Columbia University
in the City of New York

DEPARTMENT OF CIVIL ENGINEERING
AND ENGINEERING MECHANICS

INSTITUTE FOR THE STUDY OF FATIGUE AND RELIABILITY

INSTABILITY OF TITANIUM AND Ti/6Aℓ/4V ALLOY AT ROOM TEMPERATURE

by

W. A. Wood

Sponsoring Agencies
Office of Naval Research
Air Force Materials Laboratory
Advanced Research Projects Agency

Contract No. NONR 266(91)
Project No. NR 064-470

Technical Report No. 45

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Publications of the Institute for the Study of Fatigue and Reliability

No.

1) Shinozuka, M. On Upper and Lower Bounds of the Probability of Failure of Simple Structures under Random Excitation. December 1963.

2) Freudenthal, A.M., Weibull, W. and Payne, A.O. First Seminar on Fatigue and Fatigue Design. December 1963.

3) Wood, W.A., Reimann, W.H. and Sargant, K.R. Comparison of Fatigue Mechanisms in Bcc Iron and Fcc Metals. April 1964.

4) Heller, R.A. and Shinozuka, M. Development of Randomized Load Sequences with Transition Probabilities Based on a Markov Process. June 1964.

5) Branger, J. Second Seminar on Fatigue and Fatigue Design. June 1964.

6) Wood, W.A. and Reimann, W.H. Room Temperature Creep in Iron under Tensile Stress and a Superposed Alternating Torsion. June 1964.

7) Ronay, M., Reimann, W.H. and Wood, W.A. Mechanism of Fatigue Deformation at Elevated Temperature. June 1964.

8) Freudenthal, A.M. and Shinozuka, M. Upper and Lower Bounds of Probability of Structural Failure under Earthquake Acceleration. June 1964.

9) Ronay, M. On Strain Incompatibility and Grain Boundary Damage in Fatigue. August 1964.

10) Shinozuka, M. Random Vibration of a Beam Column. October 1964.

11) Wood, W.A. and Reimann, W.H. Some direct Observations of Cumulative Fatigue Damage in Metals. October 1964.

12) Freudenthal, A.M., Garrelts, J.M. and Shinozuka, M. The Analysis of Structural Safety. October 1964.

13) Ronay, M. and Freudenthal, A.M. Second Order Effects in Dissipative Solids. January 1965.

14) Shinozuka, M. and Nishimura, A. On General Representation of a Density Function. February 1965.

15) Wood, W.A. and Nine, H.D. Differences in Fatigue Behavior of Single Copper Crystals and Polycrystalline Copper at Elevated Temperatures. February 1965.

No.

16) Ronay, M. On Second Order Strain Accumulation in Torsion Fatigue. February 1965.

17) Heller, R.A. and Heller, A.S. A Probabilistic Approach to Cumulative Fatigue Damage in Redundant Structures. March 1965.

18) Wood, W.A. and Reimann, W.H. Extension of Copper and Brass under Tension and Cyclic Torsion. April 1965.

19) Grosskreutz, J.C., Reimann, W.H. and Wood, W.A. Correlation of Optical and Electron-Optical Observations in Torsion Fatigue of Brass. April 1965.

20) Freudenthal, A.M. and Shinozuka, M. On Fatigue Failure of a Multiple-Load-Path Redundant Structure. June 1965.

21) Shinozuka, M. and Yao, J.T.P. On the Two-Sided Time-Dependent Barrier Problem. June 1965.

22) Ronay, M. On Second Order Strain Accumulation in Aluminum in Reversed Cyclic Torsion at Elevated Temperatures. June 1965.

23) Freudenthal, A.M. Second Order Effects on Plasticity. August 1965.

24) Wood, W.A. Experimental Approach to Basic Study of Fatigue. August 1965.

25) Ronay, M. Conditions of Interaction of Cyclic Torsion with Axial Loads. August 1965.

26) Heller, R.A., and Donat, R.C. Experiments on the Fatigue Failure of a Redundant Structure. October 1965.

27) Shinozuka, M., Yao, J.T.P. and Nishimura, A. A Note on the Reliability of Redundant Structures. November 1965.

28) Mason, W.P. Internal Friction Measurements and Their Uses in Determining the Interaction of Acoustic Waves With Phonons, Electrons and Dislocations. January 1966.

29) Shinozuka, M., Hakuno, M. and Itagaki, H. Response of a Multi-Story Frame Structure to Random Excitation. February 1966.

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30) Wood, W. A., Suppression of Creep Induced by Cyclic Torsion in Copper under Tension. April 1966.

31) Shinozuka, M. and Sato, Y., On the Numerical Simulation of Nonstationary Random Processes. April 1966.

32) Mason, W. P., Effect of Electron-Damped Dislocations on the Determination of the Superconducting Energy Gaps of Metals. May 1966.

33) Murro, R. P., Creep Behavior of an Aluminum Alloy under Transient Temperatures. June 1966.

34) Ronay, M., On the Micro-Mechanism of Second-Order Extension of Aluminum in Reversed Cyclic Torsion. June 1966.

35) Wood, W. A., Yield and Second-Order Effects Induced by Cyclic Torsion in Copper under Tension. June 1966.

36) Nine, H. D., and Wood, W. A., On Improvement of Fatigue Life by Dispersal of Cyclic Strain. July 1966.

