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Reprinted from:

SHOCK WAVES AND HIGH - STRAIN - RATE PHENOMENA IN METALS (1981)
Edited by Marc A. Meyers and Lawrence E. Murr
Book available from: Plenum Publishing Corporation
233 Spring Street, New York, N.Y. 10013

Chapter 45

THERMOMECHANICAL PROCESSING BY SHOCK WAVES: AN OVERVIEW

Marc A. Meyers
Raimo N. Orava

New Mexico Institute of Mining and Technology
Socorro, New Mexico 87801, U.S.A.

South Dakota School of Mines and Technology
Rapid City, South Dakota 57701, U.S.A.

Research efforts assessing the potential of shock TMP for a number of alloy systems are reviewed. Shock loading seems to be a promising deformation technique in TMP when (a) the initial strength of the alloy is such that conventional deformation is precluded and (b) when the shock wave induces property improvements that are significantly superior to those of conventional deformation.

I. INTRODUCTION

The cooperative strengthening effects between mechanical and thermal treatments--called thermomechanical processing (TMP) or thermomechanical treatments (TMT)--have received in the past twenty years a great deal of attention; as a result, more than 1000 papers and reports have been published on this subject (1). In a 1976 article identifying the unknowns in thermomechanical processing, Zackay (2) points out that before 1966 much of the thermomechanical treatment effort was centered on laboratory studies of ultra-high-strength steels. "Fortunately, the events of the past ten years have profoundly altered this unsatisfactory situation. Two major contributing factors to the renaissance of TMT have been the availability of high resolution instrumentation for the analysis of composition and microstructure, and new developments in technology, for example, the automation of shaping and forming equipment. Indeed, the most striking characteristic of the rapid technological progress made in TMT in the last ten years is the quantitative integration of physical metallurgy, *viz.*, the structure *vs.* property correlations with process metallurgy, *i.e.*, the variables of time, temperature and deformation (2)". This approach has resulted in the successful commercialization of high-strength low-alloy (HSLA) steels and presents excellent opportunities for superalloys, aluminum, and titanium alloys.

Whereas shock hardening *per se* may be limited interest technologically, the stringent requirements currently being placed on high-strength alloys tend to override some of the adverse considerations involving the economy of explosive fabrication, and the availability of suitable facilities. The net result is that shock TMP is rendered an attractive processing technique either if conventional deformation is not feasible or if shock TMP yields improvements over and above those achievable by conventional TMP. Table I shows the alloys and objectives of most investigations on TMP during the 1960's. Shock hardening has been used as the mechanical processing means for several of these alloy systems, as will be seen in the following sections.

II. STEELS

The phase transformations that deformation and thermal treatments can impart to steels make them especially amenable to TMP. A wide range of TMP schedules is indicated by Delaey (3) and many of them could prove useful in shock loading. The BCC- HCP transformation undergone by iron at 13 GPa pressure is an added advantage of shock TMP. There have been several attempts, described below, to explore the benefits of shock TMP in steels.

Silverman, et al. (4) investigated the effects of shock loading (36 GPa) on the strength of H-11 and 25 pct. nickel steels. Yield strength increased from 235 ksi to 340 ksi in the as-shocked condition, for the H-11 steel, and from 235 ksi in the austenitized-plus-aged condition to 255 ksi in the shocked-plus-aged condition, for the 25 pct. nickel steel. Stein and Johnson (5) studied a high-strength steel (0.43 pct. C, 3.0 pct. Cr, 1.5 pct. Ni, 1.5 pct. Si), applying shock loading at both ambient and higher temperatures (470,680, and 1000°F). Their study disclosed the beneficial effects of shock aus-working the alloy. Advantage was taken of pre- and post- shocking thermal treatments. They concluded that, under certain conditions (shock-loading the metastable austenite at 470 and 680°F and then tempering at 400°F) it was possible to increase the strength of the alloy with no apparent loss in ductility. The work by Koepke, et al. (6), was directed towards the study of three types of alloys. For steels normally hardened by cold work, they found that the strengthening occurred only in the ferrite. Fully pearlitic steels did not respond well to shock hardening. Secondly, they studied steels normally strengthened by thermal treatments, and found them to be less responsive to shock hardening. And

TABLE I. Materials Subjected to Experimental TMT During the 1960's (from Henning (1).

Material	Description of TMT	Objective	Remarks
Carbon and low-alloy construction steels	Strain aging accomplished by warm finishing	Higher strength	
Low-and moderate-alloy high-strength steels	Hot-cold working or ausforming	Higher strength	
Maraging steels	Warm working by controlling reductions at lower end of hot-working temperature range	Increased ductility, reduced grain size	TMT controls both grain size and precipitation reactions
Tool steels	Hot-cold working and/or ausforming	Improved toughness	
Bearing steels	Hot-cold working	Improved fretting fatigue life of bearings	
High-speed steels	Hot-cold working above the nose of the time-temperature transformation (TTT) curve	Improved wear resistance and dimensional stability	
Nickel-base superalloys	Controlled reductions and temperatures specifically to alter metallurgical structure and precipitation hardening	Increased strength, increased resistance to low-cycle fatigue	
Titanium alloys	Working at specific temperatures and reductions to control microstructure	Optimization of mechanical properties	Properties of all-beta alloys depend on controlled warm reduction
Aluminum alloys	Warm or cold TMT processes to alter residual stresses	Reduce sensitivity to stress corrosion	
Iron-base superalloys			Respond unfavorably to TMT

thirdly, upon investigating precipitation-hardenable A-286 stainless steels they found that aging preceded by shock loading led to higher hardnesses than just aging. Also, the aging kinetics was accelerated by pre-shocking. They attributed it to faster diffusion rates due to the increase in the density of defects. Similar thermal response has been observed in other investigations and will be discussed later in this chapter. Gamma prime, Ni_3 (Al, Ti) precipitated homogeneously in the post-shocking aging. No attempt was made to precede shocking by a pre-aging step.

