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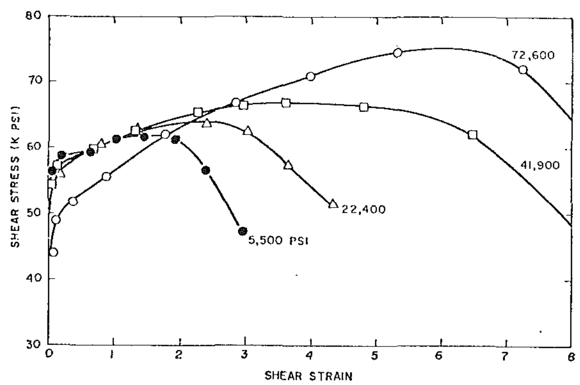

Fig. 12 Variation of shear stress with shear strain with different amounts of normal stress on the shear plane. Material = 0.08 percent C plain carbon steel.

DISCUSSION

M. C. Shaw[2]

The author's photographs of the backs of metal cutting chips are very interesting and clearly reveal the inhomogeneous character of the strain involved in cutting. This gives rise to what the author calls "shear fronts" which are regions of material separated by regions of large relative displacement.

In discussing the origin of this displacement the author refers to behavior in a tensile test of mild steel where three stages of deformation may be identified. In all of these stages plastic flow due to the motion of dislocations is the predominant mode of deformation. Only at the end of the third stage does a new mechanism appear associated with the formation of microcracks. Since the normal stress on the shear plane in a tensile test is tensile, there is little tendency for cracks to reweld and rupture occurs soon after microcracks first appear.

This is not the case in metal cutting, however, where strains can be very much higher without gross rupture due to the presence of a large normal stress on the shear plane. A materials test more appropriate to this situation is one involving linear shear with compression on the shear plane. A recent paper[3] describes such tests which lead to results such as those shown in Fig. 12. Here four curves are shown corresponding to different values of normal stress. In addition to the three stages identified for the tensile test, a fourth is clearly present here that involves negative strain hardening before gross fracture occurs. Above a strain of about one (for the curve with compressive stress of 22,400 psi on the shear plane) microcracks appear. As motion ensues microcracks reweld and reform but the number of microcracks increase with strain until gross fracture occurs.

It is now believed that the strain in a continuous metal cutting chip is not predominantly due to the movement of dislocations

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but is instead due to the movement of microcracks along planes of predominant weakness. This gives rise to the "shear fronts" shown by the author in his Figs. 1 to 3. The saw-toothed segmental chip shown in Fig. 8 is one where gross fracture followed by rewelding has occurred.

The purpose of this note is to caution against extending tensile results and dislocation theory which involves small plastic strains into the large strain regime of metal cutting where an entirely different mechanism of deformation involving microcracks pertains.

Author's Closure

The author wishes to thank Professor Shaw for his comments on this paper and for his interest in this work. Professor Shaw suggests that microcracks may be one of the contributing mechanisms by which a metal cutting chip is formed and, thus, that metal cutting is a form of ductile fracture. Indeed, the paper [27][8] referred to in Professor Shaw's comments does reiterate two very profound practical points about the phenomena of ductile fracture. First, in ductile fracture, the propagation of a ductile crack involves substantial plastic flow (whereas in a brittle fracture, the crack can proceed with a minimum of further plastic deformation). Second, the dislocation behavior in a metal or alloy is dependent on the exact conditions of stressing and thus the superposition of a hydrostatic pressure serves to increase the degree of necking prior to failure (making the material effectively more ductile). This and other similar evidence indicates that a strain criterion for ductile fracture is not valid and that regardless of the amount of cold work a material has received, it will not crack in ductile manner unless further plastic strain is imposed under favorable stress conditions

What is not clear from Professor Shaw's reference nor his comment is (*a*) by what mechanism such cracks can develop; (*b*) how and where they were observed; and (*c*) why they should appear only in the "planes of predominate weakness." (The author confesses a certain opaqueness as to the meaning of the latter term and hopes that Professor Shaw will produce a more precise description of it in his future paper on this subject.) The author will now address himself to questions (*a*) and (*b*), leaving (*c*) for Professor Shaw, as it relates directly to his work.

