Unfortunately, Schubert *et al*[5] in reporting the Ti_3Si phase do not state the interstitial contents of their alloys. We further note that the lattice parameters of the tetragonal M_5Si_3 phases are similar in magnitude to that of the present silicide. Thus we speculate that a low carbon content and/or a difference in carbon distribution may be a factor in stabilizing the present tetragonal silicide, which conceivably could have a chemistry of $(Ti, V)_5Si_3$.

Summarizing, both X-ray and electron diffraction observations indicate the presence of a tetragonal silicide which we conclude is an equilibrium phase in Ti-30 to 40V-1Si alloys aged at ~600°C. This silicide probably has a chemistry of $(Ti, V)_3Si$ but conceivably could be a tetragonal $(Ti, V)_5Si_3$ phase.

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Discussion of "Residual Strength of Shock Loaded RMI 38644"*

MARC A. MEYERS

Rack[1] recently published an article describing the results of a detailed investigation into the effect of shock-loading pressure on the mechanical properties and microstructure of the titanium alloy RMI 38644. Among other interesting results, Rack[1] clarified the residual strength decrease for shock-loading pressures below 7 GN/m^2 (70 Kbar). Seven different pressures, ranging from 2.1 to 26 GN/m^2 (21 to 260 Kbar), were used and the driver plate thickness (4.8 mm) was kept constant throughout the experiments. The pulse duration was estimated to be 1.8 μs, but no indication was given in the paper as to what expressions were used to calculate it.

The objective of this note is threefold: a) to show that the pulse duration varied significantly throughout Rack's[1] experiments since it depends not only on the driver plate thickness, but also on other shock-wave parameters; b) to succinctly present data on the effect of pulse duration on residual properties and microstructure reported in the literature; and c) to suggest, based on evidence substantiating the importance of pulse duration, that the pulse duration variation in Rack's[1] experiments has an effect on the residual tensile properties.

a) Calculated Pulse Durations for Rack's Experiments. The pulse duration at the driver plate-target interface is the time interval between the moment of impact and the instant at which the pulse that went into the driver plate arrives again at the interface (after being reflected at the free surface of the driver plate). Due to the irreversibility of the process, the shock wave undergoes attenuation as it traverses the system. Initially, there is a decrease in pulse duration. Once the duration is reduced to zero, the peak shock pressure starts decreasing. This problem has been treated by Fowles,[2] assuming hydrodynamic behavior in both driver plate and target. By combining Eqs. [11] and [12] of Fowles,[2] it is possible to determine the pulse duration at the driver-target interface. Making use of the same hydrodynamic assumptions, Orava and Wittman[3] present an expression taking into account the decrease of pulse duration due to the existence of a protective cover plate. The pulse duration is given by (Eq. [10b] of Orava and Wittman[3]):

$$t_p = d^D\left[\frac{1}{U_S^D} + \frac{\rho_o^D}{\rho^D c^D}\right] + d^C\left[\frac{\rho_o^C}{\rho^C c^C} - \frac{1}{U_S^C}\right]$$

where d is the plate thickness; U_S is the shock-wave velocity; c, the sound velocity in the compressed state; ρ_o, the initial density; and ρ, the density in the compressed state. The superscripts D and C refer to the driver and cover plate, respectively. Since the cover plate was fairly thin in Racks'[1] experiments (as compared to the driver plate) little attenuation is to be expected from it. So, the second term in the above expression could be neglected without substantial error. Nevertheless, it was included in the computations. The various parameters were obtained from plots drawn from the tables presented by Kinslow.[4] For the cover plate, the calculations were made for pure titanium. Table I shows the values of the different parameters at the pressures used by Rack.[1] The resultant pulse durations are plotted in Fig. 1, and it is clear that they vary considerably; as the pressure increases, the pulse duration drops. The highest pulse duration (for 2.1 GN/m^2, or 21 Kbar) is 2.03 μs, while the lowest is 1.63 μs (for 26 GN/m^2, or 260 Kbar). These values are approximately 10 pct above and below, respectively, of Rack's[1] estimated pulse duration.

Using a simpler and less rigorous expression by Murr and Staudhammer[5] ($t_p = 2d^D/U_S$), pulse duration values of 2.06 and 1.75 μs were found, for the lowest and highest pressures, respectively. These

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*H. J. RACK: *Met. Trans. A*, 1976, vol. 7A, pp. 1571-76.

