

THE DEFORMATION OF METALS AND ALLOYS
SUBJECTED TO EXPLOSIVE LOADING

by

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ABSTRACT

The effects of explosive ($>10^6/s, >15\text{GPa}$) and high ($10^3/s, <1\text{GPa}$) strain rates and pressures on the microstructure of pure aluminium, duralumin, tough pitch copper, pure magnetic iron and 18 chromium-8 nickel stainless steel were studied using transmission electron microscopy.

High strain rate tests were carried out using a split Hopkinson bar to apply a load pulse of one millisecond duration. An explosive assembly was designed to apply a uniaxial shock pulse whose duration varied from zero at the free surface to one microsecond at the shocked surface of a thin specimen.

The heavily deformed ($>50\%$) Hopkinson bar specimens showed typical deformation structures. There was a general reduction in dislocation cell size and deformation twins occurring in copper and 18-8 steel were often bent. Aluminium and copper underwent some recrystallization and considerable α' formation took place in 18-8 steel.

At low explosive pressures (18 - 26 GPa) deformation was minimal except in the case of aluminium which also recrystallized. The microstructures were heavily dislocated and deformation twinning occurred in copper, iron and 18-8 steel. 18-8 steel contained many body

centred cubic α' crystals and some hexagonal close-packed ϵ plates.

The hardness of copper and duralumin, and the size of the α' crystals in 18-8 steel, increased with the pulse length and this is attributed to the increasing rarefaction wave length. Two hardness steps were observed in shocked iron. These result from interactions of the double shock wave and the free surface rarefaction waves caused by the high pressure $\alpha \rightarrow \epsilon$ phase transformation.

At pressures above 30 GPa heavy deformation occurred during deceleration from the free surface velocity and recrystallization was observed in copper and iron. 18-8 steel was relatively undamaged and showed an increase in the size and number of α' crystals at 40 GPa. However, at 55 GPa relatively few α' crystals were formed and these were generally small. In both cases, the as-received α grains contained numerous short, narrow twins.

The rapid rise (10^{-9} s) and fall (10^{-7} s) times of the shock pulse permit a comparison between the postulated homogeneous and pole twinning mechanisms using the measured twin dimensions. It is suggested that, with the possible exception of α twinning in 18-8 steel, twinning occurs by homogeneous nucleation.

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SYMBOLS AND ABBREVIATIONS

A	cross-sectional area
c	sound velocity, light velocity
$\bar{c}$	specific heat
C_b	Hopkinson bar sound velocity
C_E	leading explosive rarefaction wave velocity
C_L	leading rarefaction wave velocity
C_O	zero pressure rarefaction wave velocity
C_S	sound velocity in specimen
d	specimen thickness, fringe diameter
D	detonation wave velocity
D_H	detonation wave velocity in high speed explosive
D_L	detonation wave velocity in low speed explosive
e	elastic wave, Fabry - Perot mirror separation
E	specific internal energy
f	focal length
F	finesse
G	[shear modulus]
h	twin width, beam radius
k	wave number
L	path difference, laser cavity length, twin length
L_S	initial Hopkinson bar specimen thickness
P	pressure
r	ratio of the distance between the interaction zone and the free surface and the specimen thickness
R	charge radius, reflectivity, rarefaction wave
s	explosive cylinder thickness
s_I	interface I displacement
s_{II}	interface II displacement
S	entropy
t_s	shock pulse length
t_t	twin formation time

T	temperature
T_a	isentropic release temperature
T_H	shock temperature
T_m	melting temperature
u	particle velocity
u_{fs}	free surface velocity
u_p	particle velocity after shock wave
u_r	particle velocity after rarefaction wave
U_s, U	shock wave velocity
v	dislocation velocity
v_o	impacting Hopkinson bar velocity
V	specific volume, surface velocity, voltage
V_c	cosine photomultiplier output
V_s	sine photomultiplier output
W	annular width of spall plate
W_s	maximum specimen width
W_t	explosive cylinder diameter, plate diameter
x	distance between detonation front and back rarefaction wave
X	reaction zone length
y, y'	reduction in radius of back rarefaction wave
α	plane wave generator cone angle
γ	shear strain, adiabatic constant
δ, d, Δ	differential operators
ϵ	strain
$\dot{\epsilon}$	strain rate
ϵ_s	specimen strain
ϵ_s^p	specimen plastic strain
λ	wavelength
σ	stress
σ_{bI}	impacting bar stress
σ_{bII}	specimen bar stress
σ_s	specimen stress

ρ	density
ρ_b	Hopkinson bar density
ρ_s	specimen density
ϕ	phase angle
ω	beam divergence
HMX	cyclotetramethylene tetranitramine
TNT	trinitrotoluene
RDX	cyclotrimethylene trinitramine
FSR	free spectral range
S.F.E.	stacking fault energy
CJ	Chapman - Jouguet

CHAPTER IINTRODUCTION

In aircraft crash investigation it is obviously necessary to determine whether the crash was due to accidental causes (mechanical or electrical failures, pilot error, etc.) or an exploding bomb. Often this can be determined by investigating the location and condition of the debris. Small metal fragments will punch holes in soft material under explosive pressures. Fragment surfaces may have distinctive features such as pitting, metal deposition, rippling and erosion. Metal structures near to an explosion will have distinctive fractures.

As aircraft often crash in remote places or in locations where recovery is difficult, considerable corrosion and weathering may take place before the debris is recovered. Thus the surface effects of an explosion may well have been lost. However, the internal microstructure is often undamaged and could yield evidence of the type of loading undergone.

The purpose of this work is to examine the microstructure of five metals commonly found in aircraft. Pure aluminium and duralumin (fuselage), tough pitch copper (wiring), pure magnetic iron (motor cores) and 18 chromium - 8 nickel stainless steel (luggage, fittings) were chosen. These were examined in the as-received condition, after loading

in a Hopkinson bar apparatus at a strain rate of 10^3 s^{-1} (similar loading rates occur when an aircraft crashes), and after explosive loading at a strain rate of 10^6 s^{-1} (bomb on board).

An explosive loading assembly was designed to ensure as far as possible that microstructures caused by explosive loading were not subsequently altered by gross deformation during the deceleration from the high specimen velocity caused by the explosion. These microstructures were then compared with those occurring in heavily deformed low strain rate (10^3 s^{-1}) specimens. Finally, the effect of the explosive pressure and pulse shape on the microstructures was investigated.

The use of explosives to determine the high pressure characteristics of solids has only been developed since the last war. Before 1940, high pressure experiments were mostly static, and were limited by the weakness of the available construction material. It was widely felt that explosions resulted in the catastrophic destruction of solids and that the debris was not worth inspecting. Subsequently it was realized that explosive loading is an orderly physical process which obeys Newton's laws and those of thermodynamics, chemistry and quantum mechanics. The compression produced by this dynamic loading is much greater than that caused by static loading, and this led to the revision of some

commonly held assumptions concerning the bulk properties of matter.

The advantages of explosives are that they can produce high pressure pulses which are independent of the material strength, and that there are virtually no limitations on the temperatures and pressures which could be achieved given a suitable explosive. The pressures produced can be two orders of magnitude greater than the maximum static pressure possible, and are similar to pressures existing due to gravitational forces at the centre of planets, and hence of interest to the geophysicist. This has led to the production of high pressure equation of state data and the determination of phase changes and, at very high pressures, data relating to the interatomic forces in a material. Of more practical interest is the development of high energy forming, welding and projectile impact theory.

Shock wave theory for a compressible material was first suggested by Stokes [1] who derived the conservation equations usually attributed to Rankine [2] and Hugoniot [3]. A typical shock pulse consists of a very steep shock front (rise time ca. 10^{-9} s) during which the pressure rises to the shock pressure and the density, particle velocity and internal energy of the material undergo corresponding jumps. A region of high pressure in thermodynamic equilibrium remains for a short time

(microseconds or less) after the passage of the shock front until a rarefaction wave causes the pressure to decay relatively slowly to the original value.

It might be expected that shocked materials would have vastly different terminal structures and properties from normally deformed materials. It is true that high strain rate features, such as considerable vacancy creation due to non-conservative high-speed dislocation motion and the creation of twins as well as residual structures caused by high pressure phase transformations, do occur. However, deformation mechanisms not existing at lower strain rates and pressures have not been encountered.

Generally gross deformation caused by shock is very small and shock hardening is considerably greater than hardening produced by conventional deformation to the same permanent strain.

These features can be lost to a greater or lesser extent due to the effect of the high transient shock temperature and the subsequent adiabatic release temperature as well as any gross deformation occurring during the deceleration of the material from its shock velocity [4, 5, 6, 7].

CHAPTER IIREVIEW OF THE LITERATURE

The transient nature of shock compression requires that any structural changes produced must occur extremely rapidly (within microseconds). If these changes are to be detected they must survive both the subsequent rarefaction back to atmospheric pressure (some will be caused by this rarefaction) and the residual temperature due to the adiabatic expansion from the shock temperature and pressure. In addition, these effects may be complicated by subsequent gross deformation during recovery.

Some early investigators tended to ignore the shape of the shock pulse, quoting only the peak pressure and the explosive assembly. This produced a wide variety of structural effects under ostensibly identical conditions. It was subsequently suggested that it should be possible to attribute residual structures to individual characteristics of the shock pulse, viz. shock front, pulse length, rarefaction rate, shock pressure and temperature, and the residual temperature after adiabatic expansion to atmospheric pressure.

This review comprises a survey of the microstructural effects of shock loading on the metals and alloys which are the subject of this work, as well as some proposed models for the shock compression of metals.

2.1 The Metallurgical Effects of Shock Waves

The passage of the shock and rarefaction flanks results in the generation and rapid motion of large numbers of dislocations for a short period of time. This causes the generation of large numbers of vacancies which anneal out to some extent due to the shock temperatures. The residual dislocation microstructure resembles that produced by conventional cold rolling to high strains, in that dislocation arrays or cells are formed. As the shock pressure increases the substructure undergoes refinement. Short pulse lengths tend to result in less well defined cells.

Deformation twinning is frequently observed in shock loaded metals and alloys, some of which do not twin under normal strain rates and stresses. The twin density appears to increase with shock pressure and strain rate.

In general, if it is possible for a crystal to achieve a more close-packed structure, phase transformation at high pressure may occur. Thus body centered cubic (b.c.c.) metals and alloys may transform to face centered cubic (f.c.c.) or hexagonal close-packed (h.c.p.) structures and f.c.c. metals or alloys may transform to h.c.p. structures. Whether these high pressure phases are retained after unloading depends on the driving forces required for the reverse transformation (which is not always to the original phase).

The effects of shock loading on the five metals and alloys under investigation are considered in the following sections. The effect of the shock pulse features on the microstructure is recorded where possible, and proposed models of twin formation and phase transformations are detailed.

2.1.1 Aluminium

The relatively high stacking fault energy (s.f.e) of 135 mJ/m^2 [35] and ease of cross slip would indicate that a cellular structure should form in aluminium. However, although conventional cold rolling produces such a structure, aluminium shocked to 15GPa shows only a high density of randomly distributed, heavily jogged dislocations, dislocation loops and point defect clusters [34, 39]. This degree of clustering is not apparent in copper or nickel and the lack of a cellular structure is thought to be due to the pinning of dislocations by these clusters. Tensile tests on aluminium shock loaded to 10GPa show a yield drop which could have been caused by dislocations unpinning from these clusters [36].

No twinning has been detected although the s.f.e. lies considerably below that of nickel (240 mJ/m^2 [35]) which exhibits twinning under shock loading. This is thought to be due to the narrow extended dislocation width (1.08 b where b is the Burgers vector), which means that the dislocation is effectively undissociated and thus can

cross-slip easily. The high s.f.e. of nickel is accompanied by relatively wide extended dislocations (2.1b) which can form twins under shock pressures as they do not cross-slip as easily [38].

There appears to be a contradiction in the reasons given for the lack of twinning and cellular structure. An explanation could be that the large number of dislocations generated in the shock front are capable of cross-slipping initially but do not move very far before intersecting with other dislocations. The jogs thus formed create considerable numbers of point defects which have the effect of pinning subsequent slower-moving dislocations (e.g. during rarefaction). Thus when dislocation motion ceases neither twins nor a cellular structure would have been formed. Vacancy clusters may well be the result of vacancies collecting because of high shock and residual temperatures.

2.1.2 Duralumin Al-4%Cu

Nourbakhsh and Nutting [41] have shown the various microstructures occurring in the heavily cold rolled alloy.

As the deformation was increased, the dislocation density rose sharply until at 20% deformation individual dislocations were difficult to resolve, θ' plates were bent, sheared and twisted and above 45% strain, acted as sub-grain boundaries. After 95% strain, the θ' plates had disintegrated and the structure consisted of almost spherical particles. The deformation mode then changed and deformation bands were

formed.

The onset of deformation bands in a lightly aged alloy containing G.P. zones, or in an alloy in the as-quenched condition, occurred at 40% strain. Initially, some slip bands formed having {111} habit planes. These broadened with increasing strain and secondary slip bands formed in the lightly aged alloy. After 80% strain, shear bands were formed which crossed several grains and no longer had any crystallographic relationship to the matrix, running at about 35° to the rolling direction. The G.P. zone density decreased at high strains.

Experiments on the shock loading of dural have been few and confined solely to the effect of shock pulses on the precipitate size.

Greenhut and co-workers [42,43] shocked Al-4.65%Cu at 5 GPa with a triangular pulse to produce spall fracture and found that the precipitate size (G.P. zones, θ'' , θ') increased to a maximum near the spall zone (maximum tension). In addition, the type of precipitates changed from G.P. zones to θ' .

For comparison a momentum trap was used to remove the tensile (spalling) pulse from the specimen and it was found that the precipitate size increased only slightly with distance from the shocked surface.

The growth and / or transformation of these precipitates was attributed to the formation of excess vacancies by non-conservative jog motion in the tensile pulse. These vacancies form complexes with copper atoms and diffuse to precipitate sites where they dissociate. The free vacancies diffuse back into the matrix to collect more copper atoms by reforming complexes [47].

While there is little doubt that considerable vacancy generation is caused by the tensile pulse, it should also be mentioned that the rarefaction stress wave in the compressive pulse will also produce rapidly moving dislocations capable of generating vacancies. This might account for the slight increase in precipitate size observed in specimens subjected solely to a compressive pulse. As the pulse travels away from the shocked surface, the rarefaction stress rate decreases, allowing more time for dislocation motion and the subsequent generation of vacancies. This would obviate the need for the residual tensile pulse explanation used by Greenhut.

A contrasting view is given by Stein [44] who subjected as-quenched Al-3.85%Cu to a 1.4 G Pa shock of 0.085 μ s at liquid nitrogen temperature. It was found that G.P. zones were formed. He suggested that this was caused by enhanced diffusion of copper atoms down dislocation pipes. The vacancy mechanism was ruled out as it was thought that vacancies produced by shock are quickly

annihilated by the large numbers of neighbouring dislocations also generated by the shock.

While it is true that considerable vacancy annihilation may occur at shock temperatures, there is also considerable evidence that substantial vacancy concentrations exist after shock loading [21,45,46]. It should also be noted that all measurements took place at room temperature where both mechanisms are possible.

2.1.3 Copper

At one time it was generally believed that copper, when stressed, only formed dislocation cells whose size decreased with increasing load. The absence of deformation twins in spite of the relatively low stacking fault energy (80 mJm^{-2} [62]) was put down to the theory that a unit dislocation dissociating on a (111) plane

$$a/2 [110] \rightarrow a/3 [111] + a/6 [11\bar{2}]$$

would reunite on the same plane after the Shockley partial $a/6 [11\bar{2}]$ had rotated once about the nodes of the $a/2 [110]$ dislocation bent out of the (111) plane. The monolayer twins thus formed would not grow.

However, Blewitt et al [48] provided the first undisputed observation of deformation twins in copper deformed at 78K. This led to a revision of the pole mechanism for twinning by Venables [49] (see section 2.1.4.1 for details).

Subsequently deformation twinning has been frequently observed under shock loading indicating that it is a low temperature, high strain rate phenomenon.

2.1.3.1 Shock Loading

As the shock pressure and / or shock duration is increased the visible dislocation density increases to a maximum after which it appears to drop. It has been suggested that the latter effect is caused by the generation of incoherent microtwins and the attendant mismatch dislocations at the interfaces. In general, dislocation cells are formed whose size and wall thickness decrease with increasing shock pressure and duration. Appleton and Waddington [50] have shown that similar cell sizes (and hardness) can be obtained by shocking at 5 GPa for $2\mu\text{s}$ and at 10 GPa for $1\mu\text{s}$. However, cells were less well developed at small pulse durations. In this respect it should be noted that cell size increases with temperature and dislocations will tend to move to the cell walls under thermal activation. Cell formation appeared to be insensitive to rarefaction stress rate. Brillhart et al [52] have shown that the cell size seems to follow the Kuhlmann-Wilsdorf theory of work hardening [61], that is the cell size is inversely proportional to the square of the pressure.

There appears to be considerable variation in cell sizes determined for any one pressure by different experimenters.

Johari and Thomas [51] have suggested that this is because the cell walls are not along well-defined planes and thus the viewing plane is not always the same in electron microscopical work. The cell shape will also depend on the type of loading.

Other features such as dislocation bands on {111} planes at pressures above 11.2 GPa [53] and elongated cells above 7.5 GPa [52,54] have been reported but there is little consistency in the observations.

As previously mentioned, the dislocation density increases with increasing pressure until, at some point, twins are formed. There is little consensus on the threshold pressure for twinning - values ranging from 1.6 GPa [55] to 34.5 GPa [63] have been reported. This appears to be due to differences in loading geometry and (in the case of single crystals) loading direction.

Early work [56,57] using optical methods on single crystals obliquely shocked, indicated that twins formed in compression and rarefaction and that the threshold pressure depended on the orientation; twinning in [100] oriented crystals being more difficult than twinning in [111] crystals.

However, subsequent use of X-ray techniques to generate diffraction spots showed that the [111] oriented twinning threshold (20 GPa) was higher than the [100] threshold (14.5 GPa). It was noted that the [100] crystal twinned in

compression and the $[111]$ crystal twinned in tension at 20 GPa, whereas twinning occurred in both compression and tension above 20 GPa. As the pressure was increased, the distribution changed from many fine twins to a few thick twins [58]. On the other hand, Brillhart et al [52] found that the average twin separation decreased with increasing shock pressure. This result would contradict the previous work if the decreased separation was caused by an increase in the number of twins. However, wider twins could also cause this effect. Since the manner of the measurements was not given, the true relationship is impossible to determine.

Smith [56,59] showed that the twin width increased with distance from the free surface of a shocked crystal, indicating a rarefaction stress rate dependancy which may partly explain the results of De Angelis and Cohen [58].

An interesting viewpoint is given by Zukus and Taylor [60], who suggest that twinning cannot occur in the compressive shock front. They state that twinning should be caused by the resolved shear stresses present in the zone undergoing adjustment from one-dimensional (elastic wave) to three-dimensional (plastic wave) loading. This zone does not exist long enough ($<0.01\mu s$) for twinning to occur ($0.1\mu s$).

It can be seen that considerable disagreement exists as to the source and quantity of twins under shock loading. With the exception of the work by Brillhart et al [52] and

De Angelis and Cohen [58], the results were based on optical microscopy which would not pick up micro-twins and thus these threshold pressures must be viewed with suspicion.

If a pole mechanism for twinning exists it would seem reasonable to expect numerous, short, thin twins (if any do occur) caused by the steep compressive front and larger, wider twins of various dimensions caused by the rarefaction wave. Some threshold would exist for the propagation of the twins from the original stacking fault nucleus.

2.1.4 Chromium-Nickel Steels

These alloys exist initially in the f.c.c. (γ) phase which can twin (γ_t) and / or transform to the h.c.p. (ϵ) and b.c.c. (α') phases on cooling or under stress. The degree of transformation depends on the stacking fault energy (s.f.e.) since the transformations are strongly dependent on the behaviour of partial dislocations.

Empirical equations have been devised by Rhodes and Thompson [76] to relate the s.f.e. to the chromium and nickel contents of these alloys:

$$\text{s.f.e.} = 17.0 + 2.29\text{Ni} - 0.9\text{Cr} \text{ mJm}^{-2} \quad \text{for } <18\%\text{Cr} \quad (2.1)$$

$$\text{s.f.e.} = -26.6 + 0.73\text{Ni} + 2.26\text{Cr} \text{ mJm}^{-2} \quad \text{for } >18\%\text{Cr} \quad (2.2)$$

$$\frac{\delta \text{s.f.e.}}{\delta t} = 0.05 - 0.08 \text{ mJm}^{-2} \text{ } ^\circ\text{C}^{-1} \quad (2.3)$$

Table 2.1 gives calculated stacking fault energies for the alloys considered in this review.

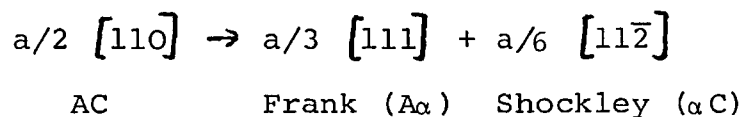
Details of the possible dislocation mechanisms responsible for these transformations, along with the postulated transformation sequences, have been the subject of much debate and the main conclusions are given below.

Finally, the effect of explosive loading and low strain rate loading on these alloys is discussed.

2.1.4.1 Twinning (γ_t) and H.C.P. (ϵ) Formation Mechanisms

(a) Pole Mechanisms

γ_t — If one considers a dislocation jogged on a (111) plane so that this jog has a Burgers vector $b = a/2[110]$ it can be seen that this jog could dissociate as follows



The Frank partial is sessile but the Shockley partial rotates about the poles formed at the ends of the jogs (Fig. 2.1 a,b). If the pole dislocation has a component b perpendicular to the (111) planes and of magnitude equal to the planar separation d_{111} then the separation of the Shockley partials

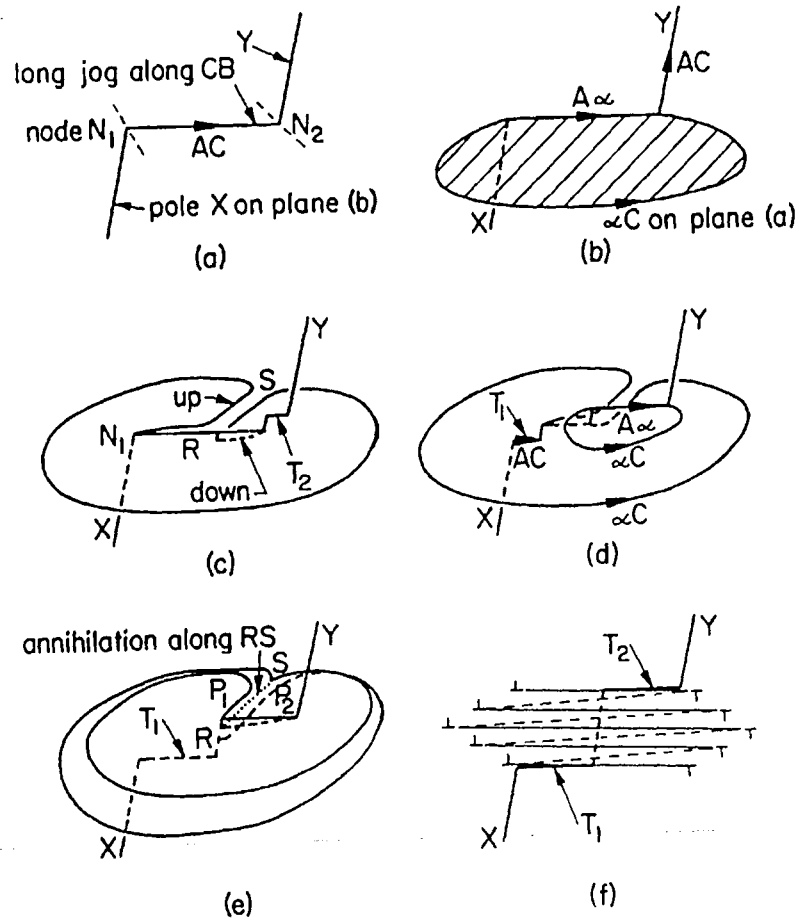


Fig. 2.1 F.C.C. twin formation by a pole mechanism
(Venables [49])

after each revolution will increase by d_{111} in the $[111]$ direction. Provided that the partials can overcome their mutual attraction at low separations, they will continue to spiral up and down to form a lenticular twin, whose thickness will be limited by the back stress and the length of the pole dislocations. The high initial barrier to the movement of the partials could be overcome by the passage of a unit jog along the Frank sessile partial (Fig. 2.1 c). As the unit jog moves, the source is recreated on the next plane (Fig. 2.1 d). This new source dissociates as before but after one revolution one of the revolving partials meets and annihilates one of the partials generated by the original source (Fig. 2.1 e). The remaining two partials are now $2d_{111}$ apart and experience less mutual attraction (Fig. 2.1 f) [49].

ϵ —A dislocation with $b = a/2 [211]$ lying in the $[111]$ direction has a screw component $2a/\sqrt{3}$ in the $[111]$ direction and an edge component $a/\sqrt{6}$ in the $[2\bar{1}\bar{1}]$ direction. If another dislocation, lying in the (111) plane, dissociates as follows

$$a/2 [\bar{1}10] \rightarrow a/6 [\bar{2}11] + a/6 [\bar{1}2\bar{1}]$$

the two partials sweep in opposite directions around the pole formed by the first dislocation to form an ϵ nucleus. After one circuit the separation is $2d_{111}$ which is equivalent to the c-axis of the ϵ phase. Growth proceeds in the $[111]$ direction provided the partials can overcome their

mutual attraction after one revolution [77,78,95].

(b) Overlapping Mechanisms (ϵ, γ_t)

1) Frank [79] has postulated that high speed dislocations can be reflected from free surfaces. On this basis, it was suggested that partial dislocations undergo stress induced reflections from stable stacking faults. These particles will reflect back one or two planes from the original slip plane giving γ_t (adjacent plane) or ϵ (alternate plane).

The newly-created partial travelling in the opposite direction collides with another fault to create another partial which moves back to the original fault, continuing the process. This would give high speed production of faults [80,81].

2) Cohen and Weertman [82] suggested that energetically favourable partials $a/6$ $[112]$ and $a/6$ $[\bar{1}21]$ on a $(11\bar{1})$ plane close up to form a perfect dislocation $a/2$ $[011]$ in a pile-up at a Lomer-Cottrell lock. This dislocation then dissociates to form a Frank sessile $a/3$ $[\bar{1}11]$ and a Shockley partial $a/6$ $[\bar{2}11]$. The Shockley partial moves in the opposite direction to the $a/6$ $[2\bar{1}\bar{1}]$ partials on the (111) plane, thus producing stacking faults. Large numbers of piled-up dislocations can generate overlapping stacking faults to produce γ_t twins or ϵ phase. Thus once a threshold stress is overcome, high speed fault generation is possible (Figure 2.2).

(c) Experimental Evidence

It is thought that the ϵ nucleus is a single intrinsic stacking fault. In-situ experiments by Brooks et al [74,75] showed that ϵ occurred where suitably-spaced stacking faults were formed. The passage of a Shockley partial produced ABAB stacking and resulted in a 1 - 2% contraction perpendicular to the $(111)_\gamma // (0001)_\epsilon$ plane. The ϵ phase grew when the faults overlapped. Few misfit dislocations were observed as the atomic spacings in the $\{0001\}_\epsilon$ planes are similar to those in the $(111)_\gamma$ plane. The observed orientation relationships were $\{111\}_\gamma // \{0001\}_\epsilon$ and $\langle 110 \rangle_\gamma // \langle 12\bar{1}0 \rangle_\epsilon$.

Enami et al. [83] showed that in-situ heating causes the ϵ phase to transform to γ by reducing along its length, leaving stacking faults behind. These subsequently shrink to form perfect dislocations at higher temperatures due to the temperature-dependent increase of the stacking fault energy.

At present there appears to be no direct evidence of pole mechanisms in the literature.

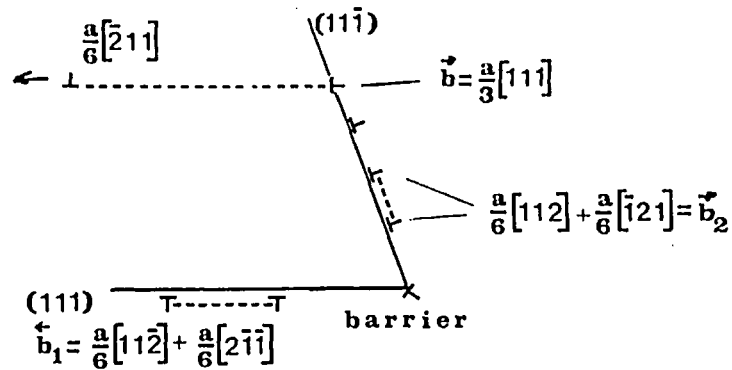
Pure ϵ only exists if the s.f.e. is less than zero but Lecroisey and Pineau [84] suggest that ϵ nucleation is preferred to γ_t formation at low temperatures and stresses and providing that the stacking fault energy is less than 30 mJm^{-2} .

Under shock loading, however, the stress is hydrostatic and thus the reduction in volume caused by the ϵ phase transformation would be favoured under compression.

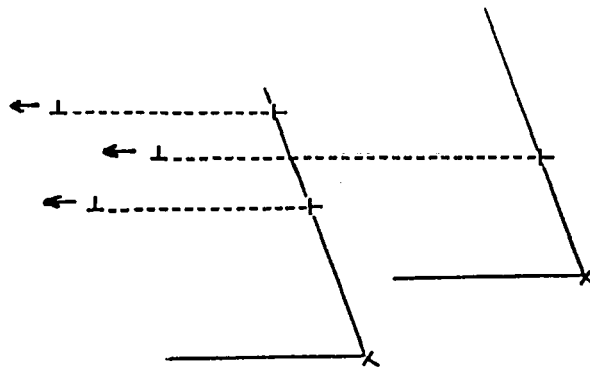
Explosively loaded flyer plate experiments conducted by Murr and co-workers [39,87,88,89,90,91,92] on 19/9 steels have shown that γ_t can form at shock pulse times of less than $0.5\mu\text{s}$, a fact which appears to favour the overlap mechanism. Below 15 GPa the stacking fault density increased with shock pressure at constant pulse duration. Above 15 GPa twins were formed whose density and width increased with pressure. Tests at 10 GPa showed that ϵ formation was more extensive with decreasing grain size but was substituted by γ_t in larger grains. ϵ was found to form preferentially in the bulk of the specimens. This was due perhaps to the high surface temperatures or the relaxation of the stress conditions at the surfaces. Above 15 GPa little ϵ phase was found.

2.1.4.2 B.C.C. (α') Formation Mechanisms

(a) Brooks et al. [74,75] suggested that α' nucleation is associated with dislocation pile-ups on an active slip plane. $a/6 \langle 112 \rangle$ partials are forced close together and the atomic structure close to the cores of the slightly separated partials is very close to the b.c.c. stacking (Figure 2.3). The volume of effective b.c.c. material increases as more dislocations join the pile-up until the nucleus reaches a critical size and rapid growth takes

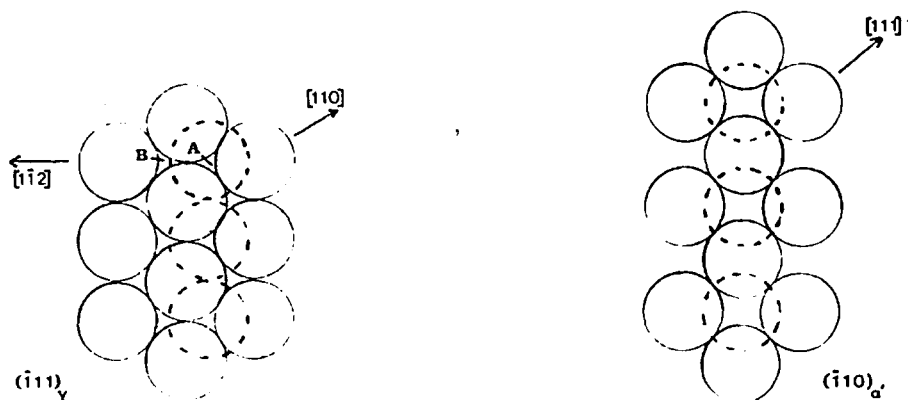


(a) Dislocation dissociation in pile-ups on (111) and $(11\bar{1})$ planes



(b) Overlap of dissociated dislocations to form twins or ϵ

Fig. 2.2 F.C.C. twin (γ_t) or h.c.p. (ϵ) phase formation by overlap mechanism (Cohen and Weertman [82])



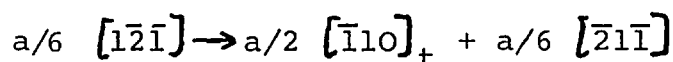
(a) F.C.C. structure

(b) Pseudo b.c.c. structure after a shear of $\frac{1}{12} [112]$ ($\frac{1}{2}AB$)

Fig. 2.3 B.C.C. (α') nucleation from an f.c.c. (γ) lattice (Brooks et al. [74])

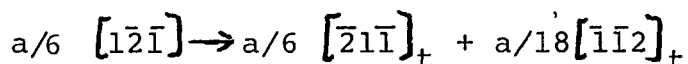
place. Dislocations required to accommodate misfit strains are formed at the interface during growth and account for the highly dislocated nature of the α' .

(b) Lacroisey and Pineau [84] suggested that α' forms at the intersections of deformation bands of any type. If, for example, a deformation twin on a (111) plane is to be crossed by a Shockley partial $a/6$ $[1\bar{2}\bar{1}]$ moving on a $(11\bar{1})$ plane, the following reaction can take place



where the subscript t indicates that the Burgers vector is in twin coordinates. The $a/2 [\bar{1}10]_t$ dislocation can dissociate further in the twinned region. The reaction above requires considerable energy - almost equal to that required for cross-slip.

However, if the α' phase is sufficiently more stable than the γ phase, the following reaction could take place:



The transformation dislocation $a/18 [\bar{1}\bar{1}2]_t$ moves along the interface creating an α' nucleus [93]. The $a/6 [\bar{2}1\bar{1}]_t$ dislocation generates another α' nucleus at the exit interface. These reactions can produce laths along the intersection of the two planes, i.e. along the $[\bar{1}10]$ direction (Figure 2.4).

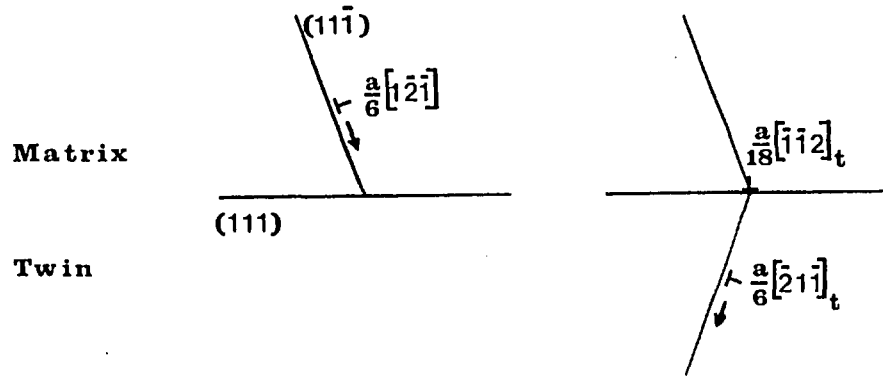


Fig. 2.4 B.C.C. (α') nucleation at the intersection of a twin and a slip plane or a twin (Bogers and Burgers [93])

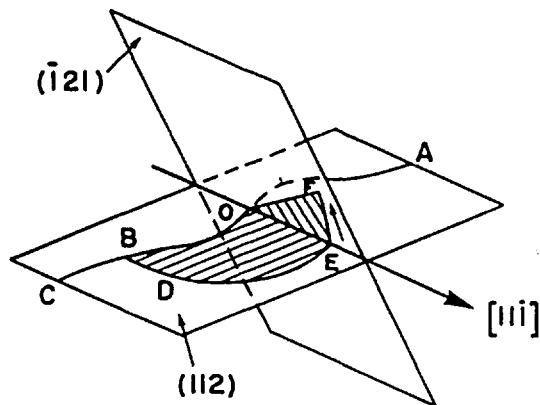


Fig. 2.5 B.C.C. twin formation by a pole mechanism (Cottrell and Bilby [97])

(c) Murr and Grace [39] suggested a simple mechanism for the generation of α' whereby partials dissociate into quarter dislocations, that is:

$$a/6 \langle 112 \rangle \rightarrow a/12 \langle 112 \rangle + a/12 \langle 112 \rangle ,$$

which pass through a twinned or hexagonal region. This was thought to occur solely to accommodate the shear and compression in cold rolling since at the time no α' had been found at shock pressures.

Two orientation relationships are most frequently detected between the α' and the γ phases. They are:

(1) The Nishiyama - Wassermann (N.W.) relationship

$$\begin{aligned} \{011\}_{\alpha'} // \{111\}_{\gamma} \\ \langle 100 \rangle_{\alpha'} // \langle 1\bar{1}0 \rangle_{\gamma} \\ \langle 01\bar{1} \rangle_{\alpha'} // \langle 11\bar{2} \rangle_{\gamma} \end{aligned}$$

and

(2) The Kurdjumov - Sachs (K.S.) relationship

$$\begin{aligned} \{011\}_{\alpha'} // \{111\}_{\gamma} \\ \langle 11\bar{1} \rangle_{\alpha'} // \langle 10\bar{1} \rangle_{\gamma} \\ \langle 2\bar{1}1 \rangle_{\alpha'} // \langle 1\bar{2}1 \rangle_{\gamma} \end{aligned}$$

These relationships are quite close to each other, being separated by a rotation of $5^{\circ}16'$ about the $\langle 011 \rangle_{\alpha'} // \langle 111 \rangle_{\gamma}$ axis. Thus the determination of the exact γ to α' relationship in

various Cr - Ni steels has not been unanimous.

Manganon and Thomas [64] , using quenched and subsequently deformed 19/9 steel, suggested that the initial orientation was N.W. which changed to K.S. during growth. However, recent work on the in situ deformation and cooling of 16/14, 16/12 and 18/10 steels favours the K.S. orientation [75]. Work on 18/8 and 18/10 cold rolled steels shows a mixture of α' crystals in both the N.W. and K.S. orientations [67,68].

Breedis [71] postulated that the apparent existence of both N.W. and K.S. orientations, as evidenced by the frequent occurrence of both the $[100]_{\alpha'}$ and the $[\bar{1}11]_{\alpha'}$ diffraction patterns superimposed on the $[1\bar{1}0]_{\gamma}$ pattern, could be explained in terms of a twin relationship between the laths. He suggested that the $[100]_{\alpha'}$ pattern is actually tilted by 5° from the $[1\bar{1}0]_{\gamma}$ direction and is caused by a crystal which is a twin-related variant of the crystal producing the $[\bar{1}11]_{\alpha'}$ pattern. Hence the true orientation relationship is K.S.