Ductile fractures can be the result of inclusions in the material, acting as nucleating centers for voids or microcracks. Fractography examinations have verified this mechanism and have shown that the main characteristic of the cup region of a ductile fracture is the formation of a continuous pattern of dimples or shallow depressions on each surface of the fracture which are clearly the result of the linkage of the cavities formed in the necked region. That is, the atomic planes on the faces of the shear zone move during deformation and form a cavity around the inclusion. This has been observed in metal cutting (see reference [7] or Fig. 21, reference [3]). Cavity forming inclusions are deliberately introduced into some metals to reduce the ease with which dislocations can move through the material, thereby reducing their ductility and so improve their ability to be machined (free cutting metals). This mechanism, however, appears to be only an adjunct mechanism since very high purity materials deform in the same fashion (shear front—lamella) as commercial purity metals.

There is also much experimental evidence to support the concept that microcracks may be developed by dislocation interactions during plastic deformation and a number of mechanisms have been advanced based on pile-ups, split bend planes, and twins. Thus, the glide mechanism may enable running cracks to be created suddenly from avalanches of glide dislocations, with concomitant acoustic emission.

[8] Numbers in brackets designate Additional References at end of last closure.

It would seem, however, that Professor Shaw has unfortunately misconstrued the reference to the three stages in plastic deformation in the interpretation of the micrographs. The three stages referred to are not connected with mild steel but with the tensile strain behavior of f.c.c. metal single crystals. In such tests, it has been well established by Schmid [28], Polanyi and Schmid [29], and others in the 1920's that the shear stress for yield is independent of the normal stress on the shear plane, be it in compression or in tension. This behavior has been verified for hexagonal crystals, cubic crystals and crystals of lower symmetry [30]. There is now no new scientific evidence to take a contrary view.

Concerning the second point, Professor Shaw refers to a recent paper co-authored by him [27] describing the deformation of certain metals at large strains which purports to show micro-cracks and negative strain hardening in course of the test. This paper fails to provide any direct experimental evidence for micro-crack formation. In fact, the authors state ". . . However, at this time it is suggested that the real area of contact is smaller than the apparent area due to the presence of micro-cracks" (page 245). This speculation is presumably supported by acoustic emission.

However, a study of the schematic drawing of the apparatus used raises serious questions regarding the source of the acoustic emission. It is well known that the low level acoustic emission is both difficult to detect and to interpret. Few materials generate sufficient emission to be audible above the machine noise and it is necessary to silence the machine to hear the emissions [31]. Specimen holding and loading fixtures call for special attention to exclude artifact emissions from the testing apparatus. The file grips (scored clamping blocks) used by Prof. Shaw are specifically considered unsatisfactory by Tatro [32]. The relaxation of the test apparatus and the growing zones of plasticity in the specimen at the clamps may very well be the source of emissions observed by him. To eliminate such difficulties, many workers frequently resort to hydrostatic loading [33], and loading by thermal expansion [34]. When such techniques are not used to locate the source of acoustic emission, time-coincident observations and triangulation methods are needed [35]. In view of this, Prof. Shaw is correct only in suggesting that micro-crack formation may occur; he has not presented evidence to claim that they indeed occur.

Transmission electron microscopy and scanning electron microscopy of the machined chips by the author [1, 2, 3] as well as by others [36, 37, 38] fail to reveal any microcracks in the chip. Thus, the evidence for the absence of microcracks in the chip is conclusive. It is therefore clear that the "shear fronts" have nothing to do with the micro-cracks postulated by Prof. Shaw.