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Table I. Shock-Wave Parameters for Driver Plate (AISI 304SS) and Cover Plate (Ti) (Based on Kinslow[4])

Pressure (GN/m^2)	ρ^D (g/cm^3)	c^D (cm/s)	U_S^D (cm/s)	ρ^C (g/cm^3)	c^C (cm/s)	U_S^C (cm/s)
2.1	8.00	4.70×10^5	4.67×10^5	4.61	5.27×10^5	5.30×10^5
4.2	8.08	4.80	4.74	4.68	5.31	5.36
7.0	8.20	4.93	4.85	4.78	5.37	5.44
11.9	8.40	5.13	5.02	4.94	5.44	5.59
16.7	8.55	5.32	5.18	5.11	5.51	5.72
21.5 (est.)	8.72	5.50	5.34	5.36	5.56	5.84
26.0	8.87	5.65	5.48	5.41	5.60	5.96

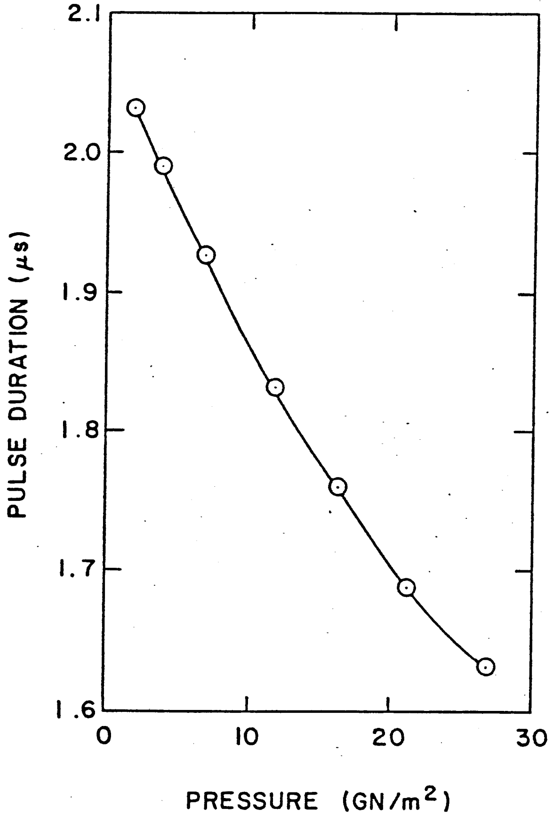

Fig. 1—Variation of pulse duration with pressure for a constant driver plate thickness (calculated assuming hydrodynamic response of the material).

results do not differ greatly from the ones using Orava and Wittman's[3] equation. The use of Fowles'[2] equations would yield results very close to the ones reported in Fig. 1.

b) Effect of Pulse Duration on Residual Substructure and Mechanical Properties. There is considerable evidence in the literature substantiating the effect of pulse duration on the substructure, hardness, and yield stress of various metals and alloys. Part of the data are schematically presented in Table II. For more details and specific experimental conditions, the reader should consult the original references. It was attempted in Table II to present the results as systematically as possible; of the available information, only the data showing the changes in residual properties with pulse duration at a constant (or nearly constant) pressure were chosen.

In the first report on the effect of pulse duration, Appleton and Waddington[6] shock loaded copper at 5.6 GN/m^2–1.23 μs and 5.0 GN/m^2–2.15 μs finding higher hardness, yield stress, and smaller cell diameter for the longer pulse duration experiment, in spite of the slightly lower pressure. Indeed, the cell size, hardness and yield stress of this event closely resembled those of the 9.8 GN/m^2–1.12 μs event, indicating that doubling the pressure resulted in similar increases. Champion and Rohde[7] systematically investigated the effects of pressure and pulse duration for Hadfield steel. Table II only presents the results at the pulse durations for which a common pressure could be found (~5.0 GN/m^2). The residual hardness and the stacking-fault probability increases with pulse duration. The differences are, however, less pronounced than the ones found by Appleton and Waddington.[6] Murr and Huang[9] studied the effect of pulse duration (1 and 2 μs) on nickel, Inconel 600 and Chromel-A at 25 GN/m^2. Again, increases in hardness and dislocation density accompanied pulse duration increases. Murr and Staudhammer[5] investigated the same effect for stainless steel, finding this time a certain decrease in hardness, when the pulse duration was increased from 1 to 2 μs. For comparison purposes, the hardness of a 35 GN/m^2 event is also shown. The difference in hardness between the 1 and 2 μs events, and 25 GN/m^2, in spite of not being dramatic, is however larger than the difference between the 25 and 35 GN/m^2 events at 2 μs. Some of the results of Marsh and Mikkola[10] (taken from his Fig. 3 at the pulse durations of 0.5 and 1.0 μs) for Cu-8.7 at. pct Ge are also shown in Table II. Marsh and Mikkola's[10] Fig. 3 and Table II clearly show that the increase in duration from 0.5 to 1.0 μs results in hardness increase at all pressures investigated (3.5 to 47.0 GN/m^2); additionally, the data indicate that the effect of pulse duration appears to become more marked at higher pressures. For lower durations, however, it seems that the effect of duration is opposite. Gray *et al*[11] investigated the effect of duration on 1008 steel; this is the only work where *both* pressure and rarefaction were maintained constant. Their results show, as predicted, an increase of residual yield stress with pulse duration; this effect is almost nonexistent at low pressures, becoming very pronounced at 350 Kbar (407 and 478 MN/m^2 for 0.5 and 1.0 μs, respectively). This increase in the effect of pulse duration for higher pressures can also be inferred from the divergent lines in Fig. 7 of Champion and Rohde.[7] Meyers[12] shock loaded nickel at the pulse durations of 1.2, 2.4, and 10.1 μs

