$l_f/h$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti</td>
<td>15.1 kbar</td>
<td>2.76 kbar</td>
<td>5.48</td>
<td>9.09</td>
</tr>
<tr>
<td>Ti-5 at.%Al</td>
<td>19.2 kbar</td>
<td>4.27 kbar</td>
<td>4.50</td>
<td>7.70</td>
</tr>
<tr>
<td>Ti-10 at.%Al</td>
<td>28.5 kbar</td>
<td>6.8 kbar</td>
<td>4.19</td>
<td>7.26</td>
</tr>
</tbody>
</table>
Table E. Dimensions of specimen used in slow compression.

<table>
<thead>
<tr>
<th></th>
<th>h</th>
<th>$l_r$</th>
<th>W</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti</td>
<td>0.100</td>
<td>0.910</td>
<td>1.050</td>
</tr>
<tr>
<td>Ti-5 at.%Al</td>
<td>0.100</td>
<td>0.770</td>
<td>0.900</td>
</tr>
<tr>
<td>Ti-10 at.%Al</td>
<td>0.100</td>
<td>0.726</td>
<td>0.780</td>
</tr>
</tbody>
</table>
Table F. Ultrasonic velocity measurements.

<table>
<thead>
<tr>
<th></th>
<th>(c_L) mm/\mu sec, ±2%</th>
<th>(c_s) mm/\mu sec, ±2%</th>
<th>(c_B) mm/\mu sec</th>
<th>(\rho_o) gm/cc</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti</td>
<td>6.07</td>
<td>3.13</td>
<td>4.88</td>
<td>4.501 ± .4%</td>
</tr>
<tr>
<td>Ti-5 at.% Al</td>
<td>6.28</td>
<td>3.24</td>
<td>5.05</td>
<td>4.454 ± .2%</td>
</tr>
<tr>
<td>Ti-10 at.% Al</td>
<td>6.49</td>
<td>3.35</td>
<td>5.22</td>
<td>4.374 ± .2%</td>
</tr>
</tbody>
</table>
Table G.

<table>
<thead>
<tr>
<th></th>
<th>Impedance, gm/cm² μsec</th>
<th>Target thickness, mm</th>
<th>Projectile velocity, mm/μsec</th>
<th>Tilt at impact, radians</th>
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</thead>
<tbody>
<tr>
<td></td>
<td>elastic</td>
<td>&quot;plastic&quot;</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Ti</td>
<td>2.73</td>
<td>2.21</td>
<td>4.293</td>
<td>0.208±0.3%</td>
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<tr>
<td>Ti-5 at.% Al</td>
<td>2.79</td>
<td>2.27</td>
<td>4.186</td>
<td>0.258±1%</td>
</tr>
<tr>
<td>Ti-10 at.% Al</td>
<td>2.84</td>
<td>2.31</td>
<td>4.013</td>
<td>0.375±0.8%</td>
</tr>
</tbody>
</table>
Table H.

<table>
<thead>
<tr>
<th></th>
<th>Hugoniot Elastic Limit</th>
<th></th>
<th>Maximum Stress</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>$\sigma_x'$</td>
<td>$\varepsilon_x'$</td>
<td>$\dot{\varepsilon}_x$</td>
</tr>
<tr>
<td></td>
<td>kbar</td>
<td>in./in.</td>
<td>in./in./sec</td>
</tr>
<tr>
<td>Ti</td>
<td>12.1</td>
<td>0.0073</td>
<td>$0.104 \times 10^6$</td>
</tr>
<tr>
<td>Ti-5 at.% Al</td>
<td>16.5</td>
<td>0.0093</td>
<td>$0.45 \times 10^6$</td>
</tr>
<tr>
<td>Ti-10 at.% Al</td>
<td>23.3</td>
<td>0.0128</td>
<td>$2.8 \times 10^6$</td>
</tr>
</tbody>
</table>
### Table I.