In summary, the author believes that in view of the present extensive literature (theoretical and experimental), dislocations need not be defended as the central mechanism for plastic deformation even up to fracture, as it is abundantly clear that fracture requires prior activity of dislocations. (In this connection it is interesting to note that the current, accepted theories of acoustic emission are themselves based on the dislocation theory. The relevant papers by P. P. Gillis [39] and A. S. Tetelman [40] are given below.) One must abide by the dictum of Ockham's razor—dislocation mechanisms are sufficient to explain the appearance of the chip and the plastic deformation which occurs during metal cutting and necessitates no other entity.

B. F. von Turkovich[4]

The paper by Dr. Black is a very important contribution toward the understanding of large plastic deformation. I wish to compliment him for the excellent scanning electron microscope photography of metal cutting chips.

The present work shows conclusively that attention must be paid to the detailed morphology of deformed material. The SEM micrographs clearly illustrate the fundamental similarity of chip formation in ductile materials, i.e., the formation of lamellar structure. In addition, the lamellar structure is remarkably regular and always well developed. The form and fine scale details of the lamallae vary, however, with different materials, as illustrated in the author's Figs. 3, 5, and 10. This fact alone is of great value in the theoretical study of plastic deformation. Also, the lamellae are separated by much slimmer layers which the author called "the shear fronts." It seems very plausible that the lamellae are deformed deeply into stage 3 of plastic flow. The shear fronts can be interpreted as layers where the plastic deformation has reached even higher strains together with a high temperature rise due to increased strain rate. A probable dislocation mechanism for such a process has been discussed in reference [26]. It is possible to carry this analysis further and arrive at a theoretical explanation of the entire process.[5]

Fig. 8 of the side of a 1020 steel chip shows the saw-tooth pattern very clearly. The pattern is limited to the edge of the chip, which appears to be the side torn from the workpiece. Is this inference correct? What explanation can be given for this particular aspect? Do such serrations occur also in f.c.c. metals?

I hope that the author will continue his investigation of this very interesting problem in plasticity. I congratulate him on his present work.

Author's Closure

The author also wishes to thank Professor von Turkovich for his generous comments concerning the merit of this work, and would like to state at this time that he has carefully considered the theoretical analyses referred to in the comments and the paper. In fact, with very minor modification in geometry, the theoretical model proposed by Professor von Turkovich in Fig. 3 of reference [26] aptly describes the deformation that has been observed in micrographs from hundreds of metal cutting chips. The most remarkable features of these studies is the consistency of the shear front-lamella structure in periodicity and magnitude. There are, however, differences in lamella width and regularity which one may be able to relate to certain metallurgical parameters, in particular, stacking fault energy and dislocation mobility in a particular metal or alloy. The author would agree with Professor von Turkovich that the shear fronts are indeed layers where the plastic deformation has reached high levels of strain and temperature due to the high levels of strain rate inherent in the process. The author is attempting to devise experiments wherein such rates can be directly measured via electron microscopy techniques, but such dynamic experiments are extremely difficult.

In his reference to the saw-toothed serration seen in Fig. 8, Professor von Turkovich has touched upon one of more facinating geometrical features of steel chips. The pattern does indeed appear on that side of the chip closest to the side cutting edge, which is to say his inference is correct. Such serrations or segmentation can often be observed in chips with the naked eye and have been observed in the SEM in aluminum and copper chips produced by turning (see Figs. 24 and 25 of reference [3]). The serrations are much larger than the lamella and are usually very periodic in steel chips. The pattern extends somewhat into the bulk of the chips and a lamella structure can be observed to exist within each serration. I do not believe that these serrations are the product of cutting or tool vibrations because they are so very periodic in form and do not extend across the bulk of the chip. Rather, I think that the serrations are the result of the changing of the applied loading from a plane strain condition in the

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bulk of the chip to a plane stress situation at the separation edge of the chip, complicated by constraints imposed on the flow by the workpiece shoulder. One observes that the lamella thin down and merge into the serration as they approach the side of the chip, suggesting that the dislocations on the shear fronts are being influenced by a different applied stress condition. Obviously, additional study of this feature is required.