37) Jacoby, G., Review of Fractographic Analysis Methods for Fatigue Fracture Surfaces. July 1966.

38) Donat, R. C. and Heller, R. A., Experiment on a Fail Safe Structural Model. September 1966.

39) Freudenthal, A. M., and Gou, P. F., Accumulation of Second Order Strain in Cyclic Loading of Viscous Bar. September 1966.

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41) Mason, W. P., and Bateman, T. B., Relation between Third-order Elastic Moduli and the Thermal Attenuation of Ultrasonic Waves in Nonconducting and Metallic Crystals. October 1966.

42) Wood, W. A., Re-examination of Mechanical Hysteresis. February, 1967.

43) Mason, W. P., and Rosenberg, A. Phonon and Electron Drag Coefficients in Single-crystal Aluminum. November 1966.

44) Mason, W. P., Mechanical Instability of Alloy Ti-6Al-4V Under Large Extensional Vibrations. February 1967.

ABSTRACT

Titanium and Ti/6Aℓ/4V under torsion exhibit unusual instability which shows itself as (a) abnormally prolonged creep on yielding, (b) ineffective strain-hardening after yielding, and (c) deviations from elastic strain at loads above a limit of proportionality which is low compared with the bulk yield stress. As a result of this inherent instability it is possible to produce significant axial extensions under low tensile stresses combined with small-amplitude cyclic torsion; and to reduce the nominal bulk yield stress in torsion to at least half its nominal value by suitable prior straining.

1. Introduction

Alpha-titanium loaded in torsion to or beyond its bulk yield-point at room temperature creeps abnormally. Typical creep curves are shown in Fig. 1 for commercially pure titanium in the form of tubular specimens obtained by turning, boring, and reamering 5/16 in. diameter rod to smoothly finished test portions 1 1/2 in. long, 1/4 in. o.d., and 3/16 in. i.d. The twist in these curves is plotted in degrees, since degrees gives a ready picture of its extent, but it can be expressed as shear strain by the relation 1 degree twist equals 0.00145 surface shear.

This creep is plotted only to 300 hours but actually was measured up to 700 hours, or until it caused undue buckling of a specimen, and though it had slowed down it was still in progress at the end of each test. This creep too refers to titanium which had been annealed in vacuo for 1/2 hour at 1300°F, but it appeared equally in unheated specimens and in specimens which were solid. Thus it seems that alpha-titanium in any form when strained above the yield becomes unstable, basically because it is incapable of strain-hardening effectively.

It was of interest therefore to test how stable the alpha-titanium might be when strained below the yield and also to test alloys containing alpha-titanium as a constituent phase. The alloy reported on here is Ti/6Aℓ/4V (AMS 4928), reduced like the titanium from 5/16 in. rod to similar tubular specimens and similarly tested in dead loading.

2. Pseudo-Elastic Range (Titanium)

An outstanding feature of titanium loaded below its yield was a low limit of proportionality relative to its bulk yield. This feature is illustrated by the torque/twist curve in Fig. 2 which shows an L.P. ~ 18 lb.-in. and a yield ~ 50 lb.-in. Because the metal thus exhibited a long drawn-out range between L.P. and yield it also exhibited easily observed permanent sets on unloading from points within this range, for example from points A and B in Fig. 2. Moreover a specimen on re-loading to A or B showed a small increment of strain as a result of the cycle, an irreversibility, clear at point B for example, which confirmed that the true linear elastic range of the metal was not as extensive as its high yield point might suggest.

3. Pseudo-Elastic Range (Ti/6Aℓ/4V)

The alloy showed even more marked evidence of the above pseudo-elasticity because its yield, but not its L.P., was much higher than that of the pure metal; so that the range between L.P. and yield was drawn out even more. This range is illustrated by Fig. 3, the L.P. now being only about 1/10th of the yield point. Consequently, as

Fig. 3 also illustrates, unloading from points such as A, B, C, D, in this range led to obvious permanent sets.

Fig. 3 refers to a specimen which was not heat treated after machining and which therefore might be taken as characteristic of a cold-worked state. However, specimens which were heat treated showed essentially similar effects, though their yield point and L.P. might be lowered a little by the heating. For example, as shown by Fig. 4, similar effects occurred in specimens heated in vacuo for 1 hour at 1300°F. Other heat treatments were 1 hour at 1200°F, 3 hours at 1200°F, 3 hours at 1300°F; so the effects were characteristic of the material and not of any particular heat treatment. Fig. 4 also serves to illustrate further that the process of unloading and re-loading from points such as A, B, C itself produces an increment of strain; thus the point B from which an unloading began does not lie on the re-loading curve to C, a further confirmation of the pseudo-elasticity.

4. Cycles of Torsion Combined With Axial Tension

If a specimen were truly elastic in torsion up to a yield torque T_o and in tension to yield σ_o it should be elastic also under combinations of T and σ corresponding to points within a Mieses type ellipse with axes T_o and σ_o, as depicted in Fig. 5a; and it should remain elastic when T is reversed to -T. If however it is beyond its elastic range under the combination T, σ it should exhibit a permanent axial extension when T is reversed. Such extension provides a sensitive indication of deviations from elasticity because it is additive in successive reversals.

