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A COMPARISON OF SHAPED CHARGE LINER CONE AND RECOVERED JET FRAGMENT MICROSTRUCTURES TO ELUCIDATE DYNAMIC RECRYSTALLIZATION PHENOMENA

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Introduction

In several previous publications, we have argued that reductions in the starting grain size of copper [1] and tantalum [2] shaped charge liner cones by as much as a factor of 10^2 in recovered slug and jet fragments following detonation provided unambiguous evidence for dynamic recrystallization (DRX) as proposed earlier by Meyers, et al [3, 4]. This evidence was based upon the fact that such a dramatic grain refinement (from roughly 35 μm to 0.3 μm) could only be explained by DRX especially in cases where the grain boundary misorientations averaged >15°. Furthermore, Chokshi and Meyers [4] argued that treatments of DRX by Sandstrom and Lagneborg [5] and Derby and Ashby [6] had developed similar expressions for a steady-state grain size proportional to the reciprocal square root of the strain rate. They utilized this proportionality along with other related assumptions to arrive at steady-state grain sizes between 0.1 and 0.01 μm for copper having undergone DRX at strain rates above 10^4 s^{-1}.

DRX in its broadest context is a high-temperature deformation phenomenon. In its simplest form, dislocations are created continuously during the recrystallization process which occurs during hot working. In much of the research which has treated DRX [7-9], the flow behavior of metals at elevated temperatures provides a signature of the presence of DRX since $\sigma \propto \sqrt{\rho}$, where σ is the flow stress and ρ is the dislocation density. These signatures are often differentiated by high or low strain rates ($\dot{\varepsilon}$), and they differ with strain (ε). Continuous or discontinuous DRX is a repeating process in which relatively strain free grains (having high-angle boundaries, misorientations, Θ, generally greater than 15 degrees) nucleate at regions of high dislocation density, grow, deform, and give rise to new nuclei. Depending upon the overall deformation conditions, this process may undergo several "cycles." Similar processes of nucleation and even cycles of nucleation characterizing DRX (or discontinuous DRX) can also occur on the grain boundaries, and the resulting recrystallization (DRX) can be observed as a kind of necklacing of smaller, recrystallized grains replacing some or all of pre-existing grain boundaries [10]. It has also been found that the complex interaction between work hardening and dynamic recovery during DRX leads to a relatively complex evolution of microstructure that cannot be described with simple analytic models [8]. Nonetheless, Rollett, et al [10, 11] and Peczak, et al [12] have applied Monte Carlo simulation techniques to investigate the effect of work hardening, nucleation rate, and strain rate on mechanical properties during DRX, as well as resulting microstructures.

The detonating shaped charge probably represents the most extreme example of DRX or discontinuous DRX. In contrast to more conventional hot working regimes where $\dot{\varepsilon}$ <10^2 s^{-1}, ε <50% (engineering strain), with flow stresses <10 GPa, the detonating shaped charge is characterized by $\dot{\varepsilon}$ ≥10^6 s^{-1}, ε >500%, and shock pressures in excess of 50 GPa. The adiabatic heating ($\Delta Q \propto (\varepsilon)(\dot{\varepsilon})$) during the liner cone collapse and jet and slug formation shown schematically in Fig. 1 has been estimated to exceed 0.7 T_M for copper [13]. In addition, the explosively generated shock pressure which collapses the cone likely produces high densities of defects (dislocations) at the outset, and a high-temperature pulse trails this high-pressure state and pre-heats the material as it is strained into the slug or jet, which is stretched to an instability condition creating necked fragments as a result of a time-varying velocity gradient along the elongating jet.

If DRX is a complex process, then DRX as a feature of the detonating shaped charge regime is an even more complex process which can only be inferred by examining the two end points: the starting cone microstructures and the recovered jet fragment or slug microstructures [1, 2]. Consequently, it might be reasonable to assume that by altering the variations which occur at these end points, additional insights into the shaped charge deformation process might be gained. Of course the simplest way to do this would be to alter the liner cone microstructure, especially the grain size. There is in fact a dearth of information about initial grain size variations or comparisons of starting and ending grain sizes in a wide variety of dynamic recrystallization studies for even conventional hot working.

