AD-A031 626		HARVARD UNIV CAMBRIDGE MASS DIV OF ENGINEERING ANDETC F/G 11/6 A MICROSCOPIC MECHANISM FOR STEADY STATE INHOMOGENEOUS FLOW INETC(U) AUG 76 F SPAEPEN TR-7											
	OF AD A031626					- Film - Market States - Market States					F		
							889- 						
				E		The second secon		นสมใหละเป็นไป ในเสนิสัยงไปเป็น ระสะวิที่สินประเทศ	END DATE FILMED				_
	5												
1							_						/





ADA031626

Office of Naval Research Contract NODO14- 67-A-0298-0036 NR-032-544 National Science Foundation Grant DMR-72-03020

A MICROSCOPIC MECHANISM FOR STEADY STATE INHOMOGENEOUS FLOW IN METALLIC GLASSES



By

Frans Spaepen

DDC Reneme NOV 5 1976 SURING B 020

August 1976

Technical Report No. 7

This document has been approved for public release and sale; its distribution is unlimited. Reproduction in whole or in part is permitted by the U. S. Government.

Division of Engineering and Applied Physics Harvard University • Cambridge, Massachusetts

REFORMUNDER D. GOVY ACCESSION NO. 1. BEEGPIENT'S CREACA AURAL Cechnical Report No. 7 TATE INHOMOGENEOUS FLOW IN Interim Report s FERD Interim Report S FERD Inter	RE	PORT DOCUMENTATION PAGE	READ INSTRUCTION BEFORE COMPLETING F
Recenter And Part And Address Recent a remove and address and addres	1. REPORT NUMBER	apart No 7	TACCESSION NO. 3. BECIPIENT'S CATALOG NUMB
Interim Xepoto Interi	Technical R		CONTRACTOR
TATE INHOMOGENEOUS FLOW IN METALLIC GLASSES. Autuonau Trans/Spaepen Trans/Spaepen PERFORMING ORGANIZATION NAME AND ADDRESS Division of Engineering and Applied Physics larvard University ambridge, Massachusetts 02138 CONTROLLING OFFICE NAME AND ADDRESS CONTROLLING STATEMENT (of the Report) This document has been approved for public release and sale; its distrist s unlimited. Reproduction in whole or in part is permitted by the U. Store number) Corriging Statement (of the Report) This document has been approved for public release and sale; its distrist s unlimited. Reproduction in whole or in part is permitted by the U. Store rument. DISTRIBUTION STATEMENT (of the abstract entered in Block 20, If different from Report) Metallic glasses ADDRESS Continue on reverse side if necessary and identify by block number) This deformation map fetallic glasses ADDRESS of deformation, homogeneous and inhomogeneous flow a geneous ased on a dynamic equilibrium between stress-driven creation and dim nihilation of structural disorder. The formalism is developed using ree volume as the order parameter. The boundary line between the heneous and inhomogeneous flow regions on the deformation map is ca ated. The stress-strain relation in inhomogeneous flow approaches TOTH - ADDRESS - Strain relation in inhomogeneous flow approaches COME - ADDRESS - Strain relation in inhomogeneous flow approaches	A MICROSC	OPIC MECHANISM FOR	STEADY
AILTADELIC GLARSES * AILTADELIC GLARSES * * rans/Spaepen <	STATE INH	OMOGENEOUS FLOW I	
Autoria Prans/Spaepen Prans/Spaepen Prans/Spaepen Presonaure controlling of the control of Engineering and Applied Physics Invision of Structural disorder. The formalism is developed using Invision of Structural disorder. The formalism is developed using a tech. The stress-strain relation in inhomogeneous flow approaches Invision of Structural Physics and Invision Physics and Invision Applied Physics Invision Physics Invision Physics and Invision Phy	METALLIC	GLASSES	6. PERFORMING ORG, REPORT N
PERFORMING ORGANIZATION NAME AND ADDRESS IN: Further DMR -72: PERFORMING ORGANIZATION NAME AND ADDRESS II: PROGRAM ELEMENT, PROJECT AREA WORK UNIT NUMBER 032=544 Iarvard University II: PROGRAM ELEMENT, PROJECT AREA WORK UNIT NUMBER 032=544 Iarvard University II: PROGRAM ELEMENT, PROJECT AREA WORK UNIT NUMBER 032=544 CONTROLLING OFFICE NAME AND ADDRESS II: ADDRESS INCHITCHING AGENCY NAME & ADDRESS(II different from Controlling Office) II: SECURITY CLASS. (of this rep Unclass ified Issuer: Distribution STATEMENT (of the Report) II: SECURITY CLASS. (of this rep Unclass ified Tiss unlimited. Reproduction in whole or in part is permitted by the U. Storernment. DISTRIBUTION STATEMENT (of the abstrect entered in Block 20, If different from Report) SUPPLEMENTARY NOTES ABSTRIEUTION STATEMENT (of the abstrect entered in Block 20, If different from Report) SUPPLEMENTARY NOTES ABSTRIEUTION STATEMENT (of the abstrect entered in Block 20, If different from Report) SUPPLEMENTARY NOTES ABSTRIEUTION STATEMENT (of the obstrect entered in Block 20, If different from Report) SUPPLEMENTARY NOTES ABSTRIEUTION STATEMENT (of the obstrect entered in Block 20, If different from Report) CONTROL (Continue on reverse side If inscensory and identify by block number) Comprise of deformation map <td>7. AUTHOR(+)</td> <td></td> <td>(15) NØØ014-67-A-0298-0</td>	7. AUTHOR(+)		(15) NØØ014-67-A-0298-0
PERFORMING ORGANIZATION NAME AND ADDRESS 10. PROGRAM ELEMENT, PROJECT Division of Engineering and Applied Physics 032=544 Cambridge, Massachusetts 02138 032=544 CONTROLLING OFFICE NAME AND ADDRESS 11. RUMBER OF PAGES MONITORING AGENCY NAME & ADDRESS 11. RUMBER OF PAGES MONITORING AGENCY NAME & ADDRESS(II different from Controlling Office) 15. SECURITY CLASS. (of the republic Cation/Downs SCHEDULE INCLUSS STRIBUTION STATEMENT (of the Report) 15. SECURITY CLASS. (of the republic cation/Downs SCHEDULE DISTRIBUTION STATEMENT (of the Report) 15. SECURITY CLASS. (of the republic cation/Downs SCHEDULE DISTRIBUTION STATEMENT (of the Report) 16. SECURITY CLASS. (of the republic cation/Downs SCHEDULE DISTRIBUTION STATEMENT (of the Report) 16. SECURITY CLASS. (of the republic cation/Downs SCHEDULE SUPPLEMENTARY NOTES 15. SECURITY CLASS. (of the republic cation of t	r rans Space		NSF-Grant DMR-72-
ANY STON OF DEGREGENER and Applied Physics C and Chive stores ity Cambridge, Massachusetts 02138 CONTROLLING OFFICE NAME AND ADDRESS MONITORING AGENCY NAME & ADDRESS(II different from Controlling Office) MONITORING AGENCY NAME & ADDRESS(II different from Controlling Office) MONITORING AGENCY NAME & ADDRESS(II different from Controlling Office) DISTRIBUTION STATEMENT (of this Report) This document has been approved for public release and sale; its distristic unclassified Is SECURITY CLASS. (of this report) This document has been approved for public release and sale; its distristic unclassified Is unclassified DISTRIBUTION STATEMENT (of the abstract entered in Block 20, II different from Report) SUPPLEMENTARY NOTES ARE WORDS (Continue on reverse side If necessary and Identify by block number) Ampriceal deformation map fetallic glasses ARESTRACT (Continue on reverse side If necessary and Identify by block number) Ampriceal deformation map fetallic glasses ARESTRACT (Continue on reverse side If necessary and Identify by block number) An empirical deformation map fetallic glasses ARESTRACT (Continue on reverse side If necessary and Identify by block number) An empirical deformation map fetallic glasses ARESTRACT (Continue on reverse side If necessary and Identify by block number) An empirical deformation map for metallic glasses is introduced ne two modes of deformation, homogeneous and inhomogeneous flow a iewed. The microscopic mechanism for steady state inhomogeneous ased on a dynamic equilibrium between stress-driven creation and difficult nuhilation of structural disorder. The formalism is developed using ree volume as the order parameter. The boundary line between the h encous and inhomogeneous flow regions on the deformation map is ca ated. The stress-strain relation in inhomogeneous flow approaches	9. PERFORMING OR	SANIZATION NAME AND ADDRESS	10. PROGRAM ELEMENT, PROJEC
CONTROLLING OFFICE NAME AND ADDRESS CONTROLLING OFFICE NAME AND ADDRESS CONTROLLING OFFICE NAME AND ADDRESS MONITORING AGENCY NAME & ADDRESS(II different from Controlling Office) DISTRIBUTION STATEMENT (of this Report) This document has been approved for public release and sale; its distr s unlimited. Reproduction in whole or in part is permitted by the U. S FOR A Control of the abstract entered in Block 20, II different from Report) SUPPLEMENTARY NOTES ABSTRACT (Continue on reverse side II necessary and Identify by block number) Impirical deformation map fetallic glasses ABSTRACT (Continue on reverse side II necessary and Identify by block number) inpirical deformation map fetallic glasses ABSTRACT (Continue on reverse side II necessary and Identify by block number) Impirical deformation map fetallic glasses ABSTRACT (Continue on reverse side II necessary and Identify by block number) inpirical deformation map fetallic glasses	Harvard Uni	versity	032-544
CONTROLLING OFFICE NAME AND ADDRESS 12. DEFORMENT 4076 August 1976 13. NUMBER OF PAGES MONITORING AGENCY NAME & ADDRESS(II different from Controlling Office) 16. SECURITY CLASS. (of this rep Unclassified 15. SECURITY CLASS. (of this rep DISTRIBUTION STATEMENT (of this Report) 16. SECURITY CLASS. (of this rep This document has been approved for public release and sale; its distr sulfinited. Reproduction in whole or in part is permitted by the U. Sovernment. DISTRIBUTION STATEMENT (of the abstract entered in Block 20, II different from Report) Supplementary notes SUPPLEMENTARY NOTES 13. Supplementary ind identify by block number) Impirical deformation map 14. Interceptory and identify by block number) AMSTRACT (Continue on reverse aids II necessary and identify by block number) Interest from Report) SUPPLEMENTARY NOTES 14. Sover and the interceptory and identify by block number) Impirical deformation map 14. Sover and the interceptory and identify by block number) Impirical deformation map 11. Sover and the interceptory and identify by block number) Impirical deformation map 13. Sover and the interceptory and identify by block number) Impirical deformation map 13. Sover and the interceptory and identify by block number) Impirical deformation map 1	Cambridge,	Massachusetts 02138	