The (α, α') phase can occur in three forms [71]:

1. Internally twinned lenticular martensite plates in low chromium-high nickel steels, e.g. 3/27.

2. Internally dislocated martensite needles or laths which can be twin-related and which form initially parallel to $\{111\}_{\gamma}$ planes and extend as needles in a direction close to the $[110]_{\gamma}$ directions. These laths occur in Cr - Ni alloys with more than 10 wt % nickel and chromium.

3. Large α crystals having little resemblance to plate or lath martensite [94]. It is thought that these crystals may not be due to a martensite reaction. They have been found in 16/8 and 8/16 Cr - Ni alloys and some Ni-Fe alloys.

The formation of lath martensite is considered in this work and will thus be reviewed in detail below.

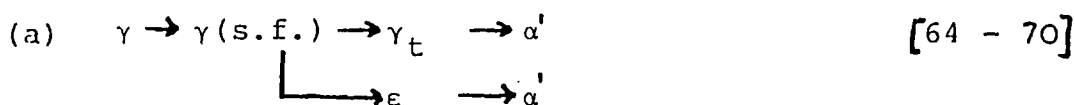
Martensite laths occur on cooling Cr - Ni alloys below the M_s temperature or as the result of cold rolling. In situ cold deformation shows that laths initially appear to grow perpendicular to the $\{111\}_{\gamma}$ nucleus plane but that the habit plane then changes toward the $\{225\}_{\gamma}$ plane. Growth starts on both sides of the slip plane but continues on only one side [74,75]. The work of Brooks et al [74,75] and others [65,66,70,71,95] suggests that the lath habit plane varies from $\{112\}_{\gamma}$ to $\{225\}_{\gamma}$. The amount of α' increases with deformation at the expense of γ_t and ϵ (if present).

It should be noted that the $\gamma \rightarrow \alpha'$ transformation involves an increase in volume of 2.45% which at first sight would seem unlikely under stress. However, under uniaxial or biaxial stress α' is thought to grow by continuous stress relief (highly stressed ϵ expands to low density α'). The transformation is however very sluggish and requires thermal activation or external pressure to continue. This has been observed by Manganon and Thomas [64].

It was initially thought that α' laths were not formed under shock loading [89] but subsequent work showed that α' formation depended on the shock pressure duration (t_s). At 25 GPa α' was first detected at $t_s = 2\mu s$ and its quantity increased with t_s up to $6\mu s$. In addition the rate of α' increase with t_s was greatest at the greatest shock pressure (45 GPa) [87,90]. The amount of α' formation also increased with decreasing grain size [88]. It has been suggested that shock sensitizes the austenite (γ) towards a strain induced α' transformation by causing the formation of small α' crystals at slip intersections during loading. These would then grow by pressure relief during rarefaction [91].

2.1.4.3 Transformation Sequence

Various theories exist concerning the transformation sequence from γ to α' and ϵ . They are as follows:



In case (a) α' is thought to form at the intersections of twins with other twins, ϵ plates and grain boundaries, and at the intersections of ϵ plates with grain boundaries and other ϵ plates.

In case (b) it has been suggested that α' forms directly from γ or $\gamma(s.f.)$ and that the stresses produced in the austenite (γ) by the formation of α' crystals above a certain size are sufficient to induce extensive dislocation dissociation leading to heavily faulted ϵ bands. Thus ϵ is not an intermediate phase.

In the final case (c) α' is thought to form directly from piled-up dissociated dislocations occurring on slip bands. ϵ and γ_t are formed by the suitable overlapping of stacking faults $\gamma(s.f.)$. Thus γ_t , ϵ and α' form independently of one another and their quantities appear to be dependent on the stacking fault energy.

Generally it has been observed that, as the amount of cold rolling deformation is increased the quantity of ϵ increases to a maximum and then decreases to zero at about 50% deformation whereas the α' content increases steadily [67]. This would tend to support (a) and (c) since increased deformation would lead to greater pile-ups of partial dislocations on $\{111\}_{\gamma}$ type planes - be they slip planes, twin plates or ϵ plates - and these partials are responsible for the formation of α' .

Thus twin and ϵ formation should increase with deformation, i.e. more stacking faults. However, as more pile-ups occur the stacking faults close up leading to the production of α' and, after sufficient deformation, the reduction of γ_t and ϵ .

In case (b) it seems reasonable to expect that the amount of α' , γ_t and ϵ would increase until most of the γ was twinned or transformed. At this point, the α' phase might continue to increase at the expense of the ϵ and γ_t . However, the results of Bowkett and Harries [67] on 18/10 steel showed a maximum amount of ϵ at 20% deformation which left approximately 40% of unaltered austenite if one assumes that the amount of γ_t equalled the amount of ϵ .

2.1.4.4 Shock Loading

Since the formation of α' involves a decrease in density, it seems unlikely that it forms under conditions of triaxial (hydrostatic) pressure existing during shock compression. Formation of the denser ϵ phase should be preferred at high pressure and subsequent relief to α' would occur during the rarefaction from the shock pressure to atmospheric pressure. The formation of stacking faults and twins could occur during both compression and rarefaction, since dislocation generation and motion would occur in both cases. The time of loading and / or the rarefaction stress rate would indicate whether a pole or an overlap mechanism was responsible for γ_t and ϵ formation. In the case of the ϵ phase, the completeness of the transformation and the threshold stress would determine if or when a double shock front was formed.

2.1.5 Iron

Iron exists in the form of a b.c.c. structure at room temperatures and pressure. It will twin on the $\{112\}$ planes at high strain rates or low temperatures and transform to h.c.p (ϵ) at high pressures. A high stacking fault energy is indicated by the absence of any reported stacking faults. In fact Vitek [111] has calculated that stable stacking faults are impossible in b.c.c. metals. This absence of dissociated

dislocations permits cross-slip, thus producing a cell structure.

2.1.5.1 Twinning

It is well known that iron twins (Neumann bands) at low temperature or high strain rates and Hockett and Zukas [96] have suggested that the empirical inequation:

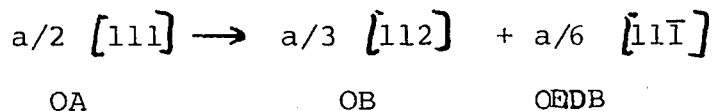
$$\log \dot{\epsilon} > -1744/T + 5.9031$$

gives the strain rates ($\dot{\epsilon}$) where twinning occurs for any temperature (T). Thus twinning would be expected to occur at Hopkinson bar (10^3s^{-1}) and explosive (10^6s^{-1}) strain rates.

Models for the generation of iron twins can be divided into pole mechanisms and homogeneous nucleation mechanisms.

(a) Pole Mechanism

Cottrell and Bilby [97] have proposed the mechanism shown in Fig. 2.5. The unit dislocation AO has a Burgers vector $a/2 [111]$ and lies in the (112) plane but not along $[11\bar{1}]$. It is dissociated at O as follows:-



where OB is sessile and OEDB is mobile. Thus OEDB moves on the (112) plane leaving a stacking fault (monolayer twin) behind. OEDB can also glide in the $(\bar{1}21)$ plane and the pure screw part OE can cross-glide onto this plane. Therefore the dislocation moves onto the $(\bar{1}21)$ plane. Now AO has a component in the $[\bar{1}21]$ direction, so that each revolution of OF results in a step of $a/6 [\bar{1}21]$ along OB. This could result in a twin of considerable thickness.

As in the case of the f.c.c. pole mechanism, the stress required to overcome the mutual attraction of two partials on neighbouring planes after one revolution would be the barrier to further growth.

This problem has been avoided by Priestner and Leslie [98] in a model where dislocation pile-ups on intersections between $\{110\}$ and $\{112\}$ or between $\{112\}$ planes can lead to the generation of $a/6 \langle 111 \rangle$ partials on three adjacent $\{112\}$ planes. This produces a three-layer twin and growth could occur by a pole mechanism about forest dislocations intersecting the $\{112\}$ planes. In this case, the partials pass each other at three planar spacings and the twinning stress is considerably reduced. The dislocation reactions for a pile-up at the junction of two $\{112\}$ slip planes are as follows. A barrier dislocation $a[001]$ is formed at the junction and this dissociates due to dislocation pile-ups on the slip planes:

$$a [001] \rightarrow a/3 [002] + a/6 [\bar{1}11] + a/b [1\bar{1}1] \quad \text{1st plane}$$

$$a/3 [002] \rightarrow a/3 [001] + a/6 [\bar{1}11] + a/6 [1\bar{1}1] \quad \text{2nd plane}$$

$$a/3 [001] \rightarrow a/6 [\bar{1}11] + a/6 [1\bar{1}1] \quad \text{3rd plane}$$

The $a/6 [\bar{1}11]$ and $a/6 [1\bar{1}1]$ partials form three-layer twins on the $(1\bar{1}2)$ and $(\bar{1}12)$ planes respectively. These partials do not glide on only one plane as this would produce a high energy transformation structure.

Similar reactions can be produced for twinning at the junction of $\{110\}$ and $\{112\}$ planes.

(b) Homogeneous Nucleation

Orowan [99] first suggested that twin nuclei could be formed where stress concentrations increase the locally applied stress to the theoretical twinning stress given by $G\gamma/2\pi$. These nuclei would be formed by local homogeneous shear of the lattice, and not by dislocation motion. Work on zinc and tin suggests that prior dislocation concentrations are not required for twin formation and that growth occurs by repeated nucleation and motion of twinning dislocations. It was found that the rate of growth perpendicular to the twinning plane was considerably less than that along the twin plane. This is thought to be due to the difficulty of nucleating new twinning dislocations which widen the twin compared to the ease of dislocation motion which lengthens the

twin $[100,101]$.

2.1.5.2 Phase Transformation

Iron transforms under pressure from a low density b.c.c. (α) structure to a high density h.c.p. (ϵ) structure at about 13 GPa. This transformation is a baric, starting at 13 GPa and completing at about 16.3 GPa. The reverse $\epsilon \rightarrow \alpha$ transformation begins at 8.1 GPa and is completed at 4.5 GPa $[102]$.

The transformation mechanism is not known but two possible models are given below.

It is known that the $\{110\}_\alpha$ plane is parallel to the $\{0001\}_\epsilon$ plane and a distorted h.c.p. cell can be picked out from the b.c.c. crystal by using every other parallel $\{110\}_\alpha$ plane.

Burgers $[110]$ proposed that if the atoms on every other $\langle 110 \rangle$ plane are sheared in the $\langle 111 \rangle_\alpha$ direction, the basal planes form a parallelogram having a 60° angle after a slight rearrangement of the atoms (Fig. 2.6 a, b, c, c'). This then becomes the $\{0001\}_\epsilon$ basal plane. The atoms in the intervening $\{110\}_\alpha$ planes must move in the $\{10\bar{1}0\}_\epsilon$ direction to the low energy position under the centre of gravity of the triangles in the $\{0001\}_\epsilon$ plane (Fig. 2.6 d).

Alternatively, a shear in the $\langle 0\bar{1}1 \rangle$ direction on the $\{011\}$ plane could move atoms from unstable saddle points to more stable positions corresponding to a close-packed structure (Figure 2.7). It has been suggested that the dislocation dissociation mechanism

$$a/2 [1\bar{1}1] \rightarrow a/8 [0\bar{1}1] + a/4 [2\bar{1}1] + a/8 [0\bar{1}1]$$

could provide this shear but such dissociations have not been observed, probably due to the high stacking fault energy $[111]$.

2.1.5.3 Shock Loading

Below 13 GPa the shock consists of an elastic shock (e) followed by a plastic shock (Pl). Twinning has been observed at pressures as low as 0.25 GPa. Generally the number of twins and their directions increases with increasing shock pressure. It has been suggested on the basis of the oblique shocking of iron that twins form between the passage of the e and Pl shocks. They are thought to be the result of an adjustment from one-dimensional to three-dimensional loading during the period between the e and Pl shocks. Twin related pressure relaxation after the e shock has been observed for shock separations of more than $0.4 \mu s$ [21,103,104].

Iron, which has not transformed under shock, contains straight, uniformly distributed dislocations running in

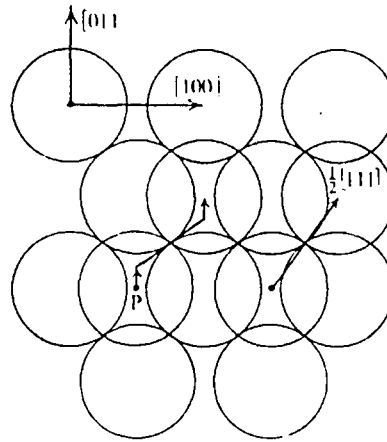


Fig. 2.7 B.C.C.(α) to h.c.p.(ϵ) phase transformation model (Kelly and Groves [129])

$\langle 111 \rangle$ directions on $\{110\}_\alpha$ planes and bands of closely-spaced dislocations interspersed with twins on $\{112\}_\alpha$ planes. Twins appear to nucleate at $\{112\}$ slip plane intersections, but are prevented from forming by prior cold work [105]. The lack of cell structure is attributed to the inability of screw dislocations to cross-slip at high strain rates. This contrasts with the cellular dislocation structure formed by cold rolling which is thought to prevent subsequent twinning at high pressures.

The $\alpha - \epsilon$ phase transformation occurs above 13 GPa. This results in a relatively large volume change, the generation of a great number of dislocations and some relaxation from the (probably overdriven P1 pressure. The single shock wave splits into two waves (see Appendix A 3), a leading P1 wave (13 GPa) and a following P2 wave which rises to the final shock pressure after the transformation. This two-wave structure exists up to about 35 GPa at which point the P2 shock velocity is greater than the P1 velocity. The ϵ wave velocity is exceeded at about 55 GPa.

Optical microscopy has shown the effect of the two-wave structure on the microstructure. When the P1 wave reaches a free surface, it reflects and meets the following P2 wave. If this reduces the pressure below the transformation pressure, a region exists near the free surface which has not transformed and reveals the microstructure caused by the passage of the P1 shock.

As previously mentioned, the P1 wave leaves twins. Single crystal studies have shown that twins appear to continue unaltered into the P2 region, thus suggesting that they persist during the transformation phase. Twin formation during rarefaction is considered unlikely since it has been found that prior deformation (in this case due to the transformation) inhibits twinning [105,106].

P2 regions for pressures below 16.3 GPa leave a two-phase region which is attributable to the abaric nature of the $\alpha \rightarrow \epsilon$ transformation. Above 16.3 GPa the structure becomes more uniform and optical microscopy shows that the structure consists of grains with fine needles and dark etched grains. Above 55 GPa the structure becomes simpler and less needle-like [56,107].

Electron microscopical work has shown traces of twins that appear to have survived the transformation. The martensite produced by the $\epsilon \rightarrow \alpha'$ transformation is sometimes twinned. These twins appear to be shorter, thicker and more widely spaced than those formed by the normal $\alpha \rightarrow \gamma \rightarrow \alpha'$ transformation sequence [105].

A thin region next to the free surface which is free of twins and transformation structure has been observed in iron shocked to above 35 GPa. This is thought to be due to elastic wave reflections or the shock front structure [108].

While ϵ usually reverts completely back to α' during unloading, investigations by Woodward and Aghan [109] on shear zones in Fe-0.5% C shocked to 30 GPa have shown white etched zones which contained a mixture of austenite, ferrite and h.c.p. (ϵ) phases. This was due to the high degree of local heating and the subsequent rapid cooling rate in the shear zones, both of which were caused by the high shear strain rate.

2.1.6 Models for the Shock Compression of Metals

(a) Smith [56] has tentatively proposed that the density discontinuity across a shock front could be accommodated by a two-dimensional array of dislocations travelling with the shock front (Fig. 2.8). This seems somewhat unlikely, since the dislocations would have to move at supersonic velocities [112]. It should be pointed out, however, that supersonic dislocations may occur in diffusionless transformations [113]. Finally, Smith's model does not predict the shock hardening of metals.

(b) Davis and Jackson [114] have suggested that the high shear stresses associated with the shock front would be sufficient to activate dislocation sources in a continuous manner.

These would generate dislocation loops. The need for supersonic dislocation movement would be avoided if the loop

expansion were limited to the shock front thickness. The dislocations left behind by the shock front would account for the shock hardness. In his work on iron, Hornbogen [115] has proposed a more detailed model. He suggests that the edge components of these loops travel with the shock front, thus drawing out the screw components in the shock directions. He has, however, neglected the reverse strain existing during rarefaction and his observations may be a consequence of recovery effects.

Cowan [116] has proposed that the shear stress will exceed the theoretical shear strength of the metal at high shock pressures causing a marked change in the mechanism of deformation. This has been related to the shock disordering of ordered alloys and to the formation of fine twins in copper [57] and nickel [117].

(c) A recent model by Meyers [118] considers that dislocations are homogeneously generated at the shock front when the deviatoric stresses exceed a critical value. The dislocations remain as the shock front moves on, creating no more dislocations until once again the critical deviatoric stress is reached. At this point, dislocations are again nucleated homogeneously (Figure 2.9).

Calculations based on this model give dislocation densities in shock loaded nickel which are an order of magnitude greater than the measured values. This is attributed to

three factors:-

- (1) The deviatoric stresses are not reduced to zero after each nucleation point since considerable elastic strain can be absorbed.
- (2) Dislocation annihilation takes place due to the high pressure and temperature existing during shock.
- (3) The nucleated dislocations will tend to be driven in the direction of the receding shock front due to stress fields caused by the previously-formed dislocation layer. This will lead to an increase in the distances between the dislocation layers because of the slower build-up of deviatoric stresses.

Meyers concludes that the plastic strains caused by the shock front are the result of dislocation generation and not of dislocation movement. The accommodation of deviatoric strains during rarefaction takes place by both the movement of existing dislocations and the generation of new dislocations. This removes the requirement put forward in Smith's model.

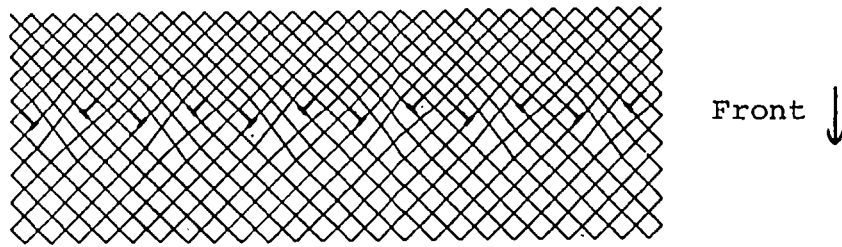
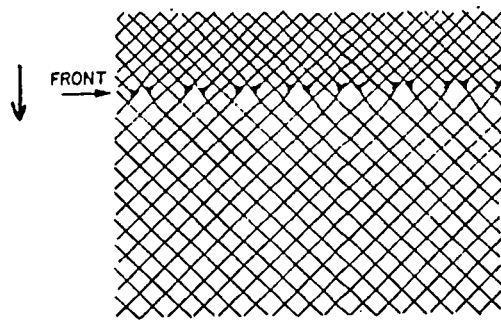
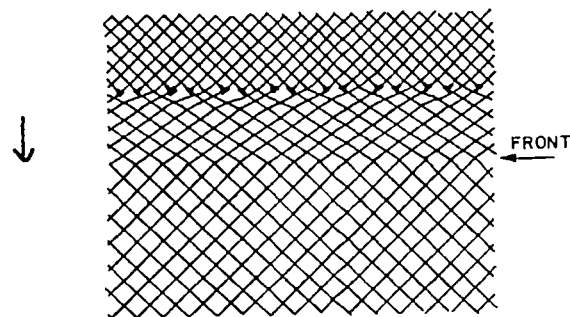


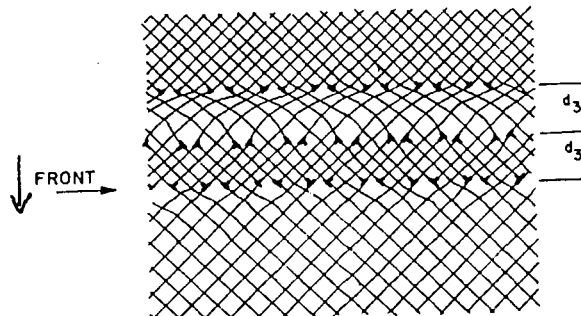
Fig. 2.8 Smith's model of dislocations in a shock front [56]



(a) dislocation interface coinciding with front.



(b) dislocation interface left behind, as front advances.



(c) new dislocation interfaces being formed.

Fig. 2.9 Meyer's model of homogeneous dislocation generation in a shock front [118]

Table 2.1 Calculated Stacking Fault Energies for
some Cr-Ni Steels

Cr(%)	Ni(%)	SFE (mJm ⁻²)
19	11	24.37
19	9	22.85
18	12	28.28
18	10	23.7
18	8	19.12
17	13	31.47
17	11	26.89
17	9	22.31
16	14	34.66
16	12	30.08
8	16	46.44
3	27	76.13

CHAPTER IIITHE SHOCK LOADING ASSEMBLY AND THE HOPKINSON BAR APPARATUS3.1 The Shock Loading Assembly

The assembly is divided into four parts:-

- (a) The explosive charge
- (b) The metal specimen
- (c) The free surface velocity measurement
- (d) The shock wave velocity measurement

The dimensions of the explosive charge and the metal specimen are determined by the requirement that the pressure applied at the explosive - metal interface be constant until the shock front and the ensuing free-surface rarefaction waves have traversed the specimen. In other words, the pressure should remain roughly constant until the last rarefaction wave, caused by the reflection of the shock front at the free surface of the specimen, reaches the explosive - metal interface. Moreover, rarefaction waves coming from the circumference of the specimen plate (spall plate) should not reach the specimen until the above sequence has been completed. In this way the specimen can be subjected to

a pressure pulse of known pressure and direction whose duration varies from zero at the specimen free surface to a maximum at the explosive - metal interface.

As the data used in the calculations for the explosive - metal assembly was gathered from the literature it was deemed advisable to confirm the results. Thus a series of experiments was undertaken to determine the shock parameters.

It is almost impossible to measure the specimen shock pressure directly and thus it must be calculated from more easily measured parameters. The shock wave velocity U_s and the free surface velocity u_{fs} are used.

The design of the explosive charge and the metal specimen, as well as the conversion of the shock wave and free surface velocities into a shock pressure, require the use of the Hugoniot equations of state and the Rankine-Hugoniot conservation equations for a shock front.

3.1.1 Shock Wave Theory

A shock front moving with velocity U_s into a medium at pressure P_0 , density ρ_0 , particle velocity u_0 and specific internal energy E_0 compresses it to P_1 , changing the remaining parameters to ρ_1 , u_1 , E_1 .

The Rankine-Hugoniot conservation equations are:-

$$U_s - u_0 = v_0 \left[\frac{P_1 - P_0}{v_0 - v_1} \right]^{\frac{1}{2}} \quad (3.1)$$

$$u_1 - u_0 = \left[(P_1 - P_0) (v_0 - v_1) \right]^{\frac{1}{2}} \quad (3.2)$$

$$E_1 - E_0 = \frac{1}{2} (P_1 + P_0) (v_0 - v_1) \quad (3.3)$$

where $v = 1/\rho$

These relationships are based on three assumptions [8,9,10,11] .

(a) Thermodynamic equilibrium is assumed to exist before and immediately after the passage of the shock wave.

This is a necessary condition for the existence of a shock wave with a constant velocity U_s . The condition is satisfied if thermodynamic equilibrium is attained in 10^{-8} seconds or less, since shock fronts are of the order of a few tenths of a millimeter thick. Although this is often the case, some polymorphic transitions can take longer than 10^{-6} seconds, and hence a state of quasi-equilibrium can exist behind the shock, giving rise to values of U_s and u_{fs} which do not reflect the true equilibrium state of the material.

(b) The shock front must attain a steady state profile and speed. Otherwise the measured value of U_s will be

an average over the distance traversed by the shock front and a spreading shock front will lead to inaccuracies in the determination of the passage of the shock. This condition is fulfilled provided that the velocity of the shock wave is greater than the sonic velocity in the medium ahead of the shock. The shock front thickness is then determined mainly by viscosity and heat conductivity.

(c) The shock compression is assumed to equal that produced by the same hydrostatic pressure. Since shock wave compression is essentially one-dimensional and a difference exists between the static and dynamic yield points, this assumption is not exact. However, if the pressure is far above the yield point, the effects of material rigidity can be neglected and the material can be treated as a fluid.

The free surface velocity is given by

$$u_{fS} = [(P_1 - P_0)(V_0 - V_1)]^{\frac{1}{2}} + \int_{P_0}^{P_1} \left[-\frac{\delta V}{\delta P} \right]_S^{\frac{1}{2}} dP \quad (3.4)$$

It has been shown that the two terms are approximately equal up to 50 GPa. Thus in most cases

$$u_{fS} \approx 2 [(P_1 - P_0)(V_0 - V_1)]^{\frac{1}{2}} = 2u_1 \quad (3.5)$$

Thus the free surface velocity can be related to the particle velocity during compression and used with the

shock velocity in the Rankine-Hugoniot equations to determine the remaining shock parameters.

The development of the above equations and the conditions necessary for shock front and rarefaction wave formation are given in Appendix A.

3.1.2 The Shape of the Pressure Pulse in a Detonating Explosive Charge [7,16,17]

Chemical reaction is initiated in an explosive by heating to bursting point a wire embedded in a sensitive explosive component. The resulting heat and pressure in the explosive produce a detonation shock which travels through the explosive at supersonic velocity. There is an almost discontinuous increase in pressure P , temperature T , and density ρ across this shock. A reaction zone follows in which the temperature rises and the pressure decreases as the chemical reaction goes to completion. This is characterised by the von Neumann spike which terminates at the Chapman-Jouguet (C.-J.) pressure at which the reaction is complete. The C.-J. state is the characteristic usually quoted for explosives. The subsequent variation of P , ρ and T is determined by the hydrodynamic expansion of the reaction products under conditions of nearly constant entropy.

Thus the shape of the pressure wave in an explosion reflects a shock followed by a rapid pressure decay to the C.-J. pressure, and then a slower decay to the initial pressure.

(Fig. 3.1)

If the explosion is considered as a shock at the C-J state moving into the unreacted explosive, the Hugoniot equations (3.1 - 3.3) and the Riemann equation (A30) apply. The pressure induced by the explosion in other media can then be calculated using these equations.

3.1.3 The Effect of an Explosive Shock on the State of an adjacent Solid [7]

Consider a solid with a typical Hugoniot (Fig. 3.2) which undergoes a shock due to an explosive at the C.-J. state P_{CJ}, V_{CJ}, u_{CJ} . Two possible cases exist. In the first case the pressure induced in the solid may be greater than P_{CJ} or, in the second case, the induced pressure may be less than P_{CJ} .

In the first case, the incident shock wave generates a shock wave in the adjoining solid corresponding to a state P_2, ρ_2, u_2 . The pressure and particle velocity across the explosive-solid interface must be continuous. Thus since $P_2 > P_{CJ}$ a compressive shock propagates back into the explosive, changing its state from $(P_{CJ}, \rho_{CJ}, u_{CJ})$ to (P_2, ρ_2', u_2') .

Let the shock wave velocities be U_1, U_2 and U_2' as given in Fig. 3.3. The conservation laws still apply and can be expressed across the reflected shock wave in the form:

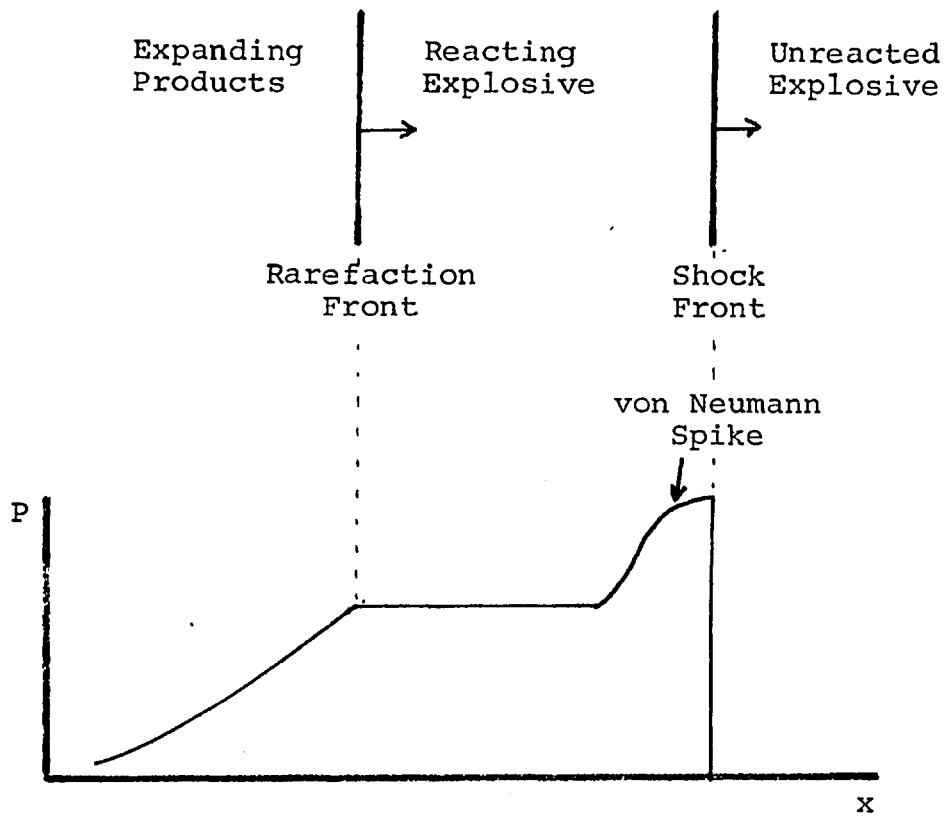


Fig. 3.1 Detonation in an explosive charge

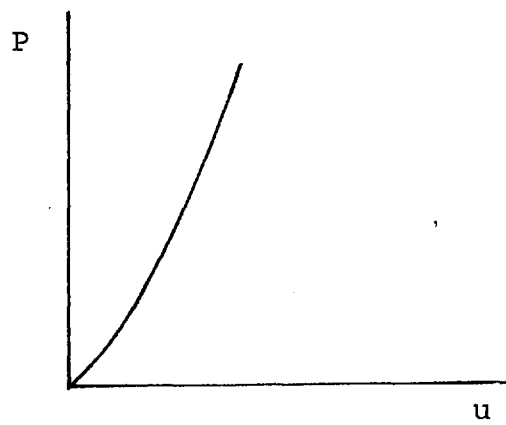


Fig. 3.2 Hugoniot curve in the P, u plane

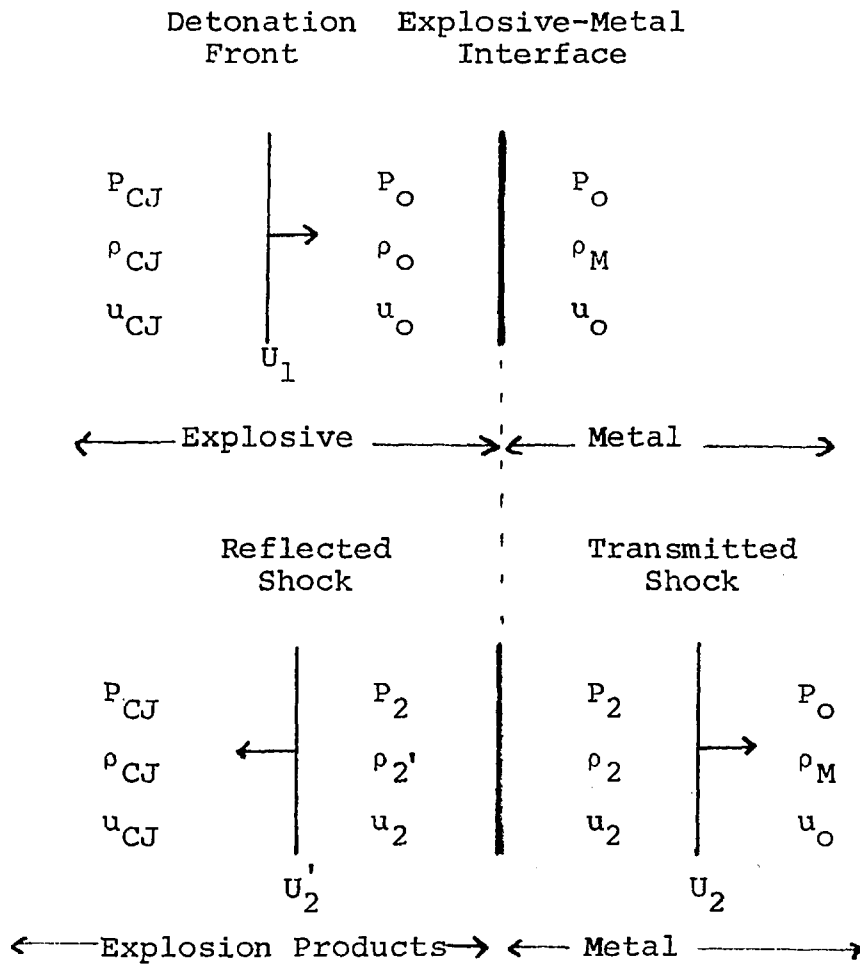


Fig. 3.3 Effect of an explosion on an adjacent metal

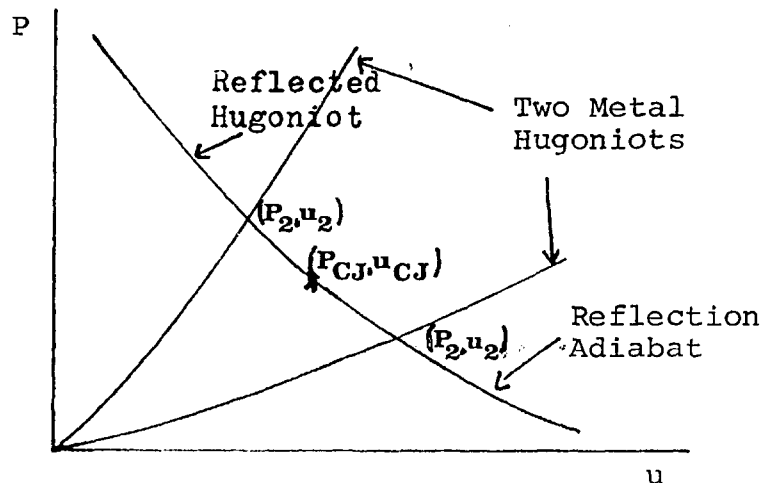


Fig. 3.4 Determination of the metal pressure and particle velocity

$$\rho_2(U_2' + u_2) = \rho_1(u_1 + U_2') \quad (3.6)$$

$$P_2 - P_1 = \rho_1(U_2' + u_1)(u_1 - u_2) \quad (3.7)$$

$$P_1 u_1 - P_2 u_2 = \rho_1(U_2' + u_1) \left[E_2 - E_1 + \frac{1}{2}(u_2^2 - u_1^2) \right] \quad (3.8)$$

The Rankine-Hugoniot relation for the reflected shock wave becomes

$$E_2 - E_1 = \frac{1}{2}(V_1 - V_2)(P_2 + P_1) \quad (3.9)$$

or in the (P, u) plane

$$u_1 - u_2 = \left[(P_2 - P_1)(V_1 - V_2) \right]^{\frac{1}{2}} \quad (3.10)$$

If the detonation gases are assumed to have the pressure-density relation of a polytropic gas [18], i.e.

$$P = a\rho^\gamma \quad ; \quad (3.11)$$

eliminating ρ between equations (3.10) and (3.11) gives the reflected Hugoniot

$$(u_{CJ} - u)^2 = \left[(P/P_{CJ}) - 1 \right] \left[1 - (P_{CJ}/P)^{1/\gamma} \right] (P_{CJ}/\rho_{CJ}) \quad (3.12)$$

This gives the locus of all possible P, u points above P_{CJ}, u_{CJ} for the explosive (Fig. 3.4).

In the second case, $P_2 < P_{CJ}$ and the reflected shock undergoes rarefaction until P_2 is reached. Thus the Riemann equation applies, and hence

$$u_2 - u_1 = \int_{P_1}^{P_2} \frac{dP}{\rho c} \quad (3.13)$$

Eliminating ρ between equations (3.11) and (3.13) and substituting the relation derived by Taylor[16] for the C.-J. pressure

$$P_{CJ} = \frac{D^2 \gamma \rho_{CJ}}{(\gamma+1)^2} \quad (3.14)$$

gives the equation of the reflection adiabat

$$u - u_{CJ} = 2\gamma D \left[1 - (P/P_{CJ})^{(\gamma-1)/2\gamma} \right] / (\gamma^2 - 1) \quad (3.15)$$

where D is the detonation velocity.

This gives the locus of all possible P, u points below P_{CJ}, u_{CJ} for the explosive (Fig.3.4).

It can be seen from Fig. 3.4 that the pressure induced by an explosive in a solid is given by the point of intersection of the solid Hugoniot and the reflected Hugoniot or the reflection adiabat. Only at this point are the pressure and particle velocity continuous across the explosive-solid interface.

3.1.4 Spall Fracture

The pressure profile existing in a medium as a result of the passage of a shock is a discontinuous compressive front, followed by a gradual rarefaction to the free surface pressure (Fig. 3.5 a). When this wave strikes a free surface, it is reflected as a tensile wave, which propagates back into the compressed medium. The net stress at any point is given by the sum of the compressive and tensile stresses. If the compressive stress falls off rapidly after the shock, the net force becomes tensile and may exceed the rupture strength of the material causing spalling [19] (Fig. 3.5 b,c,d).