<table>
<thead>
<tr>
<th>Material</th>
<th>$\sigma_x$, kbar</th>
<th>Wave velocity, mm/$\mu$sec</th>
<th>Impedance, gm/cm$^2$/$\mu$sec</th>
<th>$\dot{\varepsilon}$, in./in./sec</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti</td>
<td>12.1</td>
<td>6.07</td>
<td>2.73</td>
<td>0.104 x 10^6</td>
</tr>
<tr>
<td></td>
<td>22.8</td>
<td>4.94</td>
<td>2.24</td>
<td>0.046 x 10^6</td>
</tr>
<tr>
<td></td>
<td>25.3</td>
<td>4.23</td>
<td>1.99</td>
<td>0.034 x 10^6</td>
</tr>
<tr>
<td>Ti-5%Al</td>
<td>16.5</td>
<td>6.28</td>
<td>2.79</td>
<td>0.45 x 10^6</td>
</tr>
<tr>
<td></td>
<td>29.0</td>
<td>5.16</td>
<td>2.32</td>
<td>0.058 x 10^6</td>
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<td></td>
<td>32.2</td>
<td>4.45</td>
<td>2.02</td>
<td>0.031 x 10^6</td>
</tr>
<tr>
<td>Ti-10%Al</td>
<td>23.3</td>
<td>6.49</td>
<td>2.84</td>
<td>2.8 x 10^6</td>
</tr>
<tr>
<td></td>
<td>44.0</td>
<td>5.40</td>
<td>2.41</td>
<td>0.158 x 10^6</td>
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<td></td>
<td>48.2</td>
<td>4.81</td>
<td>2.16</td>
<td>0.090 x 10^6</td>
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<tr>
<td>Ti-10%Al</td>
<td>23.3</td>
<td>6.49</td>
<td>2.84</td>
<td>2.8 x 10^6</td>
</tr>
<tr>
<td></td>
<td>27.1</td>
<td>5.86</td>
<td>2.61</td>
<td>0.927 x 10^6</td>
</tr>
<tr>
<td></td>
<td>42.5</td>
<td>5.22</td>
<td>2.33</td>
<td>0.169 x 10^6</td>
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<tr>
<td></td>
<td>47.7</td>
<td>4.82</td>
<td>2.17</td>
<td>0.093 x 10^6</td>
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Table J. Compression tests with ratio of hydrostatic stress to yield stress in the range 4.2 to 5.5.

<table>
<thead>
<tr>
<th></th>
<th>Incubation plastic strain before inset of twinning</th>
<th>Initial rate of twinning, vol. % twin/% plastic strain</th>
<th>Strain at transition in rate of twinning</th>
<th>Final rate of twinning, vol. % twin/% plastic strain</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti</td>
<td>0.0 - 0.1%</td>
<td>0.53</td>
<td>~0.9%</td>
<td>2.0</td>
</tr>
<tr>
<td>Ti-5 at.% Al</td>
<td>0.3 - 0.4%</td>
<td>0.3</td>
<td>~5.5%</td>
<td>&gt;0.9</td>
</tr>
<tr>
<td>Ti-10 at.% Al</td>
<td>2.3%</td>
<td>0.25</td>
<td>&gt;6.2%</td>
<td>-</td>
</tr>
</tbody>
</table>

Compression tests with ratio of hydrostatic stress to yield stress equal to 2.67.

<table>
<thead>
<tr>
<th></th>
<th>Incubation plastic strain before inset of twinning</th>
<th>Initial rate of twinning, vol. % twin/% plastic strain</th>
<th>Strain at transition in rate of twinning</th>
<th>Final rate of twinning, vol. % twin/% plastic strain</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti</td>
<td>~0.1%</td>
<td>&gt;.29</td>
<td>~0.75</td>
<td>1.6</td>
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<tr>
<td>Ti-5 at.% Al</td>
<td>&gt;0.6%</td>
<td>&lt;.47</td>
<td>&gt;1.5</td>
<td>&lt;2.1</td>
</tr>
<tr>
<td>Ti-10 at.% Al</td>
<td>&lt;2.0%</td>
<td>&gt;.05</td>
<td>&gt;2.7</td>
<td>0.6</td>
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Table K. Plastic strains accompanying loading and unloading in one-dimensional strain.

<table>
<thead>
<tr>
<th>Material</th>
<th>Loading $\varepsilon_p$, in./in.</th>
<th>Unloading $\varepsilon_p$, in./in.</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>No Bauschinger effect</td>
</tr>
<tr>
<td>Ti</td>
<td>0.0084</td>
<td>0.0034</td>
</tr>
<tr>
<td>Ti-5 at.% Al</td>
<td>0.0082</td>
<td>0.0030</td>
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<tr>
<td>Ti-10 at.% Al</td>
<td>0.0138</td>
<td>0.0006</td>
</tr>
</tbody>
</table>
ACKNOWLEDGEMENT

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31. Hysol R8-2038, Hysol Division, The Dexter Corp.


33a Channel Industries PK 14-12, 500-100 Volt output into 50Ω load.