This test was applied to the alloy at T, σ combinations well within the above type of ellipse. Nominal tensile yield according to the suppliers was 130,000 to 140,000 psi, so the σ_o axis of the ellipse in Fig. 5a has been made equivalent to 135,000 psi. Yield in torsion according to the present tests occurred at $\theta \sim 12$ degrees and the torsion axis of the ellipse for convenience has been left in degrees, with $\theta_o = 12$ degrees. Test points corresponded to points like P where θ was only 3 degrees and σ only 20,000 psi.

The test was applied in a machine designed for the method. Standard procedure was first to apply the axial tensile load and allow any resulting extension to come to rest; then to apply the cycles of torsion, normally at 1500 cpm. Extra extension induced by the cycles was plotted against cycles.

Fig. 5b, which shows the curve corresponding to point P, demonstrates a significantly large induced extension and so proves a marked deviation from elasticity at stresses well within the nominal yields of the alloy. Fig. 5b includes curves showing extensions at even less σ, θ values than those at P.

Thus the alloy is even more unstable when cyclic torsion is superposed on static tension than under static loading alone and unstable moreover under combinations well within its nominal torsion and tensile yields.

5. Instability Induced By Cycles Of Large Amplitude

The initial yield torque of the alloy could be at least halved by a prior straining conveniently described by reference to Fig. 6. A specimen was first loaded in torsion to its yield T_o, here 170 lb.-in.; then allowed to creep under T_o through $\theta = 53$ degrees, the time taken in creep being 65 hours. It was then subjected to the reverse torque $- T_o$. As Fig. 6 indicates the specimen as a result of this reversal tended to creep at loads less than $- T_o$ and, when $- T_o$ itself was reached, crept more rapidly than previously at $+ T_o$; it now crept through 53 degrees in only 23 hours. Next the load was reversed to $+ T_o$ in the original direction. Creep during this reversal was further accentuated and at $+ T_o$ now produced a twist of 53 degrees in only 2 hours; and so on in a second cycle.

If after a second cycle the specimen was held under torque 90 lb.-in., a load only about half of its original nominal yield T_o, it exhibited a slow but continuous creep which did not occur before this cyclic straining; this creep is shown in Fig. 7 for the first 300 hours. Thus the nominal high yield of this alloy is rot a fixed property. It may be reduced to at least half by prior treatments, here prior cycles of strain.

6. Discussion

Alpha-titanium and its alloy Ti/6Aℓ/4V may possibly exhibit weaknesses in service because they do not strain-harden efficiently. In particular they exhibit unduly prolonged creep on yielding, so their stress/strain relationship above yield is meaningless, and they lack the safeguard of hardening which allows normal metal to adjust itself to overstrain. It appears that in assessing whether an alloy can withstand service stresses some attention should be paid not only to its yield strength but also to its capacity for strain-hardening.

It should be possible to tell why the alloy does not strain-harden efficiently from a study of its deformed microstructure. Meanwhile it seems reasonable to make the following comments, based on current theories of mechanical strength. These would attribute its high yield-strength to interstitials, impurity atoms, and phase boundaries, which strongly pin the dislocations responsible for plastic flow; yield becomes the stress needed to pluck dislocations away. In normal metal the dislocations thus set free soon become pinned again and plastic strain ceases; but not, apparently, in titanium and the alloy. An inability to pin mobile dislocations appears to be the basic cause of its special behavior.

This inability, again according to current theories, is likely to arise when interstitial and impurity atoms do not diffuse easily to the mobile dislocations and recapture them. Such slow diffusion in titanium is perhaps to be expected, for it has strong chemical affinity for the common interstitials and impurities, oxygen, carbon, nitrogen. It might be overcome by additions of impurity atoms with less affinity.

The inability to pin dislocations could also account for specific observations. The drawn-out pseudo-elasticity between L.P. and yield would result from odd dislocations which are already free. The halving of an initially high yield stress like that described in Fig. 6 would arise because the prior creep straining at yield is likely to set free a large number of mobile dislocations; these then persist and permit plasticity under a subsequent lower stress; other prior treatments with similar results might be anticipated.

Finally it might be predicted that the instability at room temperature may diminish at higher temperatures because higher temperatures normally expedite diffusion. However, for the same reason, instability could persist at lower temperatures.