Doherty,et al. (7,8), in a study partially directed at developing methods of prevention of spall and fracture formation when shock-loading irregular shapes, found that AISI 4340, AISI H-11, and 18 pct. nickel maraging steels responded better to shock plus age treatments than just shock loading. Increases of tensile properties ranging from 29,000 to 70,000 psi above the ones obtained with conventional heat treatments were observed; the associated ductilities were either equal or slightly lower than the ones for the conventional treatments.

There seems to be some potential in the shock TMP for 17-7 PH stainless steel (9). The standard heat treatment for this alloy consists of a sequence of solution heat treatment, austenite conditioning, transformation to martensite (by cooling or cold working),and precipitation hardening. The latter step also has the purpose of tempering the martensite previously formed. Shock loading was imparted in the solution-treated and martensitic conditions with post-shock aging. The results are shown in Table II. The increase in strength are substantial.

Another exploratory shock TMP study has resulted in improvements in the ambient-temperature strength of S7 tool steel (10). This is depicted in Figure 1. Shocking (20 GPa) was preceded by quenching and tempering and followed by aging. It can be seen that a combination of high hardness and Charpy impact strength could be achieved. Conventional TMP can not be applied to this alloy due to the high initial strength.

The Soviets have devoted a considerable research effort on the development of TMT (11). Shock loading has been used in this context (12-17). This effort has included the study of shock TMP of trip steels (12-14) and nickel-chromium steels (13-15).

III. NICKEL-BASE SUPERALLOYS

In the period 1972-1975 Orava extensively studied (18-25) structural and mechanical properties of two commercial superalloys--Udimet 700 and Inconel 718--by shock TMP. If room temperature tensile properties alone had been used as the criterion to assess

TABLE II. Room Temperature Tensile Properties of 17-7 PH Stainless Steel Following Thermal Processing, Conventional and Shock TMP.

Deformation	Effective Strain	0.2% YS (ksi)	UTS (ksi)	Total Elong. (%)	R.A. (%)
(a) Solution treatment + deformation + Aging					
	0	52.0	137.1	38.8	52.7
cold rolling	23.3	107.9	158.2	29.2	40.4
shocking	23.3	131.7	169.4	21.5	42.2
(b) Solution + transformed + deformation + aging					
	0	181.8	194.9	6.5	27.0
shocking	23.3	203.1	205.8	3.2	28.2

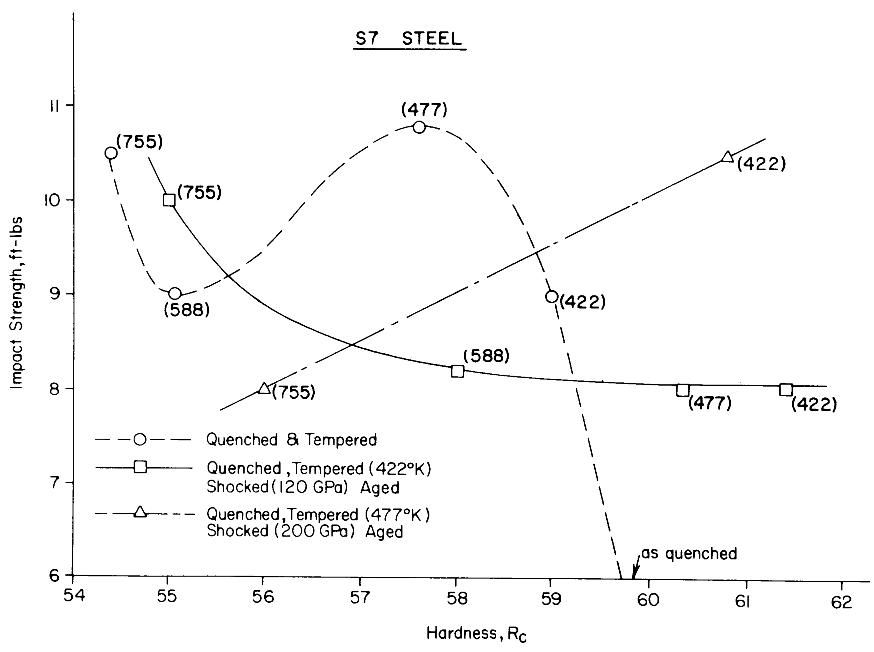

FIGURE 1. Variation of Charpy impact strength with hardness for undeformed and shock-TMP S7 tool steel (10); tempering temperature in K.

the response of Udimet 700 (18-21) and Inconel 718 (22-24) nickel-base superalloys to shock TMP, then a conclusion similar to that of Antrobus and Reid (26) could have been reached. However, an examination of other mechanical properties and thermal microstructural stability favored shock TMP over conventional TMP, particularly in the case of Udimet 700. As Table III shows, the shock-aged alloy had markedly superior elevated temperature ductility and toughness, and longer stress-rupture life than the conventionally TMP treated counterpart. The constant-load creep curves at 1200°F are shown in Figure 2. The stress rupture life of the shocked material is over twice that of the rolled one. Moreover, the low-cycle fatigue life at room and elevated temperature was increased over that of the thermally treated control material. In addition, after 50 h at 1800°F, the cold-rolled and aged material contained extensive cellular precipitate colonies whereas no cellular recrystallization could be detected in the shock-aged alloy. Figure 3 (a) shows the evidence of the occurrence of recrystallization. This is manifested by the formation of cellular precipitate during aging at 1975°F for four hours. Such a cellular reaction was observed whenever the γ' aging step followed cold rolling in the solution treated condition to reductions of 12.5 and 19.1%. Cellular recrystallization has been observed previously in cold-reduced Udimet 700 annealed thereafter at 1800°F for sixteen hours (44). In direct contrast to the above results, cellular recrystallization was not detected in any of the shock-loaded material. The photomicrograph of Figure 3 (b) tipifies microstructures resulting from the γ' aging of shock-hardened Udimet 700. The enhancement of properties was attributed to the combination of a relatively high volume fraction of fine primary γ' and a finely dispersed, thermally stabilized, dislocation substructure. The minimum planar slip band spacing in shocked material was 0.02 μm after cold-rolling.