MONITORING AGENCY NAME & ADDRESS(II different from Controlling Office) 34 MONITORING AGENCY NAME & ADDRESS(II different from Controlling Office) 15. SECURITY CLASS. (of this rep Unclass filed 15. SECURITY CLASS. (of this rep DISTRIBUTION STATEMENT (of the Report) 16. DECLASSIFICATION/DOWNG This document has been approved for public release and sale; its distr 16. DECLASSIFICATION/DOWNG SUPPLEMENT (of the abetrect entered in Block 20, II different from Report) 17. NUMBER OF PAGES SUPPLEMENTARY NOTES 18. DECLASSIFICATION from Report) SUPPLEMENTARY NOTES 19. DECLASSIFICATION from Report) SUPPLEMENTARY NOTES 10. State if the second of the	11. CONTROLLING C	FFICE NAME AND ADDRESS	August 1976
34 MONITORING AGENCY NAME & ADDRESS(II different from Controlling Office) Unclass ified Unclass ified Unclass ified DISTRIBUTION STATEMENT (of this Report) Chi is document has been approved for public release and sale; its distribution statement. DISTRIBUTION STATEMENT (of the Report) Chi is document has been approved for public release and sale; its distribution statement. DISTRIBUTION STATEMENT (of the abstract entered in Block 20, If different from Report) SUPPLEMENTARY NOTES MONITICAL deformation map for metalling by block number) Continue on reverse elde if necessary and identify by block number) SUPPLEMENTARY NOTES ABSTRIATI (continue on reverse elde if necessary and identify by block number) Continue on reverse elde if necessary and identify by block number) Continue on reverse elde if necessary and identify by block number) SUPPLEMENTARY NOTES ABSTRACT (Continue on reverse elde if necessary and identify by block number) Continue on reverse elde if necessary and identify by block number) Continue on reverse elde if necessary and identify by block number)	(12 340.	13. NUMBER OF PAGES
Unclassified Uncla	14 MONITORING AG	ENCY NAME & ADDRESS(II dillarent from (Controlling Office) 15. SECURITY CLASS. (of this repo
Ise. DECLASSIFICATION/DOWNG DISTRIBUTION STATEMENT (of the Report) This document has been approved for public release and sale; its distributed. Reproduction in whole or in part is permitted by the U.S. Sovernment. DISTRIBUTION STATEMENT (of the obstract entered in Block 20, if different from Report) SUPPLEMENTARY NOTES REY WORDS (Continue on reverse side if necessary and identify by block number) Compirical deformation map Retailing glasses ABSTRACT (Continue on reverse side if necessary and identify by block number) Compirical deformation map Getailing glasses ABSTRACT (Continue on reverse side if necessary and identify by block number) Compirical deformation map Getailing glasses ABSTRACT (Continue on reverse side if necessary and identify by block number) Compirical deformation map Getailing glasses ABSTRACT (Continue on reverse side if necessary and identify by block number) Compirical deformation map Getailing glasses Supplementary ABSTRACT (Continue on reverse side if necessary and identify by block number) Compirical deformation map Getailing glasses ABSTRACT (Continue on reverse side if necessary and identify by block number) <td>(14) E</td> <td>0 00</td> <td>Unclassified</td>	(14) E	0 00	Unclassified
SCHEDULE DISTRIBUTION STATEMENT (of this Report) This document has been approved for public release and sale; its distription in the state of the spectrum of the production in whole or in part is permitted by the U.S. invernment. DISTRIBUTION STATEMENT (of the abstract entered in Block 20, if different from Report) SUPPLEMENTARY NOTES XEY WORDS (Continue on reverse side if necessary and identify by block number) Impirical deformation map (etallic glasses) An empirical deformation map for metallic glasses is introduced to two modes of deformation, homogeneous and inhomogeneous flow regions on the deformation map is ca ated. The stress-strain relation in inhomogeneous flow approaches	(d)	17-11	154. DECLASSIFICATION DOWNG
DISTRIBUTION STATEMENT (of the Report) This document has been approved for public release and sale; its distri- s unlimited. Reproduction in whole or in part is permitted by the U.S. iovernment. DISTRIBUTION STATEMENT (of the ebetract entered in Block 20, 11 different from Report) SUPPLEMENTARY NOTES REY WORDS (Continue on reverse elde 11 necessary and identify by block number) impirical deformation map fetallic glasses ABSTRACT (Continue on reverse elde 11 necessary and identify by block number) An empirical deformation map fetallic glasses ABSTRACT (Continue on reverse elde 11 necessary and identify by block number) An empirical deformation map fetallic glasses ABSTRACT (Continue on reverse elde 11 necessary and identify by block number) An empirical deformation map fetallic glasses ABSTRACT (Continue on reverse elde 11 necessary and identify by block number) An empirical deformation map for metallic glasses is introduced iewed. The microscopic mechanism for steady state inhomogeneous ased on a dynamic equilibrium between stress-driven creation and dif nnihilation of structural disorder. The formalism is developed using ree volume as the order parameter. The boundary line between the h eneous and inhomogeneous flow regions on the deformation map is ca ated. The stress-strain relation in inhomogeneous flow approaches			SCHEDULE
SUPPLEMENTARY NOTES KEY WORDS (Continue on reverse side if necessary and identify by block number) Compirical deformation map Metallic glasses ABSTRACT (Continue on reverse side if necessary and identify by block number) ADSTRACT (Continue on reverse side if necessary and identify by block number) ADSTRACT (Continue on reverse side if necessary and identify by block number) ADSTRACT (Continue on reverse side if necessary and identify by block number) ADSTRACT (Continue on reverse side if necessary and identify by block number) ADSTRACT (Continue on reverse side if necessary and identify by block number) ADSTRACT (Continue on reverse side if necessary and identify by block number) ADSTRACT (Continue on reverse side if necessary and identify by block number) ADSTRACT (Continue on reverse side if necessary and identify by block number) ADSTRACT (Continue on reverse side if necessary and identify by block number) ADSTRACT (Continue on reverse side if necessary and identify by block number) ADSTRACT (Continue on reverse side if necessary and identify by block number) ADSTRACT (Continue on reverse side if necessary and identify by block number) at the structure of the necessary and identify by block number) ENDER a structure of the necessary and identify by block number) ENDER a structure of the necessary and identify by block number) ENDER a structure of the necessary and identify by block number) ENDER a structure of the necessary and identify by block number) ENDER a structure of the necessary and identify by block number) ENDER a structure of the necessary and identify by block number) ENDER a structure of the necessary and identify by block number) ENDER a structure of the necessary and identify by block number) ENDER a structure of the necessary and identify by block number) ENDER a structure of the necessary and identify by block number) ENDER a structure of the necessary and identify by block number) ENDER a structure of the necessary and identify by block number) ENDER a structu	 16. DISTRIBUTION S This docume is unlimited Government 17. DISTRIBUTION S 	TATEMENT (of this Report) ent has been approved for . Reproduction in whole of TATEMENT (of the abstract entered in Block	public release and sale; its distr or in part is permitted by the U.S k 20, 11 different from Report)
ABSTRACT (Continue on reverse elde if necessary and identify by block number) Ambirical deformation map Addition on reverse elde if necessary and identify by block number) An empirical deformation map for metallic glasses is introduced ne two modes of deformation, homogeneous and inhomogeneous flow a iewed. The microscopic mechanism for steady state inhomogeneous ased on a dynamic equilibrium between stress-driven creation and difficulties of structural disorder. The formalism is developed using ree volume as the order parameter. The boundary line between the h eneous and inhomogeneous flow regions on the deformation map is called. The stress-strain relation in inhomogeneous flow approaches	 16. DISTRIBUTION S This docume is unlimited Government 17. DISTRIBUTION S 	TATEMENT (of this Report) ent has been approved for . Reproduction in whole of TATEMENT (of the abstract entered in Bloc	public release and sale; its distr or in part is permitted by the U.S k 20, it different from Report)
ABSTRACT (Continue on reverse side II necessary and identify by block number) An empirical deformation map for metallic glasses is introduced he two modes of deformation, homogeneous and inhomogeneous flow a iewed. The microscopic mechanism for steady state inhomogeneous ased on a dynamic equilibrium between stress-driven creation and dif nnihilation of structural disorder. The formalism is developed using ree volume as the order parameter. The boundary line between the h eneous and inhomogeneous flow regions on the deformation map is can ated. The stress-strain relation in inhomogeneous flow approaches	 16. DISTRIBUTION S This docume is unlimited Government 17. DISTRIBUTION S 18. SUPPLEMENTAR 	TATEMENT (of this Report) ent has been approved for . Reproduction in whole of TATEMENT (of the abstract entered in Bloc Y NOTES	public release and sale; its distr or in part is permitted by the U.S k 20, 11 different from Report)
