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THE EFFECT OF STRAIN AGING ON THE DYNAMIC ELASTIC-PLASTIC

PROPERTIES OF IRON*

By de

John W. Taylor
and
Eugene G. Zukas

MASTER

Los Alamos Scientific Laboratory
of the
University of California
Los Alamos, New Mexico

Abstract

Strain-aging experiments with cold worked iron specimens show that twinning can occur during plane wave shock loading after the specimens have been permitted to recover over an extended time interval. There is a thermal activation energy of 20 kilocalories per mole associated with this recovery. The absolute times involved are, however, much too long to be explained on the basis of the Cottrell-Bilby theory of strain aging. Twinning was found to be unimportant to the dynamic yielding process. After straining, the recovery of the elastic-plastic wave structure, characteristic of fully annealed iron, proceeds with an activation energy greater than 130 kilocalories per mole.

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THE EFFECT OF STRAIN AGING ON THE DYNAMIC ELASTIC-PLASTIC PROPERTIES OF IRON

INTRODUCTION

The presence of the upper-lower yield strength phenomenon in α -iron during normal tensile testing has been known for many years. The influence of straining and straining followed by aging for various times at different temperatures has also been studied in considerable detail.

On dynamic or shock loading, a measure of the movement of the surface opposite to that impacted versus time (free-surface velocity) for annealed iron shows a similar behavior to the upper-lower yield. Such a drop or "dynamic overshoot" indicates yielding which could be caused by the volume change resulting from very extensive twinning or perhaps by a mechanism similar to that responsible for the upper-lower yield phenomenon in annealed iron.

This study was concerned with the kinetics of twinning after strain aging under dynamic loading conditions, the effect of twinning on the elastic-plastic wave structure, and the kinetics for the recovery of the dynamic overshoot after cold work.

Recent improvements in high speed photographic and electronic techniques have made possible fairly direct observations of the free surface motion of impulsively loaded metals. In particular, it has been directly observed that many metals, when shock loaded to pressures ranging from one to one hundred kilobars, exhibit a two wave structure called "elastic" and "plastic" waves. The former is propagated through the metal at roughly the speed of sound, whereas the latter appears to propagate at a speed of approximately $\sqrt{B/\rho}$ where B is the adiabatic bulk compressibility modulus of the material and ρ is the mass density. The stress amplitude of the "elastic" wave has been

called the "Hugoniot elastic limit" and it will be so designated in this paper.

The general theory of shock propagation in solids has been reviewed by Rice, Walsh and McQueen(1) and more recently by Duvall(2). More detailed treatments of shock propagation in an elastic-plastic medium have been given by Walsh(3) and Moreland(4). A dislocation model for a shock front was proposed by C. S. Smith(5). This model has recently been enlarged for the case of iron by Hornbogen(6) to include an explanation for the existence of large numbers of screw dislocations. The requirements of such a model provide the basis for calling the second shock wave "plastic"; for if any compressive disturbance is to travel at a velocity less than longitudinal sound speed, there must be some plastic shear strains in the structure. For the benefit of the reader who may not be familiar with shock wave techniques and theory, we shall sketch here the pertinent concepts with reference to Figure 1. If a one dimensional shock wave is traveling from left to right in a medium at initial pressure P_0 and specific volume V_0 , and if the shock wave is in a true steady state so that the shock front thickness is constant and the pressure and specific volume achieve constant values P_1 , V_1 behind the shock front, we may translate to a coordinate frame in which the shock front is at rest and assume that mass enters a unit area at a rate U_s/V_0 per unit time where U_s is the shock velocity. The mass leaves the shock front (moving to the left) at a velocity $U_s - U_p$, where U_p is the shock particle velocity. Conservation of mass requires that

$$\rho_0 U_s = \rho_1 (U_s - U_p) \quad (1)$$

where $\rho_0 = 1/V_0$ and $\rho_1 = 1/V_1$. Conservation of momentum requires that