Experimental Results and Discussion

In the research to be reported in this brief paper we compare a variety of shaped charge end point microstructures in copper, molybdenum, and tantalum; especially variations based on significant reductions in the starting liner cone grain sizes. These starting grain size reductions in some cases were achieved by processing liner cones by sputtering. This produced extremely small grain sizes (0.02 to 1 μm) in copper and molybdenum cones. We also produced a forged copper cone liner having a grain size of 15 μm, and the end point microstructures were compared with previous results and observations for a starting cone grain size of 35 μm. Microstructures in shaped charge starting liner cones and recovered jet fragments were observed both by optical metallography and transmission electron microscopy.

Figure 2 shows several comparative end-point microstructures for a number of copper shaped charge liner cones and recovered jet fragments. In Fig. 2(a) to (c), the grain size is refined from 35 μm in the inner cone wall to as small as 2 μm in the jet fragment cross-section. We might note that the grain size is an average of 50 measurements in several observations in the TEM which were considered "representative". Figure 2(d) to (f) shows a similar trend in grain refinement, but the grain size is larger in the jet while the starting cone grain size is smaller than Fig. 2(a) to (c). This appears to be reversed for Fig. 2(b) and (e) but there is a corresponding reversal in magnification as well. The TEM views in Fig. 2(c) and (f) show some residual deformation or stored energy in dislocations. In Fig. 2(g) to (i) a very small starting cone grain size is not refined in the recovered jet fragment, and this is also true for Fig. 2(j) to (l) for a more finely sputtered copper cone. In both Fig. 2(i) and Fig. 2(l), the ending (jet fragment) grains are devoid of microtwins (including growth and

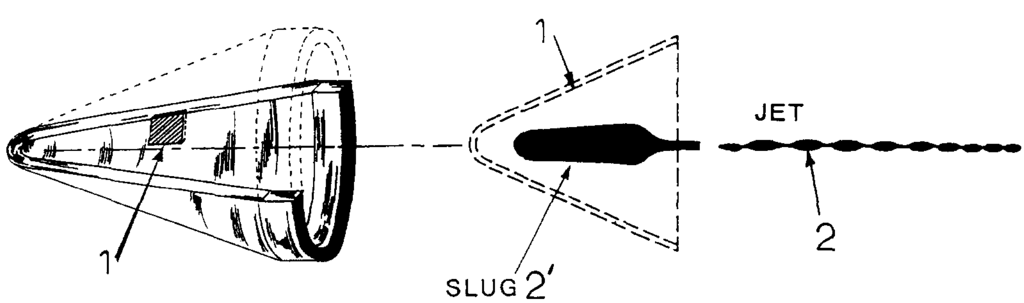

FIG. 1: Detonated shaped charge schematic showing microstructural end points marked 1 (starting liner cone) and the residual jet fragments (2) and slug (2').