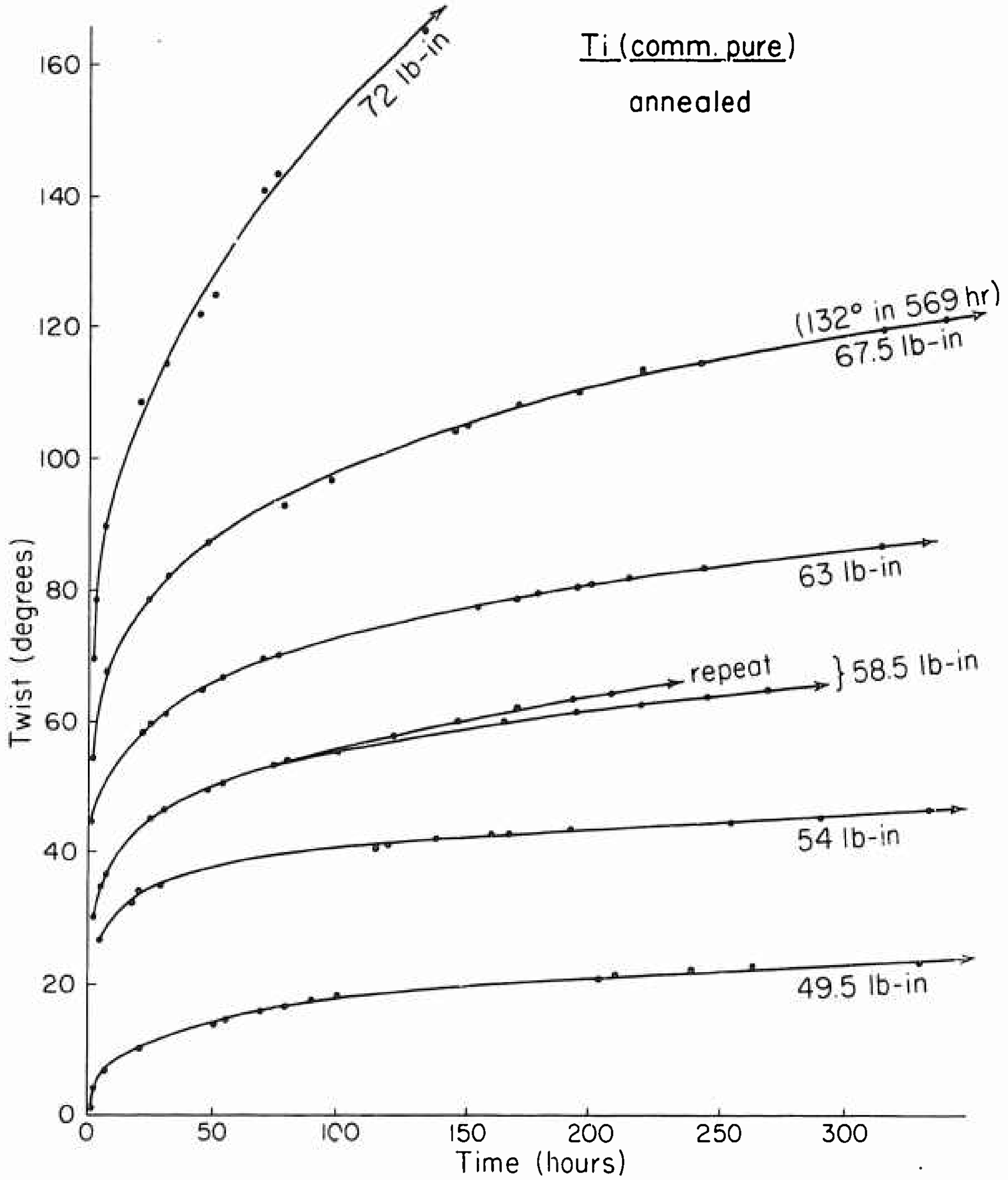

Fig. 1. Room-temperature creep of titanium at yield ~ 50 lb.-in. and above.