Multiple spalling occurs when the net tensile stress is more than twice the rupture stress. This is thought to occur as follows. When the first spall occurs the remainder of the compressive wave reflects off the newly created free surface and, if the net tensile stress is greater than the rupture strength, a second spall occurs. The process is repeated until the net tensile stress is less than the rupture strength [20]. The process appears to occur in three stages. When the tensile stress reaches a critical value, a number of individual voids or cracks form. These propagate until their projected area in the plane of the failure is equal to the area of the specimen (ignoring edge-effects). The individual fractures do not begin on the same atomic plane, and therefore do not join together. At this instant, the material between the

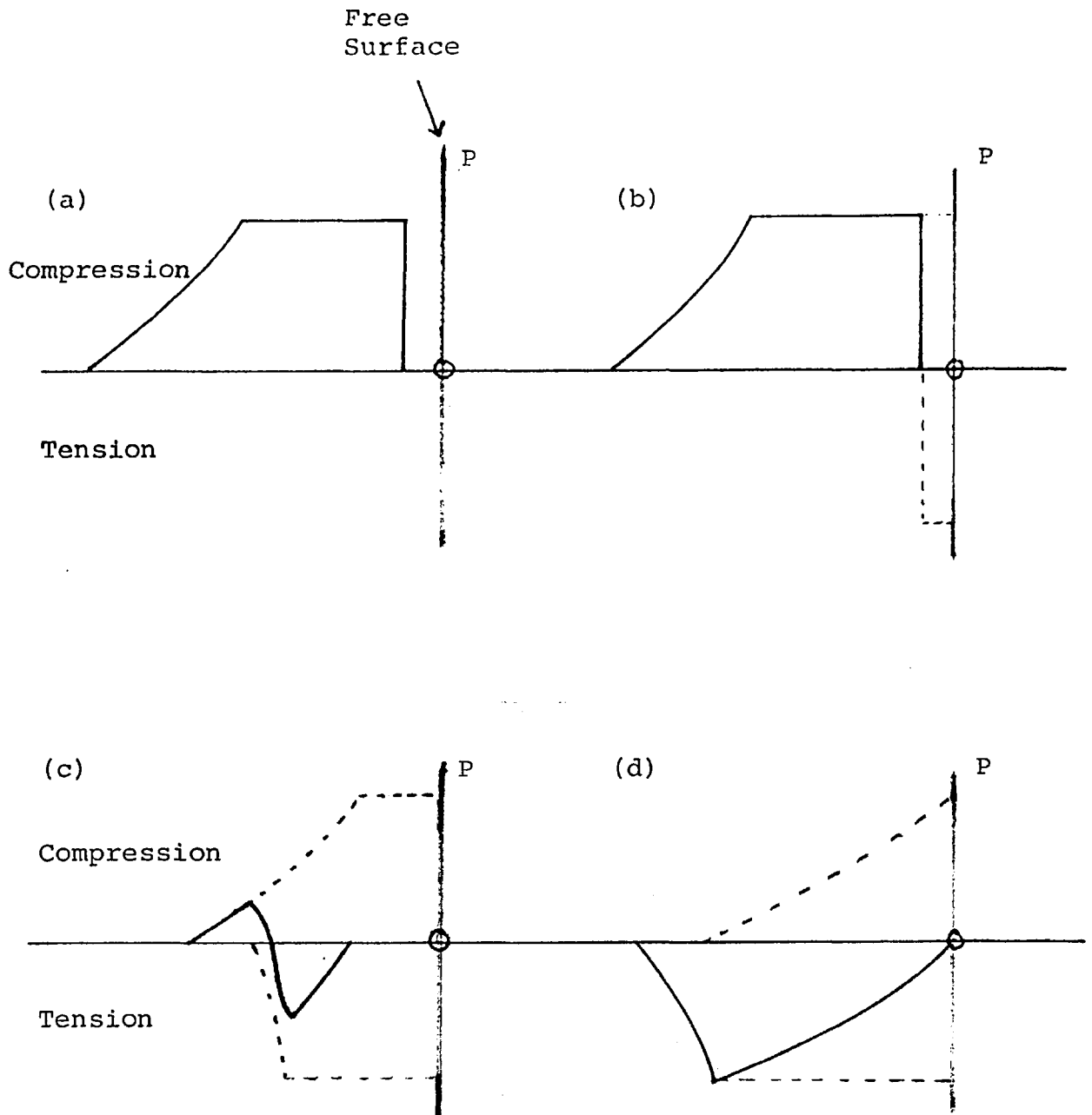


Fig. 3.5 The generation of a spall fracture

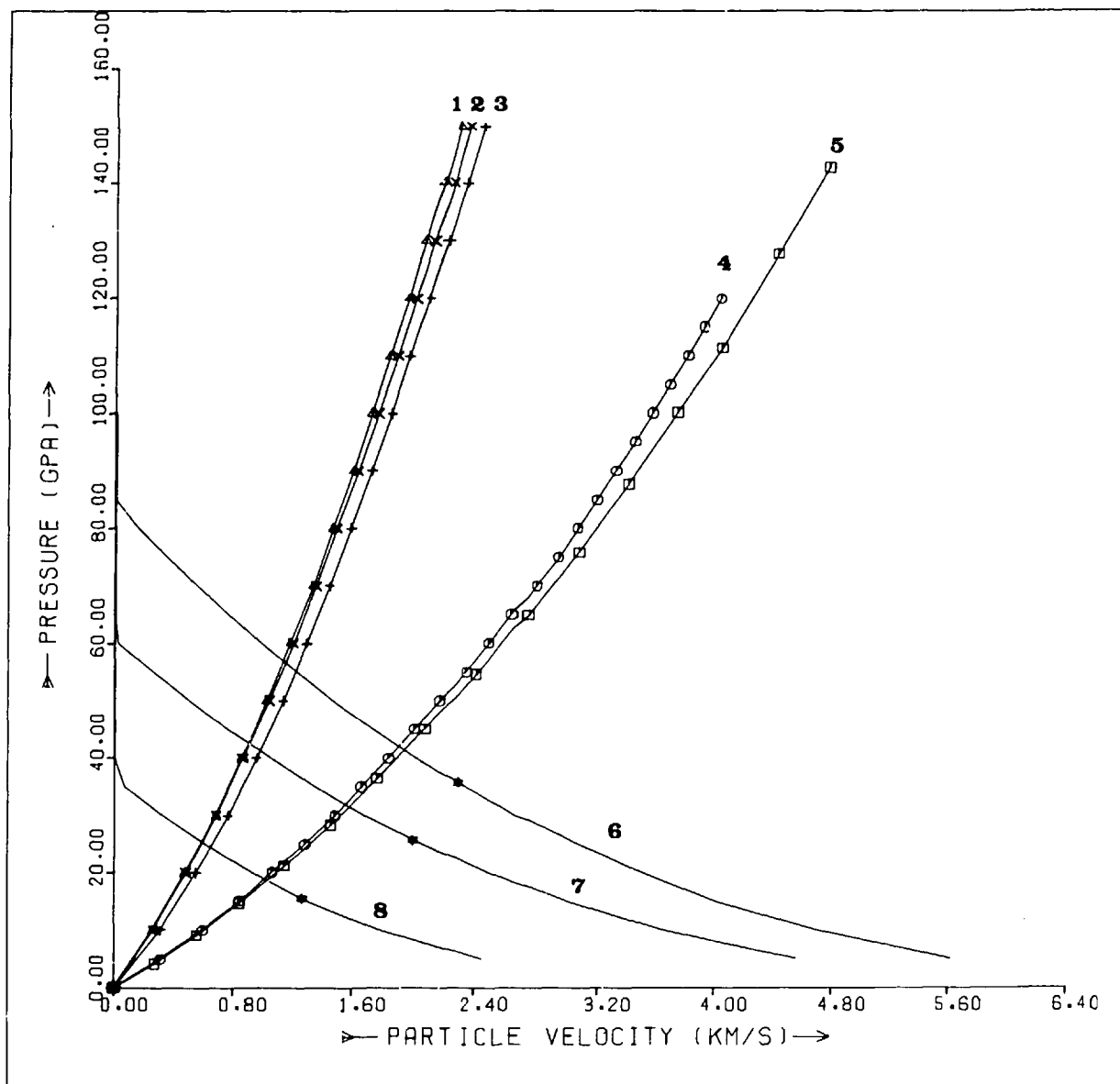
original free surface and this fracture area has trapped the most intense part of the pressure pulse, and therefore has acquired a higher forward velocity than the material behind. If this difference in velocity is large enough and applied for sufficient time, the final stage of spalling will be accomplished by deformation in the regions adjoining the individual fractures [21].

3.1.5 The Explosive Charge

Before determining the dimensions of the explosive charge it is necessary to know the C.-J. characteristics of the explosives to be used. These values are listed in Table 3.1 for baratol, cast composition B (60% RDX, 40% TNT) and a 95% HMX compound. Inserting these values in the equations of the reflected Hugoniot (3.12) and the reflection adiabat (3.15) produces the curves plotted in Fig. 3.6. The intersections of these curves and the Hugoniot equation of state curves for the metals and alloys under investigation give the specimen pressures and particle velocities listed in Tables 3.2 and 3.3. In this case, the double shock wave existing in pure iron when loaded by baratol is ignored.

The explosive charge consists of two sections:

- (a) The explosive block
- (b) The plane wave generator



- | | | | | | |
|---|-----------|---|----------------------|---|------|
| 1 | Copper | 2 | 18-8 stainless steel | 3 | Iron |
| 4 | Duralumin | 5 | Aluminium | 6 | HMX |
| 7 | Comp. B | 8 | Baratol | | |

* C.-J. pressures

Fig. 3.6 Graph of pressure v. particle velocity for the metals, alloys and explosives used in this work

(a) The Explosive Block

This is a cylinder of explosive having a diameter Wt and a thickness s . The velocities of the detonation front and the leading rarefaction wave are D and C_E . The explosive products are assumed to behave as ideal gases. Thus $P\rho^{-\gamma}$ is constant and substituting this into the equation (A12) for the sonic velocity gives:

$$C_E = \left[\frac{P_{CJ} \gamma_{CJ}}{\rho_{CJ}} \right]^{\frac{1}{2}} \quad (3.16)$$

The values of D and C_E are given in Table 3.1.

It can be seen that D is greater than C_E . Thus the duration of the pressure pulse depends on the thickness and diameter of the explosive block.

Fig. 3.7 shows the positions of the detonation wave and the leading rarefaction waves at time t after detonation. The detonation front is followed by the leading rarefaction wave originating at the back face. The region of constant pressure between these waves is reduced by rarefaction waves originating at the circumference of the charge. The locus of these wave-fronts is a straight line at an angle θ to the side of the charge. θ is given by:-

$$\sin \theta = \frac{C_E t}{Dt} = C_E / D \quad (3.17)$$

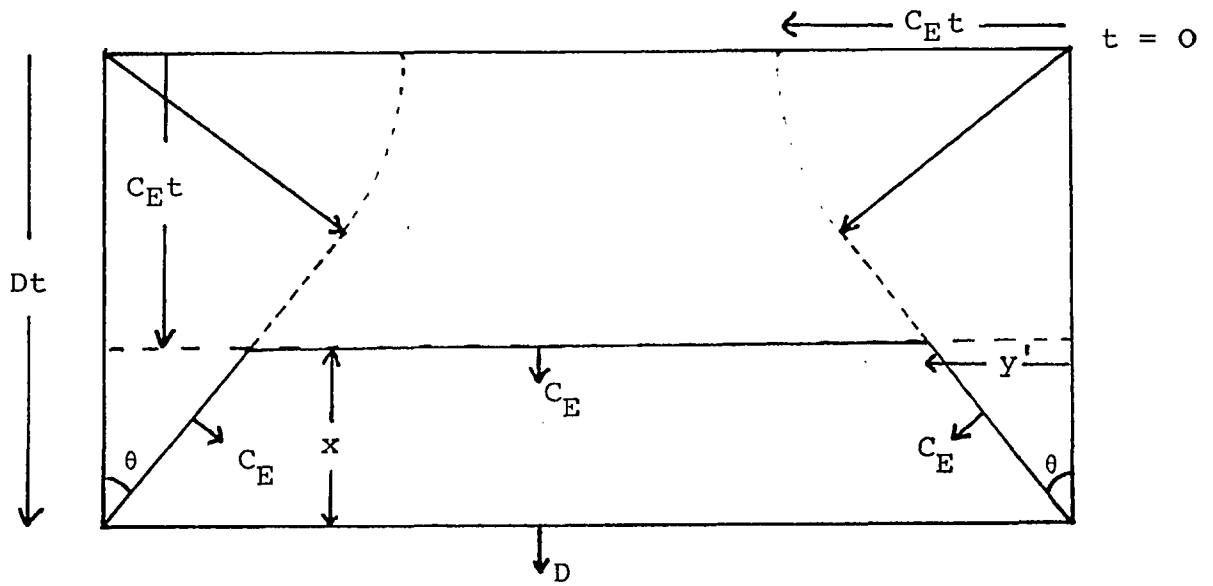


Fig. 3.7 Cross-section of the pressure profile in the explosive block

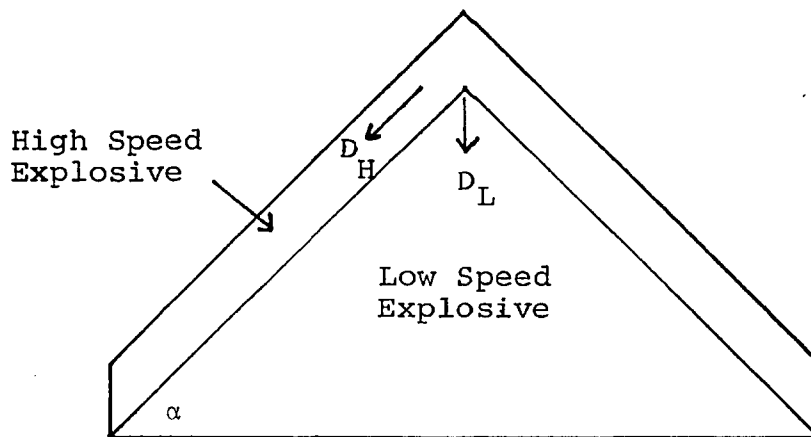


Fig. 3.8 Plane wave generator

Let y' be the reduction in the radius of the back rarefaction wave. Therefore,

$$\tan \theta = y' / (Dt - C_E t) \quad (3.18)$$

$$\therefore y' = t(D - C_E) \tan \theta = t(D - C_E) \frac{C_E}{D} \left[1 - \left(\frac{C_E}{D} \right)^2 \right]^{\frac{1}{2}} \quad (3.19)$$

It is important to determine what area of the explosive - metal interface will undergo rarefaction due to the back rarefaction wave alone. This can be determined by calculating the value of y' when the back rarefaction wave reaches the explosive-metal interface. At this point,

$$y = y' \frac{D}{C_E} = tD \left[\frac{D - C_E}{D + C_E} \right]^{\frac{1}{2}} \quad (3.20)$$

The distance between the detonation front and the back rarefaction wave is given by

$$x = (D - C_E)t \quad (3.21)$$

If t is taken as the time for the detonation front to reach the explosive-metal interface, equations (3.20) and (3.21) determine the pressure profile at this interface.

Now, $s = tD$ if t is measured in micro-seconds. Therefore, equations (3.20) and (3.21) can be rewritten as

$$y_s = \frac{y}{s} = \left[\frac{D-C_E}{D+C_E} \right]^{\frac{1}{2}} \quad (3.22)$$

$$\text{and } x' = \frac{xD}{s} = D-C_E \quad (3.23)$$

Values of y_s and x' are given in Table 3.4.

It can be seen that a suitable pressure profile is only obtained if the charge diameter (W_t) is greater than $2y_s s$.

A further limit is imposed by the effect of the charge radius on the detonation velocity [16,17,23]. The detonation velocity D differs from the ideal detonation velocity D_0 due to an infinitely wide charge as given in the following equation,

$$\left[\frac{D_0}{D} \right]^2 \approx 1 + 0.8 \frac{X^2}{R^2} \quad (3.24)$$

where X is the reaction zone length and R is the charge radius.

$$\text{Let } D = D_0 - \Delta D \quad (3.25)$$

then equation (3.24) reduces to

$$\frac{\Delta D}{D_0} = 0.4 \frac{X^2}{R^2} \quad (3.26)$$

Typical reaction zone lengths are given in Table 3.5.

If $\frac{\Delta D}{D_0}$ is to be less than 0.1%, $\frac{X}{R}$ should be less than 0.05.

This gives minimum values for R (Table 3.5). Values of X for baratol and 95% HMX are not known.

(b) The Plane Wave Generator [24]

The explosive block must be detonated by a plane wave. This can be achieved by capping the cylindrical explosive by a conically shaped charge composed of two explosives (Fig. 3.8). The high speed-explosive must be sufficiently thin to impart a sharp detonation pulse to the edge of the slow-speed explosive, but it must also be sufficiently thick to produce a detonation velocity close to D_0 . The pulse width should be determined as far as possible by the side rarefaction, and not be the back rarefaction. The angle α of the cone is given by

$$\sin \alpha = \frac{D_L}{D_H} \quad (3.27)$$

where D_L and D_H are the detonation velocities of the low-speed and high-speed explosives respectively.

The high-speed explosive has the effect of removing the retarding edge effects from the low-speed explosive.

In this way good planarity is assured. A plane wave generator having a baratol core and a composition B outer skin was used for all the explosive tests.

3.1.6 The Metal Specimen

The pressures developed in the metals and alloys are given by the intersections of the Hugoniot curves and the reflected Hugoniot curves of the explosives (Fig. 3.6). The Hugoniots for duralumin, pure copper, 18-8 stainless steel and pure iron are plotted from the data listed by Kinslow [25]. The aluminium Hugoniot is taken from the results of Goranson et al. [14]. The specimen pressures are given in Table 3.2.

The shock wave velocities U_s and the leading (P) and trailing (P = 0) rarefaction wave velocities (C_L, C_O) are calculated using equations (3.1) and (3.2), and equation (A12). These values are listed in Tables 3.6 and 3.7.

The pressure at the explosive-metal interface must be kept constant until the trailing free surface rarefaction wave reaches this interface. If this is to be achieved, the pressure pulse must last for at least a time t_s given by:

$$t_s = d \left[\frac{1}{U_s} + \frac{1}{C_O} \right] \quad (3.28)$$

where d is the specimen thickness.

Moreover the dimension W of the spall plate (Fig. 3.9) must be such that the leading rarefaction wave from the spall plate circumference does not reach the specimen disc within the time t_s . This requires that:

$$\frac{W}{d} \geq 2 \left[\frac{C_L \rho}{U_S \rho_O} + \frac{C_L \rho}{C_O \rho_O} \right] \quad (3.29)$$

where ρ_O and ρ are the metal densities before and after the passage of the shock front. Tables 3.8 and 3.9 give the minimum values of W/d and the values of t_s/d .

If the explosive is to produce a pulse of a time greater than t_s , then:

$$s \geq d \left[\frac{1}{U_S} + \frac{1}{C_O} \right] / \left[\frac{1}{C_E} - \frac{1}{D} \right] \quad (3.30)$$

This is derived from equations (3.21), (3.23) and (3.28). The minimum values of s/d are given in Table 3.10.

The diameter of the plane wave generators used was 76.2 mm (3 ins.). This determined the remaining explosive-metal assembly dimensions.

Now if the total plate diameter is W_t the diameter W_s of the plate which will be affected solely by free surface rarefaction is given by:

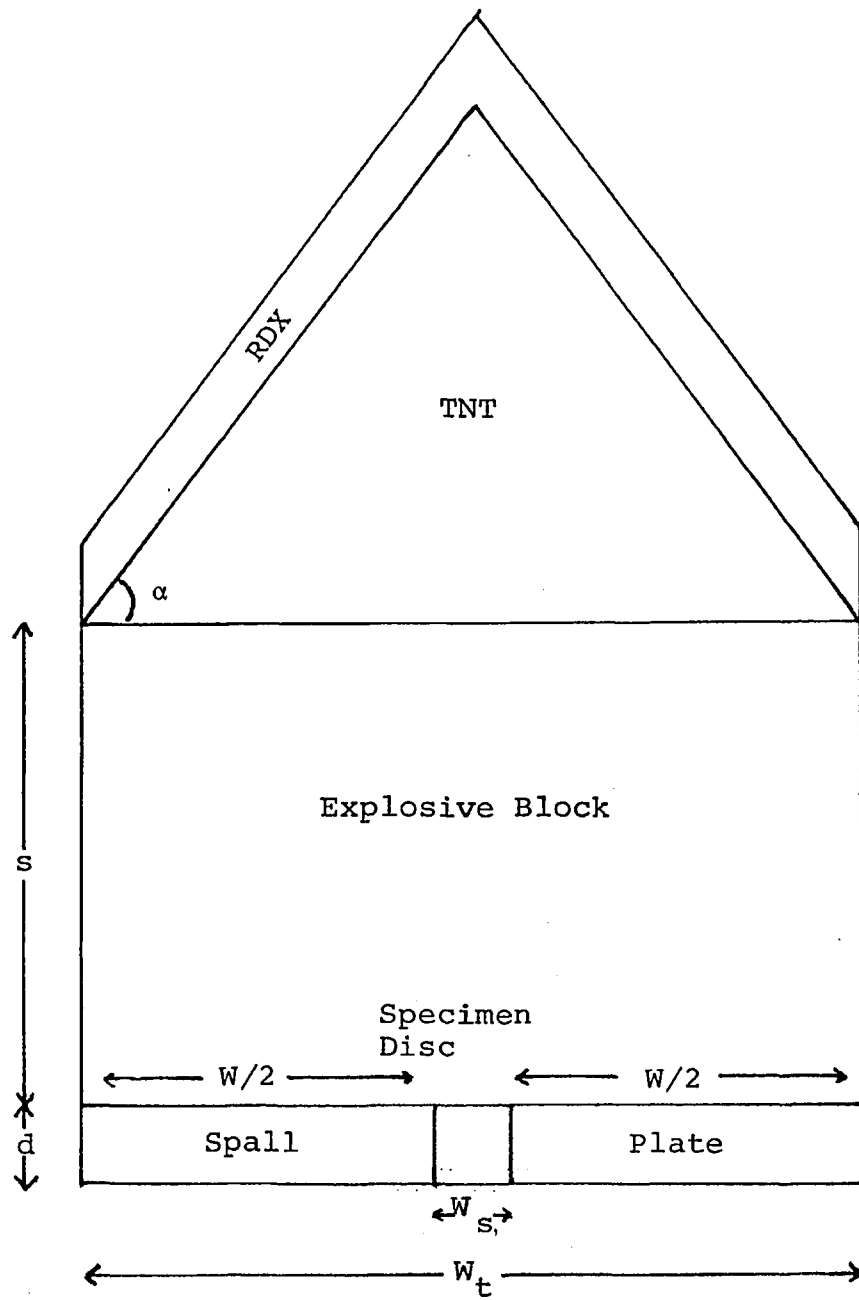


Fig. 3.9 Explosive-metal assembly

$$W_{s \max} = W_t - 2y_{s \max} s_{\min} - W_{\min} \quad (3.31)$$

Taking $y_{s \max} = 0.4$, $W_{\min} = 8d$ and $s_{\min} = 12d$ (see

Tables 3.4, 3.8 and 3.10) gives:

$$W_{s \max} = W_t - 17.6d \quad (3.32)$$

If d is 3.175 mm (1/8 in.), $W_{s \max} = 20.3$ mm and $s_{\min} = 38.1$ mm.

Circular 76.2 mm diameter plates 3.175 mm thick were cut containing a central, tight-fitting 12.7 mm (0.5 in.) diameter specimen disc of the same thickness. These were stuck onto 50.8 mm (2 in.) thick explosive blocks using grease and the blocks were capped by the plane wave generators (Fig. 3.9).

3.1.7 The Measurement of the Specimen Free Surface Velocity

The free surface of an explosively loaded metal accelerates to a high velocity u_{fs} ($>10^3$ m/s) extremely rapidly (ca. 10^{-9} s). Thus special techniques using laser interferometry are required.

A system was designed based on an idea by Laharrague, Durand and Le Bihan [26,27] using a Fabry-Perot

interferometer. Due to the sudden availability of a superior, existing system the original design was not used. It is described, however, in Appendix B. The interferometer actually used is detailed below.

3.1.7.1 The Wide Range Velocity Interferometer (WRVI)

This was designed by Amery [28] and based on the "velocity interferometer system for any reflector (VISAR)" described by Barker and Hollenbach [29]. The apparatus makes use of the Doppler effect whereby light reflected off a moving surface changes its wavelength by $\Delta\lambda$ where

$$\frac{\Delta\lambda}{\lambda} = - \frac{2V}{c} \quad (3.33)$$

c = velocity of light, V = surface velocity

A diagram of the apparatus is shown in Fig. 3.10 and is referred to in the following description.

A single frequency argon ion laser beam passes through a beam expander containing a 50Hz chopper disc. This gives a defined pulse duration every revolution and reduces the mean light level at the photomultipliers. The beam travels via the mirror M3, the beam splitter B3, the telescope and an evacuated tube to the metal-explosive assembly site. Here it is focussed (L8) onto the specimen free surface (S).

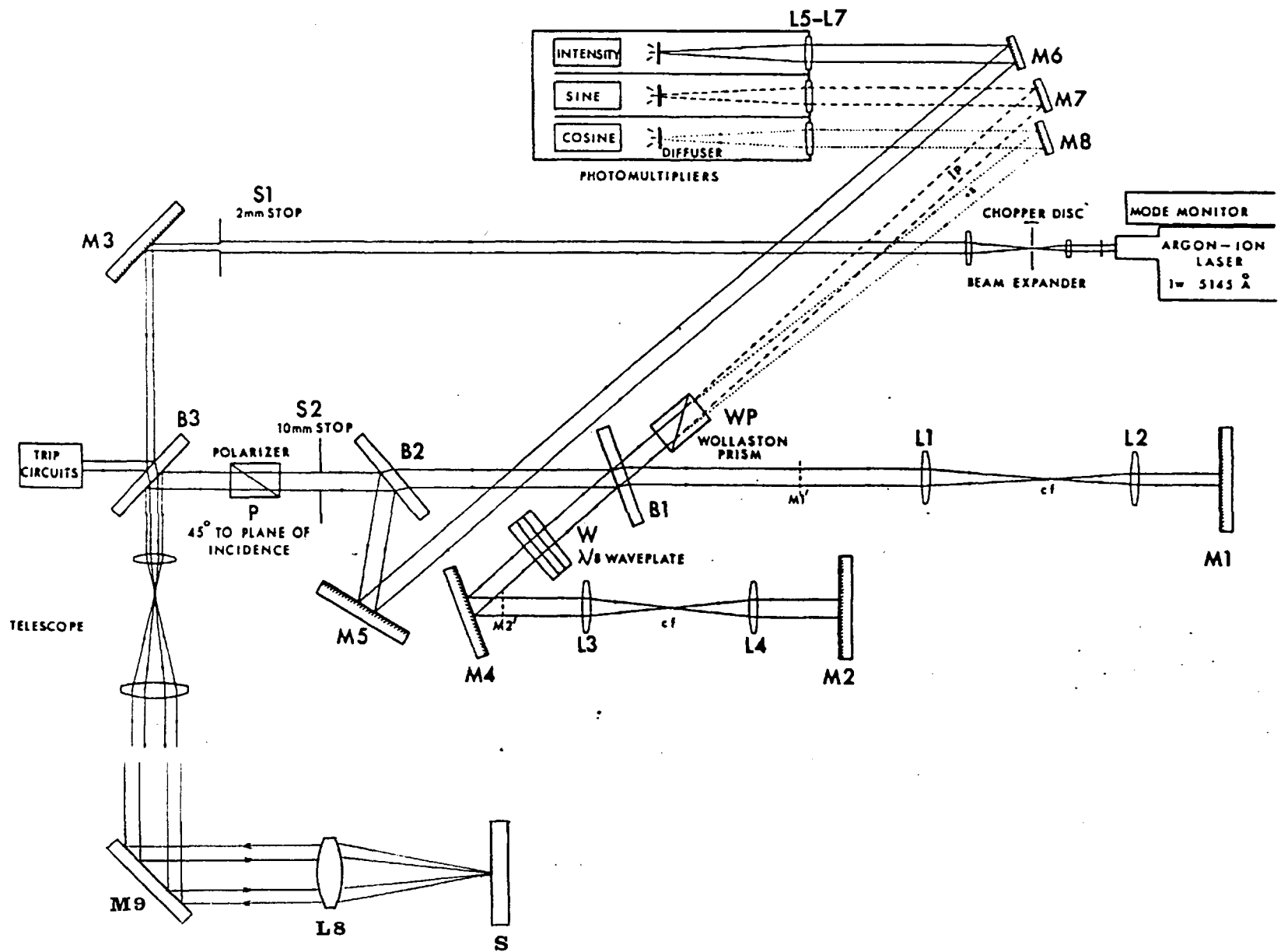


Fig. 3.10 Diagram of the wide range velocity interferometer used to measure the specimen free surface velocities (Amery [28])

The reflected beam is recollimated by the telescope, reflected off the beam splitter B3 and polarised to produce a beam with equal S (unpolarised) and P (polarised) components. A portion of the beam is removed by the beam splitter B2 and its intensity recorded by a photomultiplier. The remaining light is divided by the beam splitter B1 which together with the mirrors M1 and M2 forms a Michelson interferometer of unequal leg lengths.

A $\lambda/8$ wave plate retards the P component by a total phase angle of $\pi/2$. The recombined beams pass through the beam splitter B1 and the S and P components are separated by a Wollaston prism to be collected in the SINE and COSINE photomultipliers.

If monochromatic light of wavelength λ_0 reflects off a surface moving at velocity u_{fs} towards it, the wavelength of the reflected light λ is given by

$$\lambda = \lambda_0 (1 - 2u_{fs}/c) \quad (3.34)$$

Now if the path difference between the two legs of the interferometer is L the reflected beams interfere constructively if $L = N_0 \lambda_0$ where N_0 is an integer.

If the wavelength changes from λ_0 to λ so that there are now N wavelengths in L , the beams will again interfere constructively. Thus

$$N - N_O = L \left[\frac{1}{\lambda} - \frac{1}{\lambda_o} \right] \quad (3.35)$$

Using equation (3.35) in (3.34) gives

$$dN = \frac{L}{\lambda_o} \left[\frac{1}{1 - 2u_{fs}/c} - 1 \right]$$

$$\approx \frac{L}{\lambda_o} \frac{2u_{fs}}{c} \quad u_{fs} \ll c$$

$$\text{or} \quad u_{fs} \approx \frac{\lambda_o c}{L} dN \quad (3.36)$$

The photomultipliers detect one fringe change for a unit change in the number of wavelengths in the path difference L . The voltage outputs of the SINE and COSINE fringing photomultipliers is given by:-

$$V_S = V_O (1 + \sin (2\pi dN + \phi)) \quad (3.37)$$

and

$$V_C = V_O (1 + \cos (2\pi dN + \phi)) \quad (3.38)$$

Subtracting V_O and dividing gives

$$\frac{V_S}{V_C} = \tan (2\pi dN + \phi) \quad (3.39)$$

where ϕ is determined by V_S/V_C when $u_{fs} = 0$.

Thus the ratio of the amplitude of the SINE and COSINE photomultipliers gives

$$\frac{V_S}{V_C} = \tan \left[\frac{4\pi L u_{fS}}{\lambda_0 c} + \phi \right] \quad (3.40)$$

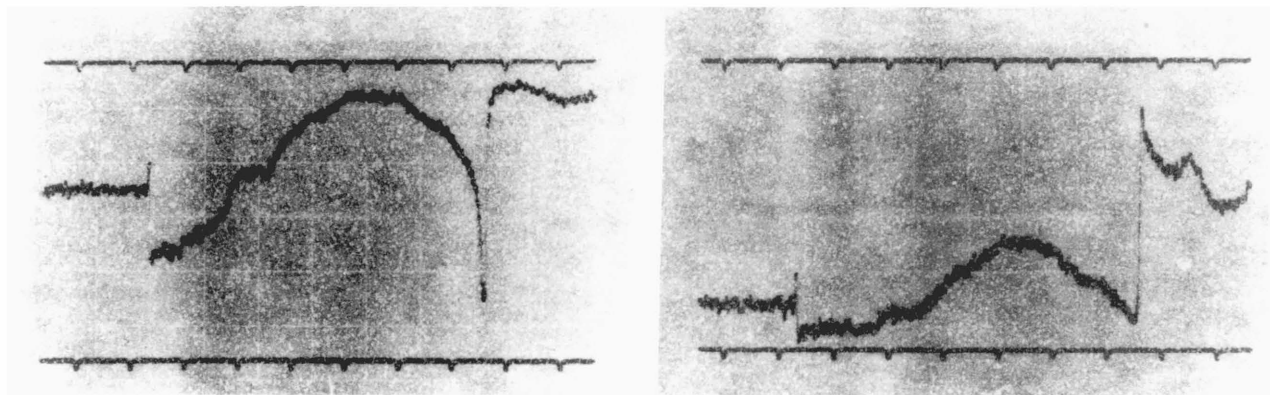
Acceleration and deceleration can be distinguished by noting whether the cosine trace leads (acceleration) or lags (deceleration) the sine trace. Typical sine and cosine traces together with timing markers are shown in Fig. 3.11.

The lenses L1 - L4 in the interferometer legs produce images M1' and M2' of M1 and M2 equidistant from the beam splitter B1. This ensures that high quality fringes are produced even if the light illuminating the interferometer is incoherent (e.g. if the specimen surface loses its mirror finish due to a phase change).

The measured values of u_{fS} for six shots are given in Table 3.11 and the behaviour of u_{fS} with time is shown in Figs. 3.12 a - f.

3.1.8 The Measurement of the Shock Wave Velocity U_S

A foil probe was inserted between the explosive block and the sample (Fig. 3.9). This was used to trigger the interferometer recording oscilloscopes. The shock wave velocity was calculated from the interval between



Sine

Cosine

Fig. 3.11 Sine and cosine traces from a typical WRVI test
(timing markers 0.2 μ s)

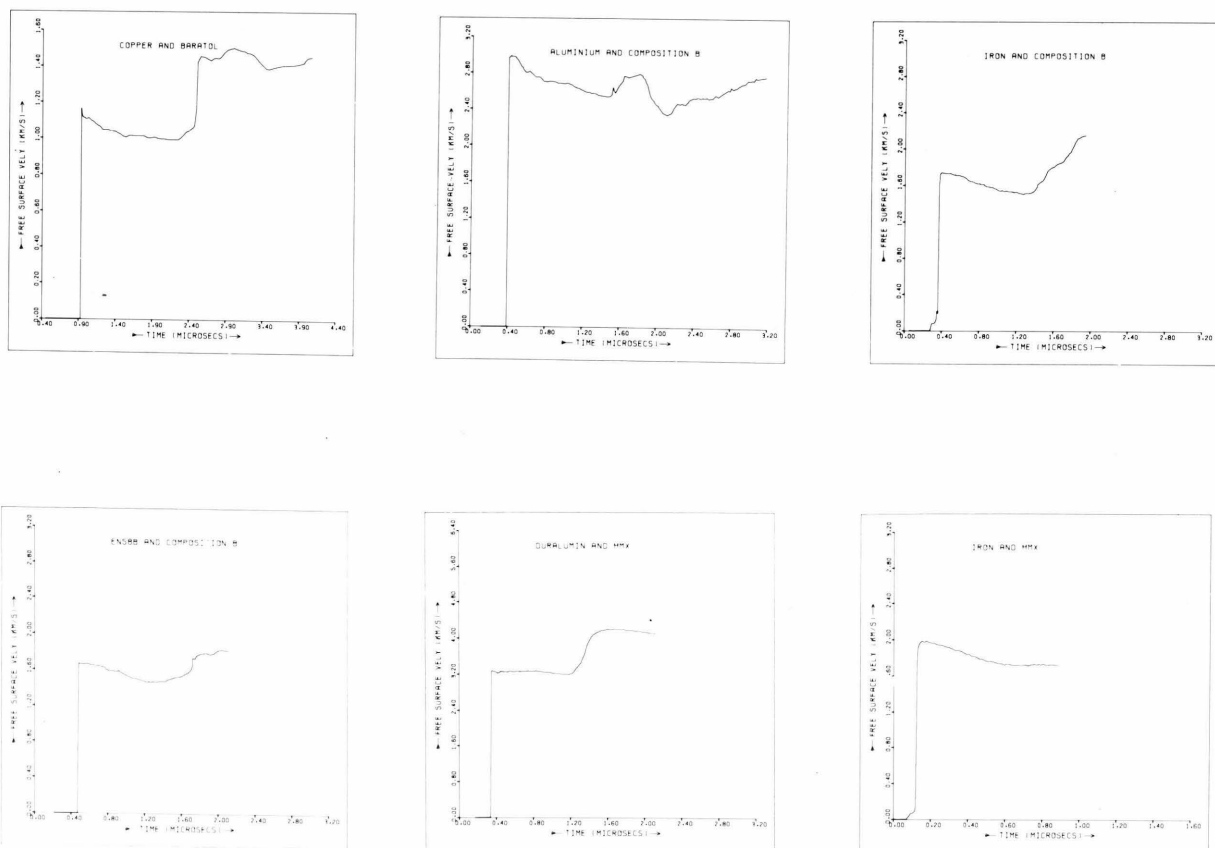


Fig. 3.12 Plots of u_{fs} v. time for the six shots given in
Table 3.11

the trigger pulse and the onset of motion of the specimen free surface after compensating for the various system delays. These values are given in Table 3.11.

3.2 The Specimen Recovery Equipment

The specimens to be used for microscopical examination were recovered from shots fired without the interferometer system.

The explosively loaded disc has an initial velocity of up to $4 \times 10^3 \text{ m/s}$ which must be reduced to zero in such a way as to keep the disc deformation to a minimum during deceleration. Fig. 3.13 shows a plan of the recovery assembly.

The explosive-metal assembly was placed on a wooden support table which stood on a thick steel plate containing a central hole for the passage of the specimen. A stack of rock fibre boards placed over a tank of water decelerated the specimen and the fragments were collected by a net in the water tank. The fibre boards were also sifted for metal debris. Fig. 3.14 shows the recovered fragments from six shots. The visible surface of the central disc is the free surface which shows deformation due to its collision with the rock fibre. The remaining fragments came from the spall plate. The condition of the recovered discs is given in Table 3.12.

