

# **Dynamic Transformation of Austenite to Ferrite above the $A_{e3}$ Temperature**

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## Abstract

The dynamic transformation (DT) of austenite to ferrite above the  $A_{e3}$  temperature was studied in a plain C-Mn and a Nb microalloyed steel. Compression and torsion tests were carried out over the temperature range 880 °C to 1350 °C applied at a strain rate of  $1 \text{ s}^{-1}$ . The deformation temperatures were all above the ortho and para-equilibrium  $A_{e3}$  transformation temperatures. These experiments were performed in a 95%Ar - 5%H<sub>2</sub> atmosphere (on the MTS machines) and in a vacuum (Gleeble 3800 system), and all the samples were water quenched immediately after straining. In this work, the optical and electron microscopy images provided evidence of the presence of dynamically formed ferrite plates. These plates have near identical orientations and coalesce into polygonal grains upon continued straining.

The occurrence of dynamic transformation is explained here by introducing the concept of transformation softening. The net softening during transformation is considered as the driving force for the phase change. This is given by the difference between the flow stress of the work hardened austenite at the critical strain and the yield stress of the fresh Widmanstätten ferrite that takes its place. The latter is estimated from hardness data. The obstacle to the transformation consists of three parts: i) the Gibbs energy difference between austenite and Widmanstätten ferrite, ii) the shear accommodation work that is imposed on the neighborhood by the transformation, and iii) the dilatation work associated with making room for the lower density ferrite. This approach was validated using test results on a plain C-Mn steel compressed at temperatures well above the  $A_{e3}$ . It is shown in this way that DT ferrite can be formed at any temperature within the austenite phase field. These observations are consistent with the calculated Gibbs energy difference between the austenite and ferrite, which displays a peak approximately midway between the  $A_{e3}$  and the delta ferrite formation temperature.

The present thermodynamic analysis can also be used to determine the effect of alloying element addition. In order to assess the effects of Mn and Si addition on the transformation, the high temperature flow stress data obtained by Wray (1984) by means of tensile tests on

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four Fe-Mn-Si alloys were employed to identify how these alloying additions affect the driving and opposing forces. The present model shows that the addition of Mn increases both the driving force for transformation as well as the energy barrier to ferrite formation. Conversely, while Si addition increases the driving force, it simultaneously decreases the free energy obstacle to the transformation. This then increases the temperature range over which DT can take place.

Torsion simulations of strip rolling were performed on a C-Mn and a Nb microalloyed steel to determine the effect of DT on the rolling load. Pass strains of 0.4 were applied at a strain rate of  $1 \text{ s}^{-1}$  with interpass times in the range 0.5s to 5s. For additional realism, simulations were carried out under both isothermal and continuous cooling conditions. It was observed that the flow stress levels and MFS's decrease significantly during the simulations. This softening is associated with the formation of ferrite. Optical microscopy revealed that the volume fraction of DT ferrite increased continuously from pass to pass. The fraction of DT ferrite formed and retained was significantly higher when short interpass times were employed. When long intervals were used, the ferrite tends to retransform back into austenite. These observations indicate that the dynamic transformation of ferrite occurs displacively while the retransformation back into austenite takes place by a diffusional mechanism. The nucleation and growth of the ferrite reduce the rolling load and modify the microstructure. Comparison of the behaviors of the C-Mn and Nb steels indicates that Nb addition retards both the forward as well as the reverse transformation.

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## Résumé

La transformation dynamique (TD) de l'austénite en ferrite au-dessus de la température  $A_{e3}$  a été étudiée dans un alliage microallié au Nb et dans un acier au C-Mn. Des essais de torsion et de compression ont été réalisés pour une gamme de températures allant de 880°C à 1350°C avec une vitesse de déformation de  $1 \text{ s}^{-1}$ . Les températures de déformation étaient toutes au-dessus de la température  $A_{e3}$ . Ces expériences ont été réalisées sous une atmosphère de composition 95%Ar - 5%H<sub>2</sub> (dans une machine MTS) et sous vide (dans une machine Gleeble 3800), et l'ensemble des éprouvettes ont été trempées à l'eau immédiatement après la fin de la déformation. Dans ce travail, les observations au microscope optique et électronique fournissent la preuve de la présence de plaques de ferrite formées dynamiquement. Ces plaques ont des orientations presque identiques et se transforment en grains polygonaux pendant la déformation.

La présence de la transformation dynamique est expliquée en introduisant le concept de la transformation par adoucissement. L'adoucissement présent lors de la transformation est considéré comme la force motrice du changement de phase. Elle provient de la différence entre la contrainte d'écoulement de l'austénite écrouie à la déformation critique et la limite d'élasticité de la nouvelle ferrite de Widmanstätten qui la remplace. La limite d'élasticité est estimée par des essais de dureté. Trois phénomènes s'opposent à cette transformation : i) la différence d'énergie de Gibbs entre l'austénite et la ferrite de Widmanstätten, ii) le cisaillement imposé au voisinage de la zone en transformation, iii) la dilatation due à la plus faible densité de la ferrite. Cette approche a été validée en utilisant des résultats de tests de compression sur l'acier C-Mn à des températures supérieures à la température  $A_{e3}$ . Ces résultats montrent que la transformation dynamique de la ferrite dans la phase austénitique peut avoir lieu à n'importe quelle température. Ces observations sont en accord avec les calculs de différence d'énergie de

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Gibbs entre l'austénite et la ferrite, qui présente un maximum à mi-chemin entre la température  $Ae_3$  et la température de formation de la ferrite delta.

Cette analyse thermodynamique peut également être utilisée pour déterminer les effets des éléments d'addition. Wray (1984) a évalué les contraintes d'écoulement à haute température par l'intermédiaire d'essais de traction sur quatre alliages Fe-Mn-Si. Ces résultats sont utilisés pour identifier comment le Mn et le Si impactent les forces motrices et les forces d'opposition. Le modèle montre que l'addition de Mn augmente à la fois la force motrice de la transformation dynamique ainsi que l'énergie d'activation de la formation de la ferrite. Inversement, si l'addition de Si augmente également la force motrice, elle diminue simultanément l'énergie d'activation de la transformation dynamique. Ce qui a pour conséquence d'augmenter la gamme de températures dans laquelle la transformation dynamique peut avoir lieu.

Des simulations de laminage par l'intermédiaire d'essais de torsion ont été réalisées sur des aciers microalliés au Nb et au C-Mn afin de déterminer les effets de la transformation dynamique sur la force de laminage. Des passes de déformation de 0,4 ont été appliquées avec une vitesse de déformation de  $1 \text{ s}^{-1}$  et entrecoupées par des interpasses sans déformation ayant un temps allant de 0,5 à 5 s. Pour plus de réalisme, des simulations ont été réalisées dans des conditions de refroidissements continus. Il a été observé que les niveaux des contraintes d'écoulement ainsi que la contrainte d'écoulement moyenne diminuent significativement pendant les simulations. Cet adoucissement est associé à la formation de ferrite. Les observations au microscope optique ont révélées que la fraction volumique de ferrite transformée dynamiquement augmente de manière continue à chaque passe. La fraction de ferrite transformée dynamiquement était significativement plus importante lors de temps d'interpasses courts. Lors de longs intervalles de temps, la ferrite a tendance à se transformer à nouveau en austénite. Ces observations indiquent que la transformation dynamique de la ferrite se réalise de manière 'displacive', tandis que la transformation de la ferrite en austénite se réalise selon un mécanisme diffusif. La germination et la croissance de la ferrite réduisent la force de laminage et modifient la microstructure. Les comparaisons entre les aciers au C-Mn et au Nb indiquent que l'ajout de Nb retarde aussi bien les deux transformations.

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## Contributions of the Authors

This dissertation was prepared according to the guidelines of the Faculty of Graduate and Postdoctoral Studies of McGill University for a manuscript-based thesis. The current document is comprised principally of 6 published manuscripts:

1. **C. Aranas Jr.**, T. Nguyen-Minh, R. Grewal and J.J. Jonas, "Flow softening-based formation of Widmanstätten ferrite in a 0.06%C steel deformed above the  $A_{e3}$ ," ISIJ International, 55 (2015) 300-307. (Chapter 3)
2. **C. Aranas Jr.** and J.J. Jonas, "Effect of Mn and Si on the dynamic transformation of austenite above the  $A_{e3}$  temperature," Acta Materialia, 82 (2015) 1-10. (Chapter 4)
3. **C. Aranas Jr.**, R. Grewal, K. Chadha, D. Shahriari, M. Jahazi and J.J. Jonas, "Formation of Widmanstätten Ferrite in a C-Mn Steel at Temperatures High in the Austenite Phase Field," Proceedings of the International Conference on Solid-Solid Phase Transformations in Inorganic Materials (PTM 2015), Whistler, BC, (2015) 613-620. (Chapter 5)
4. **C. Aranas Jr.** and J.J. Jonas, "Dynamic transformation of austenite to ferrite during the torsion simulation of strip rolling," Proceedings of the Materials Science and Technology (MS&T 2014) Conference, Pittsburgh, USA, (2014) 461-468. (Chapter 6)
5. **C. Aranas Jr.**, T. Wang and J.J. Jonas, "Effect of Interpass Time on the Dynamic Transformation of a plain C-Mn and a Nb Microalloyed Steel," ISIJ International, 55 (2015) 647-654. (Chapter 6)
6. **C. Aranas Jr.**, S. Rodrigues, R. Grewal and J.J. Jonas, "Ferrite Formation above the  $A_{e3}$  Temperature during the Torsion Simulation of Strip Rolling," ISIJ International, 55 (2015) 2426-2434. (Chapter 7)

All the manuscripts were co-authored with Prof. John J. Jonas, who actively supervised my research. With regard to Reference 1, Dr. Tuan Nguyen-Minh of Ghent University provided me with some calculations regarding the range of Schmid factors applicable to the formation of

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Widmanstätten ferrite. Ms. Rupanjit Grewal, my colleague, was involved in stimulating discussions and she assisted me in carrying out the MTS compression tests. The Gleeble compression results in Reference 3 were performed at the École de Technologie Supérieure, in collaboration with Prof. Mohammad Jahazi together with his postdoctoral fellow, Dr. Davood Shahriari and PhD student, Mr. Kanwal Chadha. They provided me with expert advice and assistance in performing my experimental work. This article is the preliminary result of Ms. Grewal's MEng project, which I am actively co-supervising. The EBSD micrographs presented in Reference 5 were produced together with Ms. Tong Wang, who taught me how to optimize the various parameters in the electron microscope. With regard to Reference 6, Mr. Samuel Rodrigues and Ms. Rupanjit Grewal are both working on this topic. They were heavily involved in the discussions and helped me in quenching my torsion samples. Nevertheless, all the data presented in the above papers were obtained by the present author.

During my work, I also co-authored the 9 articles listed below, whose abstracts are provided in Appendix B:

1. C. Ghosh, **C. Aranas** and J.J. Jonas, "Dynamic Transformation of Deformed Austenite at Temperatures above the  $A_{e3}$ ," Progress in Materials Science, (2015) ongoing revision.
2. R. Grewal, **C. Aranas Jr.**, K. Chadha, D. Shahriari, M. Jahazi and J.J. Jonas, "Formation of Widmanstätten Ferrite at very High Temperatures in the austenite phase field" Acta Materialia, (2015) submitted.
3. J.J. Jonas and **C. Aranas Jr.**, "Problems with use of the logarithmic equivalent strain in high pressure torsion," IOP Conf. Series: Materials Science and Engineering 63 (2014) 012048.
4. J.J. Jonas, **C. Aranas Jr.**, V.V. Basabe and C. Ghosh, "Dynamic Transformation during the Torsion Simulation of Strip Rolling," Materials Science Forum, 783-786 (2014), 39-44.
5. J.J. Jonas, **C. Aranas Jr.** and C. Ghosh, "Dynamic Transformation of Austenite at Temperatures Above the  $A_{e3}$ ," Proceedings of the International Conference on Solid-Solid Phase Transformations in Inorganic Materials (PTM 2015), Whistler, BC, (2015) 573-580.

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6. J.J. Jonas, **C. Aranas Jr.** and S.F. Rodrigues, "Dynamic Transformation During the Plate and Strip Rolling," Proceedings of the 4<sup>th</sup> International Conference on Thermo-Mechanical Simulation and Processing of Steels (SimPro'16), Ranchi, India, (2015) in press.
  7. J.J. Jonas and **C. Aranas Jr.**, "Dynamic transformation of austenite to ferrite during the torsion simulation of strip rolling," Proceedings of the 6th Baosteel Biennial Academic Conference and 10th CSM Steel Congress, Shanghai, China, (2015) in press.
  8. J.J. Jonas, **C. Aranas Jr.**, S.F. Rodrigues and I.-H. Jung, "Dynamic Transformation during Plate and Strip Rolling," Proceedings of the International Conference on Processing and Manufacturing of Advanced Materials (THERMEC 2016), Graz, Austria, (2016) in press.
  9. J.J. Jonas, V.V. Basabe, C. Ghosh and **C. Aranas Jr.**, "Dynamic Transformation during the Simulation of Strip Rolling," Proceedings of the 5th Baosteel Biennial Academic Conference, Shanghai, China, (2013) C1-C6.

All of the above articles are related to my thesis topic. Reference 1 is an invited review paper in Progress in Materials Science. In this work, I wrote 5 chapters out of 12 (with a total of 247 manuscript pages). With regard to Reference 2, as mentioned above, I am actively co-supervising my colleague Ms. Rupanjit Grewal for her MEng project. I assisted her in performing the experiments and in doing calculations on her results. With regard to Reference 3, I made the figures and derived the equation to show the problems associated with the logarithmic equivalent strain. I performed the torsion experiments regarding the simulation of strip rolling described in References 4-9. In these papers, I was also involved in the MFS and critical strain calculations.

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## Table of Contents

|                                              |          |
|----------------------------------------------|----------|
| Abstract                                     | i        |
| Résumé                                       | iii      |
| Contribution of Authors                      | v        |
| Acknowledgements                             | viii     |
| Table of Contents                            | x        |
| List of Figures                              | xv       |
| List of Tables                               | xxv      |
| <b>Chapter 1: Introduction</b>               | <b>1</b> |
| Introduction                                 | 1        |
| References                                   | 4        |
| <b>Chapter 2: Literature Review</b>          | <b>6</b> |
| 2.1 Iron-Carbon Phase Diagram                | 6        |
| 2.2 Development of Microstructures in Steels | 8        |
| 2.2.1 Diffusional Mechanism                  | 8        |
| 2.2.1.1 Ferrite                              | 9        |
| 2.2.1.2 Pearlite                             | 9        |
| 2.2.2 Displacive Mechanism                   | 10       |
| 2.2.2.1 Martensite                           | 10       |
| 2.2.2.2 Bainite                              | 12       |
| 2.2.2.3 Widmanstätten Ferrite                | 13       |
| 2.3 Hot Rolling Process                      | 14       |
| 2.3.1 Mean Flow Stress                       | 15       |
| 2.3.2 Basic Concepts of Steel Rolling        | 16       |
| 2.3.2.1 Recovery                             | 17       |
| 2.3.2.2 Static Recrystallization             | 17       |

---

|                                                                                                                                    |    |
|------------------------------------------------------------------------------------------------------------------------------------|----|
| 2.3.2.3 Dynamic Recrystallization                                                                                                  | 17 |
| 2.3.2.4 Metadynamic Recrystallization                                                                                              | 18 |
| 2.4 Dynamic Transformation                                                                                                         | 19 |
| 2.4.1 Real Time Evidence of DT                                                                                                     | 20 |
| 2.4.2 Microstructural Evidence of DT                                                                                               | 22 |
| 2.4.3 Formation Mechanism of Widmanstätten Ferrite                                                                                 | 25 |
| 2.4.4 Thermodynamics of Dynamic Transformation                                                                                     | 27 |
| 2.4.4.1 Stored Energy Model                                                                                                        | 27 |
| 2.4.4.2 Stress Activation Model                                                                                                    | 30 |
| 2.4.5 Early Strip Rolling Simulations                                                                                              | 34 |
| 2.5 References                                                                                                                     | 36 |
| <b>Chapter 3: Flow Softening-based Formation of Widmanstätten Ferrite in a 0.06%C Steel Deformed above the <math>A_{e3}</math></b> | 39 |
| 3.1 Introduction                                                                                                                   | 41 |
| 3.2 Experimental Procedure                                                                                                         | 42 |
| 3.2.1 Compression Testing                                                                                                          | 42 |
| 3.2.2 Metallography                                                                                                                | 45 |
| 3.3 Results                                                                                                                        | 45 |
| 3.3.1 Stress-Strain Curves                                                                                                         | 45 |
| 3.3.2 Microstructure                                                                                                               | 49 |
| 3.4 Discussion                                                                                                                     | 52 |
| 3.4.1 Mechanical Driving Force                                                                                                     | 52 |
| 3.4.2 Free Energy Obstacles to Transformation                                                                                      | 55 |
| 3.4.3 Driving Force versus Free Energy Barriers                                                                                    | 58 |
| 3.4.4 Limitations of the Present Model                                                                                             | 59 |
| 3.5 Conclusions                                                                                                                    | 61 |
| 3.6 References                                                                                                                     | 61 |
| <b>Chapter 4: Effect of Mn and Si on the Dynamic Transformation of Austenite above the <math>A_{e3}</math> Temperature</b>         | 63 |
| 4.1 Introduction                                                                                                                   | 65 |

---

|                                                                                                                        |     |
|------------------------------------------------------------------------------------------------------------------------|-----|
| 4.2 Influence of Mn and Si on the Driving Force for Transformation                                                     | 66  |
| 4.3 Influence of Mn and Si on the Obstacles to Ferrite Formation above the $A_{e3}$                                    | 76  |
| 4.4 Shear and Dilatation Work                                                                                          | 80  |
| 4.5 Discussion                                                                                                         | 81  |
| 4.5.1 Driving Forces versus Energy Barriers                                                                            | 81  |
| 4.5.2 Mechanical Activation and the Critical Strain                                                                    | 88  |
| 4.5.3 Summary of the Model                                                                                             | 89  |
| 4.6 Conclusions                                                                                                        | 90  |
| 4.7 References                                                                                                         | 91  |
| <b>Chapter 5: Formation of Widmanstätten Ferrite in a C-Mn Steel at Temperatures High in the Austenite Phase Field</b> | 93  |
| 5.1 Introduction                                                                                                       | 95  |
| 5.2 Experimental Procedure                                                                                             | 95  |
| 5.3 Results                                                                                                            | 97  |
| 5.4 Discussion                                                                                                         | 99  |
| 5.4.1 Critical Strains                                                                                                 | 99  |
| 5.4.2 Formation of Widmanstätten Ferrite                                                                               | 102 |
| 5.5 Conclusions                                                                                                        | 104 |
| 5.6 References                                                                                                         | 104 |
| <b>Chapter 6: Effect of Interpass Time on the Dynamic Transformation of a Plain C-Mn and a Nb Microalloyed Steel</b>   | 106 |
| 6.1 Strip Rolling Simulation without a roughing pass                                                                   | 107 |
| 6.1.1 Introduction                                                                                                     | 108 |
| 6.1.2 Experimental Procedure                                                                                           | 108 |
| 6.1.3 Results                                                                                                          | 110 |
| 6.1.4 Discussion                                                                                                       | 116 |
| 6.1.4.1 Critical Strains                                                                                               | 116 |
| 6.1.4.2 Mean Flow Stress                                                                                               | 117 |
| 6.1.5 Conclusions                                                                                                      | 120 |
| 6.1.6 References                                                                                                       | 120 |

---

|                                                                                                                              |     |
|------------------------------------------------------------------------------------------------------------------------------|-----|
| 6.2 Strip Rolling Simulation with a roughing pass                                                                            | 122 |
| 6.2.1 Introduction                                                                                                           | 123 |
| 6.2.2 Experimental Procedure                                                                                                 | 124 |
| 6.2.2.1 Torsion Testing                                                                                                      | 124 |
| 6.2.2.2 Metallography                                                                                                        | 126 |
| 6.2.3 Results                                                                                                                | 126 |
| 6.2.3.1 Strip Rolling Simulations                                                                                            | 126 |
| 6.2.3.2 Isothermal Rolling Simulations                                                                                       | 132 |
| 6.2.4 Discussion                                                                                                             | 135 |
| 6.2.4.1 Mean Flow Stress                                                                                                     | 135 |
| 6.2.4.2 Effect of Interpass Time                                                                                             | 137 |
| 6.2.4.3 Retransformation of DT Ferrite                                                                                       | 139 |
| 6.2.4.4 Industrial Implications                                                                                              | 144 |
| 6.2.5 Conclusions                                                                                                            | 144 |
| 6.2.6 References                                                                                                             | 145 |
| <b>Chapter 7: Ferrite Formation above the <math>A_{e3}</math> Temperature during the Torsion Simulation of Strip Rolling</b> | 147 |
| 7.1 Introduction                                                                                                             | 149 |
| 7.2 Experimental Procedure                                                                                                   | 150 |
| 7.2.1 Torsion Testing                                                                                                        | 150 |
| 7.2.2 Metallography                                                                                                          | 152 |
| 7.3 Results                                                                                                                  | 152 |
| 7.3.1 Torsion Simulation of Strip Rolling                                                                                    | 152 |
| 7.4 Discussion                                                                                                               | 156 |
| 7.4.1 Mean Flow Stress                                                                                                       | 156 |
| 7.4.2 Critical Strains                                                                                                       | 158 |
| 7.4.3 Volume Fraction of Ferrite Formed per Pass                                                                             | 162 |
| 7.4.4 Effect of Nb addition on the Behavior                                                                                  | 172 |
| 7.5 Conclusions                                                                                                              | 174 |

---

|                                                       |     |
|-------------------------------------------------------|-----|
| 7.6 References                                        | 175 |
| <b>Chapter 8: Conclusions</b>                         | 178 |
| <b>Chapter 9: Contributions to Original Knowledge</b> | 182 |
| <b>Appendix A</b>                                     | 185 |
| <b>Appendix B</b>                                     | 211 |

---

## List of Figures

|                                                                                                                                                                                                                                                                                  |    |
|----------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------|----|
| <b>Fig. 2.1.</b> Iron-carbon phase diagram showing the transformation temperatures and corresponding stable phases for each temperature [2.1].                                                                                                                                   | 7  |
| <b>Fig. 2.2.</b> Nucleation of allotriomorphic and idiomorphic ferrite [2.5].                                                                                                                                                                                                    | 9  |
| <b>Fig. 2.3.</b> Representation of pearlite formation showing alternating layers of lamellar ferrite and cementite plates [2.6].                                                                                                                                                 | 10 |
| <b>Fig. 2.4.</b> Microstructure of plate martensite in an Fe–1.86 wt.% C alloy [2.7].                                                                                                                                                                                            | 11 |
| <b>Fig. 2.5.</b> Diagram illustrating the growth of upper and lower bainite based on carbon diffusion [2.8].                                                                                                                                                                     | 12 |
| <b>Fig. 2.6.</b> Nucleation of primary and secondary Widmanstätten ferrite [2.10].                                                                                                                                                                                               | 13 |
| <b>Fig. 2.7.</b> A strip rolling operation with 7 finishing stands [2.11].                                                                                                                                                                                                       | 14 |
| <b>Fig. 2.8.</b> The mean flow stress behavior of a microalloyed steel during simulated Steckel mill rolling down to temperatures below the $A_{r1}$ [2.13].                                                                                                                     | 16 |
| <b>Fig. 2.9.</b> Schematic diagram of Yada’s in-situ X-ray experimental setup [2.29]                                                                                                                                                                                             | 20 |
| <b>Fig. 2.10.</b> Results of in-situ X-ray diffraction experiments (a) before the experiment, (b) at a temperature above the $A_{e3}$ , and (c) during deformation above the $A_{e3}$ [2.29].                                                                                    | 21 |
| <b>Fig. 2.11</b> Dilatation curves of 0.17%C steel specimens held isothermally after deformation at various temperatures above the $A_{e3}$ . [2.32]                                                                                                                             | 22 |
| <b>Fig. 2.12.</b> Microstructures of 0.17% steel specimens (a) quenched immediately after deformation at 850 °C (i.e. $A_{e3} + 15$ °C) and (b) subjected to 1200 s of holding following deformation and then quenched. [2.32]                                                   | 22 |
| <b>Fig. 2.13.</b> Optical micrographs of DT ferrite formed in steel samples deformed and quenched at (a) 887 °C ( $A_{e3} + 10$ °C) to a strain of 1.5 at the rate of $0.04\text{ s}^{-1}$ and (b) 917 °C ( $A_{e3} + 40$ °C) to a strain of 2.5 at $0.4\text{ s}^{-1}$ . [2.34] | 23 |
| <b>Fig. 2.14</b> Optical micrographs of martensite formed in steel samples deformed and quenched at (a) 887 °C ( $A_{e3} + 10$ °C) to a strain of 1.5 at the rate of $0.04\text{ s}^{-1}$ and (b) 917 °C ( $A_{e3} + 40$ °C) to a strain of 2.5 at $0.4\text{ s}^{-1}$ . [2.34]  | 23 |

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**Fig. 2.15.** EBSD maps of a Fe-6Ni-0.1C steel deformed to a strain of 0.96 at 750 °C ( $A_{e3} + 22^{\circ}\text{C}$ ) and  $10\text{ s}^{-1}$  showing (a) image quality and (b) inverse pole figure maps. 'F' and 'M' indicate ferrite and martensite, respectively. [2.36] 24

**Fig. 2.16.** Transmission electron micrographs of a 0.79C-0.65Mn-0.24Si steel deformed to a strain of 4 at 763 °C ( $A_{e3} + 30^{\circ}\text{C}$ ) and  $4\text{ s}^{-1}$ . The sample was quenched immediately after deformation. The TEM images show (a) the presence of ferrite and martensite, (b) the morphology of ferrite plates and (c) cementite precipitates. [2.42] 24

**Fig. 2.17.** EBSD maps of Widmanstätten ferrite plates taken from a 0.06C-0.01Si-0.3Mn-0.042Al steel sample deformed to a strain of 0.2 at 887 °C and  $0.4\text{ s}^{-1}$ . (a,b) IPF and IQ maps of ferrite, (c) IPF map of martensite phase (black regions are ferrite), and (d) misorientation profile measured along the white line in (b). [2.43] 26

**Fig. 2.18.** (a) The application of a shear stress results in the formation of a parallelogram of Widmanstätten ferrite from the rectangle of prior austenite. (b) The geometry associated with the formation of self-accommodating plates. [2.48] 26

**Fig. 2.19.** The coalescence of Widmanstätten ferrite plates into polygonal grains during dynamic transformation. [49] 27

**Fig. 2.20.** Illustration of the changes in Gibbs energy responsible for dynamic transformation based on the stored energy model. Holding at temperature after deformation permits the re-transformation back to austenite. [2.48] 30

**Fig. 2.21.** Pole figures showing three possible configurations of habit plane normal (HPN), shear plane normal (SPN) and shear direction (SD) with respect to the applied stress. a) SPN perpendicular to applied stress, b) HPN and SD perpendicular to the applied stress and, c) all axes are inclined at  $54.7^{\circ}$  with respect to the applied stress. [2.48] 31

**Fig. 2.22.** Austenite flow stress at  $\epsilon = 0.10$  as a function of  $\Delta T$  (right-hand axis). The shear accommodation and dilatation work as a function of  $\Delta T$  (left-hand axis). [2.48] 32

**Fig. 2.23.** The shear accommodation and dilatation work are added to the chemical free energy difference between austenite and ferrite as part of the total energy barrier to transformation. [2.48] 33

**Figure 2.24:** Illustration of the changes in Gibbs energy responsible for dynamic transformation based on the mechanical activation model. Holding at temperature after deformation permits the re-transformation back into austenite. [2.48] 34

**Fig. 2.25:** Dependence of the MFS on inverse absolute temperature in a 0.1C-1.27Mn-0.2Si-0.11Nb-0.08V microalloyed steel. Here interpass times of 2s and 30s were employed together with pass strains of 0.5 applied at a strain rate of  $60\text{ s}^{-1}$ . [2.64] 35

---

**Fig. 2.26:** Critical strains for the initiation of dynamic transformation (DT) and dynamic recrystallization (DRX) in torsion simulation experiments on a 0.79C-0.65Mn-0.24Si steel. [2.61] 35

**Fig. 3.1.** Compression test schedules employed to determine a) the yield stress of ferrite below the  $Ae_1$  and b) the flow stress of work hardened austenite above the  $Ae_3$ . 43

**Fig. 3.2.** Optical micrograph showing the homogeneous nature of the deformation in a sample compressed to a strain of 0.7 at  $1s^{-1}$  and  $980^{\circ}C$ . 45

**Fig. 3.3.** Stress-strain curves of the 0.06wt%C-0.3wt%Mn-0.01wt%Si steel compressed to a strain of 0.7 at  $1s^{-1}$ . Stress drops of approximately 10% are observed after the stress peaks. The horizontal lines display the temperature of the sample (on the right hand axis) during deformation. 46

**Fig. 3.4.** a) Plot of strain hardening rate ( $\theta$ ) versus  $\sigma$  derived from the stress-strain curve of the compression test carried out at  $920^{\circ}C$  and  $1s^{-1}$ . Each inflection point corresponds to the initiation of a particular softening mechanism. b) The inflection points in a) are identified more clearly by the minima in the  $-(d\theta/d\sigma)$  versus  $\sigma$  curve. 47

**Fig. 3.5.** a) The critical strains for dynamic transformation (DT) and dynamic recrystallization (DRX) determined by the double differentiation method. b) The critical stresses associated with the critical strains of Figure 3.4a. Those of DRX are temperature dependent while those of DT show no dependence. 48

**Fig. 3.6.** Optical micrographs of the 0.06wt%C-0.3wt%Mn-0.01wt%Si steel deformed to a strain of 0.7 at a strain rate of  $1s^{-1}$  at a)  $980^{\circ}C$  and b)  $960^{\circ}C$ . The volume fraction of dynamically formed ferrite is higher at the lower temperature. 50

**Fig. 3.7.** EBSD images of 0.09%C steel strained to 0.5 at  $856^{\circ}C$  ( $20^{\circ}C$  above the  $Ae_3$ ). a) and b) secondary electron images of Widmanstätten ferrite plates, c) orientation image map of the ferrite phase,  $[001]$  is along the longitudinal axis, and d) orientation image map of ferrite after  $90^{\circ}$  rotation around  $[010]$ . [3.22] 51

**Fig. 3.8.** a) Dependence of the austenite flow stress on inverse absolute temperature. The net softening associated with DT is evaluated by subtracting the yield stress of Widmanstätten ferrite ( $YS-W\alpha$ ) from the flow stress of the austenite work hardened to the DT critical strain of 0.12. b) The driving force for DRX can be estimated from the difference between the flow stress of the work hardened austenite at the DRX critical strain of 0.20 and the yield stress of undeformed austenite grains ( $YS-\gamma$ ). 52

**Fig. 3.9.** Chemical free energies of polygonal (solid line) and Widmanstätten (broken line) ferrite as a function of the temperature interval  $\Delta T$  (experimental temperature –  $Ae_{3(p)}$ ). The free

energy barrier increases with  $\Delta T$  because of the increasing instability of ferrite at elevated temperatures. 55

**Figure 3.10.** The Schmid factors pertaining to transformation on the (0.506 0.452 0.735) habit plane and along the [-0.867 0.414 0.277] shear direction displayed in inverse pole figure (IPF) form. 56

**Fig. 3.11.** Austenite flow stress at  $\epsilon = 0.12$  determined in compression displayed as a function of  $\Delta T$  (right-hand axis). The shear accommodation work (SA) and dilatation work (DW) as a function of  $\Delta T$  (left-hand axis).  $m$ -values of 0.5 and 0.2 are employed to illustrate the effect of grain orientation on the energy barriers. 57

**Fig. 3.12.** a) Dependence of the driving force (DF) at  $\epsilon = 0.12$  on temperature interval and Schmid factor in the range  $0.2 \leq m \leq 0.5$  (dotted lines). The total barriers to transformation over the same temperature range (solid lines). The intersections (labeled with points) correspond to the maximum temperatures up to which dynamic transformation can take place in grains of particular  $m$ -values. b) At a lower strain (i.e.  $\epsilon = 0.05$ ), the absence of intersections indicates that transformation is not possible. 58

**Fig. 3.13.** Illustration of the effect of assuming that the driving force for the formation of Widmanstätten ferrite remains essentially independent of temperature. The driving force and obstacle curves for  $m = 0.5$  no longer intersect, while the intersection temperatures for  $m = 0.35$  and  $m = 0.2$  have increased when compared with those of Figure 12a. 60

**Fig. 4.1.** Tensile flow curve determined by Wray [4.8] on a 1.4Mn-0.25Si steel at 600 °C and a strain rate of  $2.3 \times 10^{-2} \text{ s}^{-1}$ . Samples were tested in argon after annealing for one hour at 1050 °C. The maximum load in this case was attained at  $\epsilon = 0.17$ . Nevertheless, discernible necking was delayed until  $\epsilon = 0.35$ , permitting determination of the true strain up to this point [4.8]. 67

**Fig. 4.2.** Widmanstätten microstructures formed under the following conditions: a) 0.06wt%C steel strained to  $\epsilon=0.25$  at 887 °C ( $A_{e3(p)} + 10^\circ\text{C}$ ); b) 0.21wt%C steel strained to  $\epsilon=0.25$  at 862 °C ( $A_{e3(p)} + 40^\circ\text{C}$ ); c) and d) 0.09wt%C steel strained to  $\epsilon=0.25$  at 856 °C ( $A_{e3(p)} + 20^\circ\text{C}$ ). 68

**Fig. 4.3.** Effect of temperature on the flow stress levels attained at increasing strains determined in tension on the 0.01Mn-0.24Si-Fe steel of Ref. 8 at a strain rate of  $2.3 \times 10^{-2} \text{ s}^{-1}$ . Linear relationships are assumed between the inverse absolute temperature and the flow stress of austenite ( $\gamma$ -FS) as well as the yield stress of polygonal ( $\alpha$ -YS) ferrite. 70

**Fig. 4.4.** Examples of Vickers hardness measurements performed on the 0.21wt%C steel (Table 2) previously strained to  $\epsilon = 2$  at a rate of  $4.5 \text{ s}^{-1}$ . Samples were deformed at 895 °C ( $A_{e3(p)} + 87^\circ\text{C}$ ) and then quenched. a) Widmanstätten colony displaying a hardness of 168 HV; b) polygonal ferrite (held for 800s after deformation) displaying a hardness of 142 HV. 71

**Fig. 4.5.** Dependences of the flow stress on inverse absolute temperature in the present four Fe-Mn-Si alloys. The behavior of the LL (low Mn-low Si) alloy of Fig. 4.5a is the basis for comparison to determine the effects of Mn addition (Fig. 4.5b), Si addition (Fig. 4.5c) and the combined addition of both Mn and Si (Fig. 4.5d). Linear dependences of the various flow stresses on inverse absolute temperature are again assumed here and the fitted lines all approach zero stress at the melting point. 72

**Fig. 4.6.** The driving forces in J/mol derived from the data of Fig. 4.5 by subtracting the Widmanstätten ferrite yield stresses from the austenite flow stresses at the critical strain of 0.1. The addition of Mn (Fig. 4.6a) increases the energy available by 28-47 J/mol and that of Si (Fig. 4.6b) by 8-12 J/mol. The simultaneous addition of both alloying elements (Fig. 4.6c) increases the driving force by 34-60 J/mol. 75

**Fig. 4.7.** a) The chemical free energy differences between austenite and polygonal ferrite ( $\Delta G_{\gamma-\alpha}$ ) in the four alloys presented as a function of  $\Delta T$  (experimental temperature – paraequilibrium  $Ae_3$ ). The free energy barrier increases with  $\Delta T$  because of the increasing instability of ferrite as the temperature is increased. b) The chemical free energy between austenite and Widmanstätten ferrite ( $\Delta G_{\gamma-W\alpha}$ ) in the LL steel. 77

**Fig. 4.8.** Gibbs free energy differences derived from Fig. 4.7 showing the effects of Mn addition (Fig. 4.8a), Si addition (Fig. 4.8b) and the combined addition both Mn and Si (Fig. 4.8c). Since Mn is an austenite stabilizer, it *increases* the Gibbs energy difference, while Si, as a ferrite former has the opposite effect. The addition of both alloying elements leads to a combination of the two effects. It should be noted that the total barriers are characterized by maxima as the chemical free energy differences go to zero at the delta iron transformation temperature. 78

**Fig. 4.9.** The flow stress of the 0.01Mn-0.24Si-Fe (LL) alloy at  $\epsilon = 0.1$  determined in tension displayed as a function of  $\Delta T$  (right-hand axis). The calculated shear accommodation (SA) and lattice dilatation (DW) work in J/mol plotted with respect to  $\Delta T$  (left-hand axis). The behaviors of grain orientations with  $m$ -values of 0.5 and 0.1 are illustrated to represent the barriers associated with high and low resolved stresses. 81

**Fig. 4.10.** The driving force (DF) at yield ( $\epsilon = 0.01$ ) is compared to the obstacle to transformation at temperatures above the paraequilibrium  $Ae_3$ . It can be seen that dynamic transformation is not possible under these conditions because the stress difference (driving force) is too low, even for grains that are favorably oriented ( $m = 0.5$ ). The addition of Mn (compare Fig. 4.10a and Fig. 4.10b) decreases the probability of transformation by increasing the gap between the obstacle and the driving force whereas Si addition (compare Fig. 4.10a and Fig. 4.10c) increases the probability by decreasing the gap. 83

**Fig. 4.11.** The driving force (DF) at the critical strain of 0.1 compared with the total energy barrier at temperatures above the paraequilibrium  $Ae_3$ . The points of intersection at given values of  $m$  indicate the maximum temperatures up to which dynamic transformation can take

---

place. The temperature interval above the  $A_{e3}$  is restricted by adding Mn (compare Fig. 4.11a and Fig. 4.11b) and augmented by adding Si (compare Fig. 4.11a and Fig. 4.11c). 86

**Fig. 4.12.** The critical stresses and strains associated with  $m$ -values of 0.3, 0.4 and 0.5 when the austenite of alloy HH is being deformed at 50 °C above the paraequilibrium  $A_{e3}$ . At higher temperatures (not shown here), the critical strains move to higher and higher values. 89

**Fig. 4.13.** The dynamic transformation model based on activation by transformation softening. Here allowance is also made for the higher free energy of the Widmanstätten ferrite over that of the polygonal ferrite values provided by the FactSage thermodynamic model. The shear accommodation and dilatation work are also shown, as is the re-transformation back to austenite that takes place by diffusional processes upon removal of the applied stress. 90

**Fig. 5.1.** Thermomechanical schedule of the compression tests. 96

**Fig. 5.2.** Stress-strain curves of the C-Mn steel compressed to a strain of 0.7 at  $1\text{ s}^{-1}$ . The tests were carried out at temperatures in the range from 1000 °C to 1350 °C. 97

**Fig. 5.3.** Optical micrographs of the 0.06C-0.3Mn-0.01Si steel deformed to a strain of 0.7 applied at a strain rate of  $1\text{ s}^{-1}$ . The samples were quenched immediately after straining at (a) 1050 °C, (b) 1150 °C and (c/d) 1200 °C. 98

**Fig. 5.4.** (a) Plot of strain hardening rate ( $\theta$ ) versus  $\sigma$  derived from the flow curve of the compression test carried out at 1100 °C and  $1\text{ s}^{-1}$ . The first and third inflection points in (a) are identified more clearly in (b) by the minima in the  $-(d\theta/d\sigma)$  versus  $\sigma$  curve. 99

**Fig. 5.5.** (a) Critical strains for dynamic transformation (DT) and dynamic recrystallization (DRX) calculated by means of the double differentiation method. (b) The critical stresses associated with the critical strains of Fig. 5.5a. 101

**Fig. 5.6.** Chemical free energy difference between austenite and ferrite ( $\Delta G_{\gamma-\alpha}$ ) as a function of  $\Delta T$  (experimental temperature – paraequilibrium  $A_{e3}$ ). 102

**Fig. 5.7.** Widmanstätten microstructures formed in a 0.06 wt%C steel strained to  $\epsilon = 0.7$  at a) 1350 °C ( $A_{e3(p)} + 480\text{ °C}$ ) and (b) 1200 °C ( $A_{e3(p)} + 330\text{ °C}$ ). 103

**Fig. 6.1.1.** Thermomechanical schedule of the torsion tests. 109

**Fig. 6.1.2.** Strip rolling simulation carried out on the 0.06C-0.3Mn-0.01Si steel with interpass times of 1s and pass strains of 0.4 applied at a strain rate of  $1\text{ s}^{-1}$ . Start and finish temperatures were a) 980°C and 932°C, and b) 930°C and 889°C, respectively. 110

**Fig. 6.1.3.** The mean flow stresses associated with the flow curves in Figure 6.1.2. The MFS increases during the 2<sup>nd</sup> pass (due to strain accumulation) and then drops in the following

---

passes because of the softening associated with the transformation from austenite to ferrite as well as with the occurrence of dynamic recrystallization. 111

**Fig. 6.1.4.** Samples subjected to 7 deformations beginning at 980°C (a,b,c) and 930 °C (d,e,f). The amount of ferrite increases in moving from (a) the center to (b) 1.5mm from the surface, and finally to (c) the surface of the sample. A similar trend can be observed in moving from (d) the center, to (e) 1.5mm from the surface, and finally to (f) the surface of the sample. 112

**Fig. 6.1.5.** Optical micrographs of the 0.06C-0.3Mn-0.01Si steel deformed with interpass times of 1s and pass strains of 0.4 applied at a strain rate of  $1s^{-1}$ . The torsion simulation began at 980°C and quenching was performed after (a) 1, (b) 4 and (c) 7 passes. 115

**Fig. 6.1.6.** Critical strains for dynamic transformation (DT) and dynamic recrystallization (DRX) in the strip rolling simulations of Figure 6.1.2. It can be seen that DT and DRX are initiated at strains of approximately 0.04 and 0.11. 116

**Fig. 6.1.7.** Strip rolling simulations carried out on the 0.06C-0.3Mn-0.01Si steel with interpass times of 0.5s, 1s, 1.5s and 3s, and pass strains of 0.4 applied at a strain rate of  $1s^{-1}$ . The dependence of MFS on inverse temperature shows that short interpass times are associated with significant decreases in rolling load, indicating that ferrite is forming well above the  $Ae_3$  temperature. 118

**Fig. 6.1.8.** Strip rolling simulations carried out under isothermal conditions on the 0.06C-0.3Mn-0.01Si steel with interpass times of 1s and 3s, and pass strains of 0.4 applied at a strain rate of  $1s^{-1}$  at 1000°C. The MFS values decrease continuously with pass number for the 1s simulation. The flow stress decrease is much smaller when the interpass is increased to 3s, indicating that the additional holding time leads to some retransformation to austenite. Samples were quenched right after the last pass and revealed the presence of more DT ferrite in the 1s as opposed to the 3s interpass material. 119

**Fig. 6.2.1.** Torsion testing schedules for the simulation of strip rolling employing: a) continuous cooling at 6°C/s, and b) isothermal conditions. All the deformation temperatures are above the respective  $Ae_3$ 's. 125

**Fig. 6.2.2.** Stress-strain curves determined on the plain C steel using the 7-pass continuous cooling schedule of Figure 6.2.1a with pass strains of 0.4 applied at  $1s^{-1}$ . Interpass times of a) 0.5s, b) 1s, c) 1.5s, and d) 3s were employed. The broken line indicates the measured dependence of stress level on temperature in the absence of DT and DRX. The levels of the shorter interpass flow curves decrease more perceptibly than those associated with the longer times. 127

**Fig. 6.2.3.** Continuous cooling rolling simulations carried out on the Nb steel with interpass intervals of a) 0.5s, b) 1s, c) 1.5s, and d) 3s. Here the testing conditions of Figure 6.2.2 were employed. The presence of Nb in solution increases the flow stress levels and the amounts of

---

retained work hardening. There is less of a discrepancy with respect to the “expected” behavior than in Figure 6.2.2. 130

**Fig. 6.2.4.** Isothermal rolling simulations carried out according to the schedule in Figure 6.2.1b on the plain C steel using pass strains of 0.4 applied at  $1\text{ s}^{-1}$  and  $930^{\circ}\text{C}$ . Simulations with interpass intervals of: a) 0.5s and b) 5s were carried out. The short interpass intervals generate greater degrees of load decrease than the long times. 133

**Fig. 6.2.5.** Isothermal rolling simulations carried out on the Nb steel. The test parameters were identical to those of Figure 6.2.4. Interpass intervals of a) 0.5s and b) 5s were employed at a testing temperature of  $910^{\circ}\text{C}$ . There is considerable retention of work hardening, after which there is a decrease in the stress levels at rates that are lower than in Figure 6.2.4. 134

**Fig. 6.2.6.** The mean flow stresses (MFS's) deduced from the continuous cooling simulations of Figure 6.2.2 for the plain C steel plotted against inverse absolute temperature. The broken line indicates the dependence of the MFS in the absence of both DT and DRX. Here the MFS decreases much more sharply when the interpass time is short. 136

**Fig. 6.2.7.** The mean flow stresses (MFS's) deduced from the continuous cooling simulations of Figure 6.2.3 for the Nb steel plotted against inverse absolute temperature. The paraequilibrium and orthoequilibrium  $\text{Ae}_3$  temperatures for this steel are  $827^{\circ}\text{C}$  and  $843^{\circ}\text{C}$ , respectively. Although the flow stress levels are higher due to strain accumulation, it is clear that short interpass times generate greater decreases in rolling load than longer times. 136

**Fig. 6.2.8.** MFS's deduced from the isothermal flow curves of Figures 6.2.4 and 6.2.5. Interpass times of 0.5s and 5s were employed for a) the plain C and b) the Nb microalloyed steel. In Figure 6.2.8a, shorter interpass times lead to more strain accumulation followed by greater decreases in MFS than longer intervals. In Figure 6.2.8b, there is again more strain accumulation and more softening when the interpass times are short. 138

**Fig. 6.2.9.** Optical microstructures of the plain C steel subjected to the  $930^{\circ}\text{C}$  7-pass, 0.5s interpass time rolling simulation of Figure 6.2.4. The samples were quenched: a) immediately after straining, b) after holding for 8s, and c) after holding for 15s. Light regions are ferrite while dark regions are martensite (prior austenite). The DT ferrite is seen to retransform gradually into austenite during holding. 140

**Fig. 6.2.10.** Optical microstructures of the Nb steel subjected to the  $910^{\circ}\text{C}$  7-pass, 0.5s interpass time rolling simulation of Figure 6.2.5. Here the quenching conditions of Figure 6.2.9 were employed: a) immediately after straining, b) after holding for 8s, and c) after holding for 15s. There is less reversion to austenite than in the plain C steel. 141

**Fig. 6.2.11.** EBSD micrographs of transverse cross-sections of the C-Mn and Nb steels subjected to the 7-pass, 0.5s interpass time rolling simulation of Figures 6.2.4 and 6.2.5. The C-Mn samples were quenched: Fig. 6.2.11a - immediately after straining, and Fig. 6.2.11b - after

---

holding for 15s. The Nb steel sample of Fig. 6.2.11c was quenched after 15s of holding. In the right hand figures, the ferrite phase is red and the prior austenite is black. 143

**Fig. 7.1.** Torsion testing schedules for the simulation of strip rolling. The deformation temperatures were 930°C and 910°C for the C-Mn and Nb steel, respectively, and interpass times of 0.5s and 5s were employed. 151

**Fig. 7.2.** Two stress-strain curves derived from the torque/twist data obtained during a torsion test of the present Nb steel. The curve shown as a broken line was derived using the average  $N$  while the solid curve was obtained using the local values of  $N$ . There appears to be slightly more dynamic softening when the local  $N$  is employed and lower values of the critical strains are obtained. 153

**Fig. 7.3.** Stress-strain curves determined on the C-Mn steel according to the schedule in Fig. 7.1 using pass strains of 0.4 applied at a  $1\text{ s}^{-1}$  and 930°C. Interpass times of a) 0.5s and b) 5s were employed. More ferrite was formed and retained when shorter times were used. 153

**Fig. 7.4.** Stress-strain curves determined on the Nb microalloyed steel according to the schedule in Fig. 1 using pass strains of 0.4 applied at  $1\text{ s}^{-1}$  and 910°C. Interpass times of a) 0.5s and b) 5s were employed. The presence of Nb delays both the forward and reverse transformation. 155

**Fig. 7.5.** MFS's calculated from the stress-strain curves of Figs. 7.2 and 7.3. a) For the C-Mn steel, the MFS decreases are immediately evident as a result of ferrite formation during each pass. b) In the Nb steel, however, there is more work hardening retained from pass to pass. Because of carbonitride precipitation during the longer interpass times, there is little decrease in the MFS. 156

**Fig. 7.6.** The critical strains for dynamic transformation (DT) and dynamic recrystallization (DRX) in the C-Mn steel determined by means of double differentiation: a) 0.5s and b) 5s. Critical strains in the Nb steel subjected to interpass times of c) 0.5s and d) 5s. Here the first pass always displays the highest critical strain in all the simulations. 159

**Fig. 7.7.** a) Microstructure of a torsion sample deformed to a strain of 0.4 at the outer radius at 930°C and  $1\text{ s}^{-1}$ . The strain in this sample increases linearly from zero along the axis to 0.4 at the outer radius. Here the presence of DT ferrite can be first seen at a critical strain of about 0.1, while DRX of the prior austenite was initiated at a critical strain of approximately 0.15. b) Widmanstätten ferrite plates formed during deformation at 1200 °C. These plates coalesce into polygonal grains upon continued straining. 161

**Fig. 7.8.** Optical microstructures of the C-Mn steel subjected to the 7-pass simulation at 930°C with interpass times of 0.5s. The samples were quenched immediately after the: a) first pass, b) 3<sup>rd</sup> pass, c) 5<sup>th</sup> pass, and d) final pass. Dark regions are martensite (prior austenite) while the light regions are ferrite. The amount of ferrite increases progressively right up to the last pass. 163

---

**Fig. 7.9.** Optical microstructures of the C-Mn steel subjected to the 7-pass simulation at 930°C with interpass times of 5s. The samples were quenched immediately after the: a) 3<sup>rd</sup> pass, b) 5<sup>th</sup> pass, and c) final pass. Dark regions are martensite (prior austenite) while the light regions are ferrite. There is less ferrite than in the short time simulations because of reverse transformation. 165

**Fig. 7.10.** Optical microstructures of the Nb steel subjected to the 7-pass simulation at 910°C with interpass times of 0.5s. The samples were quenched immediately after the: a) first pass, b) 3<sup>rd</sup> pass, c) 5<sup>th</sup> pass, and d) final pass. Dark regions are martensite (prior austenite) while the light regions are ferrite. The amount of ferrite increases progressively right up to the last pass. 167

**Fig. 7.11.** Optical microstructures of the Nb steel subjected to the 7-pass simulation at 910°C with interpass times of 5s. The samples were quenched immediately after the: a) 3<sup>rd</sup> pass, b) 5<sup>th</sup> pass, and c) final pass. Dark regions are martensite (prior austenite) while the light regions are ferrite. The amounts of ferrite are comparable with those produced in the short time simulations. The presence of Nb significantly retards the reverse transformation. 169

**Fig. 7.12.** Dependence of the volume fraction of ferrite on pass number for the a) C-Mn and b) Nb steels. The increasing amounts of ferrite from pass to pass are in agreement with the hardness measurements displayed on the right hand axis. 171

**Fig. 7.13.** Vickers microhardnesses of the a) ferrite and b) martensite in the C-Mn steel. 172

**Fig. 7.14.** a) Ferrite volume fraction versus pass number when interpass times of 0.5s are employed. More ferrite is formed and retained in the C-Mn than in the Nb steel. b) Ferrite volume fraction versus pass number when interpass times of 5s are employed. In the C-Mn steel, reverse transformation decreases the amount of ferrite volume fraction. The addition of Nb impedes the forward transformation and prevents reverse transformation. 173

---

## List of Tables

|                                                                                                                              |     |
|------------------------------------------------------------------------------------------------------------------------------|-----|
| <b>Table 2.1.</b> Types of hot rolling mills based on the range of interpass times and strain rates. [2.12]                  | 15  |
| <b>Table 3.1.</b> Chemical composition (mass%) and equilibrium transformation temperatures (°C).                             | 43  |
| <b>Table 4.1.</b> Chemical compositions of the Fe-Mn-Si alloys (in wt%).                                                     | 66  |
| <b>Table 4.2.</b> Chemical compositions of the Fe-Mn-Si-Nb alloys (in wt%) in which the occurrence of DT has been confirmed. | 69  |
| <b>Table 5.1.</b> Chemical composition (mass %) and equilibrium transformation temperatures (°C).                            | 96  |
| <b>Table 6.1.1.</b> Chemical composition (mass %) and equilibrium transformation temperatures (°C).                          | 108 |
| <b>Table 6.2.1.</b> Chemical compositions (mass %) and equilibrium transformation temperatures (°C).                         | 124 |
| <b>Table 7.1.</b> Chemical compositions (mass %) and equilibrium transformation temperatures (°C).                           | 150 |

## Chapter 1

### Introduction

The dynamic transformation (DT) from austenite to ferrite, also known as Dynamic Strain-Induced Transformation (DSIT) or Deformation Induced Ferrite Transformation (DIFT), received significant interest since the work of Yada and coworkers in the 1980's [1.1-1.2]. By means of compression tests and actual rolling simulations, they pointed out that austenite dynamically transforms into ferrite during deformation up to 166°C above the  $A_{e3}$  temperature. They also showed that the reverse transformation of the metastable ferrite back into the more stable austenite took place on isothermal holding after deformation. By using a dilatometer, they ascertained that the volume fraction of DT ferrite decreased with holding time until all the DT ferrite was completely transformed back into austenite. Numerous microstructures of DT ferrite are available in the literature from researchers around the world [1.3-1.10].