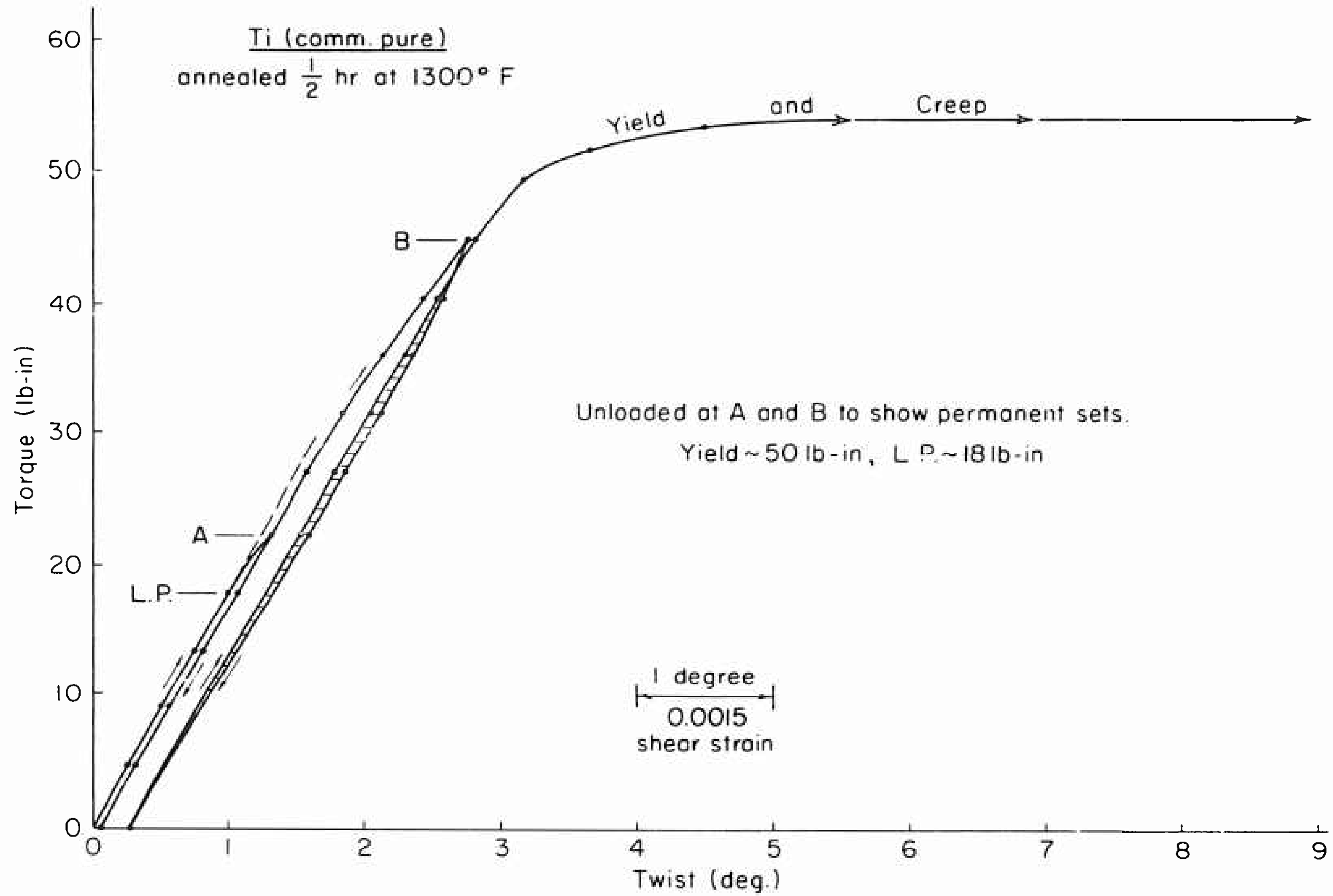

Fig. 2. Torque/twist curve of titanium showing pseudo-elastic range.

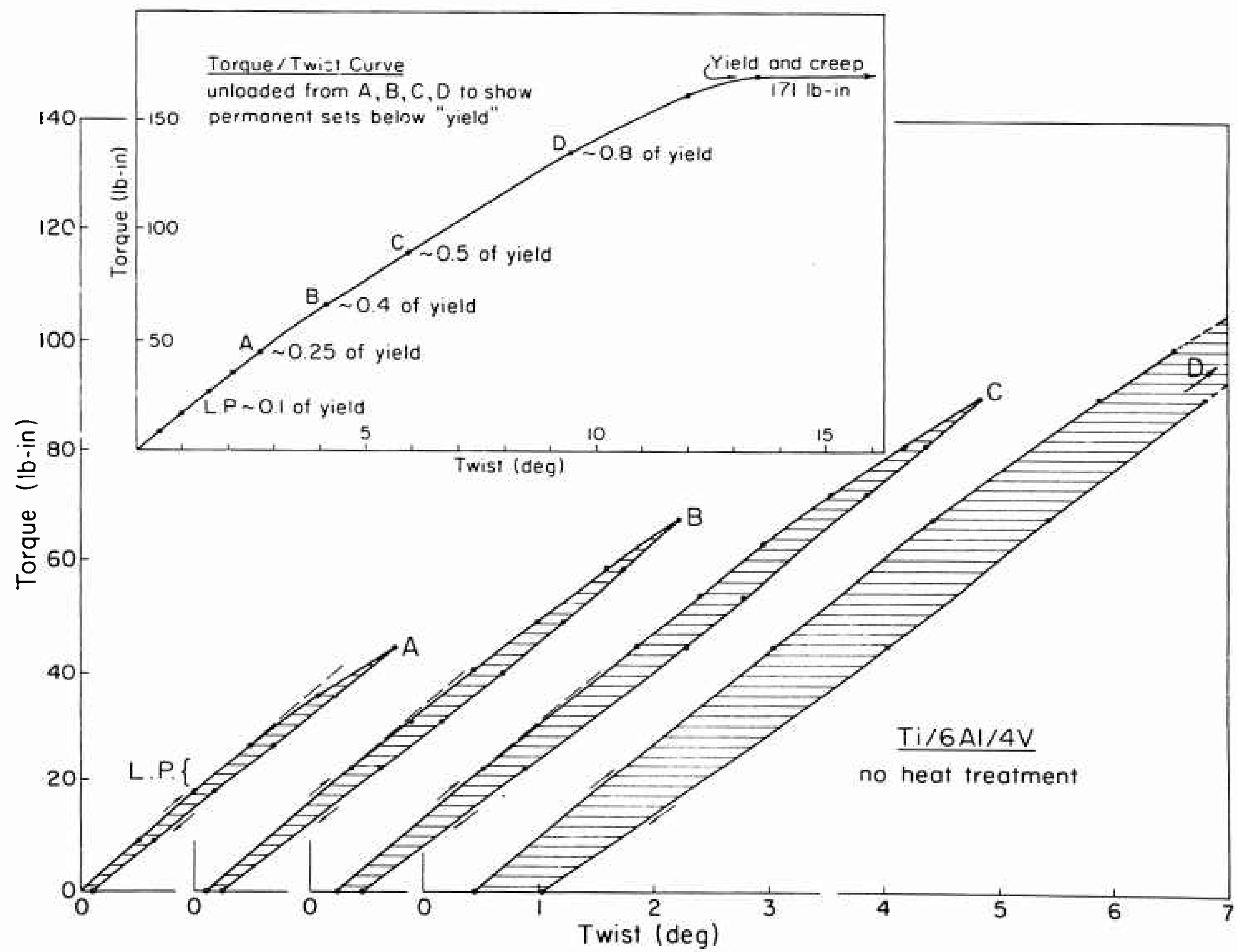

Fig. 3. Pseudo-elastic range in alloy (no heat treatment).