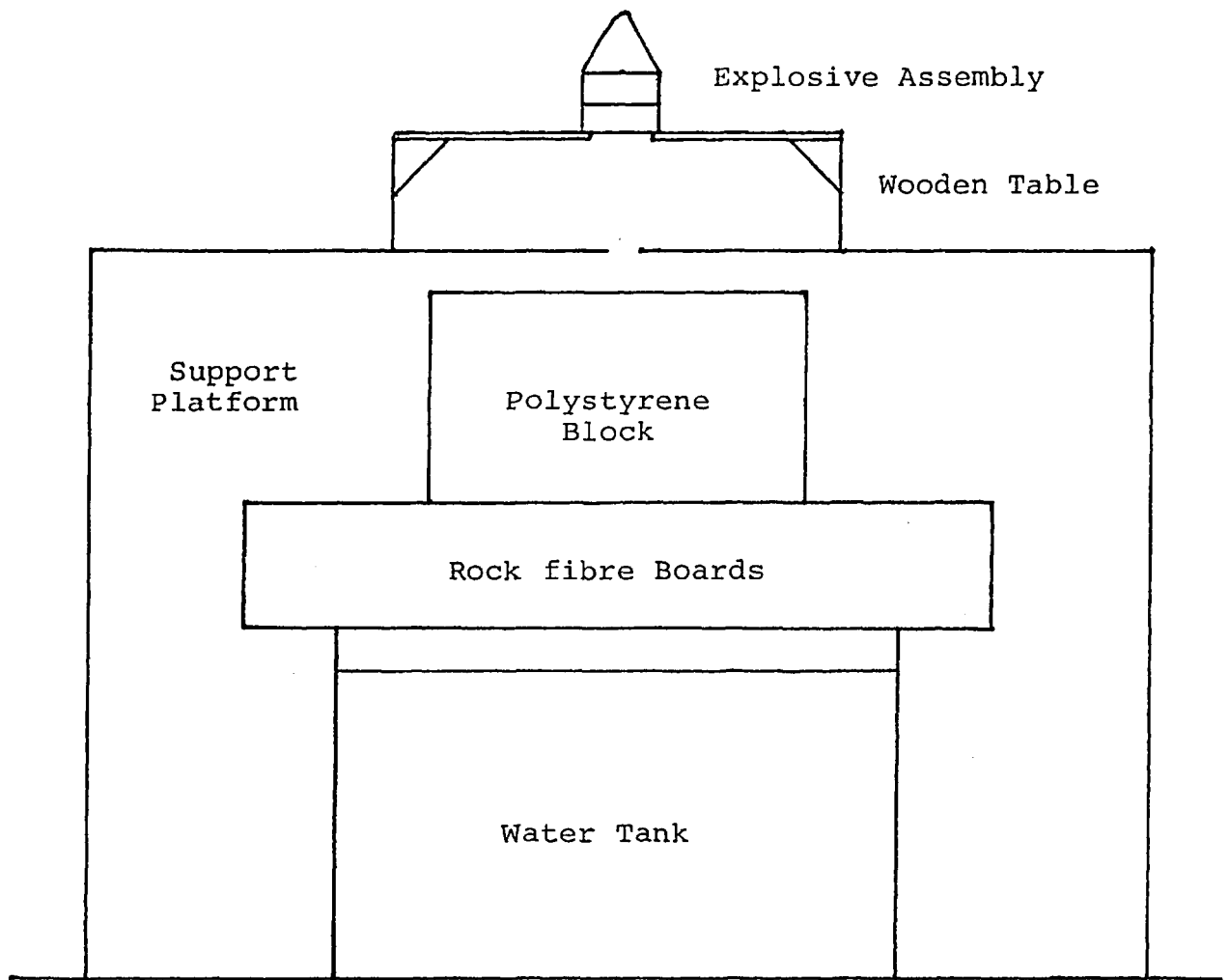


Fig. 3.13 Specimen recovery assembly



Fig. 3.14 a Fragments of baratol loaded dural

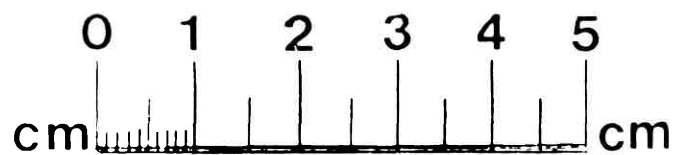


Fig. 3.14 b Fragments of baratol loaded copper



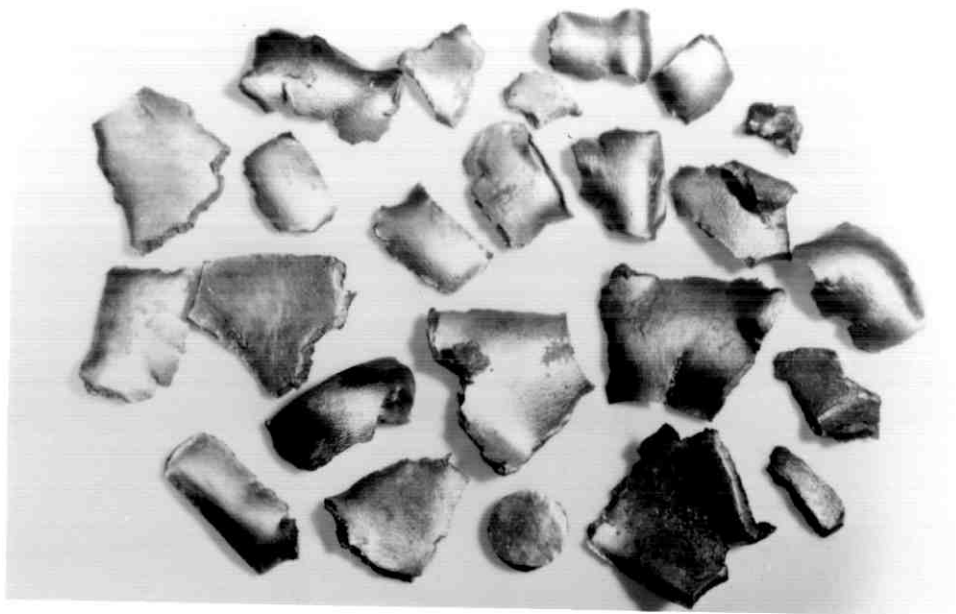


Fig. 3.14 c Fragments of baratol loaded iron

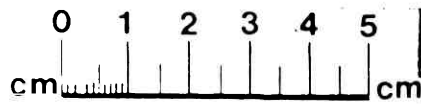


Fig. 3.14 d Fragments of HMX loaded 18/8 steel





Fig. 3.14 e Fragments of HMX loaded iron
 (disc indicated)

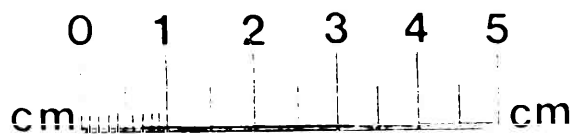
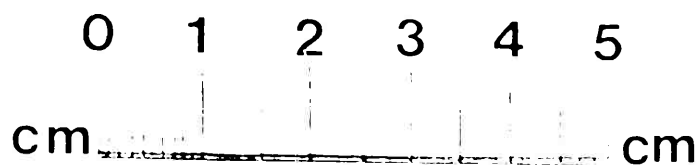


Fig. 3.14 f Fragments of HMX loaded copper
 (disc destroyed)



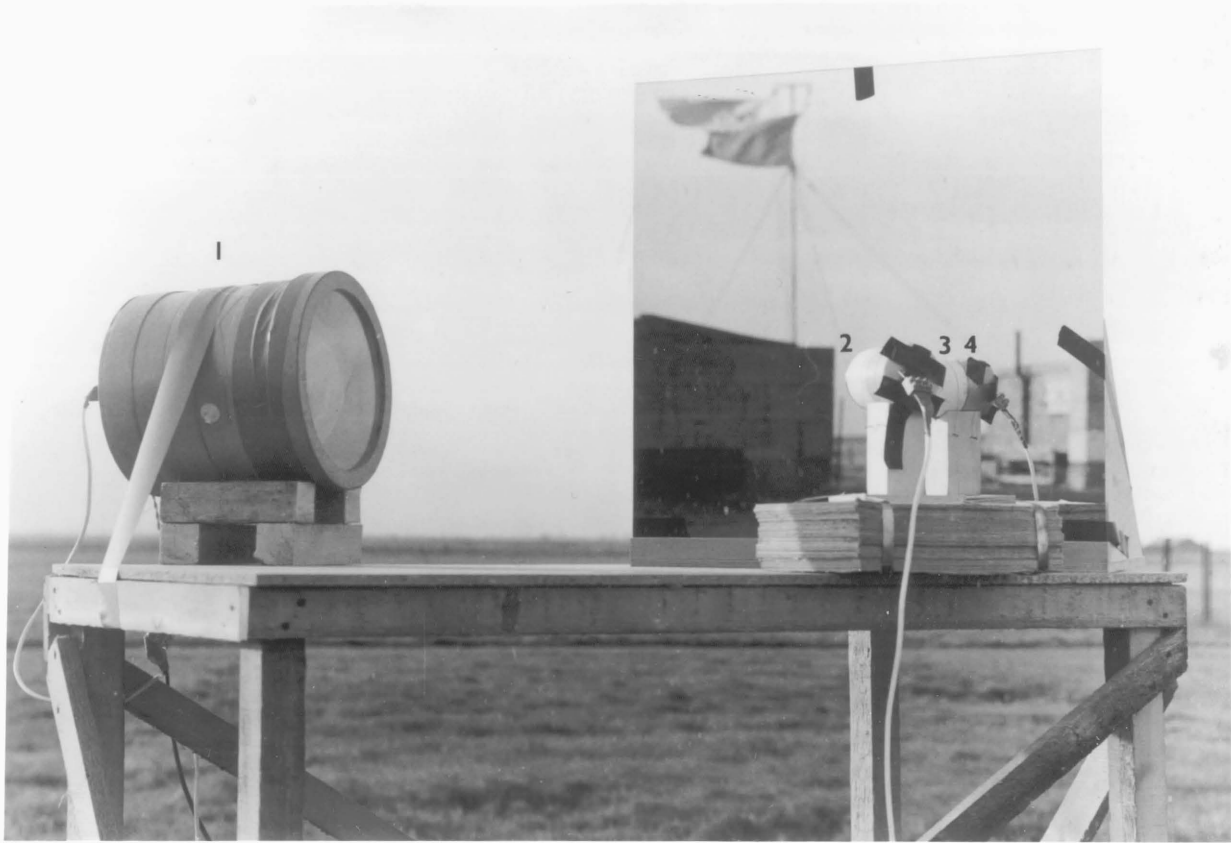
3.3 High Speed Photography of the Explosive Event

A series of high speed photographs was taken to determine whether the disc broke up due to spalling. The arrangement shown in Fig. 3.15 was used in conjunction with a high speed camera remote from the site. The explosive-metal assembly on the right is inclined at an angle to the mirror in order to produce a profile and an end-on view of the explosive event. The flash tube on the left provides illumination.

Shots using aluminium with HMX and 18-8 stainless steel with HMX were photographed. The sequences are shown in Figs. 3.16 a,b,c and d.

Frame 1 of Fig. 3.16 a shows the onset of detonation and plate motion has just begun in Frame 2. It can be seen from Frames 11 and 12 that the central aluminium disc appears to be whole up to 47 μ s after it starts to move. Since the transit time for the shock wave and the subsequent slowest rarefaction wave across the disc is about 1 μ s, many reflections would have occurred. Spall fragmentation would have taken place if this fracture mode was present.

However, in recovery shots this disc did not survive under the lowest explosive loading (baratol). This can be attributed primarily to the low strength of aluminium (VH = 25) which causes the disc to break up during deceleration.



- 1 Flash tube
- 2 Specimen disc
- 3 Explosive block
- 4 Plane wave generator

Fig. 3.15 High-speed photography set-up

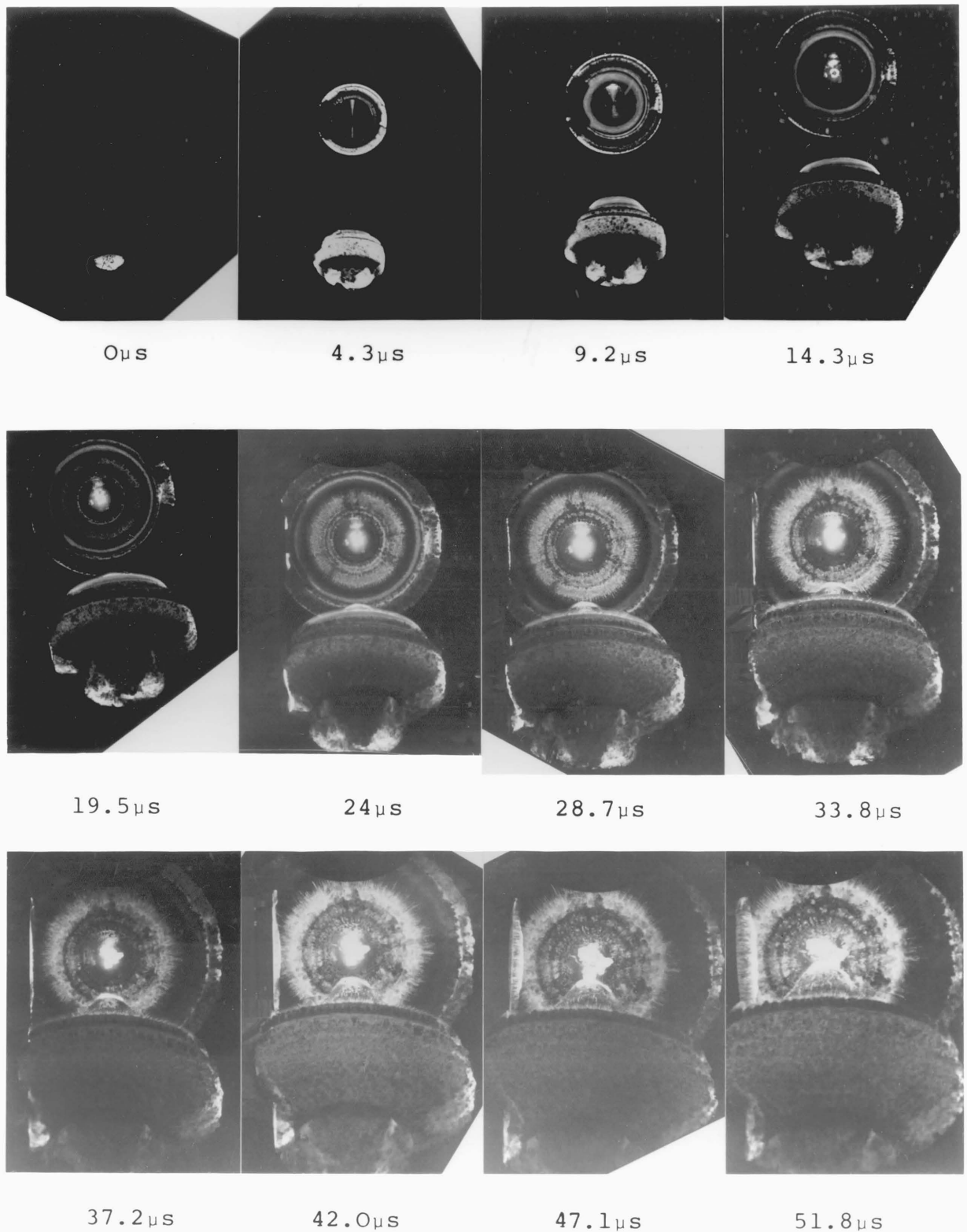


Fig. 3.16 a High-speed photographic record of HMX loaded aluminium (note central specimen disc is undamaged up to 47 μ s after it starts to move (see Fig. 3.16 c))

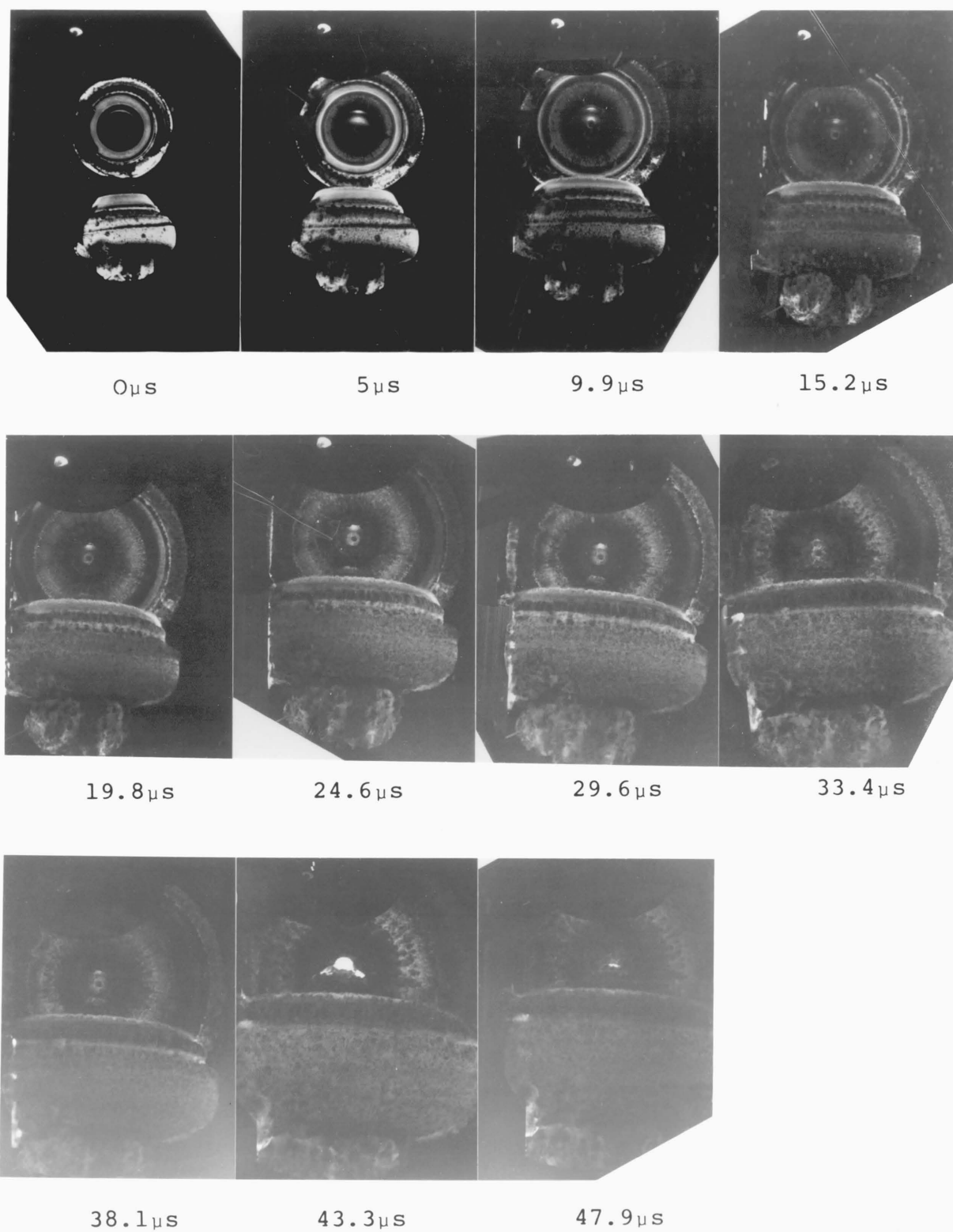


Fig. 3.16 b High-speed photographic record of HMX loaded 18-8 stainless steel (central specimen disc remains undamaged (see Fig. 3.16 d))

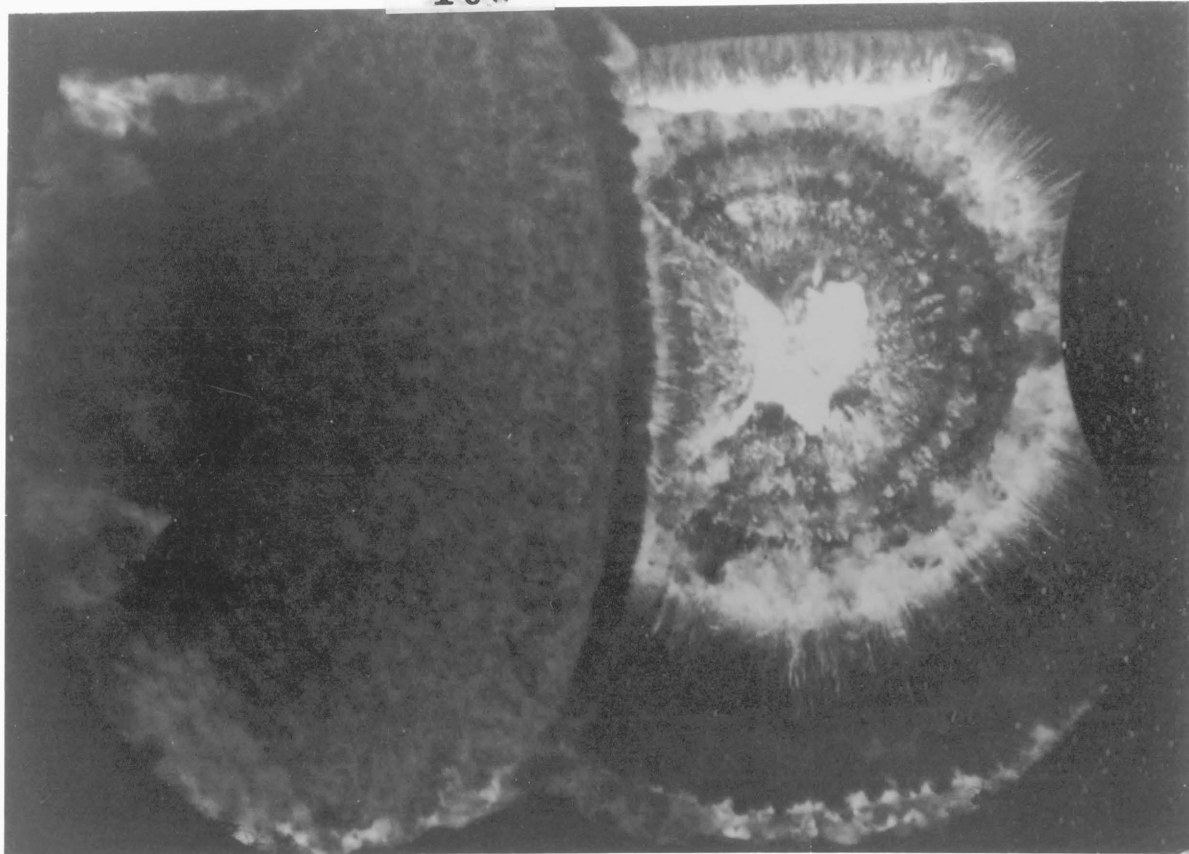


FIG.3.16c Undamaged central specimen disc of HMX loaded aluminium (photo 12, 51.8 μ s)

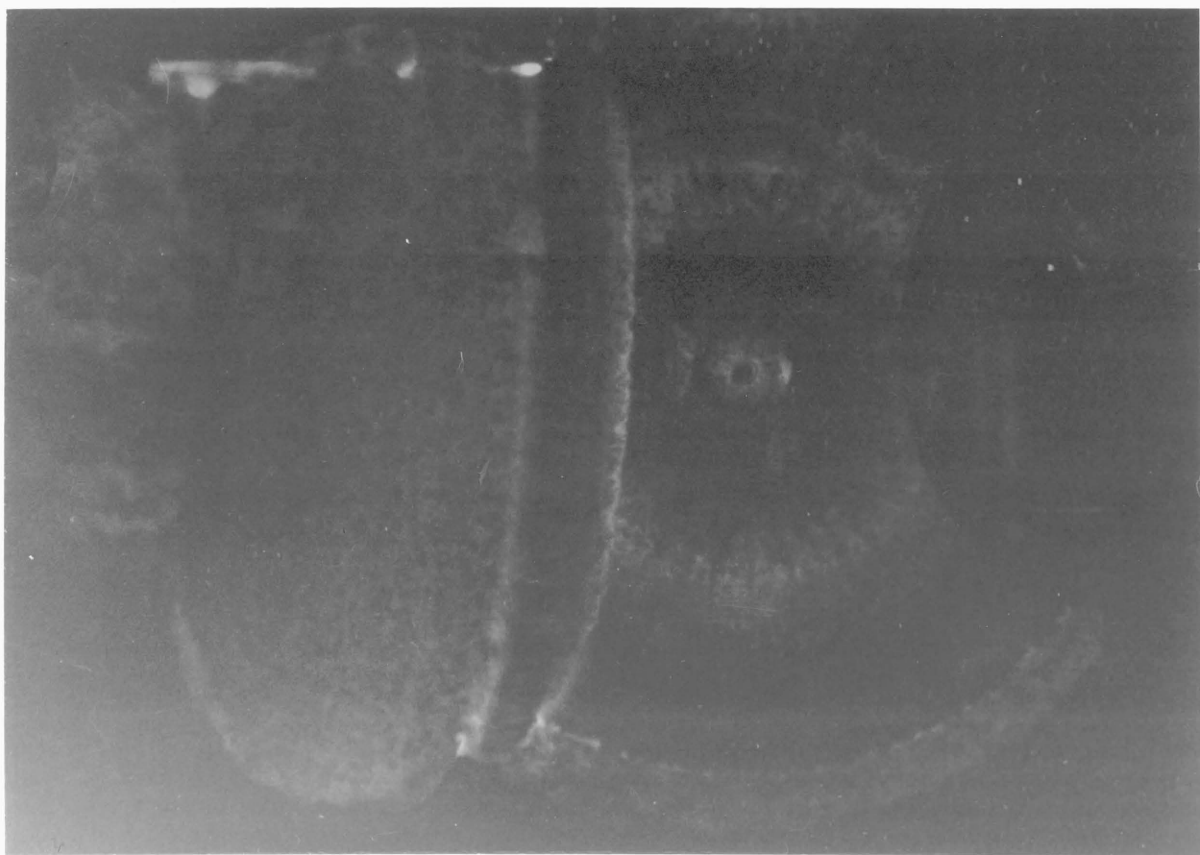


FIG.3.16d Undamaged central specimen disc of HMX loaded 18-8 steel (photo 9, 38.1 μ s)

Fig. 3.16 b shows a similar sequence for 18-8 stainless steel and HMX. In this case, however, an identical disc survived the recovery shot undergoing moderate deformation.

A rough estimate of the plate velocity can be made by plotting the separation of the plate surface profile from the middle of the central disc in the reflected image against the elapsed time. Knowing the plate width and the angle of the explosive-metal assembly to the mirror, the plate velocity can easily be calculated. Velocities of 2.35 and $3.52 \times 10^3 \text{ m/s}$ for 18-8 stainless steel and aluminium compare quite well with the calculated values (Table 3.3).

3.4 Comparison of Graphical and Measured Shock Parameters

The measured (Table 3.11) and the calculated (Table 3.6) shock velocities lie within 2% of each other. However, there is a considerably greater difference in the free surface velocities (up to 20%).

If one assumes that the measured values are accurate the results indicate that the shock pressures were lower than those calculated. Therefore the explosive curves on the (P,u) graph (Fig. 3.6) must lie closer to the origin, reflecting lower C.-J. pressures than expected.

It was not possible to check the C.-J. pressures of the explosives. In any case, these lower pressures would not affect the calculations since the specimen loading time depends on the shock velocity, which showed little difference between the measured and calculated values, and the fastest sonic velocity which would be reduced giving more time for the specimen shock sequence to occur.

3.5 The Hopkinson Bar Apparatus

A modified split Hopkinson pressure bar, designed and constructed by Yeung-Wye-Kong [30] at Leeds University, was used to load circular discs of the various metals and alloys at strain rates of 10^3 s^{-1} (Fig. 3.17).

The apparatus consisted of a gas gun which fired a striker bar, impacting and specimen bars which were supported by air bearings, and a bar-arresting anvil which contained a lead extrusion billet which dissipated the specimen bar energy.

A strain gauge bridge on the specimen bar recorded the transmitted pressure pulse. The impacting bar-end displacement was recorded by a photo-diode which detected the amount of light received from a light curtain produced by a fibre optic bundle.

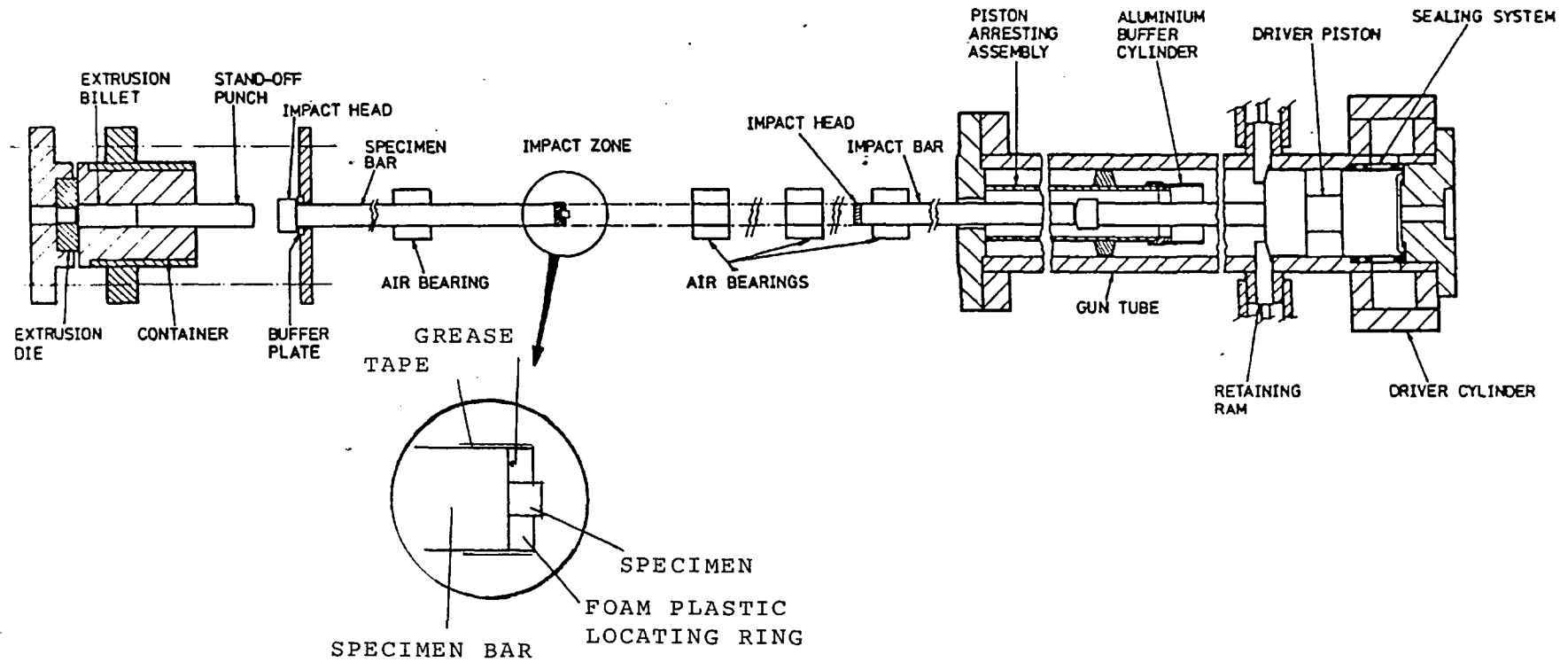


Fig. 3.17 Diagram of Hopkinson bar test facility

The impacting and specimen bars were of BS 970 EN 26 steel tempered to a minimum yield strength of 739 MPa. The bar density was 7823 kg/m^3 and the elastic modulus was 204.5 GPa. Each bar was 30 mm (1.375 ins) in diameter and 2.74 m (9 ft) long. This gave a loading pulse of 1.06 ms.

The specimen discs were fixed to the specimen bar using a little lubricating grease and a foam-plastic retaining collar. The discs were 3.175 mm (0.125 ins) thick and, with the exception of the pure aluminium discs (30 mm diameter), 9.53 mm (0.375 ins) in diameter. The smaller diameters were chosen to produce high permanent strains without excessively loading the arresting anvil. The specimen bar was pushed into the light chamber until the specimen just touched the light curtain. The gun was primed and the impacting bar fired.

The bar displacement signal and the amplified strain gauge signal were displayed on a cathode ray oscilloscope.

The acquisition of a 4K digital transient recorder permitted the design of a more sophisticated recording system. The data from each test was stored in the recorder and then transferred to a computer after the test. Here it was converted into graphical form and displayed on a visual display unit to ensure correct transfer. The data was stored on tape for further

processing at Imperial College.

3.5.1 Hopkinson Bar Theory [31,32].

The bar acts essentially as a mechanical wave-guide.

If the radius of the bar is small compared to the length of the loading stress wave, the decay or dispersion of the stress wave as it travels along the bar is minimal. In addition the bar is made sufficiently long so that all the measurements have been completed before reflected waves can interfere.

The following theory relies on the four assumptions listed below:-

- (a) Wave propagation phenomena (inertial effects) within the deforming specimen are minimal (c.f. shock wave theory).
- (b) Friction locking effects at the specimen - bar interface are insignificant.
- (c) One-dimensional wave behaviour governs the response of the pressure bars.
- (d) The pressure bars are of uniform cross-section, made of the same material and remain elastic throughout the test.

Figure 3.18 shows a diagram of the impacting bar head, the specimen and the specimen bar shortly after impact. The following refers to this diagram.

Just before impact the impacting bar is moving with a velocity v_o . After impact the forces acting on each side of interface I must be equal. Thus:-

$$\sigma_{bI} A_b = \sigma_s A_s \quad (3.41)$$

where A is the cross-sectional area and σ the stress.

After impact two pressure fronts travel away from interface I. Using the expression, given by equation (A19), for the pressure developed across a shock front in terms of the rate of change of momentum gives

$$\sigma_{bI} = \rho_b C_b (v_o - u) \quad (3.42)$$

and

$$\sigma_s = \rho_s C_s u \quad (3.43)$$

where C is the velocity of sound, ρ is the unloaded density and u is the partical velocity after impact.

Combining equations (3.41), (3.42) and (3.43) gives

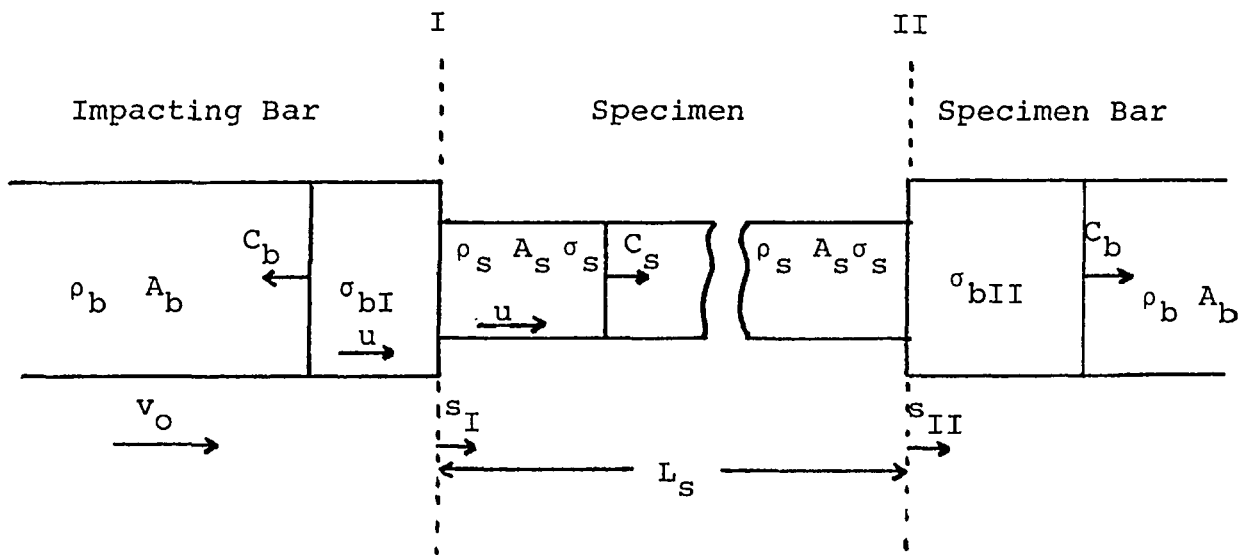


Fig. 3.18 Determination of pressures and strains due to Hopkinson bar loading

$$u = v_o - \frac{\sigma_s A_s}{\rho_b C_b A_b} \quad (3.44)$$

Now the displacement of this interface is given by:-

$$s_I = \int_0^t u \, dt \quad (3.45)$$

so that:-

$$s_I = \int_0^t \left[v_o - \frac{\sigma_s A_s}{\rho_b C_b A_b} \right] dt \quad (3.46)$$

At interface II the displacement is given by:-

$$s_{II} = \int_0^t \frac{\sigma_{bII}}{\rho_b C_b} \, dt \quad (3.47)$$

Thus the total reduction in specimen thickness is

$$s = s_I - s_{II} = \int_0^t \left[v_o - \frac{2\sigma_{bII}}{\rho_b C_b} \right] dt \quad (3.48)$$

This gives the final strain

$$\epsilon_s = \frac{s}{L_s} = \frac{v_o t}{L_s} - \frac{2}{\rho_b C_b L_s} \int_0^t \sigma_{bII} \, dt \quad (3.49)$$

and the strain rate

$$\dot{\epsilon}_s = \left[v_o - \frac{2\sigma_{bII}}{\rho_b C_b} \right] / L_s \quad (3.50)$$

where L_s is the specimen thickness.

Now σ_{bII} and v_o are determined from the strain gauge output and the light curtain photo-diode respectively. Thus ϵ_s , $\dot{\epsilon}_s$ and σ_s can be calculated if the volume of the specimen is assumed constant during the test. A computer program was written to generate values of ϵ_s , $\dot{\epsilon}_s$ and σ_s from the bar displacement and strain gauge data. A typical set of results is shown in Fig. 3.19. The maximum strain is given in Fig. 3.19 f. The jagged appearance of some of the curves is due to the relatively large integration interval.

If the specimen compression is assumed to be adiabatic at these strain rates, 85 - 95% of the plastic work represented by the area under the stress - strain curve appears as thermal energy. The remainder is stored as latent energy [33].

Thus the temperature rise ΔT due to compression can be estimated using

$$\Delta T = \frac{0.9}{\rho_{s0} C} \int_0^{\epsilon_s^p} \sigma d\epsilon_s^p \quad (3.51)$$

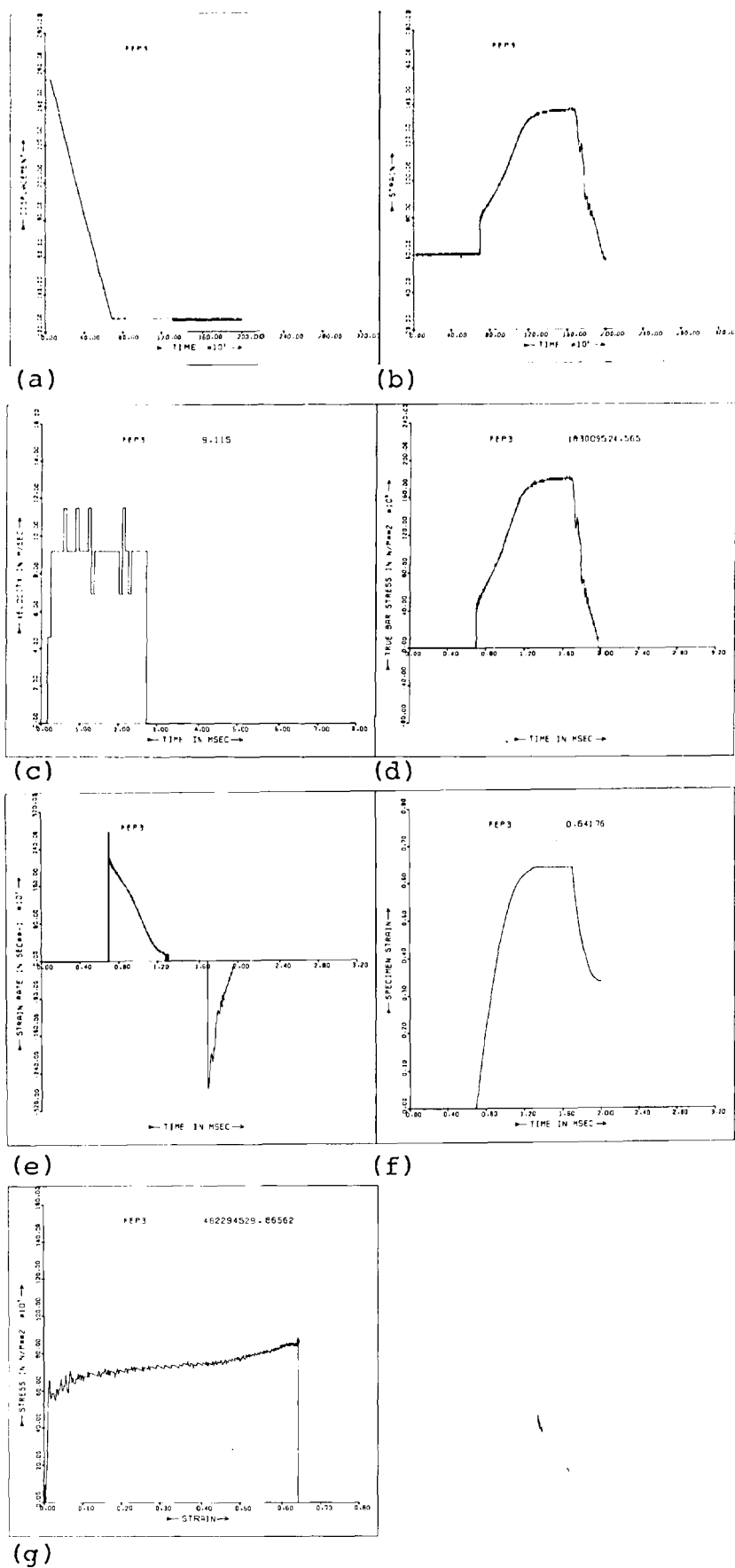


Fig. 3.19 Raw loading and bar velocity curves and their conversion into strain rate and stress - strain curves for a typical specimen (iron)

Where $\bar{c}$ is the specific heat and ϵ_s^P is the plastic strain.