In 2000, Yada and co-workers provided real time evidence for DT by employing in-situ X-ray diffraction experiments [1.11]. In this work, they were able to capture the diffraction patterns associated with  $(110)_{\alpha-Fe}$  during deformation above the  $A_{e3}$  temperature. This was followed by a series of dilatometric measurements by Chen and Chen in 2003 [1.12], Liu et al. in 2007 [1.13] and Sun et al. in 2008 [1.14].

This unusual phenomenon was initially explained in terms of the dislocation density generated during straining of the austenite. In 2008, Sun et al. [1.14] calculated that the increase in Gibbs energy of the deformed austenite was about 22.3 J/mol. Hanlon et al. [1.15] extended this approach by converting the free energy increase into an increase in the  $A_{e3}$  temperature. In this way, they were able to explain how the transformation can take place up to 10 °C above the  $A_{e3}$  temperature.

However, dynamic transformation has been shown to take place well above the transformation temperature. More recently, Ghosh et al. [1.16] used the same method but in their case, they allowed for the inhomogeneous distribution of dislocations. This resulted in up to 197 J/mol of estimated local energy, which is more than enough to explain the occurrence of DT up to 100 °C above the  $A_{e3}$ .

A limitation of the stored energy model, however, is that it requires large amounts of applied strain, which is inconsistent with the experimentally determined DT critical strains of about 10%. This discrepancy was partially resolved by introducing the concept of mechanical activation [1.17]. In this model, the applied stress plays the role of the driving force. The chemical free energy difference between austenite and ferrite and the work of shear accommodation and lattice dilatation that accompany the transformation were identified as the energy obstacles to the transformation. In this way, they were able to explain the rapid formation rate of ferrite above the  $A_{e3}$  temperature. However, this method did not lead to a satisfactory quantitative description of this phenomenon.

Early strip rolling simulations had indicated that the length of the interpass intervals has a significant effect on the rolling load [1.18-1.21]. Short interpass times lead to reductions in the mean flow stresses (MFS's) [1.22]. At this time, the softening was solely associated with recovery and recrystallization (dynamic and static) [1.23]. No consideration was given to the possibility that austenite was transforming dynamically into the softer ferrite phase.

Many metallurgists are still skeptical about the concept of dynamic transformation despite the evidences available in the literature. The traditional way of transforming austenite to ferrite is by slowly cooling the austenite (i.e. statically) down to temperatures below the  $A_{e3}$ . This reticence is similar to the reaction of metallurgists when the idea that dynamic recrystallization plays a role during hot rolling was first introduced. Clarifying the contribution of dynamic transformation to hot rolling was thus the main motivation for this work. Here, the driving force and obstacles to dynamic transformation are studied in detail. In addition, strip rolling simulations were performed to determine the effect of DT on rolling operations.

This thesis consists of nine chapters. A detailed literature review is presented in **Chapter 2**; this consists of five sections, namely: i) the iron-carbon phase diagram, ii) the development of microstructure in steels, iii) hot rolling processes, iv) dynamic transformation, and v) early strip rolling simulations. The concept of transformation softening to explain the occurrence of dynamic transformation is introduced in **Chapters 3 and 4**. In **Chapter 3**, it is shown that the driving force for dynamic transformation is the flow stress difference between the work hardened austenite at the critical strain and the yield stress of the fresh Widmanstätten plates that take its place. This force has to overcome the energy barrier, which consists of the Gibbs energy difference between austenite and Widmanstätten ferrite, as well as the shear accommodation and dilatation work that accompany the transformation. These values are calculated using data from compression tests carried out on a plain C-Mn steel. In **Chapter 4**, the transformation softening model is applied to published tensile test results on four different steel grades with contrasting levels of Mn and Si to determine the individual effects of the alloying elements. The results show that the addition of Mn increases both the driving force for transformation as well as the energy barrier to ferrite formation. On the other hand, while Si addition increases the driving force, it simultaneously decreases the energy barrier to the transformation. This then increases the temperature range over which DT can take place. This analysis indicates how the transformation softening model can be used to predict the effect on the occurrence of DT of adding various alloying elements.

Another objective of this study was to determine the maximum temperature up to which DT can occur. Some preliminary results are on this topic presented in **Chapter 5**. In this work, compression tests were performed at very high temperatures in the austenite phase field. It was shown here that DT ferrite can be formed at any temperature within the austenite phase field. This behavior is supported by the shape of the free energy obstacle, which displays a peak approximately midway between the  $A_{e3}$  and the delta ferrite formation temperature. A more detailed investigation of this range can be seen in Appendix A.

The rolling load during strip rolling depends on the composition, temperature, reduction and strain rate. The rolling load and mean flow stress (MFS) do not increase as expected with decreasing temperature in strip mills, although it has not been clear why. Recently, various

researchers around the world have shown that austenite can be transformed into ferrite by deformation at temperatures above the  $A_{e3}$ , as briefly described above. The nucleation and growth of the softer ferrite reduce the rolling load and modify the microstructure. In **Chapters 6** and **7**, torsion simulations of strip rolling are described carried out on a plain C-Mn and a Nb microalloyed steel. The aim was to determine the effect of interpass time on dynamic transformation. **Chapter 6** is divided into two parts, rolling simulation: i) without a roughing pass (preliminary results) and ii) with a roughing pass (a more detailed analysis). The torsion tests were carried out under both isothermal and continuous cooling conditions. The microstructures reveal that more ferrite is formed and retained when short holding periods are employed. The results also imply that the addition of Nb retards both the forward and reverse transformation.

This investigation was later extended in **Chapter 7**. Here the volume fractions of ferrite formed and retained from pass to pass are elucidated. These observations are complemented with critical strains, hardness measurements and EBSD and optical micrographs. The present observations clearly demonstrate that the lack of increase in rolling load cannot be attributed solely to dynamic recrystallization. It is also pointed out that the dilatation associated with ferrite formation is responsible for an increase in the volume flow rate (but not the mass flow rate) as a bar passes through a mill.

Finally, the conclusions as well as the contributions to original knowledge are stated in **Chapters 8** and **9**, respectively. In summary, it has been shown in this work that DT occurs at all temperatures over the whole temperature range of the austenite phase field. The effects of DT on the rolling load during strip rolling have been clarified and finally a novel thermodynamic model explaining these observations in terms of transformation softening has been put forward.

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## Chapter 2

### Literature Review

Many steel products are manufactured by strip rolling. This hot deformation technique involves a series of deformations until the final shape and thickness of the steel are attained. In this process, it is important to identify the composition of steel as well as its microstructure at specific temperatures in order to design and optimize the process. The amount of load for each deformation should be predictable so that shape control problems during the hot rolling process can be avoided. As the focus of the present investigation was hot rolling, the various metallurgical phenomena associated with this process are first reviewed.

#### 2.1 Iron-Carbon Phase Diagram

The iron-carbon phase diagram is presented in Figure 2.1, indicating the equilibrium phases present at each temperature. Once the temperature of pure iron reaches 910 °C, the crystal structure transforms from body centered cubic (BCC) to the face centered cubic (FCC) structure known as austenite, or  $\gamma$  iron. Iron is fully austenitic up to 1390 °C. The FCC structure of iron transforms back into the BCC structure, known as  $\delta$  iron, above 1390 °C until it melts at 1539 °C. The addition of carbon as an alloying element lowers the austenitizing temperature down to 723 °C when it reaches the eutectoid point with 0.8 wt% carbon content. The boundary between pure austenite and the initial formation of ferrite is known as the  $A_{e3}$  temperature, a very important temperature in the present work. As the temperature is slowly decreased, all the austenite transforms into a ferrite + pearlite microstructure at the  $A_{e1}$  temperature or 723 °C for the Fe-C system. Pearlite consists of a lamellar structure of ferrite and cementite ( $Fe_3C$ ). The maximum solubility of carbon in the ferrite phase is 0.02 wt% and 2.14 wt% for the

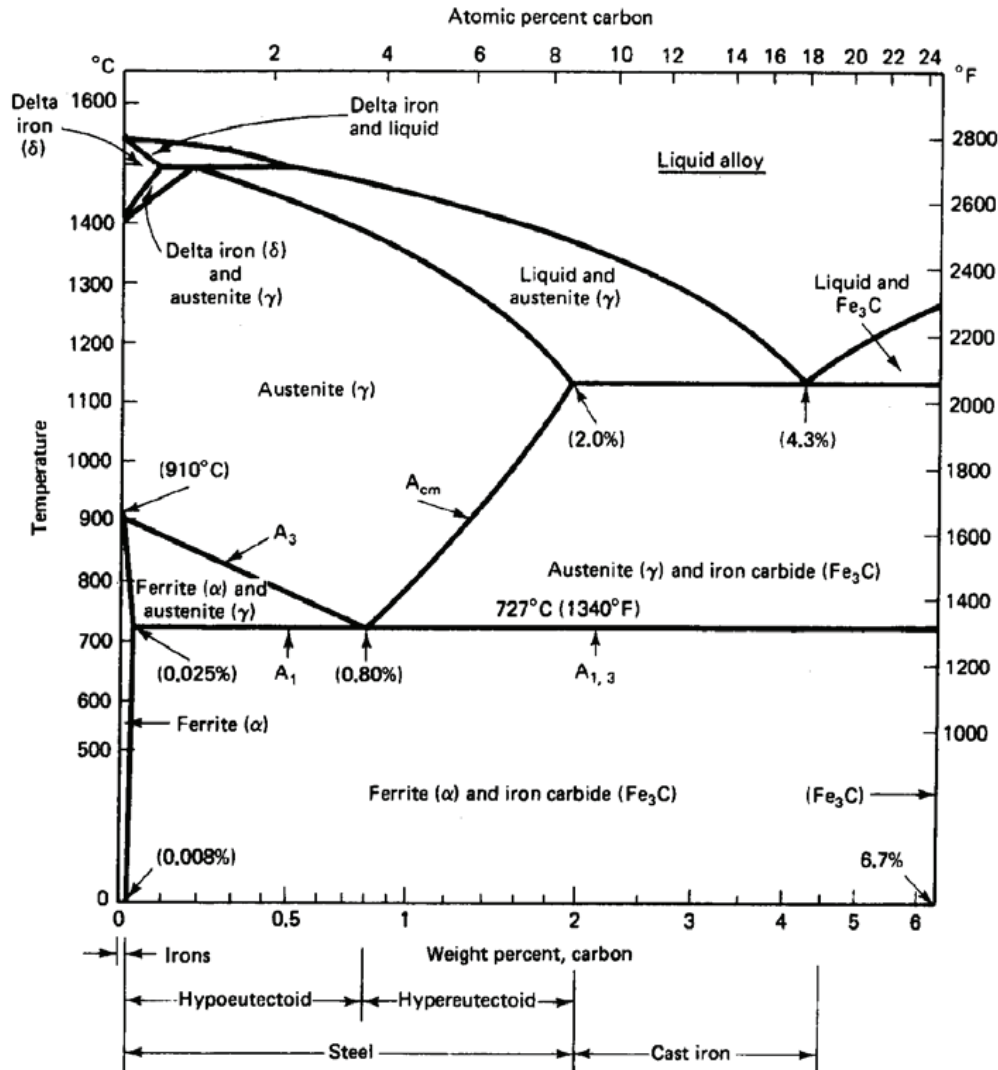


**Fig. 2.1.** Iron-carbon phase diagram showing the transformation temperatures and corresponding stable phases for each temperature [2.1].

austenite phase. This solubility limit is dependent on the crystal structure and the shapes of the interstitial sites available for the carbon atoms to occupy.

The addition of other alloying elements significantly changes the Fe-C phase diagram, especially with regard to the stability of the ferrite and austenite phases. The shift in phase boundaries depends on the alloying element and its concentration. Alloying elements known as the austenite stabilizers (i.e. C, Co, Cu, Mn, N and Ni) lower the eutectoid temperature and widen the range of temperature over which the austenite phase is stable. On the other hand, ferrite

formers (i.e. Al, Cr, Mo, Si and W) increase the eutectoid temperature, restricting the formation of the austenite phase. Alloying elements are also added to improve material properties such as the corrosion resistance.

This kind of phase diagram is usually determined based on orthoequilibrium conditions. This involves the diffusion of both the substitutional and interstitial atoms (full equilibrium) during transformation, which can take considerable time. Note that the diffusional movement of interstitial elements such as C and N is orders of magnitudes faster than that of the substitutional ones [2.2]. On the other hand, paraequilibrium is a condition where only partial equilibrium is attained [2.3]. In this case, only the interstitial elements are able to partition due to time constraints. The paraequilibrium temperatures are lower than the orthoequilibrium ones by about 5 °C to 30 °C, depending on the alloying element. In the present work, transformation occurs under paraequilibrium conditions.

## **2.2 Development of Microstructure in Steels**

In the discussion above, the presence of full equilibrium cooling conditions is assumed and that sufficient time has been given for the transformation to occur. For most applications, non-equilibrium cooling is preferred in order to modify the mechanical properties for specific functions. Various microstructures are formed in steels depending on the cooling rates imposed during processing. The development of microstructure in steels can be divided into two types of mechanism; i) diffusional and ii) displacive, as will now be described briefly.

### **2.2.1 Diffusional mechanisms**

The diffusional mechanisms involve the diffusion of all atoms during transformation. This requires very slow cooling from the austenite phase field down to temperatures below the  $Ae_3$  and  $Ae_1$  lines. The phases formed during this type of transformation are allotriomorphic and idiomorphic ferrite and pearlite.

### 2.2.1.1 Ferrite

Ferrite is the thermodynamically stable phase of steel below the  $A_{e3}$  temperature. It is nucleated within austenite grains during cooling from temperatures above the  $A_{e3}$  and is commonly seen as equiaxed shapes under the optical microscope [2.4]. Nucleation of the ferrite phase can occur at grain boundaries when it is known as *Allotriomorphic Ferrite*. Nucleation inside the austenite grains leads to *Idiomorphic Ferrite* [2.5]. The austenite grain boundaries and the surfaces of inclusions are preferred sites for the nucleation of ferrite. Ferrite forming inside grains nucleates heterogeneously on non-metallic inclusions present in the steel. Figure 2.2 illustrates the formation of ferrite [2.5].

### 2.2.1.2 Pearlite

Pearlite consists of a lamellar structure of ferrite and cementite. At temperatures above the  $A_{e3}$ , steel is fully austenitic. The formation of pearlite starts when the austenite reaches the  $A_{e1}$  temperature. The solubility of carbon in ferrite is only 0.02%, and any excess is rejected into the

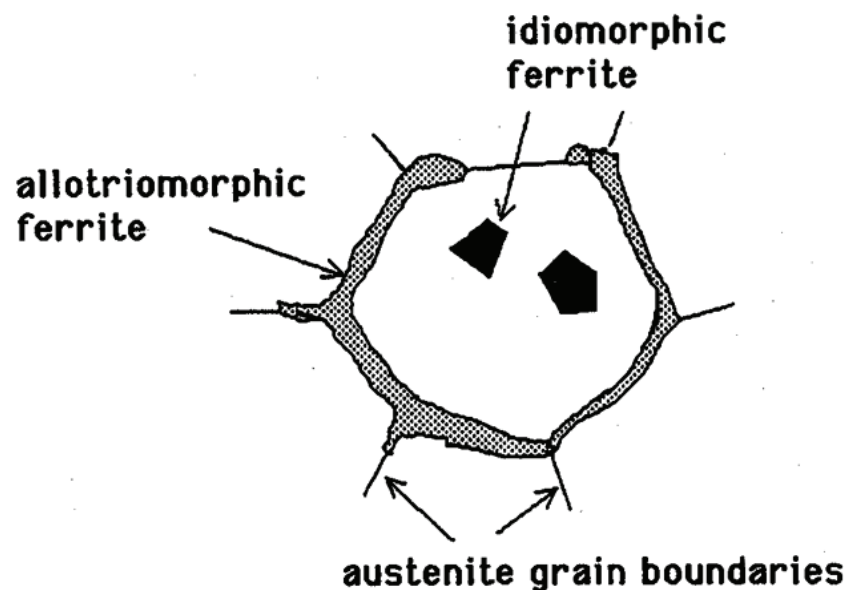


**Fig. 2.2.** Nucleation of allotriomorphic and idiomorphic ferrite [2.5].

austenite. This results in the formation of cementite on both sides of the ferrite plate. As the carbon concentration on both sides of the cementite plate decreases, this leads to the formation of another ferrite plate. This is the sidewise nucleation of pairs of lamellae responsible for the pearlite structure. The microstructure is formed by a reconstructive transformation, where the atomic mobility of carbon is sufficiently high and there is enough time for all the atoms to diffuse. Figure 2.3 shows a representation of the pearlite structure in steel [2.6].

### 2.2.2 Displacive mechanisms

The displacive transformation takes place when there is insufficient time for any diffusion to occur. This can be achieved by rapidly cooling or quenching the austenite phase. The non-equilibrium phases of steels are martensite, bainite and Widmanstätten ferrite.

#### 2.2.2.1 Martensite

Martensite is a non-equilibrium single phase structure produced by rapid cooling or quenching austenite to a low temperature. It is formed by a diffusionless or displacive transformation. The

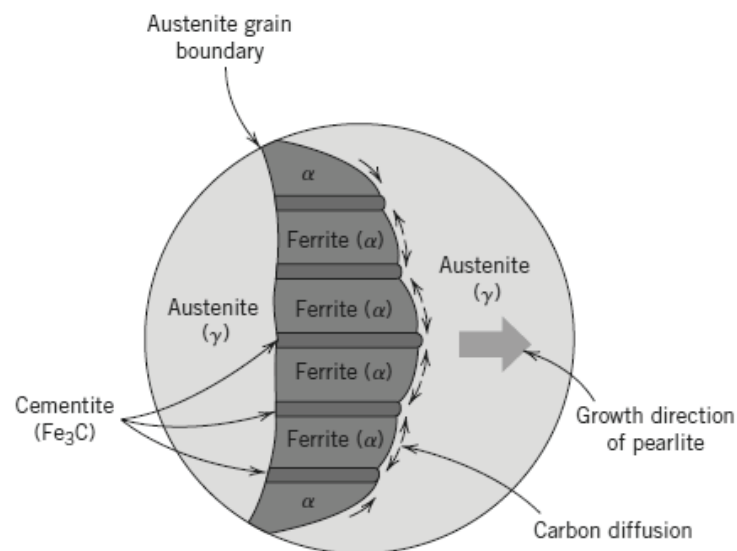


**Fig. 2.3.** Representation of pearlite formation showing alternating layers of lamellar ferrite and cementite plates [2.6].

initial FCC austenite undergoes polymorphic transformation to the body centered tetragonal (BCT) martensite phase. This transformation increases the volume because the austenite is denser than the martensite. High cooling rates prevent the carbon atoms from diffusing, thus preventing the formation of ferrite or cementite. Martensite with a supersaturated solid solution of carbon is formed and the carbon atoms remain in the interstices causing lattice strain and distortion. It is brittle and considered as the hardest phase of steel with needle-like grains appearing under the microscope. Figure 2.4 shows an example of martensite taken from an optical microscope. There is a very limited application of martensite because of its brittleness. The toughness can be recovered by tempering the as-quenched martensite between 250 °C – 600 °C. The microstructure transforms into tempered martensite with better mechanical properties [2.7].

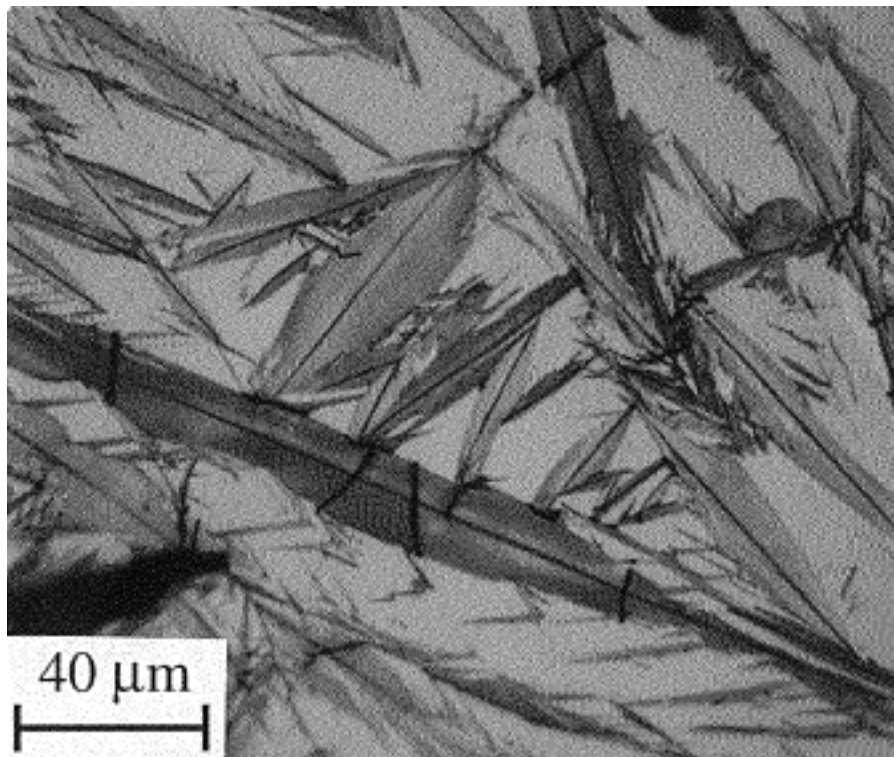


**Fig. 2.4.** Microstructure of plate martensite in an Fe–1.86 wt.% C alloy [2.7].

### 2.2.2.2 Bainite

Bainite consists of ferrite and cementite and is also formed under non-equilibrium cooling conditions. The cooling rate required to form bainite from austenite is between the cooling rates required for pearlite and martensite formation. Bainite has an orientation relationship with respect to the prior austenite phase (as does martensite) which differentiates it from pearlite. The bainite microstructure nucleates in a manner similar to martensite but in this case carbon diffusion also takes place. It can nucleate at higher temperatures, when it is known as *upper bainite* or at lower temperatures, when it is known as *lower bainite*. The mechanism of formation of the carbide precipitates is the main difference between these. Figure 2.5 shows a schematic diagram of bainite formation [2.8,2.9].

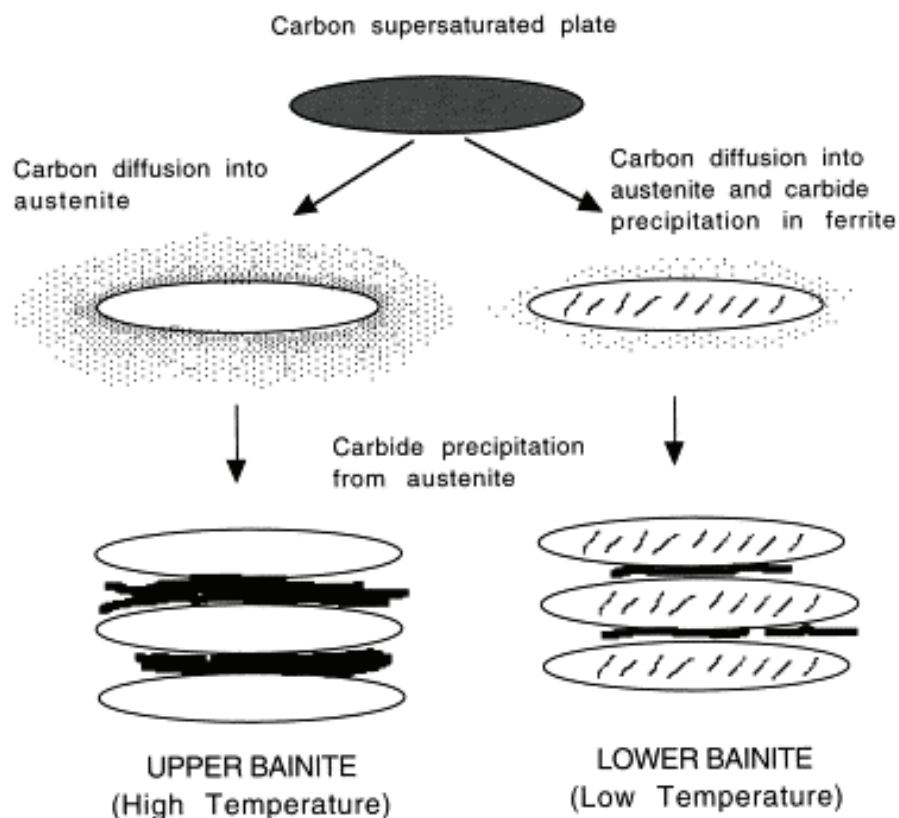


**Fig. 2.5.** Diagram illustrating the growth of upper and lower bainite based on carbon diffusion [2.8].

Upper bainite is a mixture of needle-like ferrite with cementite particles in between the laths. The needle-like ferrite nucleates at the austenite grain boundaries and the mechanism is similar to that involved in the formation of Widmanstätten ferrite. Growth of the ferrite laths increases the carbon concentration of the remaining austenite. The cementite nucleates when the austenite reaches a critical level of carbon. Lower bainite forms at temperatures below about 350 °C. The initial ferrite laths formed transform into ferrite plates at lower temperatures. Carbon diffusion is sluggish at low temperatures resulting in carbide precipitation inside the ferrite plates and in the remaining austenite [2.8,2.9].

### 2.2.2.3 Widmanstätten Ferrite

Widmanstätten ferrite consists of long and thin pointed plates of ferrite nucleated at high temperatures, usually just below the  $A_{e3}$  temperature. The growth and length of the ferrite plate depends on the paraequilibrium mechanism of carbon diffusion in the austenite. The ferrite plates can nucleate on an austenite grain boundary, when it is known as *primary Widmanstätten ferrite* or it can form from allotriomorphic ferrite, when it is known as *secondary Widmanstätten ferrite*. It grows by a reconstructive mechanism, but it nucleates by a displacive mechanism. Figure 2.6 illustrates the nucleation of Widmanstätten ferrite according

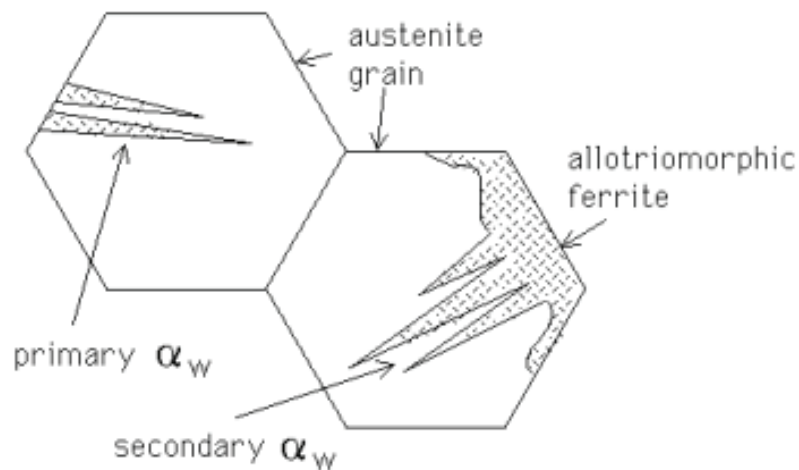


**Fig. 2.6.** Nucleation of primary and secondary Widmanstätten ferrite [2.10].

to these two types of transformation [2.10]. The present work is also concerned with the formation of Widmanstätten ferrite but, in this case, formation takes place entirely *above* the  $A_{e3}$  and the plates are much finer.

### 2.3 Hot Rolling Processes

Steel strip products are manufactured by strip rolling, as shown in Figure 2.7. The process involves reheating the steel slab at the austenitization temperature of 1100-1250 °C. The austenite phase then passes through the roughing mill at temperatures above that of “no-recrystallization” ( $T_{nr}$ ), which allows recrystallization of the grains. The final step involves the finishing mill, normally composed of 7 mill stands, which reduces the thickness to its final shape. The time it takes from one deformation to another in the finishing mill is known as the interpass time. The steel then enters the run-out table to cool down using water jets. The cooling rate dictates the strength and toughness of the final product by modifying the microstructure.

Hot rolling mills can be divided into different types depending on the interpass time and strain rates [2.12]. This is summarized in Table 2.1 below.

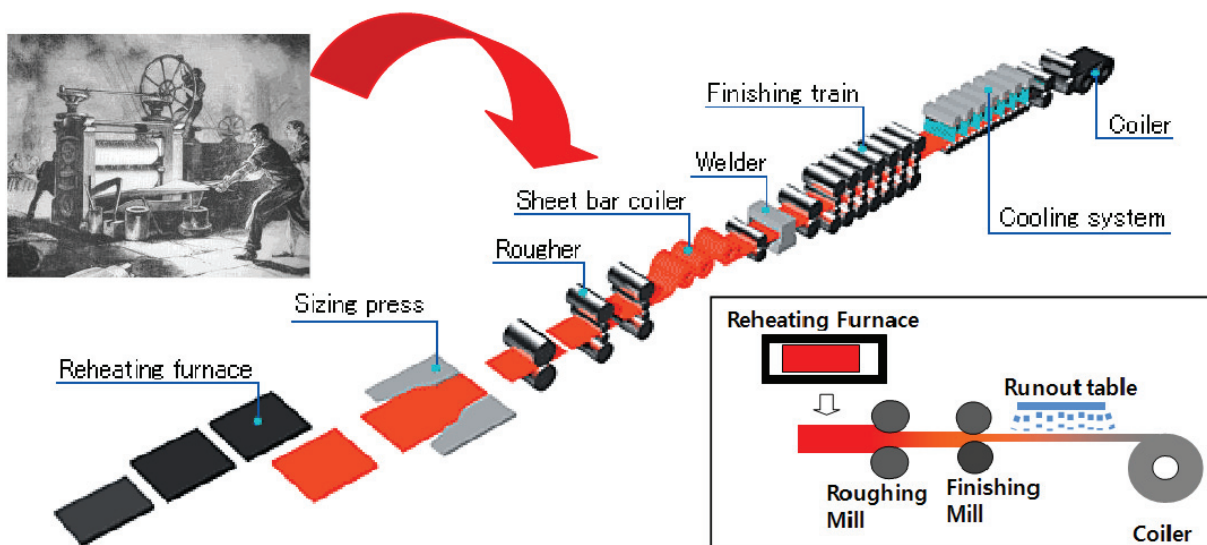


**Fig. 2.7.** A strip rolling operation with 7 finishing stands [2.11].

**Table 2.1.** Types of hot rolling mills based on the range of interpass times and strain rates [2.12]

| Mill Type                         | Strain Rate ( $s^{-1}$ ) | Interpass Time (s) |
|-----------------------------------|--------------------------|--------------------|
| Reversing Mills (plate, roughing) | 1-30                     | 8-20               |
| Hot Strip Mills                   | 10-200                   | 0.4-4              |
| Wire Rod Mills                    | 10-1000                  | 0.01-1             |

During rolling, various hardening and softening mechanisms operate. The basic concepts associated with the controlled rolling of microalloyed steels as well as with strip rolling are discussed in the section that follows.

### 2.3.1 Mean Flow Stress

The Mean Flow Stress (MFS) depends on the composition, temperature of deformation, strain and strain rate. This can be calculated by taking the area under the stress-strain curve ( $\sigma$ - $\epsilon$ ) normalized by the corresponding strain, as given by the equation [2.13]:

$$MFS = \frac{1}{\epsilon_a - \epsilon_b} \sum_a^b \sigma_{av} \Delta \epsilon \quad (2.1)$$

where  $\epsilon_a$  and  $\epsilon_b$  are the start and end of the strain interval, respectively.

In the hot rolling industry, the mean flow stress of each pass can provide information about microstructure changes such as phase transformation, precipitation, recrystallization, and strain accumulation. Knowledge of the microstructural changes at specific temperatures can lead to optimized pass schedules and help to avoid shape control problems during hot rolling. Some measured MFS values as a microalloyed steel is deformed at various temperatures from 1265°C down to 680 °C are illustrated in Figure 2.8. The MFS values in Region 1 and Region 2 increase with decreasing temperature. The rate of MFS decrease in Region 1 is less than that of Region 2 because of the occurrence of recrystallization at high temperatures. This results in softening of the steel between passes and is responsible for the lower rate of MFS increase. Upon reaching

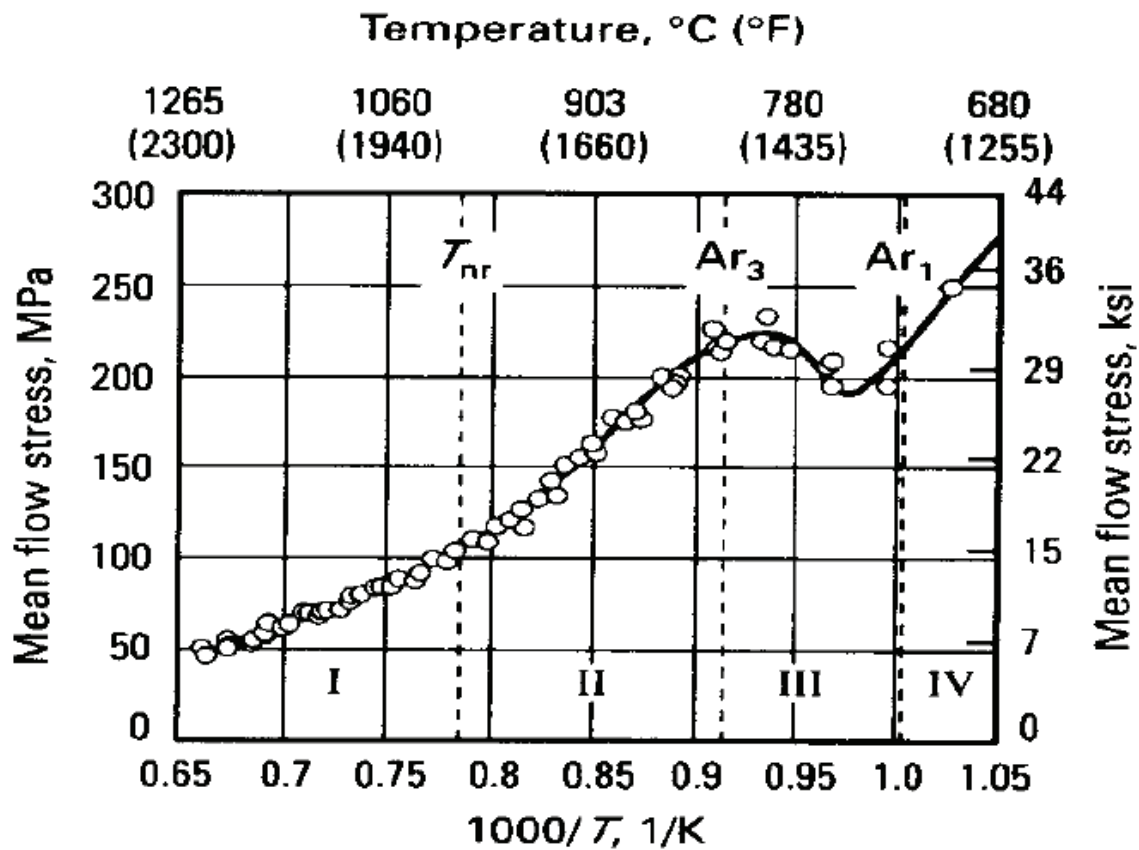


**Fig. 2.8.** The mean flow stress behavior of a microalloyed steel during simulated Steckel mill rolling down to temperatures below the  $Ar_1$  [2.13].

region 2, recrystallization stops and the significant increase in MFS is caused by strain accumulation in addition to the decrease in temperature. In Region 3, which is below the  $Ar_3$  temperature, nucleation and growth of the softer ferrite phase has begun, resulting in a decrease in the MFS values [2.13].

### 2.3.2 Basics Concepts in Steel Rolling

Both hardening and softening take place during hot rolling. Hardening is associated with the accumulation of dislocations during the application of strain. Softening occurs by means of static and dynamic recovery, static recrystallization, dynamic recrystallization and metadynamic recrystallization.

### 2.3.2.1 Recovery

Hot deformation leads to the storage of strain energy in the deformed material. This energy is released by the annihilation and rearrangement of dislocations during (dynamic recovery) or after (static recovery) deformation. An increase in temperature, strain and strain rate increases the rate of recovery. There is no movement of grain boundaries but the sub-boundaries become sharper and the dislocation density decreases with very little change in the size and shape of the grains. The recovery process takes place below the recrystallization temperature and also prior to recrystallization.

### 2.3.2.2 Static Recrystallization

Static recrystallization occurs after deformation at high temperatures. The deformed grains are replaced as a result of the nucleation and growth of new stress-free grains. Critical values of strain and temperature are required for static recrystallization to occur. The rate of static recrystallization is dependent on the strain, strain rate, temperature and grain size. Many researches have been conducted to study these effects [2.14-2.16]. Although it is difficult to calculate the fraction of recrystallization, most of the studies of the rate of static recrystallization follow the Avrami equation

$$X_{rex} = 1 - \exp \left[ -C \left( \frac{t}{t_F} \right)^k \right] \quad (2.2)$$

where  $X_{rex}$  is the fraction of recrystallized grains at time  $t$ ,  $t_F$  is the time for a specific fraction of recrystallization,  $C = -\ln(1-F)$  and  $k$  is a constant. Static recrystallization is an important softening mechanism in rolling as it can account for as much as 75% softening [2.17].

### 2.3.2.3 Dynamic Recrystallization

Dynamic recrystallization takes place *during* deformation unlike static recrystallization, which occurs after straining. When a critical value of strain is attained, the deformed grains begin to

recrystallize, a process that is initiated at the austenite grain boundaries [2.18]. Recrystallization continues during deformation until all the deformed austenite grains are converted into a new microstructure [2.19]. The critical strain for C-Mn steels can be estimated as about 80% of the strain at the peak stress,  $\epsilon_p$ , or from the equation

$$\epsilon_c = 0.57d_0^{0.17}\epsilon_p \quad (2.3)$$

where  $d_0$  is the initial grain size and  $\epsilon_p$  can be measured on the stress-strain curve. The grain size and peak strain are also related to the Zener-Hollomon parameter, given by

$$\epsilon_p = 6.97 \times 10^{-4} d_0^{0.3} Z^{0.17} \quad (2.4)$$

$$Z = \dot{\epsilon} \exp\left(\frac{312000}{RT}\right) \quad (2.5)$$

When dynamic recrystallization is initiated, the rate of dynamic recrystallization as well as the recrystallized grain size can be calculated by the equations:

$$t_{0.5} = 1.06 \times 10^{-5} Z^{-0.6} \exp\left(\frac{300000}{RT}\right) \quad (2.6)$$

$$d_{rex} = 1.8 \times 10^3 Z^{-0.15} \quad (2.7)$$

where  $t_{0.5}$  is the time to attain 50% recrystallized grains and  $d_{rex}$  is the recrystallized grain size. Equation 2.6 shows that the grain size of dynamically recrystallized grains is independent of strain but an increase in strain beyond  $\epsilon_c$  and  $\epsilon_p$  results in an increase in the volume fraction of dynamically recrystallized grains.

#### 2.3.2.4 Metadynamic Recrystallization

Metadynamic recrystallization is a kind of combination of static and dynamic recrystallization. During dynamic recrystallization, many nuclei are generated. These nuclei continue to grow

after unloading, which results in the continued recrystallization of the austenite [2.20,2.21]. This phenomenon is normally observed during hot deformation when the peak strain is exceeded.

## **2.4 Dynamic Transformation**

Deformation and phase transformation are usually two separate solid state processes in steels. The former is defined in terms of the strain, strain rate and temperature while the latter is associated with the temperature, time and heating and cooling rates. The combination of these two processing techniques led to the development of Thermo Mechanical Controlled Processing (TMCP) in the 1960's [2.22]. This industrial method provides higher strength and toughness in steel products.

It has been widely understood that only recrystallization and recovery take place during and after deformation. However, in 1981, Priestner showed that the austenite to ferrite transformation can take place during straining [2.23]. They initially called this "Strain Induced Transformation to Ferrite" (SITF). Other researchers referred to this phenomenon as "Deformation Induced Ferrite Transformation" (DIFT) and "Dynamic Strain Induced Transformation" (DSIT) [2.22,2.24-2.26]. Most of these experiments were performed at temperatures between the  $A_{e1}$  and  $A_{e3}$ , where austenite and ferrite co-exist.

The dynamic transformation (DT) from austenite to ferrite only received significant interest since the work of Yada and coworkers in the late 1980's [2.27]. By means of compression tests and actual rolling simulations, they showed that austenite transforms dynamically into ferrite during deformation as much as 166 °C above the  $A_{e3}$  temperature. Strip rolling simulations using interpass times of 1 s were also carried out by Yada et al. in 1988 [2.28]. These experiments were done above the  $A_{e3}$  temperature and revealed similar results. By using a dilatometer, they also showed that reverse static transformation takes place. The fraction of dynamically transformed ferrite decreased with increasing holding time after deformation until all the DT ferrite was completely transformed back into austenite. The forward dynamic transformation was shown to be much faster than the reverse static transformation. The

microstructure produced by dynamic transformation is ultra-fine-grained ferrite with potential industrial applications since grain size refinement improves the mechanical properties.

#### 2.4.1 Real time evidence of DT

In Yada et al.'s more recent 2000 paper, they used in-situ X-ray diffraction experiments to provide real time evidence of dynamic transformation [2.29]. The schematic diagram of their experimental setup is presented in Figure 2.9. In this work, iron alloys with varying carbon contents with approximately 6 wt% nickel were subjected to torsion tests about 40 °C above the paraequilibrium  $A_{e3}$  temperature. It is important to note that their flow stresses were lower than those for a fully austenite phase. The results of their in-situ x-ray diffraction are summarized in Figure 2.10. The diffraction pattern at a temperature below the  $A_{e3}$  is shown in Figure 2.10a, where a strong diffraction line corresponding to  $(110)_{\alpha-Fe}$  is clearly visible. When the sample was fully austenitized, this was replaced by several lines associated with  $(111)_{\gamma-Fe}$  (Figure 2.10b). When deformation was initiated at this temperature, another line appeared, which corresponds to the  $(110)_{\alpha-Fe}$  line (Figure 2.10c).

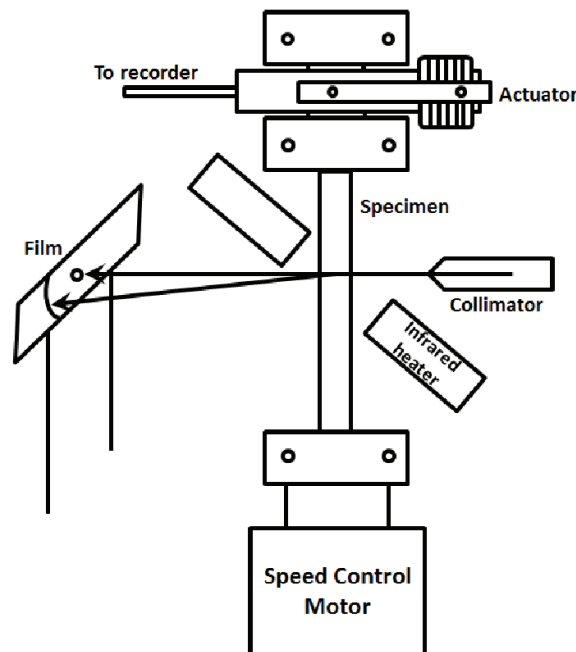


**Fig. 2.9.** Schematic diagram of Yada's in-situ X-ray experimental setup [2.29]

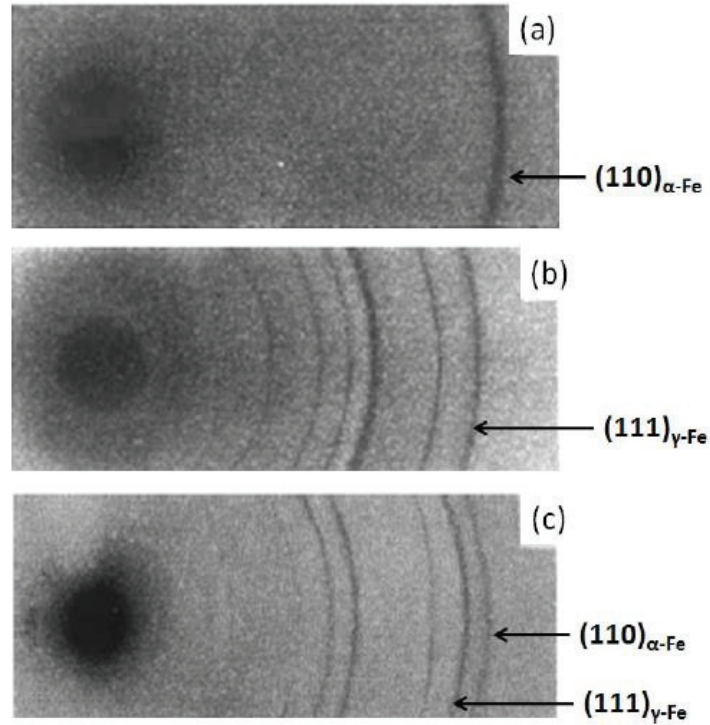


**Fig. 2.10.** Results of in-situ X-ray diffraction experiments (a) before the experiment, (b) at a temperature above the  $A_{e3}$ , and (c) during deformation above the  $A_{e3}$  [2.29].

In 2003, Chen and Chen studied the reverse static transformation [2.30]. They followed the conversion of the dynamically formed ferrite back into austenite by means of laser dilatometry. This took about 45 seconds while the forward dynamic transformation only took a fraction of a second. This was followed by a series of experiments by Liu et al. in 2007 [2.31] and Sun et al. in 2008 [2.32] to further investigate the reverse static transformation.

Examples of dilatation curves by Sun et al. [2.32] are displayed in Figure 2.11. These samples were deformed at temperatures above the  $A_{e3}$  and held isothermally to observe the occurrence of reverse transformation. At 950 °C ( $A_{e3} + 115^{\circ}\text{C}$ ), there was no change in the sample dimensions, which suggested the absence of dynamic transformation. However, for the samples strained at 840 °C ( $A_{e3} + 5^{\circ}\text{C}$ ) and 860 °C ( $A_{e3} + 25^{\circ}\text{C}$ ), the results showed significant changes in diameter which correspond to the reverse transformation. These conclusions were supported by optical microstructures and samples of these are illustrated in Figure 2.12.