The results presented in Figure 4 (21) demonstrate that further improvement in certain properties of Udimet 700 could be achieved by aging before working in order to control the size distribution of the primary and secondary γ', $Ni_3(Al, Ti)$ strengthening precipitate. Since the γ' already exists after solution, and does not precipitate on dislocations with the result that predeformation aging is essential for optimizing the TMP strengthening effect, regardless of working rate (Figure 4). Some results of the TMP schedule for Inconel 718 represented in Figure 4 are listed in Table IV. The stress rupture curves at 1200°F are shown in Figure 5. The benefits due to shock TMP can be clearly seen.

Shock TMP using the modified schedule (aging + shocking + aging) resulted in a stress-rupture life improvement of 50 pct over conventionally deformed material.

The γ' precipitate in Inconel 718 is a metastable phase. Upon overaging, it transforms into a lath-like δ precipitate. One-hundred

TABLE III. Shock-Aging Effects In Udimet 700 Nickel-Base Superalloy; 2150°F, 4 h + cold work + 1550°F, 24 h + 1400°F, 16 h (21).

Property	Temp. (°F)	STA** Control	Method of Cold Working in TMP: Rolled (19% Red.)	Shocked (527 kbar)
0.2% YS (ksi)	70	134	165	169
	1200	108	160	144
	1400	112	146	129
UTS (ksi)	70	177	203	207
	1200	134	175	170
	1400	126	152	140
Elongation (%)	70	11	11	9
	1200	4	2	12
	1400	7	10	16
RA (%)	70	17	15	19
	1200	10	12	30
	1400	16	25	43
Toughness Index* (in-kip/in^3)	70	33	34	44
	1200	14	23	62
	1400	22	44	86
Rupture Life, 120 ksi	1200	4 h	146 h	326 h
LCF Life, $k_t = 2$				
85 ± 65 ksi	70	5500 cycles		7400 cycles
75 ± 55 ksi	1200	1200 cycles		2200 cycles

*Strain energy to fracture from a true stress-true strain tensile curve.

**Solution treated and aged.

TABLE IV. Shock-Aging Effects In Inconel 718 Nickel-Base Superalloy; 1750°F, 1 h + 1300°F, 4 h + cold work + 1250°F, 8 h F.C. to 1150°F, total 18 h (22-24).

	Temp. (°F)	STA Control	Method of Cold Working in TMP: Rolled (19% Red.)	Shocked (510 kbar)
0.2% YS (ksi)	70	177	232	218
	1000	153	199	185
	1200	141	182	174
UTS (ksi)	70	207	238	223
	1000	177	204	191
	1200	154	188	180
Elong. (%)	70	17	5	5
	1000	15	4	5
	1200	11	14	7
RA (%)	70	41	46	31
	1000	35	23	33
	1200	11	14	7
Toughness Index (in-kip/in^3)	70	128	100	127
	1000	85	59	90
	1200	28	115	81
Rupture Life				
170 ksi	1000	43	1023	458
110 ksi	1200	111	147	202
KCF Life, $k_t = 2$				
105 $\pm$ 85 ksi	70	3200 cycles	3800 cycles	3800 cycles
80 $\pm$ 60 ksi	1200	720 cycles	1200 cycles	1100 cycles

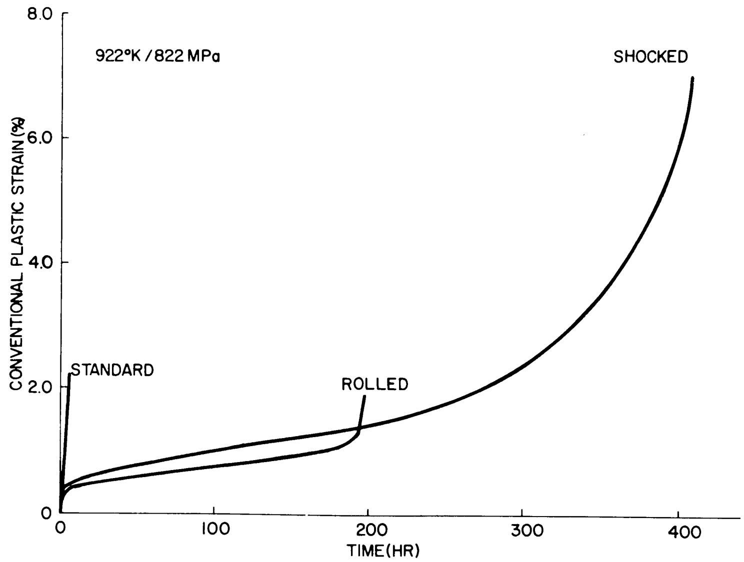

FIGURE 2. Constant-load creep curves for Udimet 700 at 1200°F (TMP schedule: 2150°F, 4 h + cold work + 1550°F, 24 h + 1400°F, 16 h): Stress: 120 ksi.