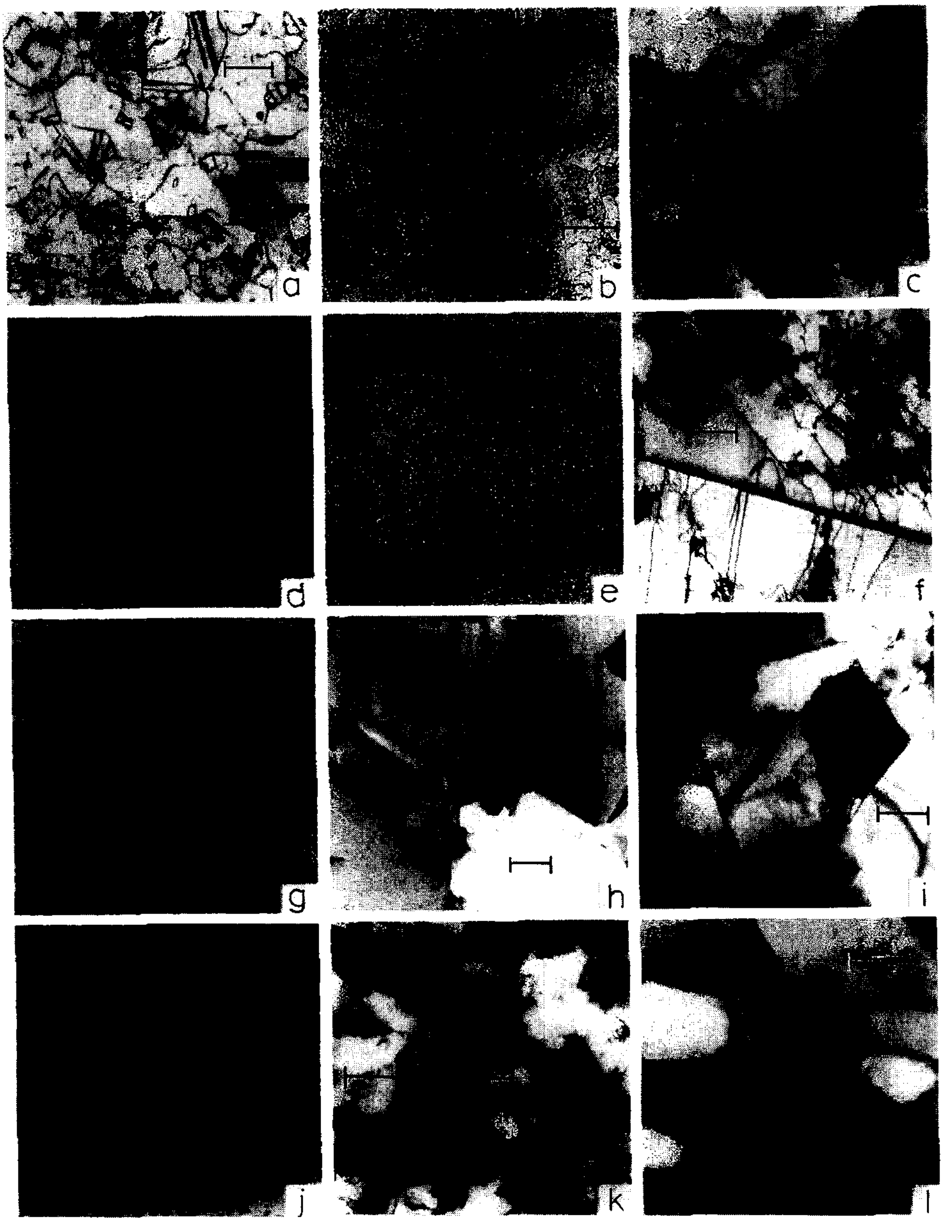

FIG. 2: Comparison of shaped charge end-point microstructures by optical metallography and TEM. (a) forged Cu cone #1, (b) jet fragment for (a), (c) is TEM corresponding to (b). (d) to (f) show a sequence for forged Cu starting cone #2 similar to (a) to (c). (g) to (i) and (j) to (l) show similar comparisons outlined above for two different sputtered Cu liner cones respectively. However (h) and (i) and (k) and (l) show TEM views for the corresponding starting cones and recovered jet fragments respectively. For all optical microscope views markers are as follows: (a) 40 μm, (d) 100 μm, (e) 1000 μm, (g) 10 μm, (j) 10 μm. All markers for TEM views remaining are 0.4 μm.

annealing twins) in comparison to the starting grain structure (Fig. 2(h) and (k) respectively). This is a particularly interesting observation because there are few twins in the entire residual microstructural regime, and this could only occur through recrystallization. However, there are many examples where the grain boundary misorientations are considerably less than 15°, and this either points up the fact that these microstructures consist of mixtures of dynamically recovered substructures and dynamically recrystallized grains or recrystallization in the shaped charge regime is quite different from that in ordinary hot working. Of course this process could be further complicated by static recrystallization immediately after the cessation of deformation when fragmentation of the jet occurs. However, in the absence of more direct and complimentary information, the nature of the recrystallization is not unambiguous and could involve evolution of microstructure implicit in concepts such as low-energy dislocation structure theory [14,15]. There is certainly enough time for requisite grain boundary movement since the jetting alone requires in excess of 100 μs [7-11].