The energy input into the specimen is given in Figure 3.19 g. Values of ϵ_s calculated from equation (3.49) and those measured from the recovered specimens are listed in Table 3.13. The calculated values of ΔT are given along with the shock values in Table 3.14.

3.5.2 Comparison of Predicted and Actual Final Strains

The calculated and measured strains did not differ by more than 10% which was sufficient to give reasonable values for the stored energies and the adiabatic release temperatures.

When calculating the final strains it was assumed that the maximum strain occurred when the loading curve flattened out (Fig. 3.19 d). If necessary, the value of the impact bar velocity (v_0) was changed so that the strain rate became zero at this point (Fig. 3.19 e). This gave calculated results close to the measured strains and thus appeared to justify this assumption while suggesting some inaccuracy in the measurement of the bar velocity.

The flat part of the loading curve could, of course, indicate that the specimen has ceased to work harden and begun to fail. However, calculations based on this assumption led to nominal strains greater than one and thus it was assumed that

compressive failure did not occur. This was supported by low strain rate tests (10^{-2}s^{-1}) which showed no failure up to the final Hopkinson bar strain for all the metals and alloys except duralumin. Duralumin failed in shear at a strain of about 0.55, which suggests that the duralumin Hopkinson bar results must be viewed with some suspicion.

Table 3.1 Explosive Characteristics [22]

Explosive	ρ_0 (g/cc)	ρ_{CJ} (g/cc)	P_{CJ} (GPa)	u_{CJ} (km/s)	D (km/s)	C_E (km/s)	γ
Baratol	2.46	3.29	15.5	1.26	5.00	3.73	2.96
Comp B	1.67	2.26	25.7	2.01	7.69	5.67	2.83
95% HMX	1.78	2.41	35.7	2.29	8.76	6.48	2.83

Table 3.2 Specimen Pressures (GRAPH)

Explosive	P (GPa)				
	Al	Al-4.5%Cu	Cu	Fe-18Cr-8Ni	Fe
Baratol	18.8	19.0	25.6	25.6	24.8
Comp B	31.7	32.1	42.6	42.6	41.3
95% HMX	41.3	41.9	56.7	56.7	55.0

Table 3.3 Specimen Particle Velocities (GRAPH)

Explosive	u (km/s)				
	Al	Al-4.5%Cu	Cu	Fe-18Cr-8Ni	Fe
Baratol	1.03	1.02	0.59	0.59	0.65
Comp B	1.58	1.55	0.91	0.91	0.98
95% HMX	1.94	1.91	1.14	1.15	1.22

Table 3.4 Explosive Block Parameters

Explosive	y_s	x' (km/s)
Baratol	0.381	1.27
Comp B	0.389	2.02
95% HMX	0.387	2.29

Table 3.5 Reaction Zone Lengths [24] and Minimum Charge Radii

Explosive	X	R
	mm	mm
RDX	2.0	40.0
Comp B	0.13	2.6*
TNT	3.0	60.0

* Private communication K. Burrows AWRE

Table 3.6 Specimen Shock Velocities (GRAPH)

Explosive	U_S /(km/s)				
	Al	Al-4.5%Cu	Cu	Fe-18Cr-8Ni	Fe
Baratol	6.76	6.69	4.84	5.48	4.88
Comp B	7.44	7.43	5.27	5.96	5.37
95% HMX	7.89	7.89	5.58	6.24	5.74

Table 3.7 Specimen Rarefaction Velocities

C_L = Leading wave C_O = Trailing wave (P = 0)

Explosive		C (km/s)				
		Al	Al-4.5%Cu	Cu	Fe-18Cr-8Ni	Fe
Baratol	C _L	6.95	6.74	4.96	5.61	5.20
Comp B	C _L	7.64	7.44	5.48	6.13	5.91
95% HMX	C _L	8.05	7.87	5.84	6.50	6.41
	C _O	5.28	5.33	3.94	4.57	3.57

Table 3.8 Minimum Values of W/d

$$\frac{W}{d} \geq 2 \left[\frac{C_p}{U_s \rho_o} + \frac{C_p}{C_{\rho_o} \rho_o} \right]$$

Explosive	Al	Al-4.5%Cu	Cu	Fe-18Cr-8Ni	Fe
Baratol	5.53	5.35	5.20	5.05	5.82
Comp B	6.28	6.06	5.87	5.59	6.74
95% HMX	6.75	6.53	6.35	6.04	7.39

Table 3.9 Values of t_s/d ($\mu s/mm$)

$$\frac{t_s}{d} = \frac{1}{U_s} + \frac{1}{C_{\rho_o}}$$

Explosive	Al	Al-4.5%Cu	Cu	Fe-18Cr-8Ni	Fe
Baratol	0.339	0.337	0.460	0.401	0.485
Comp B	0.324	0.322	0.444	0.387	0.466
95% HMX	0.316	0.314	0.433	0.379	0.454

Table 3.10 Minimum Values of s/d

$$\frac{s}{d} \geq \left[\frac{1}{U_s} + \frac{1}{C_{\rho_o}} \right] / \left[\frac{1}{C_E} - \frac{1}{D} \right]$$

Explosive	Al	Al-4.5%Cu	Cu	Fe-18Cr-8Ni	Fe
Baratol	5.00	4.97	6.78	5.91	7.15
Comp. B	7.01	6.96	9.60	8.37	10.08
95% HMX	7.84	7.79	10.75	9.41	11.27

Table 3.11 Velocity Interferometer Results

	u_{fs} (km/s)	U_s (km/s)
Copper + baratol	1.16	4.7
Aluminium + comp. B	2.96	7.3
Iron + comp. B	1.75	
18-8 steel + comp. B	1.66	6.0
Dural + 95% HMX	3.22	8.0
Iron + 95% HMX	1.98	

Table 3.12 Condition of Recovered Central Discs

Explosive	Metal				
	Al	Al-4.5%Cu	Cu	Fe-18Cr-8Ni	Fe
Baratol	Destroyed	Complete	Complete	Complete	Complete
Comp. B	-	Destroyed	Fragments	Complete	Fragments
95% HMX	-	-	Destroyed	Complete but damaged	Fragments

Table 3.13 Calculated and Measured Final Strains (ϵ_s)
due to Hopkinson Bar Loading

	Strain ϵ_s	
	Calculated	Measured
Aluminium	0.39	0.43
Duralumin	0.76	0.75
Copper	0.70	0.71
Iron	0.64	0.64
18-8 stainless steel	0.57	0.53

Table 3.14 Shock and Isentropic Release Temperatures (T_H, T_a) [25, 33]

T_m is melting temperature

	Al			Al-4.5%Cu			Cu			18-8 Steel			Fe		
	$T_H(^{\circ}\text{C})$	$T_a(^{\circ}\text{C})$	T_a/T_m	$T_H(^{\circ}\text{C})$	$T_a(^{\circ}\text{C})$	T_a/T_m	$T_H(^{\circ}\text{C})$	$T_a(^{\circ}\text{C})$	T_a/T_m	$T_H(^{\circ}\text{C})$	$T_a(^{\circ}\text{C})$	T_a/T_m	$T_H(^{\circ}\text{C})$	$T_a(^{\circ}\text{C})$	T_a/T_m
Hopkinson Bar	-	54	0.35	-	189	0.56	-	112	0.28	-	171	0.25	-	136	0.22
Baratol	239	100	0.40	217	91	0.44	165	69	0.25	145	57	0.18	190	93	0.20
Comp. B	477	204	0.51	486	212	0.59	352	163	0.32	301	132	0.23	405	213	0.27
95% HMX	703	300	0.61	753	329	0.73	560	269	0.40	476	218	0.27	639	343	0.34

CHAPTER IVRESULTS4.1 Sample Preparation

Five metals and alloys were loaded using the Hopkinson bar and explosive assemblies detailed in Chapter III. The specifications and compositions are listed in Table 4.1. All specimens were cut from circular 76.2 mm (3") diameter bars.

The Hopkinson bar discs were cut or spark-machined in the longitudinal bar direction from plates sliced off the bars. The discs were then machined to within 25 μ m (0.001") of the required thickness and the faces were made flat to within less than 6 μ m.

The explosively loaded discs and surrounding plates were cut in the same manner but machined to an optically flat finish (1 μ).

After loading, 3 mm cylinders were spark-machined from the deformed specimen discs. These cylinders were sliced if possible and the sections were ground to thicknesses suitable for jet thinning.

In the case of the explosively loaded specimen discs which remained intact, the 3 mm cylinders were sliced so that a traverse from the shocked surface to the free surface was obtained. In this way it was hoped to observe the effect of pulse length and / or rarefaction stress rate on the specimens. This was deemed unnecessary for the Hopkinson bar discs, as the variation of loading time and rarefaction rate across the disc is negligible compared to the actual loading time and rarefaction rate.

A Streuers Tenupol jet thinning machine was used to provide thin foils suitable for electron microscopical work. Table 4.2 gives the polishing solutions and conditions used for each material. The currents were nominal and values between 80 and 150 mA were used to achieve good polishing conditions.

All specimens except pure iron were examined using a Phillips EM301, 100 kV transmission electron microscope. On account of their high magnetism, the pure iron discs were examined in an AEI EM7, 1MV transmission electron microscope. The beam deflection and image distortion caused by the magnetic iron are reduced to reasonable amounts due to the high accelerating voltage. The double tilt stage facility was used in both cases.

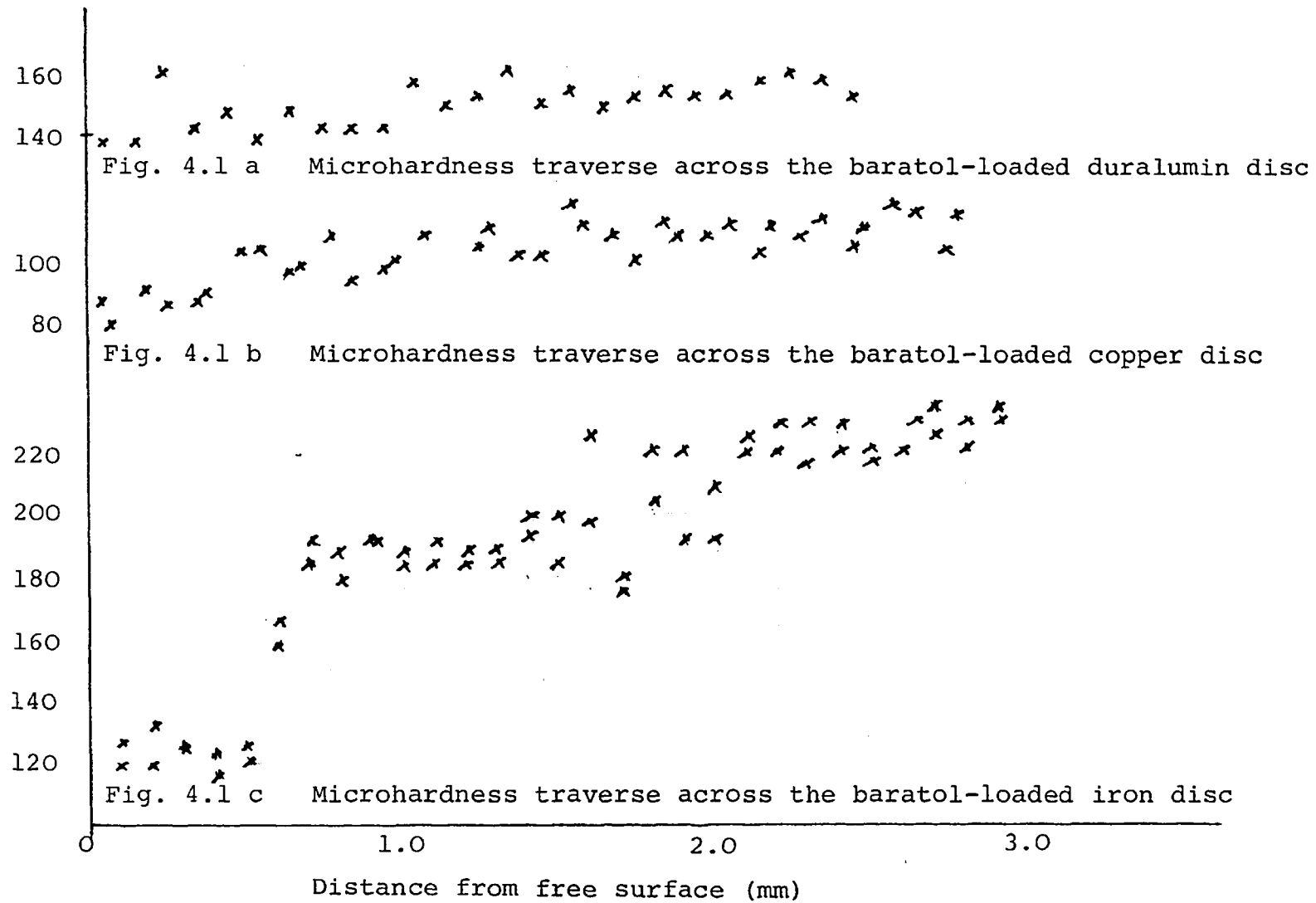
After the removal of the 3mm cylinders, the remainders of the specimen discs were sectioned perpendicular to their

loaded surfaces, polished and etched. Microhardness traverses were run perpendicular to the loaded surfaces of the deformed discs and these were compared with the as-received hardnesses. Optical micrographs were taken of all the polished and etched samples using a Carl Zeiss Jena Neophot 21 optical microscope. The microhardness measurements were taken using a Reichert microhardness tester on a Reichert Me F2 optical microscope.

4.2 Optical Microscopy and Hardness Measurements

Microhardness values are listed in Table 4.3. Where two or more numbers are given for one test, the first was taken at the free surface and the last at the shocked surface. In the case of iron shocked by baratol, the values correspond to three hardness plateaux. The hardness changes from the free surface to the shocked surface for duralumin, copper and iron loaded with baratol are shown in Figure 4.1. Those for 18 - 8 stainless steel, and to some extent iron shocked above the transformation pressure, showed considerable scatter ($\pm 10\%$). This could have been due to local variations in phase proportions (γ, α, α') present after shock loading.

The duralumin and pure copper traverses (Fig. 4.1 a,b) show a gradual increase in hardness from the free surface to the shocked surface. The iron specimen (Fig. 4.1 c)



has three hardness levels. The first level (VH 125) exists up to 0.55 mm from the free surface. There follows an abrupt rise to VH 187 which remains constant for the next 0.85 mm. From this point, the hardness rises erratically within the next 0.6 mm to approx. VH 225, after which it remains constant up to the shocked surface.

All other hardness values listed in Table 4.3 are averages taken where no hardness trends were observed.

The optical micrographs of baratol-loaded iron etched with 2% nital show a close connection with the microhardness values. The lowest hardnesses correspond to a lightly etched region within 0.55 mm of the free surface. This region is divided into an untwinned section within about 0.28 mm of the free surface, and a twinned section between 0.28 and 0.55 mm from the free surface (Fig. 4.2 a,b). The abrupt change of hardness at 0.55 mm is accompanied by an equally abrupt change in microstructure. This becomes a mixture of darkly-etched grains containing a combination of large bands (twins ?) and a fine structure, and what appear to be heavily twinned grains (Fig. 4.2 a,c). Between 1.4 mm and 2.0 mm from the free surface the hardness increases gradually to a maximum. As the shocked surface is approached, the twinned grains become less numerous and the darkly etched grains contain a simpler, finer structure (Fig. 4.2 d). The region next to the shocked surface shows almost no evidence of twins and the dark-

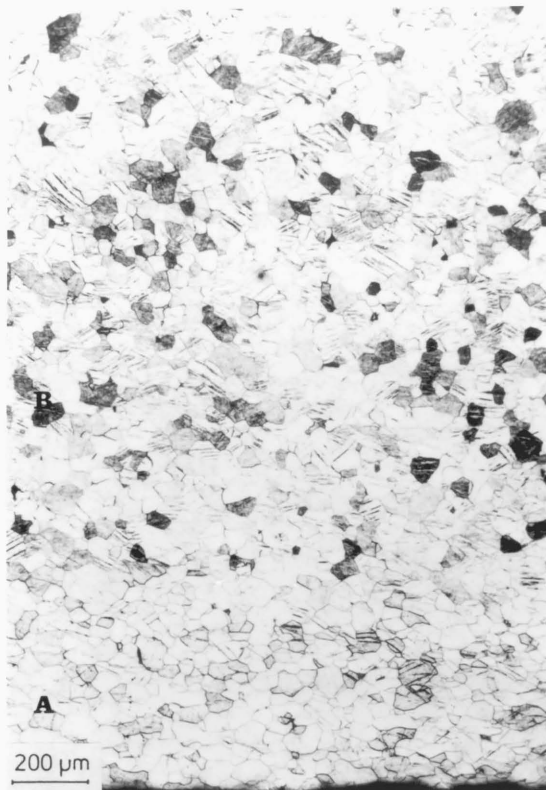


Fig. 4.2 a Baratol loaded free surface side



Fig. 4.2 b Onset of twinning (A in Fig. 4.2 a)



Fig. 4.2 c Heavily twinned and darkly etched grains in region corresponding to second hardness plateau (B in Fig. 4.2 a)



Fig. 4.2 d Darkly etched grains in region corresponding to third hardness plateau (C in Fig. 4.2 e)

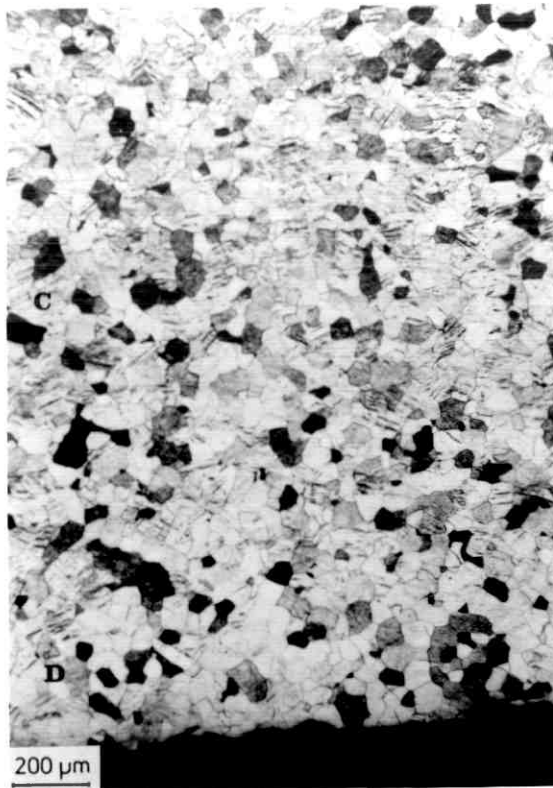


Fig. 4.2 e Baratol loaded iron-shocked surface side

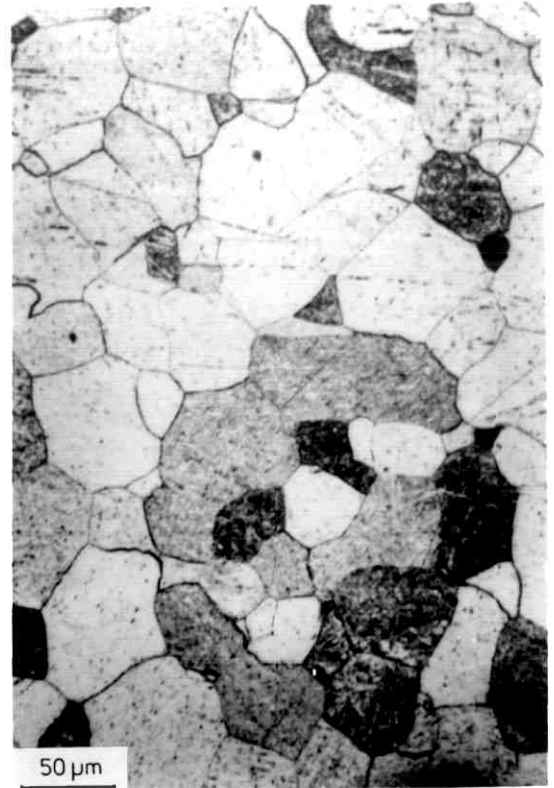


Fig. 4.2 f Needle-like grain structure near shocked surface (D in Fig. 4.2 e)

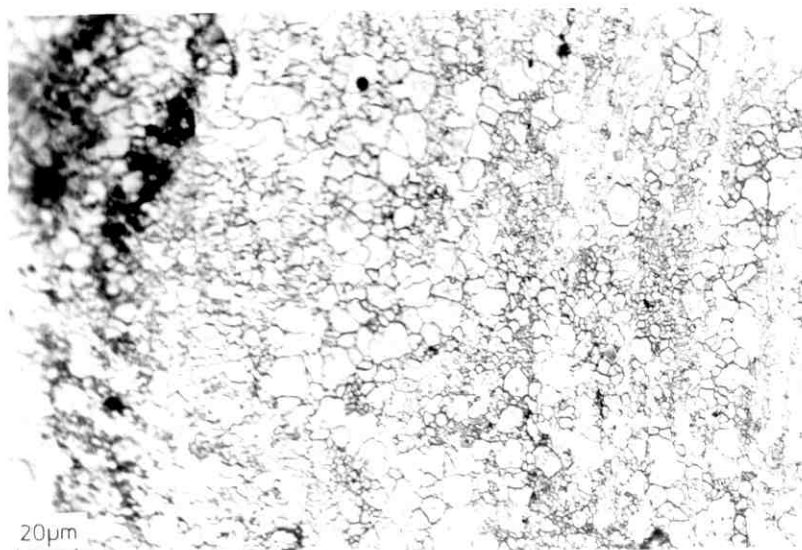


Fig. 4.3 Recrystallized composition B loaded iron at free surface

etched grains contain a very fine needle-like structure (Fig. 4.2 e,f).

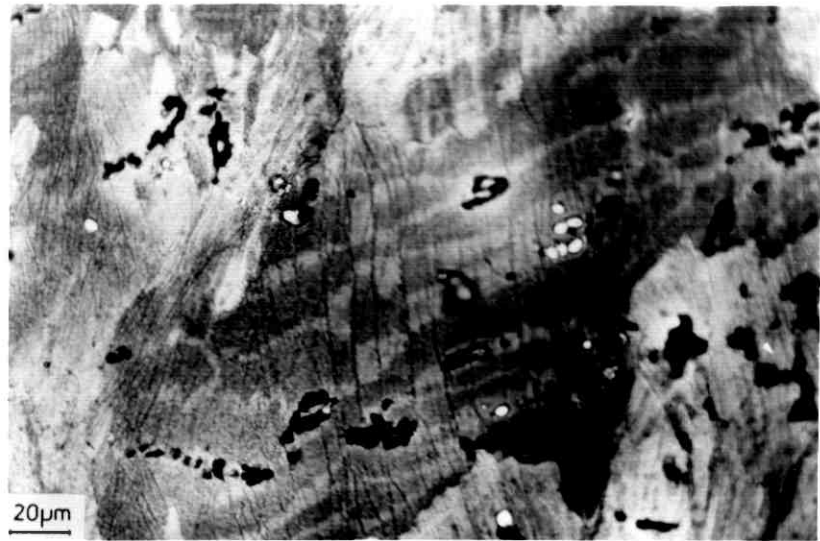
The position of these hardness steps and the onset of twinning permit a more detailed analysis of the shock wave behaviour in baratol-loaded iron. This is carried out in Section 4.2.1.

The three hardness levels do not occur in composition B and HMX loaded iron. The discs were considerably deformed and the microstructures contained large numbers of heavily deformed, darkly etched grains, along with areas of recrystallization in regions of very high deformation such as at the free surface (Fig. 4.3). Micrographs of Hopkinson bar loaded specimens show no evidence of twinning.

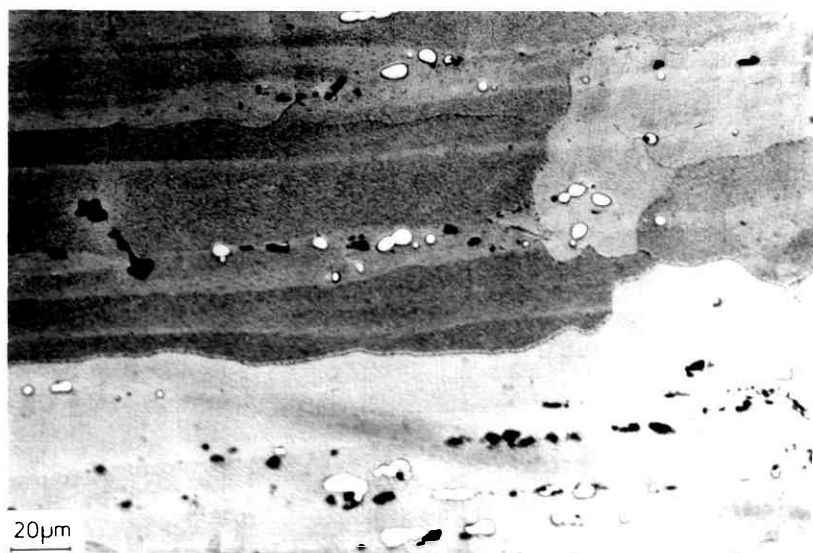
Micrographs of dural etched in HF show small white areas which are believed to be caused by CuAl_2 [119].

Whereas shock appears to have little observable effect on these areas, the high deformation undergone under Hopkinson bar loading appears to have caused a reduction in their size (Fig. 4.4).

Micrographs of copper etched with alcoholic ferric chloride show the typical effects of Hopkinson bar and shock loading (Fig. 4.5). The heavily deformed grains of the Hopkinson bar sample (Fig. 4.5 a) contrast sharply with the almost undeformed grains of the baratol loaded sample (Fig. 4.5 b).



(a) Hopkinson bar



(b) Baratol

Fig. 4.4 Comparison of CuAl_2 particles (white regions) in Hopkinson bar and baratol loaded duralumin

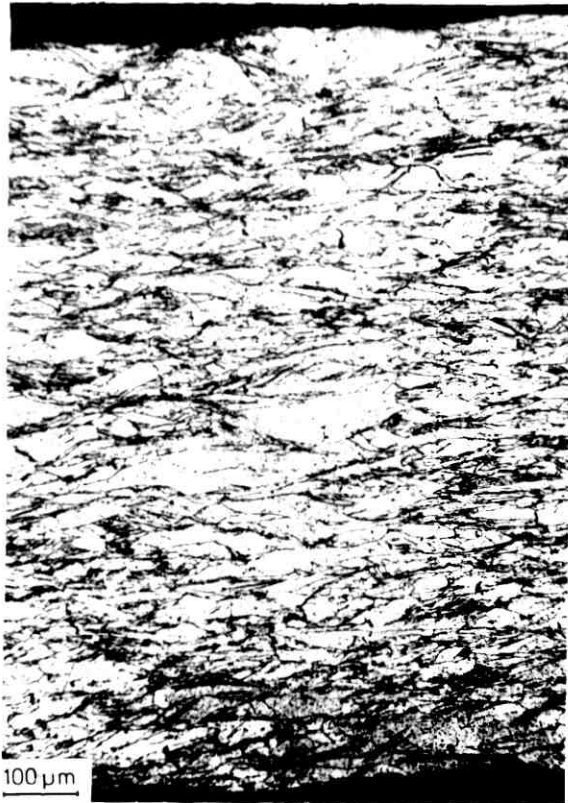


Fig. 4.5 a Hopkinson bar loaded copper



Fig. 4.5 b Baratol loaded copper



Fig. 4.5 c Recrystallization in HMX loaded copper

However, at higher explosive loads the extreme deformation of the copper during deceleration led to recrystallization at the free surface (Fig. 4.5 c).

Nothing of interest was seen on micrographs of aluminium and 18 - 8 stainless steel.

4.2.1 Analysis of the Shock Wave Profile in Baratol-loaded Iron (Fig. 4.6)

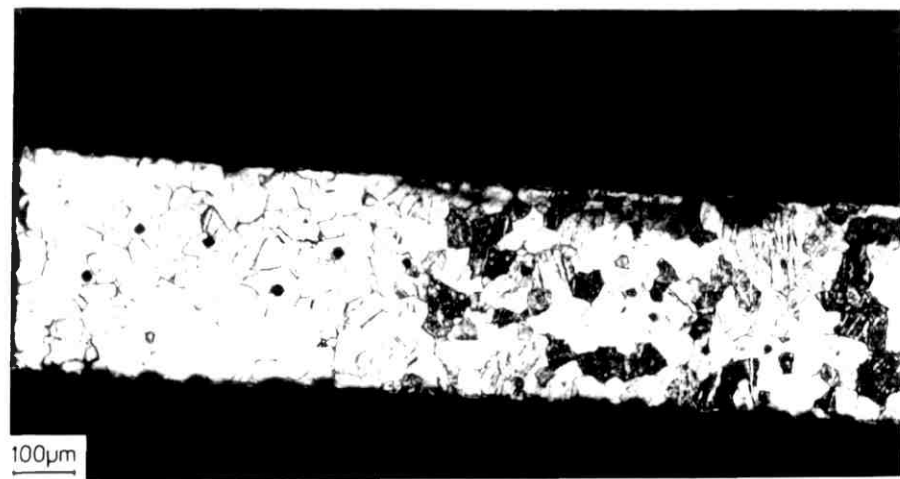
The shock profile consists of three successive shock fronts. The leading elastic (e) shock ($P_e = 0.68$ GPa) and the second 13 GPa, P1 shock have velocities which are largely independent of the final loading pressure. These were determined by Bancroft et al. [131] and are $U_e = 6.04$ km/s and $U_{P1} = 5.05$ km/s. The velocity of the third, transformation shock (P2) depends on the final shock pressure.

The position of the interaction between the reflected e-wave (R_e) and the P1 wave can be calculated using the equation

$$\frac{U'_{P1}}{U_e} = (1-r) / [1 + (U_e/R_e)r] \quad (4.1)$$

developed by Fowler et al. [128] where

$$U'_{P1} = \frac{\rho_e}{\rho_o} (u_{P1} - u_e) , \quad r \text{ is the ratio of the distance}$$

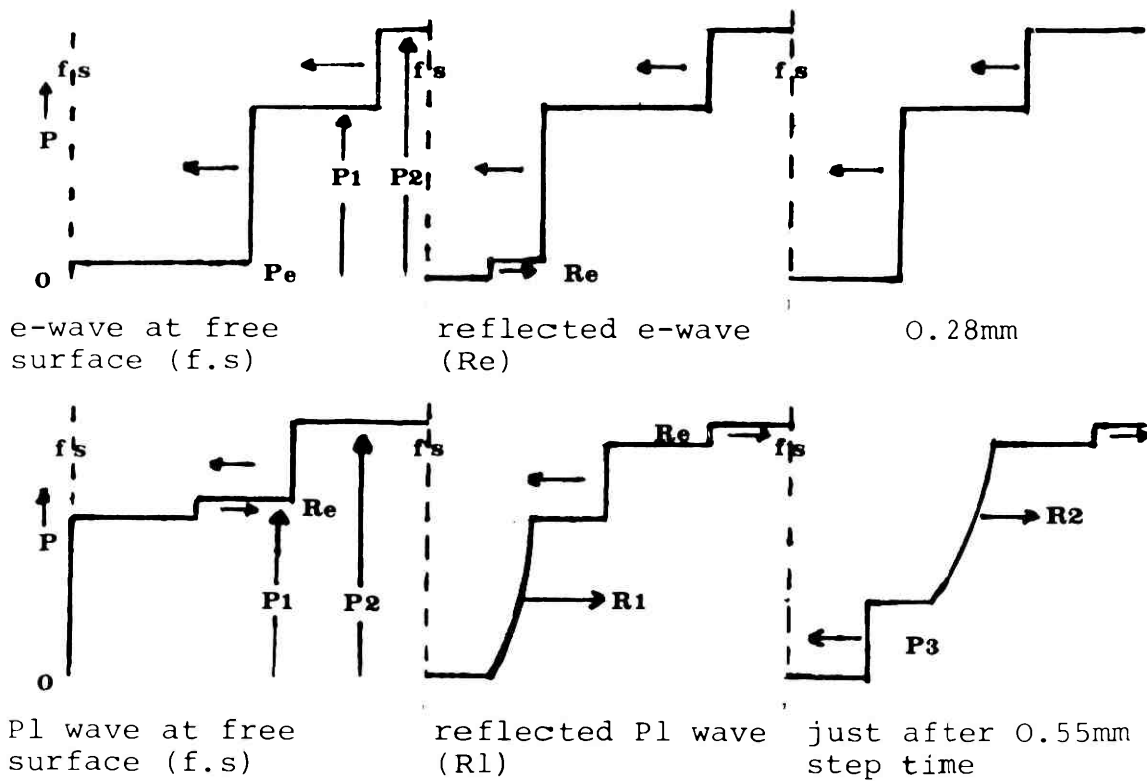


Elastic Zone | P1 Zone | P2 Zone

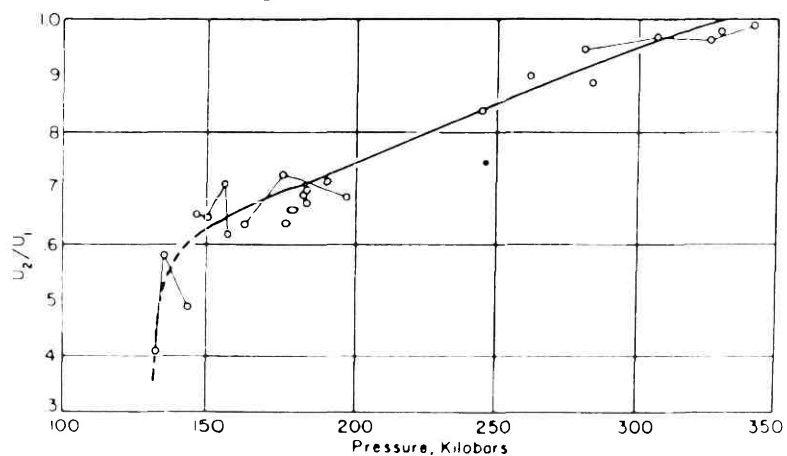
0.28mm

0.55mm

(a) Micrograph of free surface region



(b) Wave interaction diagrams



(c) Graph of U_2/U_1 v pressure (Smith and Fowler [59])

Fig. 4.6 Wave interaction in baratol loaded iron

between the interaction position and the free surface, and the specimen width, and R_e is the elastic rarefaction wave velocity.

Using the values of ρ_e, u_e, R_e, U_e and U_{p1} determined by Bancroft et al. in equation 4.1 places the interaction position 0.28 mm from the specimen free surface. This corresponds to the onset of twinning (Fig. 4.6 a).

Equation 4.1 can also be used to determine the P2 shock wave velocity U_{p2} from the location of the R1/P2 wave interaction. This occurs 0.55 mm from the free surface and gives U_{p2} as 3.80 km/s. The U_{p2}/U_{p1} ratio of 0.75 lies below the curve of Smith and Fowler [59] (Fig. 4.6 c) at the calculated shock pressure (24.8 GPa) suggesting that the actual shock pressure may have been about 21 GPa.

4.3 Electron Microscopy

4.3.1 Aluminium

The as-received structure consisted of a mixture of large (>20 μ m) and clusters of small (4 - 6 μ m), generally dislocation-free grains. Hopkinson bar loading resulted in a mixture of elongated cells and recovered or recrystallized grains (Fig. 4.7 a). When shock loaded with baratol, aluminium recrystallized totally, producing some grain refinement (1 - 3 μ m) (Fig. 4.7 b).

4.3.2 Duralumin

The structure of the as-received metal contained evenly distributed spherical CuAl_2 precipitates and easily resolvable dislocations (Fig. 4.8 a).

After Hopkinson bar loading, individual dislocations were no longer resolvable and a cellular structure (Fig. 4.8 b) was formed. No shear bands were observed.

The specimen disc survived shock loading with baratol, and thus a scan across the disc was possible. In general, the structure contained bands of dislocations lying on $\{111\}$ planes (Fig. 4.8 c). There appeared to be a gradual increase in band definition and frequency with decreasing distance from the shocked surface, whereas a more cellular structure was found in regions further away from the shocked surface. Close to the free surface there existed a mixture of isolated slip bands and a few groups of narrow bands whose orientation differed by one or two degrees from that of the matrix (Fig. 4.8 d).