$$P_1 - P_0 = \rho_0 U_s U_p. \quad (2)$$

If the material behaved as an ideal fluid, the shock would be relieved at a free surface by a rarefaction wave which could be mathematically described as a centered simple wave(7). It has been shown for shocks in the pressure range of interest in the present investigation that the final free surface velocity will be very nearly twice the shock particle velocity(2). It follows therefore that the shock pressure and the density behind the shock front are well determined if the shock velocity and free surface velocity are measured. If there are two shock waves, one behind the other, the analysis is a simple generalization of the foregoing and one finds that

$$P_1 - P_0 = \rho_0 U_{s1} U_{p1} \quad (3)$$

$$\rho_0 U_{s1} = \rho_1 (U_{s1} - U_p) \quad (4)$$

$$P_2 - P_1 = \rho_1 (U_{s2} - U_{p1}) (U_{p2} - U_{p1}) \quad (5)$$

$$\rho_1 (U_{s2} - U_{p1}) = \rho_2 (U_{s2} - U_{p1} - U_{p2}) \quad (6)$$

where U_{s2} and U_{p2} are measured relative to the state P_0 , ρ_0 ahead of the shock. The derivation of the free surface velocity is similar except that the final free surface velocity due to the second shock should be achieved in a series of steps of increasing frequency and of magnitude equal to the increment due to the weaker first shock. Therefore graphic plots of free surface velocity will closely approximate the pressure distribution inside the metal.

In three recent papers, it has been made apparent that the Hugoniot elastic wave in ingot iron and low carbon steels exhibits a very peculiar feature(8,9,10). As the wave passes a given point in the material, the

stress passes through a distinct maximum and is then reduced by 10% to 15% before the arrival of a plastic wave. In two of the papers, this reduction in pressure was found by time resolved measurements of the free surface velocity of shock loaded plates. The third method of measurement which employs a quartz transducer is also essentially a measurement of the free surface velocity because the acoustic impedance of iron is very much higher than that of quartz. The quartz technique is of particular importance, however, because when it is employed in a study of the elastic-plastic behavior of aluminum alloys, the results are quite credible. They agree well with the results of free surface velocity measurements and are almost a direct measure of the stress level in the aluminum because of the nearly perfect acoustic impedance match. These experimental results make one quite confident that the free-surface velocity versus time plots are essentially maps of pressure versus distance in the wave. It should be noted that in all of the above mentioned experiments and in all our experiments, great care was taken to insure that the target samples were of sufficient lateral extent that all waves can be considered one dimensional.

The mechanism by which the metal forms a two wave structure with the dynamic overshoot in the Hugoniot elastic wave is clearly a matter of considerable interest. It has been suggested that perhaps the effect is accomplished by the Cottrell-Bilby mechanism of tearing dislocations free from impurity atmospheres(9). A second suggestion has been that the magnitude of the dynamic yield point and the details of the overshoot are due to the rather profuse twinning which has long been known to accompany impact loading of iron and low carbon steels(10). This suggestion is probably based on the fact that the static yield point increases steadily as the

temperature is reduced until the twinning phenomenon appears at about 140°K.

In 1926 Pfiel found that under impact loading, low carbon steels deform at least partly by twinning(11). He further noted that when a sample is prestrained at room temperature, the twinning phenomenon does not occur under impact loading until after the material has rested for many weeks at room temperature. The elimination of twinning under conditions of slow deformation at low temperature by room temperature cold work was reported by Churchman and Cottrell(12), and the return of the phenomenon after room temperature aging was interpreted by them as a manifestation of the effect of diffusion of carbon and nitrogen to dislocation cores. (Cottrell-Bilby theory of strain aging of iron and low carbon steels)(13).