ABSTRACT (Continue on reverse side II necessary and identify by block number) An empirical deformation map for metallic glasses is introduced he two modes of deformation, homogeneous and inhomogeneous flow a iewed. The microscopic mechanism for steady state inhomogeneous ased on a dynamic equilibrium between stress-driven creation and dif nnihilation of structural disorder. The formalism is developed using ree volume as the order parameter. The boundary line between the h eneous and inhomogeneous flow regions on the deformation map is can ated. The stress-strain relation in inhomogeneous flow approaches	 DISTRIBUTION S This docume is unlimited Government 17. DISTRIBUTION S 18. SUPPLEMENTAR 19. KEY WORDS (Contemport 	TATEMENT (of this Report) ent has been approved for . Reproduction in whole of TATEMENT (of the abstract entered in Bloc Y NOTES	public release and sale; its distr or in part is permitted by the U.S k 20, if different from Report)
ABSTRACT (Continue on reverse side II necessary and identify by block number) An empirical deformation map for metallic glasses is introduced he two modes of deformation, homogeneous and inhomogeneous flow a iewed. The microscopic mechanism for steady state inhomogeneous ased on a dynamic equilibrium between stress-driven creation and dif nnihilation of structural disorder. The formalism is developed using ree volume as the order parameter. The boundary line between the h eneous and inhomogeneous flow regions on the deformation map is can ated. The stress-strain relation in inhomogeneous flow approaches	 DISTRIBUTION S This docume is unlimited Government 17. DISTRIBUTION S 18. SUPPLEMENTAR 19. KEY WORDS (Con Empirical d 	TATEMENT (of this Report) ent has been approved for . Reproduction in whole TATEMENT (of the abstract entered in Block Y NOTES whotes whose on reverse elde if necessary and ident eformation map	public release and sale; its distr or in part is permitted by the U.S k 20, 11 different from Report)
An empirical deformation map for metallic glasses is introduced he two modes of deformation, homogeneous and inhomogeneous flow a iewed. The microscopic mechanism for steady state inhomogeneous ased on a dynamic equilibrium between stress-driven creation and dif nnihilation of structural disorder. The formalism is developed using ree volume as the order parameter. The boundary line between the h eneous and inhomogeneous flow regions on the deformation map is can ated. The stress-strain relation in inhomogeneous flow approaches	 16. DISTRIBUTION S This docume is unlimited Government 17. DISTRIBUTION S 18. SUPPLEMENTAR 19. KEY WORDS (Con Empirical d Metallic gla 	TATEMENT (of this Report) ent has been approved for . Reproduction in whole of TATEMENT (of the ebstrect entered in Block Y NOTES WINDES Elinue on reverse elde if necessary and ident eformation map SSES	public release and sale; its distr or in part is permitted by the U.S k 20, 11 different from Report)
iewed. The microscopic mechanism for steady state inhomogeneous ased on a dynamic equilibrium between stress-driven creation and din nnihilation of structural disorder. The formalism is developed using ree volume as the order parameter. The boundary line between the h eneous and inhomogeneous flow regions on the deformation map is can lated. The stress-strain relation in inhomogeneous flow approaches	 16. DISTRIBUTION S This docume is unlimited Government 17. DISTRIBUTION S 18. SUPPLEMENTAR 19. KEY WORDS (Con Empirical d Metallic gla 20. ABSTRACT (Cont 	TATEMENT (of this Report) ent has been approved for . Reproduction in whole of TATEMENT (of the obstract entered in Bloc Y NOTES VINUE on reverse elde if necessary and identi eformation map Sses	public release and sale; its distr or in part is permitted by the U.S k 20, if different from Report)
ased on a dynamic equilibrium between stress-driven creation and di- nnihilation of structural disorder. The formalism is developed using ree volume as the order parameter. The boundary line between the h eneous and inhomogeneous flow regions on the deformation map is ca lated. The stress-strain relation in inhomogeneous flow approaches	 16. DISTRIBUTION S This docume is unlimited Government 17. DISTRIBUTION S 18. SUPPLEMENTAR 19. KEY WORDS (Con Empirical d Metallic gla 20. ABSTRACT (Cont the two mod 	TATEMENT (of this Report) ent has been approved for . Reproduction in whole of TATEMENT (of the ebstrect entered in Block TATEMENT (of the ebstrect entered in Block Y NOTES inve on reverse eide if necessary and identi eformation map sses	public release and sale; its distr or in part is permitted by the U.S k 20, if different from Report)
ree volume as the order parameter. The boundary line between the h eneous and inhomogeneous flow regions on the deformation map is ca lated. The stress-strain relation in inhomogeneous flow approaches	 16. DISTRIBUTION S This docume is unlimited Government 17. DISTRIBUTION S 18. SUPPLEMENTAR 19. KEY WORDS (Con Empirical d Metallic gla 20. ABSTRACT (Cont the two mod viewed. Th 	TATEMENT (of this Report) ent has been approved for . Reproduction in whole TATEMENT (of the abstract entered in Bloc Y NOTES	public release and sale; its distr or in part is permitted by the U.S k 20, 11 different from Report)
ated. The stress-strain relation in inhomogeneous flow approaches	 16. DISTRIBUTION S This docume is unlimited Government 17. DISTRIBUTION S 18. SUPPLEMENTAR 19. KEY WORDS (Con Empirical d Metallic gla 20. ABSTRACT (Cont the two mod viewed. Th based on a d annihilation 	TATEMENT (of this Report) ent has been approved for . Reproduction in whole of TATEMENT (of the obstract entered in Block Y NOTES WINDES WINDES Inter on reverse elde If necessary and identi eformation map sses Inter on reverse elde If necessary and identi pirical deformation map es of deformation, homoge e microscopic mechanism ynamic equilibrium betwee of structural disorder.	public release and sale; its distr or in part is permitted by the U.S k 20, if different from Report)
FORM A USA	 DISTRIBUTION S This docume is unlimited Government DISTRIBUTION S DISTRIBUTION S SUPPLEMENTAR SUPPLEMENTAR SUPPLEMENTAR SUPPLEMENTAR SUPPLEMENTAR ABSTRACT (Cont the two mod viewed. Th based on a d annihilation free volume 	TATEMENT (of this Report) ent has been approved for . Reproduction in whole of TATEMENT (of the ebetrect entered in Block TATEMENT (of the ebetrect entered in Block Y NOTES inue on reverse elde if necessary and identi- eformation map sses inue on reverse elde if necessary and identi- eformation map sses	public release and sale; its distr or in part is permitted by the U.S * 20, 11 different from Report) (by by block number) (by by block number) (by by block number) (c) metallic glasses is introduced eneous and inhomogeneous flow and if or steady state inhomogeneous flow and the formalism is developed using The boundary line between the ho
TOTAL TATE EDITION OF I NOV OF IS OPPOLETE	 DISTRIBUTION S This docume is unlimited Government DISTRIBUTION S DISTRIBUTION S DISTRIBUTION S SUPPLEMENTAR SUPPLEMENTAR SUPPLEMENTAR SUPPLEMENTAR ABSTRACT (Cont An em the two mod viewed. Th based on a d annihilation free volume geneous and lated. The 	TATEMENT (of this Report) ent has been approved for . Reproduction in whole of TATEMENT (of the abstract entered in Block TATEMENT (of the abstract entered in Block Y NOTES VINOTES Unue on reverse elde If necessary and identify eformation map sses Inue on reverse elde If necessary and identify pirical deformation, homog e microscopic mechanism ynamic equilibrium betwee of structural disorder. To as the order parameter. inhomogeneous flow regis stress-strain relation in	public release and sale; its distr or in part is permitted by the U.S * 20, if different from Report) (by by block number) (by by block number) (by by block number) (by by block number) (c) metallic glasses is introduced eneous and inhomogeneous flow an a for steady state inhomogeneous f en stress-driven creation and diff (c) for metallism is developed using The boundary line between the ho ons on the deformation map is cal inhomogeneous flow approaches i
JAN 73 14/3 S/N 0102-014-6601 SECURITY CLASSIFICATION OF THIS PAGE (Then	 16. DISTRIBUTION S This docume is unlimited Government 17. DISTRIBUTION S 18. SUPPLEMENTAR 19. KEY WORDS (Con Empirical d Metallic gla 20. ABSTRACT (Conf Empirical d Metallic gla 	TATEMENT (of this Report) ent has been approved for . Reproduction in whole of TATEMENT (of the obstract entered in Block TATEMENT (of the obstract entered in Block Y NOTES tinue on reverse side if necessary and identi- eformation map sses inve on reverse side if necessary and identi- pirical deformation map es of deformation, homog e microscopic mechanism ynamic equilibrium betwee of structural disorder. The as the order parameter. inhomogeneous flow regis stress-strain relation in	public release and sale; its distr or in part is permitted by the U.S * 20, if different from Report) (by by block number) (by by block number) (by by block number) (c) metallic glasses is introduced eneous and inhomogeneous flow and a for steady state inhomogeneous flow and infor metallism is developed using The boundary line between the he ons on the deformation map is call inhomogeneous flow approaches is (c) a flow