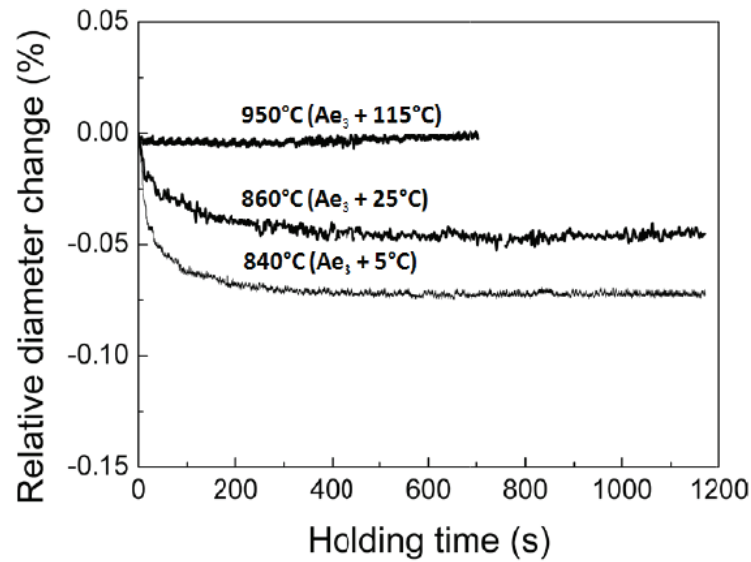


**Fig. 2.11** Dilatation curves of 0.17%C steel specimens held isothermally after deformation at various temperatures above the  $Ae_3$ . [2.32]

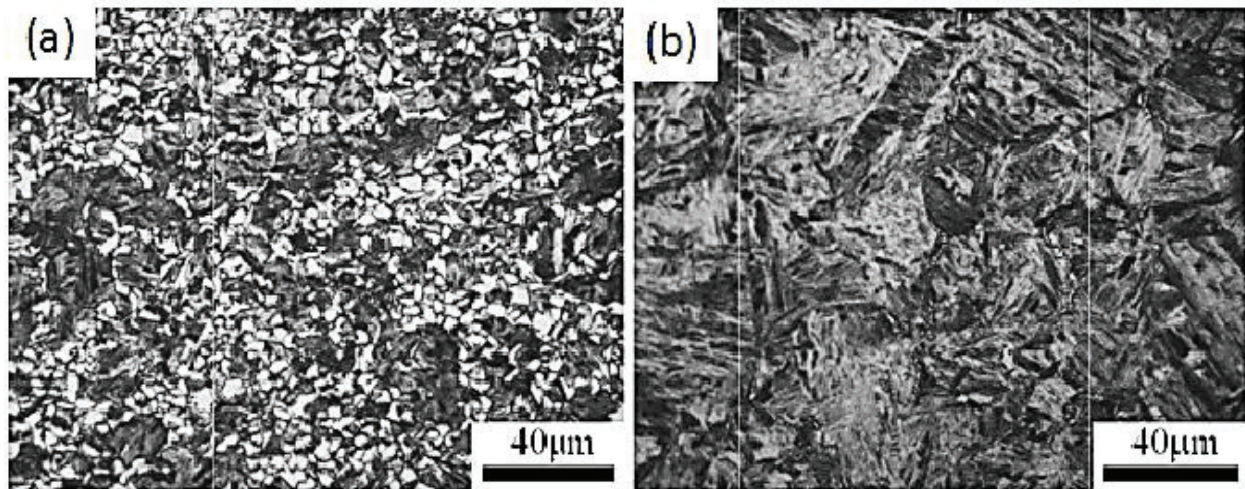


**Fig. 2.12.** Microstructures of 0.17% steel specimens (a) quenched immediately after deformation at 850 °C (i.e.  $Ae_3 + 15$  °C) and (b) subjected to 1200 s of holding following deformation and then quenched. [2.32]

#### 2.4.2 Microstructural evidence of DT

More recently, Basabe et. al. presented various optical micrographs [2.33-2.35] taken from samples deformed and quenched above the paraequilibrium  $Ae_3$  temperature. An example of

these is shown in Figure 2.13 [2.34]. In this work, torsion testing was performed on a low carbon steel strained to 1.5 and 2.5 at rates of  $0.04 \text{ s}^{-1}$  and  $0.4 \text{ s}^{-1}$ , respectively. These micrographs revealed that significant amounts of polygonal DT ferrite had formed. In the same study, they also described the characteristics of the needle-like martensitic structures formed during quenching, as illustrated in Figure 2.14. These were located near the centers of the cross sections of the torsion sample, where the strain values were below the critical strain. Various researchers around the world also deformed specimens above the  $A_{e3}$  and were able to obtain similar optical microstructures [2.36-2.39].

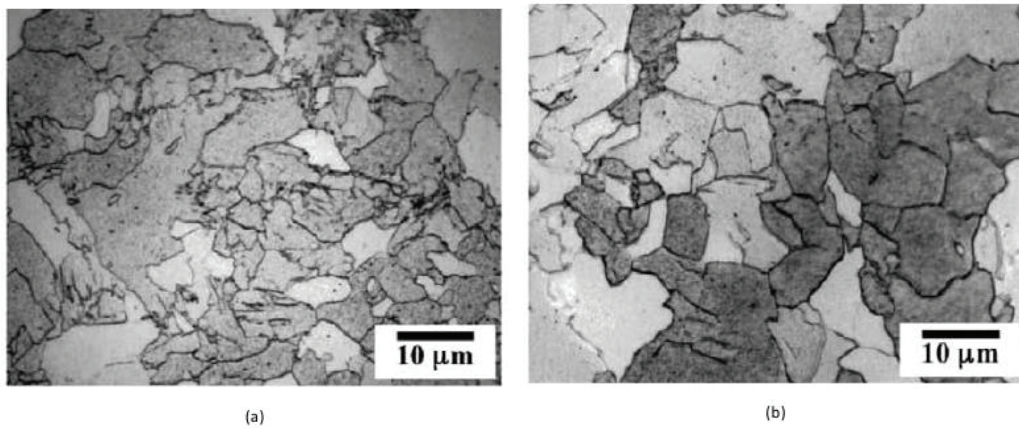


**Fig. 2.13.** Optical micrographs of DT ferrite formed in steel samples deformed and quenched at (a) 887 °C ( $A_{e3} + 10 \text{ °C}$ ) to a strain of 1.5 at the rate of  $0.04 \text{ s}^{-1}$  and (b) 917 °C ( $A_{e3} + 40 \text{ °C}$ ) to a strain of 2.5 at  $0.4 \text{ s}^{-1}$ . [2.34]

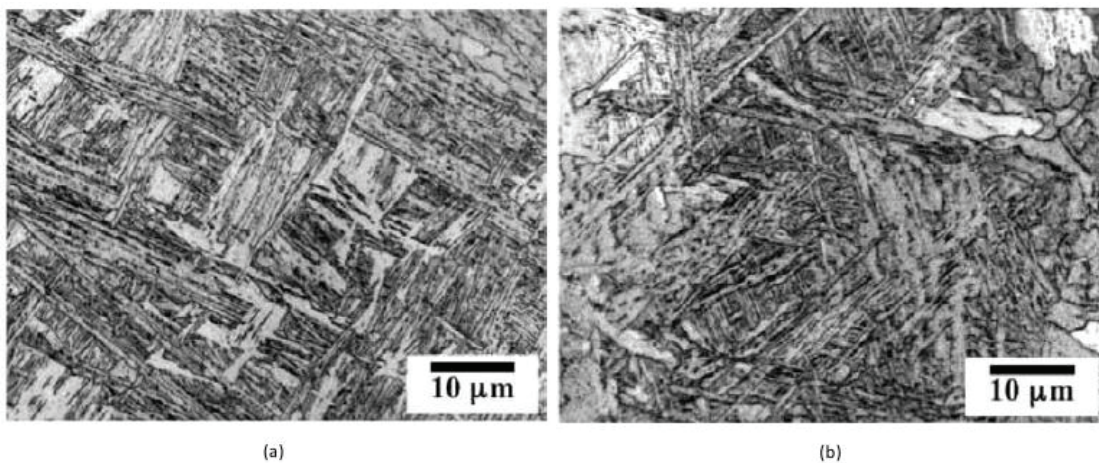


**Fig. 2.14** Optical micrographs of martensite formed in steel samples deformed and quenched at (a) 887 °C ( $A_{e3} + 10 \text{ °C}$ ) to a strain of 1.5 at the rate of  $0.04 \text{ s}^{-1}$  and (b) 917 °C ( $A_{e3} + 40 \text{ °C}$ ) to a strain of 2.5 at  $0.4 \text{ s}^{-1}$ . [2.34]

These studies were followed by a series of electron microscopy investigations using both EBSD and TEM techniques. An example of an EBSD micrograph taken by Park and co-workers in a Fe-6Ni-0.1C is displayed in Figure 2.15 [2.36]. Here the samples were deformed to a strain of 0.96 at 750 °C ( $A_{e3} + 22^{\circ}\text{C}$ ) and  $10\text{ s}^{-1}$ . The micrographs confirmed the presence of both martensite and ferrite. Similar EBSD results are also available in the literature [2.40,2.43]. In order to further verify the presence of metastable ferrite, Ghosh et al presented TEM images with diffraction patterns of a 0.79C-0.65Mn-0.24Si steel deformed to a strain of 4 at 763 °C ( $A_{e3} + 30^{\circ}\text{C}$ ) and  $4\text{ s}^{-1}$ , as illustrated in Figure 2.16 [2.42]. In this work, they were able to identify the diffraction patterns associated with ferrite, martensite and cementite.

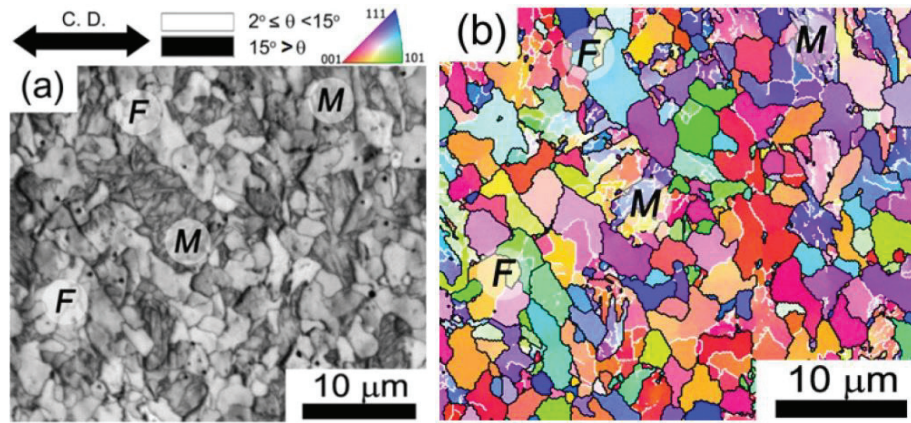


**Fig. 2.15.** EBSD maps of a Fe-6Ni-0.1C steel deformed to a strain of 0.96 at 750 °C ( $A_{e3} + 22^{\circ}\text{C}$ ) and  $10\text{ s}^{-1}$  showing (a) image quality and (b) inverse pole figure maps. 'F' and 'M' indicate ferrite and martensite, respectively. [2.36]

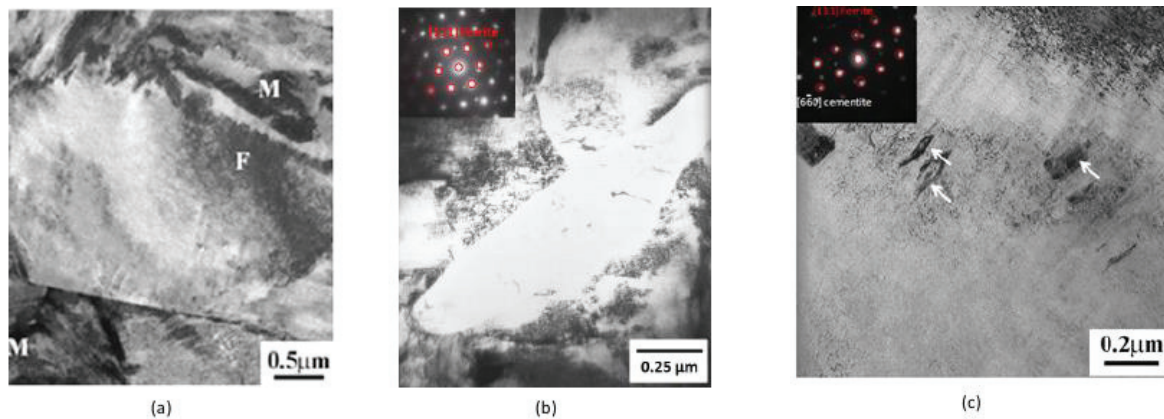


**Fig. 2.16.** Transmission electron micrographs of a 0.79C-0.65Mn-0.24Si steel deformed to a strain of 4 at 763 °C ( $A_{e3} + 30^{\circ}\text{C}$ ) and  $4\text{ s}^{-1}$ . The sample was quenched immediately after deformation. The TEM images show (a) the presence of ferrite and martensite, (b) the morphology of ferrite plates and (c) cementite precipitates. [2.42]

### 2.4.3 Formation Mechanism of Widmanstätten Ferrite

The formation of DT ferrite involves the nucleation of Widmanstätten ferrite plates at lower strains followed by coalescence of the plates into polygonal grains upon further straining [2.43]. These plates are a few hundred nanometers wide and can only be detected using electron microscopy [2.43]. This plate formation results in carbon rejection into the adjacent austenite phase [2.44]. The carbon-rich area then transforms into martensite on water quenching.

The habit plane of this type of transformation can be identified by determining the orientations of the prior austenite as well as of the transformed ferrite together with that of the interface. In a 0.45wt%C steel, the austenite habit plane was shown to be  $\{0.5057, 0.4523, 0.7346\}$  and the shear direction in the austenite to be  $\langle -0.8670, 0.4143, 0.2770 \rangle$  [2.45]. In the case of ferrite, this corresponds to  $\{0.0188, 0.5415, 0.8408\}$   $\langle -0.9397, -0.3026, 0.1589 \rangle$  [2.46]. Note that the habit planes and shear directions are not perfectly orthogonal because of the dilatation involved during transformation.

An example of an EBSD map showing the orientations and shapes of Widmanstätten ferrite plates is illustrated in Figure 2.17. One important feature of DT ferrite plates is that they have near-identical orientations. These plates form in self-accommodating pairs as a result of the opposite shears generated during deformation [2.47]. Here it is important to note the role of the opposite shears in producing the near-identical orientations. A diagram showing the geometry of plate formation is described in Figure 2.18. When a shear stress is applied and a plate is formed (see Figure 2.18a), shear accommodation is necessary. This results in the development of an internal stress, which is responsible for the formation of the adjoining plate (see Figure 2.18b). Here the austenite to the left and right of the growing plate has to accommodate the associated shear strain. [2.48]

More recently, the mechanism of plate formation was investigated by Basabe et al. by carrying out torsion tests on a 0.036 wt% Nb microalloyed steel [2.49]. In this work, they used EBSD techniques to show the two stages of ferrite formation. This is summarized in Figure 2.19. The first stage is the formation of Widmanstätten ferrite plates about 200 nm wide at strains below

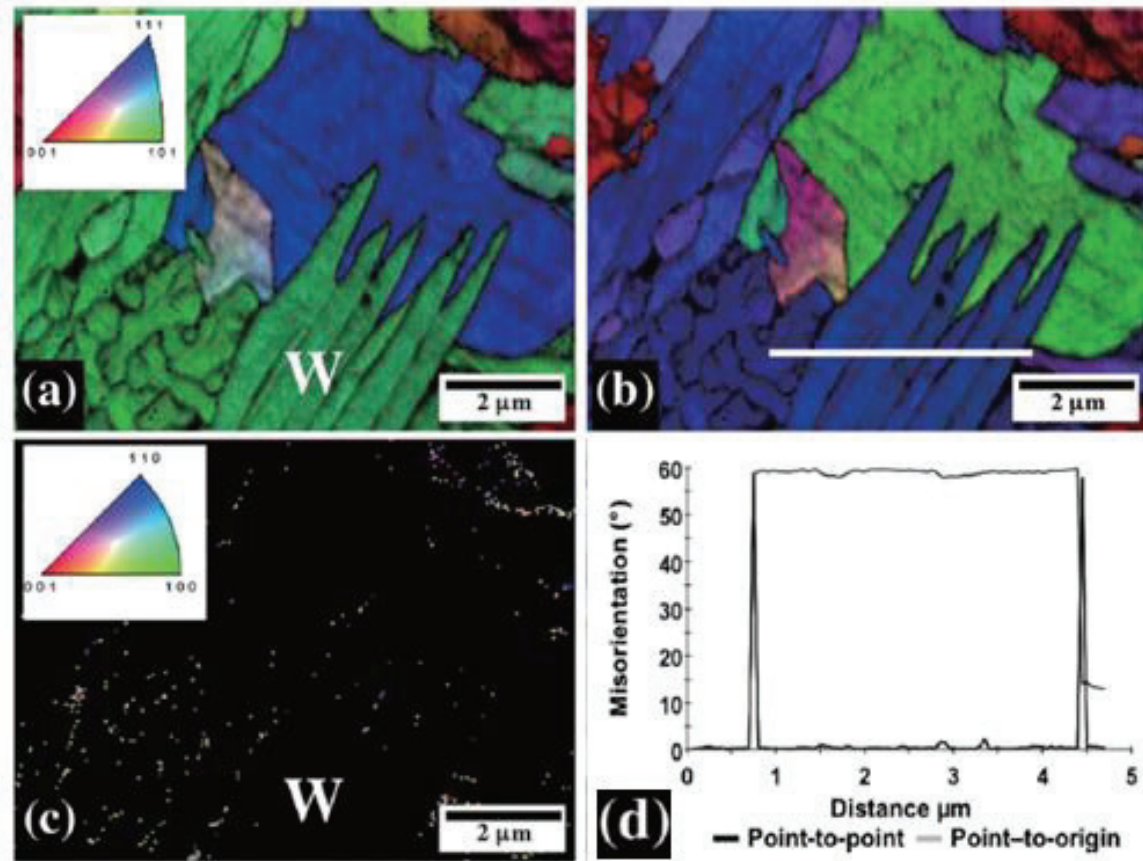


**Fig. 2.17.** EBSD maps of Widmanstätten ferrite plates taken from a 0.06C-0.01Si-0.3Mn-0.042Al steel sample deformed to a strain of 0.2 at 887 °C and  $0.4 \text{ s}^{-1}$ . (a,b) IPF and IQ maps of ferrite, (c) IPF map of martensite phase (black regions are ferrite), and (d) misorientation profile measured along the white line in (b). [2.43]

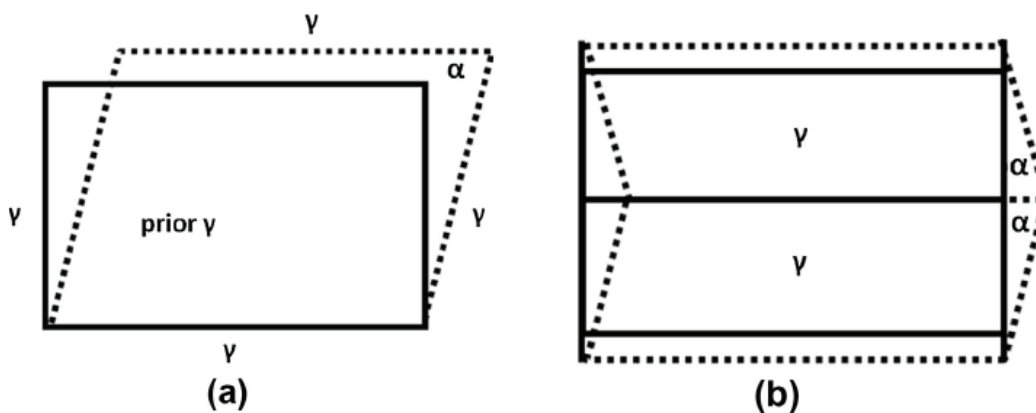


**Fig. 2.18.** (a) The application of a shear stress results in the formation of a parallelogram of Widmanstätten ferrite from the rectangle of prior austenite. (b) The geometry associated with the formation of self-accommodating plates. [2.48]

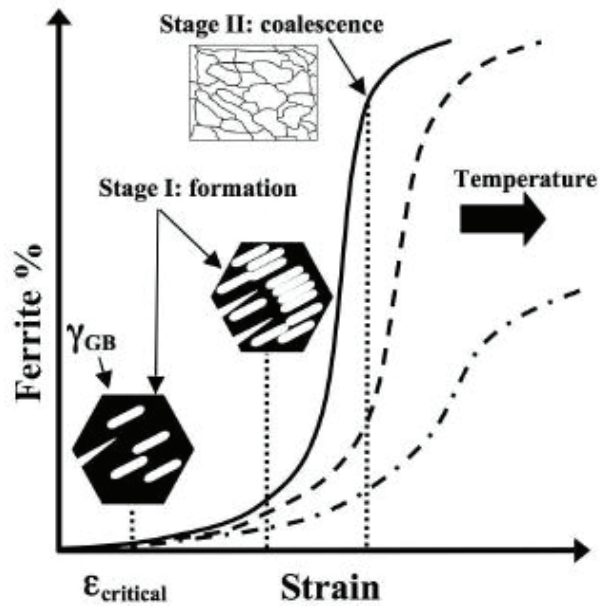


**Fig. 2.19.** The coalescence of Widmanstätten ferrite plates into polygonal grains during dynamic transformation [49].

0.5. The second stage involves coalescence of the plates into polygonal grains at strains greater than 0.5. This takes place both inside the grains and along the grain boundaries.

#### 2.4.4 Thermodynamics of Dynamic Transformation

Based on equilibrium thermodynamics, it is not possible to transform undeformed austenite into ferrite at temperatures above the  $A_{e3}$ . However, the application of strain can change the free energy of the austenite, thus resulting in the formation of metastable ferrite. Various researchers have published evidence for DT, as discussed above. So far, two types of thermodynamic models have been proposed to explain this unusual phenomenon; i) the stored energy [2.50] and ii) the stress activation models [2.48].

##### 2.4.4.1 Stored Energy Model

It has been shown that the application of strain shifts the  $A_{e3}$  temperature upwards and initiates the dynamic transformation of austenite to ferrite [2.50]. This phenomenon can be explained thermodynamically in terms of changes in the Gibbs energy. Deformation of the

austenite generates excess Gibbs energy because of the increase in the dislocation density. When this additional energy is taken into account, the Gibbs energy of the austenite becomes more positive (less stable), thus the nucleation of ferrite becomes possible.

The minimum Gibbs energy required to initiate the dynamic transformation of austenite to ferrite is given by:

$$\Delta G = \Delta G_{\gamma-\alpha} \quad (2.8)$$

where  $\Delta G_{\gamma-\alpha}$  is the chemical driving force opposing the transformation. This value is  $< 0$  below the  $A_{e3}$  but  $> 0$  above the  $A_{e3}$ . The positive value of the free energy difference means that the formation of ferrite is not possible and that an amount of energy is needed to initiate the transformation. When austenite is being deformed, the work done provides stored energy to the austenite, which increases the Gibbs energy of the system. This makes the austenite less stable. The modification can be expressed as:

$$\Delta G = \Delta G_{\gamma-\alpha} - \Delta G_{D(in \gamma)} \quad (2.9)$$

where  $\Delta G_{D(in \gamma)}$  is the stored energy from deformation and is  $> 0$ . When the  $\Delta G_{D(in \gamma)} > \Delta G_{\gamma-\alpha}$ , there is sufficient energy for the transformation to take place. The stored energy consists of three components: i) dislocation energy ( $\Delta G_{dislocations}$ ), ii) the increase in grain boundary surface energy ( $\Delta G_{grain \ boundary}$ ) and iii) vacancy formation energy ( $\Delta G_{vacancy}$ ). These components can be described by the following equations:

$$\Delta G_{D(in \gamma)} = \Delta G_{dislocation} + \Delta G_{grain \ boundary} + \Delta G_{vacancy} \quad (2.10)$$

$$\Delta G_{dislocation} = \mu_T b_T (\rho - \rho_0) / 2 \quad (2.11)$$

$$\Delta G_{grain \ boundary} = [(S_v^{gb} + S_v^{db}) - S_v^0] \times 0.8 \quad (2.12)$$

$$\Delta G_{vacancy} = Q_f \times C_s \quad (2.13)$$

where  $\mu_T$  is the temperature-dependent shear modulus,  $b_T$  is the magnitude of the Burgers vector at a particular temperature  $T$ ,  $\rho$  and  $\rho_0$  are the dislocation densities before and after deformation,  $S_v^0$  is the surface area per unit volume in fully recrystallized austenite,  $S_v^{gb}$  and  $S_v^{db}$  are the contributions from the grain boundaries and deformation bands,  $C_s$  is the excess vacancy concentration per unit volume, and  $Q_f$  is the vacancy formation energy in austenite.

Calculations showed that the dislocations provide the most significant contribution to the deformation energy [2.50]. The dislocation density can be measured experimentally but this value can also be estimated by using the equation:

$$\rho = (\sigma / (k_m \times M \times \mu_T \times b_T))^2 \quad (2.14)$$

where  $\sigma$  is the flow stress of the austenite,  $k_m$  is a constant taken as 0.20, and  $M$  is the Taylor factor. For austenite (FCC), the  $M$  value is 3.08 [2.51]. For temperatures above the  $Ae_3$ , the flow stress of the work hardened austenite can be employed in the calculations. However, for temperatures between the  $Ae_1$  and  $Ae_3$ , which corresponds to the two phase region, the method described in Ref 2.52 can be used [2.52].

The concept of stored energy to explain the occurrence of DT was first proposed by Hanlon and co-workers in 2001 [2.53]. They considered the dislocation density to be the driving force for the transformation. In this way, they were able to explain an increase in the  $Ae_3$  temperature of up to 10 °C. However, this approach cannot explain the dynamic transformations observed at temperatures well above the  $Ae_3$  [2.27]. More recently, Ghosh et al. extended this method by allowing for the inhomogeneous distribution of dislocations. They were able to increase the local deformation energy of the deformed austenite in this way to as much as 200 J/mol. This can explain the increase in the  $Ae_3$  temperature by up to 200 °C. Although the calculated values satisfied the experimental results, this approach needs the application of very large amounts of

strain. By contrast, the critical strain for dynamic transformation has been shown to be only in the range 0.10 to 0.20, depending on the alloying elements [2.54-2.57].

The stored energy model discussed above can be summarized using the diagram in Figure 2.20 [2.48]. The deformation of austenite above the  $A_{e3}$  temperature increases its free energy, which makes it less stable. When the chemical free energy of the austenite approaches that of the ferrite, the formation of metastable DT ferrite becomes favorable. However, upon isothermal holding, the DT ferrite retransforms into the more stable austenite phase.

#### 2.4.4.2 Stress Activation Model

The stored energy model was based on the increase in dislocation density produced during straining. The presence of dislocations increases the free energy of the austenite (at temperatures above the  $A_{e3}$ ), which makes the ferrite more stable. A limitation of this method, however, was that it required large amounts of strain far above the critical strains of 0.1 that have been observed to initiate the transformation [2.55,2.56]. This limitation was overcome by

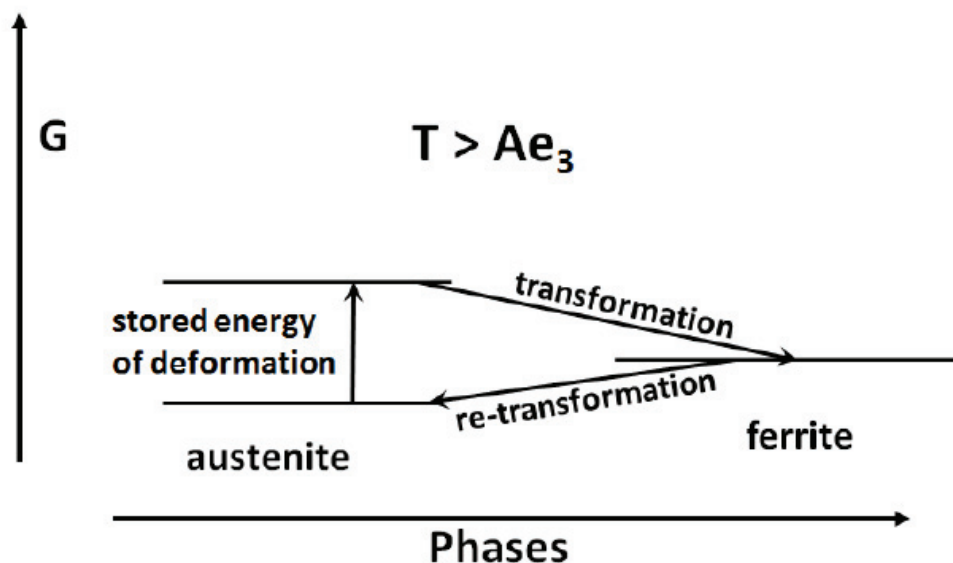


**Fig. 2.20.** Illustration of the changes in Gibbs energy responsible for dynamic transformation based on the stored energy model. Holding at temperature after deformation permits the re-transformation back to austenite [2.48].

Introducing the concept of the stress activation model, in which the applied stress acts as the driving force for the transformation [2.48]. This driving force is opposed by the energy obstacle, which consists of the chemical free energy difference between the austenite and ferrite as well as the lattice dilatation work and shear accommodation work that accompany the transformation. Note that there is a decrease in density associated with ferrite formation.

When Widmanstätten ferrite is being formed, the austenite undergoes a shear strain of 0.36 [2.45] and a dilatation of about 3% [2.4,2.58]. A diagram illustrating this was shown in Fig. 2.18 of section 2.4.3. The habit plane and shear direction of this transformation were also described in the previous section. The dilatation and shear accommodation work were first analyzed by Jonas and Ghosh in 2013 [2.48]. In this work, three possible configurations of habit plane normal (HPN), shear plane normal (SPN) and shear direction (SD) with respect to the applied stress were considered and analyzed. However, for simplicity, they used the configuration in Figure 2.21c, where all the axes are inclined at  $54.7^\circ$  with respect to the applied stress. The

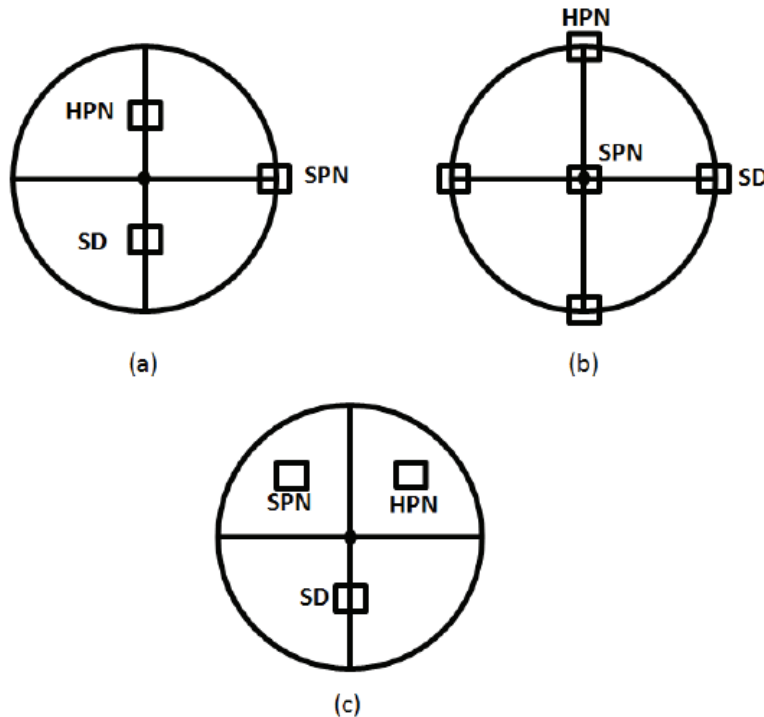


**Fig. 2.21.** Pole figures showing three possible configurations of habit plane normal (HPN), shear plane normal (SPN) and shear direction (SD) with respect to the applied stress. a) SPN perpendicular to applied stress, b) HPN and SD perpendicular to the applied stress and, c) all axes are inclined at  $54.7^\circ$  with respect to the applied stress. [2.48]

dilatation work per unit volume was estimated using the relation [2.48]:

$$W/V = \lambda \times \sigma_{0.1} \times \epsilon \quad (2.15)$$

where  $\sigma_{0.1}$  is the flow stress at the DT critical strain,  $\epsilon$  is the dilatation strain ( $\epsilon = 0.03$ ) and  $\lambda$  is the orientation factor ( $\lambda = \cos 54.7^\circ = 0.578$ ). The work done per unit volume in MPa ( $\text{MN}\cdot\text{m}\cdot\text{m}^{-3} = \text{MJ}\cdot\text{m}^{-3}$ ) was converted into J/mol using the conversion factor  $1 \text{ MPa} = 1 \text{ MJ m}^{-3} = 7.2 \text{ J/mol}$ . This conversion factor is dependent on the composition of the alloy. The relation between the calculated dilatation work per unit volume and temperature is illustrated in Figure 2.22 [2.48].

The shear accommodation work per unit volume in austenite during the formation of plates was calculated using the equation [2.48]:

$$W/V = \tau \times \gamma = m \times \sigma_{0.1} \times \gamma \quad (2.16)$$

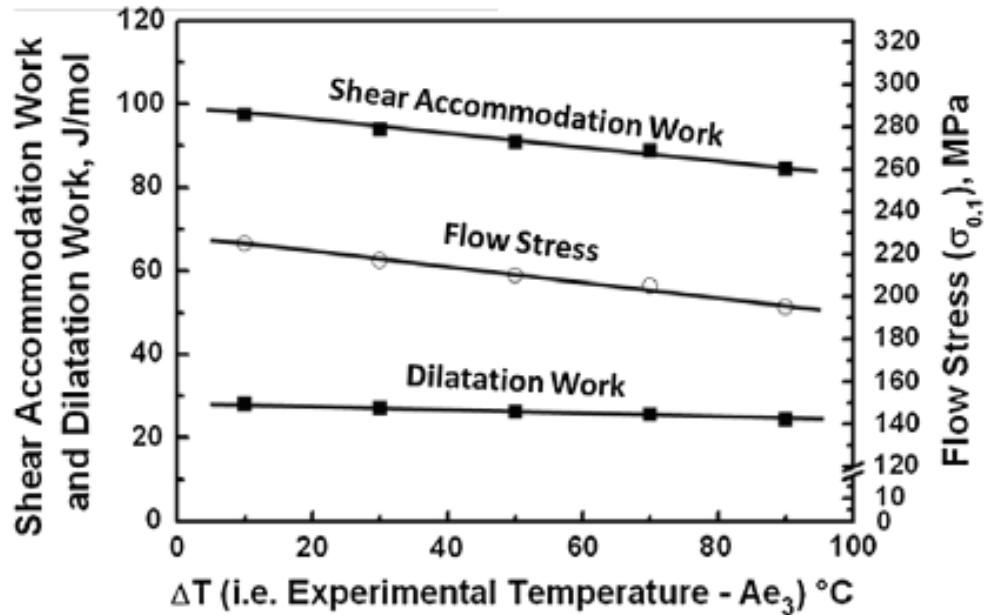


**Fig. 2.22.** Austenite flow stress at  $\epsilon = 0.10$  as a function of  $\Delta T$  (right-hand axis). The shear accommodation and dilatation work as a function of  $\Delta T$  (left-hand axis). [2.48]

where  $\sigma_{0.1}$  is the flow stress at the DT critical strain,  $\gamma$  is the average shear strain ( $\gamma = 0.18$ ) and  $m$  is the Schmid factor ( $m = 0.334$ ). These values were also converted into J/mol and are plotted in Figure 2.22 [2.48]. It was observed that the contribution of the shear accommodation work is significantly higher than that for the dilatation work.

The calculated total energy obstacle is illustrated in Figure 2.23 [2.48]. This consists of the sum of the chemical free energy difference between the austenite and ferrite and the dilatation work and shear accommodation work associated with the transformation. The flow stress at the critical strain acts as the driving force to overcome this obstacle. The changes in the free energies of the phases are summarized in Figure 2.24. This approach was able to account in a qualitative way for the dynamic transformations observed well above the  $A_{e3}$  temperature. However, it was not until the present investigations that a fully quantitative analysis of dynamic transformation, based on transformation softening, could be developed and proposed. The steps involved in attaining this objective will be described below.

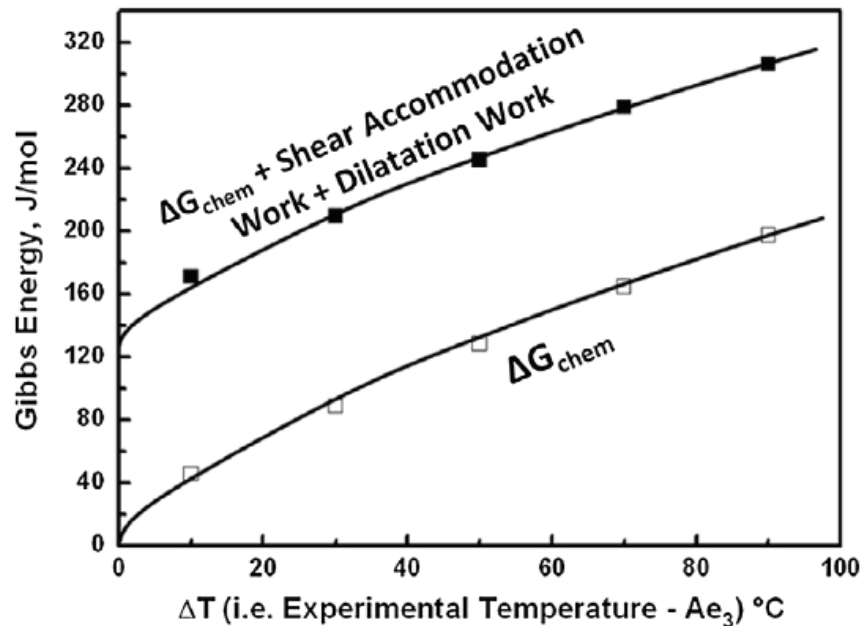


**Fig. 2.23.** The shear accommodation and dilatation work are added to the chemical free energy difference between austenite and ferrite as part of the total energy barrier to transformation. [2.48]

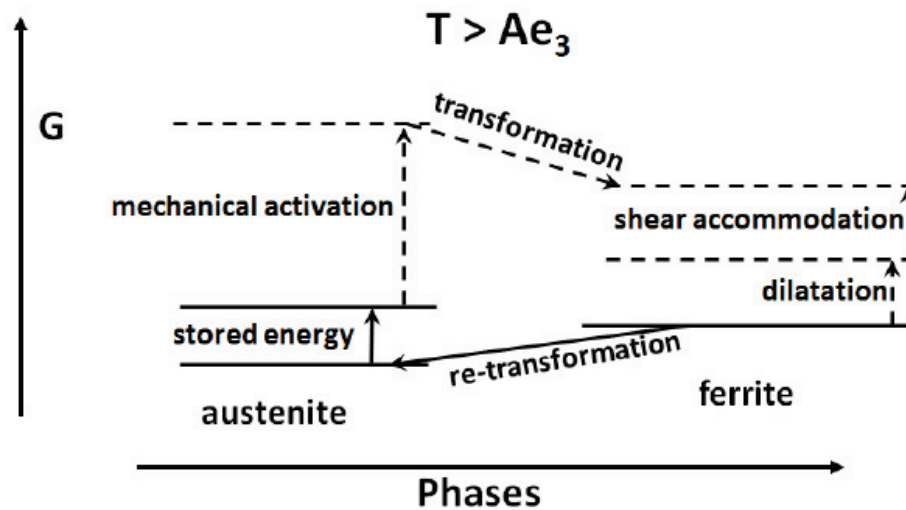


**Fig. 2.24:** Illustration of the changes in Gibbs energy responsible for dynamic transformation based on the mechanical activation model. Holding at temperature after deformation permits the re-transformation back into austenite [2.48].

#### 2.4.5 Early Strip Rolling Simulations

Previous studies have indicated that the interpass time has a significant effect on the rolling load [2.59-2.62]. Thus different separation forces may be developed at the same temperature, strain and strain rate, depending on the circumstances. The mean flow stresses developed during 30s plate rolling are illustrated in Figure 2.25, where they are compared with strip mill (2s) values. It is evident that the MFS values for the 30s interpass times are higher compared to the 2s times. In the literature, the effect of the interpass time was attributed to differences in the amount of recrystallization and of carbonitride precipitation. No consideration was given to the possible dynamic transformation of austenite to ferrite [2.63,2.64]. More recently, it has been suggested that dynamic transformation takes place during rolling. The critical strain required to induce dynamic transformation has been shown to be around 0.05, less than half the critical strain required to initiate dynamic recrystallization [2.61], see Figure 2.26. These values were calculated using the double differentiation method [2.61]. The continuous transformation of austenite to ferrite during rolling leads to softening and contributes to the observed low rate of MFS increase, as will be demonstrated in more detail below.

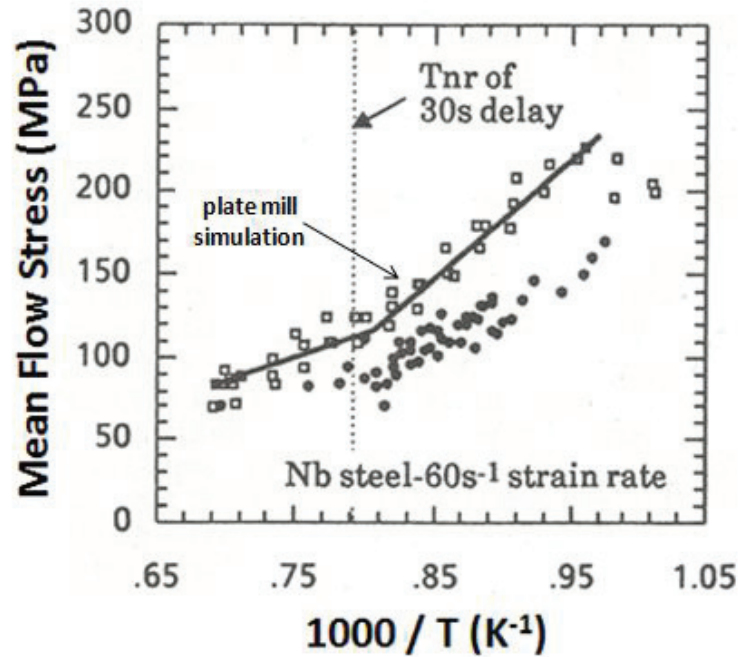


**Fig. 2.25:** Dependence of the MFS on inverse absolute temperature in a 0.1C-1.27Mn-0.2Si-0.11Nb-0.08V microalloyed steel. Here interpass times of 2s and 30s were employed together with pass strains of 0.5 applied at a strain rate of  $60\text{s}^{-1}$  [2.64].

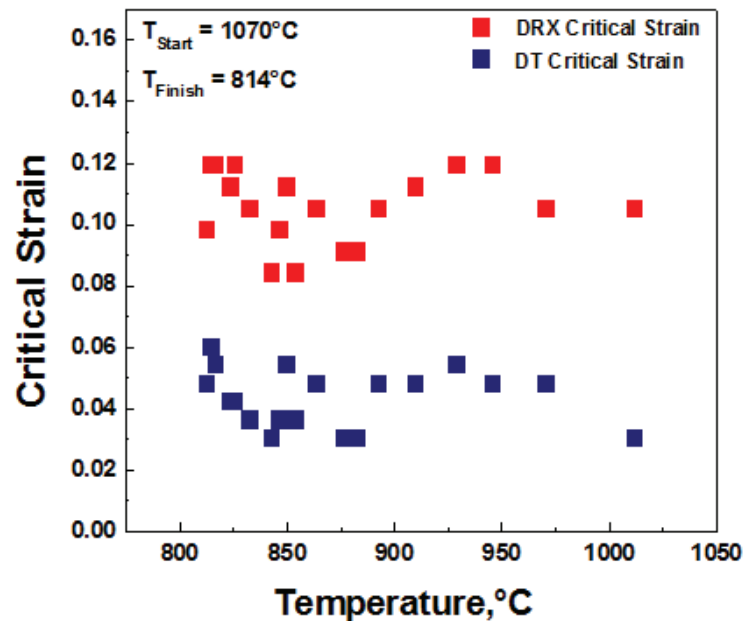


**Fig. 2.26:** Critical strains for the initiation of dynamic transformation (DT) and dynamic recrystallization (DRX) in torsion simulation experiments on a 0.79C-0.65Mn-0.24Si steel [2.61].

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## Chapter 3

### **Flow Softening-based Formation of Widmanstätten Ferrite in a 0.06%C Steel Deformed above the $A_{e3}$**

In this chapter, the concept of transformation softening during the dynamic transformation of austenite to ferrite is presented in detail. In this work, it was shown that the difference between the flow stress of austenite at the DT critical strain and the yield stress of the Widmanstätten ferrite that replaces it acts as the driving force that overcomes the energy barrier to the transformation. The obstacle consists of the chemical free energy difference between the austenite and ferrite, the lattice dilatation work and the shear accommodation work that accompany the transformation. Here, compression tests were carried out on a C-Mn steel at temperatures above the  $A_{e3}$  to observe the occurrence of DT. The results are interpreted in terms of the transformation softening model by calculating the magnitudes of both the driving force and the energy obstacles.

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### **Abstract**

Compression tests were carried out at a strain rate of  $1s^{-1}$  on a 0.06%C-0.3%Mn-0.01%Si steel over two temperature ranges: i) 920 °C to 980 °C, and ii) 500 to 750 °C. Optical and scanning electron microscopy images indicated that significant volume fractions of Widmanstätten ferrite were formed dynamically above the  $A_{e3}$  temperature. The ferrite plates coalesced into polygonal grains during straining. The double differentiation method was applied to the stress-strain curves, providing average values for the dynamic transformation (DT) and dynamic recrystallization (DRX) critical strains of 0.12 and 0.20, respectively. These results are interpreted in terms of the flow softening-based transformation model by calculating both the driving forces promoting the transformation as well as the energy barriers that oppose it. The model predicts the temperature range over which DT can occur as well as the observed critical strains.

### 3.1 Introduction

Yada and co-workers showed in the 1980's that austenite can be transformed into ferrite by deformation at temperatures above the  $A_{e3}$ . By means of rolling simulations and compression tests, they demonstrated that such dynamic transformation (DT) can take place at temperatures as much as 166 °C above the  $A_{e3}$  [3.1,3.2]. By applying accumulated strains of as much as 3 to 4, they were able to produce ferrite volume fractions of 70% to 80% with grain sizes in the range 1 $\mu$ m to 2 $\mu$ m. In a follow-up study, they provided real time evidence for DT by carrying out in-situ X-ray diffraction experiments in which ferrite peaks appeared during hot deformation in the austenite region [3.3]. Further experimental evidence for DT was reported during the decade that followed [3.4-3.5], accompanied by Monte Carlo simulations [3.7,3.8] in which it was shown that DT is generally expected to precede dynamic recrystallization (DRX) under finish rolling conditions.

This unusual phenomenon was initially justified in terms of the stored energy of the dislocations introduced into the deformed austenite. For example, Sun et al. [3.9] arrived at a free energy increase of 22.3 J/mol. Hanlon et al. [3.10] converted such energies into  $A_{e3}$  temperature increases of up to 10 °C, values that are too low given Yada's observations regarding the increase in the transformation temperature [3.1,3.2].

For this reason, Ghosh et al. [3.11] extended Hanlon's analysis by allowing for the dislocation density distribution to be inhomogeneous, increasing the driving force in this way to as much as 197 J/mol. However, this method is based on the application of very large strains, significantly greater than the critical strains of about 0.10 that have been reported to be associated with the initiation of the transformation [3.12,3.13]. This limitation was later resolved by introducing the concept of mechanical or stress activation [3.14]. According to this view, the flow stress developed at the critical strain acts as the driving force that overcomes the various obstacles to the transformation. In this way, the model was able to account for the rapid forward displacive transformation [3.11,3.12], which stands in sharp contrast to the much slower, diffusional reverse transformation.

In the present work, the driving force is taken instead as the *net softening* associated with the transformation [3.15]. This is defined as the difference between the austenite flow stress at the critical strain and the yield stress of the fresh Widmanstätten ferrite that replaces it. In this way, the driving force consists only of the *softening* that takes place during the transformation. This force is opposed by the sum of the chemical free energy difference between the austenite and the Widmanstätten ferrite and the dilatation work and shear accommodation work that are associated with the process.

For this purpose, compression tests were carried out on a 0.06wt%C-0.30wt%Mn-0.01wt%Si steel and the driving force required to initiate DT was determined experimentally from the difference between the work hardened austenite flow stress and the yield stress of the Widmanstätten ferrite projected into the same temperature range. The latter was estimated from that of polygonal ferrite specimens deformed below the  $A_{e1}$ , over the range 500 to 750 °C. The polygonal ferrite values were in turn corrected for the difference in flow stress between Widmanstätten and polygonal ferrite, as determined by means of hardness tests carried out at room temperature [3.15]. In this way, the critical strains predicted by the model could be compared with the experimental values obtained by means of the double differentiation method [3.16], described in more detail below.

## **3.2 Experimental Procedure**

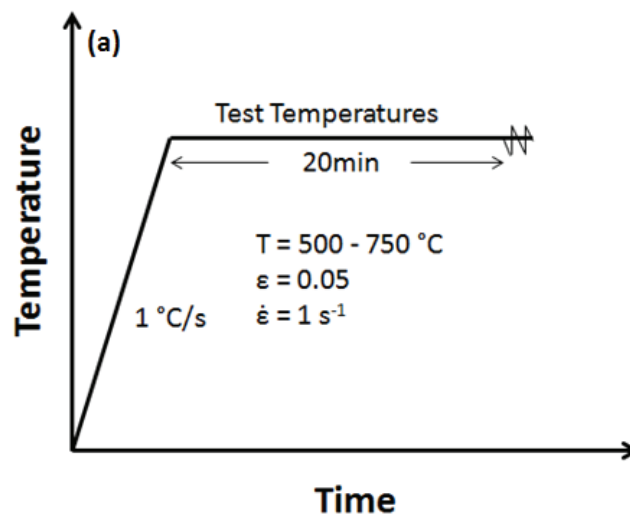
### **3.2.1 Compression Testing**

A 0.06wt%C steel containing 0.01 wt% Si and 0.30 wt% Mn was investigated in the current study. This was received in the form of hot rolled plates. The complete chemical composition, including the corresponding paraequilibrium and orthoequilibrium  $A_{e3}$  and  $A_{e1}$  temperatures, is listed in Table 3.1. The transformation temperatures were calculated using the FSstel database of the FactSage thermodynamic software [3.17]. All the analyses that follow are based on paraequilibrium conditions, under which only the interstitials (in this case C) are able to partition between the product and parent phases [3.18].