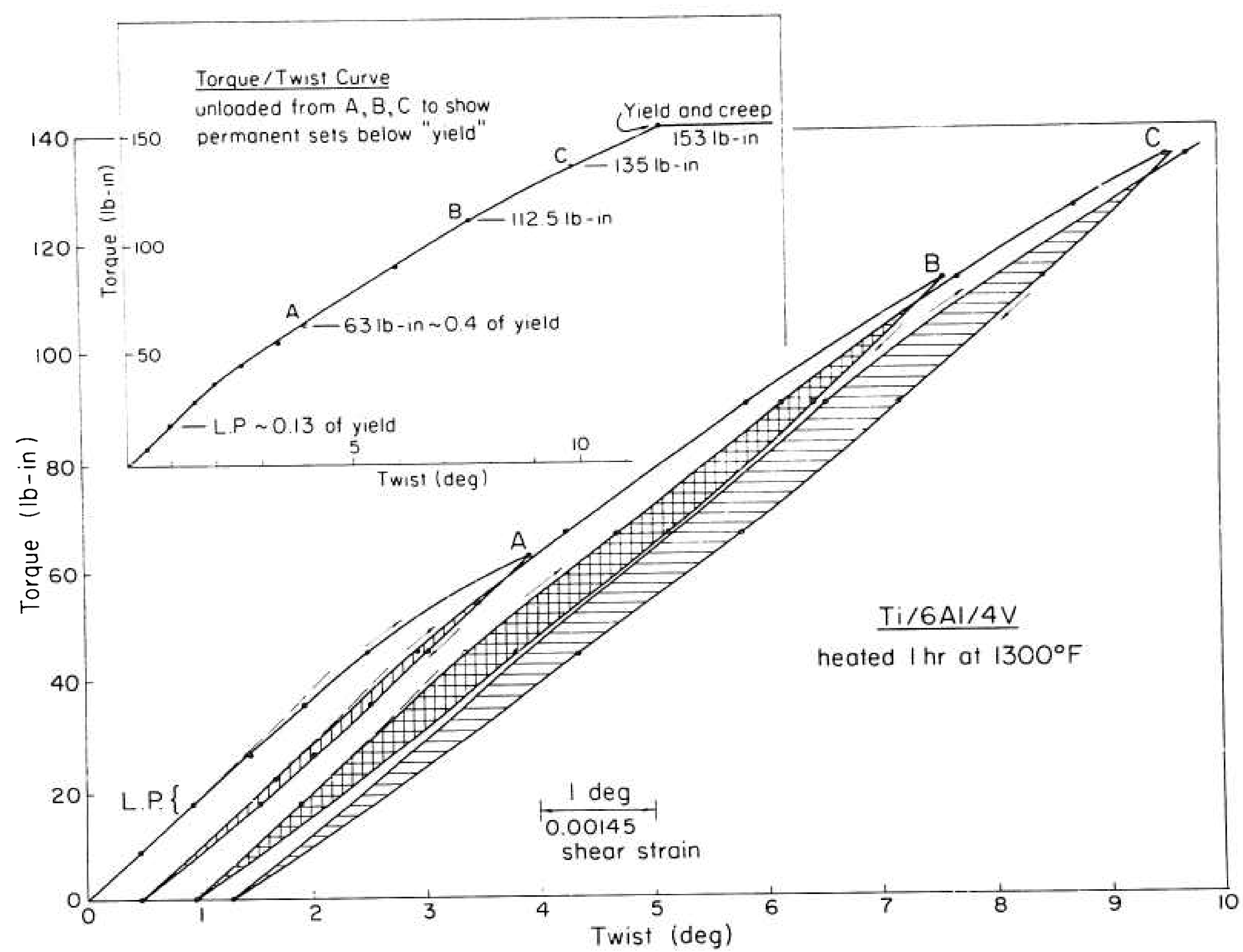

Fig. 4. Pseudo-elastic range in alloy (heat treated).