1 y .

たいの理論の語りにないという

.....

Unclassified SECURITY CLASSIFICATION OF THIS PAGE(When Data Entered) plastic behavior. Unclassified SECURITY CLASSIFICATION OF THIS PAGE(When Date Entered) 44 - Street -THE REAL PROPERTY AND A DECIMAL OF A DECIMALO OF A DECIMALO OF A DECIMALO OF A DECIMAL OF A DECIMAL OF A DECI THE PLANE

.

-

Office of Naval Research

Contract N00014-67-A-0298-0036 NR-032-544 National Science Foundation Grant DMR-72-03020

A MICROSCOPIC MECHANISM FOR STEADY STATE INHOMOGENEOUS FLOW IN METALLIC GLASSES

k,

	ACCESSION for			
Ву	NTIS WINE Section DDC Buil Conten UNANHOUNCED JUSTIFICATION			
Frans Spaepen	BY DISTRIBUTION/AVAILABILITY CODEB Dist. AVAIL and, or Special			
Technical Report No. 7	A			
This document has been approved for public and sale; its distribution is unlimited. Reprodu whole or in part is permitted by the U. S. Gove	release action in arnment.			

August 1976

The research reported in this document was made possible through support extended the Division of Engineering and Applied Physics, Harvard University by the U. S. Army Research Office, the U. S. Air Force Office of Scientific Research and the U. S. Office of Naval Research under the Joint Services Electronics Program by Contract N00014-67-A-0298-0036 and by the National Science Foundation under Grant DMR-72-03020.

Division of Engineering and Applied Physics

Harvard University · Cambridge, Massachusetts

ABSTRACT

An empirical deformation map for metallic glasses is introduced and the two modes of deformation, homogeneous and inhomogeneous flow are reviewed. The microscopic mechanism for steady state inhomogeneous flow is based on a dynamic equilibrium between stressdriven creation and diffusional annihilation of structural disorder. The formalism is developed using the free volume as the order parameter. The boundary line between the homogeneous and inhomogeneous flow regions on the deformation map is calculated. The stress-strain relation in inhomogeneous flow approaches ideally plastic behavior.

INTRODUCTION: AN EMPIRICAL DEFORMATION MAP FOR METALLIC GLASSES

The deformation map, developed by Ashby and Frost for crystalline systems ¹⁻³ is a convenient means of surveying the various modes of plastic deformation of a material. To construct a map, each mode of deformation is described by a steady state constitutive flow law:

$\dot{\gamma} = f(\tau, T, structure)$

where $\dot{\gamma}$ is the shear strain rate, τ the shear stress, T the temperature; "structure" represents all the relevant structural parameters of the material. The steady state condition implies that the structural parameters are uniquely determined by the external parameters stress and temperature, and hence remain constant during the course of the flow. An appropriate superposition of the flow mechanisms makes it possible to draw constant $\dot{\gamma}$ contours on a map with axes τ and T (usually normalized by the shear modulus μ and the melting point T_{M} , respectively), and to outline regions in which each of the various mechanisms is dominant.

Plotting experimental flow data $y(\tau, T)$ in the same coordinate system results in an empirical deformation map. A comparison of the calculated and empirical maps provides a test of the flow equations. But even if no calculated map is available yet, an empirical map is useful for surveying the data and outlining the various flow mechanisms. Therefore, the available flow data for Pd-based metallic glasses are presented on such an empirical map in Fig. 1. Ideally, only data obtained from samples with identical composition should be compared,

1.



Empirical deformation map for Pd-based metallic glasses. The number at each data point is $\log \epsilon$ (in sec⁻¹). Both stress (σ) and strain rate (ϵ) are uniaxial. The glass transition temperature (T_g) and the crystallization temperature (T_c) are for Pd₈₀Si₂₀ (Ref. 13).

Figure 1

but the relative paucity of the data, together with the large flexibility in preparing amorphous metal alloy compositions make it necessary to group the data together according to their major constituent. With one exception, all the alloys represented in Fig. 1 contain 74-82 at % Pd and 16.5-20 at % Si. Their structural characteristics (density, radial distribution function, modulus, ...) and transformation parameters (glass transition temperature T_g , crystallization temperature T_c , heat of crystallization) vary within acceptable limits, given the typical accuracy of mechanical testing. Since all the data were obtained in either uniaxial tension or compression, the stress (σ) and strain rate (ϵ) are plotted as such and have not been converted to the equivalent shear quantities. Where possible, some approximate constant strain rate contours have been drawn in. The flow stresses obtained from creep tests, 4,7 compression tests, 9, 10, 11 and ultimate tensile strength measurements (the three highest temperature points in Masumoto and Maddin's tensile tests⁵) are probably close to the steady state values. In the other tensile tests, 5-8, 12 however, fracture intervenes before steady state is reached. In these cases, the fracture stress has been plotted, since it comes closest to the steady state value. Above T (the liquidus temperature of the crystalline material with the same composition--not shown in Fig. 1) flow is measured by viscometric methods. No measurements of this type have been reported for Pdbased glasses, but experience from other liquid metal alloys¹⁴ shows that the viscosity in this temperature regime is usually around 10⁻³ Nm⁻²sec.

11

Based on this empirical information, it is now possible to show an approximate picture of what a calculated map should look like. Such

-3-

a schematic map, with stress and temperature axes extended further than in Fig. 1, is shown in Fig. 2. The temperature region between T_c and T_l is unaccessible for mechanical measurements because crystallization intervenes, but it is possible to make a reasonable extrapolation of the strain rate contours through this region.

2. THE FLOW MECHANISMS

Inspection of the samples and fracture surfaces after testing shows that there are two basic modes of deformation: homogeneous flow, where each volume element of the specimen contributes to the strain, and inhomogeneous flow, where the strain is localized in a few very thin shear bands.

a. Homogeneous Flow

1

A and the first of

As indicated on Fig. 2, this mode of deformation occurs at low stresses and high temperatures. In this regime, the flow is close to Newtonian viscous (i.e., $\dot{\gamma} \propto \tau$), as is clear from the spacing of the strain rate contours on the map. In uniaxial tension, a specimen thins down uniformly during deformation. Fracture occurs after extensive plastic flow, when some section of the specimen has narrowed down to zero thickness. For convenience, the region of homogeneous deformation can be subdivided, admittedly somewhat arbitrarily, in three parts, depending on the viscosity of the material.

(i) for $T > T_{\ell}$, the material has a viscosity $\eta \approx 10^{-3} \text{ Nm}^{-2} \text{ sec.}$, and can be characterized as "<u>fluid</u>". Flow measurements must

-4-



1

the state

2 45 -

Figure 2.

-5-

be performed by viscometric methods. Since the viscosity changes only slowly with temperature, ¹⁴ the strain rate contours in this regime are almost horizontal.

(ii) for $T \approx T_g$, the material has, by definition of T_g , a viscosity between 10^{15} and 10^{10} Nm⁻²sec., and could be called "<u>viscous</u>". In this regime, the viscosity, which is measured in a creep test, falls very steeply with increasing temperature. ^{4, 15} Consequently, the strain rate contours are almost vertical around T_g .

(iii) for $T < T_g$, the material has a viscosity $\eta > 10^{15} \text{ Nm}^{-2} \text{sec.}$, and can be called "<u>solid</u>". From the limited number of creep tests available in this regime, ^{7, 16} it seems that the viscosity does not change rapidly with temperature, which is reflected on the map by the strain rate contours becoming more horizontal. However, conclusions about flow in this region must be regarded with caution because of the limited number of experiments available, and the largely unstudied influence of the thermal and mechanical history of the material on the properties. Some careful investigations of these effects are necessary.

b. Inhomogeneous Flow

where is Prairie

This mode of deformation occurs at high stress levels (see Fig. 2). In this regime, the stress is very strain rate insensitive (or, in other words: the stress exponent $n = \partial \log \dot{\gamma} / \partial \log \tau$ is very large), so that the flow is almost ideally plastic. The flow stress for a given strain rate, when normalized by the temperature dependent shear modulus (i.e., τ/μ), is almost constant with temperature, except at low temperatures and near T_g. On the deformation map, this results in very closely spaced, almost horizontal strain rate contours.

In inhomogeneous flow the strain is localized in a few very thin shear bands. In ribbons pulled in tension these shear bands are planes, usually perpendicular to the thin side of the ribbon and at a 45° angle with the tensile axis. ^{5, 6, 17} Slip on these planes can be very extensive. This weakens the specimen locally by decreasing the cross section, until finally fracture occurs along the planes of those shear bands. ^{6, 12} The fracture surface exhibits a typical vein-like pattern, ^{6, 12, 18} which indicates that the mechanism of fracture is the Taylor instability. ¹⁸⁻²⁰

The basic physical process underlying this inhomogeneous flow phenomenon is a local <u>softening</u> of the material. This becomes clear from a number of observations.

(i) localization of the flow in shear bands requires that there is some structural change of the material inside the bands, such that they deform much faster there than in the rest of the specimen.²⁰

(ii) fracture in a tensile specimen occurs, as indicated above, along the plane of a shear band, which makes a 45° angle with the tensile axis. Ordinarily, one would expect fracture to occur along the plane of maximum normal stress, i. e., normal to the tensile axis. This means that if fracture occurs along a shear plane with a smaller resolved normal stress, the material along this plane must have been weakened by some shear-related structural change.