Micrographs taken from an area within 0.1 mm of the shocked surface showed twin-like features (Fig. 4.8 e). Analysis of the diffraction pattern shows that if these features were twins then twinning occurred on the $\{\bar{1}11\}$ plane. The $[\bar{1}92]$ matrix zone axis lies in the same direction as the $[\bar{7}\bar{1}6]$ twin zone axis. The $[\bar{7}\bar{1}8]$ direction corresponds to the line of intersection of the $(\bar{1}11)$ and

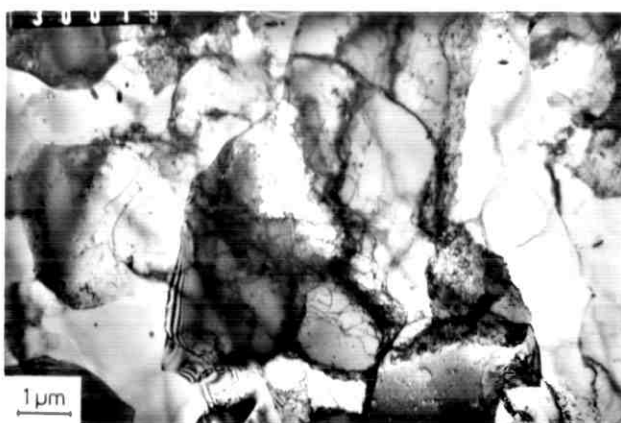


Fig. 4.7 a Hopkinson bar loaded aluminium

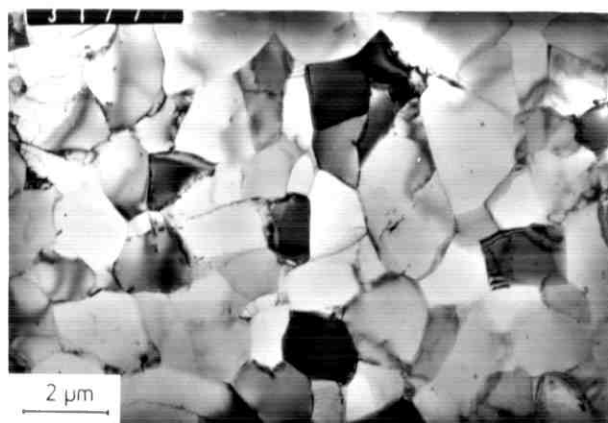


Fig. 4.7 b Baratol loaded aluminium

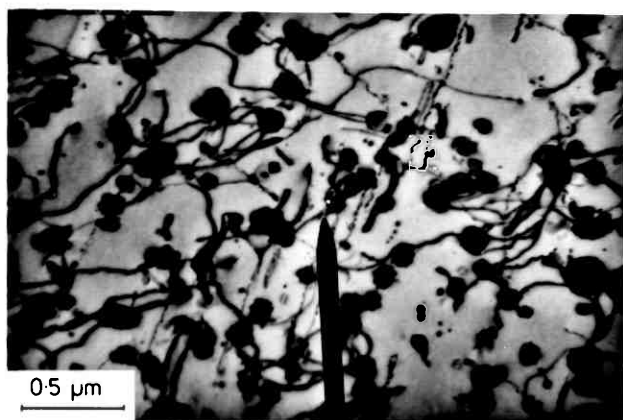


Fig. 4.8 a As-received duralumin

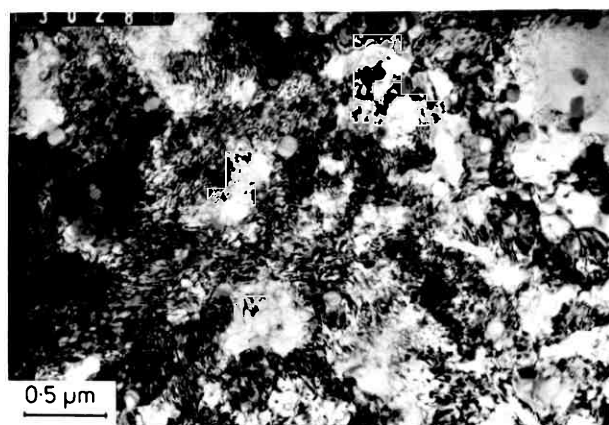


Fig. 4.8 b Hopkinson bar loaded duralumin

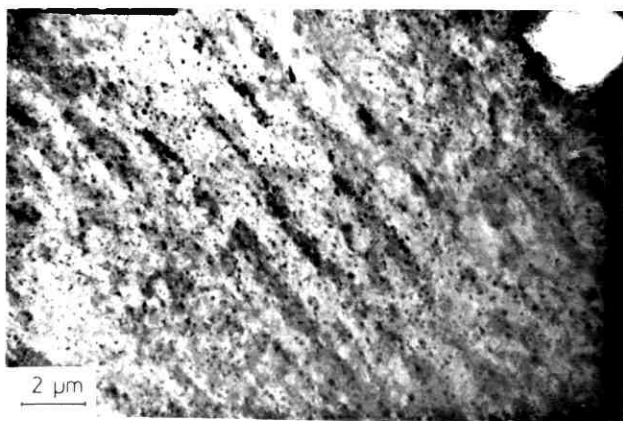


Fig. 4.8 c $\{111\}$ bands in baratol loaded duralumin

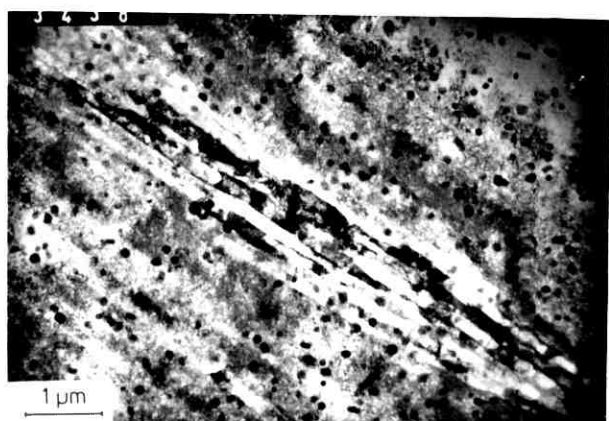
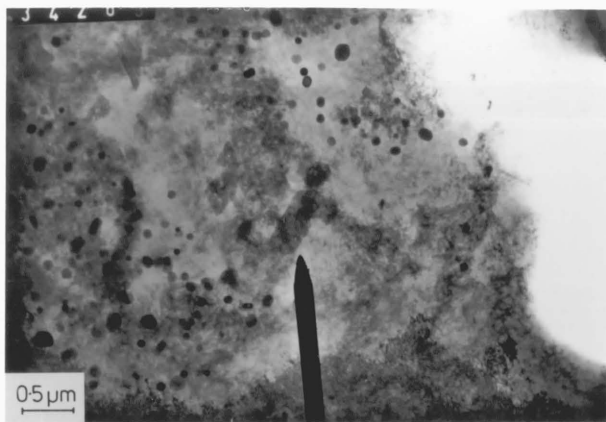
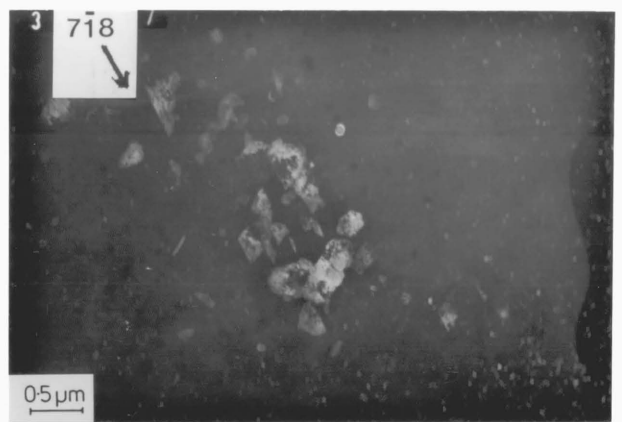


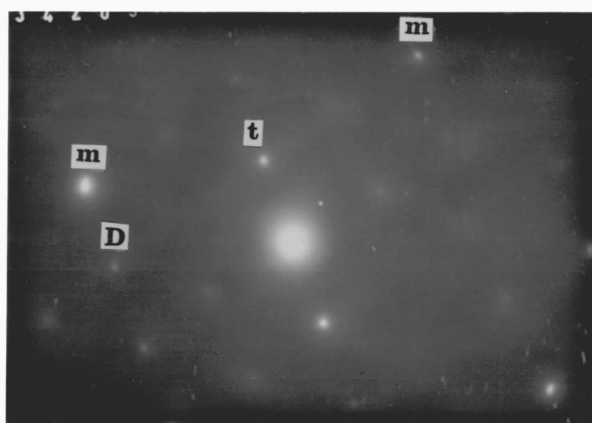
Fig. 4.8 d Narrow bands near baratol loaded duralumin free surface



Bright field
(twinned region indicated)



Dark field



Diffraction pattern and
computer plot

D = double diffraction
x = $\begin{bmatrix} 1 & 9 & 2 \end{bmatrix}$ matrix (m)
* = $\begin{bmatrix} 7 & 1 & 6 \end{bmatrix}$ twin (t)

			.888	
			.777	
			.666	919,
			.555	
939,	228,		.444	517,
	626,		.333	
			.222	804,
	313,		.111	.759
.395			000,000	.668
.486	711,			.577
.577			.111	.486
.668	402,		.222	.395
.759	804,		.333	
			.444	
	517,		.555	
			.666	
919,			.777	
			.888	
			.999	

Fig. 4.8 e Analysis of twin features in baratol loaded duralumin

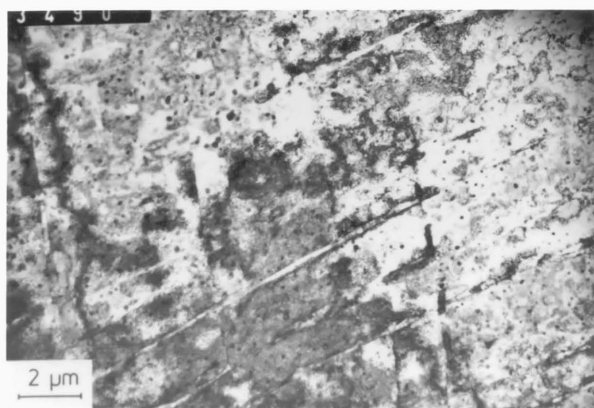


Fig. 4.8 f Shear bands in composition B loaded duralumin

the ($\bar{1}92$) planes.

When shock-loaded with composition B, the disc broke up, leaving only a few fragments. These contained large numbers of shear bands which ran in as many as three directions (Fig. 4.8 f) and had orientations quite different from the matrix. Similar bands occurred to a considerably lesser extent in the heavily-deformed outer plate fragments of the baratol-loaded assembly.

Under all the above loading conditions there was no observable variation in precipitate size and no visible destruction of the particles occurred.

4.3.3 Copper

The structure of the as-received copper consisted of elongated dislocation cells of approximately 0.6 μm in width (Figure 4.9 a).

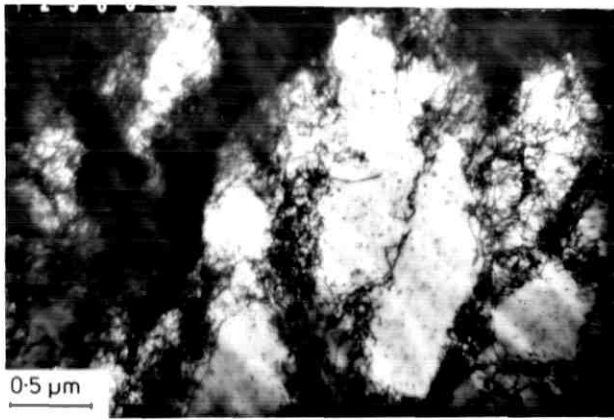
After Hopkinson bar loading, the microstructure contained equiaxed cells (0.2 μ - 0.5 μm diameter), frequent bands of dislocations and the occasional long shear band containing fine twins (0.01 μm), which were sometimes accompanied by cross-twins. In a few cases, some areas of recovery and recrystallization were observed (Figure 4.9 b - f).

When copper was shocked with baratol, the recovered disc was undamaged and thus a structure scan was impossible. The microstructure consisted of incompletely-formed cells ($0.35\text{ }\mu\text{m}$) and dense distribution tangles (Fig. 4.10 a). In addition, bands of thin twins 0.015 to $0.003\text{ }\mu\text{m}$ thick were observed in all regions except those close to the free and shocked surfaces (Fig. 4.10 b,c). Well-defined, equiaxed cells occurred near the free surface and the microstructure near to the shocked surface consisted of a mixture of loops and tangled dislocations (Fig. 4.10 d). The overall dislocation density appeared to increase with distance from the free surface.

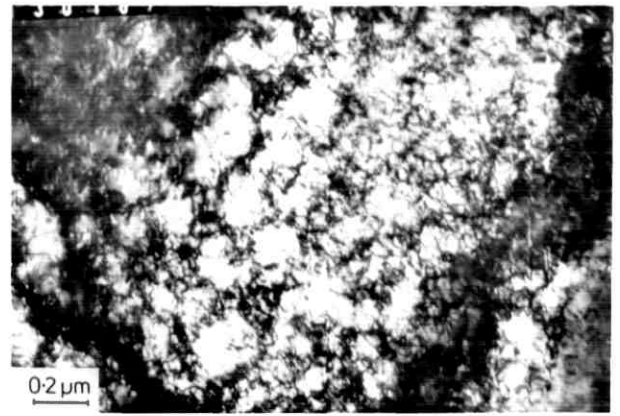
Copper discs shocked by composition B and HMX did not survive recovery intact. Extensive recrystallization occurred in both cases. The average grain size was approximately $5\text{ }\mu\text{m}$ and considerable numbers of stacking faults and annealing twins were present.

Two interesting features of the recrystallized grains were noted. There was considerable grain boundary etching by the thinning solution (Fig. 4.10 e) - this was also noticed by Murr and Korbonski in shocked stainless steel [121] - and long straight dislocations were observed running parallel to the annealing twins (Fig. 4.10 f).

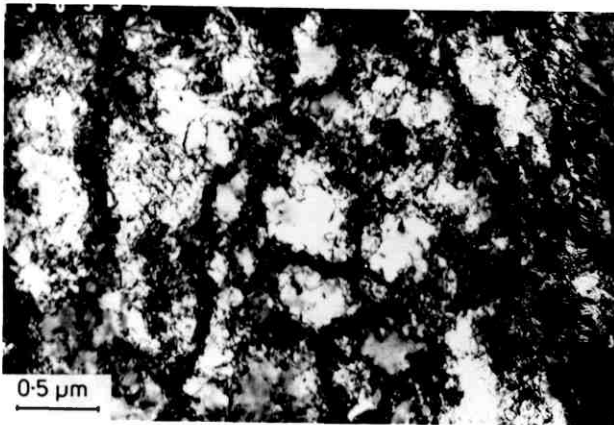
Finally, fine twinning occurred in the unrecrystallized grains of the composition B-loaded copper and some grains in the HMX-loaded copper contained dislocation loops and



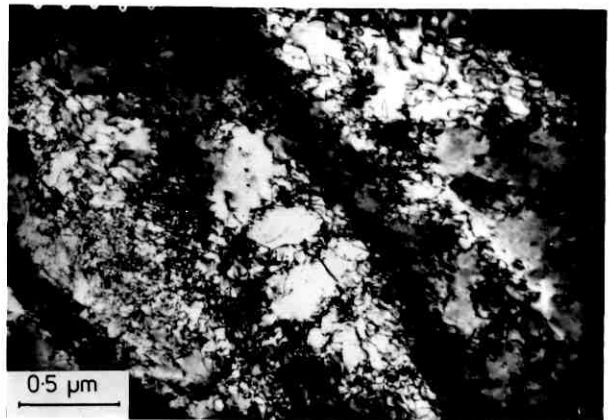
(a) Elongated dislocation cells



(b) Equiaxed dislocation cells



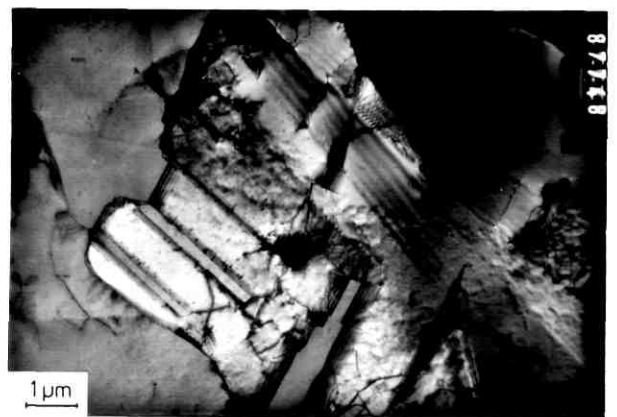
(c) Dislocation cells



(d) Dislocation bands

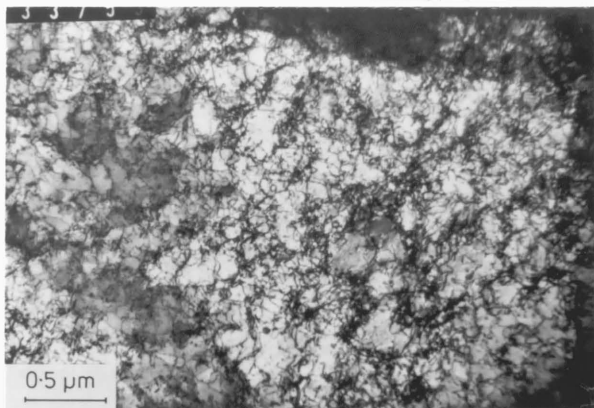


(e) Bands of fine twins

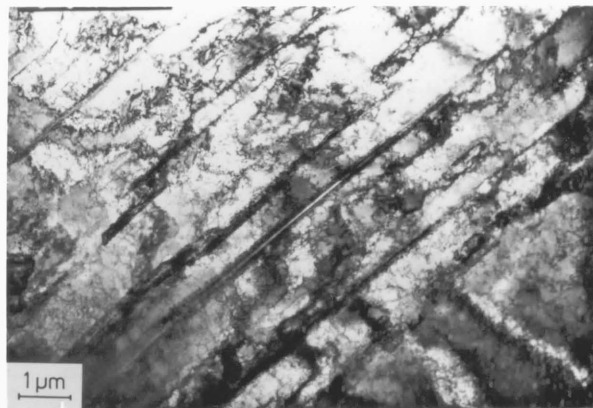


(f) Recrystallized grain

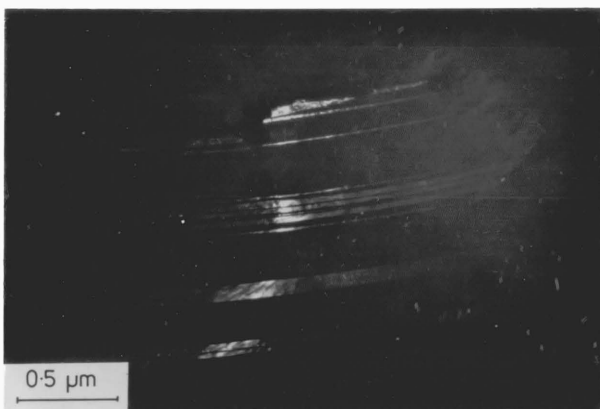
Fig. 4.9 As-received (a) and Hopkinson bar loaded (b - f) copper



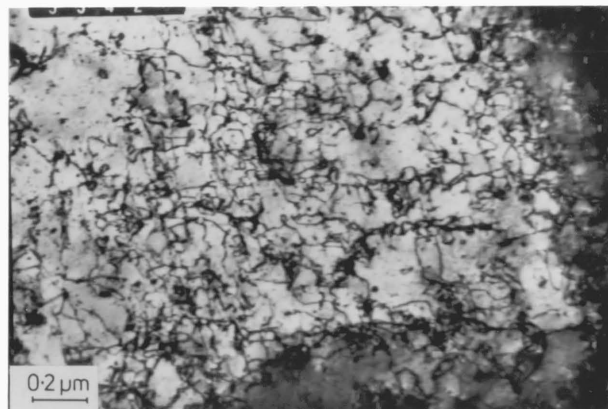
(a) Dislocation structure in bulk



(b) Twin bands



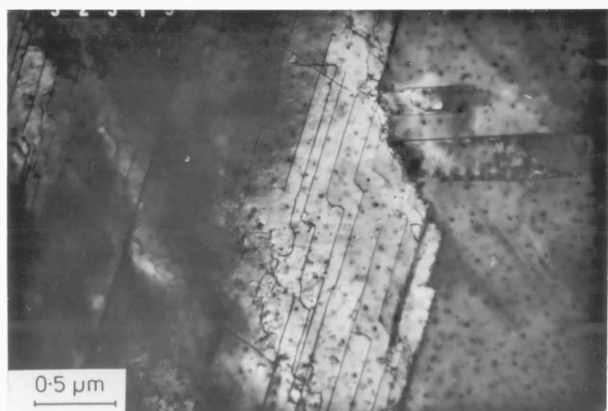
(c) Dark field of twin bands



(d) Loops and tangles near shocked surface



(e) Recrystallized grains



(f) Long straight dislocations in recrystallized grains

Fig. 4.10 Explosively loaded copper-baratol (a - d), composition B (e, f)

tangles similar to those found near the shocked surface of the baratol-loaded disc.

4.3.4 18-8 Stainless Steel

The differentiation between the ϵ phase and γ twins is only possible in a few orientations. These have been determined by Kestenbach [120]. In the present work the $\langle 125 \rangle \gamma$ orientation was used to detect the ϵ phase, as it is the lowest order orientation which is not affected by reciprocal lattice streaks or double diffraction spots. Several low order orientations (e.g. $\langle 110 \rangle \gamma$, $\langle 122 \rangle \gamma$) were available for the unambiguous detection of γ twins.

In order to assist in the interpretation of the composite diffraction patterns produced by the phases and twins in this steel, a computer program was written to generate correctly scaled composite diffraction patterns from up to three separate phase orientations. Thus a postulated composite pattern would be generated by the computer and the negative of the actual pattern could be placed over it to check whether the diffraction spots coincided. This removed much of the tedium associated with the measurement of angles and distances in diffraction pattern determination.

Microstructure

The as-received microstructure consisted of a mixture of stacking faults, planar dislocation arrays, γ twins and

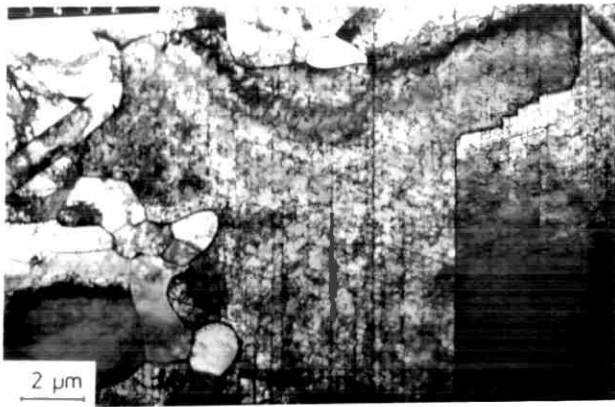
large α grains (Fig. 4.11 a). The α grains contained rows of untangled dislocations (Fig. 4.11 b). X-ray analysis showed no measurable ϵ content and ϵ plates were not detected by electron microscopy.

Hopkinson bar-loaded specimens showed a heavily-deformed structure, making the analysis of diffraction patterns very difficult at times. The alloy was heavily twinned and contained large quantities of α' . In a few areas it was possible to observe α' laths lying within $\{111\}\gamma$ bands or twins (Fig. 4.11 c). The γ twins were frequently bent and fragmented and occasionally occurred as plates within a $\{111\}\gamma$ band (Fig. 4.11 d - e). There was no electron microscopic or X-ray evidence of the ϵ phase.

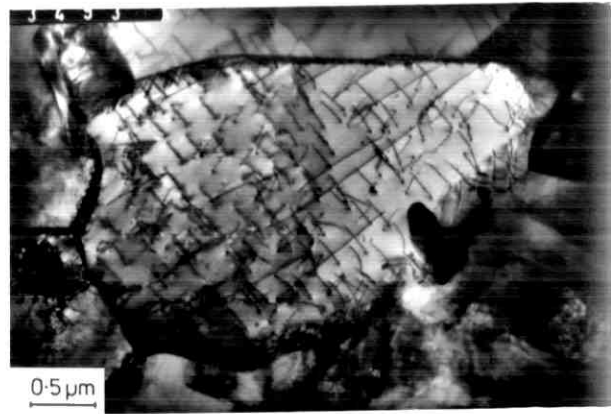
Shock loaded 18-8 steel generally contained a dense distribution of $\{111\}\gamma$ twins, and to a considerably lesser extent, ϵ plates on $\{111\}\gamma$ planes. α' laths of varying sizes were found at fault intersections in most grains. In some cases, the as-received α grains contained $\{112\}_{\alpha}$ twins.

It can be seen from Figures 4.12 a and 4.12 b that ϵ and twin plates are indistinguishable unless the diffraction pattern is viewed. They may even occur in the same grain running in the same direction (Figure 4.12 c).

α' crystals were formed at the intersections of:



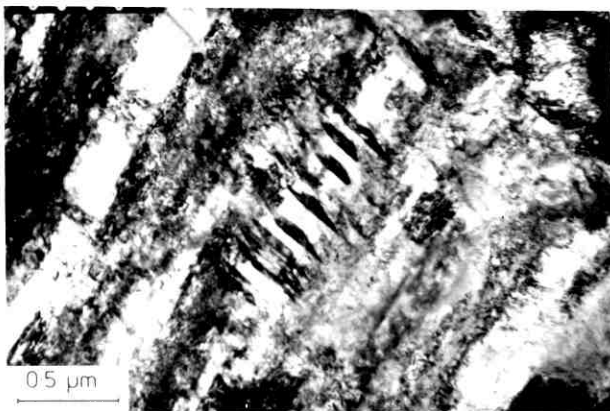
(a) Dislocation arrays and fine twins



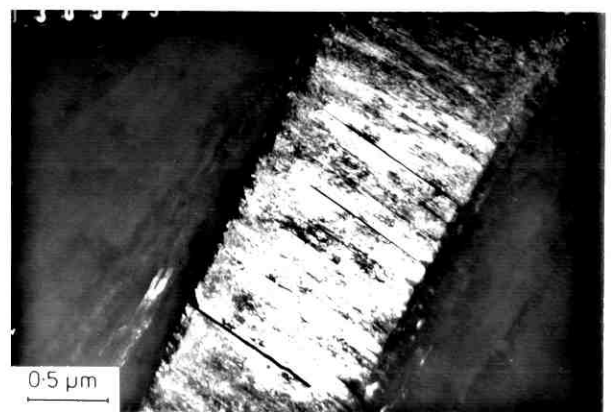
(b) α -grain



(c) α' laths in $\{111\}_{\gamma}$ bands



(d) Twins in $\{111\}_{\gamma}$ bands



(e) Band of adjacent twins (dark field)

Fig. 4.11 As-received (a,b) and Hopkinson bar loaded (c - e) 18/8 steel

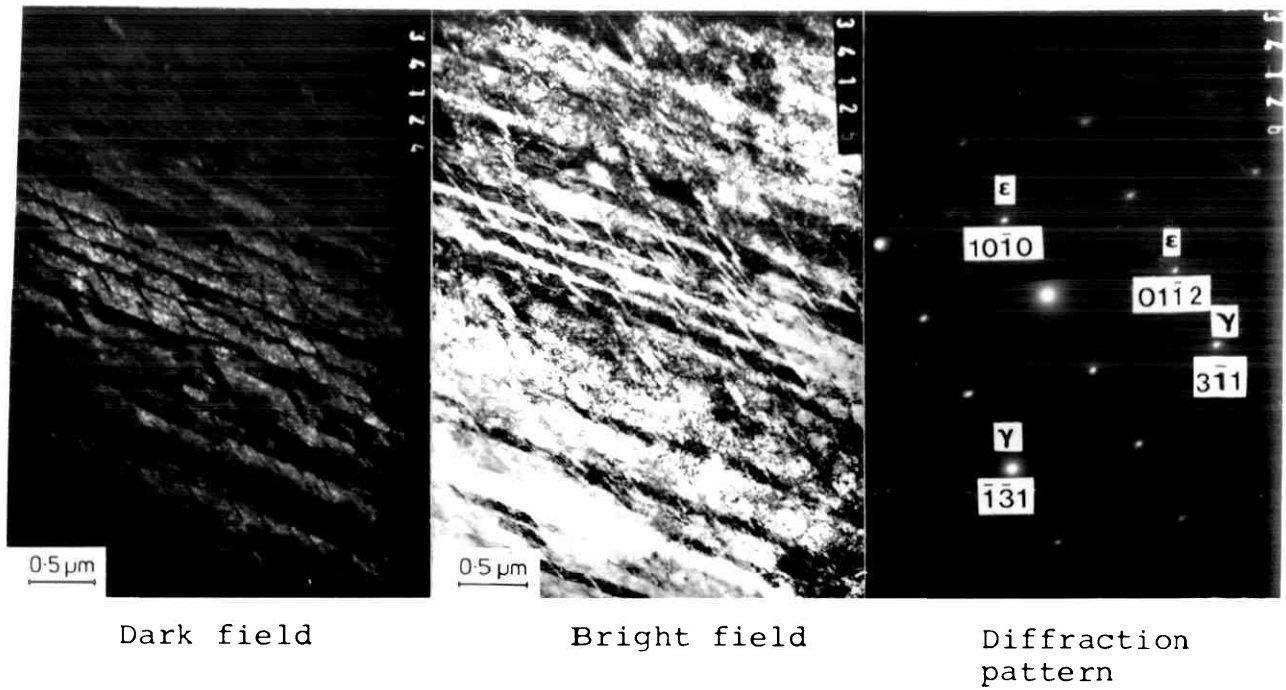


Fig. 4.12 a ϵ plates $[\bar{1}25]_{\gamma} // [\bar{2}4\bar{3}]_{\epsilon}$

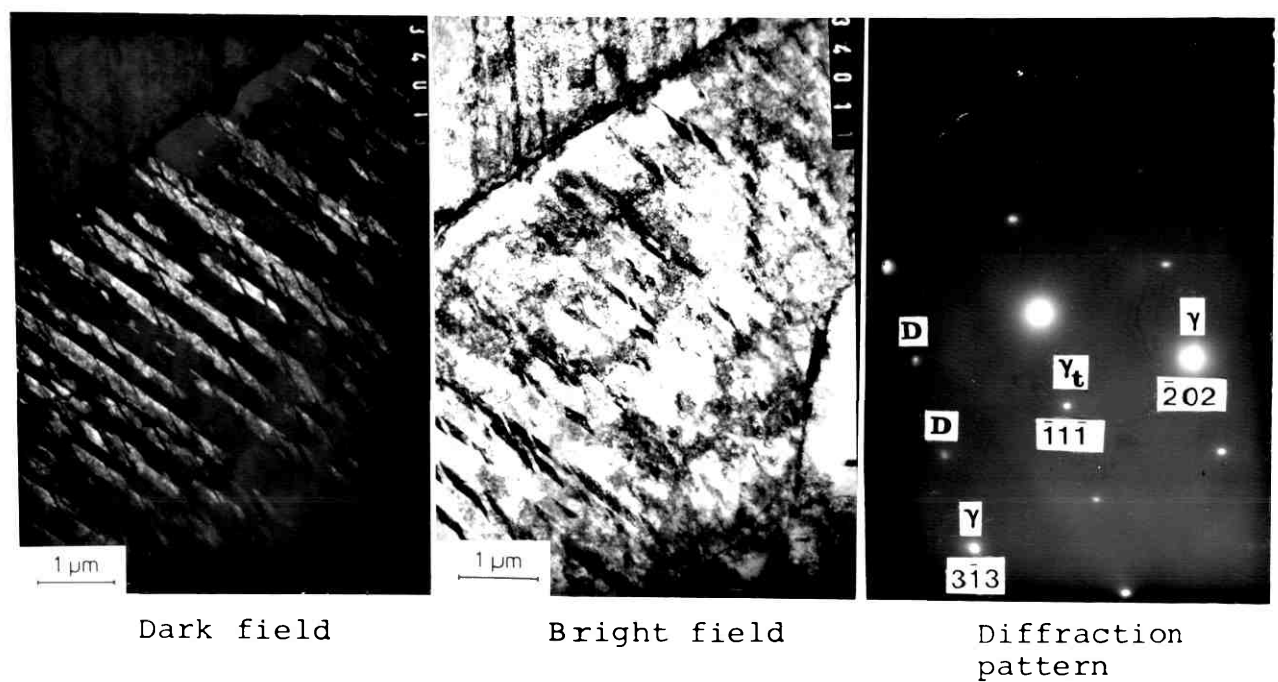
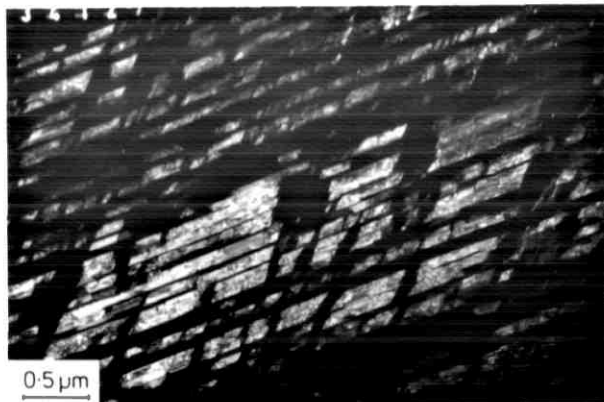
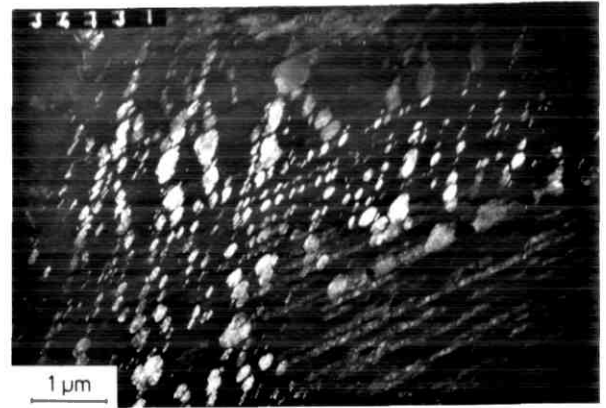
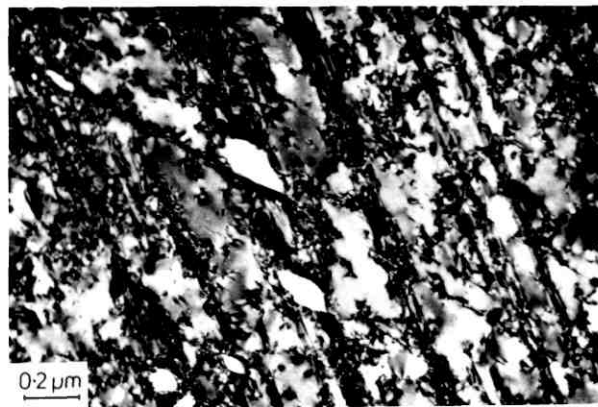
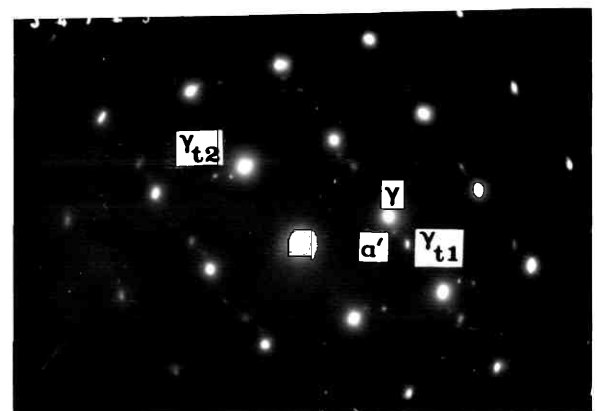
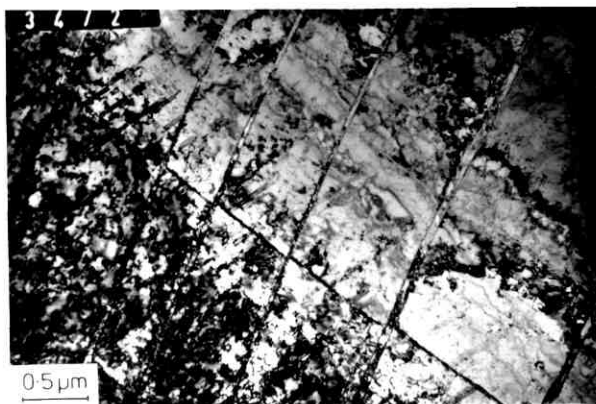


Fig. 4.12 b Twin plates $[161]_{\gamma} // [\bar{5}\bar{2}3]_t$



Twins


 ϵ plates (α' and ϵ spots overlap showing mixture of ϵ plates and α' grains)
Fig. 4.12 c Twins and ϵ plates in the same grainFig. 4.12 d α' crystals at twin - ϵ intersectionsFig. 4.12 e α' crystals at twin - twin intersections (K.-S. orientation)

(a) Twin and ϵ bands (Figure 4.12 d)

and

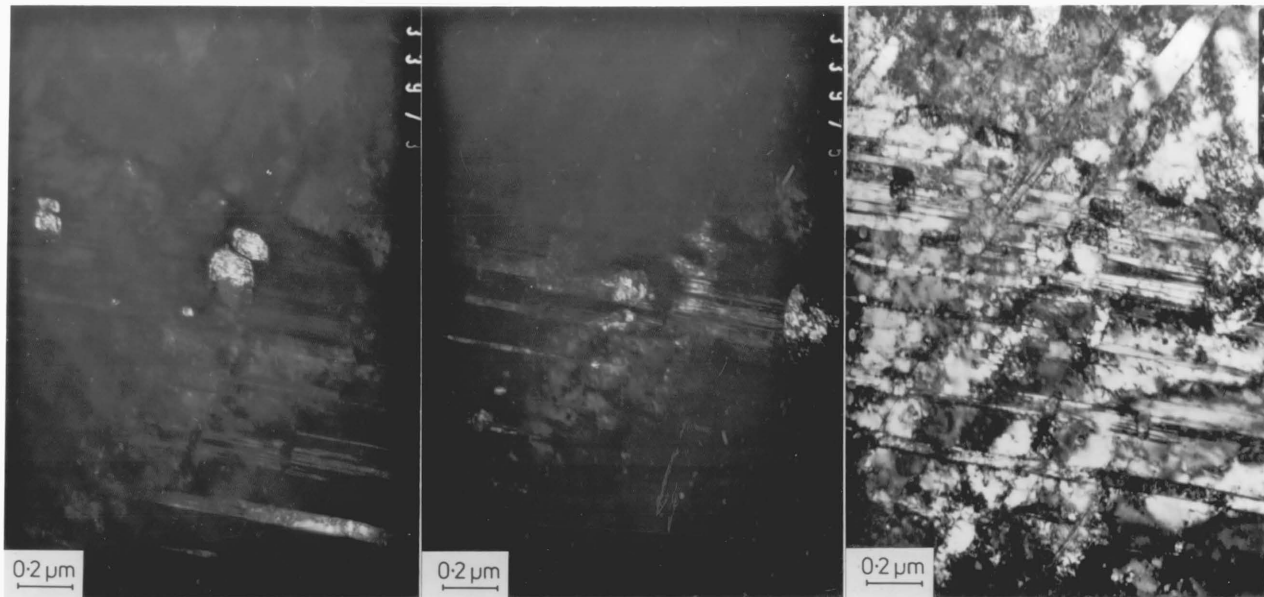
(b) Pairs of twin bands (Figure 4.12 e)

The α' crystals occurred in both the N.-W. and K.-S. orientations and sometimes on the same fault band (Fig. 4.13), but the K.-S. orientation appeared to be favoured by the more fully-developed crystals (Fig. 4.14 a).

It can be seen that small α' crystals are confined to one fault band, but that the larger crystals cross several faults and appear to have grown in a direction almost parallel with one of the intersecting bands. In some cases, these α' crystals grew in two $\{112\}_{\alpha'}$ twin-related directions (Fig. 4.14 b). There was no evidence of internal twinning in the α' grains.

After shock loading, the as-received α grains often contained short α twins twinned about $\{112\}_{\alpha}$ planes (Fig. 4.15 a). Shock loading with composition B and HMX produced, in addition to the short α twins, a "twin sheet" structure (Fig. 4.15 b). It is not certain whether these "sheets" were in the as-received α grains or caused by the meeting of large α' grains.