**Table 3.1.** Chemical composition (mass%) and equilibrium transformation temperatures ( $^{\circ}\text{C}$ )

| C    | Mn   | Si   | $A_{e1}$               | Paraequilibrium $A_{e3}$ | Orthoequilibrium $A_{e3}$ |
|------|------|------|------------------------|--------------------------|---------------------------|
| 0.06 | 0.30 | 0.01 | 724 $^{\circ}\text{C}$ | 870 $^{\circ}\text{C}$   | 877 $^{\circ}\text{C}$    |

Cylindrical compression specimens, 6 mm in diameter and 9 mm long were machined from the hot rolled plates with their axes parallel to the rolling direction. Hot compression tests were conducted using a 100kN MTS servo-hydraulic testing machine equipped with a vertical radiation furnace and a temperature controller. Two sets of experiments were carried out. In the first, the yield stress of the specimens was established in the ferrite region below the  $A_{e1}$  temperature (Figure 3.1a): in the second, the flow stress of the austenite phase was measured above the  $A_{e3}$  (Figure 3.1b). In the experiments of Figure 3.1a, the samples were heated at the rate of 1  $^{\circ}\text{C/s}$  up to test temperatures in the range 500  $^{\circ}\text{C}$  to 750  $^{\circ}\text{C}$ ; they were then held for 20 min prior to deformation.



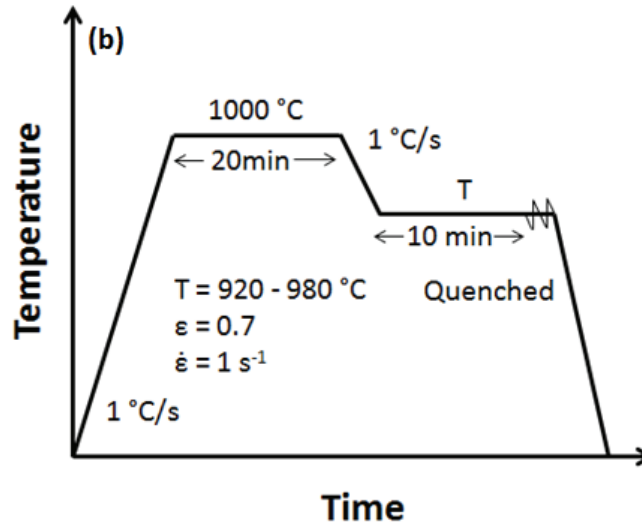


**Fig. 3.1.** Compression test schedules employed to determine a) the yield stress of ferrite below the  $A_{e1}$  and b) the flow stress of work hardened austenite above the  $A_{e3}$ .

The same heating rate was employed in the tests of Figure 3.1b, in which case the specimens were heated up to 1000 °C and then held for 20 minutes to ensure that the transformation to austenite was complete. The samples were then cooled at 1 °C/s down to test temperatures in range 920 °C to 980 °C and held for 10 minutes before straining. In this case, much larger strains of 0.7 were applied at the same strain rate. Prior to testing, the top and bottom surfaces of the specimens were covered with Mo sheets and lubricated by applying a mixture of boron nitride and ethanol. In this way, relatively homogeneous deformation was achieved, as shown in Figure 3.2. In order to minimize oxidation and decarburization during testing, an argon protective atmosphere was employed in all cases. The compression samples were rapidly water quenched (in about 1s) after testing so as to preserve the deformation microstructure. As static (post deformation) transformations involve alloying element partitioning by substitutional diffusion, such quenches are considered sufficiently fast to prevent the formation of polygonal ferrite during cooling [3.18].

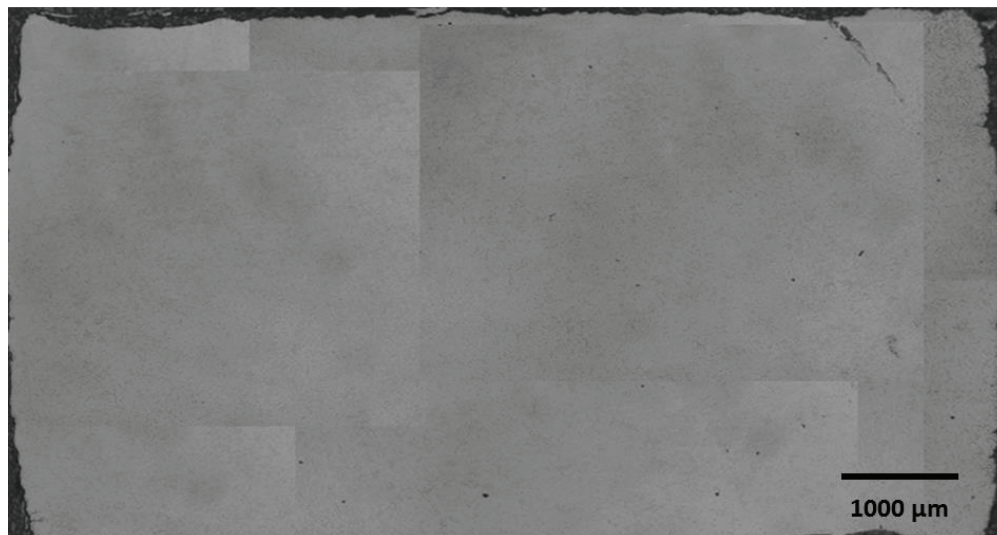


**Fig. 3.2.** Optical micrograph showing the homogeneous nature of the deformation in a sample compressed to a strain of 0.7 at  $1s^{-1}$  and 980 °C.

### 3.2.2 Metallography

Cross-sections containing the longitudinal axis were cut from the deformed compression samples. These were then mounted using a phenolic hot mounting resin. Rough polishing was carried out up to 1200 grit SiC paper. Both 3μm and 1 μm diamond suspensions were used for final polishing before etching with 2% nital to reveal the microstructure for optical and scanning electron microscopy. The volume fractions of dynamically formed ferrite were estimated from the optical micrographs by means of area measurement using the ImageJ software.

## 3.3 Results

### 3.3.1 Stress-Strain Curves

Some typical true stress - true strain curves determined in the austenite phase field are shown in Figure 3.3. These samples were deformed as much as 110 °C above the paraequilibrium  $A_{e3}$  temperature. The stress levels of the curves increase with decrease in temperature in the normal way. Here there is softening after the peak stress is attained, leading to a drop of around 10% in the stress. Such softening involves both dynamic transformation as well as dynamic recrystallization, as discussed elsewhere [3.19]. The sample temperatures measured

during deformation are also shown in this diagram and can be read on the right hand axis. It can be seen that these temperatures remained substantially constant during the tests carried out over the range from 920 °C to 980 °C. These temperatures are 50 °C to 100 °C above the paraequilibrium  $A_{e3}$ . Note that the work done per unit volume can be deduced from the experimental stress-strain curves. When converted into temperature increases, these lead to maximum increments of only 2-3 °C at the critical strain. Thus transformation was essentially initiated at the temperatures shown.

In order to identify the moments of initiation of these two softening mechanisms, the double differentiation method [3.16,3.19] was used. For this purpose, the flow curves of Figure 3.3 were fitted with 8<sup>th</sup> order polynomials using the MatLab software. In order to determine the values of the critical stresses associated with these mechanisms,  $\theta$  versus  $\sigma$  curves were prepared, as shown in Figure 3.4a, where  $\theta$  is the strain hardening rate ( $\delta\sigma / \delta\epsilon$ ). The moment

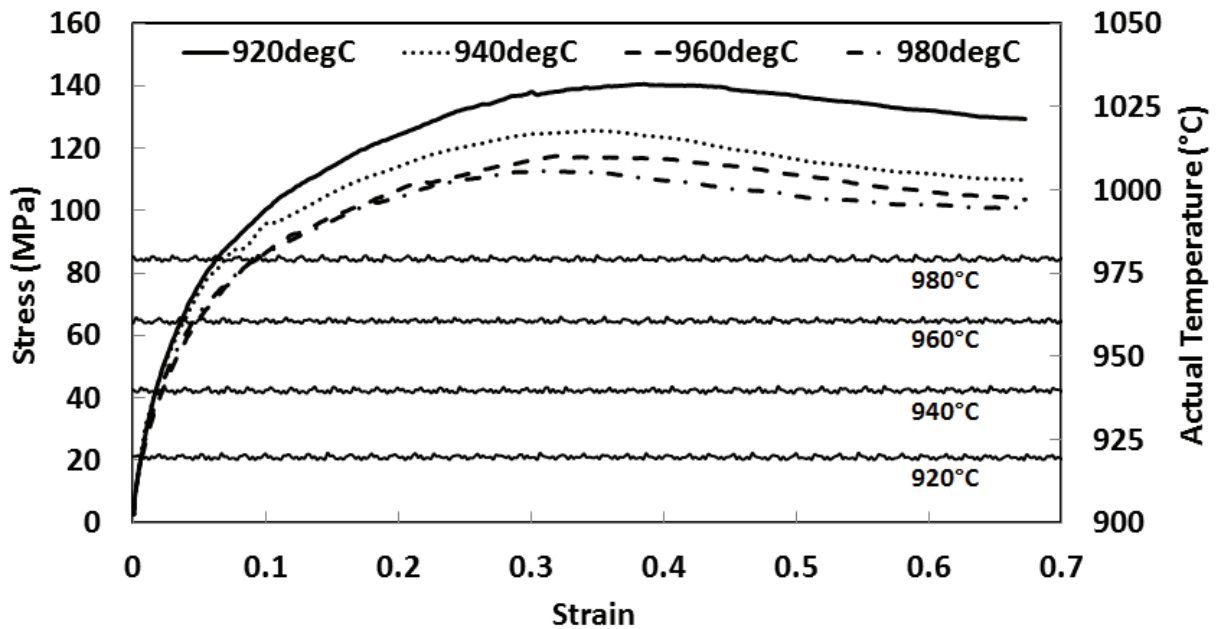


**Fig. 3.3.** Stress-strain curves of the 0.06wt%C-0.3wt%Mn-0.01wt%Si steel compressed to a strain of 0.7 at  $1s^{-1}$ . Stress drops of approximately 10% are observed after the stress peaks. The horizontal lines display the temperature of the sample (on the right hand axis) during deformation.

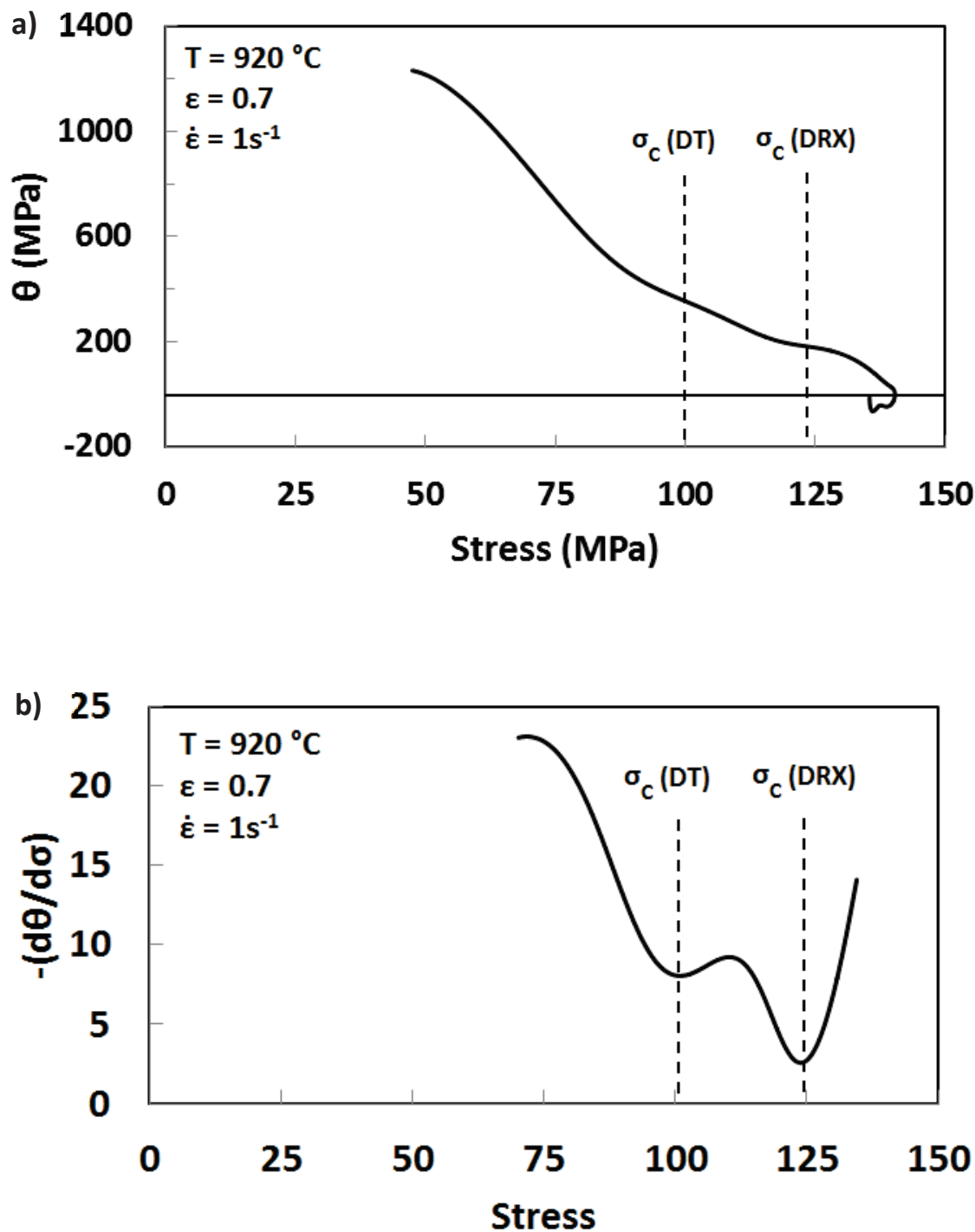


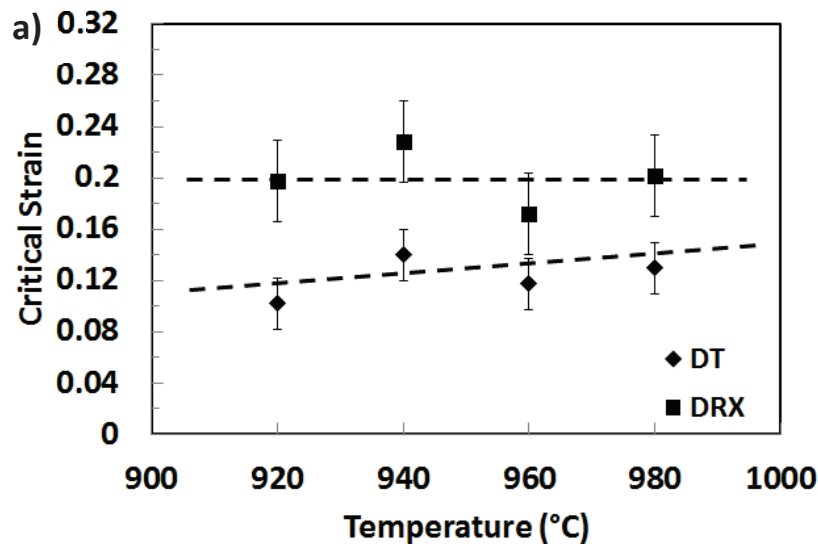
**Fig. 3.4.** a) Plot of strain hardening rate ( $\theta$ ) versus  $\sigma$  derived from the stress-strain curve of the compression test carried out at  $920\text{ }^{\circ}\text{C}$  and  $1\text{ s}^{-1}$ . Each inflection point corresponds to the initiation of a particular softening mechanism. b) The inflection points in a) are identified more clearly by the minima in the  $-(d\theta/d\sigma)$  versus  $\sigma$  curve.

of initiation of each mechanism is identified by a particular inflection point in the  $\theta$ - $\sigma$  plot and corresponds to the condition:

$$\frac{\delta}{\delta\sigma} \left( \frac{\delta\theta}{\delta\sigma} \right) = 0 \quad (3.1)$$

The two minima in the  $-\delta\theta/\delta\sigma$  versus  $\sigma$  curve of Figure 3.4b identify the onsets of first DT and then DRX. As shown earlier by means of microstructural analysis [3.12,3.13], the critical strain for DT is significantly lower than that for DRX.

The dependences on temperature of the critical strains are illustrated in Figure 3.5a. Here the average critical strain for DT is about 0.12 while that for DRX is roughly 0.20. The critical stresses that correspond to these strains are displayed in Figure 3.5b. Here it can be seen that the DRX stress depends on temperature in the manner expected for a thermally activated mechanism. By contrast, DT, being a displacive and athermal process, displays no obvious temperature dependence. The critical strains and stresses of Figure 3.5a will be employed below in order to calculate the mechanical driving force for the initiation of DT.



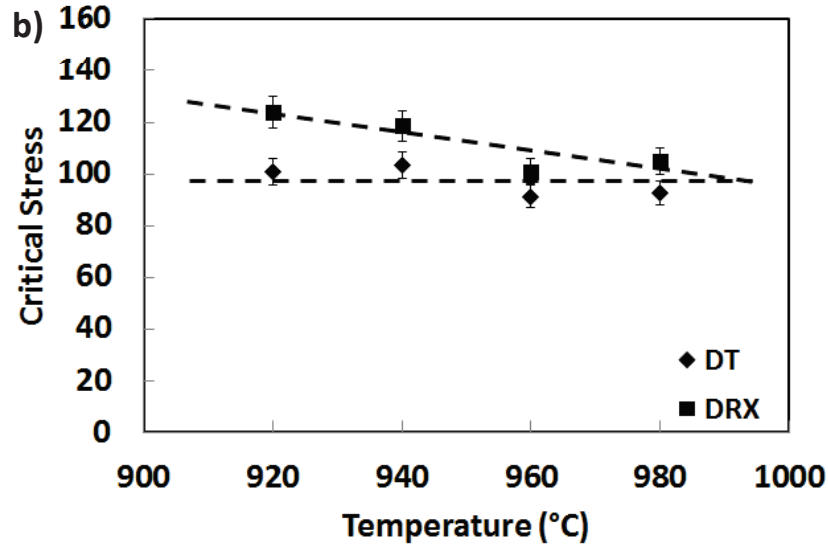


**Fig. 3.5.** a) The critical strains for dynamic transformation (DT) and dynamic recrystallization (DRX) determined by the double differentiation method. b) The critical stresses associated with the critical strains of Figure 3.4a. Those of DRX are temperature dependent while those of DT show no dependence.

### 3.3.2 Microstructure

It was shown earlier [3.6] that the volume fraction of dynamically formed ferrite increases with strain and time and decreases with the temperature interval above the  $A_{e3}$ . This phenomenon involves the displacive/diffusional transformation of austenite to ferrite. The *nucleation* of Widmanstätten ferrite takes place by a displacive mechanism [3.20] characterized by an invariant-plane strain with a large shear component [3.21]. However, the *growth* of the plates involves the diffusion of carbon [3.20], but not of the substitutional alloying elements. Because of these characteristics, the overall process can be described as a paraequilibrium displacive transformation [3.21].

Some typical optical micrographs illustrating this phenomenon in the current material are presented in Figure 3.6. The martensite (retained austenite) phase appears dark while the ferrite is light. The microstructure of the material deformed at 980 °C (110 °C above the  $A_{e3}$ ) to a strain of 0.7 (Figure 3.6a) contains approximately 40% ferrite. The same strain and strain rate were used to produce the microstructure of Figure 3.6b but straining was carried out at 960 °C

(i.e. 90 °C above the  $A_{e3}$ ). In this case, there is around 60% of ferrite present, illustrating the temperature dependence of ferrite formation. As will be seen in more detail below, the driving force *increases* with decreasing temperature, while the obstacle free energy barrier *decreases* as the  $A_{e3}$  temperature is approached. Thus both trends contribute to the increased formation of Widmanstätten ferrite as the temperature interval above the  $A_{e3}$  is reduced.

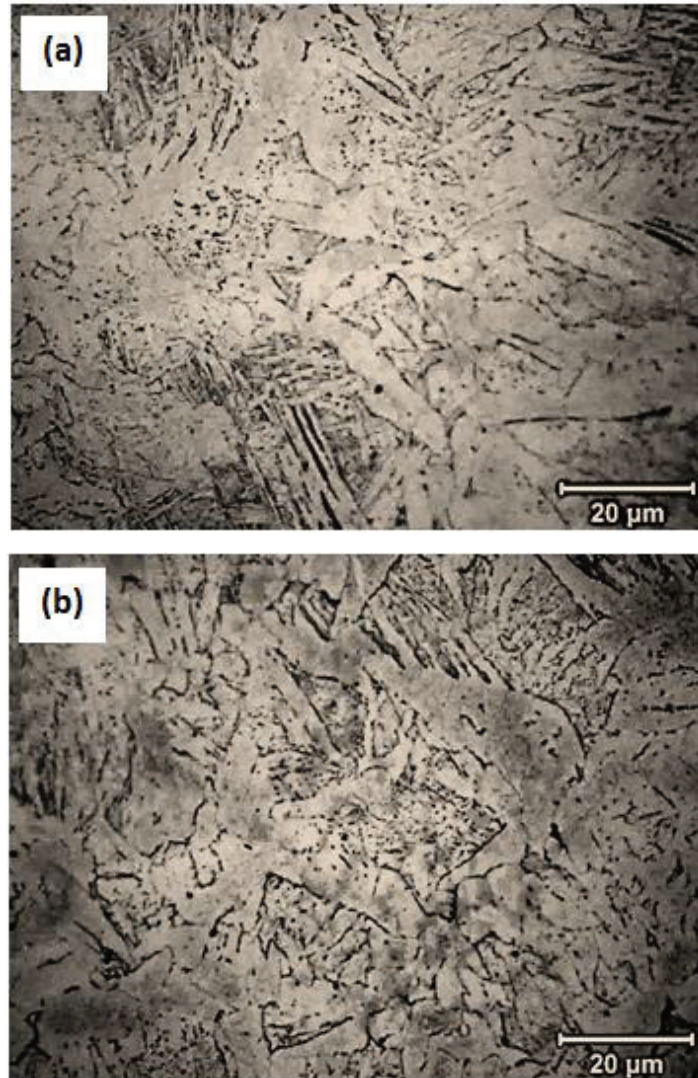


**Fig. 3.6.** Optical micrographs of the 0.06wt%C-0.3wt%Mn-0.01wt%Si steel deformed to a strain of 0.7 at a strain rate of  $1s^{-1}$  at a) 980 °C and b) 960 °C. The volume fraction of dynamically formed ferrite is higher at the lower temperature.

The appearance of the displacive ferrite plates that form during straining is illustrated in Figure 3.7. These are approximately 200nm in width [3.22] and gradually coalesce into polygonal ferrite grains during straining [3.22]. The Widmanstätten ferrite produced in this way differs significantly from that formed conventionally below the  $A_{e3}$ . Here dynamic plate formation *precedes* the appearance of polygonal ferrite, whereas static Widmanstätten is normally nucleated on the allotriomorphic ferrite that appeared first. The Widmanstätten ferrite plates produced by mechanical activation form in self-accommodating pairs of near-identical orientation [3.23].

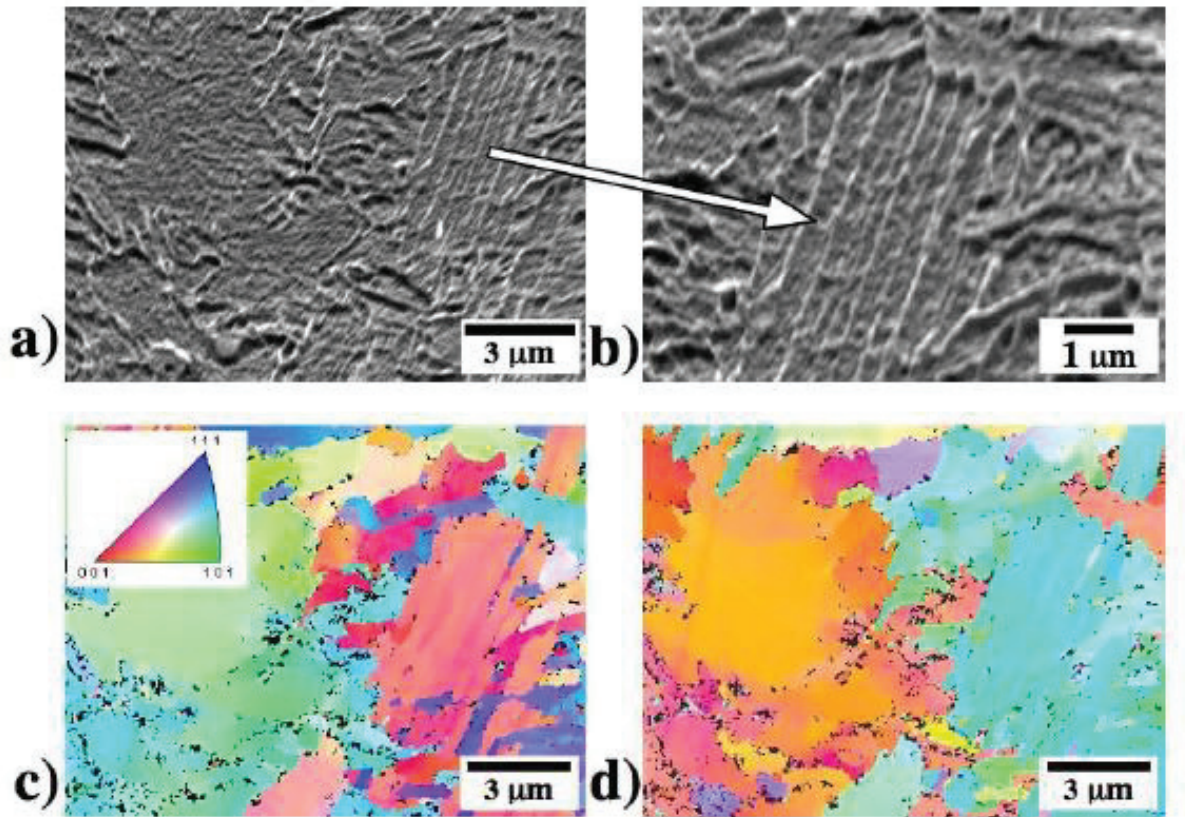
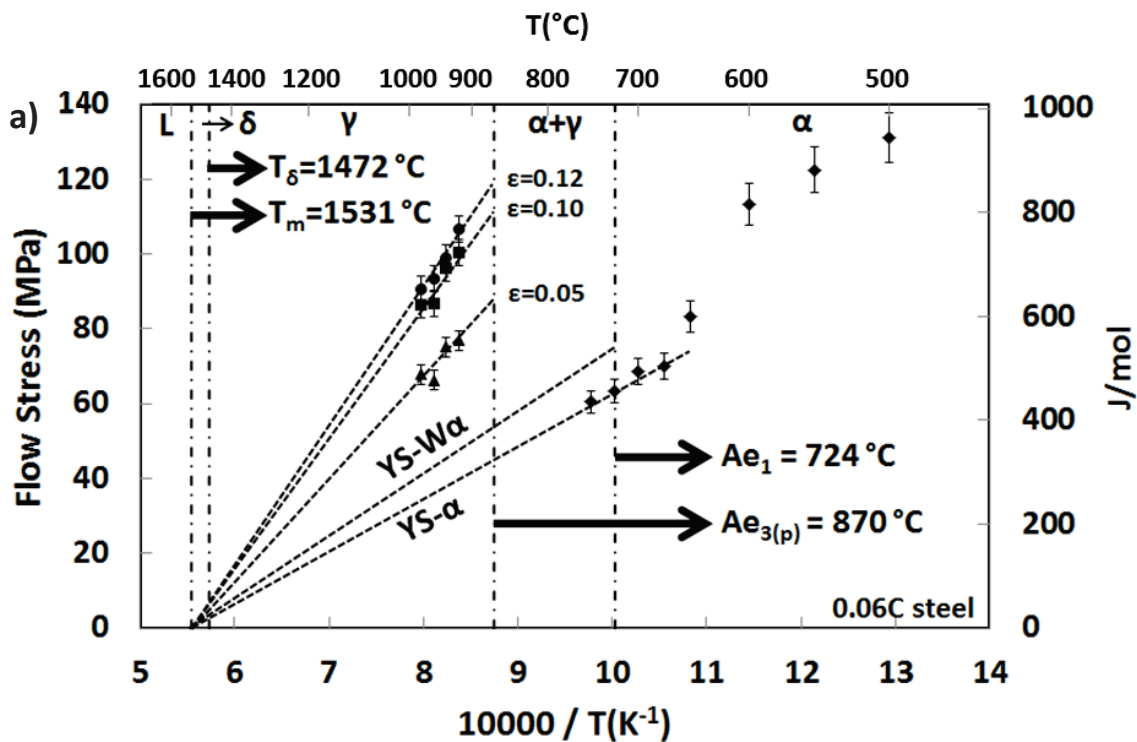


**Fig. 3.7.** EBSD images of 0.09%C steel strained to 0.5 at 856 °C (20 °C above the  $A_{e3}$ ). a) and b) secondary electron images of Widmanstätten ferrite plates, c) orientation image map of the ferrite phase, [001] is along the longitudinal axis, and d) orientation image map of ferrite after 90° rotation around [010]. [3.22]

### 3.4 Discussion

#### 3.4.1 Mechanical Driving Force

In what follows, the driving force is evaluated as the difference between the flow stress of the austenite work hardened up to the critical strain and the yield stress of the fresh Widmanstätten ferrite that takes its place. As a first step, the yield stress of polygonal ferrite (YS- $\alpha$ ) at temperatures below the  $Ae_1$  was taken from the experimental curves (not shown here) by applying the 0.2% offset rule. These values are plotted against inverse absolute temperature in Figures 3.8a and 3.8b. The flow stresses of the work hardened austenite at strains of 0.05, 0.10 and 0.12 are displayed in Figure 3.8a, while those for strains of 0.002, 0.05, 0.10 and 0.20 are illustrated in Figure 3.8b. The various transformation temperatures are also indicated in these diagrams.



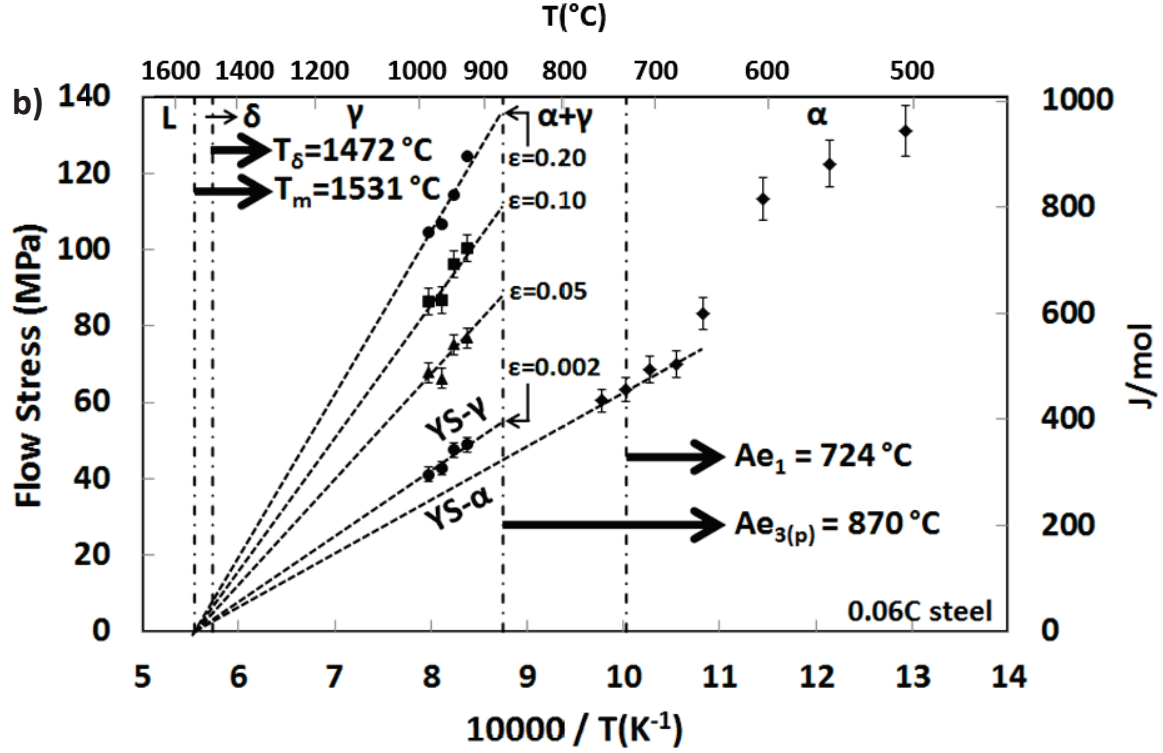


**Fig. 3.8.** a) Dependence of the austenite flow stress on inverse absolute temperature. The net softening associated with DT is evaluated by subtracting the yield stress of Widmanstätten ferrite (YS-W $\alpha$ ) from the flow stress of the austenite work hardened to the DT critical strain of 0.12. b) The driving force for DRX can be estimated from the difference between the flow stress of the work hardened austenite at the DRX critical strain of 0.20 and the yield stress of undeformed austenite grains (YS- $\gamma$ ).

The linear dependence on inverse temperature is based on the so-called exponential stress law for creep [3.24]:

$$\dot{\epsilon} = B \exp(C\sigma) \exp\left(\frac{-Q}{RT}\right) \quad (3.2)$$

which can also be expressed as

$$\sigma = \frac{\ln(\dot{\epsilon}/B)}{C} + \frac{Q}{CR} \left(\frac{1}{T}\right) \quad (3.3)$$

where B and C are material constants, R is the universal gas constant and Q is the activation energy of flow. This kind of linear relationship has been observed to apply to numerous steels [3.25] and is commonly used to determine the  $T_{nr}$  in rolling mills.

In order to estimate the yield stress of Widmanstätten ferrite at temperatures above the  $A_{e3}$ , two assumptions were made here. The first is that the flow stress of the displacive ferrite is also linearly dependent on inverse temperature. The second is that this stress is 18% higher than that of conventional ferrite, an estimate that is based on hardness measurements carried out in an earlier investigation [3.15].

Once the various flow stresses displayed in Figure 3.8a are available, the mechanical driving force for DT can be deduced by subtracting the Widmanstätten yield stress ( $YS-W\alpha$ ) from the austenite flow stress at a strain of 0.12 (the critical strain for DT). This difference is highest, about 480 J/mol, at temperatures close to the paraequilibrium  $A_{e3}$  and decreases linearly as the inverse temperature is decreased. The driving force per unit volume,  $(W/V)_{\text{driving force}}$ , also varies with grain orientation, as it depends on the Schmid factor ( $m$ ), as given by [3.14]:

$$(W/V)_{\text{driving force}} = \Delta\text{Flow Stress} \times m \quad (3.4)$$

Here  $m$  is the Schmid factor and  $\Delta\text{Flow Stress}$  is the difference between the flow stress of work hardened austenite at a strain of 0.12 and the yield stress of the fresh Widmanstätten ferrite. The flow stresses are converted into J/mol (1MPa=7.2J/mol) for comparison with the energy obstacles introduced in the next section.

An interesting application of the present approach is that the mechanical driving force for DRX can also be deduced in this way. Although DRX is diffusion-controlled and therefore thermally activated, it is the stored energy of the dislocations introduced by straining that provides the driving force. This can be estimated from the net softening resulting from the motion of a grain boundary while it is annihilating dislocations. This has the effect of reducing the local flow stress from its current work hardened value down to the dislocation-free yield stress. It is accordingly given by the vertical distance between the flow stress at  $\epsilon = 0.20$  in Figure 3.8b (the

critical strain for DRX) and the austenite yield stress. Here it falls in the range 320 to 60 J/mol, decreasing with increasing temperature. By contrast, DT is an entirely athermally-activated phenomenon; the mechanical driving force of eq. 3.4 will now be compared with the energy barriers opposing the transformation.

### 3.4.2 Free Energy Obstacles to Transformation

The free energy obstacle to the formation of ferrite at temperatures above the  $A_{e3}$  ( $\Delta G_{\gamma-\alpha}$ ) is illustrated in Figure 3.9 (solid line). At room temperature, Widmanstätten ferrite, a non-equilibrium phase, has a stored energy of around 50 J/mol [3.26] when compared with conventional ferrite. Here it is assumed that, at the high temperatures of interest in this work, namely 870-1020 °C, this difference decreases to approximately 15 J/mol (at 870 °C), declining further to 5 J/mol at 1020 °C. This additional free energy gap is added to  $\Delta G_{\gamma-\alpha}$  in Figure 3.9 to provide an estimate of the free energy barrier opposing the formation of displacive ferrite ( $\Delta G_{\gamma-w\alpha}$ ) in this temperature range, shown as a broken line.

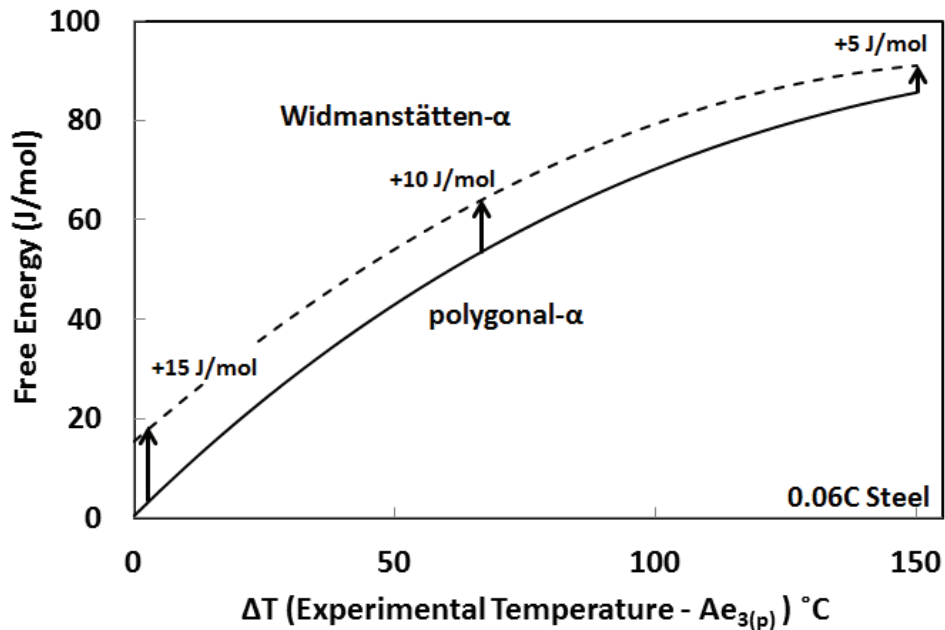


**Fig. 3.9.** Chemical free energies of polygonal (solid line) and Widmanstätten (broken line) ferrite as a function of the temperature interval  $\Delta T$  (experimental temperature –  $A_{e3(p)}$ ). The free energy barrier increases with  $\Delta T$  because of the increasing instability of ferrite at elevated temperatures.

The transformation also requires the expenditure of work to accommodate shearing of the plates as well as the dilatation that accompanies ferrite formation [3.14]. A shear strain of  $\gamma = 0.36$  is associated with the transformation of austenite into Widmanstätten ferrite [3.21,3.27], accompanied by a dilatation of about 3%, perpendicular to the habit plane [3.27]. In order to calculate the work done, the transformation habit plane was taken to be (0.506, 0.452, 0.735) and the shear direction to be (-0.867, 0.414, 0.277) [3.21,3.27]. These values were employed to evaluate the Schmid and orientation factors,  $m$  and  $\lambda$ , required in these calculations. The values of the shear accommodation,  $(W/V)_{SA}$ , and dilatation,  $(W/V)_D$ , work per unit volume were then calculated using the following relations [3.14]:

$$(W/V)_{SA} = m \times \sigma_{0.12} \times 0.36 \quad (3.5)$$

$$(W/V)_D = \lambda \times \sigma_{0.12} \times 0.03 \quad (3.6)$$

Here  $\sigma_{0.12}$  is the flow stress at the DT critical strain of 0.12.

The Schmid factors governing the formation of Widmanstätten ferrite were calculated using the method of Ref [3.28] applied to the transformation habit plane and shear direction specified above. These are displayed in inverse pole figure (IPF) form in Figure 3.10. Here the maximum

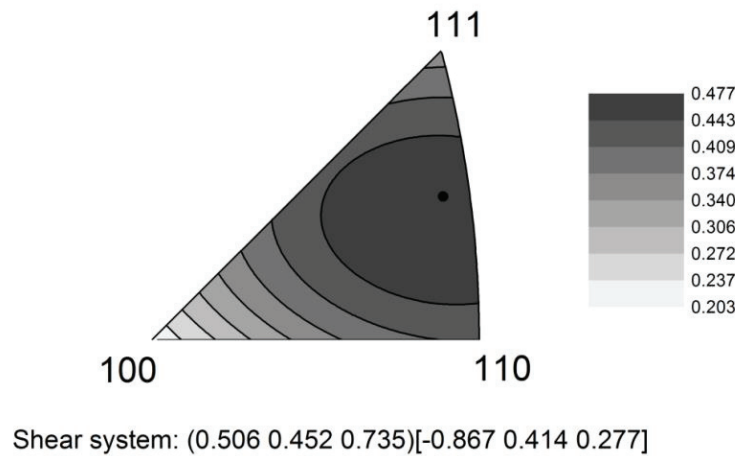


**Figure 3.10.** The Schmid factors pertaining to transformation on the (0.506 0.452 0.735) habit plane and along the [-0.867 0.414 0.277] shear direction displayed in inverse pole figure (IPF) form.

and minimum  $m$ -values are 0.477 and 0.203, respectively. Note that the habit plane normal and the shear vector are not orthogonal because of the rigid body rotation and dilatation associated with the present type of displacive transformation.

The dependence of the flow stress on temperature interval above the  $A_{e3(p)}$  (required to solve eqs. 3.5 and 3.6) is illustrated in Figure 3.11 (solid line). The values of the shear accommodation work (broken line) and dilatation work (dotted line) obtained from these equations are also displayed in Figure 3.11 for  $m$  values of 0.5 and 0.2 and orientation factors,  $\lambda = \sqrt{m}$ , of 0.707 and 0.447. The former quantities represent the maximum amount of work that can be done by the applied stress in an optimally oriented grain ( $m = 0.5$ ;  $\lambda = 0.707$ ). The lowest possible Schmid factor as shown above is 0.203. Here  $m = 0.2$  is used to calculate the work performed in the least well-oriented grain. These values are used below to estimate the extent to which DT can occur at a strain of 0.12 in grains of specific orientations.

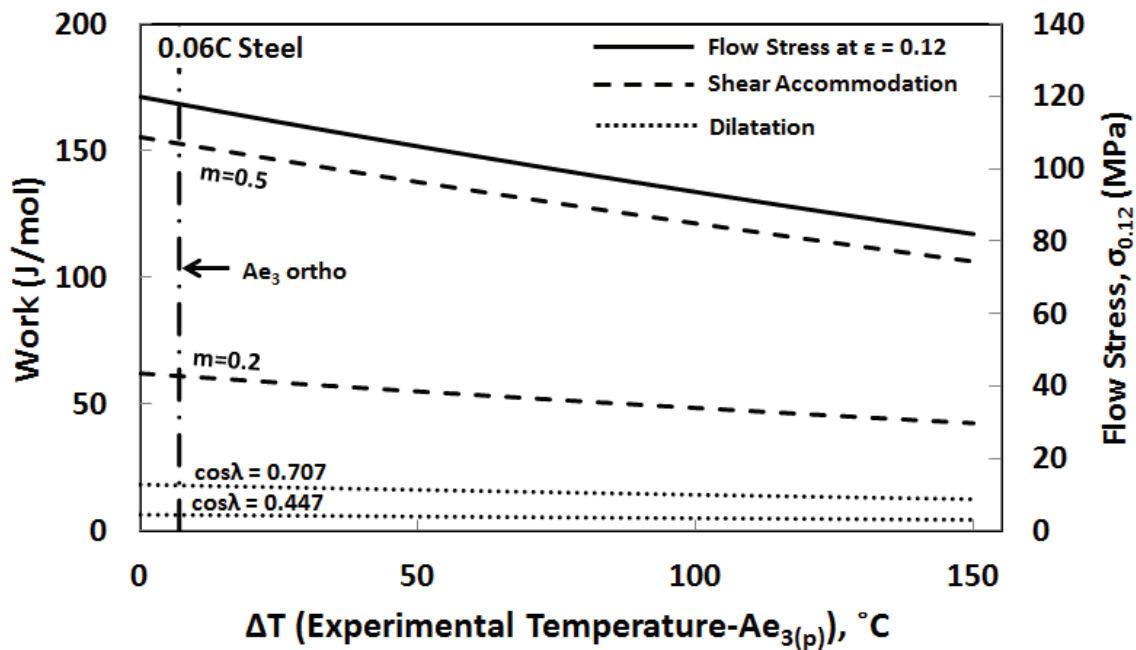


**Fig. 3.11.** Austenite flow stress at  $\epsilon = 0.12$  determined in compression displayed as a function of  $\Delta T$  (right-hand axis). The shear accommodation work (SA) and dilatation work (DW) as a function of  $\Delta T$  (left-hand axis).  $m$ -values of 0.5 and 0.2 are employed to illustrate the effect of grain orientation on the energy barriers.

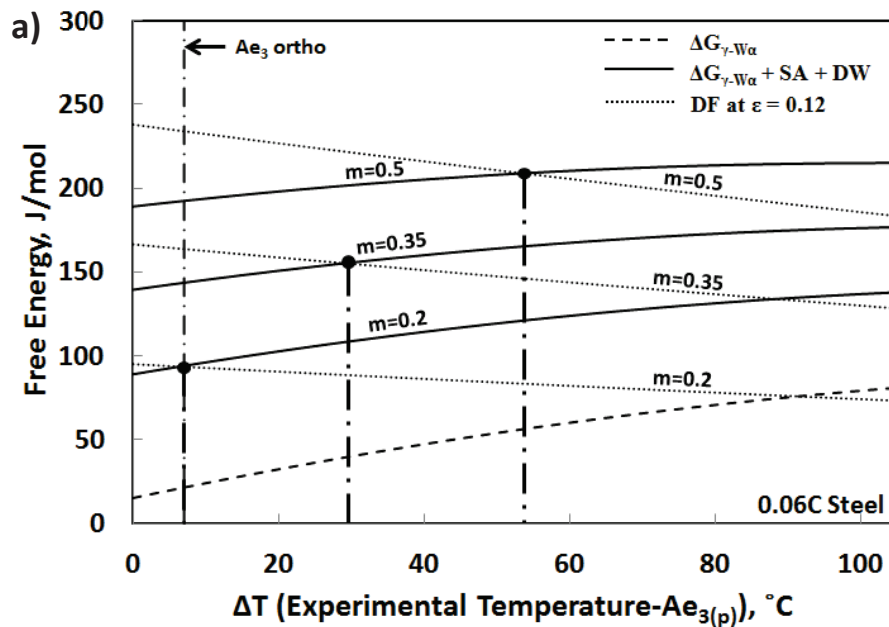
### 3.4.3 Driving Forces versus Free Energy Barriers

The mechanical driving forces developed at a constant strain of 0.12 in grains with  $m$  values in the range 0.2 to 0.5 are illustrated in Figure 3.12a (dotted lines). Here they are compared with the total free energy barriers (solid lines) as given by the sum of the chemical free energy difference ( $\Delta G_{\gamma-\alpha}$ ), shear accommodation work and dilatation work:

$$\text{Total Energy Barrier} = \Delta G_{\gamma-\alpha} + (W/V)_{SA} + (W/V)_D \quad (3.7)$$

The intersections in the plot, labeled here with points, indicate the maximum temperature up to which DT can take place in grains of particular orientations. For example, poorly ( $m = 0.2$ ) and well ( $m = 0.5$ ) oriented grains can undergo transformation at temperatures up to 7 and 54 °C above the  $Ae_3$ , respectively.

The effect of loading the material to a stress level below that of the critical stress is depicted in Figure 3.12b. Taking the flow stress at a strain of  $\epsilon = 0.05$  as an example (Figure 3.8a), it can be seen that the driving force is too low to overcome the barrier to the transformation even in the most ideally oriented grains ( $m = 0.5$ ). This state of affairs also applies to strains of 0.06 and 0.07.



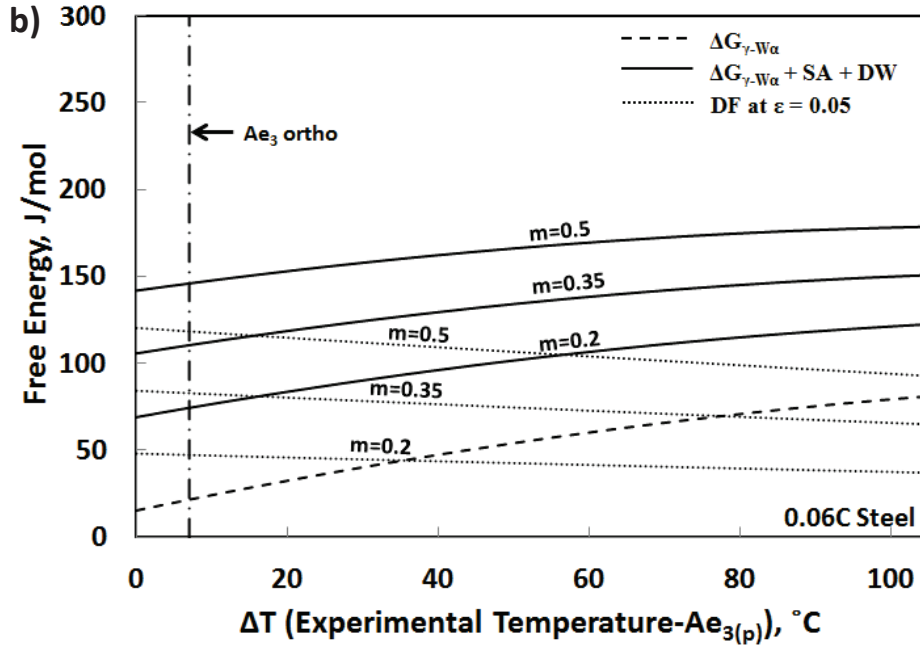


**Fig. 3.12.** a) Dependence of the driving force (DF) at  $\epsilon = 0.12$  on temperature interval and Schmid factor in the range  $0.2 \leq m \leq 0.5$  (dotted lines). The total barriers to transformation over the same temperature range (solid lines). The intersections (labeled with points) correspond to the maximum temperatures up to which dynamic transformation can take place in grains of particular  $m$ -values. b) At a lower strain (i.e.  $\epsilon = 0.05$ ), the absence of intersections indicates that transformation is not possible.

