Determination of the Sources of Acoustic Emission Generated during the Deformation of Titanium

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An investigation of the acoustic emission (AE) generated during the deformation of pure polycrystalline alpha-titanium rod has been carried out. The goals of the investigation were to characterize the AE produced during deformation and identify the specific deformation processes responsible for the measured AE. The AE was found to depend on sample orientation, mode of testing, strain rate, test temperature, and purity. In all cases twinning and/or the initiation of dislocation slip processes were found to be the principal sources of acoustic emission.

I. INTRODUCTION

It is well known that crystalline materials deform by specific deformation mechanisms and processes, and that detectible AE may be generated if the deformation processes are accompanied by a sudden stress relaxation and/or strain accommodation. Alpha-phase (HCP) titanium has been observed to deform by twinning, dislocation slip, and grain boundary sliding, although the latter process has been shown not to be a significant source of AE.[1] By simultaneously measuring a wide range of AE parameters along with other experimental variables sensitive to specific deformation mechanisms, it is hoped that the measured AE can be ascribed to specific deformation processes with some degree of confidence. The purpose of this paper is to present data and analyses which can be used to identify the sources of AE generated during deformation of alpha-phase titanium.

II. EQUIPMENT AND PROCEDURE

The data gathering and analyzing system used throughout this investigation is based on a Dunegan-Endevco (D/E) S140B high sensitivity transducer, D/E 3000 series signal processing equipment, a Hewlett-Packard 3400A voltmeter, a Motorola M6800 microprocessor, and a Dillon Universal testing machine. A detailed description of the experimental apparatus is given elsewhere.[2] With this experimental set-up it was possible to measure and record simultaneously in digital form the applied load, the sample displacement, the RMS voltage (analog only), the AE count rate, the AE event rate, the AE energy rate, and the AE signal amplitude distributions. All deformation tests were carried out at constant crosshead speeds with the total system gain at 100 db or 112 db.

Two purity grades of alpha-titanium were tested: both VP grade (impurities <665 ppm) and Marz grade (impurities <275 ppm) were purchased from MRC Corp., Orangeburg, NY.The materials were supplied in the form of polycrystalline rods. One Marz rod (#1) was 0.95 cm in diameter, while the VP and a second Marz rod (#2) were 1.27 cm in diameter. X-ray analyses indicated that the crystal c axes tended to be oriented radially but pointing at some small angle above the cross-sectional plane. The preferred orientation was less pronounced in the Marz material. Metallographic analyses showed that the grains were uniformly equiaxed and on the order of 10 microns in diameter in all of the materials tested. Microholes, similar to those described by Zhu Zu-Ming *et al.*,[3] were found in both grades of material in the as-received condition.

Due to the rod geometry, tensile specimens could be machined only with the gauge length parallel to the rod axis. The tensile samples had a 2.54 cm gauge length, 0.26 cm^2 circular cross-section, and a flat, 0.02 cm deep, on the gauge length which served as a seat for the transducer. Compression samples were machined into 1.15 cm $\times$ 0.26 cm^2 parallelepipeds oriented in the axial direction. All observable traces of twins and microholes were removed prior to testing by vacuum annealing the test specimens at 810 K for four hours after which they were allowed to furnace cool.

III. EXPERIMENTAL RESULTS

The measured AE was found to be dependent on the purity of the test sample and on the testing mode used. For these reasons and to avoid confusion the following labeling system will be used:

C-V = VP grade material, compression parallel to the rod axis.
T-V = VP grade material, tension parallel to the rod axis.
C-M = Marz grade material, compression parallel to the rod axis.
T-M = Marz grade material, tension parallel to the rod axis.

C-V tests were characterized by stress/strain curves, with one or more distinct hardening regions, similar to those reported by Wasilewski.[4] Figure 1 (see arrow) provides a clear example of the hardening behavior. The data shown in Figure 1 are for a C-V test run at a total strain rate of 0.5 per hour. Notice that two distinct hardening regions can be observed. The AE data for this particular test along with the stress/strain curve are shown in Figure 2. After some very early activity, there is very little AE in the elastic region (except for occasional bursts). However, as yield is approached the AE rises to a maximum near yield and then suddenly drops at the onset of hardening (see Figure 1). The AE then begins to increase again until the onset of the