The ending grain size in Fig. 2(i) and (l) is actually larger than the starting, liner cone grain size. It would appear generally from the comparisons provided in Fig. 2 for copper, that as the starting liner grain size is reduced to the micron range and smaller, the ending jet fragment grain size remains about the same or is increased in proportion as the starting grain size is reduced. This feature seems to be generally true for bcc metals as well, as shown in Fig. 3. Although Fig. 3 compares two different metals (Ta and Mo), the reduction of the starting Mo liner cone grain size to the micron range produces an ending jet fragment grain size which is essentially commensurate.

Table 1 summarizes a more quantitative analysis of grain sizes for each of the end-point microstructures shown in Figs. 2 and 3 for comparison. Table 1 also shows the ratio of starting liner (average) grain size, D_o, to the ending jet fragment (average) grain size, D_s. It is apparent from Table 1 that if we ignore the specific metals and look only at the grain sizes, smaller starting liner cone grain sizes produce correspondingly smaller grain size ratios, and correspondingly larger end-point jet fragment grain sizes.

This effect would seem to be associated with the larger volume stored energy associated with extremely small (sub-micron) starting grain sizes and microstructures which could be increased by the initial liner cone deformation in the detonating shaped charge. That is, the total volume stored energy would be expressed generally by a grain boundary contribution ($\Gamma\gamma_{gb}/D$), where Γ is a constant which includes a (grain boundary) geometrical factor, γ_{gb} is the specific grain boundary free energy, and D is the grain size. An additional term would include the uniform stored energy due to deformation-induced defects (most prominently dislocations or dislocation density). It is likely that deformation also changes γ_{gb} as well as the geometrically-related constant Γ. Certainly the shaped charge detonation/deformation process is not an equilibrium situation, because the necked jet fragments as well as the final slug (Fig. 1) contain a significant amount of residual, stored energy. Consequently, the initial, uniform stored energy will have a significant effect on the recovery and recrystallization processes or the overall discontinuous dynamic recrystallization process as may be the case. Since there is apparently always residual stored energy in the shaped charge regime, there is never normal grain growth. However there may be some kind of deformation enhanced grain growth as described by Sherwood and Hamilton [16].

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References

1. L. E. Murr, H. K. Shih, C-S. Niou, and L. Zernow, *Scripta Metall. et Materialia, 29,* 567(1993).
2. H. K. Shih, L. E. Murr, C-S. Niou, and L. Zernow, *Scripta Metall. et. Materialia, 29,* 1291(1993).
3. M. A. Meyers, et al, *J. de Physique, IV,* pp. 3-11(1991).