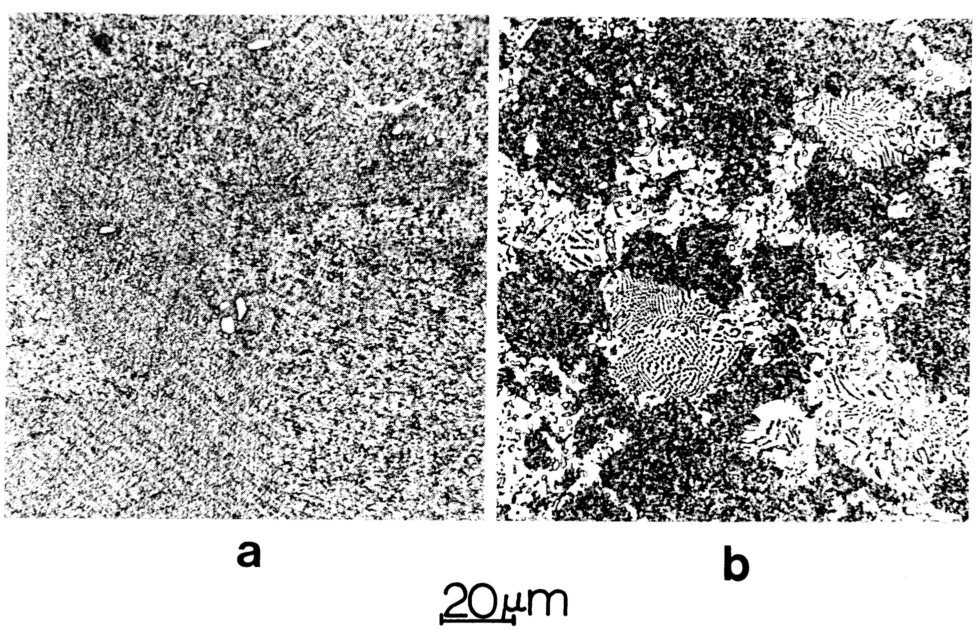

FIGURE 3(a), (b). Cellular precipitate in Udimet 700 aged at 1975°F for four hours following cold work in the solution treated condition. a. Cold rolled to 12.5% reduction in thickness b. Shock loaded to 287 kbar

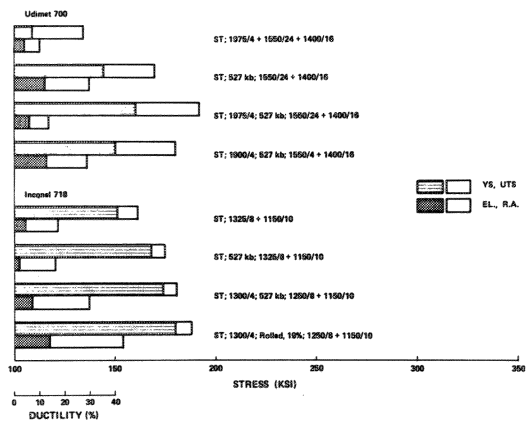

FIGURE 4. Effect of predeformation on the TMP strengthening response of nickel-base alloys at 1200°F.

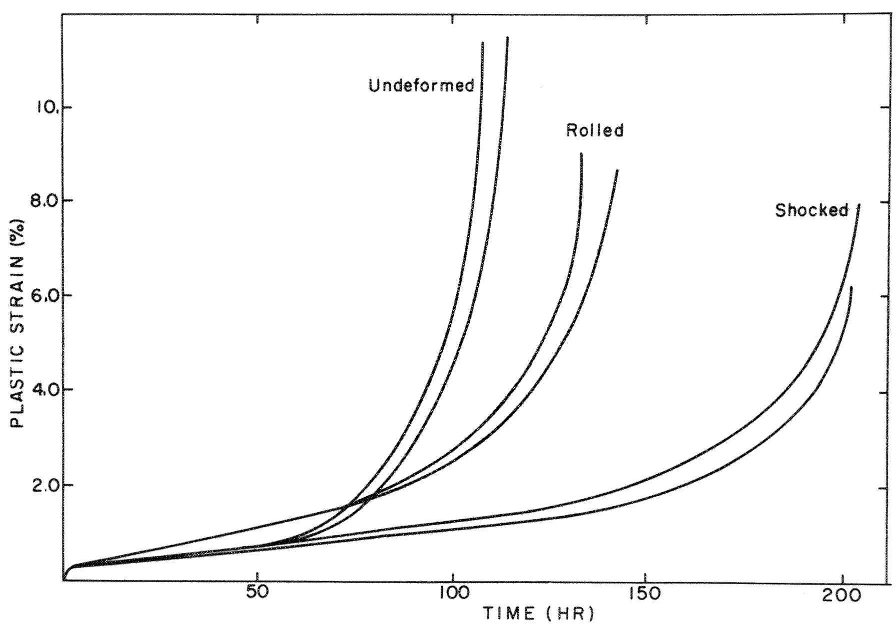

FIGURE 5. Constant-load creep curves at 1200°F for Inconel 718 shocked and rolled: 1750°F, 1 h + 1300°F, 4 h + cold work + 1250°F, 8 h F.C. to 1150°F, total 18 h; undeformed, σ = 110 ksi.

hour aging treatments at 1450°F disclosed extensive δ formation for the shocked material; it was totally absent from the undeformed and somewhat present in the cold-rolled one (see Figure 6). δ is known to form along the {111} planes of the matrix. So, deformation enhanced δ precipitation. Figure 7 shows some of the δ plates, in a background of overgrown γ' particles. δ has a deleterious effect on the mechanical properties of undeformed material.