-7-

(iii) the vein pattern on the fracture surface is very similar to that obtained by the pulling apart of two solid surfaces containing a thin viscous layer between them. ^{8,21,22} This observation is consistent with (ii): the shear band consists of a layer of material with a viscosity lower than that of the rest of the specimen, which weakens the plane against fracture by the Taylor instability mechanism.

(iv) differential etching of the shear bands after deformation^{23,24}
 indicates that there are chemical changes associated with flow.
 This is to be expected if the shear-induced structural changes
 do not completely relax out after removal of the load.

3. THE GENERAL FLOW EQUATION

The microscopic mechanism which governs both homogeneous and inhomogeneous flow is illustrated by Fig. 3. It is assumed that macroscopic flow occurs as a result of a number of individual atomic jumps. In order for an atom to jump, it must have a nearest neighbor environment as shown in Fig. 3, i. e., next to it there must be a hole large enough to accomodate its (appropriate hard-sphere equivalent) atomic volume v*. It is reasonable to assume that the atomic positions before and after the jump are positions of relative stability, i. e., local free energy minima. In order to make the atom jump, some activation energy of motion ΔG^{m} must be supplied. If no external force is present, this is obtained from thermal fluctuations; the number of jumps across the activation barrier is the same in both directions; this is the basic microscopic mechanism for diffusion. When an



external force, e.g., a shear stress, is applied, the atomic jumps are biased in the direction of the force; the number of forward jumps across the activation barrier is larger than the number of backward jumps; this results in a net forward flux of atoms and forms the basic mechanism for flow. Quantitatively, this can be expressed as follows. When <u>all</u> the atoms in a specimen, deformed in shear, make one jump each of length λ (\approx l atomic diameter) in the direction of the shear, this would result in a macroscopic shear strain $\gamma \approx 1$. If only a fraction of the atoms jump, the shear strain rate is: $\dot{\gamma} =$ fraction of atoms that make forward jumps/second. Since only a fraction of the total number of atoms in the sample are potential jump sites, i.e., have a large enough hole next to them as in Fig. 3, this can be written as

 $\dot{\gamma}$ = (fraction of potential jump sites) \times (net number of forward jumps on each of those sites per second)

(1)

The fraction of potential jump sites is calculated in the free volume theory of Turnbull and Cohen. $^{25-27}$ The free volume of an atom is, intuitively, that part of its nearest neighbor cage in which the atom can move around without an energy change. In an amorphous system, the free volume is distributed statistically among all atoms. In Cohen and Turnbull's theory²⁵ the probability p(v) dv of finding an atom with a free volume between v and v + dv is calculated to be:

 $p(v) dv = \frac{\gamma}{v_f} \exp \left(-\frac{\gamma v}{v_f}\right) dv$

-10-

where γ is a geometrical factor between 1 and 1/2, and v_f is the average free volume of an atom.

In order for an atom to be on a potential jump site, its free volume must be larger than v*, the effective hard-sphere size of the atom. Therefore, the total probability that an atom is on a potential jump site is:

$$\int_{\mathbf{v}^{*}}^{\infty} \frac{\gamma}{\mathbf{v}_{f}} \exp\left(-\frac{\gamma \mathbf{v}}{\mathbf{v}_{f}}\right) d\mathbf{v} = \exp\left(-\frac{\gamma \mathbf{v}^{*}}{\mathbf{v}_{f}}\right)$$
(2)

It is also necessary to include a factor Δf , the fraction of the sample volume in which potential jump sites can be found. For example, in homogeneous flow, where the total volume contributes to flow, $\Delta f = 1$; in inhomogeneous flow, where everything happens in a few very thin bands, $\Delta f \ll 1$. Therefore,

fraction of potential jump sites = $\Delta f \exp(-\gamma v_f)$ (3)

The net number of forward jumps per second on a potential jump site is obtained from simple rate theory (see Fig. 3). The shear stress τ exerts a force τa on an atom, where a is the projected area of the atom onto the plane of shear. When this atom makes a jump of length λ , the work done is $\tau a \lambda$. Since λ is roughly equal to an atomic diameter, the work done is $\tau \Omega$ where Ω is the atomic volume. The free energy of the atom after the jump is therefore decreased by $\Delta G = \tau \Omega$. The net number of forward jumps per second can now be calculated as the difference between a forward flux over an activation barrier $\Delta G^{m} - \Delta G/2$ and a backward flux over an activation barrier $\Delta G^{m} + \Delta G/2$. Assuming an equal distribution of atoms over the two

-11-

equilibrium positions gives:

net number of forward jumps per second per potential jump site

$$= \upsilon \exp\left(-\frac{\Delta G^{m} - \tau \Omega/2}{kT}\right) - \exp\left(-\frac{\Delta G^{m} + \tau \Omega/2}{kT}\right)$$
(4)

where υ is the frequency of atomic vibration (~ Debye frequency). Substituting (3) and (4) into (1) gives:

$$\dot{\gamma} = \Delta f \exp\left(-\frac{\gamma v^*}{v_f}\right) 2_{\cup} \sinh\left(\frac{\tau \Omega}{2kT}\right) \exp\left(-\frac{\Delta G^m}{kT}\right)$$
 (5)

This is the general flow equation. In the case of homogeneous deformation ($\Delta f = 1$) at low stress levels ($\tau \Omega \ll 2kT$), this leads to Newtonian viscous behavior (2 sinh($\tau \Omega/2kT$) $\approx \tau \Omega/kT$), and an expression for the viscosity can be derived:

$$\eta = \frac{\tau}{\dot{\gamma}} \approx \frac{kT}{\upsilon \Omega} \exp\left(\frac{\gamma \upsilon^*}{v_f}\right) \exp\left(\frac{\Delta G^m}{kT}\right)$$
(6)

The diffusion coefficient in a three dimensional random walk process is defined as $D = \frac{1}{6} \Gamma \lambda^2$ where Γ and λ are the jump frequency and distance, respectively. An analysis, similar to that of the flow problem gives:

> Γ = (probability that an atom is on a potential jump site) × (number of jumps an atom on this site makes per second)

or:

$$\Gamma = \exp\left(-\frac{\gamma v^*}{v_f}\right) \cup \exp\left(-\frac{\Delta G^m}{kT}\right)$$
(7)

This gives for the diffusion coefficient:

$$D = \frac{1}{6} \upsilon \exp\left(-\frac{\gamma \upsilon^*}{\upsilon_f}\right) \exp\left(-\frac{\Delta G^m}{kT}\right) \lambda^2$$
(8)

It is worth noting that combining Equations (6) and (8) gives

$$D\eta = \frac{kT}{6(\Omega/\lambda^2)}$$

which is, save for a numerical factor of order 1, the Stokes-Einstein equation.

4. THE SOFTENING MECHANISM IN INHOMOGENEOUS FLOW

As has been discussed in Section 2b, during inhomogeneous flow, the material in the shear bands undergoes some structural change that leads to a local lowering of the viscosity.

Polk and Turnbull¹⁴ have proposed that this structural change is the net result of two competing processes: a shear-induced disordering and a diffusion controlled reordering process. This concept will be extended here, by deriving a formal expression for the dynamic steady state equilibrium between the two processes. The free volume will be chosen as the order parameter, since it makes the problem mathematically tractable. It should be kept in mind, however, that to the extent that the free volume is equivalent to other possible order parameters (compositional, entropic,...), the result should be quite generally valid.

In the free volume formulation of the flow Equation (5), the structural parameter which governs the viscosity is the average free volume v_f . Therefore, if there is to be a lowering of the viscosity in the shear bands, there must be an increase of the free volume.

It is proposed that, at high stress levels, free volume can be created by the mechanism shown on Fig. 4. At a sufficiently high stress, an atom with hard-sphere volume v* can be squeezed into a neighboring hole with a smaller volume v. This makes the neighbors of the new position move out, and creates a certain amount of free volume.

Competing with this is a relaxation process, which tends to annihilate the extra free volume created and restore the system to its initial structural state. The structural rearrangements necessary to annihilate free volume consist of a series of diffusional jumps.

In steady state a dynamic equilibrium is reached between the two competing processes: an equal amount of free volume is created by the stress-driven process, as is annihilated by the diffusional process.

a. Creation of Free Volume

The energy necessary to squeeze an atom with volume v^* into a smaller hole of volume v (see Fig. 4) can be approximated by the elastic distortion energy required to squeeze a sphere with volume v^* into a spherical hole with volume v in a matrix of the same material. This has been calculated by Eshelby:²⁸

$$\Delta G^{e} = S \frac{(v^{*} - v)^{2}}{v}$$

with

$$S = \frac{2}{3}\mu \frac{1+\upsilon}{1-\upsilon}$$

If an atom makes the jump over the activation barrier in the direction of shear, its free energy is decreased by the driving term $\tau\Omega$ (see

-14-



Figure 4.

$$\Delta G = \tau \Omega - S \frac{(v^* - v)^2}{v}$$

The volume v_m of the smallest hole that an atom can still be squeezed into by a given stress is determined by $\Delta G = 0$, or:

$$\tau \Omega = S \frac{\left(v^* - v_m\right)^2}{v_m}$$
(7)

If $\tau \ll S$, this gives:

$$v_m \approx v^*(1 - \sqrt{\tau/S})$$

The amount of free volume created per second, $\Delta^+ v_f$, can now be calculated from simple rate theory.

Consider the amount of free volume created per second associated with the squeezing open of holes with a volume between v and v + dv.