In the late 1940's while investigating the effects of projectile impacts on mild steel, G. I. Taylor and Carrington and Gayler(14) found that twinning occurred in iron grains which did not appear to have suffered other plastic deformation. They suggested that twinning may occur between the elastic and plastic waves. More recently, deformation twinning and strain aging effects were investigated by Zukas and Fowler(15). They noted that when samples were tested at liquid nitrogen temperature, the recovery of the ability to twin occurred in two or three weeks after 25% deformation at room temperature, and that the percentage of grains twinned appeared to be a significant parameter. Correlations of twinning with the Hugoniot elastic-plastic wave structure in iron were made by Taylor, Rice and Zukas(10) who found that samples of annealed iron shocked to pressures to just below and to just above the Hugoniot elastic limit did not twin and did twin respectively; and by Zukas, Fowler, O'Rourke and Minshall in an investigation of Fe-3% Si single crystals(16). The

latter authors noted that if two orientations of Fe-Si crystals are simultaneously shocked to a pressure such that the plastic shock velocity exceeds the longitudinal sound speed through one crystal and is less than the sound speed through the other, the former crystal does not twin and the latter does.

EXPERIMENTAL

The iron used for samples was made in the following way: commercial ingot iron was vacuum melted with 3% to 4% of 1095 steel, the exact addition depending on the original composition of each material. In this way, 20 kilogram castings were obtained in which the iron was comparatively free of oxide inclusions and was not excessively contaminated by other elements. A typical casting contained about 120 ppm carbon, 180 ppm oxygen, both free and combined, 300 ppm manganese, and minor amounts of other constituents.

Samples were cut from the castings and annealed. Right cylinders 0.800 inch long by 0.875 inch in diameter were used. Concentric grooves 0.015 inch deep and spaced 0.050 inch apart were machined on the ends to prevent barreling during pressing. After lubricating the grooved ends with a molybdenum disulfide paste, the samples were pressed to a 25% reduction in height. Samples 0.890 inch in diameter by 0.500 inch thick were each mounted in a 5 inch diameter by 0.500 inch thick low carbon steel plate. These assemblies were surface ground prior to shock loading to insure an absolute minimum error due to averaging effects caused by surface irregularities in the capacitor gap.

Two procedures were used for the elevated temperature aging treatments. For temperatures up to 100°C, a silicone oil bath of adequate size was used.

The pressed samples were dropped into the oil bath which raised the sample to the holding temperature quickly. For temperatures above 100°C, the samples, in copper boats, were placed inside a vacuum tube furnace.

In order to reduce the cost of the type of experiments reported in Reference 10 and to facilitate recovery of the shocked samples, the shock loading experiments were carried out with two-inch diameter projectiles which were gas fired from a small gun. At this small diameter, the elimination of two dimensional edge effects from the region of interest required that the target samples be not more than one-half inch thick. Initially, assemblies were almost completely prepared prior to cold working and aging and were then kept at dry ice temperature until within one hour of shock loading. After it was realized that several weeks at room temperature are necessary to produce any effects observable by our techniques, specimens were stored at room temperature.

The presence of the low carbon steel mounting plates, which served primarily for lateral confinement of the slugs, did not perturb the measurements significantly. This is because the elastic-plastic properties of this steel are virtually identical with the properties of ingot iron. This was checked by comparing the free surface velocity profile from a fully annealed sample imbedded in such a plate with that obtained from a large diameter iron slug of the same thickness.

The experimental geometry for shock loading and recovery is shown in Figure 2. Time resolved free surface velocities were measured electronically by a dc capacitor technique. The target assemblies were mounted directly on the muzzle of the gun and were driven into the recovery chamber by the projectile. The projectiles were capped by 0.250 inch thick low carbon steel

driver plates which guaranteed constant particle velocity at the collision interface during the entire run of the compression wave through the target. The target slugs invariably impacted the brass or steel measuring capacitors after being loaded by plane shocks and the pressure released. The geometry was such that only the outer diameters could have been seriously affected by this collision but no significant effects from this source were noted. The recovery chamber was a two foot cubical steel box with walls two inches thick. The box was lined with pine two-by-sixes and the target assembly was actually driven into low density (5 lb/ft^3) polyurethane foam plastic inside the chamber. No gross effects of impacts within the recovery chamber were noted. In cases where only metallographic examination was desired, the target slug was driven directly into the polyurethane foam and was thus decelerated gently.