Table II. Summary of Results on the Effect of Pulse Duration Reported in the Literature

Material	Pressure (GN/m^2)	Pulse Duration (μs)	Substructure	Hardness	Yield Stress (MN/m^2)	Investigators
Copper	5.6	1.23	0.53 (cell size in μm)	81 (VHN)	170	Appleton and Waddington[6] (Taken from Table I)
	5.0	2.15	0.31	93	242	
	9.8	1.12	0.31	90	263	
Hadfield Steel	4.8	2.2	7.1 (stacking fault prob. X 10^3)	26 (Rc)		Champion and Rohde[7] (taken from Table III)
	4.25	0.23	4.2	18.7		
	5.4	0.065	5.7	16.8		
Nickel	25.0	1	6.3 X 10^{10}(Disl. dens. in cm^{-2})	214 (VHN)		Murr and Huang[9] (taken from Fig. 1 and Table II)
	24.0	2	8.6 X 10^{10}	214		
Inconel 600	25.0	1	8.1 X 10^{10}	310		
	24.0	2	8.2 X 10^{10}	331		
Chromel-A	25.0	1	1.1 X 10^{11}	371		
	24.0	2	1.3 X 10^{11}	400		
340 S.S.	25.0	1		382 (VHN)		Murr and Staudhammer[5] (taken from Fig. 1)
	25.0	2		354		
	35.0	2		363		
Cu-8.7 at. pct Ge	3.5	0.5		161.5		Marsh and Mikkola[10]
	3.5	1		162.3		
	10.0	0.5		172.1		
	10.0	1		174.2		
	18.5	0.5		176.2		
	18.5	1		184.1		
	32.5	0.5		180.8		
	32.5	1		201.2		
	47.0	0.5		199		
	47.0	1		220		
1008 Steel	10.0	1			305	G. T. Gray III, G. A. Stone, and R. N. Orava[11]
	10.0	2			324	
	13.0	1			307	
	13.0	2			307	
	18.0	0.5			292	
	18.0	1.0			307	
	25.0	0.5			304	
	25.0	1.0			330	
	35.0	0.5			407	
	35.0	1.0			478	
Nickel	20.0	1.2			395	Meyers[12]
	20.0	2.4			405	
	20.0	10.1			480	

(20 GN/m^2 pressure); the yield stresses were 395, 405 and 480 MN/m^2, respectively. Mantaroshin[8] *et al* also found an effect of pulse duration on the hardness of copper.

Table II and the foregoing discussion allow the following conclusions to be drawn: in the pulse duration range of 0.5 to 2.0 μs the residual strength of shock-loaded metals and alloys increases with pulse duration at a constant pressure (except Ref. 5). The effect of pulse duration on the residual hardness and tensile properties seems to be of the same order of magnitude than pressure. It seems that, as the pressure becomes higher, the effect of pulse duration becomes more pronounced.

