FORMATION, MICROSTRUCTURE AND DEVELOPMENT OF THE LOCALIZED SHEAR DEFORMATION IN LOW-CARBON STEELS

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(Received 8 February 1995; in revised form 18 July 1995)

Abstract—An investigation has been made into the effect of microstructural parameters on the propensity for forming shear localization produced during high speed torsional testing by split Hopkinson bar with different average rates of 610, 650 and 1500 s⁻¹ in low carbon steels. These steels received the quenched, quenched and tempered as well as normalized treatments that provide wide microstructural parameters and mechanical properties. The results indicate that the occurrence of the shear localization is susceptible to the strength of the steels. In other words, the tendency of the quenched steel to form a shear band is higher than that of the other two steels. It is also found that there is a critical strain at which the shear localization occurs in the steels. The critical strain value is strongly dependent on the strength of the steels. Before arriving at this point, the material undergoes a slow work-hardening. After this point, the material suffers work-softening, corresponding to a process during which the deformation is gradually localized and eventually becomes spatially correlated to form a macroscopic shear band.

Examinations by SEM reveal that the shear localization within the band involves a series of sequential crystallographic and non-crystallographic events including the change in crystal orientation, misorientation, generation and even perhaps damage in microstructures such as the initiation, growth and coalescence of the microcracks. It is expected that the sharp drop in the load-carrying capacity is associated with the growth and coalescence of the microcracks rather than the occurrence of the shear localization, but the shear localization is seen to accelerate the growth and coalescence of the microcracks.

The thin foil observations by TEM reveal that the density of dislocations in the band is extremely high and the tangled arrangement and cell structure of dislocations tends to align along the shear direction. The multiplication and interaction of dislocations seems to be responsible for work-hardening of the steels. The avalanche of the dislocation cells corresponds to the sharp drop in shear stress at which the deformed specimen is broken.

Double shear bands and kink bands are also observed in the present study. The principal band develops first and its width is narrower than that of the secondary band.

INTRODUCTION

When the work-softening rate arising from an increase in temperature exceeds the work-hardening and strain rates in a material which was subjected to deformation at high strain rates, the plastic strain might be limited into a narrow region, termed the “adiabatic shear band”. The term “adiabatic” is only used because a small amount of heat is generated by the motion, multiplication and interaction of dislocations are considered to be transformed into the surrounding matrix during ongoing deformation.

It should be pointed out firstly that because of the same mechanisms of localized plastic flow, however, the slip and shear bands are both easily confused with each other. A great deal of investigation indicates that they can be distinguished at least in the following aspects:

(1) The slip bands usually occur on a defined crystallographic plane and are often limited into one

grain with a defined crystallographic direction. However, the development of the shear localization shows a non-crystallographic deformation feature, arising from the microscopic areas of misorientation that could have been generated by the operation of multiple slip systems and often involves over 10 or 100 grains. In other words, the shear band starts with a crystallographic slip event within an individual grain but expands over the whole cross-section of the specimen along the shear direction with a non-crystallographic feature.

(2) Thermal softening caused by the rise in temperature in the band plays an important role in the occurrence of the shear localization but it is not considered to be necessary for the slip band formation.

Since the shear band was firstly recognized in 1944 [1], the investigations on the shear localization have, in general, focused on the following two aspects.

Fig. 1. The profile and dimension of the specimen in millimetres.

Scientists in mechanics have focused their efforts on the modeling description developing a criterion for instability of plastic flow, according to the qualitative analysis of Zener and Hollomon [1], Recht [2], Culver [3], Bai [4], Stake [5], Wu [6], Fressengeas [7] and Clifton [8]. However, metallurgists have emphasized the relationship of the shear localization to the microstructures, pursuing the role of the localized shear in engineering applications [9–27].

It is known that materials failure by shear localization frequently occurs in matching, forming, punching, high speed deformation and ballistic impact in many materials, such as aluminium alloys, titanium and its alloys, copper, brass, plastics and most commonly in steels. Thus, the mechanism of the localized shear formation has been greatly studied.

The purpose of this study is to investigate the effect of the microstructural parameters on the formation of the shear localization in low-carbon steels.