### 3.4.4 Limitations of the present model

According to the current simplified model, see Figure 3.12, dynamic transformation can be initiated at temperatures up to 54 °C above the  $A_{e3(p)}$ . However, these predictions conflict with experimental observations, according to which DT can be induced as much as 200 °C above this temperature. As there are no adjustable parameters in this model, it seems reasonable to question the accuracy of the experimental data employed in the calculations. Although measurements of the flow stresses are subject to error, these are generally small, even with regard to the extrapolations employed for the Widmanstätten ferrite. The problem is probably associated instead with the simplifying assumptions employed in the model.

The most direct way to correct for this discrepancy is to set aside the assumption that the critical strain has an approximately constant value of 0.12 and to allow for the observation that it increases with temperature, as displayed in Figure 3.5a. This is consistent with the

observations of other workers [3.19]. Concurrently, the critical stress remains approximately constant over the experimental temperature range, see Figure 3.5b, in keeping with the nature of a displacive process. If the driving forces derived using Figure 3.8a (based on a *constant* critical strain) are now modified to allow for a constant critical stress instead, it can be seen that the former actually *increase* slightly with temperature (because the Widmanstätten ferrite yield stress decreases with increasing temperature). For the present purpose, for simplicity however, the driving force can be taken as approximately constant over the current temperature range.

The effect of driving forces that are essentially independent of temperature is depicted in Figure 3.13. Here it can be seen that the driving force and total obstacle curves do not intersect at all for  $m = 0.5$ , indicating that dynamic transformation can take place at temperatures as much as 150 °C and more above the  $Ae_3$ . For grains with  $m$  values of 0.2 or 0.35, the intersection temperatures have all increased when compared with those of Figure 3.12a.

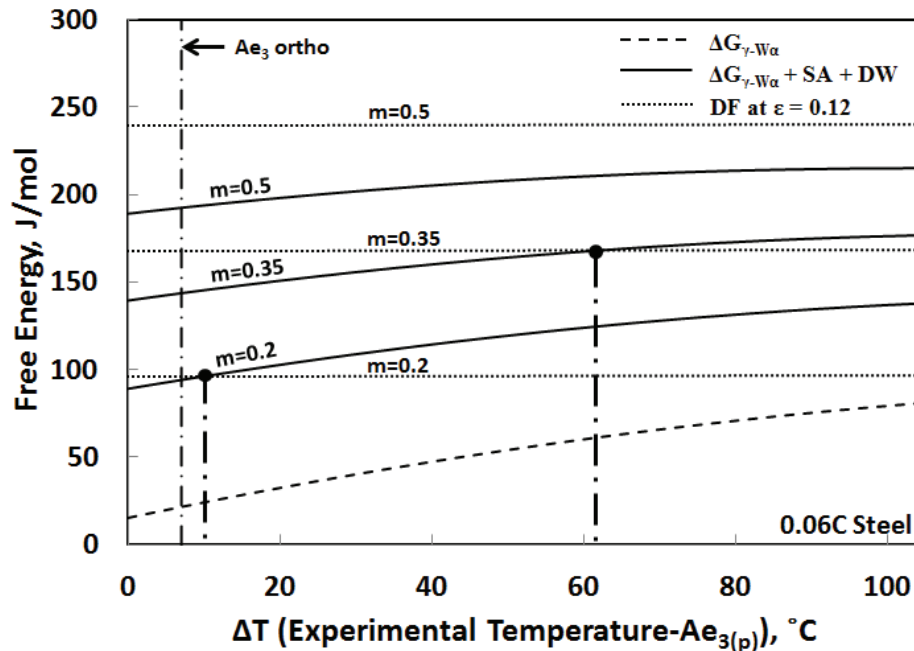


**Fig. 3.13.** Illustration of the effect of assuming that the driving force for the formation of Widmanstätten ferrite remains essentially independent of temperature. The driving force and obstacle curves for  $m = 0.5$  no longer intersect, while the intersection temperatures for  $m = 0.35$  and  $m = 0.2$  have increased when compared with those of Figure 12a.

### 3.5 Conclusions

1. The net softening during transformation can be considered as the driving force for phase change. This is given by the difference between the flow stress of the work hardened austenite at the critical strain and the yield stress of the fresh Widmanstätten ferrite that takes its place. The obstacle to the transformation consists of three parts: i) the free energy difference between austenite and Widmanstätten ferrite, ii) the shear accommodation work that is imposed on the neighborhood by the transformation, and iii) the dilatation work associated with making room for the lower density ferrite.
2. The mechanical driving force for DT can be readily evaluated by plotting the flow stresses and yield stresses against inverse absolute temperature.
3. In the present material, dynamic transformation begins at a critical strain of about 0.11 (at 920 °C), increasing to about 0.14 at 980 °C. At lower strains, the driving force is too low to permit the obstacles to the transformation to be overcome.
4. The use of a constant critical strain in the simplified model presented here accounts for the occurrence of dynamic transformation at temperatures up to 54 °C above the  $A_{e3}$ . However, if allowance is made for the increase in the critical strain with temperature (together with an approximately constant critical stress), then the dynamic transformation observations reported in the literature at temperatures as high as 150-200 °C above the  $A_{e3}$  can be readily justified.
5. The current transformation model also leads to evaluation of the stored energy that is the driving force for dynamic recrystallization. This falls in the range 60 to 320 J/mol.

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## Chapter 4

### Effect of Mn and Si on the Dynamic Transformation of Austenite above the $A_{e3}$ Temperature

In this study, the effects of Mn and Si addition on dynamic transformation were assessed by calculating both the driving forces and the chemical free energy obstacles employing the transformation softening model presented in the previous chapter. For this purpose, the tensile test results of Wray (1984) using 4 steels with contrasting levels of Mn and Si were employed. It was shown there that the addition of Mn increases both the driving force for transformation as well as the energy barrier to ferrite formation. Conversely, while Si addition increases the driving force, it simultaneously *decreases* the free energy obstacle to the transformation. This then increases the temperature range over which DT can take place. This approach shows how to employ the present thermodynamic analysis to predict the effect of adding alloying elements on the occurrence of DT.

- This section has been published: **C. Aranas Jr.** and J.J. Jonas, "Effect of Mn and Si on the dynamic transformation of austenite above the  $A_{e3}$  temperature," *Acta Materialia*, 82 (2015) 1-10.

### Abstract

In the mechanical activation model of dynamic transformation, the stress applied to the austenite is considered to be the driving force for the transformation. Here the yield stress of the newly formed Widmanstätten ferrite that replaces the austenite and that must continue to be deformed is taken into account as an additional obstacle to the transformation. This is estimated from hardness data. In this way, the driving force is redefined to consist only of the *softening* that takes place during the process and therefore only of the *difference* between the current flow stress of the work hardened austenite and the yield stress of the fresh Widmanstätten ferrite that takes its place. According to this model, this driving force is opposed by the chemical free energy difference between the austenite and the Widmanstätten ferrite as well as the lattice dilatation work and shear accommodation work that accompany the transformation. In order to assess the effects of Mn and Si addition on the transformation, the high temperature flow stress data obtained by Wray (1984) by means of tensile tests on four Fe-Mn-Si alloys are employed to determine how these alloying additions affect the driving and opposing forces. The present type of model provides a more detailed analysis of the phenomenon than any available heretofore and in this way shows that Mn addition is only expected to lead to a small decrease in the temperature range over which displacive ferrite can form while Si addition is predicted to have a much larger effect on this interval.

#### 4.1. Introduction

The dynamic transformation (DT) of austenite to ferrite at temperatures above the  $A_{e3}$  was first reported by Yada and coworkers in the 1980's [4.1,4.2]. By means of both rolling simulations and compression tests, they showed that austenite can be transformed into ferrite by deforming it at temperatures as high as 166°C above the  $A_{e3}$ . The thermodynamics of this type of transformation were first analyzed by Hanlon et al. in 2001 [4.3]. By taking the stored energy associated with the dislocations introduced during straining into account, they were able to show that dynamic transformation can readily occur up to 10°C above the  $A_{e3}$  temperature. This approach was later extended by Ghosh et al. [4.4], who increased the justifiable temperature range to 100 °C above the  $A_{e3}$  by allowing for the dislocations introduced during straining to be heterogeneously distributed. These dislocations had to be numerous enough to offset free energy differences of as much as 200 J/mol between the stable austenite and the unstable Widmanstätten ferrite.

A limitation of this approach, however, was that it required levels of strain far above the critical strains of about 0.1 that have been observed to initiate the transformation [4.5,4.6]. This limitation was overcome recently by introducing the concept of mechanical activation, in which the applied stress plays the role of the driving force [4.7]. In this model, the shear accommodation and dilatation work expended during the transformation were identified as barriers to the transformation while the flow stress at the experimental temperature was considered to provide the energy required for the transformation. This model was able to account for the large difference between the rapid (displacive) rate of the forward transformation and the much slower (diffusional) rate of the reverse transformation. It did not, however, lead to a satisfactory quantitative description of this phenomenon.

This shortcoming is overcome here by introducing an additional obstacle to the transformation, namely the flow stress of the newly formed Widmanstätten ferrite. Recognizing that work must be done to deform the ferrite that has now replaced the austenite, the expression for the driving force is modified here to involve only the *difference* between the flow stress of the work hardened austenite being replaced and that of the fresh ferrite that takes its place. Thus the

actual driving force is considered to be the *reduction* in the local flow stress brought about by the transformation and not the full value of the applied stress itself. This model is applied here to four steels of a range of Mn and Si contents so as to examine the effects of the addition of these elements on the transformation.

## 4.2 Influence of Mn and Si addition on the driving force for transformation

Because the flow stress of ferrite above the  $A_{e3}$  cannot be measured experimentally, it must be obtained by extrapolation from tests performed below the  $A_{e3}$ . For this purpose, use is made here of the extensive data reported by Wray [4.8] determined in tension on a series of 18 steels. These experiments were carried out at a strain rate of  $2.3 \times 10^{-2} \text{ s}^{-1}$  over the temperature range 600 to 1300 °C.

The compositions of the four ultra-low carbon steels selected for the present analysis are listed in Table 4.1. All the transformation temperatures were calculated using the FSstel database of FactSage thermodynamic software [4.9]. Two of the alloys contained low levels of Mn (about 0.01wt%) and two contained high levels (about 1.4wt%). Comparison of their behaviors allows the effect of adding 1.4wt% of Mn to be assessed. The two corresponding Si levels were 0.24 and 0.95wt%, leading to an evaluation of the effect of adding 0.7wt% Si. The four alloys are referred to here as LL, LH, HL, & HH, see Table 4.1, where L and H refer to “low” and “high”, so LL is the base line composition and HH represents the combined effect of adding both elements.

**Table 4.1.** Chemical compositions of the Fe-Mn-Si alloys (in wt%).

| Code | C     | Mn    | Si   | S     | P     | O     | $A_{e1}$<br>(°C) | Para $A_{e3}$<br>(°C) | Ortho $A_{e3}$<br>(°C) | $T_m$<br>(°C) |
|------|-------|-------|------|-------|-------|-------|------------------|-----------------------|------------------------|---------------|
| LL   | 0.007 | 0.011 | 0.24 | 0.002 | 0.001 | 0.004 | 863              | 924                   | 927                    | 1533          |
| HL   | 0.007 | 1.40  | 0.25 | 0.002 | 0.001 | 0.004 | 776              | 843                   | 862                    | 1526          |
| LH   | 0.005 | 0.013 | 0.95 | 0.001 | 0.001 | 0.008 | 932              | 981                   | 992                    | 1522          |
| HH   | 0.005 | 1.38  | 0.94 | 0.002 | 0.001 | 0.008 | 826              | 877                   | 901                    | 1515          |

A feature of importance for the present purpose, as will be seen in more detail below, is that Wray reported values of the flow stresses he measured at true strains of 0.001, 0.002, 0.005,

0.01, 0.02, 0.05, and 0.1. As the critical strain for the initiation of dynamic transformation is about 0.1 [4.5,4.6] and the newly transformed ferrite is as yet undeformed, this is the exact range of strains over which the flow stresses must be known for derivation of the driving force. Also of importance is the dependence on temperature of these flow stresses, also provided by Wray [4.8], as this is required for assessment of how the driving force depends on temperature. An example of one of the flow curves determined on the 1.4Mn-0.25Si steel reported in this work is reproduced here as Fig. 4.1. Here the following five strains, 0.002, 0.01, 0.02, 0.05 and 0.1 have been identified on this curve, as these play important roles in the calculations to follow. As mentioned above, dynamic transformation is only observed when the critical strain of about 0.1 is exceeded. Conversely, at strains below 0.1, the conditions required for the nucleation of Widmanstätten ferrite have not yet been satisfied. Accordingly, the effect of increasing strain on the driving force for the transformation will be examined below and compared with the free energy obstacle to the transformation, which depends on the temperature, but not on the strain.

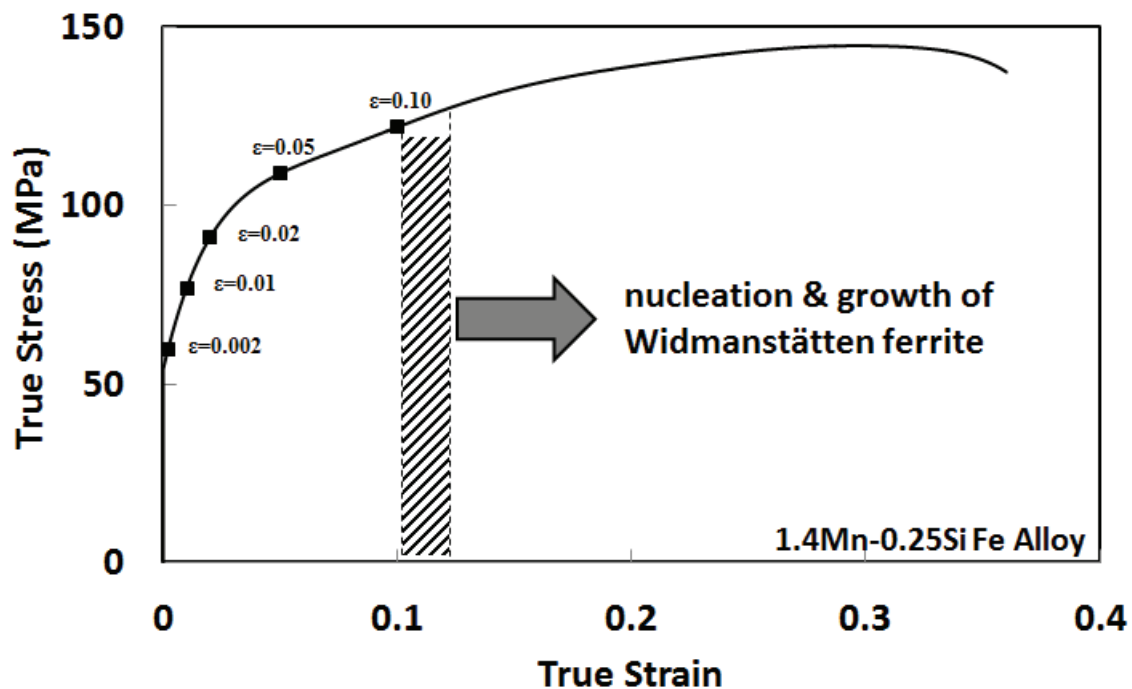


**Fig. 4.1.** Tensile flow curve determined by Wray [4.8] on a 1.4Mn-0.25Si steel at 600 °C and a strain rate of  $2.3 \times 10^{-2} \text{ s}^{-1}$ . Samples were tested in argon after annealing for one hour at 1050 °C. The maximum load in this case was attained at  $\epsilon = 0.17$ . Nevertheless, discernible necking was delayed until  $\epsilon = 0.35$ , permitting determination of the true strain up to this point [4.8].

Examples of the Widmanstätten ferrite plates formed when the critical strain is exceeded are provided in Fig. 4.2 [4.4,4.5]. This phenomenon has now been studied in detail in the eight steels listed in Table 4.2. In these tests, the C contents ranged from 0.019 to 0.79 wt% and the Mn and Si concentrations from 0.30 to 1.56 wt% and 0.01 to 1.56 wt%, respectively. However, as the individual effects of adding these elements are not linear with respect to the amount added nor are the effects of multiple additions simply additive as demonstrated below, it is of interest to consider how the influence of arbitrary additions of Mn and Si, taken one at a time, can be assessed. This will be considered in some detail below.

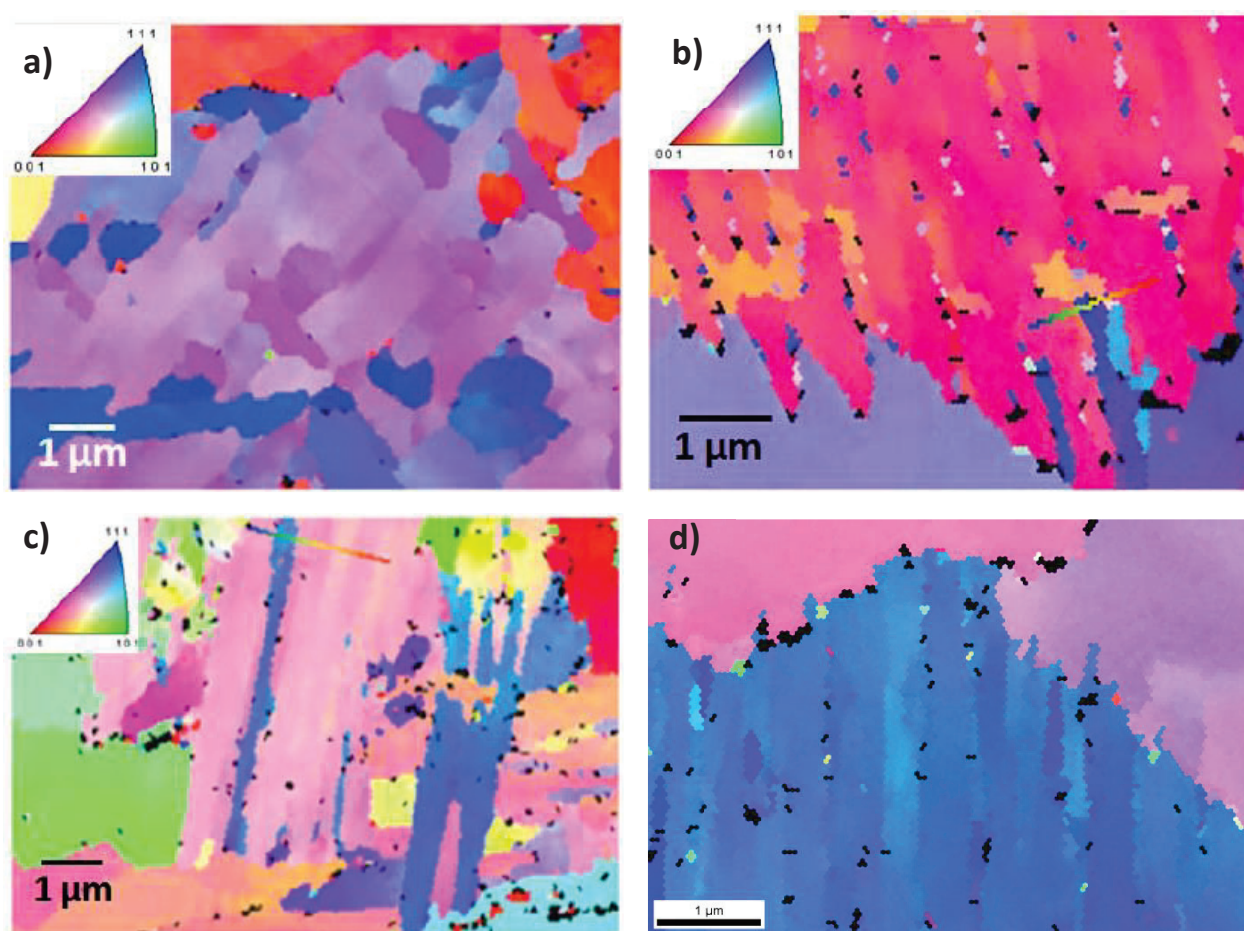


**Fig. 4.2.** Widmanstätten microstructures formed under the following conditions: a) 0.06wt%C steel strained to  $\epsilon=0.25$  at 887 °C ( $A_{e3(p)} + 10^\circ\text{C}$ ); b) 0.21wt%C steel strained to  $\epsilon=0.25$  at 862 °C ( $A_{e3(p)} + 40^\circ\text{C}$ ); c) and d) 0.09wt%C steel strained to  $\epsilon=0.25$  at 856 °C ( $A_{e3(p)} + 20^\circ\text{C}$ ).

**Table 4.2.** Chemical compositions of the Fe-Mn-Si-Nb alloys (in wt%) in which the occurrence of DT has been confirmed.

| Steel | Composition (wt%) |      |      |       | $A_{e3}$ Temperature (°C) |       | Maximum Temperature at which DT identified (°C) | $T_{max} - A_{e3(p)}$ (°C) |
|-------|-------------------|------|------|-------|---------------------------|-------|-------------------------------------------------|----------------------------|
|       | C                 | Mn   | Si   | Nb    | para                      | ortho |                                                 |                            |
| 1     | 0.019             | 1.5  | 0.2  | -     | 860                       | 872   | 1100                                            | 240 <sup>b</sup>           |
| 2     | 0.06              | 0.30 | 0.01 | -     | 870                       | 877   | 917                                             | 47 <sup>a</sup>            |
| 3     | 0.21              | 1.30 | 0.24 | -     | 808                       | 822   | 912                                             | 104 <sup>a</sup>           |
| 4     | 0.45              | 0.70 | 0.24 | -     | 753                       | 772   | 802                                             | 49 <sup>b</sup>            |
| 5     | 0.79              | 0.65 | 0.24 | -     | 709                       | 733   | 863                                             | 154 <sup>a</sup>           |
| 6     | 0.09              | 1.30 | 0.02 | 0.036 | 820                       | 836   | 896                                             | 76 <sup>a</sup>            |
| 7     | 0.11              | 1.10 | 0.26 | 0.038 | 836                       | 848   | 1100                                            | 264 <sup>b</sup>           |
| 8     | 0.22              | 1.56 | 1.56 | 0.045 | 850                       | 860   | 1100                                            | 250 <sup>b</sup>           |

<sup>a</sup> – Widmanstätten ferrite microstructure observed using electron microscopy. [4.4]

<sup>b</sup> – Dynamic transformation detected by means of the double differentiation method [4.10]

Hundreds of tests have been carried out and analyzed using the double differentiation technique to determine the critical strain for the initiation of dynamic transformation [4.5,4.6,4.10,4.11]. These have all led to values in the range  $\epsilon = 0.10$  to  $0.12$  as shown in Fig. 4.1.

In order to determine the driving force for the transformation, the amount of softening associated with the phase change must first be established. The dependence on temperature and composition must also be evaluated. This is accomplished by following the procedure outlined in Fig. 4.3 for alloy LL, where the austenite flow stresses determined by Wray are plotted against inverse absolute temperature. Here the black lines represent linear fits to these data. For assistance in fitting, the flow stresses were considered to go to zero at the melting point, although they actually have small (and unknown) non-zero values at this temperature. Also indicated on this diagram is a linear fit to the flow stresses determined on polygonal ferrite both below and just above the  $A_{e1}$  temperature. These are also assumed to go to zero at the melting point, as shown by the extrapolation into the austenite phase field. Similar data were analyzed for the other three materials, but are not illustrated here.

As will be shown below, the absence of significant work hardening at small strains (e.g.  $0.002$  to  $0.02$ ) does not lead to flow stress increases that are large enough to provide sufficient energy

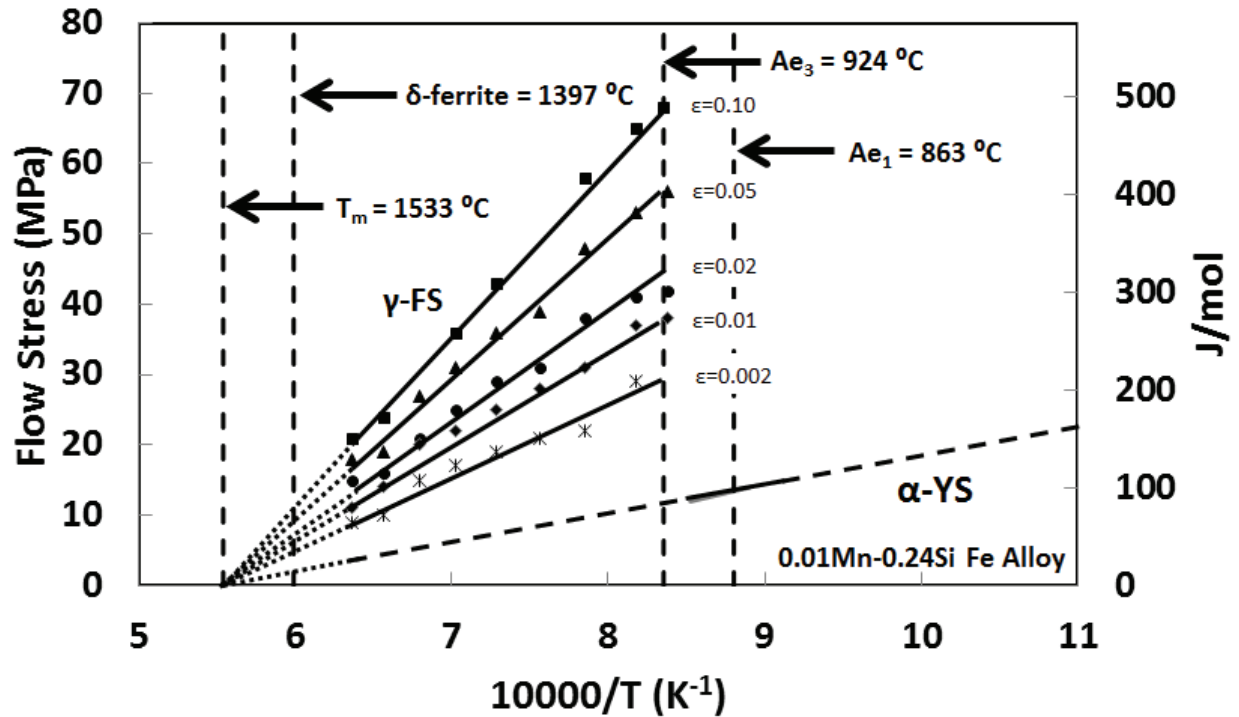


**Fig. 4.3.** Effect of temperature on the flow stress levels attained at increasing strains determined in tension on the 0.01Mn-0.24Si-Fe steel of Ref. 8 at a strain rate of  $2.3 \times 10^{-2} s^{-1}$ . Linear relationships are assumed between the inverse absolute temperature and the flow stress of austenite ( $\gamma$ -FS) as well as the yield stress of polyagonal ( $\alpha$ -YS) ferrite.

to initiate the transformation. Conversely, the amount of work hardening associated with the critical strain will be seen to provide enough softening that the obstacles to the transformation can be overcome. One further input is required in order to carry out the driving force calculation and that is the flow stress of the fresh (undeformed) Widmanstätten ferrite. Because of the presence of the dislocations introduced during the shear transformation, such ferrite is harder (and has a higher flow stress) than conventional polyagonal ferrite. The room temperature hardnesses of the Widmanstätten and polyagonal ferrite were therefore determined on steel no. 3 (Table 4.2), as shown in Fig. 4.4. This led to a hardness (and therefore UTS [4.12]) ratio of  $168/142 = 1.18$ . The latter ratio was then employed to estimate the yield strength of the Widmanstätten ferrite in the austenite phase field.

Using information similar to that displayed in Fig. 4.3, the austenite flow stresses at a strain of 0.1 (the approximate critical strain for dynamic transformation in commercial steels) and the

Widmanstätten ferrite yield stresses were plotted against inverse absolute temperature for the four materials, as illustrated in Fig. 4.5. Here the paraequilibrium  $A_{e3}$  and  $A_{e1}$  temperatures are also shown, together with the melting and delta ferrite transformation temperatures.

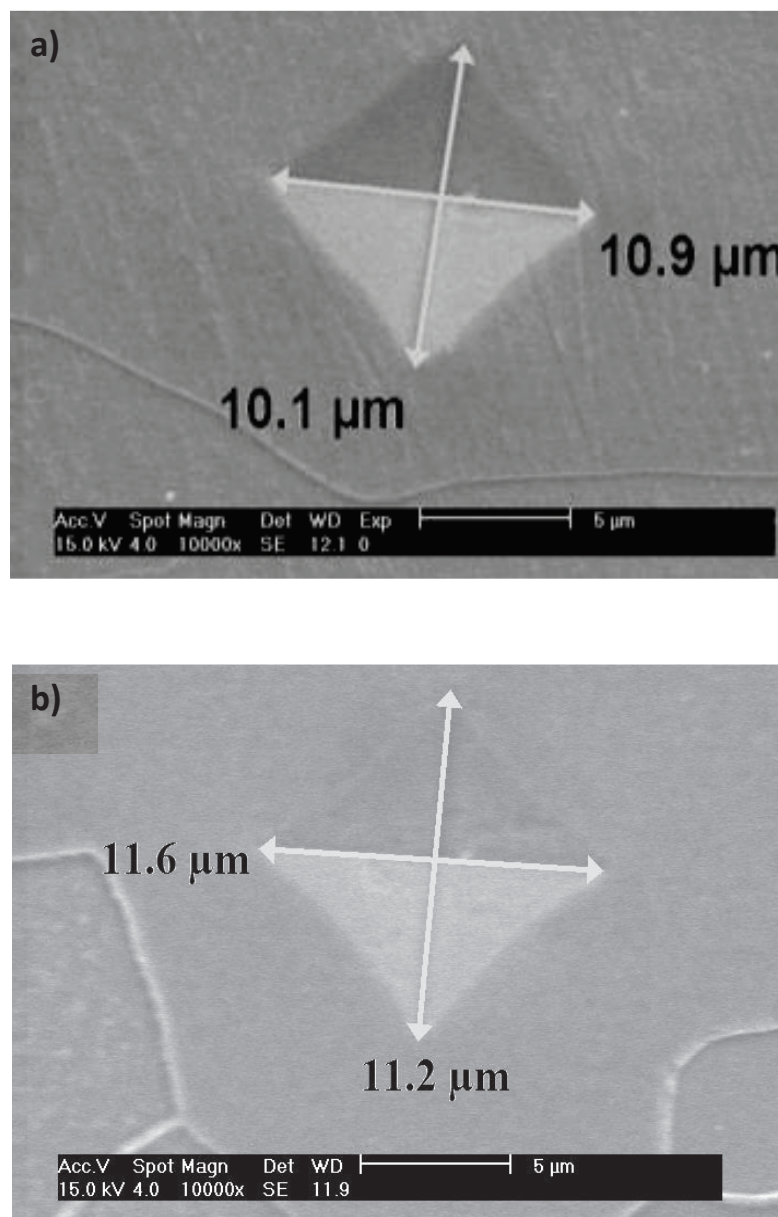
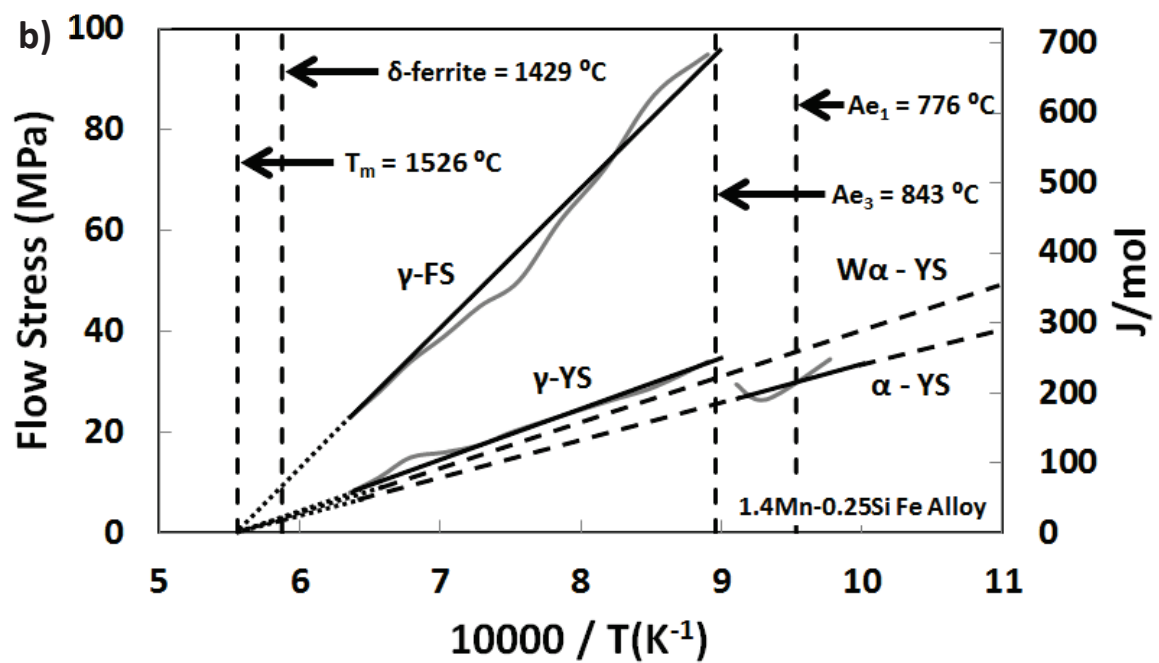
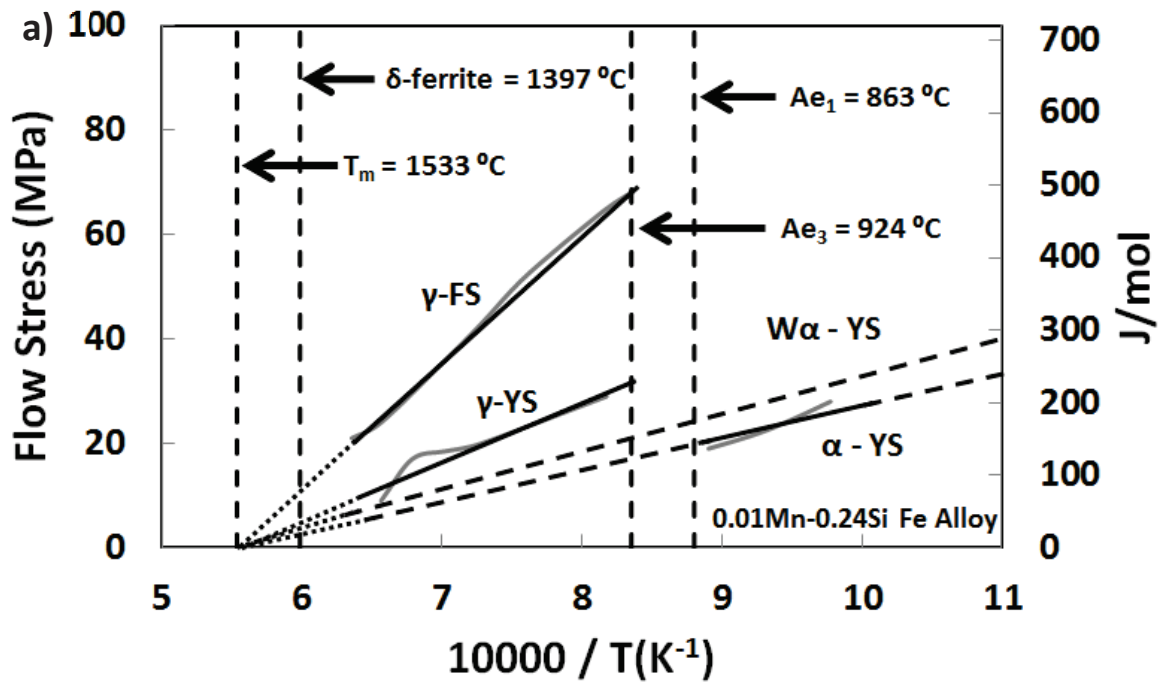


**Fig. 4.4.** Examples of Vickers hardness measurements performed on the 0.21wt%C steel (Table 2) previously strained to  $\epsilon = 2$  at a rate of  $4.5 \text{ s}^{-1}$ . Samples were deformed at  $895^\circ\text{C}$  ( $A_{e3(p)} + 87^\circ\text{C}$ ) and then quenched. a) Widmanstätten colony displaying a hardness of 168 HV; b) polygonal ferrite (held for 800s after deformation) displaying a hardness of 142 HV.



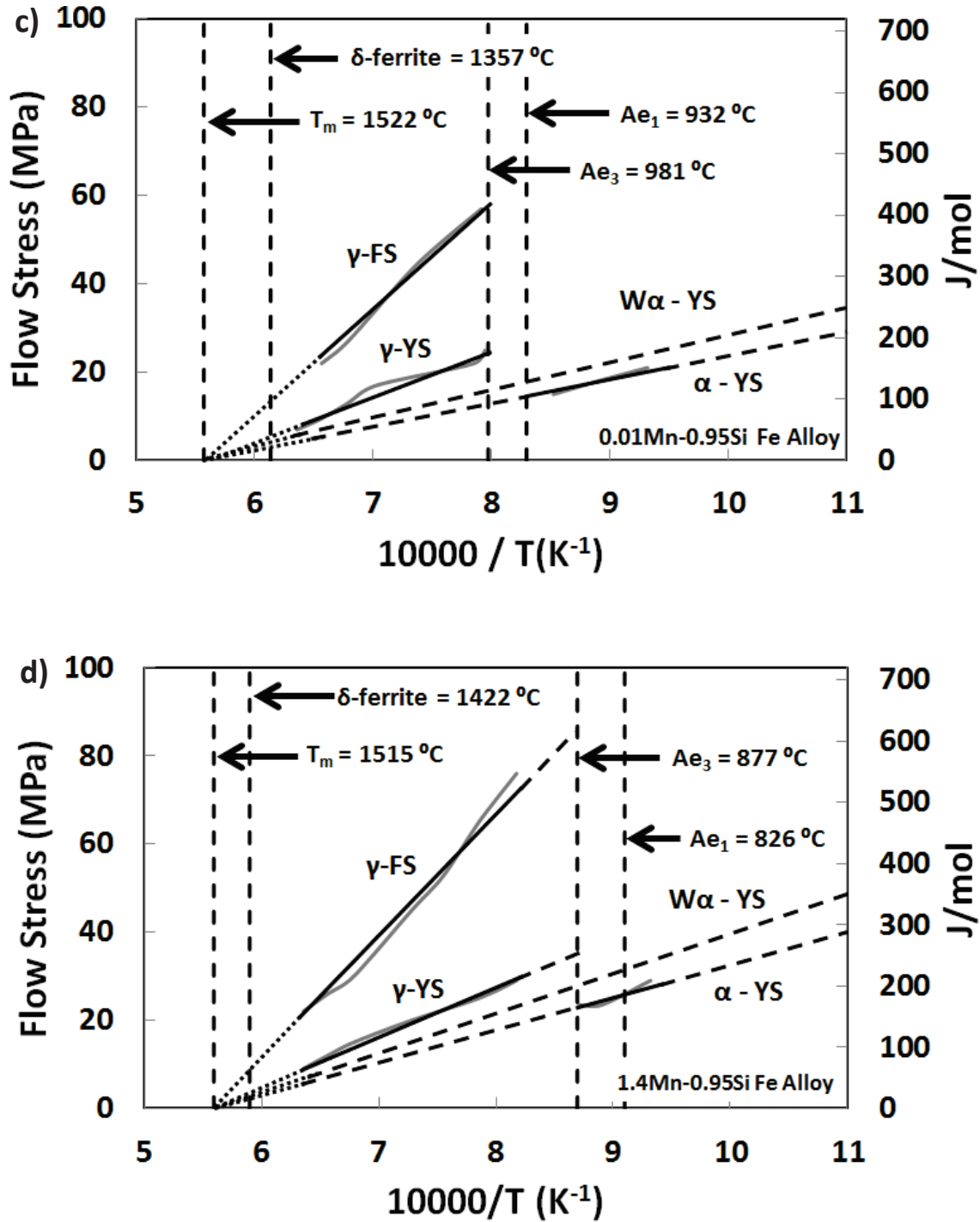


**Fig. 4.5.** Dependences of the flow stress on inverse absolute temperature in the present four Fe-Mn-Si alloys. The behavior of the LL (low Mn-low Si) alloy of Fig. 4.5a is the basis for comparison to determine the effects of Mn addition (Fig. 4.5b), Si addition (Fig. 4.5c) and the combined addition of both Mn and Si (Fig. 4.5d). Linear dependences of the various flow stresses on inverse absolute temperature are again assumed here and the fitted lines all approach zero stress at the melting point.

The driving force for transformation at a given temperature is represented in these diagrams by the vertical distance between the austenite flow stress at 0.1% strain and the Widmanstätten ferrite yield stress, estimated as described above. By comparison with the driving forces in the LL case (0.01Mn-0.24Si), Fig. 4.5a, it can be seen that the addition of 1.4 wt% Mn, HL, Fig. 4.5b, leads to a *large increase* in the driving force for ferrite formation, while the addition of 0.7 wt% Si, LH, Fig. 4.5c, also increases the driving force at a given temperature but to a lesser extent. When Mn and Si are added in combination, HH, Fig. 4.5d, the driving force is further increased beyond the amounts associated with single Mn or Si addition.

These effects are at first sight somewhat counter-intuitive, as Mn, being an austenite stabilizer, should *hinder* (not promote) ferrite formation. As will be shown below in more detail, the usual effects of adding Mn and Si are to increase and decrease, respectively, the chemical free energy barrier opposing the formation of ferrite above the  $A_{e3}$ . However, here the measurements refer to the effects of these additions on the respective *high temperature flow stresses*. It is the significantly greater strengthening effect of adding Mn (the austenite stabilizer) on the austenite flow stress than on the ferrite flow stress *at elevated temperatures* that is responsible for the driving force increase that accompanies its addition. Concurrently, the chemical free energy difference resisting the transformation is also increased, so the two effects of Mn addition act in opposition to each other. By contrast, the addition of Si increases the driving force for ferrite formation on the one hand while decreasing the resistance to the transformation via its effect on the chemical free energy on the other, thus promoting the occurrence of dynamic transformation.

These effects of Mn and Si addition on the driving force at the critical strain of 0.1 are summarized in Fig. 4.6. Here, in order to permit comparison with the chemical free energy differences employed below, the driving forces of Fig. 4.5 (in MPa) have been converted into units of J/mol using the conversion factor  $1 \text{ MPa} = 1 \text{ MJ m}^{-3} = 7.2 \text{ J/mol}$ . As can be seen from Fig. 4.6a, Mn addition increases the driving force by about 28 to 47 J/mol over the present temperature range. In a similar manner, the addition of Si, Fig. 4.6b, increases the driving force by about 8 to 12 J/mol. Finally, in the case of the combined addition, Fig. 4.6c, the driving force

increase is somewhat greater than that of simple Mn addition because both Mn and Si addition act to strengthen the austenite.

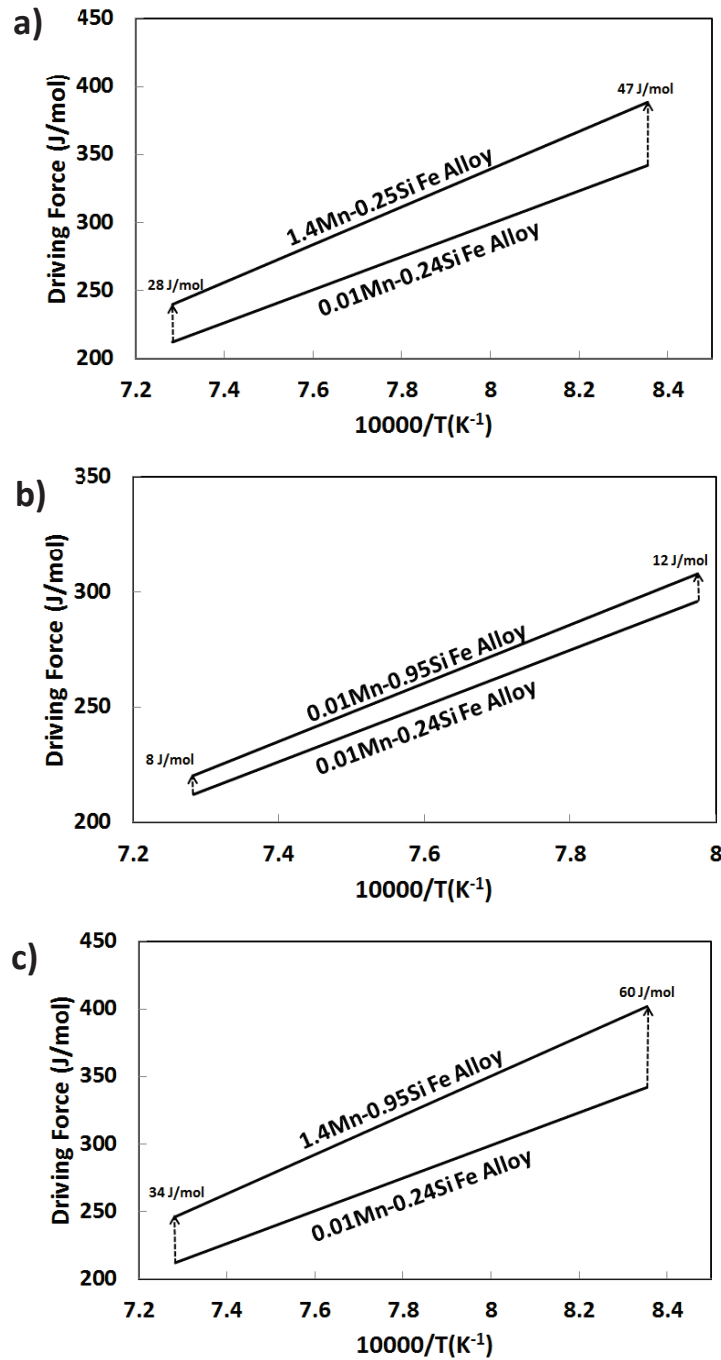


**Fig. 4.6.** The driving forces in J/mol derived from the data of Fig. 4.5 by subtracting the Widmanstätten ferrite yield stresses from the austenite flow stresses at the critical strain of 0.1. The addition of Mn (Fig. 4.6a) increases the energy available by 28-47 J/mol and that of Si (Fig. 4.6b) by 8-12 J/mol. The simultaneous addition of both alloying elements (Fig. 4.6c) increases the driving force by 34-60 J/mol.

It is evident from the above that Mn and Si addition both increase the driving forces for ferrite formation. Nevertheless, the net effect of these additions cannot be known until the influence of adding these alloying elements on the obstacles to ferrite formation is also evaluated. This is the task described in the section that follows.

### **4.3 Influence of Mn and Si addition on the obstacles to ferrite formation above the $A_{e3}$**

The three most important obstacles to the formation of ferrite under the present conditions are: i) the chemical free energy difference between the phases; ii) the work done to accommodate localized shearing during the transformation; and iii) the dilatation work imposed on the surrounding austenite by the expansion of the lower density ferrite [4.7]. These will now be evaluated in turn for the four materials.

The chemical free energy barriers to the transformation were first calculated using the FSstel database of the FactSage thermodynamics software [4.9]. All these were derived under the condition that paraequilibrium conditions prevail, as the previous work [4.4-4.7] showed that the nature of the displacive transformation did not permit any partitioning of the substitutional elements during the formation of Widmanstätten ferrite, as discussed by others [4.13]. The differences are also based on the formation of polygonal (and not Widmanstätten) ferrite, as the latter values were not available in the database.

The dependences of these free energy barriers on temperature above the  $A_{e3}$  are displayed in Fig. 4.7a, where it should be noted that the actual  $A_{e3}$  temperature (and therefore the temperature at a particular abscissa) varies somewhat with Mn and Si content, as shown in Table 1. At this point, the thermodynamic quantities displayed in Fig. 4.7a must be modified to account for the free energy differences between Widmanstätten and polygonal ferrite. This can be done by taking the room temperature value for this difference, about 50 J/mol [4.14], and adjusting for the approximate effect on the dislocation density of raising the temperature by 900 to 1000 °C. Here the free energy difference is assumed to be relatively constant at about 10 J/mol, where the effect of temperature over the current experimental range is neglected as

being of secondary importance<sup>\*</sup>. It should also be noted that there are no adjustable parameters in the present model, so that any changes to this estimate can be considered to play the role of an adjustable parameter. The effect of adding 10 J/mol to the obstacle heights is illustrated in Fig. 4.7b for steel LL, where the values shown here are employed in the calculations that follow.