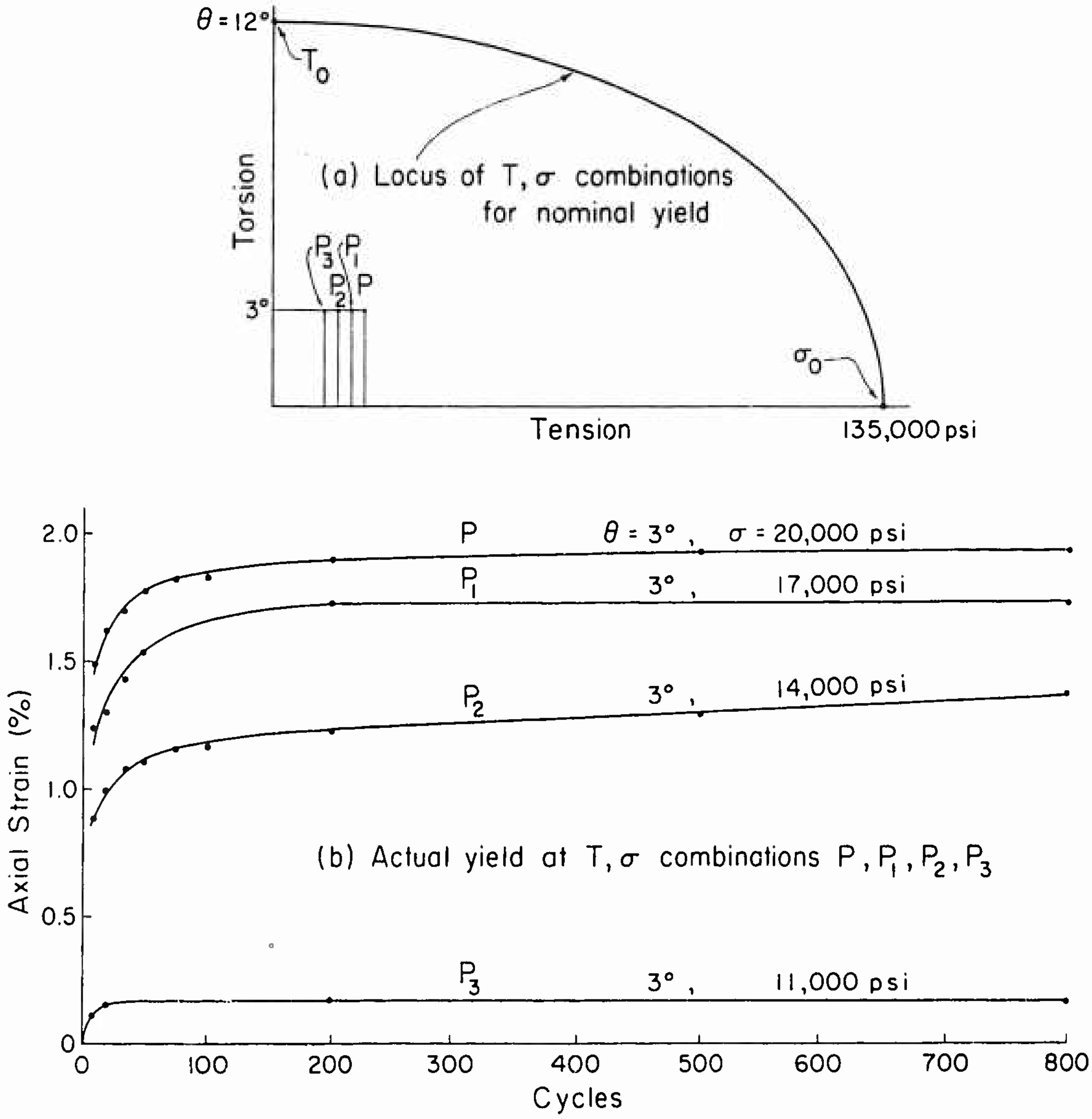

Fig. 5. (a) Schematic Mieses ellipse; (b) Creep for points well within ellipse.