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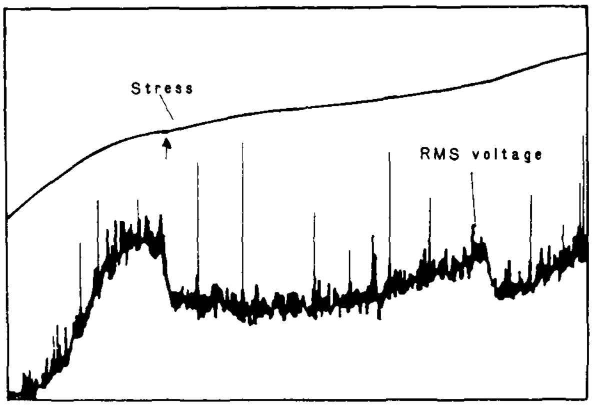

Fig. 1—Applied stress and the AE measured in terms of the RMS voltage from yield to the start of the second hardening region plotted against total strain. VP grade titanium; type C-V test.

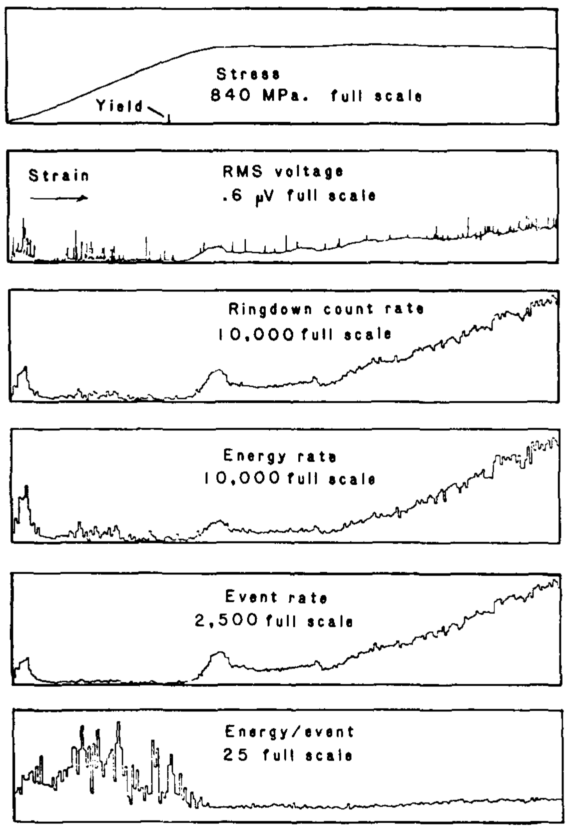

Fig. 2—Typical data for a type C-M test. AE parameters and stress as a function of total strain. All AE rates are in counts per 5 sec.

second hardening region where it suddenly decreases again. The AE then increases until the conclusion of the test. The AE appears to have a continuous background with significant burst type AE occurring during the test. The burst type of emission appears to increase as the test progresses to higher strains.

A part of the investigation was to determine the effects of strain rate on the measured AE. For the C-V tests the AE was determined by the value of the RMS voltage at the AE maximum near yield, the RMS voltage at the AE maximum immediately preceding the second hardening region and the RMS voltage at a predetermined value of plastic strain past the second hardening region. The data are shown

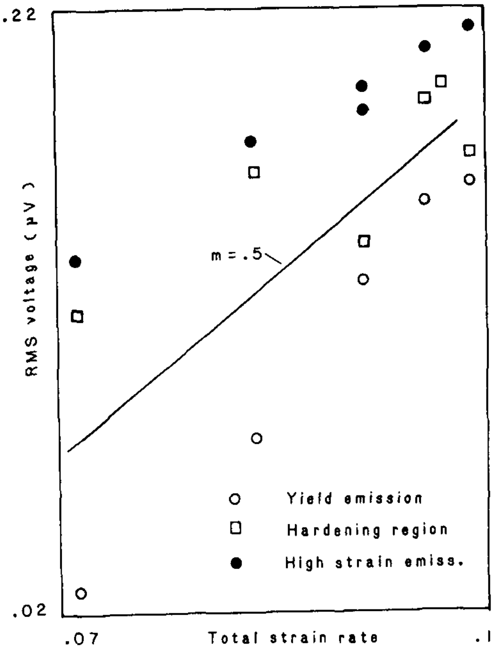

Fig. 3—Natural logarithm of the RMS voltage *vs* the natural logarithm of the total strain rate for C-M type tests.

in Figure 3. The solid line is what would be expected if the RMS voltage were proportional to the strain rate raised to the one-half power. This relationship is consistent with a mechanism which does not change as a result of a change in strain rate.[5] Clearly, the AE increases with increasing strain rate, although the amount of increase is less at higher values of strain. A metallographic examination of a sample deformed exactly like the sample which produced the data shown in Figure 2 revealed very few twins (less than 10) on the sample surfaces; however, after testing well past yield areas with high twin densities were observed. It was impossible to determine actual twin densities due to the large amount of plastic deformation.