MATERIALS AND PROCEDURE

The steel used in this study has a composition (wt.%): 0.22%C, 0.009%S, 0.35%Si, 0.53%Mn, 0.018%P and balance Fe. It was thermally treated as follows:

- at 850°C holding for 30 min followed by air-cooling;
- at 850°C holding for 30 min followed by quenching in water;
- at 850°C holding for 30 min, followed by quenching in water and then tempering at 300°C for 30 min.

The microstructures corresponding to these treatments are F + P (ferrite + pearlite), F + M_q

Fig. 2. Shear stress-strain response for the three steels. A: quenched; B: quenched and tempered and C: normalized steels.

(ferrite + quenched martensite) and F + M_t (ferrite + tempered martensite).

The specimen used in high speed torsional testing is a thin-walled tube machined from steel bars, with integral circular flanges matching the torsional impedance of the aluminium bar. It has a gage length of 2.0 mm, an internal diameter of 10 mm and a wall thickness of 0.8 mm. The details of the specimen are given in Fig. 1. It should be emphasized that the inside and outside wall of the specimen must be polished carefully prior to testing to remove the surface defects. In order to explore the developing process of the shear localization, fine lines are scribed on the inside surface of the wall parallel to its shear stress axis.

The torsional tests were performed on all three steels in a split Hopkinson bar with the average strain rates of 610, 650 and 1500 s⁻¹. The basic apparatus parameters can be obtained from a previous paper [18].

In order to investigate the relation of the microstructures to the developing process of the shear localization, the torsional tests were interrupted at different strains and then the specimens machined from the deformed cross-section were used to characterize the microstructure by TEM and SEM. For comparison, the mechanical properties of the steels were also determined and are listed in Table 1.

RESULTS AND DISCUSSION

Shear stress-strain response

Figure 2 is a typical shear stress-strain response that occurs during dynamic loading in the three

Table 1. Mechanical properties and microstructural parameters of the three kind of steels tested

Microstructure	σ_F (MPa)	σ_Y (MPa)	n	δ (%)	$\dot{\gamma}$ S ⁻¹	τ^* (MPa)	γ^*	γ_c	W_M (μ m)	ΔT (°C)
F + M _q	699	427	0.179	22.5	650	550	1.5	0.30	75 ~ 100	219
F + M _t	649	426	0.189	23.5	610	510	1.72	0.36	130 ~ 150	265
F + P	498	325	0.252	35.9	1500	530	1.95	0.83	350 ~ 400	275

σ_F : Fracture strength; σ_Y : yield strength; n : work-hardening exponent; δ : elongation; $\dot{\gamma}$: strain rate; τ^* : shear strain; γ^* : average strain in the band; γ_c : average critical shear strain; W_M : width of the band; ΔT : the rise in temperature in the band.

steels. It appears that the torsional stress-strain response has almost the same trend. In other words, they were all subjected to work-hardening, after arriving at the maximum stress, corresponding to the critical strain, the specimens undergo slow work-softening with further straining and finally the load-carrying ability of the specimen drops rapidly until fracture occurs along the shear direction. The major difference in stress-strain response is that the critical

strain which corresponds to the occurrence of the shear localization is strongly dependent on the strength of the steels.

It can be found from Table 1 that the critical strain values for the quenched, quenched and tempered, and normalized steels are 0.30, 0.36 and 0.83, respectively. This implies that the higher the strength of the steels, the more susceptible they are to the shear localization. These results are in good agreement with the

Fig. 3. Localized shear bands in the low-carbon steels: (a) normalized steel; (b) quenched and tempered steel; (c) quenched steel.

Fig. 4. Double shear bands observed in quenched and tempered steel.

observations of Costin *et al.* in cold rolled steel that the shear band develops at a critical strain of 0.30 [16] and Clifton in hot rolled steel at a critical strain of 0.80–1.0 [22] and also similar to the result of Hartley *et al.* [23].

Temperature rise within the shear band

It is known that high speed plastic deformation is caused by creation, multiplication and interaction of dislocations. Assuming that a small part of the work of plastic deformation is stored in the material as elastic strain energy, the remaining 90–95% is converted into heat [9]. If the shear localization is proposed to be an “adiabatic” or a “quasi-adiabatic” process the heat loss in the band may be neglected. In other words, the plastic work is converted into heat completely. Therefore, the temperature rise

within the band may be calculated by the following equation:

$$\Delta T = \tau^* \gamma^* / \rho C_v$$

where τ^* and γ^* correspond to the shear stress and shear strain, respectively, ρ is the material density and C_v is the heat capacity.