 $d(\Delta^{+}v_{f}) = (number of potential sites) \times (net number of forward jumps on each of those sites per second)$

× (amount of free volume created per jump)

The number of potential sites is equal to the total number of atoms (N) multiplied by the probability that they have a neighboring hole with volume between v and v + dv. This factor is obtained from the free volume theory (see Equation (2)).

So

number of potential sites =
$$N \frac{Y}{v_f} \exp\left(-\frac{Yv}{v_f}\right) dv$$
 (9)

The net number of forward jumps is calculated as the difference between a forward and a backward flux, analogously to Equation (4), but with a different $\Delta G = \tau \Omega - \Delta G^{e}$. So:

net number of forward jumps per potential site per second

$$= \upsilon \left[\exp \left(-\frac{\Delta G^{m} - \tau \Omega/2 + \Delta G^{e}/2}{kT} \right) - \exp \left(-\frac{\Delta G^{m} + \tau \Omega/2 - \Delta G^{e}/2}{kT} \right) \right]$$
(10)

The amount of free volume created per jump is simply the difference between the volume of the atom and the volume of the hole:²⁸

The total amount of free volume created per second is obtained by integrating all the elementary contributions $d(\Delta^+ v_f)$ over all the hole sizes v, from v_m (the minimum that can be squeezed open) to v* (above which no free volume is created). This, combined with Equations (9), (10), and (11) gives:

$$\Delta^{+} \mathbf{v}_{\mathbf{f}} = \int_{\mathbf{v}_{\mathbf{m}}}^{\mathbf{v}^{*}} \mathbf{d}(\Delta^{+} \mathbf{v}_{\mathbf{f}})$$

$$= \int_{\mathbf{v}_{\mathbf{m}}}^{\mathbf{v}^{*}} N \frac{\gamma}{\mathbf{v}_{\mathbf{f}}} \exp\left(-\frac{\gamma \mathbf{v}}{\mathbf{v}_{\mathbf{f}}}\right) \cup \left[\exp\left(-\frac{\Delta \mathbf{G}^{\mathbf{m}} - \tau \Omega/2 + S(\mathbf{v}^{*} - \mathbf{v})^{2}/2\mathbf{v}}{\mathbf{k}T}\right)\right]$$

$$- \exp\left(-\frac{\Delta \mathbf{G}^{\mathbf{m}} + \tau \Omega/2 - S(\mathbf{v}^{*} - \mathbf{v})^{2}/2\mathbf{v}}{\mathbf{k}T}\right) \left[(\mathbf{v}^{*} - \mathbf{v}) \mathbf{d}\mathbf{v}\right]$$

To simplify the integration, the following assumption is made:

 $\tau \ll S$ and $v \approx v^*$

-17-

Therefore,

$$\frac{\gamma}{v_{f}} \exp\left(-\frac{\gamma v}{v_{f}}\right) \approx \frac{\gamma}{v_{f}} \exp\left(-\frac{\gamma v^{*}}{v_{f}}\right)$$
$$S \frac{\left(v^{*} - v\right)^{2}}{2v} \approx S \frac{\left(v^{*} - v\right)^{2}}{2v^{*}}$$

This gives:

Acres . . Planet

$$\Delta^{+} \mathbf{v}_{f} = \mathbf{N}_{\upsilon} \exp\left(-\frac{\Delta \mathbf{G}^{\mathbf{m}}}{\mathbf{k}\mathbf{T}}\right) \frac{\gamma}{\mathbf{v}_{f}} \exp\left(-\frac{\gamma \mathbf{v}^{*}}{\mathbf{v}_{f}}\right) \left\{ \exp\left(\frac{\tau \Omega}{2\mathbf{k}\mathbf{T}}\right) \frac{\mathbf{v}^{*}\mathbf{k}\mathbf{T}}{\mathbf{S}} \left[1 - \exp\left(-\frac{\mathbf{S}\left(\mathbf{v}^{*}-\mathbf{v}_{\mathbf{m}}\right)^{2}}{2\mathbf{k}\mathbf{T}\mathbf{v}^{*}}\right)\right] + \exp\left(-\frac{\tau \Omega}{2\mathbf{k}\mathbf{T}}\right) \frac{\mathbf{v}^{*}\mathbf{k}\mathbf{T}}{\mathbf{S}} \left[1 - \exp\left(\frac{\mathbf{S}\left(\mathbf{v}^{*}-\mathbf{v}_{\mathbf{m}}\right)^{2}}{2\mathbf{k}\mathbf{T}\mathbf{v}^{*}}\right)\right] \right\}$$

In view of Equation (7), which defines v_m , this reduces to the final result:

$$\Delta^{+} v_{f} = \frac{\gamma v^{*}}{v_{f}} \frac{2kT}{S} \left[\cosh\left(\frac{\tau \Omega}{2kT}\right) - 1 \right] N_{\upsilon} \exp\left(-\frac{\Delta G^{m}}{kT}\right) \exp\left(-\frac{\gamma v^{*}}{v_{f}}\right) \quad (12)$$

b. Annihilation of Free Volume

The structural rearrangemetns, required for annihilation of extra free volume, consist of a series of diffusive jumps. Let n_D be the number of diffusive jumps necessary to annihilate a free volume equal to v*. It is expected that n_D should be a small number, between 1 and 10.

This expectation is made plausible by the observation of a twodimensional amorphous dynamic hard-sphere model.²⁹ In two dimensions, an amorphous state of high density can only be obtained by using

-18-

different size spheres. In contrast with a crystalline system, where a vacancy is a stable entity after many diffusive jumps, in an amorphous system an artificially created "vacancy" is annihilated after a small number of diffusive jumps.

Another useful analogy that can be made comes from Nabarro-Herring creep in crystals. (This analogy must obviously be used with caution, since amorphous systems are topologically quite different from an assembly of microcrystals.) In N. -H. creep, vacancies travel a distance on the order of the grain diameter, and are annihilated at the grain boundaries. In an amorphous system something similar can be imagined, but with a very small equivalent grain size. In crystalline systems, vacancies can annihilate at the grain boundaries because there the structural requirement of crystalline translational symmetry is relaxed. In an amorphous system, this requirement does not exist at all, and hence it is expected that "vacancies" or "extra free volume" can be annihilated in just a few diffusional jumps.

The amount of free volume annihilated per second is

 $\Delta^{-}v_{f}^{-}$ = (free volume annihilated per jump) × (number of jumps per second) The amount of free volume annihilated per jump is, given the above definitions, v*/n_D. The number of jumps is NГ, where the jump frequency Γ is given by Equation (7). Therefore,

$$\Delta \bar{v}_{f} = \frac{v^{*}}{n_{D}} N \cup \exp\left(-\frac{\gamma v^{*}}{v_{f}}\right) \exp\left(-\frac{\Delta G^{m}}{kT}\right)$$
(13)

-19-

c. Steady State

In steady state, the amount of free volume created is equal to the amount of free volume annihilated:

$$\Delta^+ \mathbf{v}_{\mathbf{f}} = \Delta^- \mathbf{v}_{\mathbf{f}}$$

Or, from Equations (12) and (13):

$$\frac{\gamma v^*}{v_f} \frac{2kT}{S} \left[\cosh\left(\frac{\tau \Omega}{2kT}\right) - 1 \right] = \frac{v^*}{n_D}$$
(14)

Before the sample is loaded, the average free volume v_f is set by the structure of the quenched metallic glass. At low stresses, the left hand term in (14) is smaller than v^*/n_D , i.e., the extra free volume created by the stress can easily be annihilated by the diffusion, and the average free volume stays at the initial value v_f . As soon as τ exceeds the value for which (14) is obeyed, however, more free volume is being created than can be annihilated by diffusion; hence v_f will keep on increasing until condition (14) is obeyed again. Or, in other words: the free volume v_f is set by the stress τ .

Formally, this can be seen by rewriting Equation (14):

$$\frac{\gamma v^*}{v_f} = \frac{v^*}{n_D} \frac{S}{2kT} \left[\cosh\left(\frac{\tau \Omega}{2kT}\right) - 1 \right]^{-1}$$
(15)

When this explicit expression for $\gamma v^*/v_f$ is substituted in the term $\exp(-\gamma v^*/v_f)$ of the general flow Equation (5), a flow equation for inhomogeneous flow is obtained.

5. DISCUSSION

The only unknown parameter in the flow equations which prevents us from constructing a complete calculated deformation map is ΔG^{m} . However, if it is assumed that ΔG^{m} , which may be a function of T and τ , is the same for both homogeneous and inhomogeneous flow, it becomes possible to calculate the boundary between the regions in which each of the mechanism s dominates. The boundary between the two mechanism regions on the deformation map is a line $\tau(T)$, where

$$\dot{Y}_{homogeneous} (\tau, T) = \dot{Y}_{inhomogeneous} (\tau, T)$$
 (16)

The only factors in the general flow Equation (5) that are different from one mechanism to the other are Δf and $\exp(-\gamma v^*/v_f)$. So condition (16) becomes then

$$\Delta f_{h} \exp\left(-\frac{\gamma v^{*}}{v_{f}}\right)_{h} = \Delta f_{i} \exp\left(-\frac{\gamma v^{*}}{v_{f}}\right)_{i}$$

or

$$\left(\frac{\gamma \mathbf{v}^*}{\mathbf{v}_f}\right)_i = \left(\frac{\gamma \mathbf{v}^*}{\mathbf{v}_f}\right)_h - \ln\left(\frac{\Delta f_h}{\Delta f_i}\right)$$
(17)

Since the free volume v_f is a structural parameter, and since all the data in the vicinity of the mechanism boundary are obtained at temperatures far enough below T_g where the configuration can be considered frozen, $(\gamma v^*/v_f)_h$ is a constant in the equation of the boundary line. Using Equation (15), which determines $(v^*/v_f)_i$, Equation (17) can be rewritten as:

$$\frac{\tau}{\mu} = \frac{4kT}{\mu\Omega} \operatorname{arcsinh} \left(\alpha \, \frac{kT}{S\Omega} \right)^{-1/2} \tag{18}$$

with

$$\mathbf{x} = 4 \left[\left(\frac{\gamma \mathbf{v}^*}{\mathbf{v}_f} \right)_h - \ell n \frac{\Delta f_h}{\Delta f_i} \right] \frac{n_D}{\mathbf{v}^* / \Omega}$$
(19)

(18) is the equation of the mechanism boundary line on the deformation map. It contains only one adjustable parameter: α . The other quantities are known from experiment.