C-M tests and data were equivalent to the C-V tests and data with two exceptions. First, there was an overall decrease in the magnitude of the AE and second, the 0.2 pct yield stress measured in a C-M test was found to be approximately 82 pct of the yield stress measured in a C-V test.

The measured AE and stress/strain curve for a T-M test are somewhat different as shown in Figure 4. The T-M tests were run with a system gain of 112 db while all of the other tests were carried out with a system gain of 100 db. A sharp yield point with a pronounced flow region followed by hardening was typical for this type of test. The AE is characterized by a significant peak at yield with a slight increase at the onset of hardening. After hardening the AE increases very slightly during the remainder of the test. The peak at yield appears to be composed of continuous emission; however, past yield the AE contains an increasing amount of burst type activity. Test temperature was found to have a significant effect on both the amount and location of burst type AE for the T-M samples. A reduction of test temperature was found to enhance the amount of burst type AE and move it to lower strains. Testing at 198 K produced a significant amount of burst type AE at yield. Testing at various strain rates at room temperature indicated that the AE, measured in terms of the maximum RMS voltage at

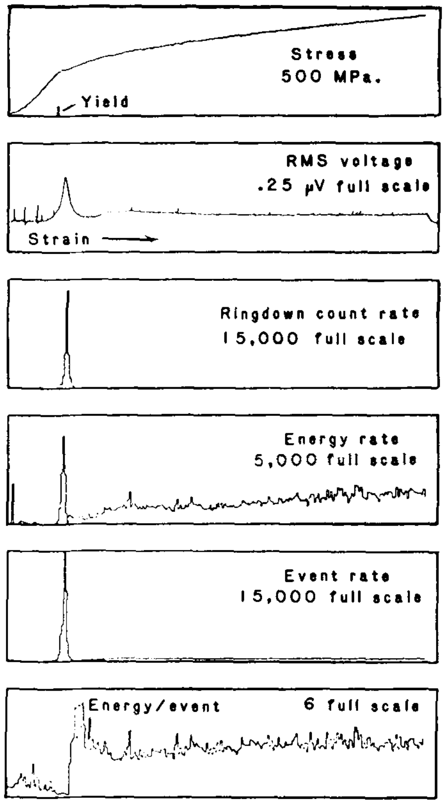

Fig. 4—Typical data for a T-M type test. AE parameters and stress as a function of total strain. All AE rates are in counts per 5 sec.

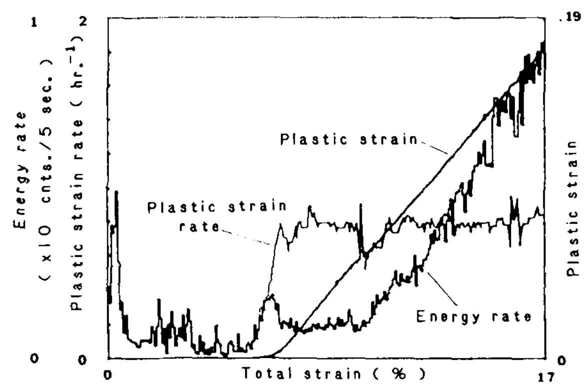

Fig. 5—AE energy rate, plastic strain, and plastic strain rate as a function of total strain for a C-V type test.