The results calculated from the three steels are 219, 265 and 275°C, respectively. As a matter of fact, the temperature rise in the band may be somewhat higher than these values because of thermal diffusion.

Formation of the shear band

Figure 3 displays the shear localization observed in the three steels. It may be found that the quenched steel with high strength and low work-hardening capability does develop a pronounced shear band

Fig. 5. Developing process of the shear localization corresponding to different loading times. (a) 750 μs; (b) 850 μs; (c) 890 μs in the quenched and tempered steel.

Fig. 6. Formation of the kink bands caused by the reversed loading.

with a width of 70–100 μm shown in Fig. 3(c). However, the width of the band in the normalized steel with high capability of work-hardening is much wider than those of the other two steels, approximately 300–400 μm . The width in the quenched and tempered steel is in between them. It should be mentioned that the boundary of the band and the alloy matrix is not well defined but as relative determinations these values are available.

Costin *et al.* [16] found that the localized shear band developed in the cold rolled steel, but was not formed in the hot rolled steel when deformation is limited to about 30% strain. Shawki *et al.* [19] considered that the reason no localized shear was found in the hot rolled steel may be attributed to the low strain they used. They predicted that if the nominal strain increased to 80–100%, shear localization could also be formed in the hot rolled steels. This proposal was justified by the results of Shawki [24] and Hartley [23]. However, they pointed out that the width of the band in the hot rolled steel is narrower than that in the cold rolled steel. This is quite different from the result in the present study,

showing that the smaller the work-hardening capability, the narrower the shear band is.

It is interesting to find that the double localized shear bands developed in all three steels. Figure 4 represents a typical field which was found in the quenched and tempered steel. It can be noted that the development of these two bands is not of equilibrium. One of them develops firstly into the major band and the secondary band develops slowly. Determinations show that the width of the principal band is about 100 μm with a local strain of 1.84 and the secondary one is about 250 μm with a local strain of 2.75. This deformation phenomenon was also found in Ti-alloy during dynamic loading [18]. The mechanism for the occurrence of the double bands in one piece of the tested specimen is not clear and further work is still needed.

Evolution and development of the shear localization

In order to characterize the developing process of the shear localization produced during dynamic loading, interrupted tests are conducted on all three steels. It was found that when the average critical strain is

Fig. 7. Precipitate rotation in the band towards the shear direction.

Fig. 8. Microcrack initiation at the interfaces between the ferrite-inclusion (indicated by A) and the ferrite-cementite (indicated by B).

Fig. 9. Fractographs of the three steels. (a) Quenched martensite; (b) quenched and tempered martensite; and (c) ferrite plus pearlite.

Fig. 10. The interfaces between the ferrite-cementite and the ferrite-pearlite as the sources release a great deal of dislocation.

Fig. 11. Hexagonal network formation by the interaction of dislocations.

approximately 0.34, corresponding to the loading time of 550 μ s, the prescribed line does show homogeneous deformation and there is no shear localization to be found on the cross-section of the specimen at this moment, implying that the specimen tested still has work-hardening capability. However, as shear strain increases to about 0.36, corresponding to the loading time of 750 μ s, shear localization does occur. After this point, the specimen appears to be work-softening with increasing shear stress. Figure 5 shows a developing process of the shear localization in the

Fig. 13. Avalanche of dislocation cells corresponding to the broken specimen.

quenched and tempered steel, corresponding to different loading duration. The width of the bands are 310,

Fig. 12. Elongated dislocation cells along the shear direction.

Fig. 14. Dislocation cells and non-crystallographic deformation feature in the localized shear band in the quenched steel.

110 and 100 μm . This implies that the localized deformation is a progressive process during which the shear deformation becomes gradually localized. The longer the loading time, the narrower the band is.

One point that should be mentioned is that the reversed load plays an important role in the shear deformation. Figure 6 gives a typical example indicating the formation of the kink deformation bands which is caused by the reversed load during loading.