The baratol and composition B-loaded discs survived intact, whereas the HMX-loaded disc was quite heavily deformed. This enabled microstructural scans to be completed on all of the discs.



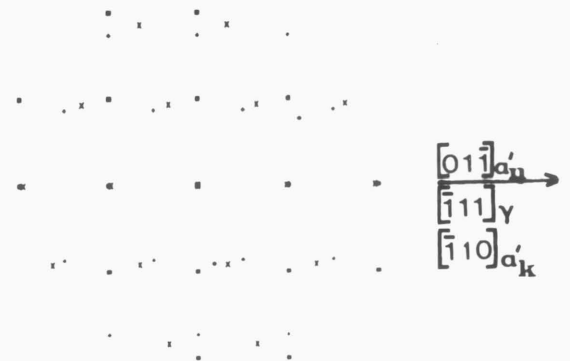
N.-W.

K.-S.

Bright field



α'_n 100 BEAM DIRECTION(■)
 γ 110 BEAM DIRECTION(X)
 α'_k 111 BEAM DIRECTION(+)



Diffraction pattern and computer plot (twin spots not shown) $[\bar{1}10]_{\gamma} // [100]_{\alpha'_{NW}} // [111]_{\alpha'_{KS}}$

Fig. 4.13 α' crystals in the N.-W. and K.-S. orientations

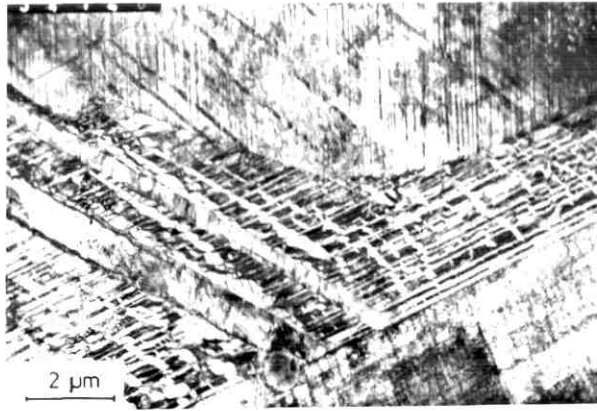
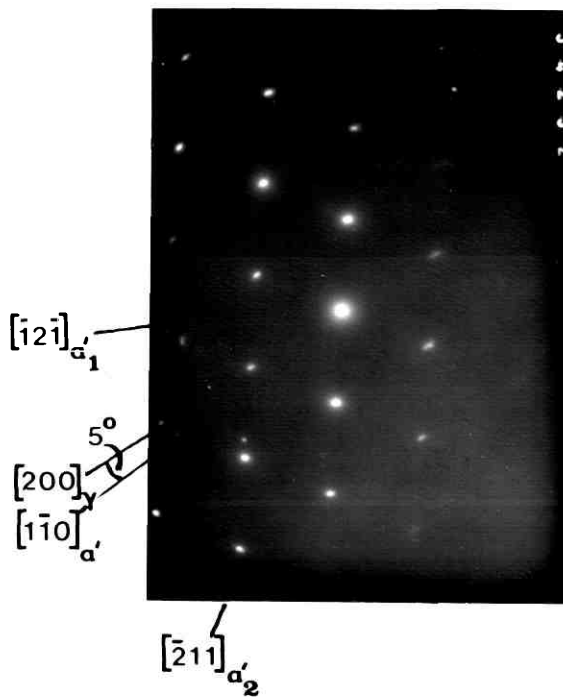


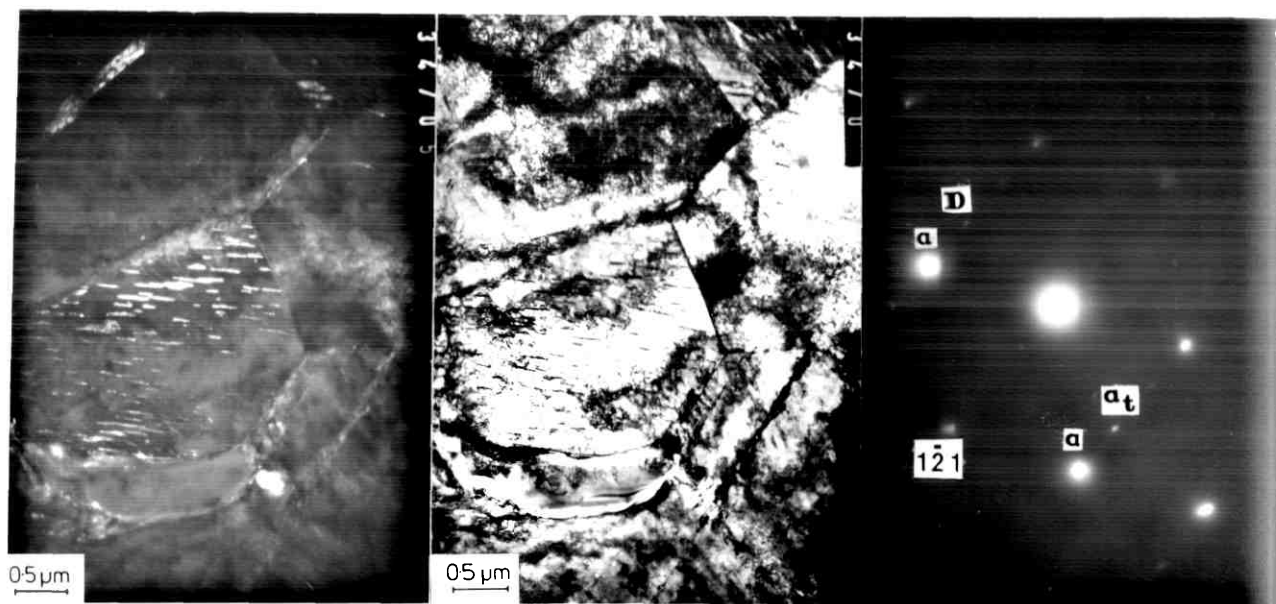
Fig. 4.14 a Large α' crystals in K.-S. orientation



$$(011)_{\gamma} // (111)_{\alpha'} // (\bar{1}\bar{1}\bar{1})_{\alpha'_t}$$

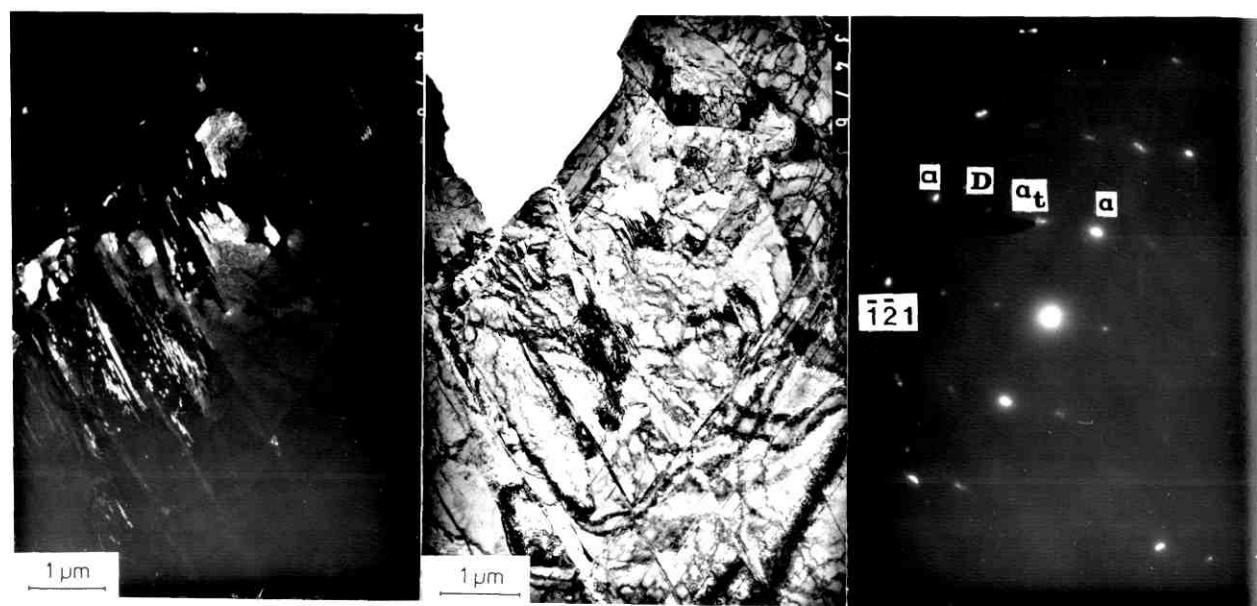


Fig. 4.14 b Twin related α' crystals $[\bar{1}2\bar{1}]_{\alpha'_1}$ direction reflects about $(11\bar{2})_{\alpha'}$ plane to produce $[\bar{2}11]_{\alpha'_2}$ direction



Dark field

Bright field

Diffraction
pattern
 $[01\bar{2}]_{\alpha} // [0\bar{1}2]_{\alpha_t}$
Fig. 4.15 a α twins, $(1\bar{2}1)_{\alpha}$ twin plane

Dark field

Bright field

Diffraction
pattern
 $[11\bar{3}]_{\alpha} // [\bar{1}1\bar{3}]_{\alpha_t}$
Fig. 4.15 b α twin "sheet" structure, $(\bar{1}2\bar{1})_{\alpha}$ twin plane

Baratol

The microstructure was heavily faulted but fault bending only occurred near to the free surface. The faults were usually γ twins. ϵ plates, when they were detected, were usually faint. Close to the free surface the α' phase occurred as small crystals lying within the fault bands. As the distance from the free surface increased the laths became larger and began to overlap neighbouring fault bands to produce elongated crystals. As the shocked surface was approached, the crystals appeared to form in two, twin-related directions. However, within 0.1 mm of the shocked surface the structure once again contained small α' crystals. The anomalous existence of overlapping α' crystals was observed within 0.1 mm of the free surface.

Twinned α grains only occurred near to the free surface.

Composition B

In this case the structure was heavily faulted and contained large amounts of α' . γ twins predominated, there being very little evidence of ϵ plates. The α' laths usually overlapped several fault bands and twin-related laths were frequently formed. Extensive twinning occurred in the as-received α grains and the "twin sheet" structure was also observed.

With the exception of the region near the shocked surface, there was little observable variation of microstructure across the disc. Near to the shocked surface some ϵ phase was detected and the α' laths lay within the fault bands.

HMX

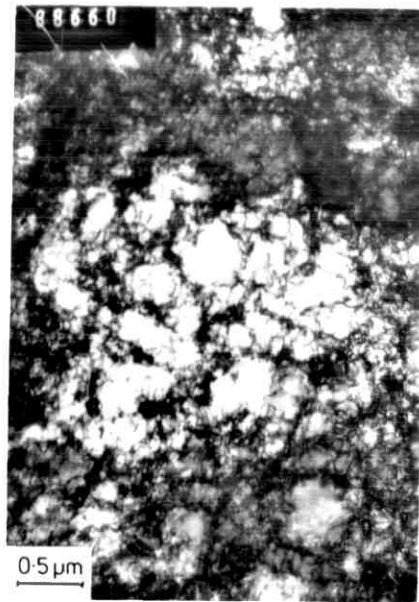
Due to the heavy deformation of the disc during recovery, the microstructural scan was probably not perpendicular to the original faces of the disc.

The structure consisted of large numbers of γ twins and occasional traces of ϵ . There was general evidence of gross deformation reminiscent of the Hopkinson bar specimens. In those grains where diffraction pattern analysis was possible, there appeared to be a considerable reduction in the size and number of α' laths compared to the composition B specimen. The γ grains were heavily twinned and only rarely contained α' laths. The as-received α grains were often twinned and the twin sheet structure was also observed.

4.3.5 Iron

Due to technical difficulties with the LMV electron microscope, it proved impossible to conduct a thorough investigation of Hopkinson bar and shock-loaded iron.

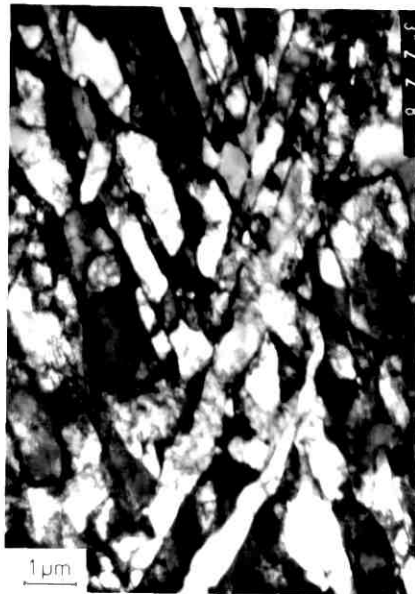
Thus the micrographs shown in Fig. 4.16 represent typical microstructures encountered after the four loading tests. It can be seen that the well-formed dislocation cells in the Hopkinson bar loaded iron (Fig. 4.16 a) are replaced by plates and elongated grains after composition B loading, and by a denser mixture of plates and elongated grains after HMX loading (Fig. 4.16 b,c). These are probably caused by the reverse $\epsilon \rightarrow \alpha'$ transformation. The microstructure in the transformed region of the baratol loaded iron also contained plates but these appeared to contain short fragmented twins (Fig. 4.16 d). Microstructures within the untransformed region contained isolated twins and no plates.



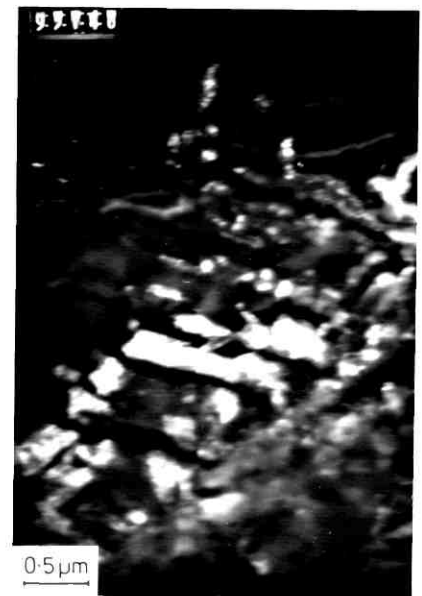
(a) Hopkinson
bar loaded



(b) Composition B
loaded



(c) HMX loaded



(d) Baratol loaded
fragmented twins
near to shocked
surface

Fig. 4.16 Typical iron structures

Table 4.1 Material Specifications and Compositions

<u>Material</u>	<u>Specification</u>	<u>Composition</u>
Aluminium	BS 1476 E1CM as rolled bar	99.0% Al
Duralumin	BS 1476 HE 15 TB Solution-treated naturally aged bar	94% Al, 4% Cu, 0.5% Mg, 0.5% Si, 1% Mn
Copper	BS 1433 tough pitch	99.9% Cu
18-8 Stainless Steel	EN 58 B	Fe, 17 - 20% Cr, 7 - 10% Ni, 0.045% P (max), 0.045% S (max), 2.00% Mn (max), 0.2% Si (max), 0.15% C (max), 0.6% Ti
Iron	REMCO magnetic	99.79% Fe, 0.025% Cu, 0.03% Si, 0.16% Mn

Table 4.2 Polishing Solutions and Conditions

<u>Material</u>	<u>Polishing Solution</u>	<u>Temp. (°C)</u>	<u>Current (mA)</u>
Aluminium	30% Nitric acid in methanol	- 40	100
Duralumin	30% Nitric acid in methanol	- 40	100
Copper	30% Nitric acid in methanol	- 40	100
18-8 Stainless Steel	5% Perchloric acid in acetic acid	15	100
Iron	5% Perchloric acid in acetic acid	15	100

Table 4.3 Hardness Values of as-received, Hopkinson Bar, and Shock-loaded Specimens

<u>Material</u>	<u>as-rec.</u>	<u>Hop. Bar</u>	<u>Baratol</u>	<u>Comp. B</u>	<u>HMX</u>
Aluminium	24	41	38	-	-
Duralumin	120	152	138 - 160	122	-
Copper	87	114	87 - 115	95	87
18-8 Stainless Steel	277	348	302	305	302
Iron	118	167	125, 187, 224	172	177

CHAPTER VDISCUSSION

Before discussing the individual results in Chapter IV, it is worth considering the possible effects of the passage of a shock wave through a metallic material.

The passage of a stable shock front implies a change of state which occurs within approximately 10^{-9} s. It seems reasonable to assume that the abrupt density change must be accompanied by the homogeneous generation of large numbers of dislocations which accommodate the mismatch between the shocked and unshocked medium.

If, as suggested by Meyers [118], the newly-generated dislocations achieve considerable velocities due to their mutual repulsion, they will also generate new dislocations but by the more usual multiplication methods. The narrow shock front will restrict the movement and generation of these dislocations.

Since the shock wave velocities lie between 5 and 8 km/s, the width of the shock front will be between 5 and 8 μm . Thus, with the exception of a region within 4 μm of the specimen free surface, all of the specimen discs will have been subjected to a shock pressure for some period of time.

Under the large shock pressures used, the material undergoes a change from one-dimensional to three-dimensional loading on the passage of the shock front. The pressure is hydrostatic and there appears to be no reason why dislocations should move during this constant pressure period. However, many of the point defects created by rapidly moving dislocations in the shock front would be annihilated at the high shock temperatures and pressures.

As previously indicated, high hydrostatic pressure results in high densities. These can be achieved by a reduction in the existing lattice dimensions or by a transformation to a higher density phase, but not by the creation of vacancies or dislocations, which lead to a decrease in density. Thus a reduction of the dislocation and point defect densities would occur during the high pressure part of the shock pulse. This would lead to a slight relaxation of the shock pressure.

The shock pressure is reduced to zero by the passage of the rarefaction wave. This is an envelope of the individual pressure release waves, travelling at sonic velocities which correspond to the existing material pressure. Thus all release wave velocities are sonic and considerable dislocation motion should take place. Maximum rarefaction times of between 0.1 and 0.2 μ s would occur in the specimens tested in this work. The rarefaction times vary linearly between zero at the

specimen free surface and a maximum at the shocked surface. Thus the greatest dislocation motion should occur at the shocked surface.

These rapidly moving dislocations would be responsible for the generation of large numbers of excess point defects and dislocations whose concentration after unloading would depend on the adiabatic release temperature (T_a).

The final microstructure of a shocked material will only reflect the loading conditions if the adiabatic release temperature does not cause a significant change in the microstructure. As Table 3.14 shows, these temperatures can be a considerable fraction of the melting point.

McElroy and Szkopiak [37] have divided the processes occurring during the annealing of cold worked metals into six groups:-

1. Low Temperature Recovery ($<0.2 T_m$)

Large-scale elimination and redistribution of deformation-induced point defects by vacancy-interstitial recombination, clustering and defect migration to dislocations, causing loops, stacking faults and helical dislocations.

2. Intermediate Temperature Recovery ($0.2 T_m - 0.3 T_m$)

Minor dislocation rearrangements; cell growth, reduction in thickness and raggedness of cell walls, due to annihilation by glide and cross-slip.

3. High Temperature Recovery ($<0.4 T_m$)

Large scale rearrangements by climb and cross-slip, formation of low angle grain boundaries, polygonization.

4. Recrystallization ($>0.4 T_m$)

5. Grain Growth ($>0.5 T_m$)

6. Secondary Recrystallization ($>0.8 T_m$)

In the case of aluminium and duralumin, room temperature is above $0.3 T_m$ and thus considerable recovery must have occurred between loading and examination, although the precipitates in duralumin probably reduced this effect. Similarly, copper would have undergone some intermediate temperature recovery.

When comparing the values of T_a/T_m for the shocked and Hopkinson bar-loaded specimens with the above classification, it must be noted that McElroy and Szkopiak's review considers annealing times of an hour or more. The

shocked specimens were quenched from T_a to the water tank temperature within a millisecond of loading, and the Hopkinson bar specimens probably took only a few minutes to cool to room temperature. However, in the shocked case, the driving forces for recovery and recrystallization are probably considerably larger than those due to conventional loading.

The electron and optical micrographs of baratol-loaded aluminium (Fig. 4.7 b) and the heavily deformed free surfaces of copper and iron loaded by composition B and HMX (Figs. 4.5 c, 4.3) show recrystallization. The extreme deformation occurring during specimen deceleration is probably responsible for this, although the recrystallization time is unknown. This is supported to some extent by the discovery of recrystallized grains in Hopkinson bar loaded aluminium and copper (Figs. 4.7 a, 4.9 f). This problem of the time of recrystallization is highlighted by the work of Cohen et al [122] which showed that shock-loaded copper, which was left for several months at room temperature, underwent isolated recrystallization.

The absence of recrystallization in Hopkinson bar loaded iron can be attributed to the low value of T_a/T_m .

In conclusion, it can be said that the observed microstructure of pure aluminium, and duralumin and copper at the

higher explosive pressures, were probably not those which would have been observed immediately after Hopkinson bar or shock loading. They were the result of subsequent recovery or recrystallization caused by the large numbers of point defects and dislocations generated during shock loading, the extreme deformation undergone during deceleration, or the high deformation in Hopkinson bar loading.

The 18-8 stainless steel, iron, baratol-loaded copper and to a lesser extent the baratol-loaded duralumin specimen microstructures were probably very similar to those existing immediately after loading.

The consideration of the residual microstructures after Hopkinson bar and shock loading can be divided into three parts:- (1) dislocation structure, (2) twin formation and (3) phase transformations

5.1 Dislocation Structure

The results for copper and duralumin shocked with baratol indicate that the hardness increase from the free surface to the shocked surface is accompanied by a corresponding increase in dislocation density.

Since the shock rise time and the shock pressure remain constant as the shock wave traverses the specimen, the

dislocation density must depend on either the pulse length or the rarefaction time. As previously indicated, an increased pulse length would tend to favour recovery and thus a hardness reduction as the shocked surface is neared. Thus it seems reasonable to attribute the increasing hardness and dislocation density to the increasing rarefaction time.

All the dislocations can move within the rarefaction tail and the distance that they move will depend on the rarefaction time. Thus in a single obstacle-free crystal dislocations could move distances varying from zero at the free surface to a few hundred microns at the shocked surface, if one assumes an average dislocation velocity of 1 km/s. In the case of baratol-loaded copper and duralumin, the maximum distance would be about 150 μm .

Obviously, a typical dislocation would never travel this distance in these heavily-dislocated, polycrystalline specimens. However, the longer the rarefaction time, the greater would be the number of new dislocations and point defects produced by sources caused by the interaction of rapidly moving dislocations with obstacles such as other dislocations, grain boundaries, precipitates or twins.

The electron micrographs of baratol-loaded copper show incomplete cells, and dislocation loops and tangles. The incomplete cells have been attributed previously to the lack of sufficient time (pulse length) for their formation. Certainly, the 1 μ s maximum pulse length in these experiments is at the low end of the pulse range investigated by Appleton and Waddington [50] and Murr and Kuhlmann-Wilsdorf [123] for shocked copper and nickel respectively.

However, even if one assumes a pulse length long enough to permit the dislocations created in the shock front to form a cell structure, there seems little reason to suppose that this would survive the considerable high-speed motion and generation of dislocations occurring during the unloading, rarefaction tail. Thus, the completeness of the dislocation cells should depend on the rarefaction time and, sometimes more importantly, on the value and duration of the isentropic release temperature and the subsequent room temperature behaviour of the microstructure. If cell formation does occur during rarefaction, it is likely to take place at low rarefaction rates (low pressures) when dislocations are moving relatively slowly and few new dislocations are being generated.

In conclusion, the incomplete cells found in baratol-loaded copper may be the result of the short rarefaction

time (0.16 μ s) and the low isentropic release temperature (0.25 T_m). However, in baratol-loaded duralumin considerable recovery and recrystallization should have occurred since the isentropic release temperature (0.44 T_m) and the room temperature (0.33 T_m) are large fractions of the melting point. The hardness values suggest that this has not occurred, which probably reflects the effect of the large precipitate concentration.

The anomalous microstructures at the free surfaces of the copper and duralumin specimens (well-formed dislocation-free cells in copper and bundles of shear bands in duralumin) were caused by the high strains and high local temperatures occurring during specimen deceleration. This is supported by the occurrence of large numbers of shear bands observed in the heavily-deformed fragments of composition B loaded duralumin.

The dislocation loops and tangles in copper and the twin-like features in duralumin in regions close to the shocked surface may have been the result of the von Neumann spike. In the specimens this would take the form of a shock front rising well above the stable shock pressure. This would be rapidly reduced to the shock pressure by a subsequent rarefaction tail and thus only a small region close to the shocked surface would be affected. Micrographs of composition B and HMX loaded copper show similar features in uncrystallized regions. This tends to support the above explanation, although

the effect of the longer rarefaction tail and the higher isentropic release temperatures should not be forgotten.

Finally, if one compares the hardness and the dislocation structure of the heavily deformed Hopkinson bar loaded copper with that of the lightly deformed baratol loaded copper, it can be seen that, at longer explosive pulse lengths, the hardnesses are similar. However, the dislocation structures are different. The Hopkinson bar copper contains dislocation free, equiaxed, complete cells whereas the baratol-loaded copper contains incomplete cells with few dislocation-free areas.

This can be explained in terms of the loading pulse and the final release temperatures. In Hopkinson bar loading the pulse front lasts for about 0.5 ms. Thus inertial effects can be ignored and loading is at constant volume. The slow rise time does not require the homogeneous generation of dislocations, there being sufficient time for dislocations to move and multiply by more conventional means and for microscopic grain distortion to occur. Also after maximum load is reached there is sufficient time during the constant load period (0.5 ms) for dislocations to form into complete cells. The unloading wave will have far less effect on the microstructure than the shock rarefaction wave and the high adiabatic release temperature cooling slowly to room temperature would permit some recovery.

In the shocked case the high speed dislocations interact to produce large numbers of point defects, jogs and tangles. Thus subsequent dislocation realignment is more difficult and occurs at room temperature since the specimen is rapidly quenched in the water tank.

It can be seen that although similar hardness values occur in specimens subjected to long pulse shock loading and Hopkinson bar loading, the source of the hardnesses are different. In Hopkinson bar specimens hardening is caused by well-formed dislocation cells and grain size reduction in the loading direction, whereas point defects, dislocation jogs, loops and tangles are responsible for the shock-loaded hardness.

5.2 Twin Formation

Deformation twins in b.c.c. and high stacking fault energy f.c.c. metals and alloys tend to be long and thin, usually running between grain boundaries. Twins in low stacking fault energy metals and alloys are short and often terminate within grains. The reason is thought to be the relative ease of twin nucleation and the velocity of the twinning dislocations when they encounter forest dislocations in their path.

In the first case, the driving force on the twinning dislocation rapidly accelerates it so that propagation

through forest dislocations is possible and twins only stop at grain boundaries. In the second case, nucleation is easy and dislocation acceleration is low, thus enabling forest dislocations to halt twinning dislocations [124, 125, 126].

It can be seen from Chapter II that two methods of twin generation have been proposed. These are the pole mechanism and the homogeneous nucleation mechanism.

As previously mentioned, dislocations mostly move and multiply during the shock front and the rarefaction tail. Since these times are within the possible twin formation times, it is interesting to compare the theoretical formation times of twins by the pole mechanism and homogeneous nucleation with the rise and fall times of the shock front and the rarefaction tail.

If one considers a twin of thickness h and length L , the formation times t_t are as follows:-

(1) Pole Mechanism

- (a) Straight dislocation of length $L/2$ sweeping out equal areas during each revolution.

$$t_t = \frac{h\pi L}{dv}$$

where d is the habit plane spacing and v the dislocation velocity [82]

- (b) Dislocation takes on the shape of an Archimedes spiral which forms a twin whose oblateness (h/L) depends on the applied stress [127].

$$t_t = \frac{2\pi L}{v}$$

(2) Homogeneous Nucleation

Large numbers of overlapping dislocation loops generated at the same time grow to radius $L/2$.

$$t_t = \frac{L}{2v}$$

These are minimum times and assume that the twin nucleus was at a point half way along the final twin length. Obviously, if twin growth is halted in any direction by an obstacle, the final value of L will no longer reflect the formation time.

The micrographs of iron and copper show that the twins formed are indeed long and thin, relatively infrequent,

and stopped only by grain boundaries and sometimes, in the case of iron, by other twins. However, in γ stainless steel long unbroken twins are also seen to completely traverse grains although most of the twins are short and often imperfect. The twins in α stainless steel appear to begin and end within the α grains, but apparently not at intersections with other twins (cf. γ stainless steel).

The theoretical twin formation times for copper, iron and γ and α phase stainless steel are given along with the typical twin lengths and widths in Table 5.1. If these are compared with the rise time of the shock pulse ($0.001 \mu\text{s}$) and the fall time of the rarefaction tail ($0.1 - 0.2 \mu\text{s}$ maximum) some general observations can be made.

Firstly, the straight dislocation model (a) could not apply to copper, iron or γ -stainless steel twins as the formation times are too long.

Secondly, if twins were formed during the shock front they would most likely be the result of homogeneous nucleation and their concentration would be independent of their position in the specimens. However, the copper and iron twinning times appear to be too long even for homogeneous nucleation and thus it is possible that no shock front twins form in these cases.

Thirdly, the spiral pole mechanism could only operate successfully during the rarefaction tail (with the possible exception of the α -stainless steel). In this case the average twin length might be expected to increase with distance from the specimen free surface. One might then expect to see a preference for short twins running between grain boundaries nearer to the free surface and a mixture of short and longer twins nearer to the shocked surface.

This does, however, pose the question as to what happens to twins which do not reach the grain boundaries due to lack of time. In the absence of an applied force, the twin would attempt to contract, the degree of contraction being dependent on the obstacles in its path. The apparent absence of short twins starting and / or ending within copper or iron grains suggests that either complete untwinning occurs (this would seem unlikely as considerable dislocation motion must have occurred inside and outside the twinned region), or else the pole mechanism is not responsible for twinning.

This conclusion cannot be made in the case of γ twinning in stainless steel since twins are terminated not only by grain boundaries but also by other twins or bands within the grains. Thus twinning could have been caused by the pole mechanism, although the plentiful supply of stacking faults seems to favour an overlapping mechanism.

α twinning in stainless steel provides an interesting comparison with twinning in pure iron. The α twins are short, thin and of reasonably constant length. They do not appear to start or end on other twin or slip planes (c.f. γ twins). Thus their lengths are determined either by the available twinning time or by dislocation tangles which, in contrast to those in pure iron, are capable of stopping twinning dislocations. In the former case, it would then be reasonable to assume that twinning must have occurred within the shock front by means of a pole mechanism, as the α twins would have been considerably longer had they been formed in the rarefaction tail or by homogeneous nucleation. If the latter possibility is correct, one can only assume that the dislocation density in the α grains is considerably higher than that in pure iron and / or the nucleation stresses are much lower (although one would still expect long twins due to the very high dislocation velocities occurring during shock).

Finally, consideration should be given to the twin nucleation stress. If this forms a substantial fraction of the final load, the available twin propagation time would be considerably reduced. Thus twins would be shorter. However, this is not the case in copper, which twins at Hopkinson bar loads, or in γ stainless steel.

In pure iron, twinning does not occur at Hopkinson bar loads but does occur at a 13 GPa shock pressure (P1). The calculated position of the interaction between the reflected elastic wave and the P1 shock front corresponds almost exactly to the onset of twinning in baratol-loaded iron. Thus, twinning appears to nucleate during rarefaction from 13 GPa to about 12.3 GPa, giving considerable scope for twin growth during the remainder of the rarefaction tail.

The possible occurrence of twins near to the shocked surface of baratol-loaded duralumin may be due to the high stresses which are caused by the von Neumann spike and which exceed the twinning stress.

In conclusion, homogeneous twin nucleation appears to be more likely in copper and γ stainless steel (either in shock front or rarefaction tail) and in iron (in rarefaction tail), whereas the pole mechanism appears more likely in α stainless steel (in shock front).

5.3 Phase Transformation

Both iron and stainless steel undergo pressure-induced phase transformations. This is evidenced by the hardness step in baratol-loaded iron and the existence of α' and ϵ phases in the residual microstructure of the stainless steel.

5.3.1 Iron

In iron, the high pressure ϵ phase transforms back to α' -martensite on unloading and hence its presence can only be detected by the effect of the double shock front on both the microhardness and the etching characteristics. This was only observed at the lowest shock pressure (25 GPa, baratol). At the higher loading pressures (41 GPa, comp. B and 55 GPa, HMX) the P2 transformation wave travels faster than the P1 wave, thus generating only one shock front (P2). This is confirmed by the absence of hardness steps in these specimens.

In the baratol-loaded specimen the value of U_{P1}/U_{P2} determined from the position of the interaction between the reflected P1 wave (R1) and the oncoming P2 wave lies somewhat below the curve of Smith and Fowler [59]; see Fig. 4.6 c. This suggests that the actual P2 pressure may have been as low as 21 GPa, although Smith and Fowler's results do show considerable scatter.

The hardness measurements show a second, more gradual rise in a region between 1.4 mm and 2.0 mm from the free surface. Calculations show that the reflected P3 wave (R3) catches up with the R2 rarefaction wave in this region. The leading part of the reflected wave overtakes the R2-wave at about 1.4 mm whereas the trailing part overtakes it at about 2.0 mm. This leads to an increase in the pressure drop between 1.4 mm and 2.0 mm

from the free surface (Fig. 5.1). Now the pressure of the P3 wave is greater than 8.1 GPa which means that the $\epsilon \rightarrow \alpha'$ transformation does not occur until the R3 rarefaction wave passes through the transformed iron.

On reflection from the free surface, the R3 wave provides a gradual rarefaction to the atmospheric pressure. However, when it reaches the high pressure ϵ -phase at 0.55 mm from the free surface it causes the $\epsilon \rightarrow \alpha'$ phase change. The wave shape changes to a rarefaction shock from the P3 pressure to 4.5 GPa followed by a gradual rarefaction to the atmospheric pressure. This transformation shock will produce a hardness increase which remains constant until the R3 wave catches up to the R2 wave at 1.4 mm from the free surface. From this point until 2.0 mm from the free surface, the rarefaction shock increases from about 4 GPa to about 20 GPa. Thus, the driving forces for the transformation will be increased considerably and hence the deformation and hardness will increase. Beyond 2.0 mm the rarefaction shock is constant and no further hardness increase occurs.

Finally, the hardnesses of the composition B and HMX-loaded specimens are less than that of the transformed baratol-loaded specimen, reflecting probably some degree of recovery. The plates visible in the electron micrographs of the composition B and HMX-loaded iron are probably caused by the $\epsilon \rightarrow \alpha'$ martensite transformation.

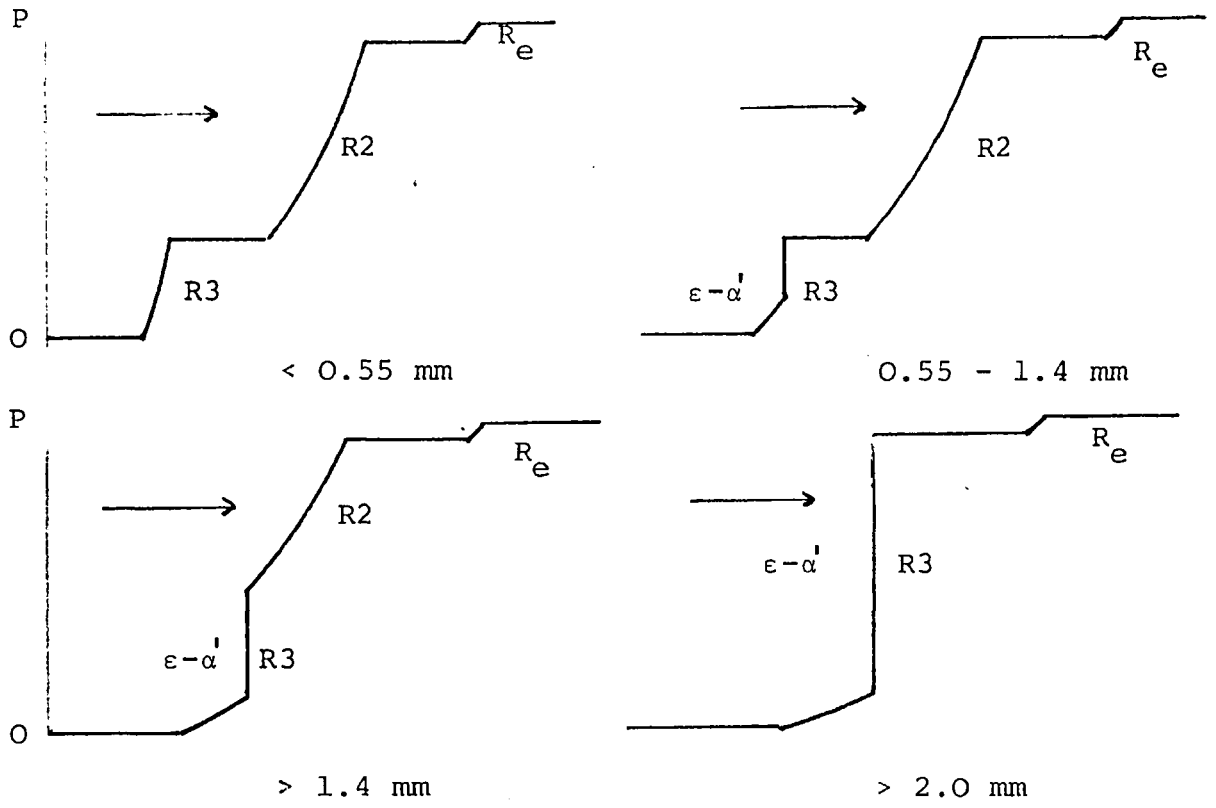


Fig. 5.1 Tensile shock interaction

5.3.2 18-8 Stainless Steel

The absence of any hardness steps in the explosively loaded specimens indicates the absence of a two-wave structure at these pressures. This was also noted by Fowler et al. [128] in their shock tests on commercial 18-8 steel containing up to 40% α phase. However, two points should be mentioned. Firstly, ϵ plates have been detected under conventional, low strain rate loading conditions indicating a very low transformation stress relative to that in iron. Secondly, the density change for the $\gamma \rightarrow \epsilon$ transformation in 18-8 steel is about - 0.81% compared to an estimated - 5.9% for the $\alpha \rightarrow \epsilon$ transformation in iron [67,109]. Thus the break in the Hugoniot curve, and hence the pressure range for a possible double shock formation, would both be considerably less than those for iron. Therefore it is not surprising that a two-wave structure is absent at the pressures used here.

The absence of the ϵ phase in Hopkinson bar loaded specimens ($\epsilon = 0.52$) is in agreement with the results of Bowkett and Harries [67] who showed that the ϵ phase ceases to exist in EN 58 B cold-rolled to greater than 50% deformation.