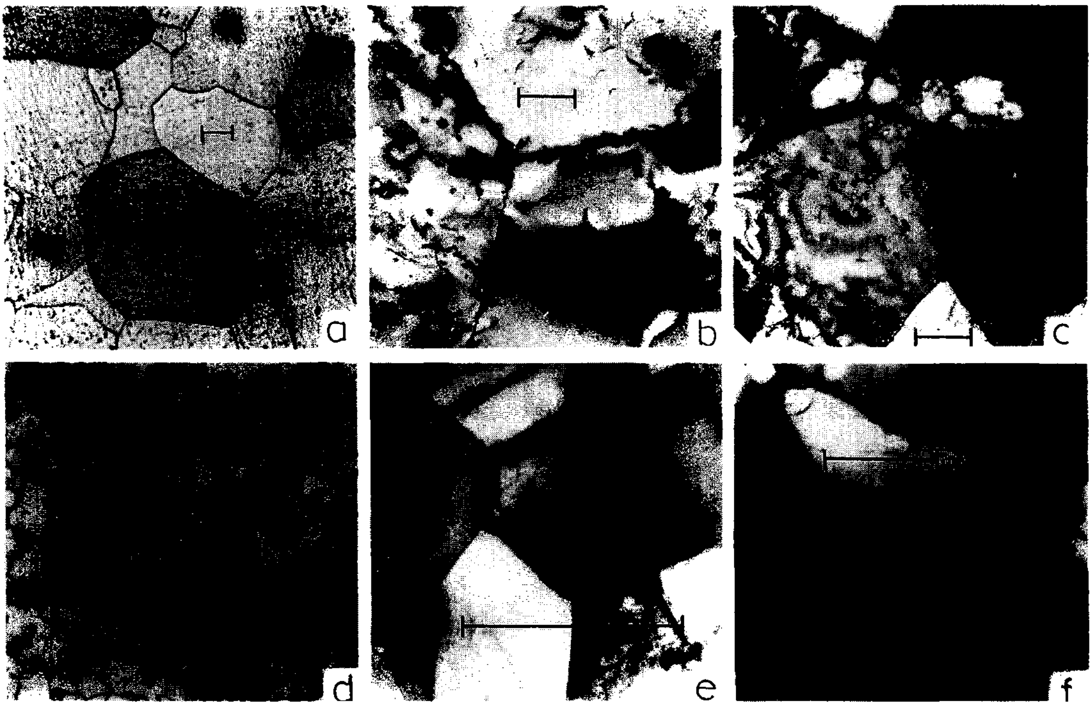

FIG. 3: Comparison of forged Ta((a) to (c)) and a sputtered Mo((d) to (f)) end-point microstructures respectively as in Fig. 2. (a) and (b) correspond to the Ta cone, (c) is a TEM of Ta jet fragment. (d) and (e) correspond to the Mo cone, (f) is a TEM of Mo jet fragment. In (a) and (d) the marker is 10 μm. In the TEM images (b), (c), (e) and (f) the marker is 0.4 μm.

Table 1: Comparison of End-Point Grain Sizes in Detonated Shaped Charges

System (Figs. 2 and 3)	D_o(μm)*	D_s(μm)*	D_o/D_s	System (Figs. 2 and 3)	D_o(μm)*	D_s(μm)	D_o/D_s
Forged Cu (1) Fig. 2(a) to (c)	35	2	18	Sputtered Cu(2) Fig. 2(j) to(l)	0.1	1.5	0.07
Forged Cu (2) Fig. 2(d) to (f)	15	8	1.9	Forged Ta Fig. 3(a) to(c)	35	0.35	100
Sputtered Cu (1) Fig. 2(g) to (i)	1	1.5	0.7	Sputtered Mo Fig. 3(d) to(f)	0.5	1	0.5

*D_o is the initial, average liner (inner wall) grain size while D_s is the smallest, average residual grain size observed in the corresponding, recovered jet fragment; considered to represent the steady-state DRX grain size. Grain sizes were measured by averaging maximum and minimum lengths for a statistical sampling of individual grains.

4. A. H. Chokshi and M. A. Meyers, *Scripta Metall. et Materialia, 24,* 605 (1990).

5. R. Sandstrom and R. Lagneborg, *Acta Metall., 23,* 387(1975).

6. B. Derby and M. F. Ashby, *Scripta Metall., 21,* 879(1987).

7. R. W. Cahn, *Recovery and Recrystallization,* Physical Metallurgy, Part II, Ed. R. W. Cahn and P. Haasen, Elsevier, New York, 9183, p. 1595.

8. T. Sakai and J. J. Jonas, *Acta Metall., 32,* 189(1984).

9. J. J. Jonas, C. M. Sellars and W. J. Mc. G. Tegart, *Metall. Rev., 14,* 130(1969).

10. A. D. Rollett, M. J. Lston, and D. J. Srolovitz, *Acta Metall., 40(1),* 43(1990).

11. A. D. Rollett, D. J. Srolovitz, and M. P. Anderson, *Acta Metall., 37(4),* 1227(1989).

12. P. Peczak and M. J. Lutton, *Acta Metall. et Materialia, 41(1), 59(1993).*

13. L. Zernow and L. Lowry, Chap. 46 in *Shock Wave and High-Strain-Rate Phenomena in Materials,* Ed. M. A. Meyers, L. E. Murr, and K. P. Staudhammer, Marcel Dekker, Inc., New York, 1992, p. 521.

14. D. Kuhlmann-Wilsdorf, *Phys. Stat. Sol. (a), 104,* 121 (1987).

15. D. Kuhlmann-Wilsdorf, *Mater. Sci. Engr., A113,* 1 (1989).

16. D. J. Sherwood, and C. H. Hamilton, *Scripta Metall. et Materialia, 25,* 873(1991).