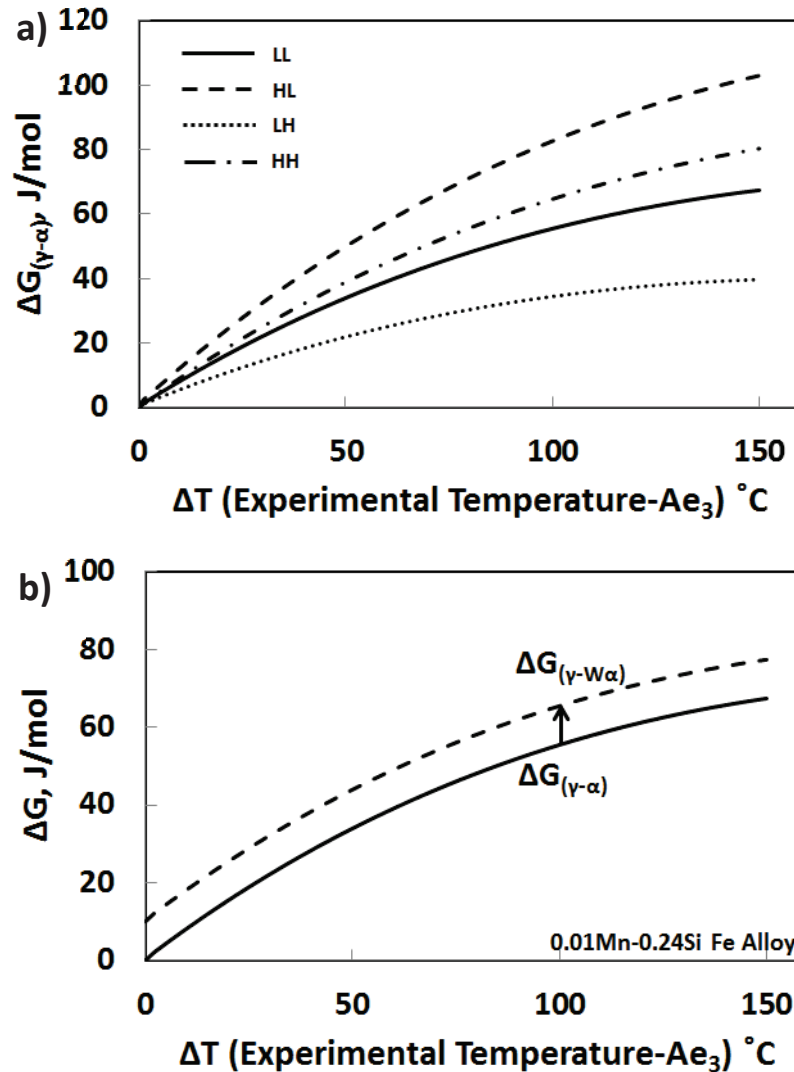
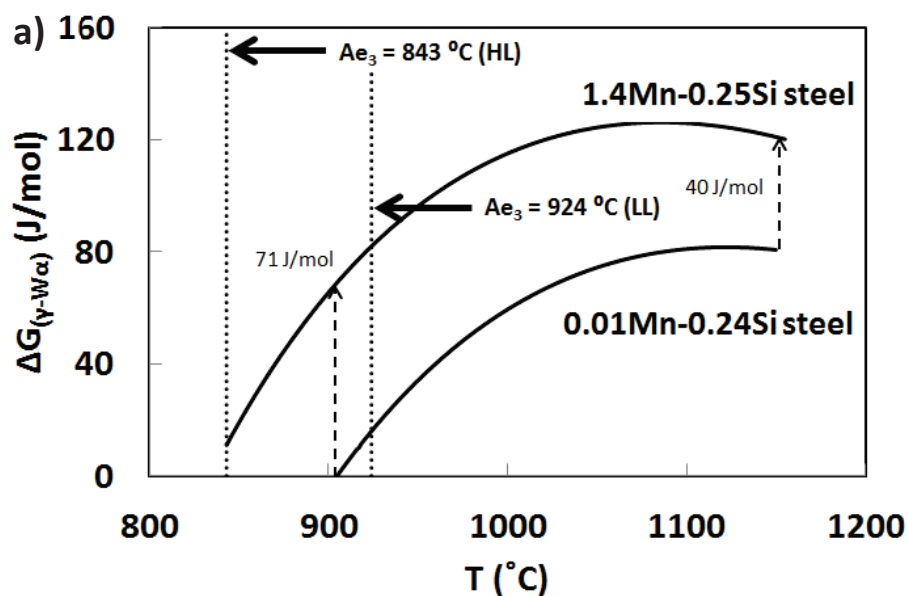


**Fig. 4.7.** a) The chemical free energy differences between austenite and polygonal ferrite ( $\Delta G_{\gamma-\alpha}$ ) in the four alloys presented as a function of  $\Delta T$  (experimental temperature – paraequilibrium  $A_{e3}$ ). The free energy barrier increases with  $\Delta T$  because of the increasing instability of ferrite as the temperature is increased. b) The chemical free energy between austenite and Widmanstätten ferrite ( $\Delta G_{\gamma-W\alpha}$ ) in the LL steel.

<sup>\*</sup>Allowing for the temperature dependence of the free energy difference would lead to slight increases in the predicted temperature range (see Fig. 4.11below) over which displacive ferrite can be formed.

The Widmanstätten alpha values were used in the preparation of Fig. 4.8, where the effects of Mn and Si addition are illustrated more directly in these diagrams. Here it can be seen that Mn addition, Fig. 4.8a, increases the free energy barrier by about 40 to 70 J/mol, that is, by more than its effect on the driving force. When it comes to the Si, see Fig. 4.8b, it is evident that its addition leads to barrier decreases, as expected, of about 35 J/mol, changes that are opposite in sign to the effects of Si addition on the flow stress. Finally, when both elements are added in combination, Fig. 4.8c, the magnitude of the barrier increase associated with Mn addition is decreased somewhat by the Si addition.

The presence of maxima in the barrier curves reflects that fact that the chemical free energy obstacles go to zero (the dilation and shear accommodation work do not go to zero) at the delta ferrite transformation temperature.



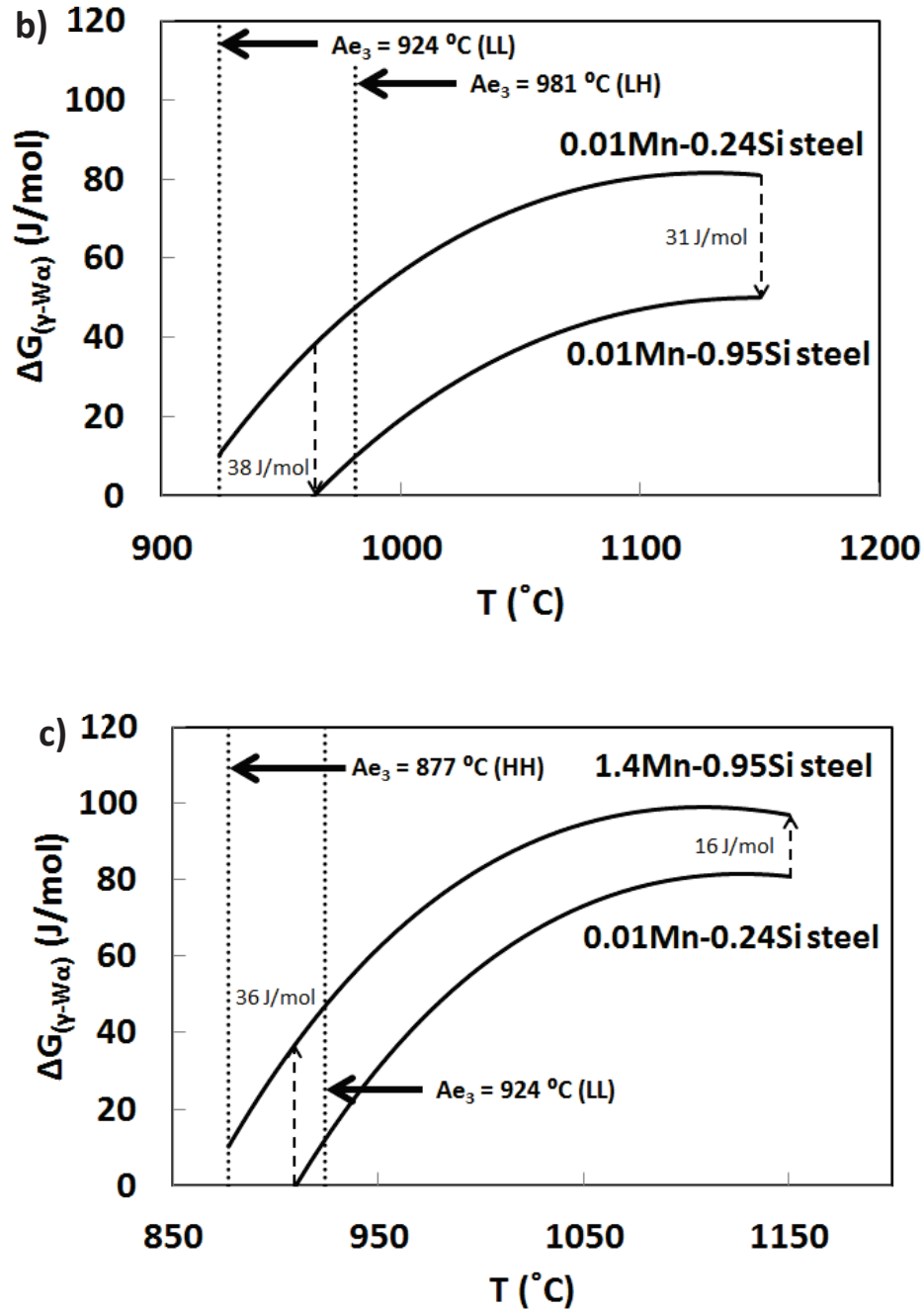


**Fig. 4.8.** Gibbs free energy differences derived from Fig. 4.7 showing the effects of Mn addition (Fig. 4.8a), Si addition (Fig. 4.8b) and the combined addition both Mn and Si (Fig. 4.8c). Since Mn is an austenite stabilizer, it *increases* the Gibbs energy difference, while Si, as a ferrite former has the opposite effect. The addition of both alloying elements leads to a combination of the two effects. It should be noted that the total barriers are characterized by maxima as the chemical free energy differences go to zero at the delta iron transformation temperature.

#### 4.4 Shear and dilatation work

The work expended in accommodating the shearing process as well as with the dilatation associated with ferrite formation were deduced using the austenite flow stresses developed at the measured critical strain of 0.1. The transformation of austenite to Widmanstätten ferrite involves a shear strain  $\gamma = 0.36$ , [4.7], which was employed here to calculate the accommodation work. The dilatation strain was taken to be 3% and was considered to take place perpendicular to the habit plane [4.15,4.16]. For this purpose, appropriate Schmid and orientation factors,  $m$  and  $\lambda$ , must be employed, which are defined here with respect to the following transformation habit plane (0.506, 0.452, 0.735) and shear direction (-0.867, 0.414, 0.277) [4.17].

In this way, the shear accommodation work per unit volume  $(W/V)_{SA}$ , was calculated using the following relation [4.7]:

$$(W/V)_{SA} = m \times \sigma_{0.1} \times 0.36 \quad (4.1)$$

where the product is expressed in units of J/mol. The results obtained in this way for the LL grade are displayed in Fig. 4.9 for  $m$  values of 0.5 and 0.1. Similar values were obtained for the three other materials (not shown here), which will be employed below.

The dilatation work per unit volume,  $(W/V)_D$ , is also illustrated in Fig. 4.9 for the same material, where orientation factors of  $\lambda = \sqrt{m} = 0.707$  (corresponding to  $m = 0.5$ ) and  $\lambda = \sqrt{m} = 0.316$  (corresponding to  $m = 0.1$ ) were employed. These were calculated for a dilatation strain of 3% using the relation [4.7]:

$$(W/V)_D = \cos \lambda \times \sigma_{0.1} \times 0.03 \quad (4.2)$$

These values of the dilatation and shear accommodation work will now be employed to calculate whether ferrite nucleation is possible or not at strains of: i) 0.01 (i.e. before the critical strain); and ii) 0.10, i.e. at the critical strain.

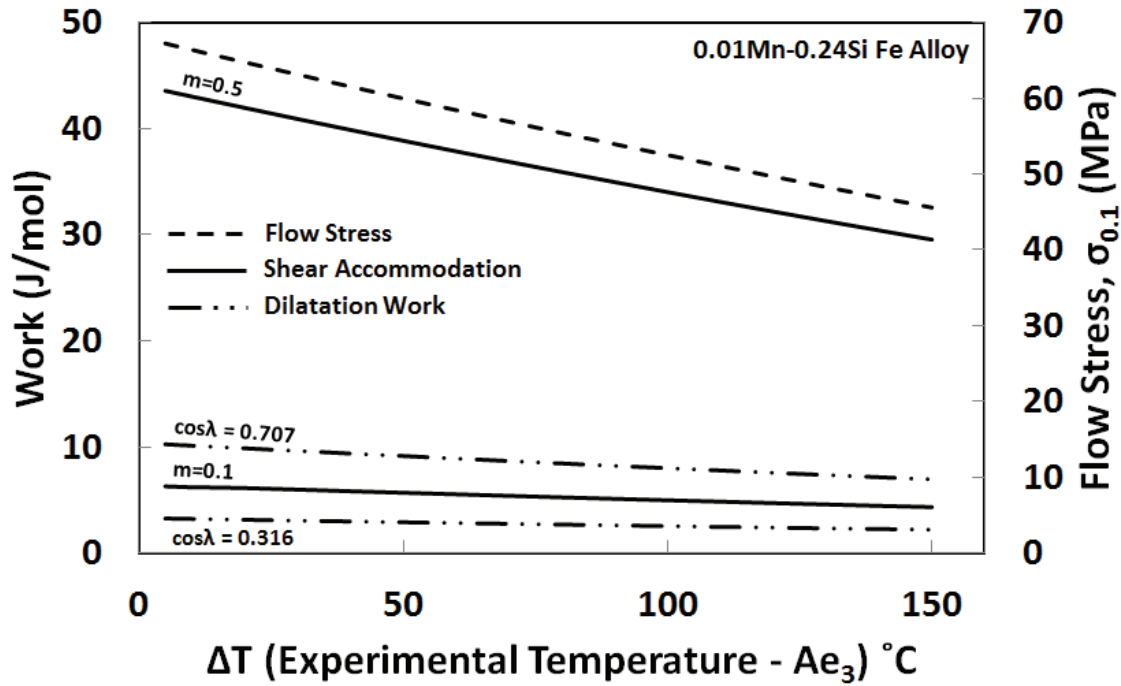


**Fig. 4.9.** The flow stress of the 0.01Mn-0.24Si-Fe (LL) alloy at  $\epsilon = 0.1$  determined in tension displayed as a function of  $\Delta T$  (right-hand axis). The calculated shear accommodation (SA) and lattice dilatation (DW) work in J/mol plotted with respect to  $\Delta T$  (left-hand axis). The behaviors of grain orientations with  $m$ -values of 0.5 and 0.1 are illustrated to represent the barriers associated with high and low resolved stresses.

## 4.5 Discussion

### 4.5.1 Driving Forces vs. Energy Barriers

One of the questions considered by the present model involves the minimum value of the driving force that can provoke the transformation. This essentially defines the critical strain, although initiation may begin well before the type of critical strain detected by the “double differentiation method” [4.18,4.19]. This is because the latter requires the presence of a minimum volume fraction of transformed material before it has a measurable effect on the work hardening rate and can therefore be detected experimentally. In what follows, the observed dependence of the driving force on strain will first be used to question whether transformation can be initiated at or shortly after yielding. Later, the conditions under which

the driving force becomes equal to the resistance provided by the obstacles to the process will also be derived.

Turning first to the possibility of transformation at low strains, the above three components of the energy barrier, as expressed by eq. 4.3 below, are illustrated in Fig. 4.10 for a strain of 0.01 for the four materials.

$$\text{Energy Barrier to the Transformation} = \Delta G_{\gamma - w\alpha} + (W/V)_{SA} + (W/V)_D \quad (4.3)$$

where  $\Delta G_{\gamma - w\alpha}$  is the free energy difference between austenite and Widmanstätten ferrite. As already indicated earlier, the addition of Mn raises the energy barrier opposing the transformation while that of Si lowers it.

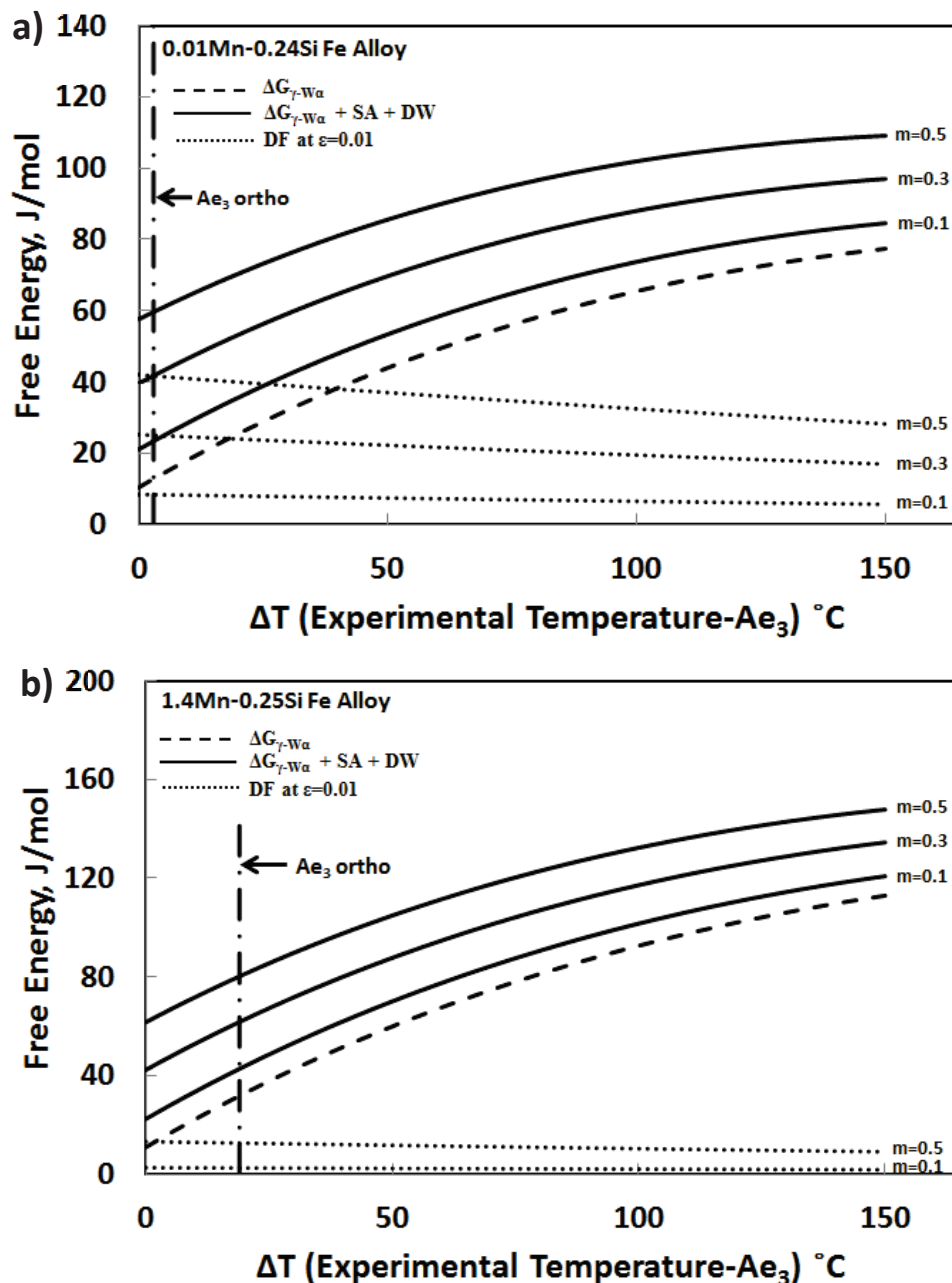
The driving forces present at the yield stress ( $\epsilon = 0.01$ ) are also displayed in Fig. 4.10, where they are compared with the total free energy barriers for the four materials. The results are illustrated for  $m$  values of 0.1, 0.3, and 0.5, where these represent the behaviors of grains that are: i) poorly oriented; ii) moderately well oriented; and iii) particularly well oriented, respectively, with respect to the applied stress. The driving force per unit volume,  $(W/V)_{\text{driving force}}$ , is accordingly considered to vary with grain orientation and therefore to depend on  $m$ -value, as specified by the following relation:

$$(W/V)_{\text{driving force}} = \Delta \text{Flow Stress} \times m \quad (4.4)$$

Here the austenite *yield* stress ( $\epsilon = 0.01$ ) was employed in the work terms.

It can be seen from Fig. 4.10 that, even in the most favorably oriented grains, where  $m = 0.5$ , the driving force is insufficient to initiate the transformation. This is indicated by the absence of an intersection between the driving force and barrier curves. The absence supports the view that there is a critical stress as well as a critical strain involved in provoking dynamic transformation.

It is of interest that the gap between the energy required and that provided by the driving force is greatest in the two Mn added alloys, as can be expected from the austenite stabilizing properties of this element. Thus it appears that the addition of Mn can act to prevent the formation of mechanically activated ferrite while Si can promote it. This effect is demonstrated below in more detail when the driving forces at larger strains are compared with the obstacles to the transformation.



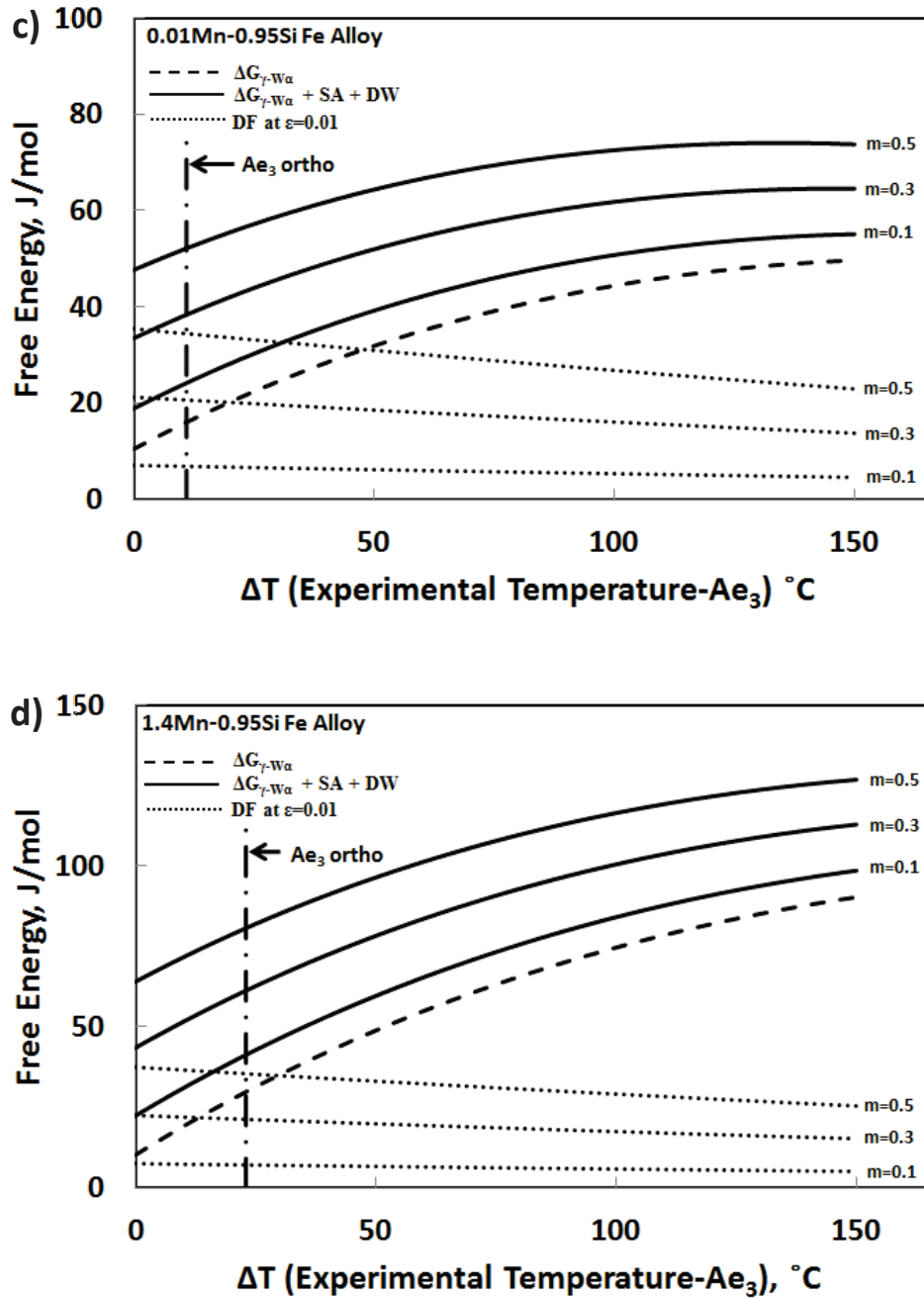
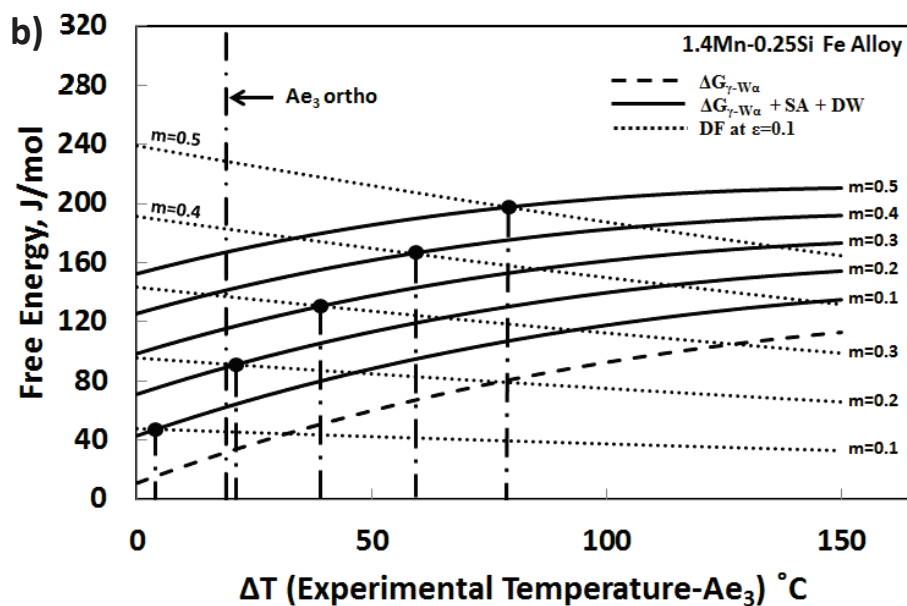
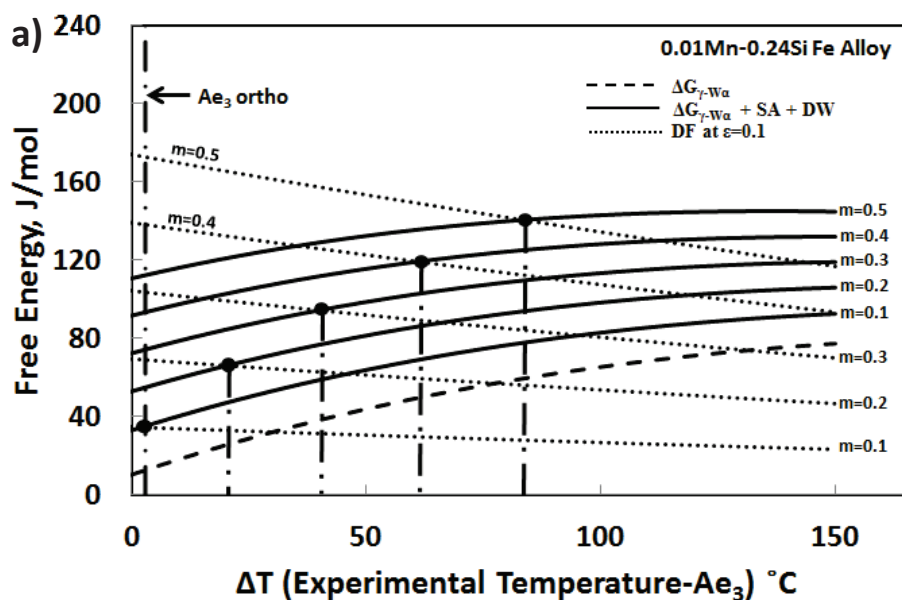


**Fig. 4.10.** The driving force (DF) at yield ( $\epsilon = 0.01$ ) is compared to the obstacle to transformation at temperatures above the paraequilibrium  $A_{e3}$ . It can be seen that dynamic transformation is not possible under these conditions because the stress difference (driving force) is too low, even for grains that are favorably oriented ( $m = 0.5$ ). The addition of Mn (compare Fig. 4.10a and Fig. 4.10b) decreases the probability of transformation by increasing the gap between the obstacle and the driving force whereas Si addition (compare Fig. 4.10a and Fig. 4.10c) increases the probability by decreasing the gap.

The driving forces expected to be present when the critical strain of 0.1 is attained were also calculated for the four materials. These are compared to the corresponding energy barriers (evaluated using the austenite flow stress at  $\epsilon = 0.1$ ) and illustrated in Fig. 4.11. Here, the behaviors pertaining to the grains associated with Schmid factors in the range 0.1 to 0.5 are depicted. As can be seen, there is enough driving force in the moderately well oriented grains to permit activation, i.e. intersection takes place, at the lower temperature intervals above the  $A_{e3}$ . Poorly oriented ( $m < 0.2$ ) grains cannot undergo transformation at all (except at temperatures close to the  $A_{e3}$ ). At the higher temperatures, however, only the most favorably oriented ( $m = 0.5$ ) grains can be activated and finally, at still higher temperatures, mechanical activation is no longer possible.

The driving force/obstacle intersections in Fig. 4.11 enable some conclusions to be drawn regarding the effects of Mn and Si addition on the tendency for the formation of displacive ferrite above the  $A_{e3}$ . By comparing the  $m = 0.5$  intersections in Figs. 4.11a and 4.11b, it is evident that the addition of 1.4 wt% Mn *decreases* the temperature range over which Widmanstätten ferrite can be formed by 4 °C, i.e. by about 3 °C for a 1 wt% addition, although this effect is not expected to be completely linear. Conversely, by comparing the  $m = 0.5$  intersections in Figs. 4.11a and 4.11c, the effect of a 1 wt% addition of Si can be described as leading to an *increase* of 70 °C in the expected temperature range. Finally, from Figs. 4.11a and 4.11d, the effect of the combined addition of 1.4 wt% Mn and 0.7 wt% Si in the present case is seen to involve a range increase of 19 °C. This can be compared with an expected increase of 45 °C based on a simple linear superposition of their effects. It is of interest that the normalized effect of Mn addition is very small while that of Si addition is quite large. This can be attributed directly to the large driving force (flow stress) increase associated with Mn addition, which opposes the effect of Mn on the obstacle height. By contrast, in the case of Si, the changes in driving force and obstacle free energy are cumulative and not in opposition.



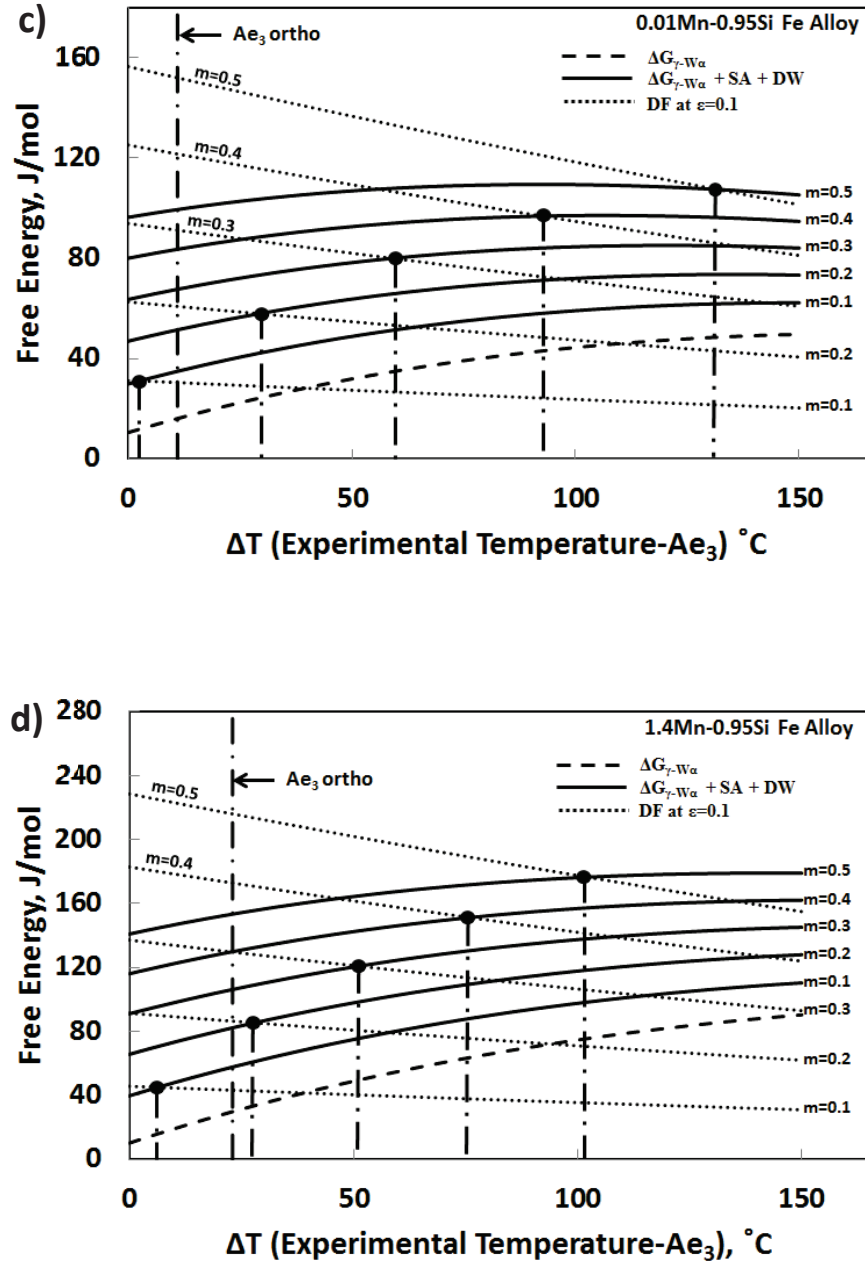


**Fig. 4.11.** The driving force (DF) at the critical strain of 0.1 compared with the total energy barrier at temperatures above the paraequilibrium  $A_{e3}$ . The points of intersection at given values of  $m$  indicate the maximum temperatures up to which dynamic transformation can take place. The temperature interval above the  $A_{e3}$  is restricted by adding Mn (compare Fig. 4.11a and Fig. 4.11b) and augmented by adding Si (compare Fig. 4.11a and Fig. 4.11c).

#### 4.5.2 Mechanical activation and the critical strain

A more detailed demonstration of the role of the Schmid factor on the initiation of the transformation is depicted in Fig. 4.12. Here, the HH alloy is being deformed 50 °C above the paraequilibrium  $A_{e3}$  and the flow stress (i.e. the driving force) at which intersection takes place has been deduced for three  $m$  values. It can be seen that the driving force first matches the free energy barrier at about 0.06 strain, but only in grains where the  $m$ -value = 0.5. This is physically reasonable, a trend that is analogous to the role of  $m$ -value in twinning in Ti [4.20]. As the strain is gradually increased, first to 0.08 and then to 0.10, transformation is enabled in grains of  $m$ -value 0.4 and then 0.3.

Such modeled behavior must now be compared with critical strains of about 0.1 observed in commercial steels [4.5,4.6,4.10,4.11]. As has been suggested above, the transformation of a few suitably oriented grains at strains below the observed 'critical' suggests that the volume fraction of the first grains to transform is too small to be detectable by the double differentiation method. The critical strain may also be less than 0.1 in ultra-low C steels of the type considered here. It is also possible that the present model, being somewhat oversimplified, is unable to represent all aspects of the dynamic transformation behavior. Further refinement is required, not only to shift the intersections to larger strains and higher temperatures, but also to allow the chemical free energies of Widmanstätten ferrite at temperatures well above the  $A_{e3}$  to be estimated more accurately.

Another shortcoming of the model concerns the effect of temperature on the intersection strain. Plots such as that of Fig. 4.12 for higher and higher temperatures (not shown here) depict the critical strains rising to large and larger values. This is because more work hardening would be required to attain the higher barrier levels associated with higher temperatures, particularly as the overall stress level of the flow curve has also decreased. This is in conflict with the observation that the critical strain remains nearly constant, or decreases slightly, with increasing temperature [4.5,4.6,4.10,4.11]. A possible explanation of this observation (not allowed for in the model) is that, while there is less nucleation at a given strain at higher temperatures, there is considerably more growth [4.21]. In this way, the faster growth at

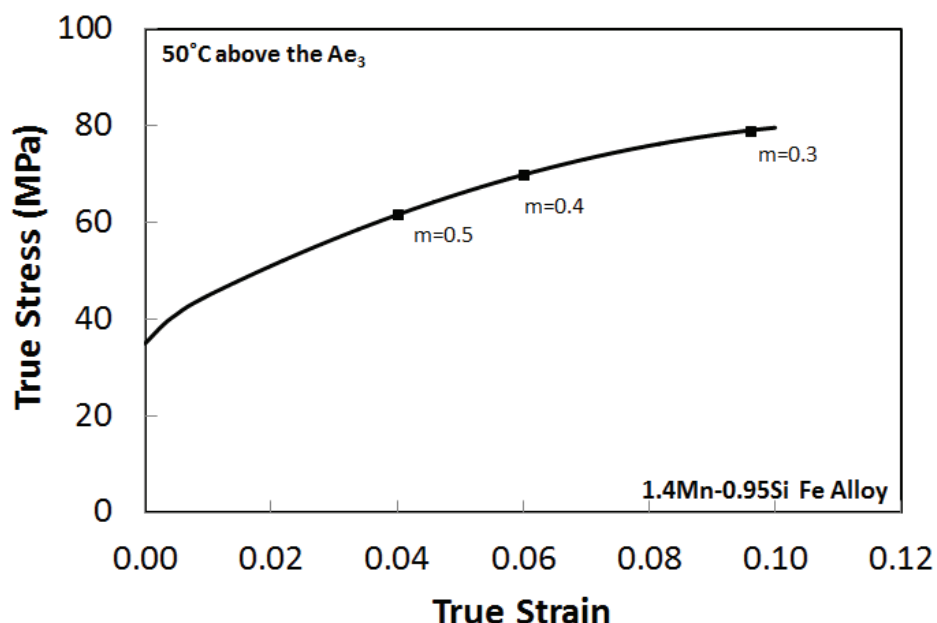


**Fig. 4.12.** The critical stresses and strains associated with m-values of 0.3, 0.4 and 0.5 when the austenite of alloy HH is being deformed at 50 °C above the paraequilibrium  $A_{e3}$ . At higher temperatures (not shown here), the critical strains move to higher and higher values.

higher temperatures could lead to volume fraction increases that compensate for the smaller number of nuclei.

#### 4.5.3 Summary of the model

The main features of the present model are summarized in Fig. 4.13. Here the driving force is depicted as the *softening* that accompanies the transformation; this contrasts sharply with the earlier approach in which the resistance to deformation of the newly formed Widmanstätten ferrite did not play a role [4.7]. This more accurate representation of the forces involved is one of the reasons why the present model is able to provide a quantitative description of the transformation, while the previous version was restricted to qualitative predictions. As shown in the diagram, the model accounts for both the dilatation work required to accommodate the density decrease associated with ferrite formation as well as the shear work required to allow plate formation.

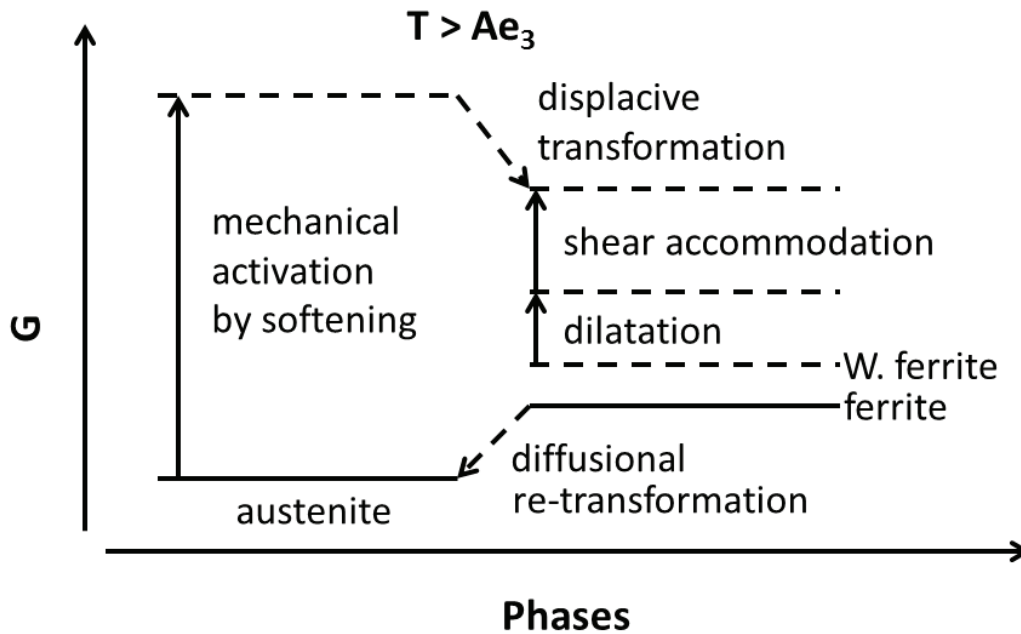


**Fig. 4.13.** The dynamic transformation model based on activation by transformation softening. Here allowance is also made for the higher free energy of the Widmanstätten ferrite over that of the polygonal ferrite values provided by the FactSage thermodynamic model. The shear accommodation and dilatation work are also shown, as is the re-transformation back to austenite that takes place by diffusional processes upon removal of the applied stress.

#### 4.6 Conclusions

1. The current value of the applied stress is shown to be an inappropriate driving force in the mechanical activation model of dynamic transformation. It should, instead, be replaced by the flow stress *difference* between the work hardened austenite and the yield stress of the fresh Widmanstätten ferrite that takes its place. According to this view, it is the *softening* produced by the transformation that provides the energy required to overcome the obstacles to the phase change.
2. The flow stress of Widmanstätten ferrite at temperatures above the  $A_{e3}$  can be estimated by linear extrapolation against inverse absolute temperature from polygonal and Widmanstätten ferrite data obtained below the  $A_{e1}$ . This enables the derivation of the driving force required by the present model.
3. The stress-difference model indicates that dynamic transformation cannot be induced at applied stresses in the vicinity of the yield stress. Instead, the driving force must be

increased by work hardening until a minimum critical stress, and therefore a critical strain, is attained.

4. At the critical strain, both moderately well oriented (Schmid factor,  $m = 0.3$ ) and ideally oriented ( $m = 0.5$ ) grains can participate in the transformation. As the temperature interval above the  $A_{e3}$  is increased, transformation can only take place in grains of more and more ideal orientation, that is those associated with higher and higher  $m$ -values.
5. The addition of Mn increases both the driving force for the transformation as well as the free energy obstacle to ferrite formation. This leads to a small net decrease in the driving force and therefore in the temperature interval above the  $A_{e3}$  over which dynamic transformation can take place. Conversely, Si addition increases the driving force and decreases the obstacle height, thus leading to a large increase in the temperature interval. The present thermodynamically-based model thus leads to insights regarding the effects of Mn and Si addition that are not intuitively obvious.

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## Chapter 5

### **Formation of Widmanstätten Ferrite in a C-Mn Steel at Temperatures High in the Austenite Phase Field**

In the previous chapters, it was shown that the driving force for dynamic transformation is the flow stress difference between the work hardened austenite at the critical strain and the yield stress of the fresh Widmanstätten plates that take its place. The energy obstacle consists of the chemical free energy difference between austenite and ferrite, as well as the shear accommodation and dilatation work that accompany the transformation. In this study, compression tests were carried out using a Gleeble 3800 system (at the École de Technologie Supérieure) in the temperature range 1000 °C to 1350 °C to determine the maximum temperature above the  $A_{e3}$  at which DT can take place. It was shown that DT ferrite can be formed at *any temperature* within the austenite phase field. These findings were supported by the calculated Gibbs energy difference between the austenite and ferrite, which displays a peak approximately midway between the  $A_{e3}$  and the delta ferrite formation temperature. The shape of the barrier also explains why less ferrite is formed and why higher critical strains are observed at temperatures in the vicinity of the peak than at higher or lower temperatures. A more detailed account of these observations is presented in Appendix A.

- This section has been published: **C. Aranas Jr.**, R. Grewal, K. Chadha, D. Shahriari, M. Jahazi and J.J. Jonas, "Formation of Widmanstätten Ferrite in a C-Mn Steel at Temperatures High in the Austenite Phase Field," Proceedings of the International Conference on Solid-Solid Phase Transformations in Inorganic Materials (PTM 2015), Whistler, BC, (2015) 613-620.

### **Abstract**

Gleeble compression tests were carried out on a 0.06%C-0.3%Mn-0.01%Si steel over the temperature range 1000°C to 1350°C. Strains of 0.7 were applied at a strain rate of  $1\text{s}^{-1}$ . The double differentiation method was employed to determine the critical strains for the initiation of dynamic transformation (0.05 - 0.12) and dynamic recrystallization (0.10 - 0.18). The occurrence of dynamic transformation was detected at temperatures up to 1350°C. Optical and electron microscopy images indicated that Widmanstätten ferrite plates were being formed in the austenite phase field and that the plates coalesced into polygonal grains during straining. These results are interpreted in terms of the flow softening model of dynamic transformation.

## 5.1 Introduction

In the late 1980's, Yada and co-workers [5.1,5.2] showed that austenite can be transformed into ferrite by deformation. They were able to show that dynamic transformation (DT) can be initiated up to 166°C above the  $A_{e3}$  temperature. This was followed by in-situ X-ray diffraction studies [5.3] and dilatometric measurements [5.4] to provide real time evidence for the occurrence of this unusual phenomenon. Later experiments [5.5,5.6] using optical and EBSD microscopy confirmed that the ferrite was being formed displacively and dynamically.

Dynamic transformation was initially attributed to the stored energy of the dislocations introduced by the deformation. In this way, Sun et al. [5.7] calculated free energy increases of about 22.3 J/mol and Hanlon et al. [5.8] converted this energy into an  $A_{e3}$  temperature increase of 10°C. These values are too low given the observations of Yada and co-workers of transformation temperatures well above the  $A_{e3}$  [5.1,5.2]. Ghosh et al. resolved this limitation by allowing for the inhomogeneous distribution of the dislocations, leading to driving forces as high as 197 J/mol [5.9]. However, this approach is based on the application of very large strains, higher than the measured critical strains of about 0.10 associated with the initiation of DT [5.10,5.11]. More recently, the present authors have proposed that the driving force promoting the transformation is the softening brought about by the transformation [5.12,5.13]. This method can readily justify the occurrence of DT as much as 150-200 °C above the  $A_{e3}$  reported in the literature.

In the present study, compression tests were carried out on a Gleeble machine on a 0.06%C-0.30%Mn-0.01%Si steel. The aim was to determine the maximum temperature above the  $A_{e3}$  at which DT could take place readily. In this way, the softening-based transformation model could be tested using the experimental results described in more detail below.

## 5.2 Experimental Procedure

A low carbon C-Mn steel was investigated in the current study. This was received in the form of a hot rolled plate with a thickness of 12.5 mm. The steel plate was machined into cylindrical specimens, 10mm in diameter and 15mm long, with their axes parallel to the rolling direction.

The complete chemical composition is listed in Table 5.1 together with its corresponding paraequilibrium and orthoequilibrium  $Ae_3$  temperatures. These transformation temperatures were calculated using the FSstel database of the FactSage thermodynamic software [5.14]. The results and discussions in the next sections are based on the paraequilibrium temperature, as only the interstitial atoms are able to partition between the product and parent phases under displacive transformation conditions [5.9].

**Table 5.1.** Chemical composition (mass%) and equilibrium transformation temperatures (°C)

| C    | Mn   | Si   | Orthoequilibrium $Ae_3$ | Paraequilibrium $Ae_3$ |
|------|------|------|-------------------------|------------------------|
| 0.06 | 0.30 | 0.01 | 877 °C                  | 870 °C                 |

The hot compression tests were conducted using a Gleeble 3800 at temperatures high in the austenite phase field. The samples were heated to test temperatures in the range 1000°C to 1350°C at the rate of 4°C/s. They were then held for 3 minutes, after which a strain of 0.7 was applied at a strain rate of  $1\text{ s}^{-1}$ . The thermomechanical schedule followed in the compression tests is illustrated in Figure 5.1.

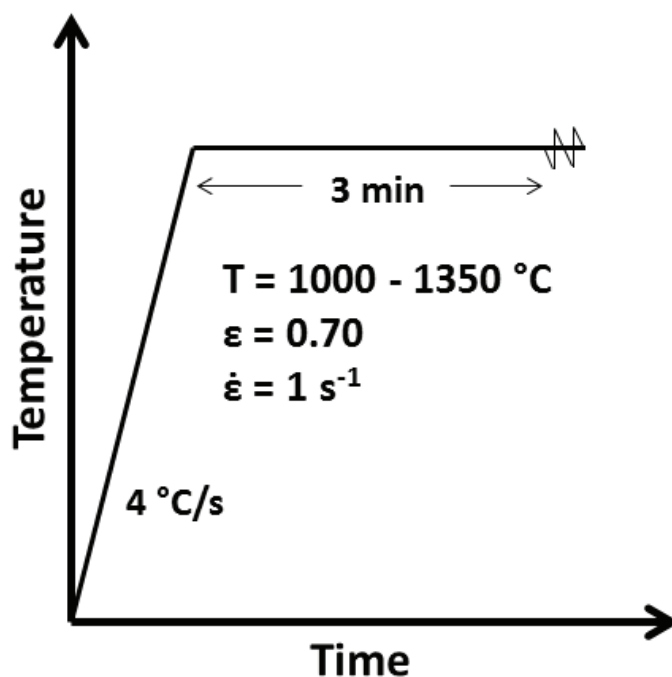


**Fig. 5.1.** Thermomechanical schedule of the compression tests

### 5.3 Results

Some of the stress-strain curves determined in this way are presented in Figure 5.2. Here the samples were deformed in the range 1000 °C to 1350 °C, high in the austenite phase field. The flow stress levels decrease with increasing temperature and softening takes place after the peak stress. This led to stress drops of as much as 10% after the maximum stress was attained. Such softening involves both dynamic transformation and dynamic recrystallization, as shown elsewhere [5.10,5.15].