Damage and fracture

The initiation of the localized shear band is proposed to start with a crystallographic slip in a favored individual grain and then it propagates into the adjacent grains by co-operative slip or cross-slip [20, 21], leading to spread of shear localization over the total cross-section of the deformed specimen, and eventually becomes a macroscopic shear band. It is expected that the formation of the shear localization is a very complex deformation phenomenon involving a series of crystallographic and non-crystallographic events such as orientation change, crystal rotation, misorientation generation and even perhaps possible initiation, growth and coalescence of the microcracks. Figure 7 shows a typical TEM micrograph displaying the rotation of the precipitates in the band towards the shear direction.

Observations by SEM show that there are several ways for a microcrack nucleation to be found in the present study, including the decohesion of the interfaces at the inclusions, precipitates and grain boundaries as shown in Fig. 8 where the microcracks are initiated at the inclusion–matrix and ferrite–pearlite interfaces. It is seen that the nucleation and growth of the microcracks could promote the developing process of the shear localization [18]. As Cowie *et al.* pointed out, the strain localization in UHS steels under pure shear loading conditions is driven by microvoid softening. They found that the enhanced resistance to shear instability was caused by full dissolution of the cementite particles [25]. In the present study there has not been any dissolution of precipitates by shearing to be observed.

It is generally accepted that the fracture in morphology in the quenched steel appears to be cleavage or quasi-cleavage [26]. However, it is surprising to note in the present study that there are no cleavage faces or other brittle features to be observed on the fracture surface in all three steels, particularly in the case of the quenched steel. As a result of an intensive localized shear deformation, the fracture occurring in all these steels exhibits a ductile transgranular as characterized by the elongated shear dimples on the fracture surface shown in Fig. 9. This implies that heat softening plays an important role in the deformation of the shear localization, leading to the final shear fracture. In other words, the material in the band subjected to a large accumulated strain under high speed rate, resulting in temperature rise seems to suffer an annealed or tempered treatment. Therefore

the fracture in morphology in the quenched steel which was deformed at high speed rate and large strain appears to be similar to those in the annealed or tempered steels in tensile and fatigue tests.

Dislocation structure

Wittman *et al.* [27] investigated the transformation band produced during torsional expansion in AISI 4340 steel by High Voltage Transmission Microscopy and found that, except for a major band, there are small incipient bands associated with the major band about 0.5 μm wide. These small bands travel 30–50° to the band, eventually joining it.

One of the important deformation features observed in the present study is shown in Fig. 10, indicating that the interfaces between the ferrite and cementite as emission sources could release a great deal of dislocation. Diffraction contrast analysis shows that the majority of dislocations are single dislocations with Burgers vector $1/2[111]$ lying on a plane $(1\bar{1}0)$ and $1/2[1\bar{1}\bar{1}]$ lying on a plane (110) . These dislocations are interact with each other to form a stable hexagonal dislocation network (Fig. 11) expressed by

$$1/2[111] + 1/2[1\bar{1}\bar{1}] \rightarrow [100].$$

Another prominent feature found during high speed torsional deformation is the formation of the tangled structure and dislocation cells. Their elongation and arrangement tend to be aligned along the length of the shear direction as shown in Fig. 12. It might be expected that the formation of the tangles, cells and networks of the dislocations could be responsible for the work-hardening of the material during dynamic loading but the avalanche of the dislocation cells (Fig. 13) may be associated with the sharp drop in the shear stress. Because of large accumulated plastic strain leading to the operation of multiple slip systems, the deformation in some regions in the band appears to be a non-crystallographic feature in a form which is characterized by a well developed spotty ring pattern as shown in Fig. 14 (marked by B and C). On the other hand in region A, the shear band still keeps its crystallographic feature. This implies that deformation distribution in the band is inhomogeneous.

CONCLUSIONS

- (1) The localized shear bands were observed in the quenched, quenched and tempered as well as normalized steels with widths of 75–100, 130–150 and 350–400 μm respectively. The shear localization is susceptible to the strength of the steels. The higher the strength of the steels, the easier it is for the shear band to occur.
- (2) Double shear bands are found in the present study. The principal band develops quickly and its width is narrower than that of the secondary band, but they are parallel to each other.
- (3) The reversed loading results in the formation of kink bands.