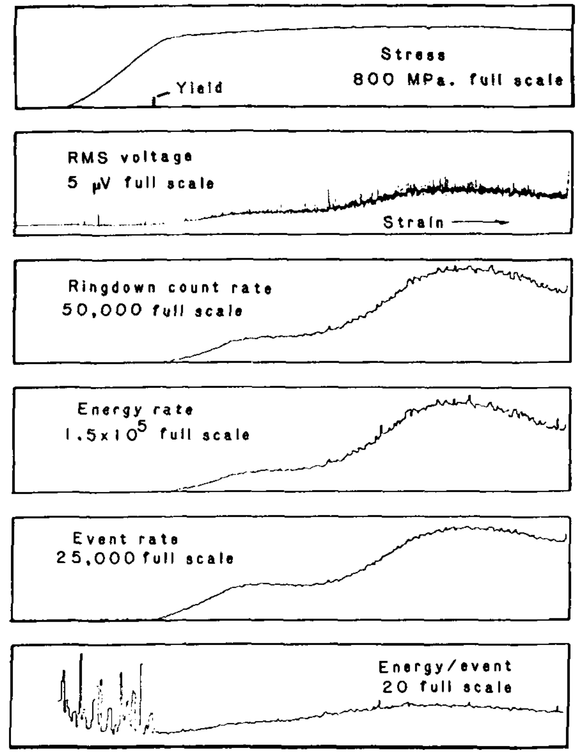

Fig. 6—Typical data for a T-V type test. AE parameters and stress as a function of total strain. All AE rates are in counts per 5 sec.

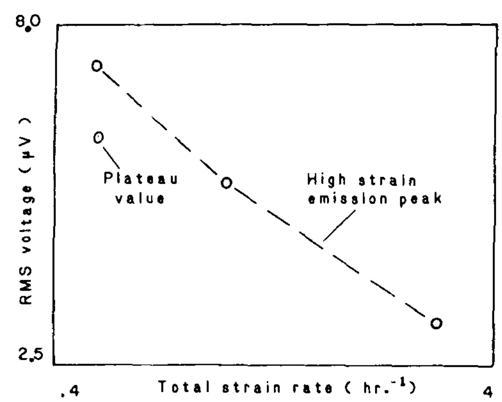

Fig. 7—Natural logarithm of the RMS voltage at the high strain maximum of AE *vs* the natural logarithm of the total strain for T-V type tests.

yield, increased with increasing strain rate. The data were found to agree quite well with the relationship: RMS voltage proportional to strain rate to the one-half power. All of the T-M tests and C-V tests were characterized by an AE maximum near yield. In all cases there was found to be a close relationship between the AE and the plastic strain rate up to the maximum of the AE. Such a relationship is shown in Figure 5, which shows data for a typical C-V test.

The AE measured during the deformation of a T-V sample is significantly different from that observed in any of the other tests. Typical data for a T-V test are shown in Figure 6. The AE is significantly higher, appears to begin at yield, increases to a plateau past yield, and then increases further to a maximum at high strains. It is not clear if absence of an AE peak at yield is because none exists or if it is swamped out by the higher levels of AE (notice the difference in scales for Figures 2, 4, and 6). The measured AE contains a significant amount of burst type emission. The AE as measured by the RMS voltage had a considerable amount of variation in magnitude for similarly prepared and tested T-V samples. No obvious relationship between any portion of the AE and the plastic strain rate could be found. Testing the T-V samples at a variety of strain rates at room temperature produced a surprising result. An increase in the strain rate was observed to produce a decrease in the measured AE; data are given in Figure 7. A decrease in the test temperature for the T-V type of sample produced the same result as increasing the strain rate, *i.e.*, a decrease in the measured AE. Data showing the AE as a function of test temperature are shown in Figure 8.

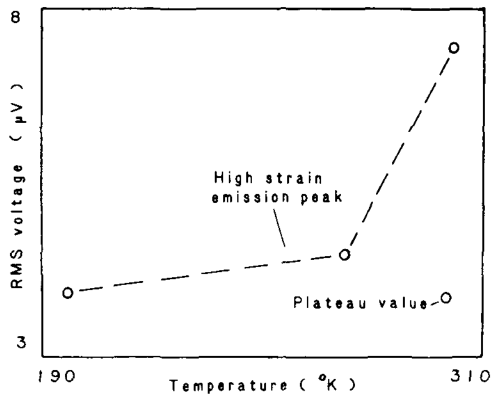

Fig. 8—RMS voltage at the high strain maximum of AE *vs* test temperature for T-V type tests.