Optical micrographs of samples quenched right after deformation display significant volume fractions of ferrite, as illustrated in Figure 5.3. There was insufficient time during cooling for the ferrite to have formed by a diffusional process (which requires substitutional diffusion), confirming that it was nucleated displacively [5.16]. The martensite (prior austenite) appears dark while the ferrite is light. As shown in an earlier study [5.13], the volume fraction of DT ferrite decreased as the temperature was increased over the temperature range from 920 °C to

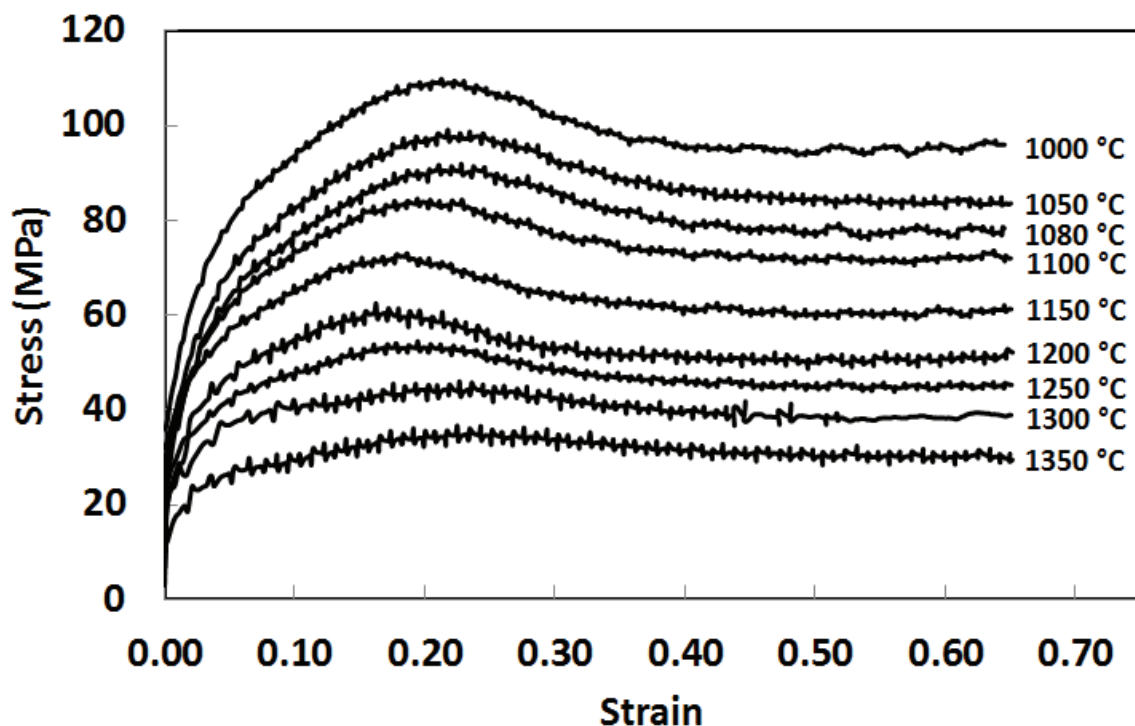


**Fig. 5.2.** Stress-strain curves of the C-Mn steel compressed to a strain of 0.7 at  $1 \text{ s}^{-1}$ . The tests were carried out at temperatures in the range from 1000 °C to 1350 °C.

980 °C (up to about 110 °C above the  $A_{e3}$ ). Here, at 1050 °C (i.e. 180 °C above the  $A_{e3}$ ), the volume fraction of ferrite was only about 5-10% and most of the structure was martensitic, as illustrated in Figure 5.3a. However, it can be clearly observed that at the higher temperatures of 1150 °C and 1200 °C (Figures 5.3b, 5.3c and 5.3d), the volume fraction of ferrite increases to about 15% and 30%, respectively. This interesting behavior will be analyzed in more detail in the next section.

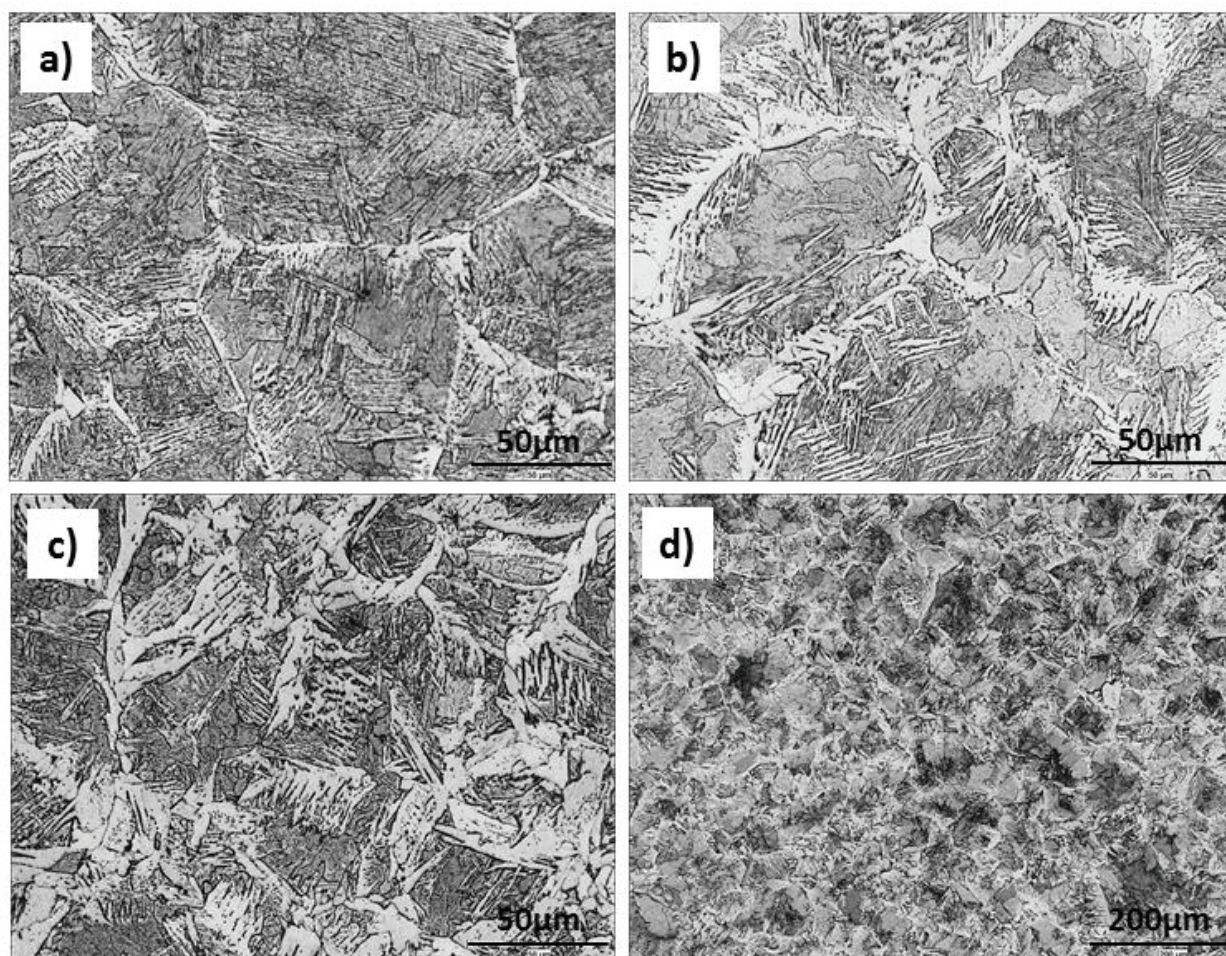
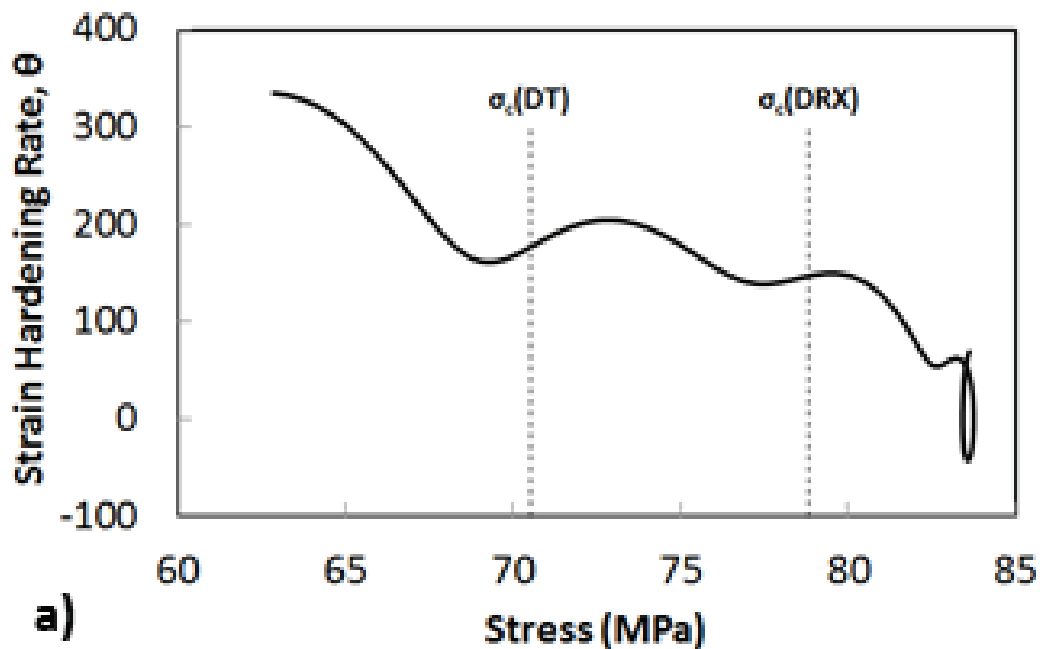


**Fig. 5.3.** Optical micrographs of the 0.06C-0.3Mn-0.01Si steel deformed to a strain of 0.7 applied at a strain rate of  $1s^{-1}$ . The samples were quenched immediately after straining at (a) 1050 °C, (b) 1150 °C and (c/d) 1200 °C.

## 5.4 Discussion

### 5.4.1 Critical Strains

The optical microstructures of Figure 5.3 indicate that significant volume fractions of ferrite were formed during deformation. Support for this interpretation was gained from the shapes of the stress-strain curves presented in Figure 5.2. This was accomplished by carrying out double differentiations [5.17] on the curves. For this purpose, the flow curves were fitted with 9<sup>th</sup> order polynomials (higher orders in some cases) using the MatLab software. Plots of  $\theta$  versus  $\sigma$  were then prepared, as shown in Figure 5.4a, where  $\theta$  is the strain hardening rate  $(\delta\sigma/\delta\epsilon)_\epsilon$ . The point of initiation of each softening mechanism is then identified by an inflection point in the  $\theta$ - $\sigma$  plot using the condition  $\delta/\delta\sigma(\delta\sigma/\delta\epsilon) = 0$ . The two minima in the  $-\delta\theta/\delta\sigma$  versus  $\sigma$  curve of Figure 5.4b identify the strains more precisely at which first DT and then DRX are initiated. Several previous studies [5.10,5.15] have shown that the critical strain for DT is less than that for DRX.



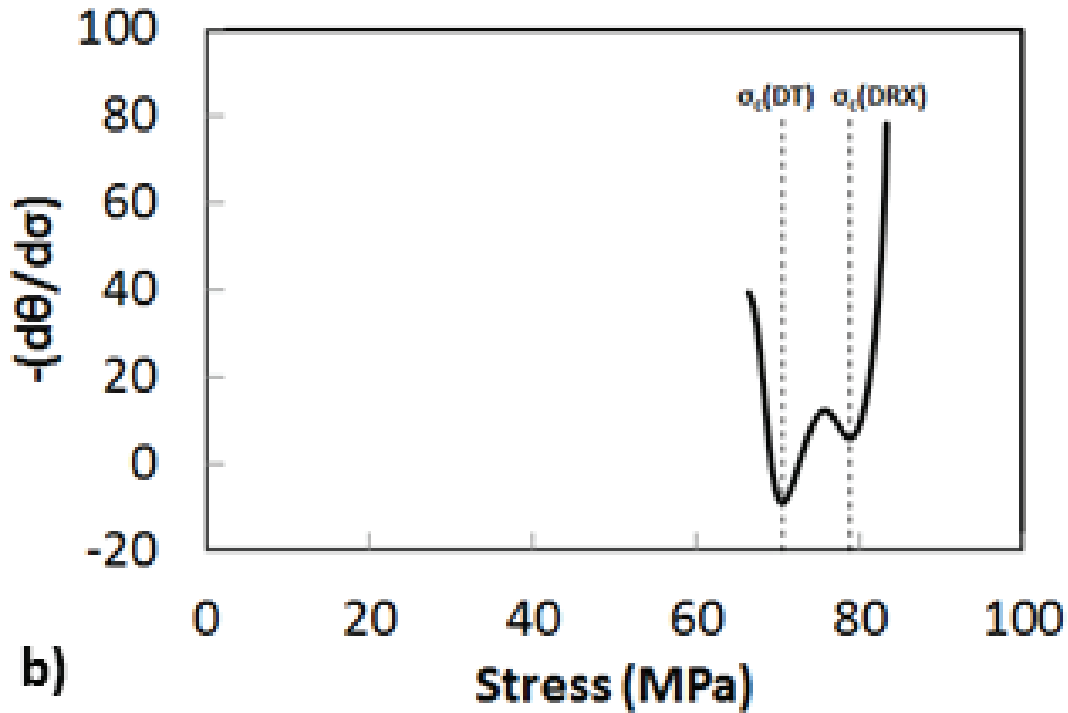


**Fig. 5.4.** (a) Plot of strain hardening rate ( $\theta$ ) versus  $\sigma$  derived from the flow curve of the compression test carried out at 1100 °C and  $1s^{-1}$ . The first and third inflection points in (a) are identified more clearly in (b) by the minima in the  $-(d\theta/d\sigma)$  versus  $\sigma$  curve.

The dependences on temperature of the critical strains and stresses are illustrated in Figure 5.5. In Figure 5.5a, the critical strain for DRX decreases with increasing temperature in the normal way since recrystallization occurs more easily at elevated temperatures. The DT critical strain normally *increases* with temperature, as denoted by the shape of the energy barrier [5.13] opposing the transformation (see below). However, the present results indicate that the DT critical strain actually *decreases* with temperature beyond 1080 °C. This is a trend that is discussed in greater detail below.

The free energy obstacle opposing dynamic transformation in the present material is illustrated in Figure 5.6. This was calculated using the FSstel database of the FactSage thermodynamic software [5.14]. It can be seen that the obstacle is highest at around 1080 °C and then the barrier decreases in height as the temperature for transformation into delta ferrite is approached. The increasing fractions of ferrite formed at temperatures above 1080 °C can thus be understood in terms of the shape of the energy barrier.

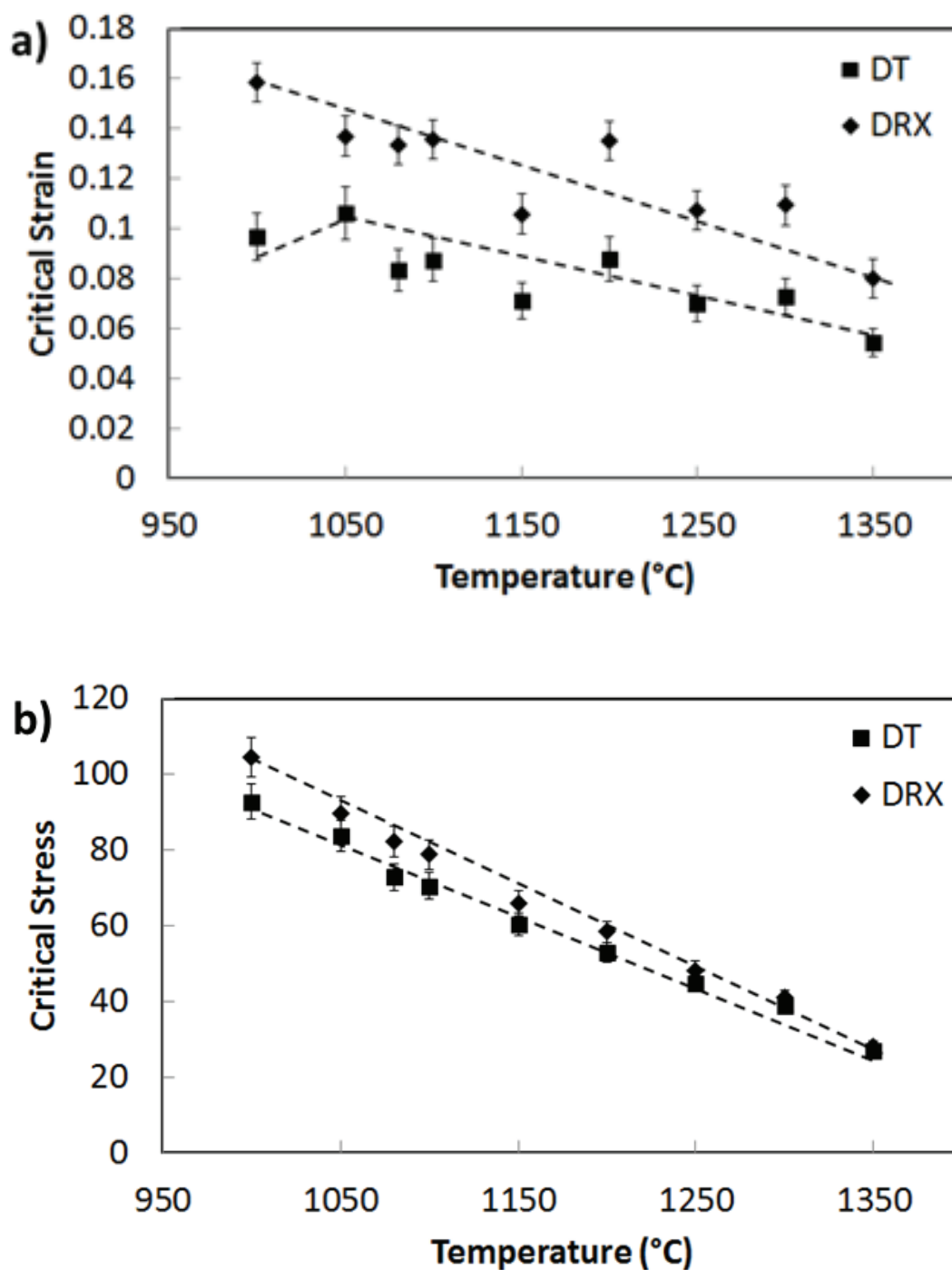


**Fig. 5.5.** (a) Critical strains for dynamic transformation (DT) and dynamic recrystallization (DRX) calculated by means of the double differentiation method. (b) The critical stresses associated with the critical strains of Fig. 5.5a.

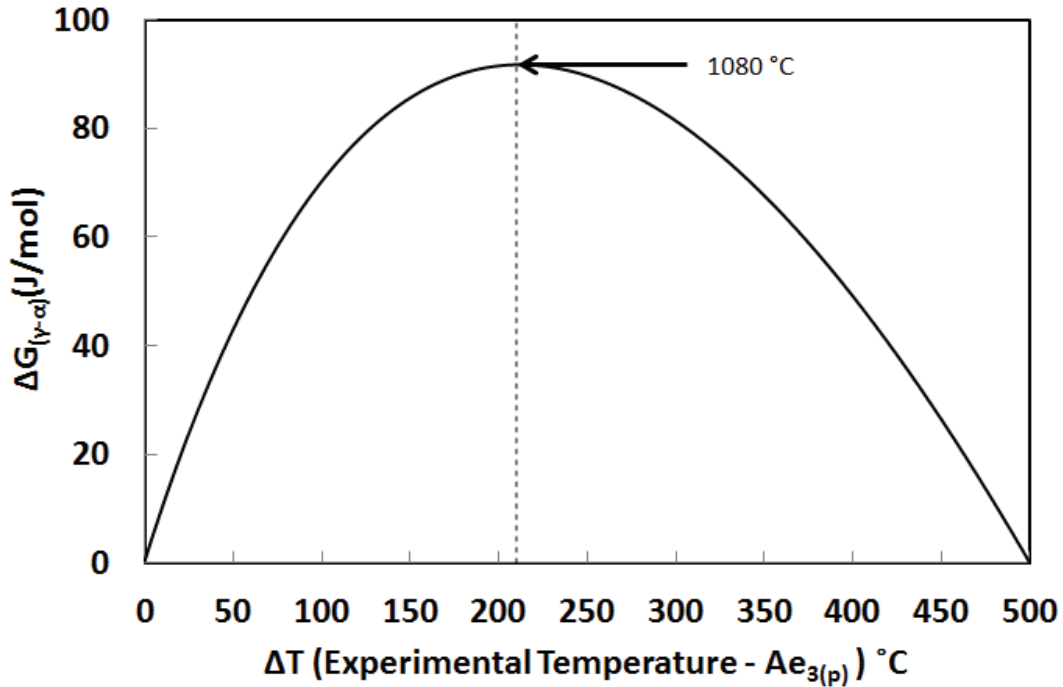


**Fig. 5.6.** Chemical free energy difference between austenite and ferrite ( $\Delta G_{\gamma-\alpha}$ ) as a function of  $\Delta T$  (experimental temperature – paraequilibrium  $A_{e3}$ )

#### 5.4.2 Formation of Widmanstätten ferrite

It was shown in an earlier study [5.13] that the volume fraction of ferrite decreases with increasing temperature just above the  $A_{e3}$ . However, the present investigation has revealed that this trend does not apply to temperatures high in the austenite phase field. More DT ferrite was observed to form as the temperature was progressively increased beyond the peak of the free energy obstacle. The nucleation and growth of DT ferrite leads to the formation of Widmanstätten ferrite plates approximately 200nm in width [5.5]. The displacive ferrite then gradually coalesces into polygonal grains during continued straining. As this process begins displacively, it differs from the conventional diffusional manner of forming polygonal ferrite that is observed below the  $A_{e3}$  temperature. The plates produced by deformation form in self-accommodating pairs of near-identical orientation [5.18]. Examples of such Widmanstätten colonies are illustrated in Figure 5.7.

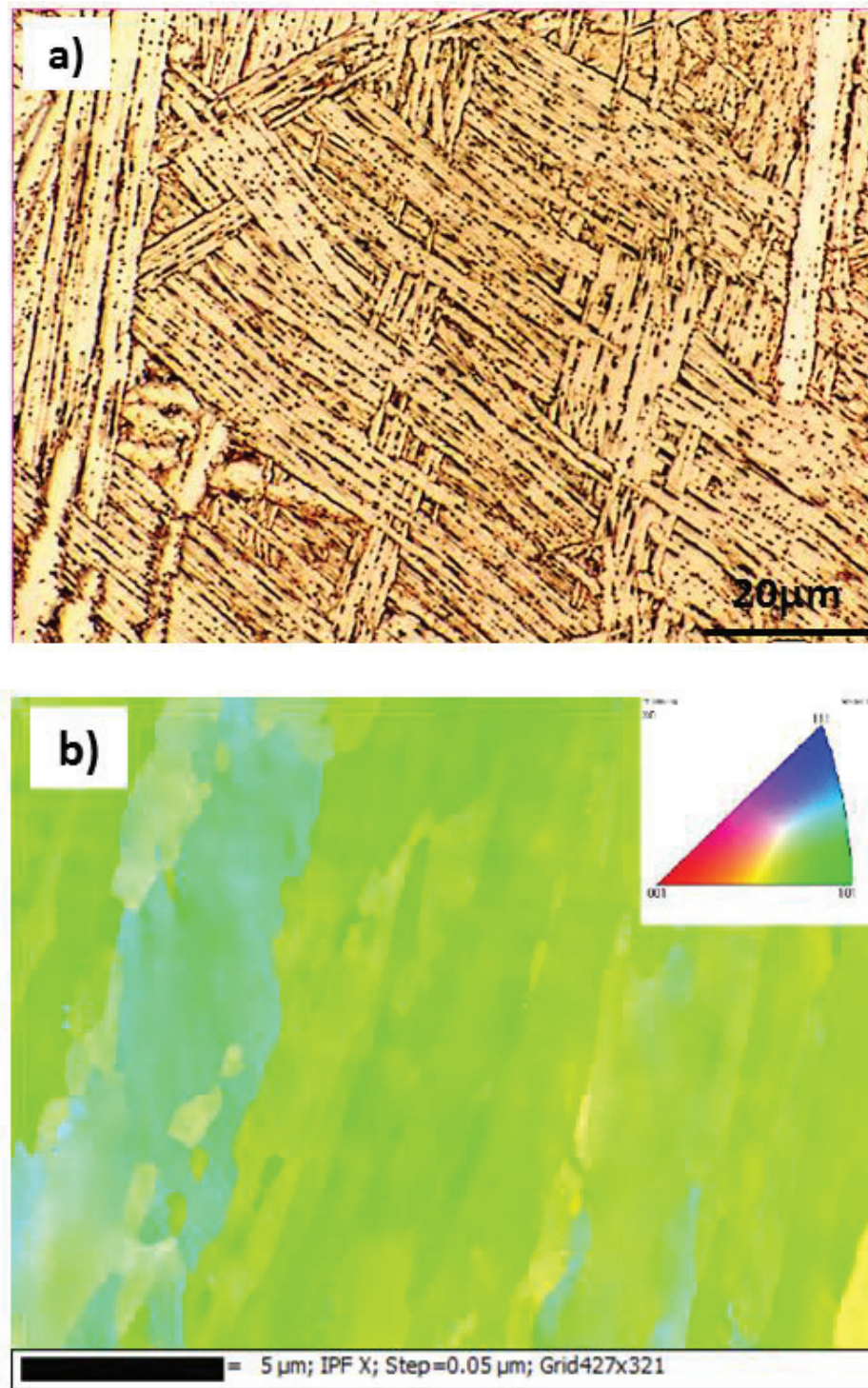


**Fig. 5.7.** Widmanstätten microstructures formed in a 0.06 wt%C steel strained to  $\epsilon = 0.7$  at a) 1350 °C ( $A_{e3(p)} + 480$  °C) and (b) 1200 °C ( $A_{e3(p)} + 330$  °C)

## 5.5 Conclusions

1. Both dynamic transformation and dynamic recrystallization can be seen to operate in the temperature range 1000 °C to 1350 °C. By employing the double differentiation method, these are seen to be initiated at average critical strains of about 0.10 and 0.15, respectively. The critical strain for DRX decreases monotonically with increasing temperature. By contrast, the critical strain for DT first increases with temperature up to about 1080 °C and then decreases as the delta ferrite formation temperature is approached.
2. These observations are consistent with the shape of the free energy barrier opposing the transformation, which rises to a maximum at 1080 °C. The shape of the barrier also explains why less ferrite is formed at temperatures in the vicinity of the peak than at higher or lower temperatures.
3. The present results indicate that Widmanstätten ferrite forms at temperatures as high as 1350 °C, i.e. 480 °C above the Ae<sub>3</sub> temperature. Thus it appears that DT ferrite can be formed at any temperature within the austenite phase field.

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## Chapter 6

### **Effect of Interpass Time on the Dynamic Transformation of a Plain C-Mn and a Nb Microalloyed Steel**

In most available studies regarding strip rolling, only dynamic recovery and recrystallization are taken into account as softening mechanisms. It was shown in the previous chapters that dynamic transformation and dynamic recrystallization occur concurrently during the straining of austenite above the  $A_{e3}$  temperature. Numerous researchers have demonstrated that dynamic transformation takes place above the  $A_{e3}$ , but these studies have mostly been limited to single pass deformations. Here 7-pass torsion simulations of strip rolling were carried out in order to establish the effects on rolling load of long and short interpass times. For additional realism, tests were carried out under both isothermal and continuous cooling conditions. It was shown that more ferrite is formed and retained when short interpass times are employed. In plain C steels, DT ferrite retransforms into austenite during the interpass intervals, as shown by the optical micrographs presented here. On the other hand, the addition of Nb retards both the forward and reverse transformation.

This section is composed of two parts, strip rolling simulations: i) without a roughing pass (preliminary results) and, ii) with a roughing pass (a more detailed analysis).

### 6.1 Strip Rolling Simulation without a roughing pass

- This section has been published: **C. Aranas Jr.** and J.J. Jonas, "Dynamic transformation of austenite to ferrite during the torsion simulation of strip rolling," Proceedings of the Materials Science and Technology (MS&T 2014) Conference, Pittsburgh, USA, (2014) 461-468.

#### Abstract

Torsion simulations of strip rolling were carried out on a 0.06%C-0.3%Mn-0.01%Si steel over the temperature range 930°C to 1000°C. Pass strains of 0.4 were applied at a strain rate of  $1\text{s}^{-1}$ . Interpass times of 0.5s, 1s, 1.5s and 3s were employed to determine the mean flow stresses (MFS's) applicable to strip rolling. By means of double differentiation, critical strains of 0.04 and 0.11 were established for the initiation of dynamic transformation and dynamic recrystallization, respectively. It is shown that the dynamic transformation of austenite to ferrite takes place above the  $A_{e3}$  temperature, resulting in significantly lower than expected MFS values. The nucleation of ferrite reduces the rolling load and modifies the microstructure. Shorter interpass times produce larger decreases in MFS than longer ones because of reduced recovery during short interpass intervals.

### 6.1.1 Introduction

The dynamic transformation (DT) of austenite to ferrite has attracted significant interest since the work of Yada and coworkers in the 1980's [6.1.1]. They showed by means of compression tests and rolling simulations, that austenite can transform into ferrite during deformation at temperatures as high as 193°C above the  $A_{e3}$ . Strip rolling simulations using interpass times of 1s were also carried out by Yada et al. in 1988 [6.1.2] leading to similar results. In a more recent paper in 2000, they used in-situ X-ray diffraction experiments to provide real time evidence for the occurrence of dynamic transformation [6.1.3]. In this work, iron alloys were subjected to torsion tests about 40°C above the  $A_{e3}$  temperature. The increasing presence of  $(110)_{\alpha-Fe}$  diffraction patterns revealed that ferrite was forming during deformation.

More recently, Ghosh and Basabe and the present authors reported on the results of hot torsion tests in which the characteristics of the ferrite produced dynamically during multiple successive deformations separated by short interpass times were described [6.1.4]. Two distinct softening mechanisms were identified by using the double differentiation method developed by Poliak and Jonas in 1996 [6.1.5]. These were shown to be dynamic transformation and dynamic recrystallization and were determined to be initiated at critical strains of 0.05 and 0.10, respectively [6.1.4].

### 6.1.2 Experimental Procedure

The present steel was supplied as hot rolled plates with a thickness of 12.5 mm. The chemical composition of this steel is presented in Table 6.1.1 together with its corresponding paraequilibrium and orthoequilibrium  $A_{e3}$  temperatures. The  $A_{e3}$  temperatures were calculated using the Thermo-Calc software.

**Table 6.1.1.** Chemical composition (mass%) and equilibrium transformation temperatures (°C)

| C    | Mn   | Si   | Orthoequilibrium $A_{e3}$ | Paraequilibrium $A_{e3}$ |
|------|------|------|---------------------------|--------------------------|
| 0.06 | 0.30 | 0.01 | 877 °C                    | 870 °C                   |

The steel plate was machined into torsion specimens with diameters of 6.3 mm and gauge lengths of 22.2 mm with the cylindrical axes parallel to the rolling direction.

Specially programmed torsion tests were performed for the simulation of strip rolling. The samples were heated to 1200°C at 1°C/s. After austenitization for 20 minutes, the samples were cooled at 1°C/s down to test temperatures in the range 1000°C to 930°C. They were then held for 10 minutes, after which multiple deformations were applied while the sample was cooled at approximately 6 °C/s. The number of passes (with 1, 4 and 7 deformations) was varied so that samples could be quenched leading to quantification of the volume fraction of ferrite produced. Interpass times of 0.5, 1, 1.5 and 3 seconds were employed. The thermomechanical schedule followed in the torsion tests is illustrated in Figure 6.1.1. As the sample diameter increased during testing, the changes in diameter were taken into account in deriving the flow stress from the torque.

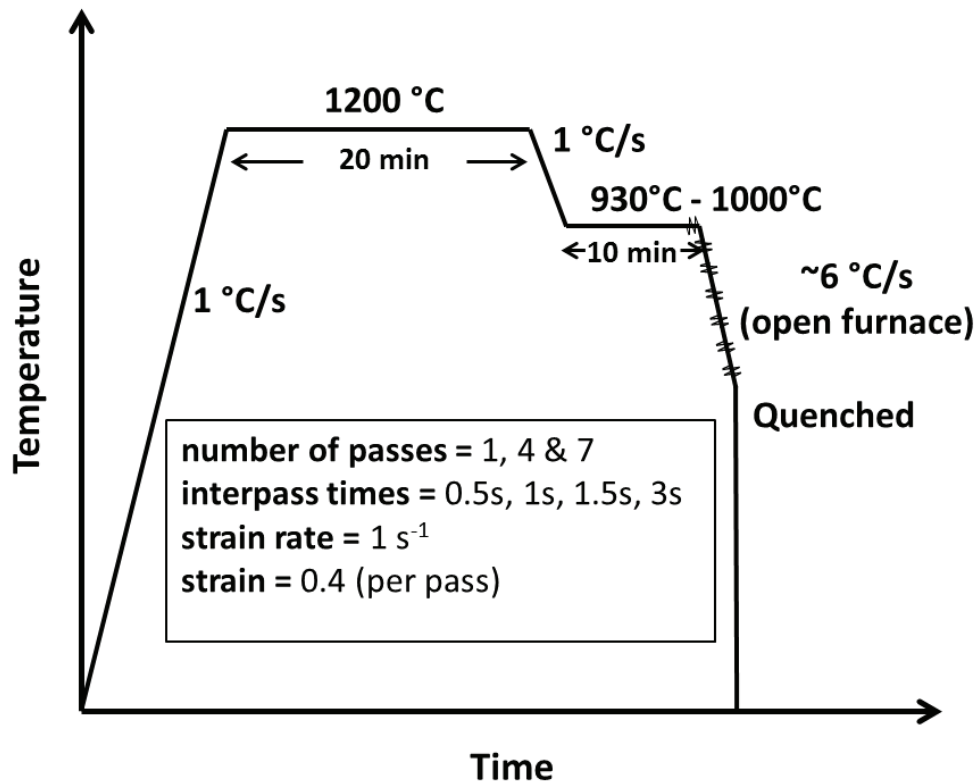
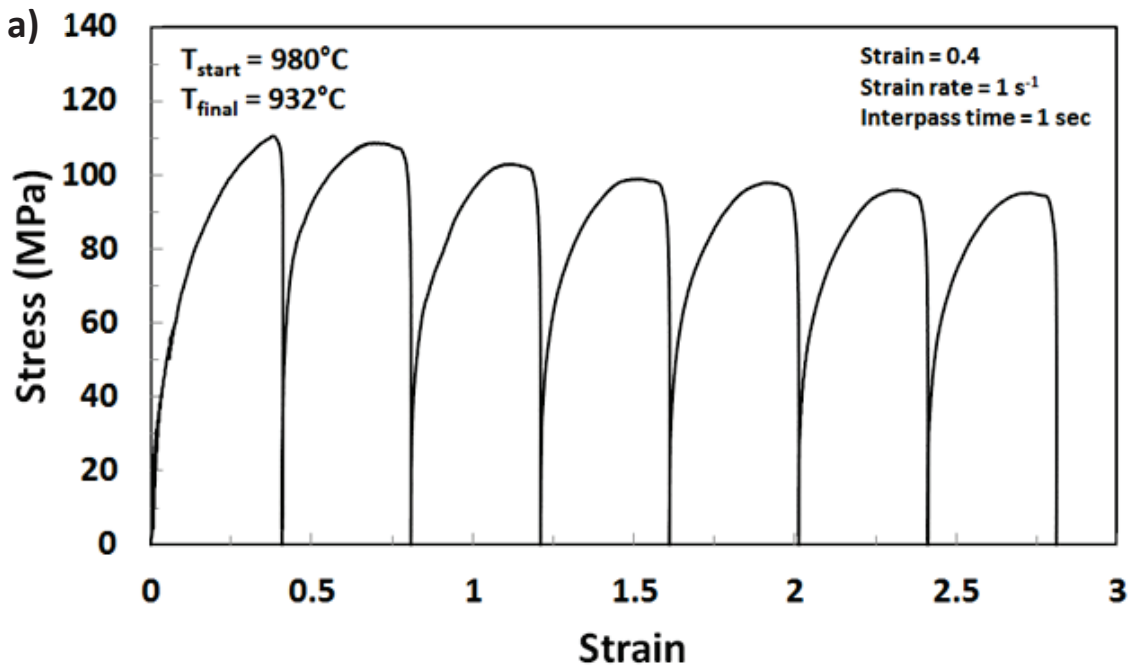


Fig. 6.1.1. Thermomechanical schedule of the torsion tests

### 6.1.3 Results

The stress-strain curves of a 7-pass strip rolling simulation with interpass pass times of 1s applied at a strain rate of  $1\text{ s}^{-1}$  are presented in Figure 6.1.2. Although the flow curve levels are normally expected to increase due to the decreasing temperature, it is clear that the levels either display low rates of increase or even decrease during the simulations. The mean flow stresses were calculated by integration of the stress-strain curves of Figure 6.1.2 and these values are plotted in the form of Boratto diagrams in Figure 6.1.3. The Pussegoda relation is also plotted here, which indicates the approximate value of the MFS for a series of first passes [6.1.6]. Although the MFS values are expected to rise as the temperature is decreased, here there is no increase in load even though the passes are being applied about  $100^\circ\text{C}$  above the  $A_{e3}$  temperature.



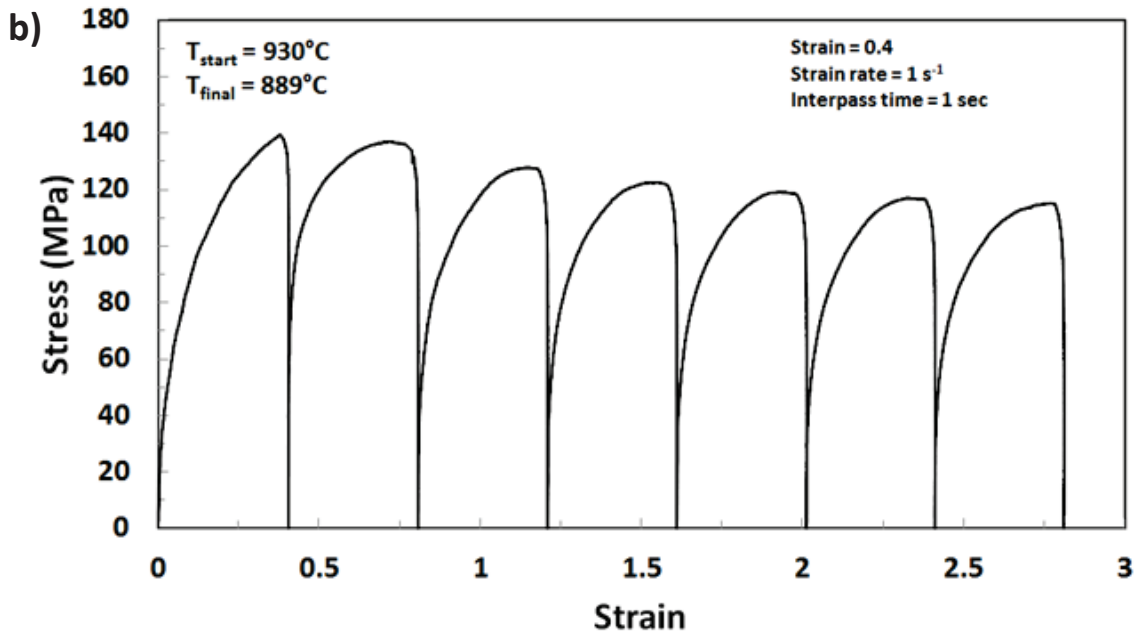


**Fig. 6.1.2.** Strip rolling simulation carried out on the 0.06C-0.3Mn-0.01Si steel with interpass times of 1s and pass strains of 0.4 applied at a strain rate of  $1 \text{ s}^{-1}$ . Start and finish temperatures were a)  $980^{\circ}\text{C}$  and  $932^{\circ}\text{C}$ , and b)  $930^{\circ}\text{C}$  and  $889^{\circ}\text{C}$ , respectively.

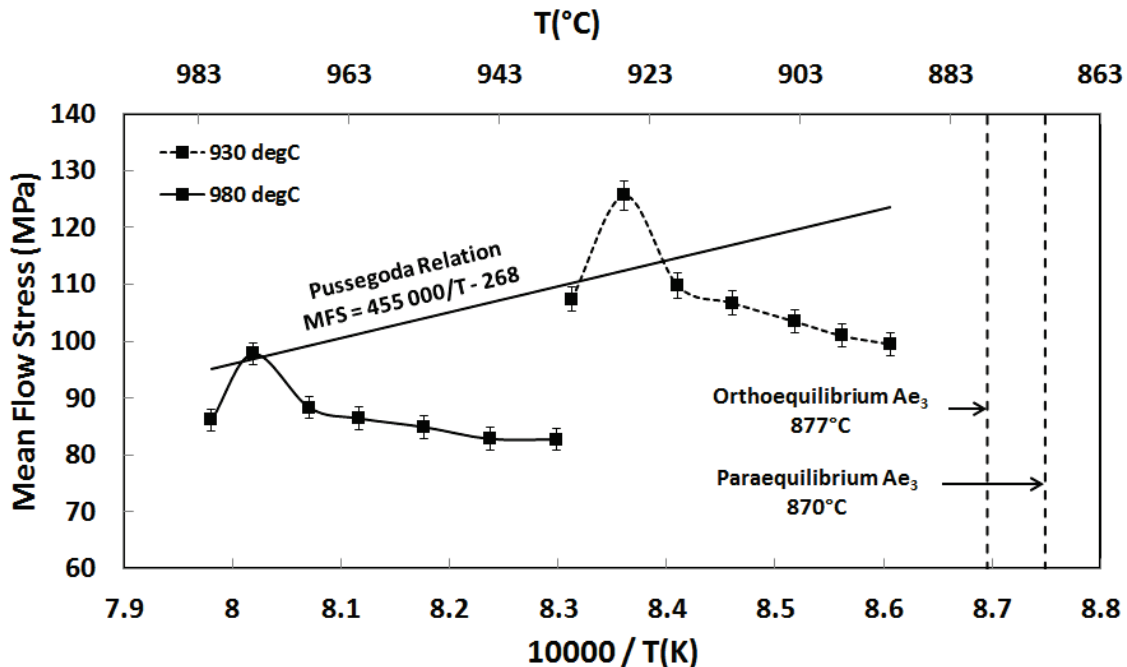
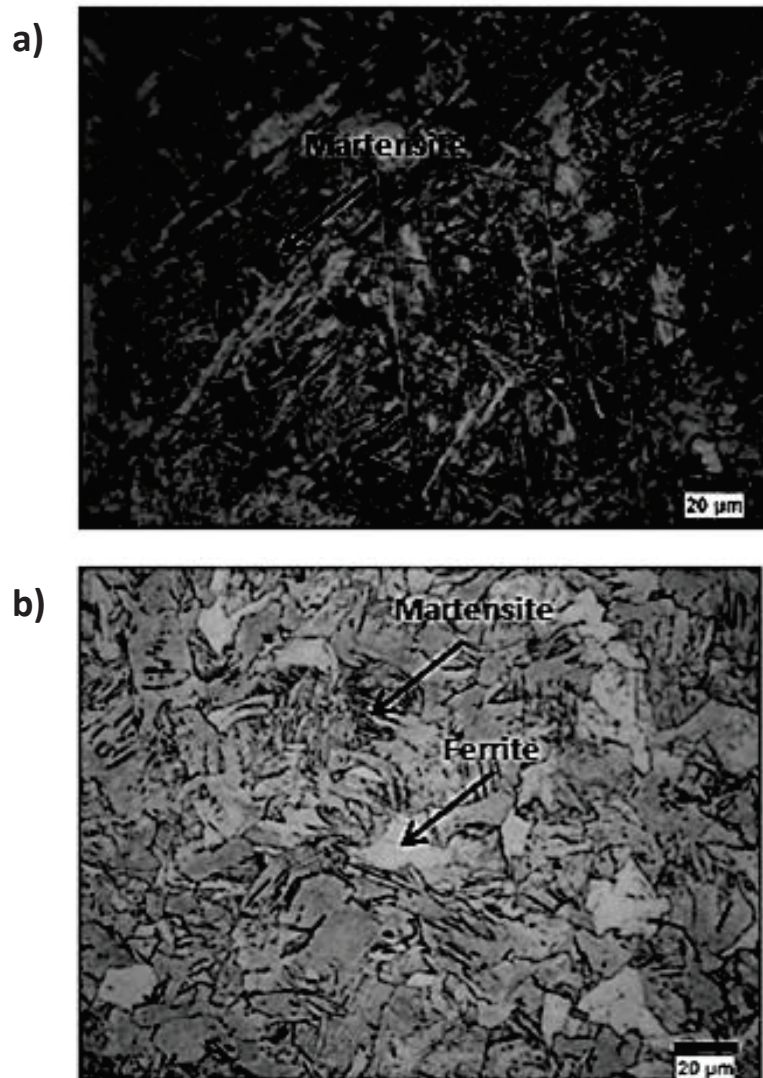
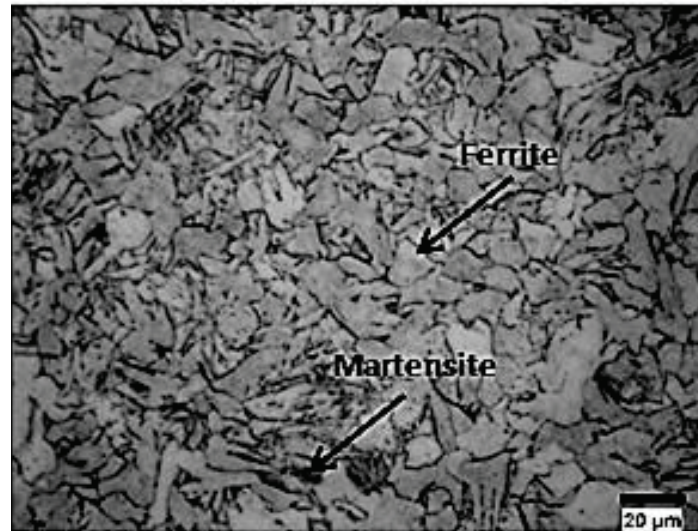


**Fig. 6.1.3.** The mean flow stresses associated with the flow curves in Figure 6.1.2. The MFS increases during the 2<sup>nd</sup> pass (due to strain accumulation) and then drops in the following passes because of the softening associated with the transformation from austenite to ferrite as well as with the occurrence of dynamic recrystallization.

In torsion simulations, the applied strain increases linearly from zero along the axis to a maximum at the surface of the sample. The increase in the volume fraction of ferrite formed with distance from the axis is illustrated in Figure 6.1.4. The martensite phase appears dark while the ferrite is light. It can be seen that only martensite is present along the axis and that the amount of ferrite increases significantly as the surface is approached. The microstructures also reveal that the volume fraction of ferrite increases progressively with pass number, as shown in Figure 6.1.5. Note that the center is still martensitic after 7 passes. This indicates that the critical strain for dynamic transformation has not yet been attained in these areas.