All shock loading experiments were performed at ambient outside temperature, which ranged from 0°C to 20°C . There is other evidence, however, that the elastic-plastic wave structure in iron is virtually independent of temperature below 20°C (17).

RESULTS

Our experiments fall rather naturally into two groups: those concerned with twinning and low temperature annealing, and those primarily concerned with the elastic-plastic transition and high temperature annealing.

For the first group of experiments, a control sample was strained and then fully annealed to a hardness of DPH 72. This sample was shock loaded to a pressure of approximately 40 kilobars and recovered. The free surface velocity profile agreed well with those reported by Taylor and Rice for equivalent shots in which Armco iron was used(10). Subsequent metallurgical

examination indicated that about 90% of the grains twinned. Samples from the same casting were then pressed 25% and held at 293°K, 335° and 373°K for various times. The percentage of grains twinned as functions of time and temperature are shown in Figure 3.

In samples that were held for eight hours or more at 373°K, 90% of the grains twinned when shock loaded. No further variation in metallographic appearance was noted in additional shock loaded specimens that had been held at 373°K for as long as 22 days. The shock loading of specimens held for the maximum time intervals at temperatures to 100°C failed to produce any wave structure comparable to that observed when fully annealed iron is shock loaded. If the point of maximum negative curvature of the time-resolved free surface velocity profile is defined as the "elastic" wave amplitude (Figure 4), then it is possible to detect a trend. A graphic plot of pressure versus aging time for several aging temperatures is shown in Figure 5. Note that the observed effect appears to be one of age softening rather than the age hardening which is observed when low carbon steels treated in this manner are tested statically. Also, the annealing times and temperatures used here are fully adequate, according to many published observations, to restore the static upper-lower yield point phenomenon in iron of comparable purity(18).

Earlier in this report, it was mentioned that perhaps the dynamic overshoot might be a result of the volume displacement brought about by extensive twinning. Referring again to Figure 4, curve A is for a sample which did not twin under shock loading, whereas about 90% of the grains were twinned in the sample represented by curve C. From these data, it can

be concluded that the twinning process is not responsible for the dynamic overshoot.

If we now hypothesize that diffusion is the metallurgical process by which the ability of iron to twin under shock loading is restored within the time scale observed here, we can postulate further and obtain the activation energy for the diffusion process in question. We assume that concentration gradients will not be the rate controlling factors because everything is initially supposed to be distributed randomly and uniformly throughout the structure. The diffusion coefficient will then enter the picture only to the extent that it determines the mobility of the causative agent in the pertinent potential gradient. Therefore, the process must go to completion as some function of the variable (Dt/kT) where T is the absolute temperature, k is Boltzmann's constant, t is time, and D is the diffusion coefficient. It is further assumed that

$$D = D_0 \exp(-E/RT) \quad (7)$$

where R is the gas constant and E is the activation energy. If one chooses a horizontal line (constant percent grains twinned) in Figure 3 and determines times t_p at each temperature T , he should be able to plot $\ln(t_p/T)$ against $1/T$ and determine the desired activation energy from the slope of the curve. The result of such an approach is shown in Figure 6. The activation energy so obtained is 20,000 cal./mole. This is known to be the activation energy for the diffusion of carbon in alpha iron(19). Thus the situation appears to be reasonably straightforward until one notes that the absolute times involved are much too long for samples having equivalent cold work when compared with the data of Harper(20) for extinction of the internal friction Snoek peak, but they are too short when compared with the data of Kamber,

Keefer and Wert(21) for the growth of the cold work, or Koster, peak(22). They are also much longer than the times reported by Zukas and Fowler for the recovery of the ability to twin during low temperature deformation(15). These comparisons make it difficult to conclude that carbon is the responsible agent. However, another experiment was suggested by the recently reported results of Barrand and Leak(23). These authors noted that if a cold worked sample is aged at temperatures of the order of 100°C for many hours and is then restrained, the Snoek peak does not reappear at its original height. They interpreted their data to mean that after carbon has diffused to dislocation cores in iron, it forms, at a slower rate, carbide precipitates. These precipitates will not subsequently decompose at moderate annealing temperatures, even though the favorable environment which allowed their formation (the dilatational strain field of edge dislocation cores) has been pulled away by further stress.