The scarcity of ϵ plates in shock-loaded 18-8 stainless steel may be due to the reported tendency of ϵ to revert

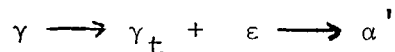
back to γ at temperatures above 150°C [67]. This temperature is exceeded (albeit for a short time) at these pressures and, in conjunction with the expected $\epsilon \rightarrow \alpha'$ transformation and the relatively high stacking fault energy, may explain the absence of ϵ plates.

The ϵ and / or γ_t to α' transformation is thought to occur by stress or pressure relief. Under the uniaxial loads provided by the Hopkinson bar, the α' phase can form by the continuous relief of highly-stressed $\epsilon - \epsilon$ and $\epsilon - \gamma_t$ intersections. Thus α' could occur at any time during the load pulse. In the shock case, the pressure is hydrostatic and thus the $\gamma_t, \epsilon \rightarrow \alpha'$ transformation will occur during the rarefaction tail. As the results show, the amount of transformation product occurring in shocked specimens is considerably less than that in Hopkinson bar specimens, reflecting the short shock pulse time.

Since α' formation results from the movement of transformation dislocations which only occurs during the rarefaction tail, it might be expected that the length or size of the α' laths would reflect the length of the rarefaction tail at any point. This appears to be the case in baratol-loaded 18-8 steel but is not obvious at composition B or HMX pressures. Due to the widely varying shapes of these α' laths and insufficient knowledge concerning their growth mechanism, no attempt was

made to make a detailed correlation between lath length and rarefaction tail length. However, some type of pole mechanism would seem to be responsible for α' growth, as a rough calculation shows that an overlap mechanism would require less than 10^{-9} s to generate the longest α' laths. Thus either the α' lath growth is limited by obstructions such as dislocations, twins or other grains (which is not supported by the findings of this work or by those of Murr and co-workers [87,90]), or else lath growth occurs by some mechanism which slows the growth time to the 10^{-7} s available within the rarefaction tail (e.g. pole mechanism).

In conclusion, it can be said that the absence of a double shock wave structure may be due to the low $\gamma \rightarrow \epsilon$ transformation pressure and the small density change involved. Thus at the shock pressures employed, the P2 wave moves faster than the P1 wave. The α' phase appears to form within the rarefaction tail and the size of the α' laths shows some dependence on the length of the rarefaction tail. The transformation sequence is probably:-



During
shock
front

During rarefaction
tail

Table 5.1 Measured Twin Dimensions and their possible
Formation Times (dislocation velocity $v = 3\text{km/s}$)

	<u>Fe</u>	<u>Cu</u>	<u>γ-SS</u>	<u>α-SS</u>
L (μm)	60	100	0.6 - 6	0.3
h (μm)	0.25	0.03	0.003 - 0.08	0.005
No. of atomic planes	2000	150	15 - 400	45
L/2v (μs)	0.01	0.02	0.001 - 0.01	0.00005
$2\pi L/v$ (μs)	0.1	0.2	0.01 - 0.10	0.0006
$h\pi L/dv$ (μs)	140	15	0.1 - 25	0.015

CHAPTER VICONCLUSIONS AND SUGGESTIONS FOR FUTURE WORK6.1 Conclusions

Unless great care is taken to ensure that shock-loaded materials are recovered undamaged, any microstructural effects of shock loading will have been either masked or completely altered by the heavy deformation caused by spalling and deceleration. Aluminium, and copper and iron shocked by composition B and HMX, underwent varying amounts of recrystallization in heavily deformed regions. The amount of grain refinement might reflect shock loading but a comparison between this and the recrystallization caused by Hopkinson bar loading and high temperatures (burning aircraft) would be required.

In the case of metals and alloys which have not been heavily deformed after shock loading, the following conclusions can be made:-

- 1) With the exception of shock and adiabatic temperature recovery effects, any microstructural changes occur within the shock front and the rarefaction tail.
- 2) The hardness of the metals and alloys not undergoing recrystallization or phase

transformations increases with the length of the rarefaction tail.

- 3) The size of the α' crystals in 18Cr-8Ni steel increases with the length of the rarefaction tail.
- 4) Three wave interactions occur in iron shocked within the double shock wave region.
 - (a) Twinning occurs when the reflected elastic wave interacts with the first (plastic) wave.
 - (b) An abrupt hardness increase occurs when the reflected first (plastic) wave interacts with the second (transformation) wave.
 - (c) Interaction between the low-pressure α and the high pressure ϵ rarefaction waves causes a gradually increasing rarefaction shock which results in a further hardness rise.
- 5) In shocked 18Cr-8Ni steel α' crystals can be formed at the intersections of twins and ϵ plates or other twins. The α' crystals occur in both the Kurdjumov-Sachs and the

Nishyama-Wasserman orientations, the two orientations sometimes being observed in the same fault band. The Kurdjumov-Sachs orientation is favoured by larger crystals which can also grow in twin-related directions.

- 6) Twinning was observed in shock-loaded copper, iron, γ and α 18Cr-8Ni steel and possibly in duralumin. It is suggested that a homogeneous twin nucleation mechanism is responsible for twinning in copper and γ 18Cr-8Ni steel (in shock front or rarefaction tail), and iron (in rarefaction tail). In α 18Cr-8Ni steel however, twinning may occur by means of a pole mechanism during the shock front.

Comparison of the Hopkinson bar and shock-loaded metals and alloys shows the effect of heavy deformation on the microstructure and hardness. With the exception of iron, Hopkinson bar hardnesses are greater than or equal to the maximum shock-loaded hardnesses and the Hopkinson bar loaded microstructures would mask any shock-loaded microstructures. The existence of martensitic plates in shocked iron and the high hardness caused by the rarefaction shock provide a good shock loading indicator.

6.2 Suggestions for Future Work

The wide scope of this research has of necessity left several areas where more detailed investigation would be desirable. Some possibilities are given below.

- 1) Electron microscopical studies into the effects of the double shock wave on baratol-loaded iron.
- 2) Quantitative investigations into the effects of the rarefaction tail on the dislocation density in shock-loaded copper, and on the α' particle size in 18Cr-8Ni steel.
- 3) Determination of the twin nucleation mechanisms in copper, iron and α 18Cr-8Ni steel by studying the effect of the grain size and the length of the rarefaction tail on their twinning behaviour.

APPENDIX ABASIC SHOCK WAVE THEORY

The theory below will consider the effects of a longitudinal plane wave in which all the displacements of material particles take place in the direction of the wave propagation. This is the simplest example to deal with, and can be easily achieved if the specimen thickness is considerably less than its diameter.

A.1 The Generation of a Shock Front [8, 12]

Consider an element of volume of length dx and of unit cross-section. In one-dimensional flow the time rate of increase of mass in this element must equal the net flux of mass into the volume. This can be expressed by the equation

$$\frac{\delta \rho}{\delta t} dx = \rho u - \left[\rho u + \frac{\delta (\rho u)}{\delta x} dx \right] \quad (A1)$$

where ρ and u are the density and mass velocity respectively, and x and t are space and time variables.

In a perfect fluid, the forces on a mass element arise from the pressure gradient and possible body forces such as gravity. In this case, only the pressure gradient is

of importance. Hence

$$P - \left[P + \frac{\delta P}{\delta x} dx \right] = \rho dx \frac{du}{dt} \text{ where } P \text{ is the pressure} \quad (A2)$$

$$\text{But } du = \frac{\delta u}{\delta x} dx + \frac{\delta u}{\delta t} dt \quad \text{when } u = f(x, t) \quad (A3)$$

$$\therefore \frac{du}{dt} = \frac{\delta u}{\delta x} \frac{dx}{dt} + \frac{\delta u}{\delta t} \quad (A4)$$

Let subscripts denote partial differentiation.

$$\text{Thus (A1) can be written as } \rho_t + u\rho_x + \rho u_x = 0 \quad (A5)$$

and after substituting (A4) into (A2) the result can be written as

$$\rho u_t + \rho u u_x = - P_x \quad (A6)$$

If heat conduction is neglected the entropy S_o of each element remains constant. Thus the pressure can be given by the isentropic equation of state

$$P = f(\rho, S_o) \quad (A7)$$

$$\text{Differentiating gives } dP = \left(\frac{\delta P}{\delta \rho} \right)_{S_o} d\rho + \left(\frac{\delta P}{\delta S_o} \right)_{\rho} dS_o \quad (A8)$$

where $dS_o = 0$.

$$\text{Now } \rho = f(x, t) \quad \therefore d\rho = \left(\frac{\delta \rho}{\delta x} \right)_t dx + \left(\frac{\delta \rho}{\delta t} \right)_x dt \quad (\text{A9})$$

Substituting into (A8) gives

$$dP = \frac{\delta P}{\delta \rho} \left(\frac{\delta \rho}{\delta x} dx + \frac{\delta \rho}{\delta t} dt \right) \quad (\text{A10})$$

Since dx and dt are independent

$$\frac{\delta P}{\delta x} = \frac{\delta P}{\delta \rho} \frac{\delta \rho}{\delta x} \quad \text{or } P_x = (P_\rho)_{S_0} \rho_x \quad (\text{A11})$$

If equation (A11) is substituted into (A6) and the result divided by the sound speed given by

$$c(\rho) = \left(\frac{\delta P}{\delta \rho} \right)_{S_0}^{\frac{1}{2}} \quad (\text{A12})$$

this can be added to, or subtracted from, equation (A1) to give

$$\rho_t + (u + c)\rho_x + \frac{\rho}{c} [u_t + (u + c)u_x] = 0 \quad (\text{A13})$$

$$\text{and } \rho_t + (u - c)\rho_x - \frac{\rho}{c} [u_t + (u - c)u_x] = 0 \quad (\text{A14})$$

Now in general:

$$\phi_t + \frac{dx}{dt} \phi_x = \frac{d\phi}{dt} \quad \text{if } \phi = f(x, t) \quad (\text{A15})$$

∴ Equations (A13) and (A14) yield the four characteristic equations governing the flow. These are:

$$d\rho + \frac{\rho}{c} du = 0 \quad \text{along} \quad \frac{dx}{dt} = u + c \quad \text{in the } +x \text{ direction} \quad (\text{A16})$$

$$\text{and } d\rho - \frac{\rho}{c} du = 0 \quad \text{along} \quad \frac{dx}{dt} = u - c \quad \text{in the } -x \text{ direction} \quad (\text{A17})$$

The equations in $\frac{dx}{dt}$ are the equations of sonic disturbance propagating in the $+x$ and $-x$ directions respectively.

Consider a semi-infinite homogeneous mass at pressure P_0 , density ρ_0 , and zero particle velocity, which is subjected to a continuously increasing pressure P_1 . This is equivalent to a piston accelerating into the mass. Initially $u = 0$ and $c = c(\rho_0)$, but after the first infinitesimal disturbance travels into the medium, the density increases. Thus the next disturbance travels at a velocity $c = c(\rho_0 + \Delta\rho)$, which is added to the increase in u due to the accelerating piston. For a

"normal equation of state" c and u increase with density, and thus $\frac{dx}{dt}$ increases in the $+x$ direction. Figure (A1) shows a series of characteristics issuing from the piston. This increase in $\frac{dx}{dt}$ implies that the characteristics will cross at some point. Since more than one value of u and ρ cannot exist for one value of x and t , a discontinuity occurs. This is the shock front. The pressure gradient increases with time, until the characteristics cross. At this point, effects such as fluid viscosity, heat conduction and thermal non-equilibrium begin to play an important part, and deviations from complete discontinuity depend on these. However, for practical purposes, the shock can be regarded as a discontinuity.

If this shock front is to remain, it must move with a velocity greater than the sound speed in the medium ahead of it. Otherwise disturbances created at the front would precede it and lead to a reduction of the pressure gradient across it.

A.2 The Conservation Equations for a Shock Wave [8,9]

Consider a shock front moving at velocity U_s through a medium originally at pressure P_0 , density ρ_0 and at a mass velocity u_0 . After the shock has passed the pressure, density and mass velocity are P_1 , ρ_1 and u_1 respectively. (Fig. A2)

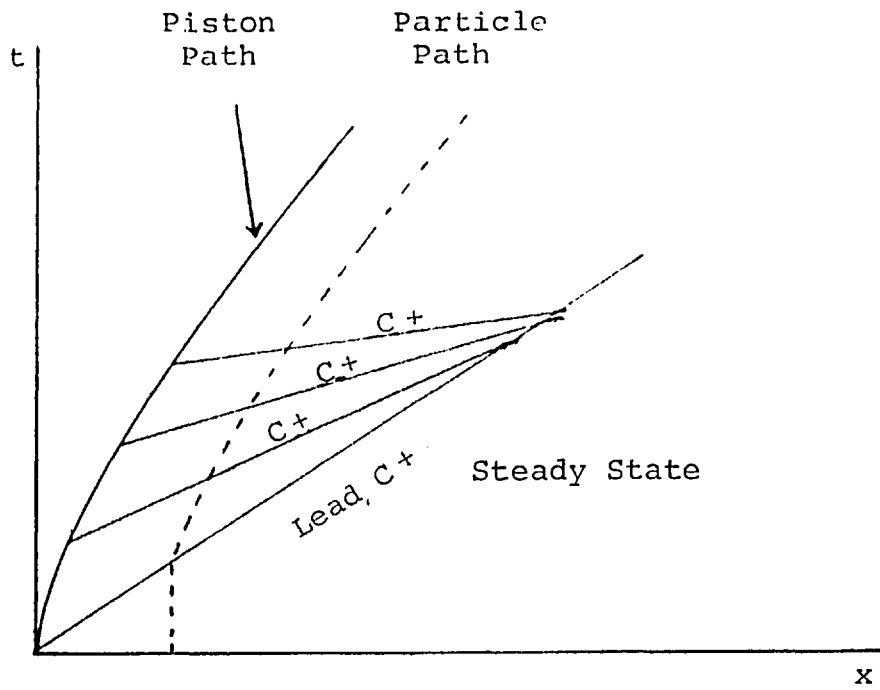


Fig. A.1 Distance-time plot of flow in a simple compression wave

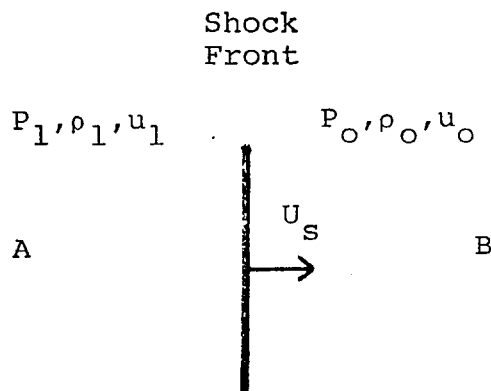


Fig. A.2 Motion of a shock front into an uncompressed medium

As before, the mass flux into and out of the shock-front must be equal. If the coordinate system moves with the shock front at velocity U_s , then in this system, mass flows into the shock front from the uncompressed region at velocity U_s , and out of the shock front at velocity $U_s - u_1$. Thus:

$$\rho_0(U_s - u_0) = \rho_1(U_s - u_1) \quad (A18)$$

The net force on a unit cross-section of material between A and B is $P_1 - P_0$. This is equal to the rate of change of momentum with time.

Thus:

$$P_1 - P_0 = \rho_0(U_s - u_0)(u_1 - u_0) \quad (A19)$$

Finally the power input to a unit cross-section of material between A and B must equal the rate of change of energy with time for the enclosed material.

Therefore:

$$P_1 U_1 - P_0 u_0 = \frac{\rho_0}{2} (U_s - u_0)(u_1^2 - u_0^2) + \rho_0(U_s - u_0)(E_1 - E_0) \quad (A20)$$

where E is the specific internal energy

Combining equations (A18) and (A19) gives:

$$u_s - u_o = v_o \left[\frac{p_1 - p_o}{v_o - v_1} \right]^{\frac{1}{2}} \quad \text{where } v = 1/\rho \quad (\text{A21})$$

$$\text{and } u_1 - u_o = [(p_1 - p_o)(v_o - v_1)]^{\frac{1}{2}} \quad (\text{A22})$$

and substituting (A18) and (A19) into (A20) gives:

$$E_1 - E_o = \frac{1}{2}(p_1 + p_o)(v_o - v_1) \quad (\text{A23})$$

Equations (A21), (A22) and (A23) are the Rankine-Hugoniot equations and express the conservation relations for a shock-wave.

In the case where a shock wave is passing through an uncompressed medium, p_o and v_o are known and $u_o = 0$. Thus if u_s and u_1 can be measured $E_1 - E_o$, p_1 and v_1 can be deduced.

A.3 The Stability of a Shock Wave [8,13]

The shock wave velocity increases with pressure for materials which have Hugoniot curves along which $dp/dv < 0$ and $d^2p/dv^2 > 0$. These conditions are fulfilled by most materials. However two important exceptions exist. They are the elastic-plastic wave structures and the two-wave structures associated with certain phase changes.

Consider two compression shocks moving through a medium. The first shock connects P_0, V_0 to P_1, V_1 and moves with a velocity

$$U_{s1} - u_1 = V_1 \left[\frac{P_1 - P_0}{V_0 - V_1} \right]^{\frac{1}{2}} \quad \text{relative to the state} \quad (A24)$$

P_1, V_1 . The second moves with a velocity

$$U_{s2} = V_1 \left[\frac{P_2 - P_1}{V_1 - V_2} \right]^{\frac{1}{2}} \quad \text{relative to the state } P_1, V_1 \quad (A25)$$

and separates P_2, V_2 from P_1, V_1 .

A two-wave structure exists if:

$$U_{s1} - u_1 > U_{s2} \quad (A26)$$

$$\text{or } \frac{P_1 - P_0}{V_0 - V_1} > \frac{P_2 - P_1}{V_1 - V_2} \quad (A27)$$

It can be seen from equation (A27) that at some pressure P_2 the two wave structure will revert to a single shock wave structure. This point is shown in Figure A3.

A.4 The Motion of a Rarefaction Wave [14,15]

Consider a semi-infinite homogeneous mass at pressure P_1 , density ρ_1 and zero particle velocity, which is subjected to a decrease of pressure to P_0 . Equations (A16) and (A17) still apply, but in this case the pressure reduction could be caused by a piston accelerating in the $-x$ direction. The first disturbance propagates at $\frac{dx}{dt} = c(\rho_1)$ since at $x = 0$, $t = 0$ and $u = 0$. Typical characteristics are shown in Figure A.4. The characteristics C^- from equations (A17) all initiate at $t = 0$, $u = 0$, $\rho = \rho_1$ and intersect the C^+ characteristics given by equations (A16). Integrating along $dx/dt = u - c$ gives:

$$u = \int_{\rho_1}^{\rho} \frac{c d\rho}{\rho} \quad (A28)$$

Since any point x, t can be reached by a C^- characteristic initiating at $t = 0$, equation (A28) gives the particle velocity at this point. Now the density ρ decreases with pressure, and thus u and c also decrease. Thus the C^+ characteristics are a series of diverging straight lines originating at the piston surface. The pressure thus decreases with dx/dt , until the surface pressure P_0 is reached. The C^+ characteristic where this occurs is given by

$$\frac{dx}{dt} = \int_{\rho_1}^{\rho_0} \frac{c d\rho}{\rho} + \left[\frac{\delta P}{\delta \rho} \right]_{S, \rho = \rho_0}^{\frac{1}{2}} \quad (A29)$$

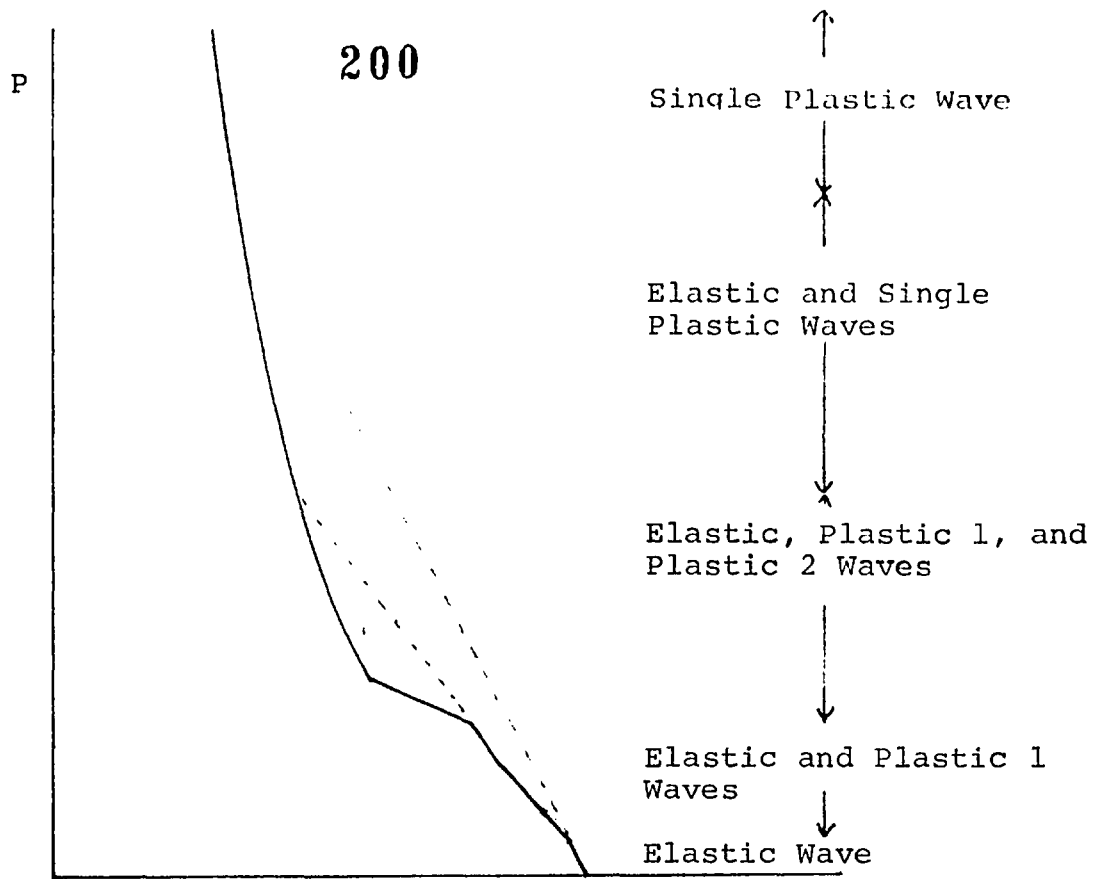


Fig. A.3 Hugoniot Curve

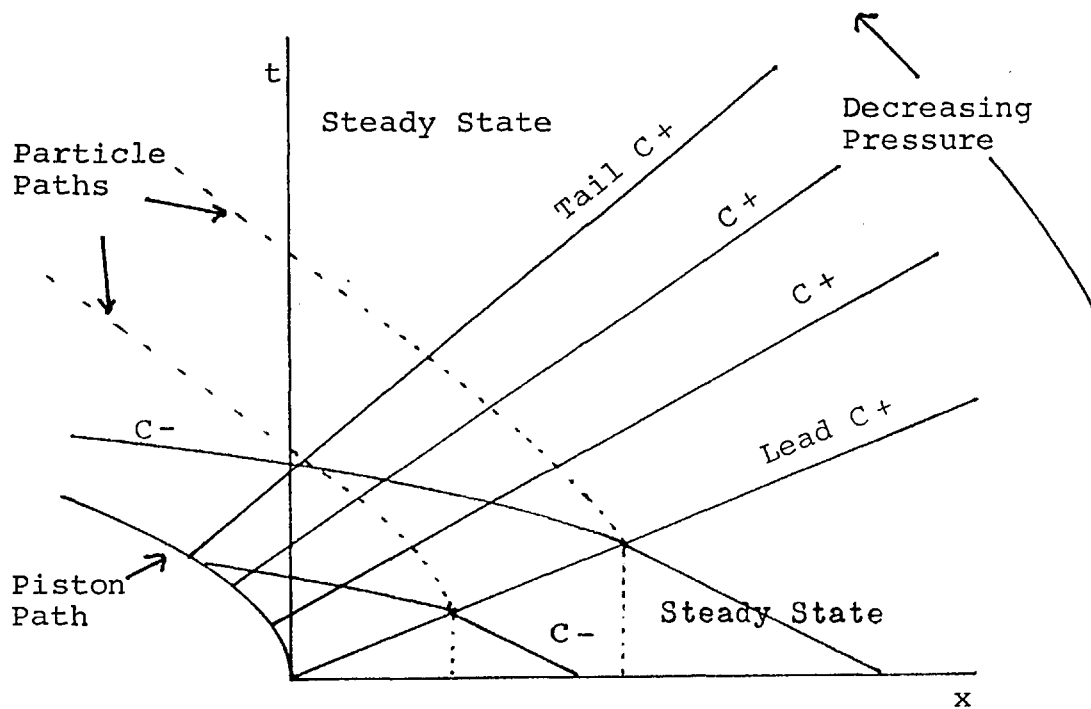


Fig. A.4 Distance-time plot of flow in a simple rarefaction wave

and the value of u_1 is given by

$$u_1 = \int_{\rho_1}^{\rho_0} \frac{c d\rho}{\rho} \quad (\text{A30})$$

This is the Riemann integral.

Since dx/dt decreases with density and pressure, the pressure gradient at the front of the rarefaction wave decreases, and hence a shock front is not formed. Moreover, a compressive shock wave is reflected at a free surface as a series of rarefaction waves. These reduce the pressure gradually to the free surface pressure, and thus the shock front spreads out (Fig.A4).

A.5 The Free Surface Velocity u_{fs} [14,15]

If a plane shock wave, connecting state P_0, V_0 and P_1, V_1 , reaches a free surface which is parallel to the shock front, the shock pressure P_1 is reduced to zero by a rarefaction wave from the free surface. The free surface velocity imparted is the sum of the particle velocity due to the shock wave (u_p) and the particle velocity due to the rarefaction wave (u_r). Therefore,

$$u_{fs} = u_p + u_r = \left[(P_1 - P_0) (V_0 - V_1) \right]^{\frac{1}{2}} + \int_{\rho_1}^{\rho_0} \frac{c d\rho}{\rho} \quad (\text{A31})$$

$$\text{or } u_{fs} = [(P_1 - P_0)(V_0 - V_1)]^{\frac{1}{2}} + \int_{P_0}^{P_1} \left[- \frac{\delta V}{\delta P} \right]_S^{\frac{1}{2}} dP \quad (A32)$$

Up to a pressure of about 50 GPa it has been shown that $u_p \approx u_r$

Thus, in most cases,

$$u_{fs} \approx 2u_p \quad (A33)$$

Hence, the free surface velocity can be related to the particle velocity, and if the shock velocity U_s is known, the remaining variables can be deduced from the Hugoniot equations (A21), (A22) and (A23).

APPENDIX BTHE MEASUREMENT OF THE FREE SURFACE AND SHOCK VELOCITIESB.1 Free Surface Velocity Measurement using a Fabry-Perot Interferometer

The system is based on an idea by Laharrague, Durand, and Le Bihan [26,27]. As previously indicated in 3.1.7, light reflected off a moving surface undergoes a wavelength change $\Delta\lambda$ given by:-

$$\frac{\Delta\lambda}{\lambda} = -\frac{2V}{c} \quad (B1)$$

When a monochromatic laser light beam thus reflected enters a Fabry - Perot interferometer it undergoes successive reflections (Fig. B1). Each time that the beam strikes the back surface a small percentage of the light is transmitted. These transmitted beams are focussed by a lens (L) to form circular interference fringes. Any change of wavelength causes the fringe diameter to alter.

B.1.1 Laser

If the interferometer is to be set up correctly for the range of free surface velocities which are expected

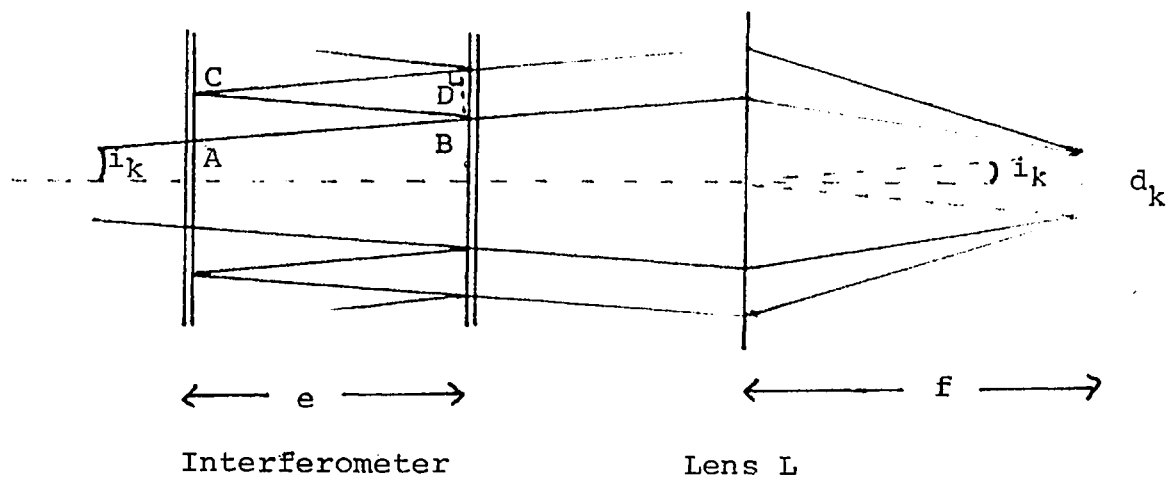


Fig. B.1 Fabry-Perot interferometer

the laser beam characteristics must be determined.

It was hoped to use a Coherent CR-2, 2 Watt argon ion laser ($\lambda = 5145\overset{\circ}{\text{\AA}}$, beam diameter 1.2mm, beam divergence 0.6 mrad., cavity length 1.15 m, line power 800 mW).

The transverse TEM₀₀ laser mode is used as Gaussian distribution is ensured in all directions perpendicular to the beam direction. However this beam has additional longitudinal harmonic modes in the beam direction and a few of these are emitted with the main beam frequency. The mode wavelength separation (free spectral range) is given by $\lambda^2/2L$ where L is the laser cavity length. Thus to prevent harmonic fringes from forming in the interferometer the bandpass of the interferometer must be made less than the free spectral range of the laser beam (FSR_L).

B.1.2. Interferometer

The interferometer must be calibrated for the expected free surface velocity range. Thus the maximum and minimum wavelength changes ($\Delta\lambda_{\text{max}}$, $\Delta\lambda_{\text{min}}$) must be calculated from equation (B1).

Now the Fabry - Perot mirror separation (e) is given by:-

$$e = \frac{\lambda cn}{4V}$$

where n is the number of static fringes crossed by one dynamic fringe during a free surface acceleration to velocity V .

Using values of $V_{\max} = 4 \text{ mm}/\mu\text{s}$, $e = 100 \text{ mm}$, $c = 3 \times 10^8 \text{ m/s}$ and $\lambda = 5145 \text{ \AA}$ gives

$$\Delta\lambda_{\max} \approx 137 \text{ m\AA} \quad \text{and } n \approx 10$$

The FSR of the interferometer is expressed as

$\text{FSR}_{\text{FP}} = F\Delta\lambda_{\text{FP}}$ where the Finesse

$$F = \frac{\pi R}{1 - R} \quad , \quad R \text{ is the reflectivity of the partially}$$

transmitting mirror and $\Delta\lambda_{\text{FP}}$ is the interferometer band pass.

The FSR_{FP} equals $\Delta\lambda_{\max}/n$ and $\Delta\lambda_{\text{FP}}$ must be less than FSR_{L} ($1 \text{ m\AA}$). Thus the Finesse should be greater than 12. In practice values of 50 or more are used to give sharp fringes.

B.1.3 Lens System (Fig. B.2)

(a) Collimator

The necessarily large path length between the laser and the interferometer requires that the beam divergence be

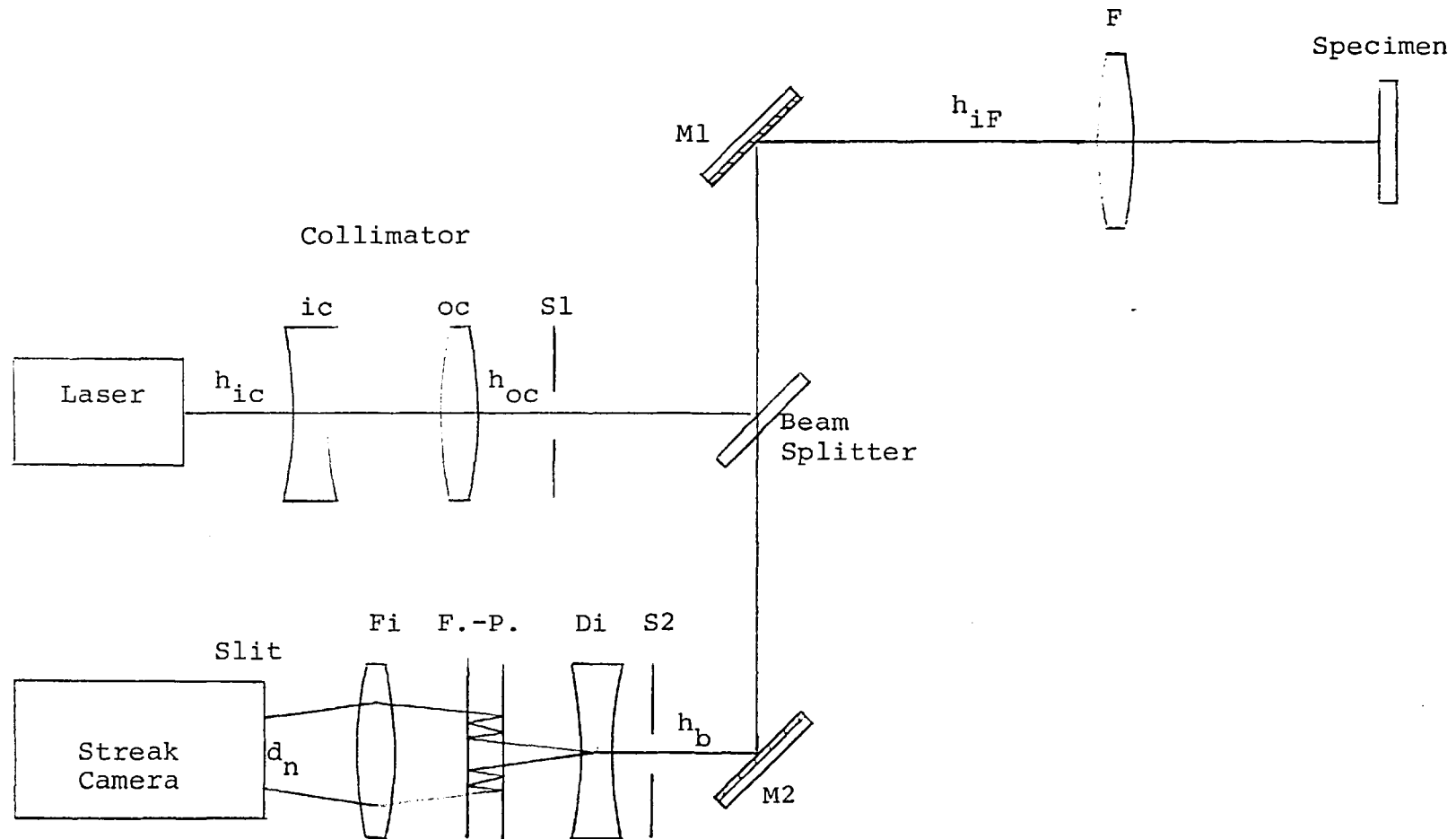


Fig. B.2 Diagram of proposed apparatus for the measurement of the specimen free surface velocity

as little as possible. This is achieved by a divergent collimator at the laser output.

An incoming beam of diverging angle ω_i and radius h_{ic} is made parallel and expanded to a radius h_{oc} where

$$\frac{h_{oc}}{h_{ic}} = \frac{f_{oc}}{f_{ic}} + \omega_i \quad \text{where } f_{oc} \text{ and } f_{ic} \text{ are the focal lengths}$$

of the output and input lenses respectively.

If there is a positional error (dx) between the two collimating lenses an output divergence ω_{oc} occurs where

$$\omega_{oc} = \frac{h_{oc}}{f_{oc}} - \frac{h_{oc}}{f_{oc} + dx} \approx \frac{h_{oc}}{f_{oc}^2} dx$$

This results in an increased beam radius of $L\omega_{oc}$ where L is the distance between the collimator and the focussing lens (F).

(b) Specimen Focussing Lens (F)

When the specimen begins to move the reflected beam can be affected in two ways.

Firstly - rotation of the specimen by ω_t will cause a sideways shift $dh_F \approx 2f_F\omega_t$ of the beam.

Secondly - a motion of the specimen towards the lens will cause a divergence of the reflected beam given by:-

$$\omega_{OF} \approx \frac{h_{iF}}{f_F^2} V dt$$

(c) Diverging Lens before Interferometer (D_i)

In order to produce n fringes from the interferometer the diverging angle (θ) of the beam must be such that:-

$$\cos \theta > \frac{n\lambda}{2e}$$

For small θ this gives:-

$$\theta_n \approx \left[\frac{n\lambda}{e} \right]^{\frac{1}{2}} \approx 7 \text{ mrad.}$$

Thus the lens should have a focal length

$$f_{Di} < \frac{h_b}{\theta_n} \text{ where } h_b \text{ is the beam radius}$$

and equals $h_{OC} + l\omega_{OC} + dh_F + l\omega_{OF}$

where l is the distance between the interferometer and the specimen.

(d) Focussing Lens after Interferometer (Fi)

The transmitted light from the interferometer is focussed to form fringes whose diameters d_n are given by:-

$$d_n = 2f_{Fi}\theta_n$$

A thin slit is placed at the focal plane of this lens and the behaviour of the ring diameter is recorded by a streak camera.

(f) Stop Apertures

Due to the inevitable divergence and deviation of the beam stop apertures are placed at the collimator output (S_1) and before the interferometer diverging lens (S_2) in order to ensure a centered beam of known diameter.

B.1.4 Determination of the Free Surface Velocity

When the specimen is stationary fringes of diameter d_k are formed where k is the wave number. The velocity of a moving specimen which produces fringes of diameter d' is given by:-

$$V = \frac{c}{4} \left[\frac{n\lambda}{e} + \frac{d'^2 - d_k^2}{4f_{Fi}} \right]$$

where n is the difference in wave number between a fictitious dynamic fringe d'_{k+n} and a static fringe d_k of the same diameter.

If n cannot be determined by inspection of the streak trace (i.e. in the case of a shock front) it can be calculated using a known approximate value of V .

Alternatively the approximate velocity can be determined experimentally by reducing the Fabry - Perot mirror separation until a change of wave number less than one occurs during acceleration to velocity V .

B2. Shock Wave Velocity Measurement using a Selenium Switch

Selenium is non-conducting at pressures below 12.8GPa and becomes metallic above this pressure. Cole and Lyle [130] have constructed a switch using this principle. The use of vapour deposition can produce a switch less than 50 μm thick.

It was proposed to deposit a similar switch onto the shocked surfaces of the specimen plates. When a shock wave reaches the switch a capacitor discharges through it to give a trigger pulse. The arrival of the shock front at the free surface is recorded by the interferometer. After correcting for delays the shock wave velocity can be determined.

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