c) Effect of Pulse Duration Changes on Rack's[1] Results. The only rigorous way of establishing the effects of pulse duration variation on the residual tensile properties of RMI 38644 would be to repeat the experiments at the pressures used by Rack[1] and a constant pulse duration of 1.8 μs; this could be accomplished by changes in the driver plate thickness as shown by Fig. 1. This is, however, beyond the scope of this note. The results presented in Table II indicate that the effect of pulse duration varies from material to material; some metals, like copper,[6] Cu-8.7 at. pct Ge (Ref. 10) seem to be more sensitive to pressure duration changes than others like Inconel 600, Chromel-A (Ref. 9) and Hadfield steel.[7] It is not known to what extent the lack of control of rarefaction rates could be responsible for the differences. These differences render a quantitative prediction of the corrected tensile properties of RMI 38644 impossible. From Fig. 1 one can see that the pressure at which the calculated pulse duration is 1.8 μs is 13.5 GN/m^2. Consequently, Rack underestimated the pulse duration for the events below this pressure (2.1, 4.2, 7.0, 11.9 GN/m^2) and overestimated it for the events at higher pressures (16.7, 21.7, and 26.0 GN/m^2). The pressure range from which the pulse duration was underestimated corresponds to the region where the tensile properties were decreased; there is no report in the literature on the effect of pulse duration in a similar situation and no prediction can be made. At pressures above 13.5 GN/m^2 the tensile properties would exceed those reported by Rack[1] if a constant pulse duration of 1.8 μs is assumed.

As a conclusion, evidence reported in the literature seems to indicate that pulse duration is an effect of the same order of magnitude than pressure. If the latter has to be precisely stated, the former should also be specified with the maximum permissible precision. It is not expected, however, that drastic changes in residual tensile properties would be obtained by maintaining a constant pulse duration; but, since the effect of pressure on these properties was not dramatic, these two effects could be confused, or the variation of one could mask the effect of the other. Finally, it should be emphasized that this note does not try to invalidate the results presented by Rack;[1] rather, it attempts to render them more complete and, therefore, more meaningful.

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Author's Reply

H. J. RACK

The author agrees with Dr. Meyers that the pulse duration during shock loading of RMI 38644 was a function of shock pressure. However, the possible effects of this ±10 pct variation on the residual mechanical behavior of RMI 38644 are difficult to predict. Indeed Table II in the discussion shows that a 100 pct change in pulse duration may either increase or decrease the hardness/yield strength after shock loading. Furthermore, the discusser suggests that the decrease in pulse duration at higher shock pressures might have led to an underestimation of the residual yield strength. If correct, this would tend to emphasize the initial reduction in mechanical strength observed at low shock pressures and would actually reinforce the conclusions of the original investigation. In summation, the principal observations of the effect of shock loading on the residual strength of RMI 38644 remain unaltered by Dr. Meyers comments. Shock loading of this alloy initially results in a reduction in strength, the most likely explanation for this decrease being the shock induced reversion of athermal omega.

Cessation of the Dislocation Splitting Mechanism for γ' Al-Ag Formation Above a Critical Temperature

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G. W. LORIMER

Nicholson and Nutting,[1] and subsequently Hren and Thomas,[2] have used transmission electron microscopy to show that the formation of hcp γ' plates in fcc α Al-Ag takes place by the splitting of $a/2\langle 110\rangle$ dislocations into two Shockley partials bounding a region of γ' two lattice planes (0.461 nm) thick and the simultaneous diffusion of Ag into the hcp regions. Using hot-stage electron microscopy, Hren and Thomas,[2] observed that this mechanism is rapid and operated throughout the range of aging temperatures, 100 to 400°C, which they investigated in an Al-20 wt pct Ag alloy. During the present study of an Al-15.0 wt pct Ag alloy this mechanism was found to cease operating above a critical temperature lying well below the solvus.

An Al-15.0 wt pct Ag alloy was melted under argon, homogenized at temperatures up to 550°C for one week, hot rolled into strip 2.5×10^{-4} m thick and then cold rolled to a thickness of 1.2×10^{-4} m. Samples were solution annealed for 180 s at 500°C and isothermally reacted at temperatures ranging from 160 to 400°C; both treatments were conducted in salt baths. Following isothermal reaction, specimens were quenched in iced water. Electropolishing using the 'window' technique[3] was carried out at 0 ± 8°C in a solution of 0.04 kg hydrated sodium thiocyanate, 2.2×10^{-4} m^3 ethyl alcohol and 10^{-5} m^3 glycerol at 26 volts, in the absence of γ' and 35 volts, in its presence. The electrolyte was stirred and the specimen agitated throughout the thinning process. Immediately afterwards, the specimen was immersed in the stripping solution described by Clark[4] for times ranging from 90 s, in the absence of GP zones, to as little as 8 s, in their presence. Electron microscopic examination was conducted with an AEI EM6G.

The evolution of γ' by the dislocation splitting mechanism has been amply documented.[1,2] The absence of γ' at dislocations after aging above the critical temperature is illustrated in Fig. 1(*a*), and the presence of γ' at dislocations at 10 and 20°C lower aging temperature is shown in Fig. 1(*b*) and (*c*). Fig. 1(*d*) shows later stages of γ' evolution at the lowest of

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