To test the possible relationship with the Barrand and Leak theory, the following experiments were conducted. Two standard samples were pressed to 12-1/2% reduction in height. Sample number one was then held for 24 hours at 100°C while sample number two was kept at 20°C for the same time. Both samples were again pressed 12-1/2% and held for 24 hours at 100°C. After shock loading the samples to 40 kilobars, it was found that the sample which had received the intermediate 100°C anneal had not twinned, whereas the second sample twinned as if there had been no long pause between the two cold work operations. The authors regard this experiment as strong evidence in support of the hypothesis that the motion of carbon is in some way responsible for the recovery of twinning. In both of these experiments, time resolved

free surface velocity data were taken. The elastic wave amplitudes were virtually identical with each other, and with those for samples which were pressed 25% and immediately aged for 24 hours at 100°C. This indicates that the gradual decrease in amplitude of the Hugoniot elastic wave as low temperature strain aging progresses is not caused by the twinning contribution to the plastic deformation, even though it may be accompanied by it. It would be interesting to know the results of a similar experiment carried out at liquid nitrogen temperature where it has long been presumed that twinning controls the static yield strength.

When it became apparent that the distribution of impurity atoms and the existence or nonexistence of twinning are probably not at all pertinent to the peculiar structure of the Hugoniot elastic-plastic wave system in iron, it was decided to investigate certain other possibilities. Since all of the detection methods used to observe the dynamic overshoot are intrinsically averages over many grains for the normal iron samples, perhaps we were observing a grain boundary effect. Tests on single crystal samples would certainly resolve this point, but it is very difficult if not impossible to obtain single crystals of iron of sufficient size for shock loading studies where edge effects are an important consideration. However, single crystals of iron-silicon alloys of an adequate size are readily available and these were used. Two single crystals of Fe-3-1/2% Si were mounted in the [100] and [111] orientations and shock loaded. Both crystals exhibited pronounced overshoots in the elastic wave, and both samples twinned (Figure 7). We take this to be conclusive evidence that the observed overshoot in normal iron is in no way caused by viscous friction at grain boundaries.

It was next attempted to determine the extent of high temperature annealing necessary to recover the dynamic overshoot effect. If it is assumed that this recovery is controlled by the diffusion of iron in iron, the process presumed to effect the removal of existing dislocations, then the activation energy for this process should be 60 kcal/mole(24). Cold worked samples were annealed for various times at temperatures to 700°C before shock loading. An anneal of one hour at 700°C did not result in the reappearance of the overshoot whereas 5 hours at 700°C did. Therefore, for a process controlled by the diffusion of iron in iron, the overshoot should reappear on samples held for about 3 days at 650°C. Samples were held at 650°C for times from 1 hour to over 6 days and the dynamic overshoot still had not reappeared. From these tests, it can be concluded that the activation energy for this recovery process is greater than 130 kcal/mole. Thus, the recovery of the overshoot of the Hugoniot elastic pressure wave after cold work does not appear to be governed by the diffusion of iron in iron.