These data disclose that the properties resulting from the conventional and shock TMP of Inconel 718 do not differ appreciably, even though shock deformation was responsible for a finer planar substructure. However, the desirability of pre-aging reduces the conventional formability of this alloy and favors shock straining. The strength differences, and a lower observed dislocation density in the shocked material, suggested that equivalent maximum shear strain rather than equivalent effective strain, as used above, would be a more realistic basis on which to compare the effects of different methods of working. An interpolation disclosed that with this approach, shock and conventional TMP strengths would have been identical. In continuation of this study, it was decided to evaluate the response of Unitemp AF2-1DA nickel alloy to shock TMP, particularly with regard to its effect on the resistance of crack propagation (25).

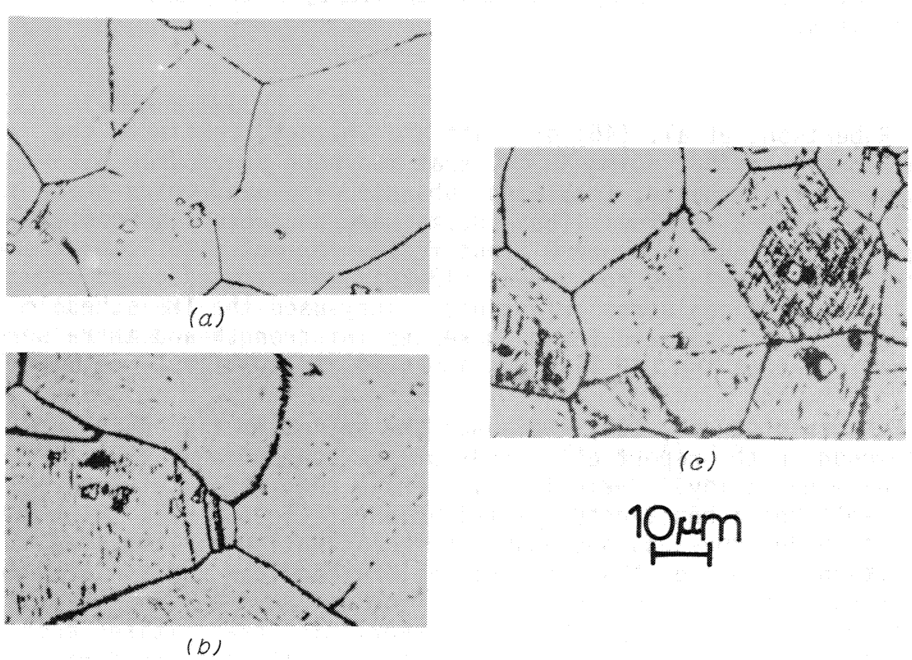

FIGURE 6. One hundred hour anneal at 1450°F for Inconel 718 (solution treated at 1950°F) a. undeformed b. rolled c. shocked

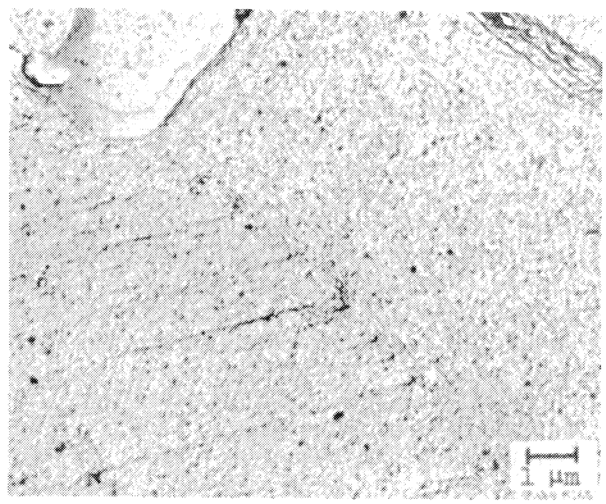

FIGURE 7. Two-stage surface replica of overaged shocked condition.

Robertson, et al. (46) of Pratt and Whitney, continued the investigative effort using IN-100 that had been previously *gatorized*. Instead of flat sheets, they used subscale first and third-stage disks for F100 B turbine. These disks were protected in lead potting. The maximum pressure without fracturing was much lower than that used by Orava and co-workers (18-25); at pressures over 150 kbar fracturing could not be avoided. They used the TMP schedule with a pre-shock age and found increases in strength and there were indications that the LCF life was improved at 1000°F (538°C).

Conserva, et al., (27) explained the predeformation aging phenomenon in the report of a study of the conventional TMP of 7075 aluminum alloy. Basically, it serves two purposes. Firstly, it establishes a homogeneous distribution of second phase which can not be destroyed by subsequent deformation. Otherwise, the nucleation of second phase on dislocations can lead, on final aging, to a non-uniform precipitate distribution which is not conducive to optimum response. Secondly, the preexisting precipitates anchor the dislocation substructure in order to inhibit recovery during the final aging treatment (s). A schematic illustration of what is envisioned to occur in Inconel 718 is provided in Figure 8 (22-24).

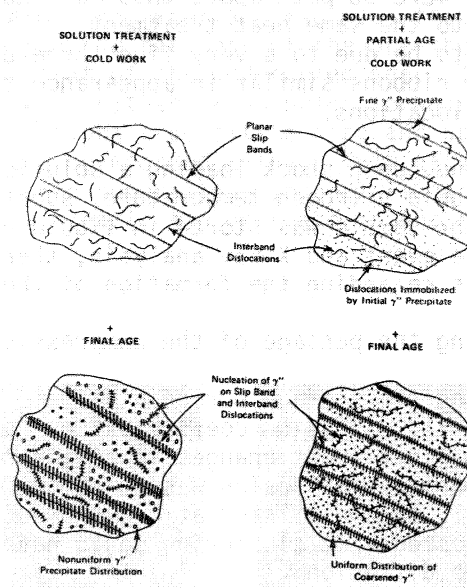

FIGURE 8. Schematic representation of the predeformation aging effect in the conventional and shock TMP of Inconel 718.