40. Dow Corning B276-V9 resin.


FIGURE CAPTIONS

1. Grain size as a function of annealing time at several temperatures. Material initially in as received, hot rolled condition. (a) commercial purity titanium-50A, (b) titanium-5 atomic percent aluminum, (c) titanium-10 atomic percent aluminum.

2. Photomicrographs of titanium-5 atomic percent aluminum alloy in (a) as received condition and (b) annealed condition.

3. Light compressed-gas gun used in high strain rate, plane wave loading experiments.

4. Projectile used in high strain rate, plane wave loading experiments.

5. Detail of front of projectile and impact side of target assembly.

6. Back side of target assembly showing the adapter plate holding an array of Ba TiO₃ crystal pins and a brass cup which contained the titanium alloy target disc and quartz gauge.

7. The stress profile of an impulsive wave.

8. Cross-section of target disc and quartz gauge showing an impulsive stress wave moving in the target with a velocity V and impinging on the quartz gauge at an angle θ.

9. Cross-section of flying plate and target at the time of first contact during a non-simultaneous impact having a tilt at impact equal to the angle α.

10. Response ω(t) of a quartz gauge to a planar impulsive stress wave input that impinges on the gauge at an angle.

11. (a) Ramp stress wave input to gauge. Ramp slope is equal to ε.

(b) Quartz gauge output due to ramp input of (a) impinging on gauge at an angle.
12. Ramp input \( r(t) \) and tilted impulse response \( \omega(t) \).

13. Oscilloscope trace of \( \text{Ba TiO}_3 \) crystal pin signals from which the velocity of impact and angle of tilt at impact were determined.

14. Specimen configuration for low strain rate compression tests.

15. Tensile specimen for quasi-static tensile testing.

16. Cross section of flying plate, target and quartz gauge showing on the left stress waves \( A \) and \( B \) shortly after impact of the flying plate onto the target, and on the right the elastic and plastic portion of the waves shortly after transmission of the elastic part of wave \( A \) into the quartz gauge.

17. Oscilloscope trace of quartz gauge output resulting from the stress history at the rear surface of the target. The duration of the trace is 1 usec.

18. Impedance matching construction in the stress-material velocity diagram used to convert the stress level recorded in quartz to the stress level in the titanium alloy target disc.

19. Stress-material velocity diagram for the impact of two identical materials for the case when the maximum stress is greater than the elastic limit of the materials.

20. Example of the representation of a stress history record by straight line segments.

21. Position-time diagram for the target and quartz gauge. The numbers designate the conditions of stress and material velocity experienced by the target and gauge caused by the stress waves that were generated at the time of impact. The sloping lines separating the numbered states are the locus of the positions of the stress waves in the target and gauge as a function of time. The times \( t_1, t_2, \) and \( t_3 \) are the arrival times of the stress waves at the target-quartz gauge interface.
22. Stress-material velocity diagram showing impedance matching to determine the numbered states of stress and material velocity for Fig. 21. In this diagram a three wave loading path was assumed.

23. Stress-strain behavior of commercially pure titanium. The $\sigma_x$, $\varepsilon_x$ curves are for plane wave, high strain rate loading in one dimensional strain. The curve having a dashed segment corresponds to a two wave loading path assumption. The curve having three solid segments corresponds to a three wave loading path assumption. The $\sigma_s$, $\varepsilon_s$ curves were derived from the $\sigma_x$, $\varepsilon_x$ curves by converting to one-dimensional stress conditions. The curve labeled 1-D stress is the loading behavior of the material in conventional quasi-static tensile stress.

24. Same as Fig. 23 for titanium-5 atomic percent aluminum.

25. Same as Fig. 23 for titanium-10 atomic percent aluminum, except that the two and three wave loading path assumption results are shown with dashed segments while the four wave loading path assumption result is shown with solid segments.

26. Average stress-displacement curves for the three materials using thin compression specimens with length to height ratio between 7.2 and 9.1. The heavy lines are the loading curves. A pair of light tangent lines is associated with each loading curve. The upper tangent line of each pair shows the elastic loading path for the entire system. The lower tangent line was constructed as described in the text to provide a horizontal displacement offset that corresponded to the plastic displacement of the center of the specimen.
27. Condition of the thin compression specimen in the transition region between totally elastic and totally plastic behavior.

28. Curves of hydrostatic stress at the center of the thin compression specimens versus plastic strain at the center of the specimen. The dashed straight lines at the left show the hydrostatic stress versus plastic strain experienced by the materials in one-dimensional strain plane wave loading.