DISCUSSION

These results indicate that the twinning associated with impact loading of iron and low carbon steels is incidental to the Hugoniot elastic-plastic transition. Strain aging experiments show that the recovery of the twinning phenomenon proceeds with what appears to be the activation energy for the diffusion of carbon in alpha iron, but the required times are so long that it is necessary to invoke the further assumption that the process is somehow controlled by the formation of carbide precipitates, probably at dislocation cores. The discrepancy between the absolute times reported here

and the times reported by Zukas and Fowler(15) for the recovery of twinning under conditions of low temperature deformation can be resolved by the following qualitative model. Twinning is a mode of plastic deformation which competes with the motion of ordinary dislocations. The dynamic generation of regular dislocations is a thermally assisted process which can occur at somewhat lower stress levels than twin formation. At low temperatures, the stress necessary to generate regular dislocations becomes comparable to that necessary to form twins, and at room temperature under impact loading conditions the required strain rates are so high that some twins will form before very many dislocations can be created and set in motion. The effect of slow deformation at room temperature is to fill a sample with dislocations which are mobile to the extent that they only partially interfere with each other. The mutual interference of dislocations is adequate to produce work hardening and increase the Hugoniot elastic limit, but it is not adequate to keep these dislocations from introducing a comparatively short range randomness to the structure. Since twinning, at least in observably large regions, involves long range atomic cooperation, this randomness of the structure prohibits the formation of observable twins either under shock loading or low temperature deformation. Low temperature annealing results in the diffusion of carbon and nitrogen atoms at least to the edge components of the dislocations, and the degree to which the dislocations are bound by these impurity atmospheres is a function of the temperature. This would appear to explain why twinning appears first at low temperature. Later stages of strain-aging result in the formation of carbide precipitates, perhaps in the fashion suggested by Bullough and Newman(25), and in stable configurations which can pin existing dislocations.

enough to permit twinning to occur. Bullough and Newman point out that the formation of carbide precipitates may be a thermally activated process which proceeds much slower than diffusion but which nevertheless has an activation energy which may coincide with that for diffusion.

The dynamic yielding mechanism remains somewhat obscure. It was found that in fully work hardened specimens the amplitude of the Hugoniot elastic wave decreased gradually over an interval of several hundred hours when samples were held at 100°C. However, the dynamic overshoot did not reappear nor did it reappear after several hundred days at 20°C. After a one hour anneal at temperatures above 400°C, the amplitude of the Hugoniot elastic wave was about 10 kilobars and did not change appreciably for longer annealing times. Recovery of the wave structure, characteristic of fully annealed iron, appears to occur with a thermal activation energy greater than 130 kcal/mole. This suggests that it is a more complicated process than the simple diffusion of iron in iron which was found by Kamber, Keefer and Wert(20) to control the decay of the cold work internal friction peak.

The authors suggest that the overshoot structure in the Hugoniot elastic wave is not connected with dislocation pinning as is apparently the case with the static upper-lower yield point effect. Instead it is suggested that the mechanism is due to the stress dependence of the dislocation velocity and generation rate, similar to the mechanism suggested by Gilman and Johnston for yielding in LiF and extended to iron by Bohn(26,27). It is apparent from all published data that some yielding begins immediately as the wave front passes through the material. If at some stress level which appears to be of the order of 10 kilobars, the dislocation multiplication

rate becomes catastrophically large, a sudden fall in stress would be natural. In a cold worked sample, the local stresses caused by dislocation tangles would probably prevent the rate of generation of new dislocations from reaching a level sufficient to do this, and the overshoot could not develop. The amplitude of the Hugoniot elastic wave in work hardened material is determined by the stress necessary to move dislocations through dislocation forests. This stress increases very rapidly for small amounts of deformation but levels off before the deformation reaches 25%. Thus, the amplitude of the Hugoniot elastic wave is approximately doubled for samples pressed 25% over that for annealed iron, but further cold work, even to 90% reduction in thickness, does not increase it appreciably.

CONCLUSIONS

1. The strain aging kinetics for the recovery of twinning has an activation energy of 20 kcal/mole. However, the absolute rate is too low to be interpreted on the basis of carbon diffusion to dislocation cores.
2. The dynamic overshoot and the amplitude of the Hugoniot elastic wave pressure are not a result of extensive twinning.
3. Twinning can be permanently inhibited by a double strain aging treatment.
4. The activation energy for the dynamic overshoot recovery after cold work is greater than 130 kcal/mole. This is much larger than that for iron-iron diffusion (60 kcal/mole).
5. Experiments with 3-1/2% silicon-iron single crystals show that the dynamic yielding process is not materially affected by grain boundaries.