IV. ALUMINUM ALLOYS

The only major study of shock TMP between 1966 and 1970 known to these authors involved the evaluation of the tensile properties and stress corrosion cracking susceptibility of as-shocked and shock-aged 7075 aluminum alloy (28). Some decreases in strength due to shock-aging were observed; however, a careful analysis of the SCC results disclosed appreciable increases in failure time of shocked and overaged material when shock deformation was imparted in the T6 condition.

Antrobus and Reid (26) compared shock and conventional TMP by cold rolling (to an equivalent strain) for an Al-2.6 pct. Cu-1.27 pct. Mg alloy. Shocking the alloy in the solution-treated condition and aging increased the strength above that of simply aged material. The peak strength was achieved at shorter aging times. On the structural level, the precipitate distribution was affected by both methods of deformation; the mean length of the lath-shaped precipitates decreased, while their density increased.

For 6061-0 and 6061-T6 aluminum shocked in the course of explosion bonding, Wittman (29) obtained some interesting results. For shock-loaded aluminum, after 650 and 775°F aging treatments, the strength levels were 50 pct. above those of undeformed control material subjected to the same heat treatment. This strength increase was thought to be due to a very fine three-dimensional network of precipitate ribbons similar in appearance to a network of widely extended dislocations.

Stein (30) found, upon shock loading a solution-treated Al-3.85 pct. Cu alloy at liquid nitrogen temperature, substantial G-P zone formation. Since the sample was stored in liquid nitrogen in the interval between the event and X-ray analysis, there are three viable possibilities regarding the formation of these zones:

a) clustering during the passage of the compressive shock wave.

b) clustering during the storage period at liquid nitrogen temperature. Since the diffusion coefficient of Cu at 77 K is too low to generate significant changes in atom positions, and since vacancy-enhanced diffusion was ruled out excess vacancies would most likely be annihilated at close-by dislocations. Stein (30) suggested that clustering could have taken place by dislocation pipe diffusion.

c) clustering during X-ray examination.

Greenhut, et al. (31) investigated the precipitation response of an aluminum alloy, and found that it was enhanced in the spalling region, that was subjected to strong tensile stresses and exhibited, consequently, a much higher vacancy density. Their study would indicate that not all excess vacancies are absorbed by adjoining dislocations; they could, therefore, have an important effect on precipitation.

V. OTHER SYSTEMS

Nordstrom, et al. (32) investigated the increases in strength of a precipitation-hardenable Cu-Be alloy by both conventional and shock hardening. Both methods of deformation were imparted in the solution-treated condition; subsequent aging increased the room temperature strength of TMP alloys above that of solely aged alloy. However, the cold-deformed conditions responded better to aging than shock-loaded ones. Gamma prime, the peak strength precipitate, is coherent with the matrix, with high coherency strains. Shock-loaded Cu-Be twins profusely at shock pressures above 20 GPa; however, twinning did not seem to contribute significantly either to the strength or to the nucleation of precipitates during aging.

On the other hand, Beta III titanium alloy (33) shock-loaded in the solution-treated condition and then aged exhibits a response similar to that obtained earlier by Kalish and Rack (34), with a slightly higher ductility. The titanium alloy RMI 38644 was shocked at 16.7 GPa by Rack (45) and then aged. A comparison of the age-hardening response of the material disclosed that, at equivalent aging times (8 hours) the hardness of the pre-shocked material was always higher than that of the undeformed one. Transmission electron microscopy showed that shocking prior to aging resulted in a decrease in the HCP alpha precipitate size and spacing relative to that observed in the undeformed alloy. This effect of shocking is similar to that in the superalloys Inconel 718 and Udimet 700. Figure 9 shows very clearly the decrease in the size of α induced by shock loading.

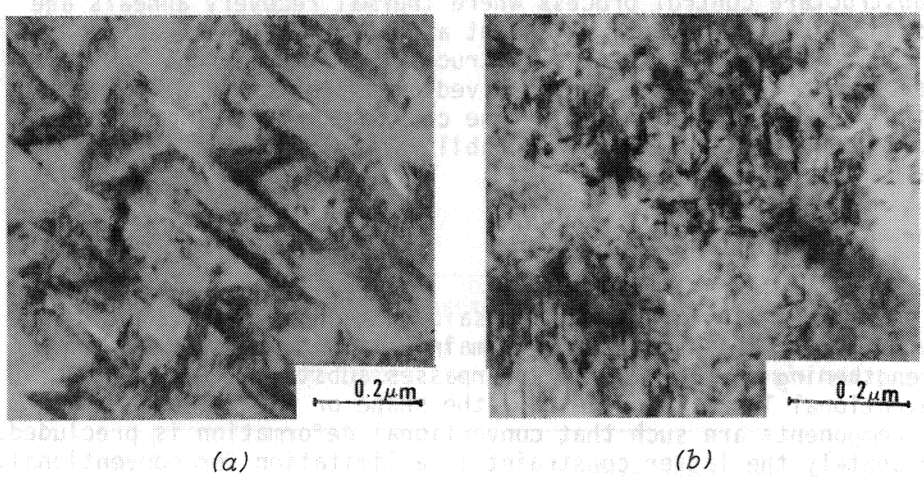

FIGURE 9. Transmission electron micrographs of RMI 386 44 at 723 K for 8 h after (a) solution treatment and (b) shock-loading at 167 kbar (16.7 GPa) (45).