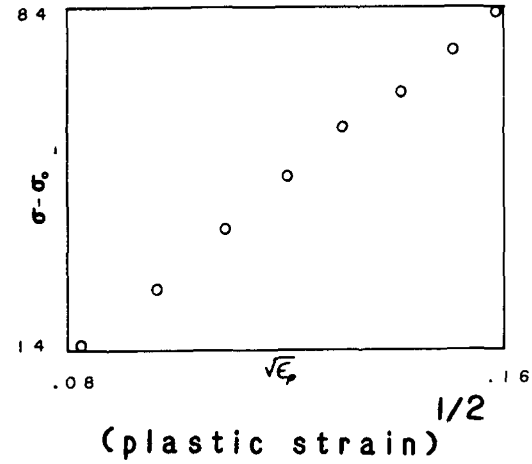

Fig. 9—Stress *vs* the square root of plastic strain in the hardening region of a T-M type test. Stress is in MPa.

IV. DISCUSSION

Analyses of the AE data for the C-V and C-M type of tests indicate that there is a correlation between changes in the AE and the onset of hardening seen in the stress-strain curves. A relationship between the sources of AE in these regions and the deformation mechanisms responsible for the stress hardening behavior could be expected. Various researchers have noted prism slip (propagated as Lüder's bands) to be the primary mechanism at yield in single and polycrystalline alpha-titanium. Past yield the samples deformed by basal or pyramidal slip, although some single crystals have been observed to deform entirely by twinning under certain conditions.[6,7] Evidence that the primary deformation mechanism at yield for the C-V and C-M tests is prism slip can be obtained from a comparison of the yield stresses of the two grades of material tested under the same conditions. The critical resolved shear stress τ_c for prism slip was found to be dependent on the concentration of alpha phase stabilizing solutes, C_i, according to[8]

$$\tau_c = K C_i \qquad [1]$$

where K is a constant. By assuming the observed 2 pct yield stress proportional to τ_c, we find

$$\sigma_c(\mathrm{VP})/\sigma_c(\mathrm{Marz}) = [C_i(\mathrm{VP})/C_i(\mathrm{Marz})]^{1/2} \qquad [2]$$

The principal impurities affecting the yield stress are oxygen, carbon, nitrogen, and iron.[9] The concentrations of these impurities for the two grades of materials are given in Table I.

Expressing the values given in Table I in oxygen equivalents and then evaluating the ratio of the concentration of solutes gives

$$[C_i(\mathrm{VP})/C_i(\mathrm{Marz})]^{1/2} = 1.26 \qquad [3]$$

Taking the ratio of the 0.2 pct yield stress for the C-V and C-M tests gives

$$\sigma_c(\mathrm{VP})/\sigma_c(\mathrm{Marz}) = 1.22 \qquad [4]$$

Table I. Concentrations in ppm

Material	O	N	C	Fe
VP	200	15	70	15
Marz	140	4	35	22

The excellent agreement between the two values provides strong support for the primary deformation mechanism at yield being prism slip. Careful analysis of the hardening region past yield indicates that the following relation exists between the applied stress and the plastic strain:

$$\sigma - \sigma_c = \varepsilon_p^{1/2} \qquad [5]$$

where σ_c is some value of threshold stress (see, for example, Figure 9). Numerous investigators have both proposed and observed this relationship for a number of different deformation processes, all of which involve dislocation slip.[10-14] Since it has already been shown that the measured AE is closely related to the behavior of the stress/strain curve, one might conclude that the AE is due to the same dislocation mechanisms responsible for the stress behavior. Further evidence for this point of view comes from the fact that the early AE correlates well with changes in the plastic strain rate. However, the presence of burst-type emission must be explained as well as the continually increasing emission at large values of strain. These observations coupled with the fact that twinning was observed metallographically after deformation indicate that both dislocation slip and deformation twinning are taking place. The most logical explanation of the C-V and C-M data is that the AE at yield is primarily due to the initiation of dislocation slip with the burst emission[15] being due to a small amount of deformation twinning. As the test progresses to higher strains, the AE from deformation twinning increases and the AE from dislocation slip decreases except where hardening is observed. At still higher strains the AE is primarily due to deformation twinning. Additional evidence for this interpretation of the AE data can be gained from analysis of the AE as a function of strain rate. The AE measured at yield fits quite well the one-half power dependence which has been reported[5] many times for AE from the initiation of dislocation slip. On the other hand, the AE measured at high strains does not fit the one-half power dependence, indicating a change in mechanism, which, from metallography, is clearly deformation twinning.