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FIGURES

Fig. 1. A. Schematic representation of an ideal two wave shock system in an elastic-plastic medium. The material, initially at density ρ_0 and pressure P_0 , is subjected to first shock (1) which carries pressure P_1 , density ρ_1 and particle velocity U_{p1} . This shock propagates at velocity U_{s1} relative to state (0). The second shock travels at velocity U_{s2} relative to state (0) and carries pressure P_2 , density ρ_2 and particle velocity U_{p2} . B. Single shock in a coordinate system in which the shock front is at rest.

Fig. 2. Sketch of the experimental geometry used in shock loading and recovery experiments. A--recovery chamber; B--capacitor assembly; C--target plate; D--specimen slug; E--gun barrel; F--steel driver; G--projectile.

Fig. 3. Percentage of grains twinned after shock loading to approximately 40 Kilobars as a function of annealing time and temperature. The error flags represent estimated spread of data.

Fig. 4. Selected free surface velocity profiles from fully annealed and strain aged iron samples. The zero of time is arbitrary.

A. Freshly work hardened sample	D. Treatment A + 8 hours at 650°C
B. Treatment A + 1/2 hour at 100°C	E. Treatment A + 5 hours at 700°C
C. Treatment A + 48 hours at 100°C	F. Furnace cooled from gamma

Fig. 5. Approximate amplitude of the Hugoniot elastic wave in strain aged iron as a function of aging time and temperature. (one kbar = 10^9 dyne/cm²).

Fig. 6. A semilogarithmic plot of t_f/T vs $1/T$ where T is absolute temperature and t_f is the time necessary for a fraction, f , of grains to twin under shock loading. The straight line is drawn as a best fit to the 50% data.

Fig. 7. Free surface velocity traces from shock loading single crystals of Fe- 3-1/2% Si in the [111] and [100] orientations. The zero of time is arbitrary.

Fig. 1. A. Schematic representation of an ideal two wave shock system in an elastic-plastic medium. The material, initially at density ρ_0 and pressure P_0 , is subjected to first shock (1) which carries pressure P_1 , density ρ_1 and particle velocity U_{p1} . This shock propagates at velocity U_{s1} relative to state (0). The second shock travels at velocity U_{s2} relative to state (0) and carries pressure P_2 , density ρ_2 and particle velocity U_{p2} .

B. Single shock in a coordinate system in which the shock front is at rest.

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Fig. 2. Sketch of the experimental geometry used in shock loading and recovery experiments showing: A--recovery chamber; B--capacitor assembly; C--target plate; D--specimen slug; E--gun barrel; F--steel driver; G--projectile.

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Fig. 3. Percentage of grains twinned after shock loading to approximately 40 kbar as a function of annealing time and temperature. The error flags represent spread of data.

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Fig. 4. Selected free surface velocity profiles from fully annealed and strain aged iron samples. The zero of time is arbitrary.

A. Freshly work hardened	D. A + 8 hours at 650°C
B. A + 1/2 hour at 100°C	E. A + 5 hours at 700°C
C. A + 48 hours at 100°C	F. Furnace cooled from gamma

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Fig. 5. Approximate amplitude of the Hugoniot elastic wave in strain aged iron as a function of aging time and temperature (one kbar = 10^9 dyne/cm²).

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Fig. 6. A semilogarithmic plot of t_f/T vs $1/T$ where T is absolute temperature and t_f is the time necessary for a fraction f of grains to twin under shock loading. The straight line is drawn as a best fit to the 50% data.

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Fig. 7. Free surface velocity traces for single crystals of Fe - 3-1/2% Si shock loaded in the [111] and [100] orientations.

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