A work-hardenable and subsequently precipitation-hardenable Co alloy was shock loaded at the Denver Research Institute for evaluation of its response by the Naval Weapons Center, China Lake. The precipitation hardening per unit prestrain was appreciably greater after shock-wave-deformation than after conventional deformation (35).

VI. THERMAL STABILIZATION OF THE SUBSTRUCTURE

It was found by Meyers (40) that the substructure developed by shock-loading is mechanically unstable, for nickel. Similar results were found by Orava, Stone and Pelton (41) for magnetic ingot iron. This unstability of the substructure results in work softening; after the yield point is reached the work-hardening rate is essentially negative, leading to immediate necking. Typical curves are shown in Figures 10 and 11. Under the action of mechanical stress, the substructure will change, and this change will be toward the substructure that is in equilibrium under the imposed conditions. Annealing under appropriate conditions will stabilize the shock-produced substructure. The effects are dramatic, as can be seen in Figures 10 and 11. Work-softening is eliminated with little loss in yield strength. These treatments do not, strictly speaking, fall under the category of TMP; they are called mechanical-thermal treatments (MTT). In contrast to TMP, MTT is strictly a substructure control process where thermal recovery anneals are used to stabilize, by rearrangement and/or impurity segregation, the dislocation network or cell structure for improved overall properties. The subject has received a comprehensive and critical review (37). Because of the unique characteristics of the shock-induced substructures, thermal stabilization should be especially advantageous for them.

VII. CONCLUSIONS

As a conclusion, it can be said that shock TMP presents potential applications. These lie mainly in situations where the strengthening due to shock TMP surpasses substantially that of conventional TMP or where either the shape or initial strength of the components are such that conventional deformation is precluded. Fortunately the latter constraint is a limitation for conventional, but not shock TMP. The preexistence of a precipate that has resulted in the greatest strength gains (22), also is a deterrant to conventional deformation, because it increases the resistance to deformation of the alloy. The importance of slip dispersal in achieving higher strengths is now being recognized (36). Another situation in which the potential benefits of shock TMP have not been completely explored is, in the authors' opinion, tool steels. The process could be applied to gun barrels, forging dies, breaches, shafts, armor, punches, cutting tools, and a wide variety of components requiring high integrity and strength.

As in any technological application, the benefits of shock TMP have to be weighed against its cost. Therefore, it is impossible to establish a general rule as to the worthiness of the process. The only viable approach to an eventual commercialization of the process would be to make in-depth studies of specific applications looking at the tooling requirements, developing techniques for shocking specific irregular shapes, optimizing thermal and shock treatments, assessing the performance of the components from the properties, and making an overall cost-benefit analysis.

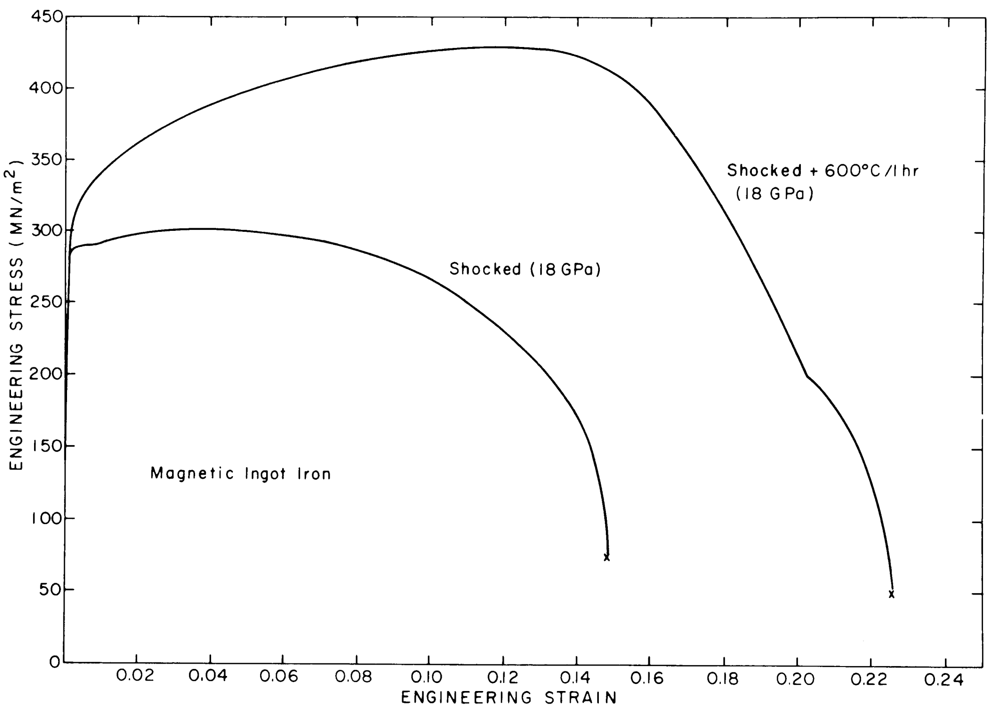

FIGURE 10. Elimination of work-softening in shock-loaded magnetic ingot iron by thermal stabilization of the substructure (41).