The AE measured in a T-M test is characterized by a pronounced peak in the AE at yield. Three factors: (1) the

AE maximum occurs at yield, (2) the AE is continuous in nature, and (3) the AE measured in terms of the RMS voltage fits very well the one-half power dependence, all strongly suggest that the AE at yield is primarily due to the initiation of dislocation slip. A careful analysis of Figure 4 shows that at the RMS voltage maximum there is a very high number of events but that the energy per event is very low. This high number of low energy events is consistent with dislocation slip. Immediately after yield, the energy per event increases and then remains fairly constant for the remainder of the test. The AE measured past yield is characterized as being burst type with a high energy per event. In all likelihood, this AE is due to deformation twinning. Because of the large amount of deformation, it was difficult to make accurate twin counts using metallographic techniques. However, other evidence indicated that this AE was indeed due to twinning. The AE measured at different test temperatures provides evidence that the AE at high strains is from deformation twinning. At 198 K, the yield peak was saturated with burst-type emission, while at 340 K, there was no observable burst-type emission. These results are consistent with $\{10\bar{1}2\}$ twinning. This twinning system is not activated much above room temperature, and the critical resolved shear stress to activate this twinning system decreases with decreasing test temperature. In fact, the critical resolved shear stress required to activate $\{10\bar{1}2\}$ twinning and the critical resolved shear stress for prism slip are nearly equal at 198 K. It is then reasonable that at 198 K the AE at yield should be saturated with burst-type AE. A consistent explanation of the measured AE from a T-M test is that the AE at yield is due to the initiation of dislocation slip while that observed past yield is due to $\{10\bar{1}2\}$ deformation twinning.

The T-V test AE data, with its unique strain rate dependence, is most interesting. As mentioned earlier, it was impossible to determine if there was a small AE peak at yield due to the high level of AE. The burst-type character of the emission, coupled with its location in strain, the fact that there is no relationship of the AE to the plastic strain rate, and the AE's inverse strain rate dependence, strongly suggest that dislocation slip is not a source of measured AE. Conversely, all of these factors would suggest deformation twinning as the source of the measured AE. Metallographic analyses indicated high twin densities as the samples were strained past yield. Again, the results of T-V tests at different test temperatures suggest twinning as the mechanism responsible for the measured AE. Suppose the AE maximum observed at higher strains is due to the nucleation of deformation twins: (1) since the critical resolved shear stress required to activate the possible twinning systems decreases with a decrease in test temperature, one would expect to reach the maximum at a lower value of strain as the test temperature is lowered (the critical resolved shear stress is reached sooner); (2) since the stress necessary to activate the twinning system decreases, it would be expected that the energy per event would decrease. Both of these results are observed for the T-V tests as can be seen in Figure 10. The measured AE in a T-V type test is thus believed to be due to deformation twinning and the unique strain rate dependence is most likely due to the temperature dependence of the critical resolved shear stress to activate the deformation twinning modes.

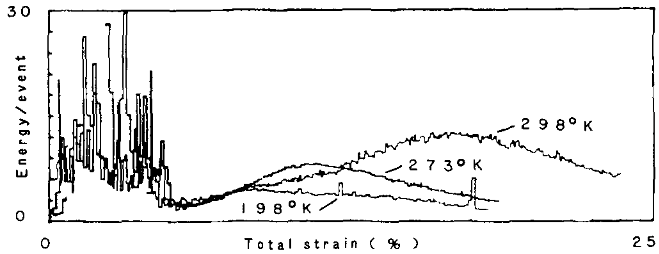

Fig. 10 — AE energy per event *vs* total strain for T-V tests at three different temperatures.

V. CONCLUSIONS

The AE generated during the deformation of alpha-titanium has been measured and characterized and possible sources of the emission suggested. Two purity grades of alpha-titanium were investigated, and the measured AE was found to be dependent on the purity, sample orientation, and mode of testing employed. In all cases the measured AE was found to be caused by either the initiation of dislocation slip or deformation twinning. In those tests where AE was observed at yield or at hardening regions of the stress strain curve (C-V, C-M, and T-M tests) it is postulated that this AE is primarily due to the initiation of dislocation slip. The AE observed at strains well past yield is believed due to deformation twinning. The unique strain rate dependence of the measured AE for the T-V tests is thought to be a result of the temperature dependence of the critical resolved shear stress necessary to activate twinning systems in alpha-titanium.

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