J. J. Jonas[6]

The author has presented a series of scanning electron micrographs which throw much light on the surface morphology of machined chips. In particular, they reveal the structure and spacing of "shear fronts" with greater clarity and resolution of detail than that provided by optical techniques. The excellence of the micrographs reintroduces the important question of just how the lamellae and shear fronts are produced. A number of alternative mechanisms present themselves for consideration.

In this paper, the author favors the mechanism of extensive single slip as the one largely responsible for lamella formation. The purpose of this discussion is to present some arguments in favor of an alternative mechanism, that of adiabatic shear [41],[7] to which the author refers briefly.

The phenomenon of "adiabatic shear" was first investigated in detail by Zener and Hollomon during a study of projectile impact [42–44]. They noted that the process of homogeneous shear can become unstable because of adiabatic heating, after which all further deformation is restricted to a narrow zone of highly localized shear strain and strain rate. The likelihood of adiabatic shearing is enhanced by large strains, such as those encountered in machining, or in the necked region of tensile specimens. This is because the rate of isothermal work hardening decreases with strain, so that the additional decrease in work hardening rate produced by adiabatic heating is more likely to make the net work hardening rate negative as the strain increases. A locally negative rate of work hardening, in turn, leads to localization of the strain. The likelihood of adiabatic shearing is also very much greater at high strain rates (e.g., $>10^3$ sec^{-1}) than at tensile testing strain rates ($\sim 10^{-3}$ sec^{-1}). The possibility that adiabatic shearing is prominently involved in the formation of the lamellae must therefore be examined more closely.

The arguments against the single slip theory can be summarized as follows:

1 The author notes that large scale dislocation activity occurs in the workpiece prior to its entrance into the region of lamella formation. Such large scale dislocation activity, taking place as it does in a region of constrained deformation, almost certainly involves slip on several slip systems (multiple slip). However, it should be noted that single slip (slip on a single slip system) in single crystals, if it occurs, always precedes and never follows multiple slip. Thus the prior occurrence of multiple slip during compressive deformation makes subsequent single slip on entry into the shear zone rather unlikely.

2 The crystallographic mechanism of single slip involves local slip plane orientations which vary fairly widely from grain to grain. Such slip planes will only rarely contain the surface normal to the direction of tool travel. Thus considerable amounts of cross slip, that is, activity on other slip systems, would be required to produce the macroscopically observed shear fronts by this means, as the author has in fact suggested in his paper. However, the simultaneous occurrence of such extensive cross slip would again make single slip or easy glide rather unlikely. By contrast, the observation that the lamellae are always perpendicular to the direction of travel of the cutting tool can be taken to favor the macroscopic mechanism of adiabatic shear, which is not tied to a particular crystallographic slip plane.

In summary, it appears from the published information that the mechanism of adiabatic shear probably plays an important role in the machining process described by the author. It therefore merits more serious examination. To clarify the alternatives, more transmission electron micrographs are required, particularly of the zone of compressive deformation in advance of the cutting tool. The extent of dislocation activity on secondary slip systems can then be assessed more fully, which would permit more reliable conclusions to be reached about the plausibility of single slip.

Author's Closure

The author would like to thank Professor Jonas for his informed commentary on this paper. Indeed, it is gratifying to those individuals who have ventured into plastic deformation studies of manufacturing processes as viewed from dislocation mechanics to know some knowledgable readers have taken the time to examine these efforts. Professor Jonas's kind words regarding the scanning electron micrographs are also appreciated. One observes that the fine structure seen herein could not have been observed prior to the application of scanning electron microscopy methods to these studies, since these deformation structures have morphological features which exceed the depth of field characteristics of even the best optical light microscopes at the magnifications required to resolve such structures.