- (4) There is a critical strain at which the shear localization does occur. Before arriving at the critical strain, the deformation goes homogeneously over the whole gage length of the specimen and the work-hardening results from the creation, multiplication and interaction of the dislocations. After this point, the deformation becomes gradually localized during which the material begins to be work-softening until fracture occurs.
- (5) The sharp drop in the load-carrying capability for the deformed specimen seems to be associated with the growth and coalescence of the microcracks rather than the occurrence of shear localization.
- (6) The formation of shear localization involves a series of sequential crystallographic and non-crystallographic deformation events within the band, including the change of the crystal orientation, the misorientation generation and even perhaps damage in microstructures such as initiation, growth and coalescence of the microcracks.
- (7) The fracture of the three steels studied in the present study appears to be ductile transgranular in nature.

Acknowledgement—This work was supported by National Nature Science Foundation of China No.19392300.

REFERENCES

1. C. Zener and J. H. Hollomon, *J. appl. Phys.* **15**, 722 (1944).
2. R. F. Recht, *J. appl. Mech. Trans. ASME (E)* **31**, 189 (1964).
3. R. S. Culver, in *Metallurgical Effect at High Strain Rates* (edited by R. W. Rohde, B. M. Butche, J. R. Holland and C. H. Karnes), p. 519. Plenum Press, New York (1973).
4. Y. Bai, *J. Mech. Phys. Solids* **30**, 195 (1982).
5. M. R. Staker, *Acta metall.* **29**, 683 (1981).
6. F. H. Wu and L. B. Freund, *J. Mech. Phys. Solids* **32**(2), 119 (1987).
7. C. Fressengeas and A. J. Molinari, *J. Mech. Phys. Solids* **35**(2), 185 (1987).
8. R. J. Clifton., Report No. NMAB-356, National Materials Advisory Board Committee, Chapter 8, p. 129 (1980).
9. H. C. Rogers, *Ann. Rev. Mater. Sci.* **9**, 283 (1979).
10. R. L. Woodward, *Metall. Trans.* **10A**, 283 (1979).
11. H. A. Grebe, H. R. Park and M. A. Meyers, *Metall. Trans.* **16A**, 761 (1985).
12. Y. Me-Bar and D. Shechtman, *Mater. Sci. Engng.* **58**, 181 (1983).
13. B. Dodd and A. G. Atkins, *Acta metall.* **31**, 9 (1983).
14. M. A. Meyers and C. L. Wittman, *Metall. Trans.* **21A**, 3153 (1990).
15. D. A. Scokey, in *The Inter. Symposium EXPLO-MENT 85*, 28 July–1 August 1985 in Portland, Oregon and Published in the *Proc. Conf. on Metallurgical Applications of Shock-Wave and High Strain-Rate Phenomena* (edited by L. E. Murr, K. T. Staudhammer and M. A. Meyers). Marcel Dekker, New York (1986).
16. L. S. Costin and J. Duffy, *J. Engng. Mater. Technol.* **101**, 258 (1979).
17. Y. B. Xu and Y. L. Bai *et al. Mater. Sci. Engng.* **114**, 81 (1989).
18. Y. L. Bai, Q. Xue, Y. B. Xu and L. T. Shen, *Mech. Mater.* **17**, 155 (1984).
19. T. Shawki, R. J. Clifton and G. Majda, Brown University Report No. ARD DAAG29 81-K-0121/3.
20. Y. B. Xu and Y. L. Bai *et al. Scripta metall.* **24**, 571 (1990).
21. K. Cho, Y. C. Chi and J. Duffy, *Metall. Trans.* **21A**, 1161 (1990).
22. R. J. Clifton, J. Duffy, K. A. Hartley and T. J. Shawki, *Scripta metall.* **18**, 443 (1984).
23. K. A. Hartley, J. Duffy and R. H. Hawley, *J. Mech. Phys. Solids* **35**, 283 (1987).
24. T. Shawki, PhD Thesis, Brown University (1985).
25. J. G. Cowie, M. Azrin and G. B. Olson, *Metall. Trans.* **20A**, 143 (1989).
26. Z. F. Zhang and Y. B. Xu, *Acta metall. sinica* **21**, 384 (1986).
27. C. L. Wittman, M. A. Meyers and H. R. Pak, *Metall. Trans.* **21A**, 707 (1990).