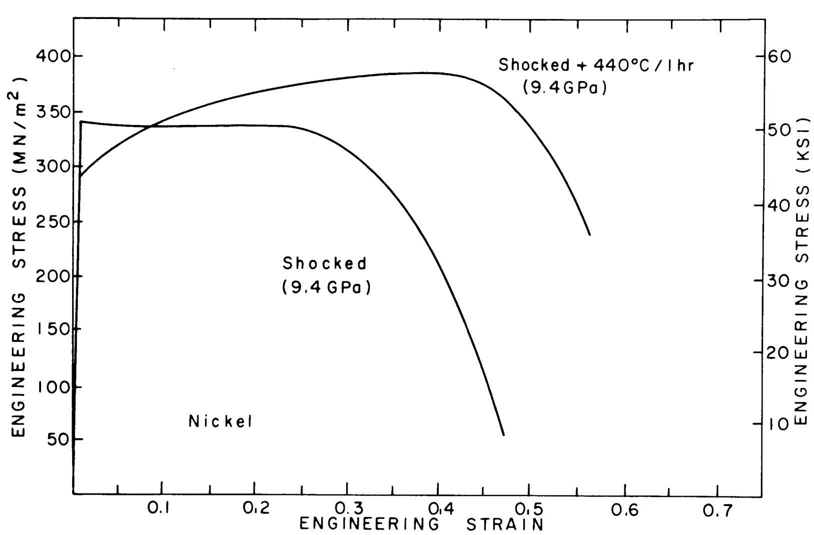

FIGURE 11. Elimination of work-softening in shock-loaded nickel by thermal stabilization of the substructure (42).

Some of the processing schedules that could be explored using shock waves are:

1. Austenite can be shocked above the M_s temperature, a process analogous to ausworking. The shock strain would be introduced at a temperature considerably below that practical be conventional ausworking, and it is known that final mechanical properties increase as austenite working temperature decreases (38). Because of the rapid processing time possible by shock deformation, steels with less sluggish transformation characteristics could be processed. It would only be necessary, in principle, to cool past the knee of the TTT curve and shock in the lower bay region as illustrated in Figure 12.

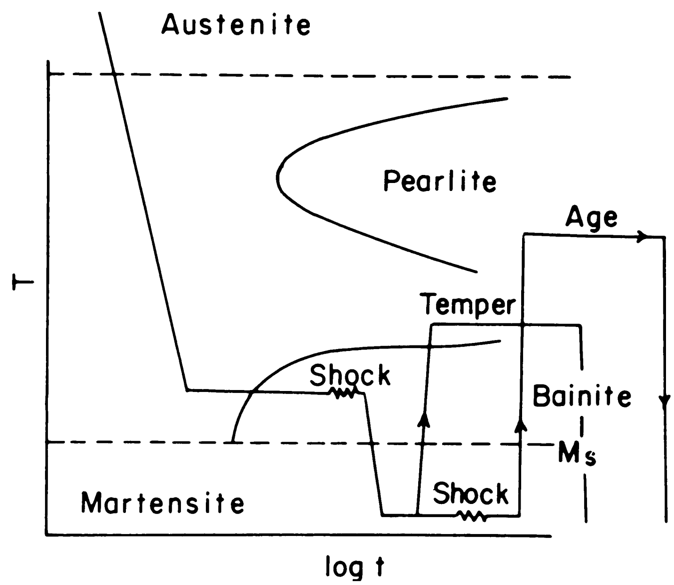

FIGURE 12. Schematic time-temperature-transformation diagrams of aus-shocking "with shock-strain-aging".

2. Another similar treatment is that suggested by Zackay (39). He anticipates that K_c to yield strength ratios in excess of 1 may be possible at yield strengths approaching 300 ksi by deforming in the lower bainitic region and transforming to martensite. It would be difficult to introduce the required 10 to 20 percent deformation by conventional means. The equivalent transient strain would be developed at shock pressures of 400 kbar and less. An example of such a processing cycle is shown in Figure 13.

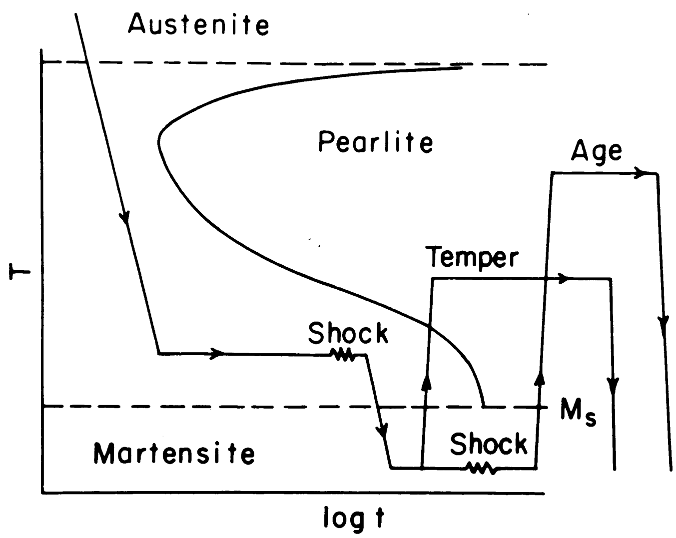

FIGURE 13. Schematic time-temperature-deformation diagram for shock TMP of austenite plus bainite and shock-strain-aging of bainite plus martensite.

3. Shock-aging (sequential shock deformation and thermal aging) treatments could be performed in any one of a number of initial material conditions, including an untempered, partially tempered, or fully tempered martensitic state. It would also be possible to shock-age a steel which has been strengthened to a high level by conventional or shock ausworking. Moreover, a shock-aging treatment can be effectively applied to a TRIP steel could be shock-aged to the same hardness level as that achievable by rolling at 850°F to an 80% reduction.

VIII. ACKNOWLEDGMENTS

The research reported herein conducted by R.N.O. (references 18-24) was sponsored by the U.S. Naval Air Systems Command. The writing of this review was supported by NSF Grant No. DMR 79102.

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