This paper has concerned itself primarily with the presentation of micrographs which reveal the structure to always be shear front-lamella in nature and not so much with the question of just how these characteristic features are produced, as indicated in my summary statements 1 through 6. Reference [26] to which Dr. Jonas has referred, concerning the adiabatic shear theory developed for the metal cutting process by von Turkovich, is the only theory which has received extended development. Irrespective of the model selected to describe the process, the dislocation activity which occurs in the workpiece prior to its entrance into the region of shear front activity, undoubtedly produces dislocations which serve to block the dislocations moving on the shear front. In fact, Lomer Cottrell blocks and jogs are a distinct possibility based on identation studies of cube faced copper [1] as extended to metal cutting.

Again, however, because the crystal is constrained from rotation, the shear deformation is constrained to a small region which must deform to satisfy the displacement imposed upon the crystal by the tool so the dislocation activity at the shear front predominates, whether or not it has reached an adiabatic state. It has, for some time, been my conviction that the tool tip or cutting edge must be one of the very strongest sources of dislocations possible, in that two new surfaces are created with extensive levels of plastic deformation in both the chip and the workpiece.

I cannot quarrel with the rest of Dr. Jonas's comments except to say that I seem to have unfortunately mislead him by what appears to be an explanation of metal cutting phenomena in terms of stress-strain diagrams, when actually, my intent here was to present those dislocation mechanisms which are favorable to the metal cutting problem in terms of their roles in tensile and compressive deformations. In this light, it was suggested that a single set of slip planes may be adequate to account for the observed deformation when microtomy levels of depth of cut are employed. Here chips whose thickness (typically less than 2000 A) is an order of magnitude less than the normal lamella spacing (typically 2 microns) could easily be formed by those dislocation (or segments thereof) produced at the tool tip and driven to the free surfaces. Even here, it is unlikely that a single slip system (that is, a (111) ⟨110⟩ combination) is operative, unless the cutting edge were absolutely perfect. More likely, 2 or 3 slip directions are activated. The product of their intersection is mobile and they may cross slip to satisfy the applied stress, with the ability to cross slip being controlled by the stacking fault energy of the material.

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[7] Numbers in brackets designate Additional References at end of discussion.

At this writing, the theory developed by von Turkovich is fairly complete but space does not permit even a cursory review here. While many of the finer details of the theory (such as the influence of stacking fault energy and the interaction of dislocations with boundaries, second phase particles, and so forth) must still be resolved, the constancy of the results produced by the theory (especially when compared to the physical evidence generated by microscopy methods) is hard to refute. Professor von Turkovich and I are presently preparing a manuscript which summarizes our joint efforts to date.

In our most recent discussions with Dr. S. Ramalingam from the State University of N.Y. at Buffalo, it was concluded that it may be misleading to interpret metal cutting in the conceptual workhardening mechanisms associated with the stress strain curves, since strain itself is not a readily measureable feature of the metal cutting process while displacement is.

Some additional physical observations are worth mentioning at this time. The effect of crystal orientation can be observed in very thick chips of copper and aluminum cut at low cutting speeds which suggests that the process may not be appropriately labelled as adiabatic when the material is a good heat conductor and cut at slower speeds. Also, during microtomy, the chips are typically flat but as the knife edge deteriorates or as the depth of cut is increased, the chips begin to curl and continue to be curled until depths of cut of 1 micron or so are reached, where the chips start to flatten out again. In conjunction with these observations, it should be pointed out that metal cutting is not strictly a symmetric deformation process even in the orthogonal case when one considers the dislocation activity produced at the tool tip. It may well be that there is a predominance of edge dislocations of one sign generated by the tip source which after annihilation processes of the symmetrical segments leaves a large number of dislocations of one sign in the chip. Additional dislocation activity occurs at the tool-chip contact region due to friction plastic deformation which will populate the bottom side of the chip with a high density of dislocations. Either of these two situations could account for the curled nature of the chip and explain why chips are alternately curled or flat.

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