The temperature dependent shear modulus is:

$$\mu(T) = \mu_0 \left(1 - \frac{1}{\mu_0} \frac{d\mu}{dT} (T - 300) \right)$$
,

where μ_0 is the shear modulus at 300 K. The values used here are for $Pd_{77.5}Cu_6Si_{16.5}$:

$$L_0 = 3.29 \times 10^{10} \text{ Nm}^{-2}$$
 (from Chen et al. ³⁰)

and

$$\frac{1}{\mu_0} \frac{d\mu}{dT} = 2.6 \times 10^4 \,\mathrm{K}^{-1} \qquad (\text{from Chen}^{31})$$

The atomic volume is calculated from the density measurements³² on $Pd_{85.5}Si_{16.5}: \Omega = 1.46 \times 10^{-29} m^3$. The quantity $S = \frac{2}{3} \mu (1+\nu/1-\nu)$ is calculated using a Poisson's ration $\nu = 0.4$, which is typical for metallic glasses.³⁰

Figure 5 is an enlarged portion of the empirical deformation map of Fig. 1, showing the boundary between the regions of homogeneous and inhomogeneous flow. The creep data at the bottom of the diagram (Maddin and Masumoto, ⁷ open squares) are obviously all in the homogeneous deformation region. The transition between inhomogeneous and homogeneous flow, however, shows up in Masumoto and Maddin's tensile data⁵ (open circles). The three data points at the highest temperature (533 - 583K) are ultimate tensile strengths obtained from stress-strain curves which show very extensive plastic deformation (several percent) before fracture. Therefore, these points belong in the region of homogeneous flow. The parameter α in Equation (13) can now be adjusted such that the resulting boundary line passes between these points and the rest of the tensile data; for the line shown in Fig. 5, $\alpha = 360$.

It remains to be shown now that this is a physically realistic number. Therefore, the quantities in (19), the defining relation for α , must be investigated.

As discussed in Section 3, for homogeneous flow $\Delta f_h = 1$. For inhomogeneous flow Δf is determined by the thickness of the softened layer in the shear band. Since no direct measurements on this are available, the thickness of this layer can only be estimated, e.g., from the height and size of the little "tributaries" in the fracture surface vein-pattern. ^{18, 12} A layer thickness of 500Å seems to be a reasonable estimate. If the sample length is 5 cm. and there are only a few shear bands, $\Delta f_i = 10^{-6}$. This is, of course, only an estimate, but, fortunately, the equation is quite insensitive to the exact value of Δf_i .

Chen and Goldstein¹⁴ have measured

$$\frac{Yv^*}{v_f} = \frac{3180}{T - 515}$$

in $Pd_{77.5}Cu_6Si_{16.5}$, for homogeneous flow in the region $T \approx T_g$. As mentioned above, in the region under consideration here v_f is considered constant, due to configurational freezing at some temperature around T_g . For T = 600, the viscosity of $Pd_{77.5}Cu_6Si_{16.5}$ is $10^{14} Nm^{-2} sec$;⁴ below this the system will be considered configurationally frozen. The appropriate quantity to be used in Equation (19)



Enlarged portion of the empirical deformation map for Pd-based metallic glasses. The equivalent shear stress (τ) and shear strain ($\dot{\gamma}$) have been calculated from the uniaxial values by the von Mises criterion ($\tau = \sigma/3$; $\dot{\gamma} = 3\dot{\epsilon}$). The strain rate contours connect data points from the same reference. The shear modulus is from Refs. 30 and 31. The heavy line is the boundary between regions of homogeneous and inhomogeneous is calculated from Eq. (18).

Figure 5.

is therefore:

$$\left(\frac{\gamma v^*}{v_f}\right)_h = \frac{3180}{600 - 515} = 37.4$$

Substituting the values for α , Δf_h , Δf_i , and $(\gamma v^*/v_f)_h$ in (19) leaves

$$\frac{\mathbf{v}^*/\Omega}{n_{\rm D}} = 0.26$$

This is a physically reasonable result. The equivalent hard-sphere volume of a late transition metal as Pd probably takes up a large fraction of the atomic volume, say, $v^*/\Omega \approx .8$. This leaves $n_D \approx 3$, which is consistent with the physical considerations of Section 4b, where it was argued that the number of diffusional jumps necessary to annihilate a free volume equal to v^* should be somewhere between 1 and 10.

It may be worth noting that the mechanism boundary on the deformation map has a positive slope, i.e., with increasing temperature it is necessary to go to higher stresses to get inhomogeneous flow. Physically, this is not unexpected, since the diffusional relaxation process becomes relatively more important at high temperatures.

At this stage, the stress exponent $n = \partial \log \dot{\gamma}/\partial \log \tau$ for the two flow modes can also be calculated, if it is assumed that ΔG^{m} is independent of τ . The results of this calculation are shown in Fig. 6. The stress exponent for homogeneous flow is $n \approx 1$, as it should be for Newtonian viscous flow. For inhomogeneous flow $n \approx 50$, which indicates that it approaches ideally plastic behavior.

Finally, it should be emphasized that the present mechanism involves monatomic systems, and does not explicitly include any chemical effects. However, chemical ordering is reflected partially in the



Figure 6.

free volume formulation, for example, if an efficient packing of atoms of different size (as Pd and Si) requires a non-random chemical environment for the two atoms. In this case, creation of free volume is equivalent to creation of chemical disorder. On the other hand, there are probably some specific chemical effects, as the metal-metalloid bonding, which should be included in ΔG^{m} . The differences in strength found by changing the metalloid components in a metallic glass are probably of this nature. However, it is also possible that an increased diffusivity of one type of metalloid to the other might lead to an increased diffusional relaxation, and hence a higher strength. For example, the exceptionally high strength of $Fe_4 B$ glass³³ might conceivably be caused by the higher diffusivity of the small B-atom.

Acknowledgements

It is a pleasure to acknowledge several useful suggestions and criticisms from Professors D. Turnbull and M.F. Ashby. I am indebted to Dr. H.J. Frost for his help in constructing the deformation maps and for several helpful discussions.

This work has been supported in part by grants from the Office of Naval Research under Contract N00014-67-A-0298-0036 and from the National Science Foundation under Contract DMR-72-03020.

I also want to acknowledge a postdoctoral Fellowship from the I. B. M. Corporation.

-27-

References

in double with a transfer and a

A LAN

1.	M.F. Ashby, <u>Acta Met</u> . <u>20</u> , 887 (1972).
2.	H. J. Frost and M. F. Ashby, <u>Proc. of the John E. Dorn Symposium</u> , J. C. M. Li and A. K. Mukherjee (eds.), A. S. M., Cleveland, Ohio, 70 (1975).
3.	M.F. Ashby and H.J. Frost, "Constitutive Equations in Plasticity", A.S. Argon (ed.), M.I.T. Press, Cambridge, Mass., 117 (1975).
4.	H.S. Chen and M. Goldstein, J. Appl. Phys. <u>43</u> , 1642 (1971).
5.	T. Masumoto and R. Maddin, Acta Met. 19, 725 (1971).
6.	H.J. Leamy, H.S. Chen and T.T. Wang, Met. Trans. 3, 699 (1972).
7.	R. Maddin and T. Masumoto, Mat. Sci. Eng. 9, 153 (1972).
8.	H.S. Chen, <u>Scripta Met.</u> 7, 931 (1973).
9.	C.A. Pampillo and H.S. Chen, Mat. Sci. Eng. 13, 181 (1974).
10.	H.S. Chen and D.E. Polk, J. Non-Cryst. Solids 15, 174 (1974).
11.	C.A. Pampillo, J. Mat. Sci. 10, 1194 (1975).
12.	CP. P. Chou and F. Spaepen, Acta Met. 23, 609 (1975).
13.	H.S. Chen and D. Turnbull, Acta Met. 17, 1021 (1969).
14.	D.E. Polk and D. Turnbull, <u>Acta Met.</u> 20, 493 (1972).
15.	H.S. Chen and D. Turnbull, J. Chem. Phys. <u>48</u> , 2560 (1968).
16.	J. Logan and M.F. Ashby, Acta Met. 22, 1047 (1974).
17.	C.A. Pampillo and D.E. Polk, Acta Met. 22, 741 (1974).
18.	C.A. Pampillo and A.C. Reimschuessel, J. Mat. Sci. 9, 718 (1974).
19.	F. Spaepen, Acta Met. 23, 615 (1975).
20.	A.S. Argon and M. Salama, "Proc. 2nd Int. Conf. on Rapidly Quenched Metals", N.J. Grant and B.C. Giessen (eds.), Section II, <u>Mat. Sci. Eng.</u> 23, 219 (1976).
21.	F. Spaepen and D. Turnbull, Scripta Met. 8, 563 (1974).
22.	T. Masumoto and R. Maddin, Mat. Sci. Eng. 19, 1 (1975).