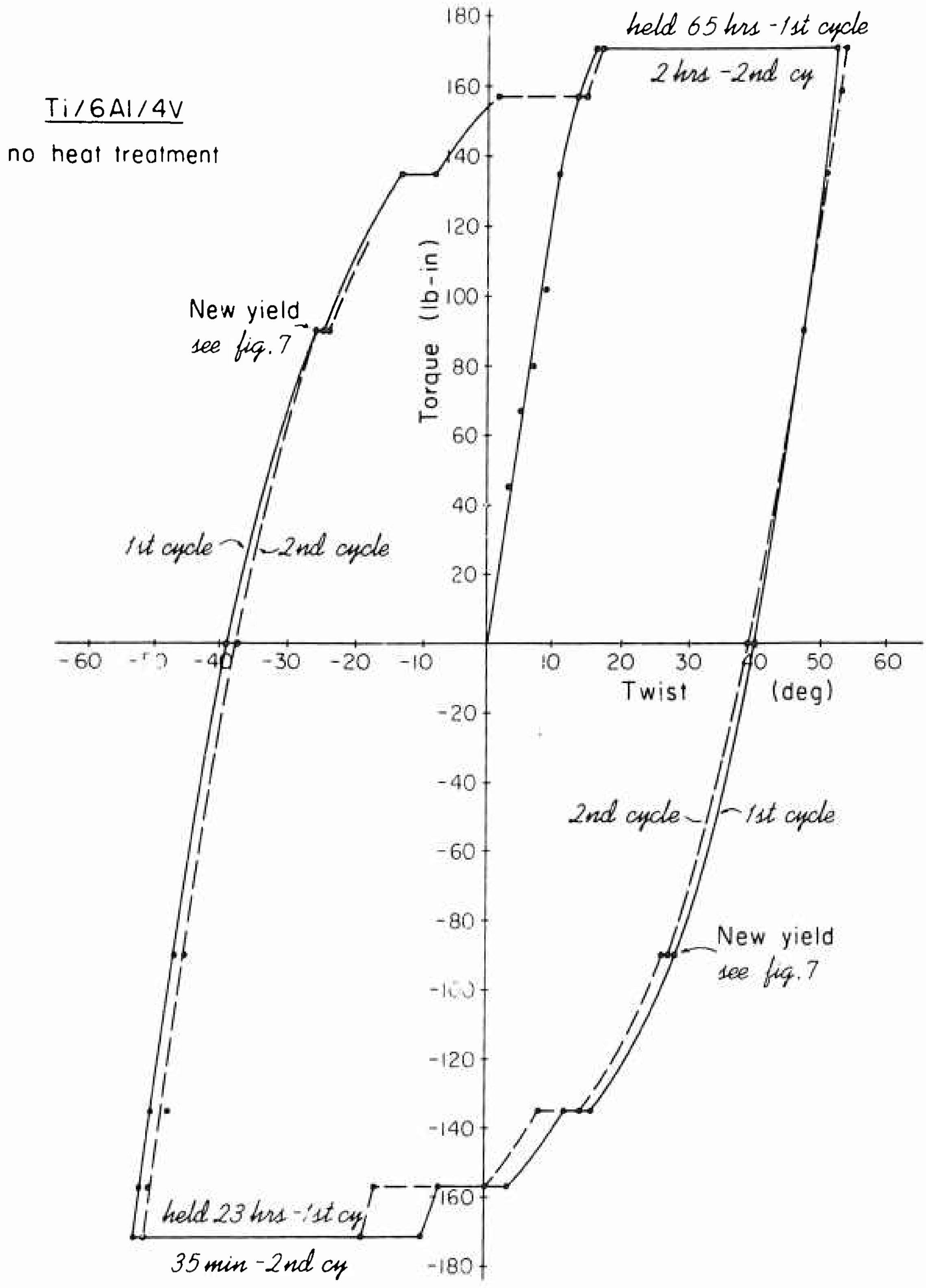

Fig. 6. Instability induced by large-amplitude cycles.

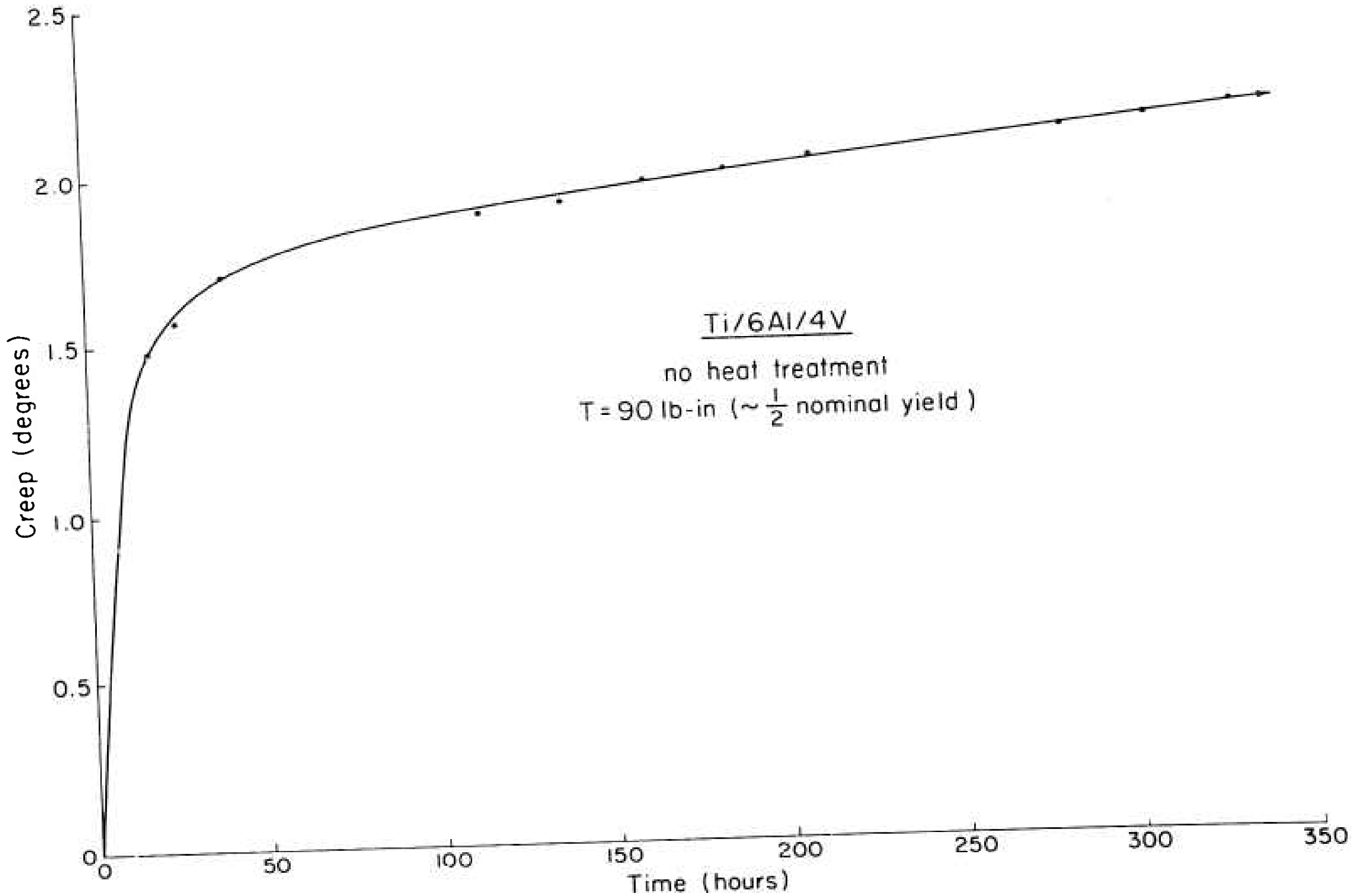

Fig. 7. Creep at yield reduced from 170 to 90 lb.-in. by pre-cycles.

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13. ABSTRACT

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