29. Photomicrographs of the cross-section at the middle of the thin compression specimens showing in (a) through (f) the deformation structure developed at increasing levels of strain. The material is commercially pure titanium-50A. The plastic strains were (a) 0.4%, (b) 0.61%, (c) 1.04%, (d) 1.54%, (e) 3.0%, (f) 5.5%.

30. Same as Fig. 29 titanium-5 atomic percent aluminum. The plastic strains were (a) 0.2%, (b) 0.8%, (c) 1.62%, (d) 2.9%, (e) 3.8%, (f) 5.7%.

31. Same as Fig. 29 for titanium-10 atomic percent aluminum. The plastic strains were (a) 1.02%, (b) 1.9%, (c) 2.73%, (d) 3.9%, (e) 4.4%, (f) 5.8%.

32. Photomicrographs of the deformation structure of thin compression specimens having the same ratio of hydrostatic stress to yield stress and at the same level of residual plastic strain of 2%. The materials are (a) commercially pure titanium-50A, (b) titanium-5 atomic percent aluminum, (c) titanium-10 atomic percent aluminum.

33. Volume fraction of twinned material for the three alloys tested in constrained compression versus residual plastic strain each at a different values of the ratio of hydrostatic stress to yield stress.
34. Same as Fig. 33 with a constant value of the ratio of hydrostatic stress to yield stress.

35. Volume fraction of twinned material for the three alloys tested in simple tension versus plastic strain.

36. Photomicrographs of a grain containing a twin and kinking taken (a) with a normal objective with polarized light illumination and (b) with a Nomarski interference contrast objective. The kinked region in (a) that appears to be shadowing due to a rumpled surface is seen in (b) to be flat.

37. Photomicrographs of deformation structure in the commercially pure titanium resulting from compression to about 5.7\% residual plastic strain under high hydrostatic pressure constraint.

38. Same as 37 for titanium-5 atomic percent aluminum alloy.

39. Same as 37 for titanium-10 atomic percent aluminum alloy.

40. A combined plot of the volume of twinned material of the three alloys each deformed the following three ways: constrained compression, simple tension, and dynamic planar compression. The volume of twinned material is shown here as a function of applied shear stress. The arrows along the shear stress axis indicate the yield stress in simple tension for each material. The filled data points connected by vertical lines indicate the range of twin volume percent found in simple tension at the shear stress shown. The open data points connected by horizontal lines indicate the volume percent twin found in dynamic planar compression and the range of applied shear stress found from the two, three or four wave loading path assumptions.
41. Recovered flying plate and target discs as well as part of the target assembly on the right and part of the projectile front on the left.

42. Idealized loading and unloading paths for the normal stress \( \sigma_x \) and the lateral stress \( \sigma_y \) in one-dimensional strain showing a case of yielding during unloading.

43. Photomicrographs of deformation structures in commercially pure titanium produced by high strain rate loading and unloading in one dimensional strain.

44. Same as Fig. 42 for titanium five atomic percent aluminum alloy.

45. Same as Fig. 42 for titanium ten atomic percent aluminum alloy.

46. Photomicrographs of deformation produced at the edge of targets of (a) commercially pure titanium, (b) titanium-five atomic percent aluminum and (c) titanium-ten atomic percent aluminum.

47. Strain rate sensitivity of titanium-aluminum alloys represented by solid line as the ratio of the dynamic yield stress to the quasi-static yield stress, as a function of the quasi-static yield stress. Dashed line is the ratio of low temperature quasi-static yield stress to room temperature quasi-static yield stress.
Fig. 1.
Fig. 7.

Fig. 8.
Fig. 9.

Fig. 10.
Fig. 11.
Fig. 12.
Fig. 14.
TENSILE SPECIMEN

SCALE: 2:1

Fig. 15.
Fig. 16.
Fig. 18.

Fig. 19.
Fig. 20.
Fig. 21.
Fig. 22.
Fig. 23.
Fig. 24.
Fig. 25.
Extrapolation of Initial Elastic System Loading Line

Plastic Displacement of Center of Specimen

Fig. 26.
ELASTIC BEHAVIOR DUE TO HYDROSTATIC CONSTRAINT

Fig. 27.
Fig. 28.
Fig. 30.
Fig. 30.
Fig. 31.
Fig. 31.
Fig. 32.
Fig. 33.
Fig. 34.

Fig. 35.
Fig. 36.
Fig. 37.
Fig. 38.
Fig. 39.
Fig. 40
Fig. 42.
Fig. 45.
Fig. 47
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