24. S. Takayama and R. Maddin, Acta Met. 23, 943 (1975).

- 25. M.H. Cohen and D. Turnbull, J. Chem. Phys. 31, 1164 (1959).
- 26. D. Turnbull and M. H. Cohen, J. Chem. Phys. 34, 120 (1961).
- 27. D. Turnbull and M. H. Cohen, J. Chem. Phys. 52, 3038 (1970).
- J. D. Eshelby, <u>Solid State Physics</u>, F. Seitz and D. Turnbull (eds.), <u>3</u>, 115 (1956).
- 29. For a description of the original mechanical model, see: D. Turnbull and R.L. Cormia, J. Appl. Phys. 31, 674 (1960).
- 30. H.S. Chen, J.T. Krause and E. Coleman, <u>J. Non-Cryst. Solids</u> <u>18</u>, 157 (1975).
- 31. H.S. Chen, Scripta Met. 9, 411 (1975).

1

- 1

a the second for

- 32. H.S. Chen, J.T. Krause and E. Coleman, <u>Scripta Met.</u> 9, 787 (1975).
- L.A. Davis, R. Ray, C.-P. Chou, and R.C. O'Handley, <u>Scripta</u> <u>Met.</u> 10, 541 (1976).

Defense Documentation Center Cameron Station Alexandria, Virginia 22314 Office of Naval Research Department of the Navy

(12)

(3)

(6)

Atta: Code 471 Code 105 Code 470

Director Office of Naval Research Branch Office Branch Office 495 Summer Street Boston, Massachusetts 02210

Director Office of Naval Research Branch Office 536 South Clark Street Chicago, Illinois 60605

Office of Naval Research San Francisco Area Office 760 Market Street, Room 447 San Francisco, California 94102

Naval Research Laboratory Washington, D.C. 20390 Attn: Code 6000 Code 6100 Code 6300 Code 6400 Code 2627

1

2

11

. .

1 . 1

Attn: Mr. F.S. Williams Naval Air Development Center Code 302 Warminster, Pennsylvania 18974

Naval Air Propulsion Test Center Trenton, New Jersey 08628 Attn: Library

Naval Weapons Laboratory Dahlgren, Virginia 22448 Attn: Research Division

Naval Construction Batallion Civil Engineering Laboratory Port Hueneme, California 93043 Attn: Materials Division

Naval Electronics Laboratory Center San Diego, California 92152 Attn: Electronic Materials Sciences Div.

Naval Missile Center Materials Consultant Code 3312-1 Point Mugu, California 93041

Commanding Officer Naval Ordnance Laboratory White Oak Silver Spring, Maryland 20910 Attn: Library

Naval Ship R. and D. Center Materials Department Annapolis, Maryland 21402

Naval Undersea Center San Diego, California 92132 Attn: Library

Naval Underwater System Center Newport, Rhode Island 02840 Attn: Library

Naval Weapons Center China Lake, California 93555 Attn: Library

Naval Postgraduate School Monterey, California 93940 Attn: Materials Sciences Dept.

Naval Air Systems Command Washington, D.C. 20360 Attn: Code 52031 Code 52032 Code 320

Naval Sea System Command Washington, D. C. 20362 Attn: Code 035

Naval Facilities Engineering Command Alexandria, Virginia 22331 Attn: Code 03

Scientific Advisor Commandant of the Marine Corps Washington, D.C. 20380 Attn: Code AX

Naval Ship Engineering Center Department of the Navy Washington, D.C. 20360 Attn: Director, Materials Sciences

Army Research Office Box CM, Duke Station Durham, North Carolina 27706 Attn: Metallurgy and Ceramics Div.

Army Materials and Mechanics Research Center Watertown, Massachusetts 02172 Attn: Res. Programs Office (AMXMR-P)

DISTRIBUTION LIST Contract N00017-67-A-0298-0036

Commanding General Department of the Army Frankford Arsenal Philadelphia, Pennsylvania 19137 Attn: ORDBA-1320

Office of Scientific Research Department of the Air Force Washington, D.C. 20333 Attn: Solid State Div. (SRPS)

Aerospace Research Labs Wright-Patterson AFB Building 450 Dayton, Ohio 45433

Air Force Materials Lab (LA) Wright-Patterson AFB Dayton, Ohio 45433

NASA Headquarters Washington, D.C. 20546 Attn: Code RRM

NASA Lewis Research Center 21000 Brookpark Road Cleveland, Ohio 44135 Attn: Library

National Bureau of Standards Washington, D.C. 20234

Attn: Metallurgy Division Inorganic Materials Division Atomic Energy Commission Washington, D.C. 20545 Attn: Metals and Materials Branch

Defense Metals and Ceramics Information Center Battelle Memorial Institute 505 King Avenue Columbus, Ohio 43201

Director Ordnance Research Laboratory P.O. Box 30 State College, Pennsylvania 16801

Director Applied Physics Lab. University of Washington 1013 Northeast Fortieth Street Seattle, Washington 98105

Metals and Ceramics Division Oak Ridge National Laboratory P.O. Box X Oak Ridge, Tennessee 37830

Los Alamos Scientific Lab. P.O. Box 1663 Los Alamos, New Mexico 87544 Attn: Report Librarian

Argonne National Laboratory Metallurgy Division P.O. Box 229 Lemont, Illinois 60439

Brookhaven National Laboratory Technical Information Division Upton, Long Island New York 11973 Attn: Research Library

Library Building 50, Room 134 Lawrence Radiation Laboratory Berkeley, California

Professor G.S. Ansell Rensselaer Polytechnic Institute Dept. of Metallurgical Engineering Troy, New York 12181

Professor H. K. Birnbaum University of Illinois Department of Metallurty Urbana, Illinois 61801

Dr. E.M. Breinan United Aircraft Corporation United Aircraft Research Lab. East Hartford, Connecticut 06108

Professor H. D. Brody University of Pittsburgh School of Engineering Pittsburgh, Pennsylvania 15213

Professor J. B. Cohen Northwestern University Dept. of Material Science Evanston, Illinois 60201

Professor M. Cohen Massachusetts Institute of Technology Department of Metallurgy Gambridge, Massachusetts 02139

Professor B. C. Giessen Northeastern University Department of Chemistry Boston, Massachusetts 02115

Dr. G.T. Hahn Battelle Memorial Institute Department of Metallurgy 515 King Avenue Columbus, Ohio 43201

Professor R. W. Heckel Carnegie-Mellon University Schenley Part Pittsburgh, Pennsylvania 15213

Dr. David G. Howden Battelle Memorial Institute Columbus Laboratories 505 King Avenue Columbus, Ohio 43201

Professor C. E. Jackson Ohio State University Dept. of Welding Engineering 190 West 19th Avenue Columbus, Ohio 43210

Professor G. Judd Rensselaer Polytechnic Institute Dept. of Materials Engineering Troy, New York 12181

Dr. C.S. Kortovich TRW, Inc. 23555 Euclid Avenue Cleveland, Ohio 44117

Professor D. A. Koss Michigan Technological University College of Engineering Houghton, Michigan 49931

Professor A. Lawley Drexel University Dept. of Metallurgical Engineering Philadelphia, Pennsylvania 19104

Dr. H. Margolin Polytechnic Institute of New York 333 Jay Street Brooklyn, New York 11201

Professor K. Masabuchi Massachusetts Institute of Technology Department of Ocean Engineering Cambridge, Massachusetts 02139

Dr. G. H. Meier University of Pittsburgh Dept. of Metallurgical and Materials Engineering Pittsburgh, Pennsylvania 15213

Professor J. W. Morris, Jr. University of California College of Engineering Berkeley, California 94720

Professor K. Ono University of California Materials Department Los Angeles, California 90024

Professor W. F. Savage Repsselaer Polytechnic Institute Rensselaer Polytechnic School of Engineering Troy, New York 12181

Dr. C. Shaw Rockwell International Corp. P.O. Box 1085 1049 Camino Dos Rios Thousand Oaks, California 91360

Dr. E. R. Thompson United Aircraft Research Lab. 400 Main Street East Hartford, Connecticut 06108

Professor David Turnbull Harvard University Division of Engineering and Applied Physics Cambridge, Massachusetts 02139

Dr. F. W. Wang Naval Ordnance Laboratory Physics Laboratory White Oak Silver Spring, Maryland 20910

Professor H. G. F. Wilsdorf University of Virginia Department of Materials Science Charlottesville, Virginia 22905

Dr. M. A. Wright University of Tennessee Space Institute Dept. of Metallurgical Engineering Tullahoma, Tennessee 37388

Professor O. D. Sherby Stanford University Materials Sciences Dept. Stanford, California 94300

Professor J. Shyne Stanford University Materials Sciences Department Stanford, California 94300

Dr. W. A. Spitzig U.S. Steel Corporation Research Laboratory Monroeville, Pennsylvania 15146

Dr. E.A. Starke, Jr. Georgia Institute of Technology School of Chemical Engineering Atlanta, Georgia 30332

Professor N.S. Stoloff Rensselaer Polytechnic Institute School of Engineering Troy, New York 12181

Dr. J.C. Williams Rockwell International Science Center P.O. Box 1085 Thousand Oaks, California 91360