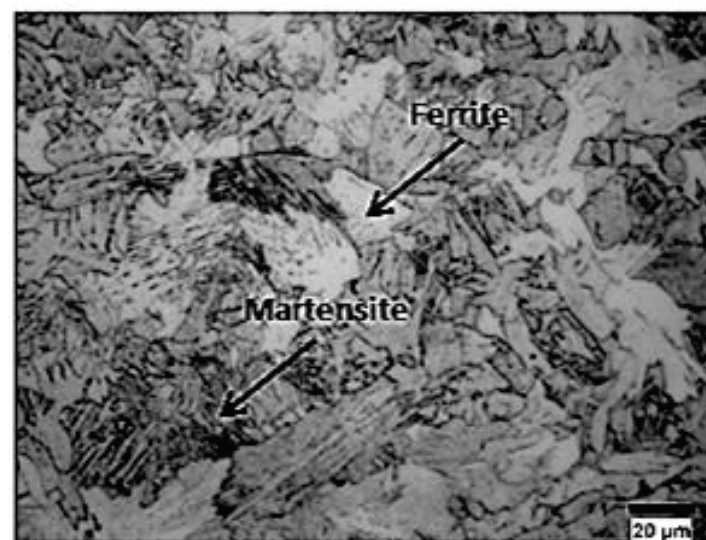
c)



d)



e)



f)

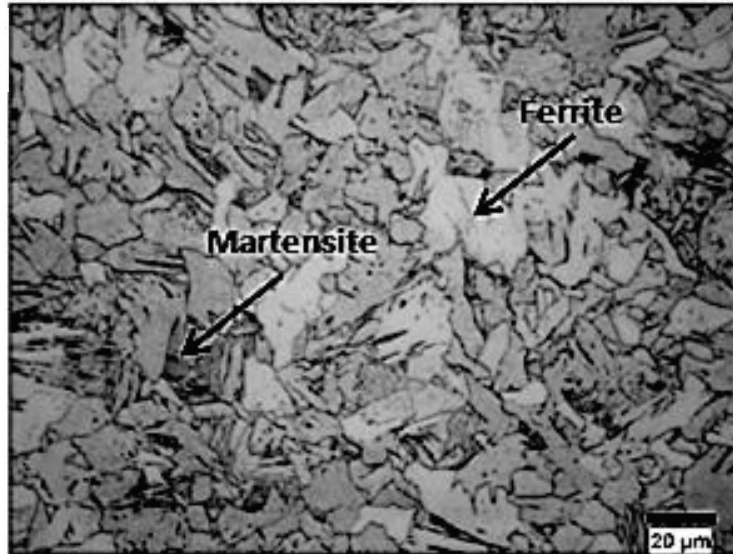
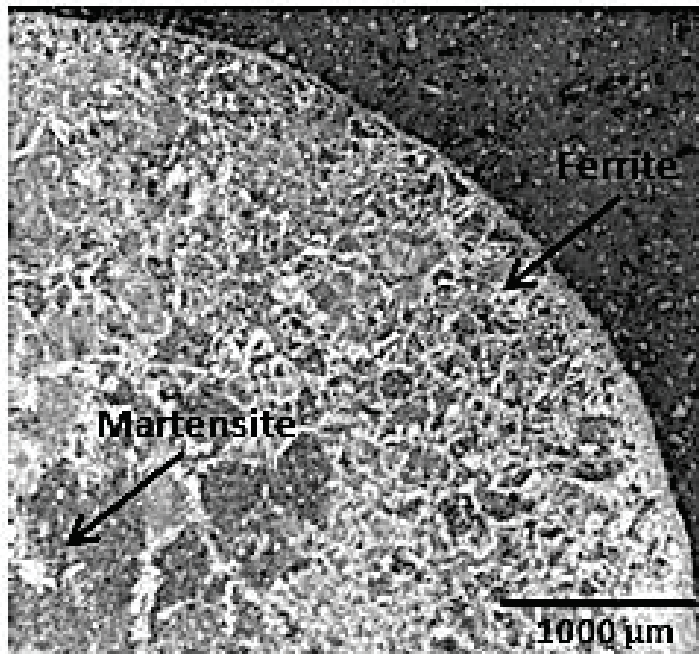


**Fig. 6.1.4.** Samples subjected to 7 deformations beginning at 980°C (a,b,c) and 930 °C (d,e,f). The amount of ferrite increases in moving from (a) the center to (b) 1.5mm from the surface, and finally to (c) the surface of the sample. A similar trend can be observed in moving from (d) the center, to (e) 1.5mm from the surface, and finally to (f) the surface of the sample.

a)



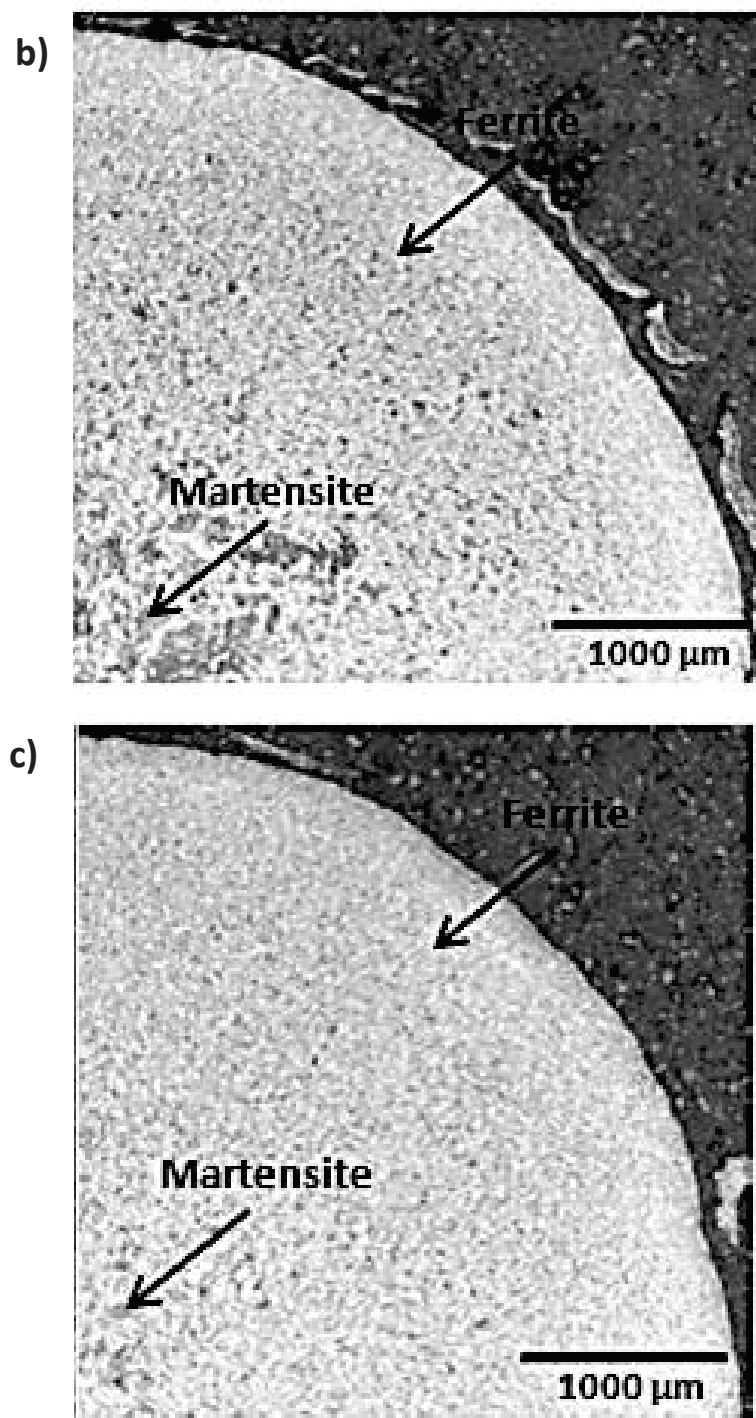
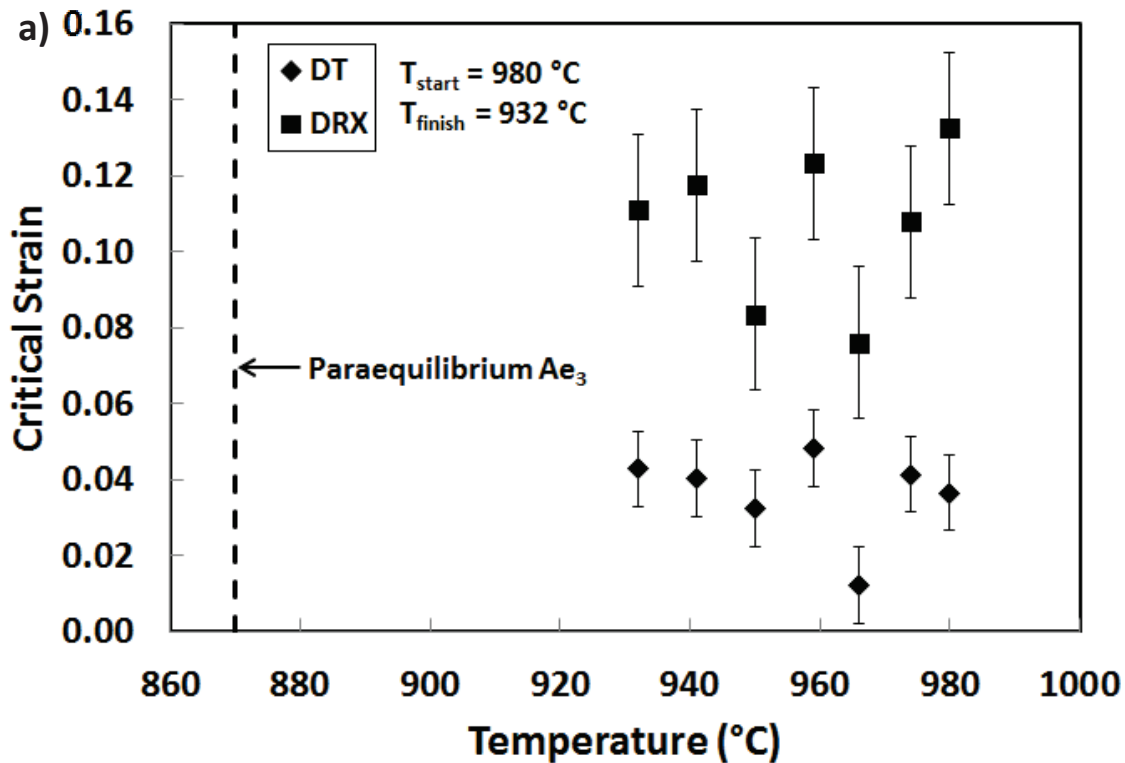


**Fig. 6.1.5.** Optical micrographs of the 0.06C-0.3Mn-0.01Si steel deformed with interpass times of 1s and pass strains of 0.4 applied at a strain rate of  $1s^{-1}$ . The torsion simulation began at 980°C and quenching was performed after (a) 1, (b) 4 and (c) 7 passes.

### 6.1.4 Discussion

#### 6.1.4.1 Critical Strains

The stress levels of the flow curves in Figure 6.1.2 do not increase with decreasing temperature. To investigate this unusual behavior, the double differentiation method of analysis was applied to these curves. It was verified that two distinct softening mechanisms were initiated at strains well before the peak strain. The first of these is dynamic transformation (DT), which converts some of the austenite into the softer ferrite phase, and the second is dynamic recrystallization (DRX). The critical strains required to initiate DT and DRX were determined to be around 0.04 and 0.11, respectively, as shown in Figure 6.1.6. Several previous studies [6.1.7-6.1.10] have shown that the critical strain for DT is less than that for DRX. The transformation is reinitiated during each simulated rolling pass. Note that transformation is not continuous with increasing strain but requires the application of a series of successive critical (initiation) strains. The occurrence of this phenomenon is supported by the optical microstructures presented in the previous section, Figures 6.1.4 and 6.1.5.



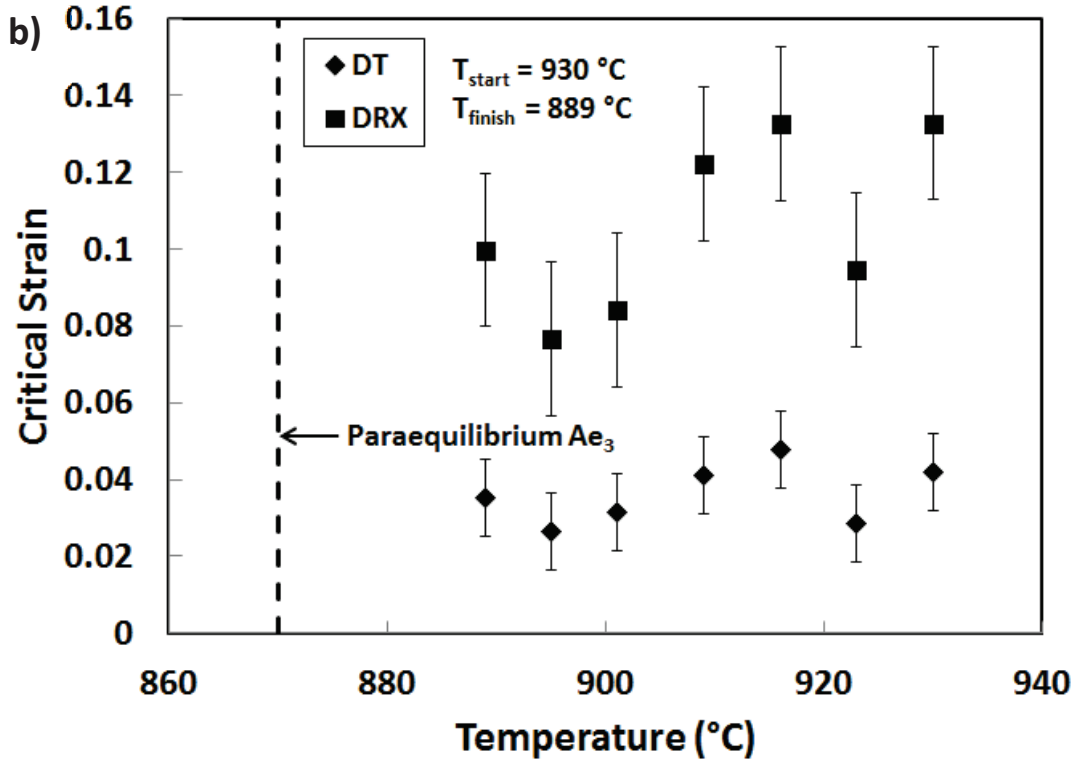


**Fig. 6.1.6.** Critical strains for dynamic transformation (DT) and dynamic recrystallization (DRX) in the strip rolling simulations of Figure 6.1.2. It can be seen that DT and DRX are initiated at strains of approximately 0.04 and 0.11.

#### 6.1.4.2 Mean Flow Stress

The mean flow stress (MFS) depends on the composition, temperature of deformation, strain and strain rate. It is important to note that, during strip rolling, the short times in the deformation zone (about 5 ms [6.1.10]) do not permit any partitioning of the substitutional alloying elements, in this case Mn and Si, and only carbon diffusion can take place. As the transformation during rolling is displacive, it occurs under paraequilibrium conditions [6.1.10]. The absence of MFS increase can thus be explained in terms of the ferrite formed above the  $Ae_3$ , as indicated by the optical micrographs presented in the previous section. Dynamic recrystallization also contributed significantly to the lack of MFS increase.

One of the major differences between strip rolling and plate rolling is the length of the interpass time. In the former case, the interpass times fall in the range 0.5s to 3s. Four sets of simulations were carried out with times of 0.5, 1.0, 1.5 and 3s to investigate the effect of increasing the interval. The MFS values determined in this way are illustrated in Figure 6.1.7. These tests were carried out using the same starting temperature, cooling rate, pass strain and strain rate. The shortest interpass time produced a flow stress decrease; as the interpass time was increased, the MFS values first displayed lesser decreases, before increasing slightly at 3s. These values should be compared with those specified by Pussegoda et. al. [6.1.6], which apply to single phase austenite. It can be estimated that a ferrite volume fraction of approximately 35% must have been formed in order to reduce the rolling load below the Pussegoda values by the amounts shown [6.1.4].

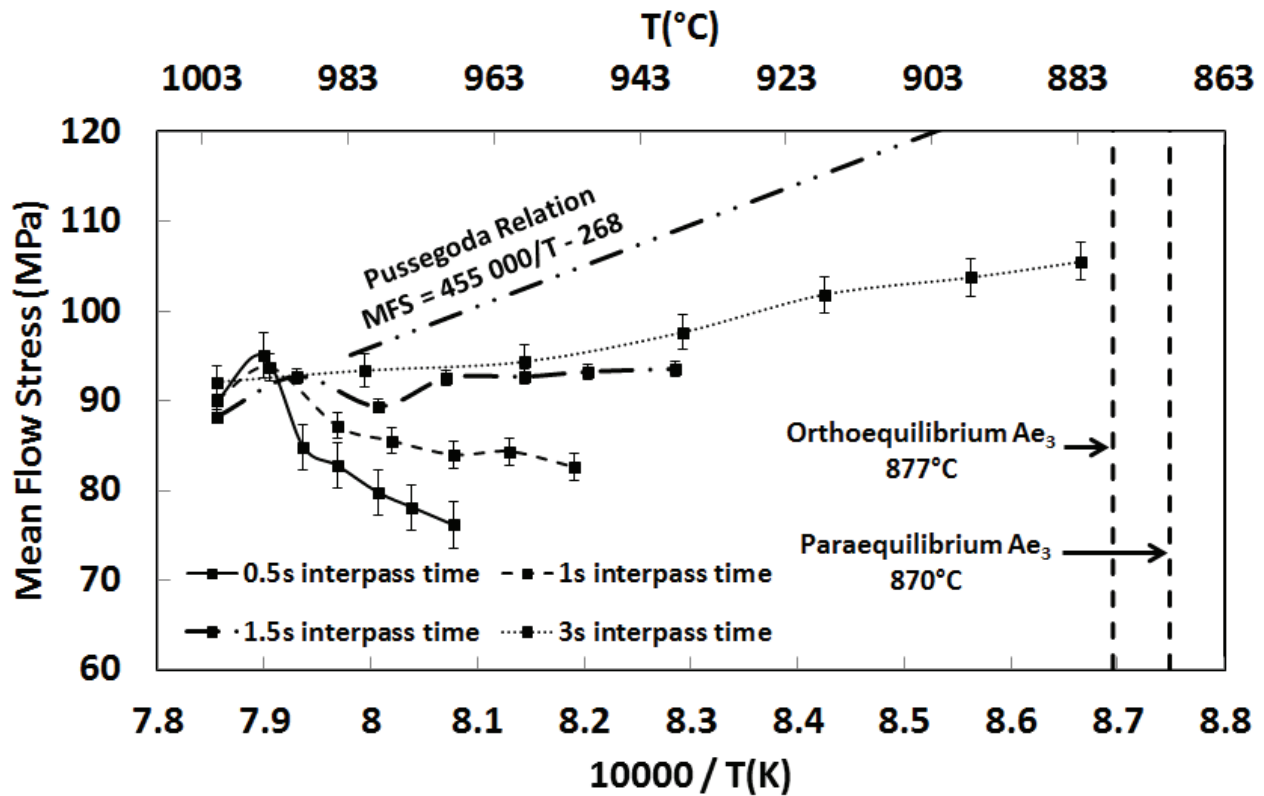


**Fig. 6.1.7.** Strip rolling simulations carried out on the 0.06C-0.3Mn-0.01Si steel with interpass times of 0.5s, 1s, 1.5s and 3s, and pass strains of 0.4 applied at a strain rate of  $1\text{s}^{-1}$ . The dependence of MFS on inverse temperature shows that short interpass times are associated with significant decreases in rolling load, indicating that ferrite is forming well above the  $\text{Ae}_3$  temperature.

In order to gain more insight into the influence of interpass time on the formation of DT ferrite, some isothermal 7 deformation strip rolling simulation were carried out at 1000°C with 1s and 3s interpass intervals. The results obtained in this way are illustrated in Figure 6.1.8. Here it can be seen that the 1s interpass times lead to about a 20% decrease in rolling load after 7 passes. The optical microstructures illustrated in Figure 6.1.8 pertaining to the last pass confirm that there is considerably more DT ferrite in the 1s than in the 3s material. The smaller proportion of ferrite produced in the 3s schedule is probably caused by reverse transformation during longer interpass times, a possibility that is under current investigation.

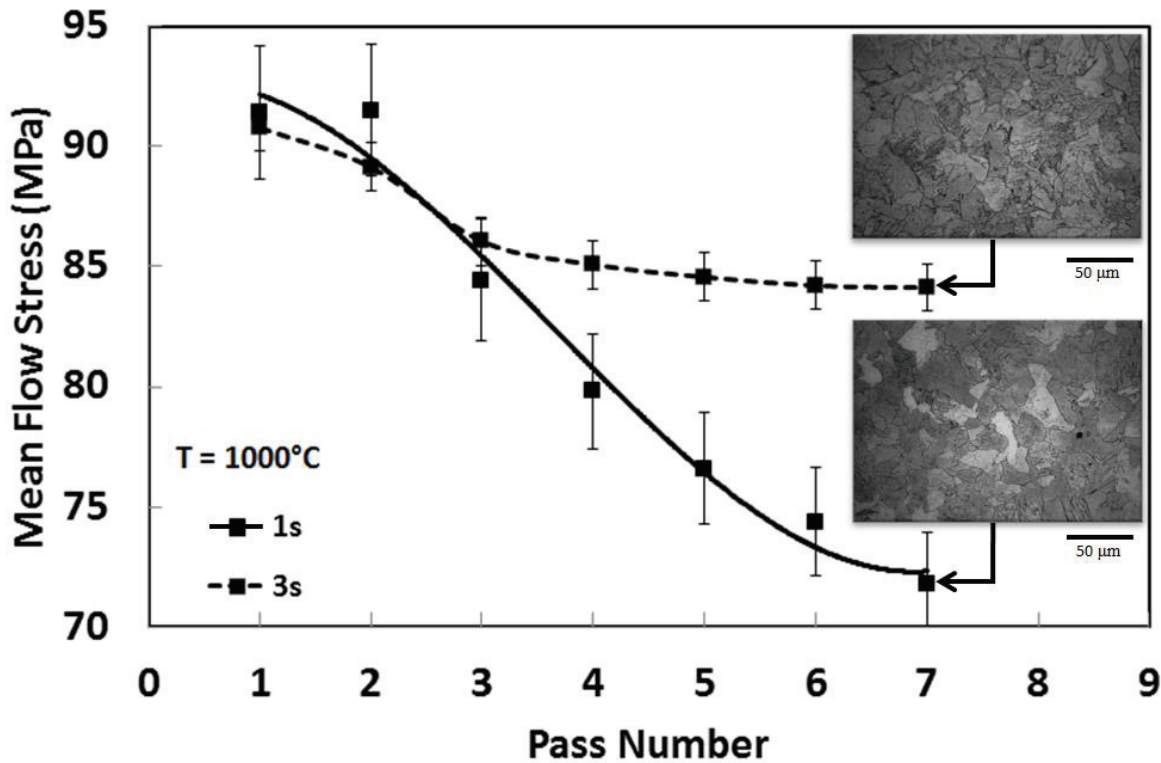


**Fig. 6.1.8.** Strip rolling simulations carried out under isothermal conditions on the 0.06C-0.3Mn-0.01Si steel with interpass times of 1s and 3s, and pass strains of 0.4 applied at a strain rate of  $1s^{-1}$  at 1000°C. The MFS values decrease continuously with pass number for the 1s simulation. The flow stress decrease is much smaller when the interpass is increased to 3s, indicating that the additional holding time leads to some retransformation to austenite. Samples were quenched right after the last pass and revealed the presence of more DT ferrite in the 1s as opposed to the 3s interpass material.

### 6.1.5 Conclusions

1. Two softening mechanisms were identified as occurring in strip mill simulations by applying the double differentiation method. Dynamic transformation and dynamic recrystallization are initiated at critical strains of 0.04 and 0.11, respectively. The dynamic transformation of austenite to ferrite above the  $A_{e3}$  temperature took place during all the simulated passes resulting in significant decreases in MFS.
2. Appreciable differences appear between 1s and 3s strip mill simulations. These are apparent from the MFS plots as well as from the optical microstructures of samples subjected to isothermal simulations. Appreciable ferrite forms during the shorter strip mill interpass time simulations, producing the softening. Less ferrite appears to form during long intervals and the DT ferrite produced probably retransforms into austenite.
3. Significant amounts of ferrite form during strip mill simulations and the volume fraction increases continuously until the last pass. As a result, there is little increase in MFS as the temperature is decreased.

### 6.1.6 References

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## 6.2 Strip Rolling Simulation with a roughing pass

- This section has been published: **C. Aranas Jr.**, T. Wang and J.J. Jonas, "Effect of Interpass Time on the Dynamic Transformation of a plain C-Mn and a Nb Microalloyed Steel," ISIJ International, 55 (2015) 647-654.

### Abstract

Strip rolling simulations were carried out on a 0.06%C-0.3%Mn-0.01%Si and a 0.11%C-1.0%Mn-0.11%Si-0.03%Al-0.034%Nb steel over the temperature range 1000°C to 883°C. Pass strains of 0.4 were applied at a strain rate of  $1\text{s}^{-1}$  with interpass times of 0.5s, 1s, 1.5s, 3s and 5s. Two different temperature schedules were employed, namely i) continuous cooling at 6°C/s and ii) isothermal holding. The mean flow stresses (MFS's) applicable to strip rolling were determined by integration. The flow stress levels and MFS's decrease when the interpass times are short. When they are long, the flow stress increases with decreasing temperature. These observations indicate that the austenite is transforming dynamically into ferrite and statically into austenite. The nucleation and growth of the ferrite reduce the rolling load and modify the microstructure. The addition of Nb in solid solution delays the occurrence of dynamic transformation (and the retransformation of ferrite back into austenite). The forward nucleation of ferrite occurs displacively while the retransformation back into austenite takes place by a diffusional mechanism.

### 6.2.1 Introduction

Deformation below the  $A_{e3}$  and above the  $A_{r3}$  temperature is known to provoke strain-induced transformation and to produce ferrite grain sizes of 1-3  $\mu\text{m}$  (or even less) [6.2.1]. However, transformation can also be initiated *above* the  $A_{e3}$ , as shown by the work of Yada and coworkers in the 1980's [6.2.2], but under these conditions, it is dynamic and not static. By means of compression tests, they demonstrated that the dynamic transformation (DT) of austenite can be made to take place as much as 166°C above the  $A_{e3}$  temperature. Strip rolling simulations using interpass times of 1s were also performed by Yada et al. in 1988 [6.2.3]. These experiments were again carried out above the  $A_{e3}$  temperature and led to similar results. They showed that the reverse transformation took place on isothermal holding after deformation. By using a dilatometer, they ascertained that the volume fraction of dynamically transformed ferrite decreased with holding time until all the DT ferrite was completely transformed back into austenite.

Earlier simulations of strip rolling had already indicated that the length of the interpass time can have a significant effect on the rolling load [6.2.4-6.2.7]. In 1988, Samuel et. al. [6.2.8] showed that short interpass times lead to significant reductions in MFS. The rolling simulations of Bowden et al. [6.2.9] carried out on microalloyed steels indicated that Nb(C,N) precipitation retarded both dynamic (DRX) and static (SRX) recrystallization. However, in these investigations, no consideration was given to the possibility that austenite was transforming dynamically into ferrite.

More recently, it has again been demonstrated that dynamic transformation can take place during deformation above the  $A_{e3}$  [6.2.10,6.2.11]. The critical strain required to initiate this phenomenon has been shown to be around 0.12 [6.2.12, 6.2.13]. These values were calculated using the double differentiation method [6.2.13, 6.2.14]. By means of rolling simulations, dynamic transformation has also been linked with low rates of MFS increase and even decreases in some cases [6.2.15]. Nevertheless, the exact influence of interpass time on this phenomenon and more particularly on the amount and nature of the softening remains to be clarified. In order to investigate these effects, two alloys were selected for the current study: i)

a plain C-Mn and ii) a Nb microalloyed steel. Here torsion simulations of strip rolling were carried out with the specific aim of establishing the effects on rolling load of long and short interpass times. The results obtained are described in detail below.

## 6.2.2 Experimental Procedure

A plain C-Mn steel and a Nb microalloyed steel were investigated in the present study. These were supplied in the form of hot rolled plates. The latter were machined into torsion specimens with diameters of 6.3 mm and gauge lengths of 22.2 mm with the cylinder axes parallel to the rolling direction. The chemical compositions of these steels, including their paraequilibrium and orthoequilibrium  $Ae_3$  temperatures, are listed in Table 6.2.1. The latter were calculated using the FSstel database of the FactSage thermodynamic software [6.2.16].

**Table 6.2.1.** Chemical compositions (mass%) and equilibrium transformation temperatures (°C)

|                        | <b>C</b> | <b>Mn</b> | <b>Si</b> | <b>Nb</b> | <b><math>Ae_{3(p)}</math></b> | <b><math>Ae_{3(o)}</math></b> |
|------------------------|----------|-----------|-----------|-----------|-------------------------------|-------------------------------|
| <b>Plain C-Mn</b>      | 0.06     | 0.30      | 0.01      | -         | 870 °C                        | 877 °C                        |
| <b>Nb microalloyed</b> | 0.11     | 1.0       | 0.11      | 0.034     | 827 °C                        | 843 °C                        |

### 6.2.2.1 Torsion Testing

Two contrasting temperature profiles were employed. In the first, the samples were air cooled at about 6°C/s during deformation, while in the second, all straining and holding was done isothermally. The torsion tests were conducted using a servo-hydraulic, computer controlled MTS torsion machine equipped with a horizontal radiation furnace and a temperature controller. The torque/twist curves were converted into stress-strain curves using the Fields and Backofen [6.2.17] formula. A thermocouple was welded to each sample to allow the deformation temperatures to be tracked with accuracy. A protective argon atmosphere was used in all tests to minimize oxidation and decarburization during heating. It is important to note that, as the sample diameter increases during free-end torsion testing (while it also shortens), the changes in diameter were taken into account in deriving the flow stresses from the torques [6.2.18].

In the continuous cooling schedule, shown in Figure 6.2.1a, the samples were heated at  $1^\circ\text{C/s}$  from room temperature to the austenitization temperature of  $1200^\circ\text{C}$ . They were then held for 20 minutes before cooling at  $1^\circ\text{C/s}$  down to the roughing temperature of  $1100^\circ\text{C}$ . After holding for 60 seconds at this temperature, the samples were strained to  $\epsilon = 0.4$  at a strain rate of  $1 \text{ s}^{-1}$  to simulate a roughing pass. After this, samples were held at this temperature for 90 seconds to allow recrystallization to take place before cooling down to  $1000^\circ\text{C}$ . They were then held at this temperature for seven minutes before beginning the 7-pass deformation schedule in which pass strains of 0.4 were applied at a strain rate of  $1 \text{ s}^{-1}$ . Tests were carried out using interpass times of 0.5s, 1s, 1.5s and 3s. The samples were deformed while being cooled at approximately

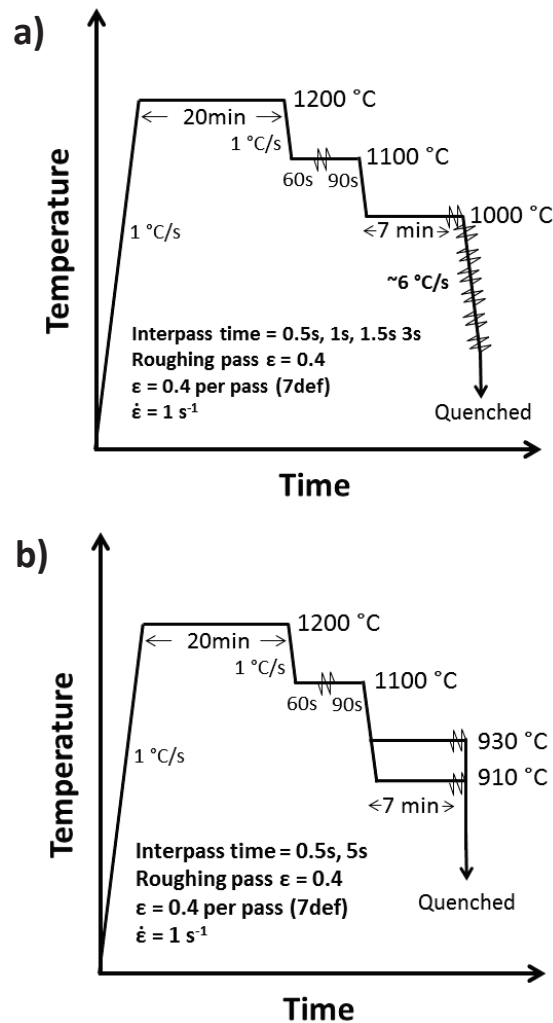


**Fig. 6.2.1.** Torsion testing schedules for the simulation of strip rolling employing: a) continuous cooling at  $6^\circ\text{C/s}$ , and b) isothermal conditions. All the deformation temperatures are above the respective  $A_{e3}'$ s.

6 °C/s. This was accomplished by lifting and opening the upper half of the radiation furnace. The first simulated pass was applied at exactly 1000°C, while the temperatures of the succeeding passes depended on the interpass time, as the cooling rate was held constant. Finally, samples were quenched right after the last pass.

For the isothermal schedule, shown in Figure 6.2.1b, the same heating and cooling rates, holding times, austenitization and roughing pass temperatures, as well as applied strains and strain rates were employed as in the continuous cooling schedule of Figure 6.2.1a. Only the finishing temperatures were different: 910°C and 930°C for the plain C-Mn and Nb-microalloyed steels, respectively. These tests were conducted isothermally with interpass times of 0.5 and 5s. After completing the seventh pass, the samples were quenched in three different ways; i) directly quenched after the last deformation, ii) quenched after 8s of holding at the test temperature, and iii) quenched after 15s of holding. This procedure was adopted for assessment of the retransformation behavior.

#### **6.2.2.2 Metallography**

The torsion samples were cut perpendicular to the longitudinal axis to provide cross-sections for microstructural examination. These were mounted using a hot phenolic mounting material. The mounted samples were ground using 400, 600, 800 and 1200 grit silicon carbide papers lubricated with water. Both 3 and 1 µm diamond suspensions were used for fine grinding before final polishing with a 0.02 µm colloidal silica suspension. The surfaces were etched with 2% nital to reveal the microstructure. This was followed by applying a 10% aqueous metabisulfite ( $\text{Na}_2\text{S}_2\text{O}_5$ ) solution to provide contrast between the martensite and the ferrite.

### **6.2.3 Results**

#### **6.2.3.1 Strip Rolling Simulations**

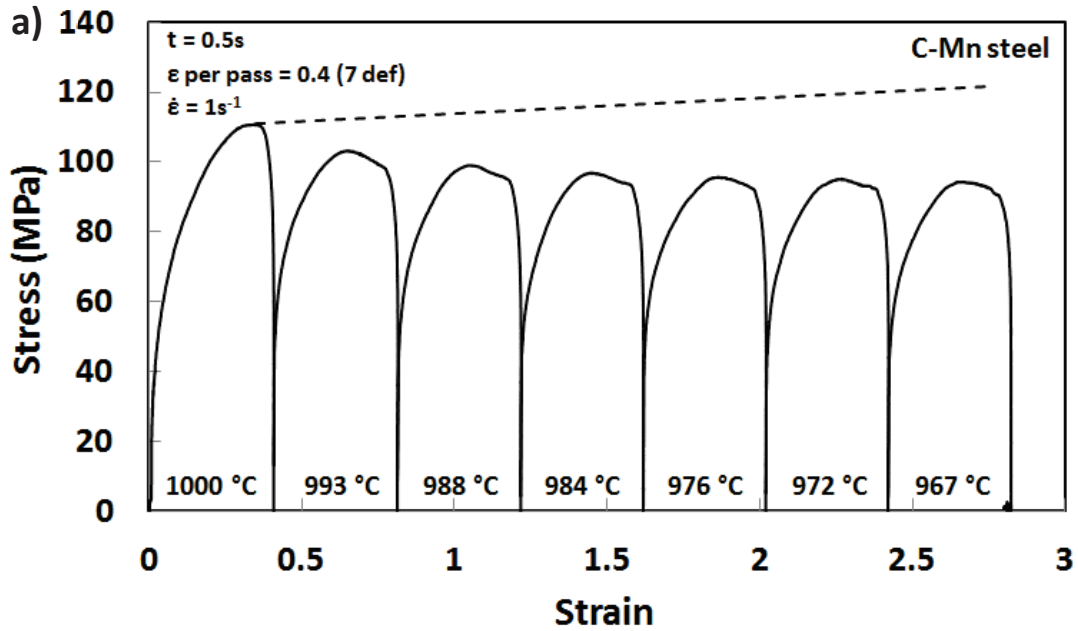
The continuous cooling flow curves determined at a strain rate of  $1\text{s}^{-1}$  on the plain C steel are displayed in Figure 6.2.2. Here interpass times of 0.5s (Figure 6.2.2a), 1s (Figure 6.2.2b), 1.5s

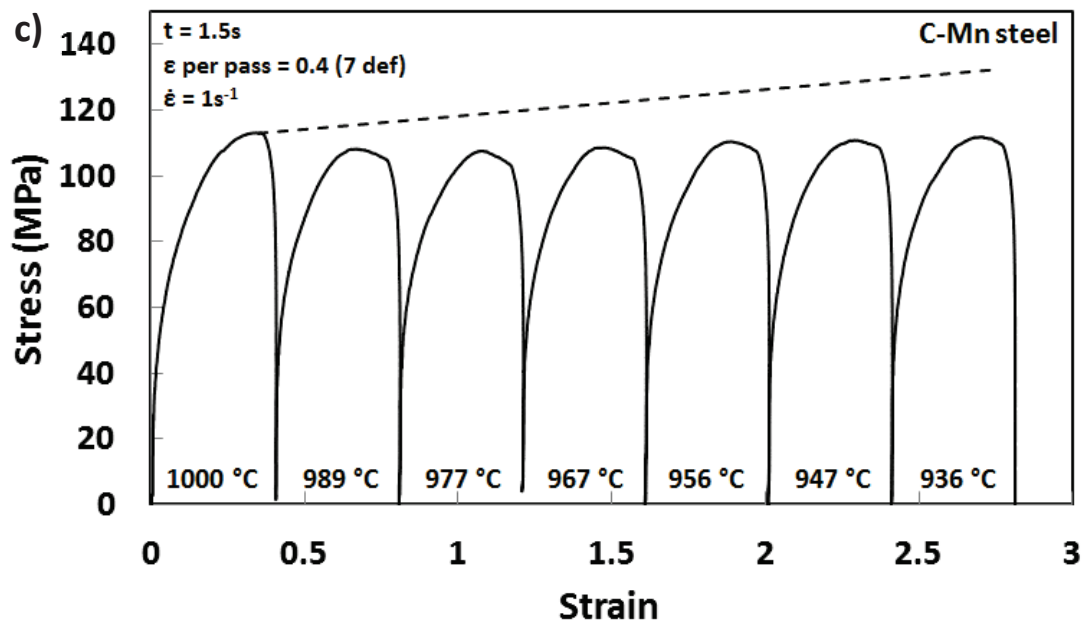
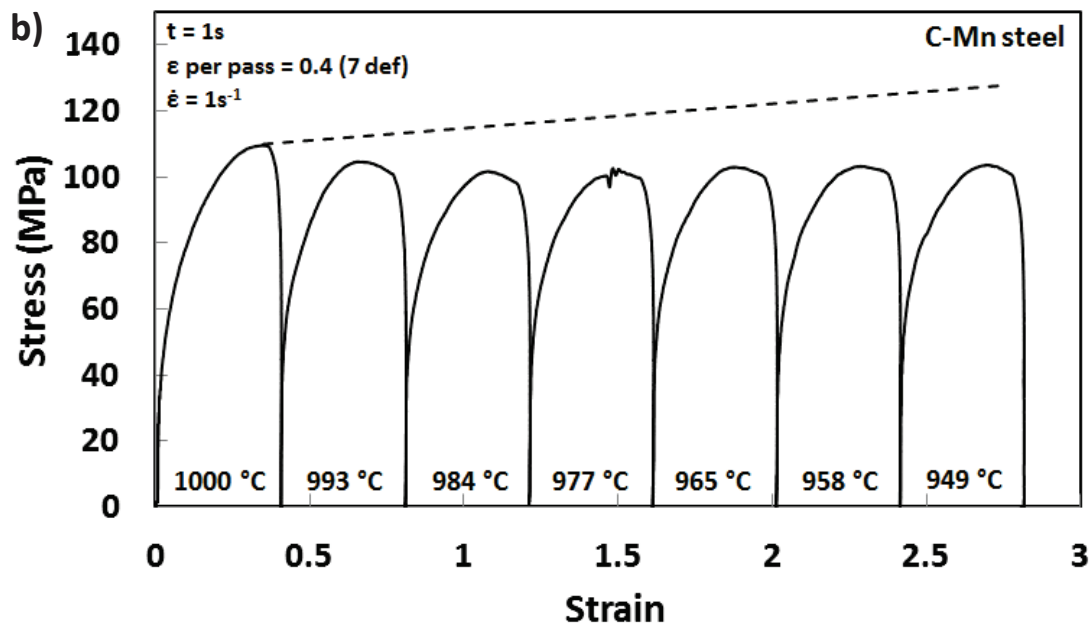
(Figure 6.2.2c) and 3s (Figure 6.2.2d) were employed. The temperature of each pass is indicated on each flow curve and it should be noted that all the deformation temperatures are above the  $A_{e3}$  for this steel, see Table 6.2.1.

In the absence of a phase change, the levels of the flow curves should increase from pass to pass because of the decreasing temperature. For reference, the broken line represents the expected first pass peak stress under these conditions. This was deduced from the first pass flow curves of rolling simulations conducted at a number of decreasing temperatures to strains of 0.10 (not shown here). This strain is lower than the critical strains for DT and DRX [6.2.12, 6.2.13]. The slope was then normalized using the relation:

$$\left(\frac{1}{\sigma_{0.1}}\right) \frac{d \ln \sigma_{0.1}}{d(1/T)} = \left(\frac{1}{\sigma_p}\right) \frac{d \ln \sigma_p}{d(1/T)} \quad (6.2.1)$$

where  $\sigma_{0.1}$  and  $\sigma_p$  are the respective flow stresses applicable to applied strains of 0.1 and the peak strain, respectively.





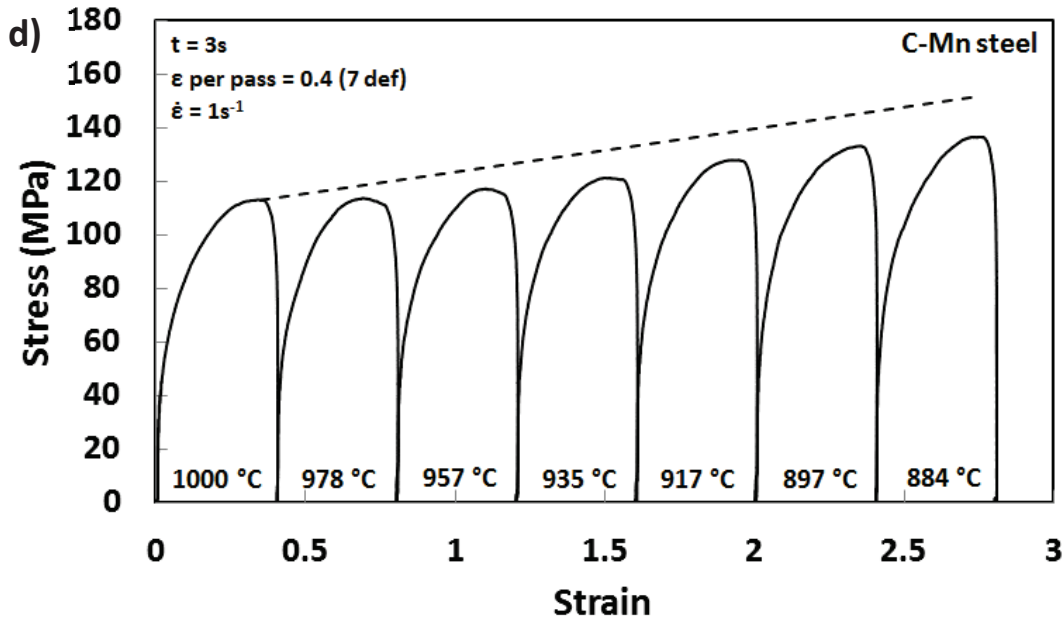


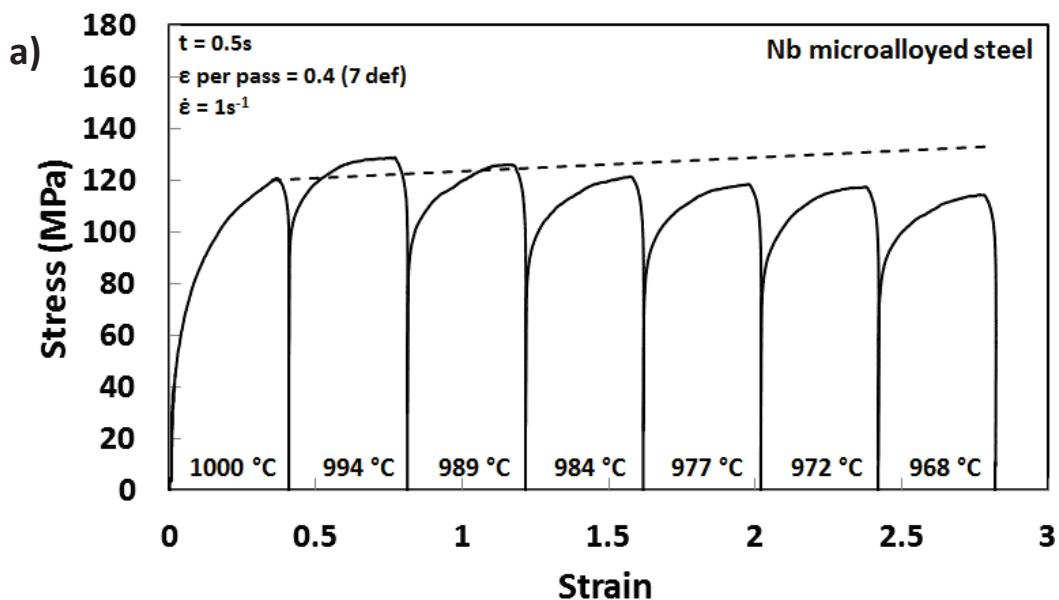
**Fig. 6.2.2.** Stress-strain curves determined on the plain C steel using the 7-pass continuous cooling schedule of Figure 6.2.1a with pass strains of 0.4 applied at  $1s^{-1}$ . Interpass times of a) 0.5s, b) 1s, c) 1.5s, and d) 3s were employed. The broken line indicates the measured dependence of stress level on temperature in the absence of DT and DRX. The levels of the shorter interpass flow curves decrease more perceptibly than those associated with the longer times.

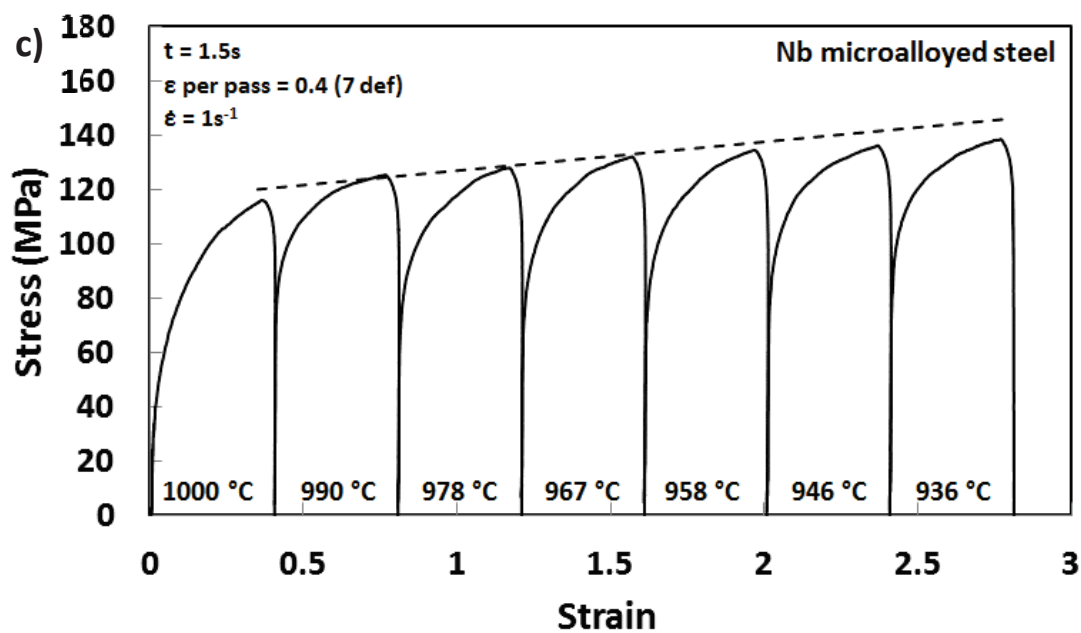
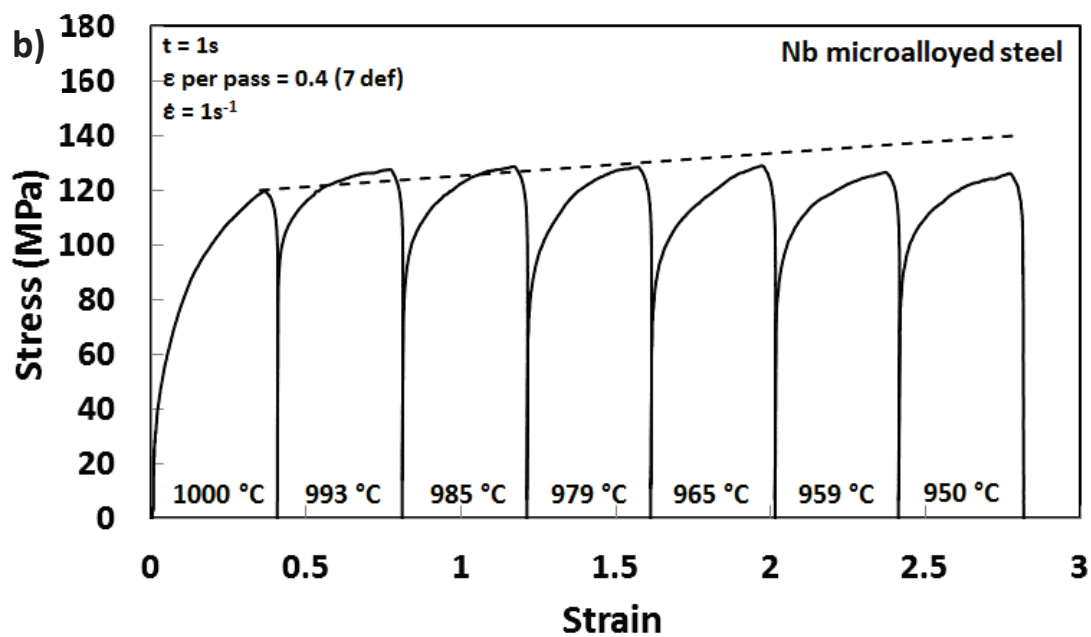
Here the final pass temperatures were 967°C (Figure 6.2.2a), 949°C (Figure 6.2.2b), 936°C (Figure 6.2.2c) and 884°C (Figure 6.2.2d). It is evident that the stress levels actually *decrease* at short interpass intervals (Figures 6.2.2a and 6.2.2b) and increase less than expected at longer times (Figures 6.2.2c and 6.2.2d). For example, in Figure 6.2.2a, the level drops by approximately 10% in going from the 1<sup>st</sup> to the 2<sup>nd</sup> pass and by about 15% when the 1<sup>st</sup> pass is compared with the last pass. There is also a noticeable change in the shapes of the curves (the 2<sup>nd</sup> to 7<sup>th</sup> display peaks). These changes indicate that strain accumulation is taking place, leading to dynamic and metadynamic recrystallization. Similar conclusions apply to Figures 6.2.2b and 6.2.2c.

It has been shown recently that two distinct softening mechanisms are contributing to the decreasing flow stress levels: these are dynamic transformation and dynamic recrystallization [6.2.10]. As mentioned above, the strains at which these are initiated have been identified

using the double differentiation method [6.2.13, 6.2.14]. In all cases, the critical strain for dynamic transformation is reached before the one for dynamic recrystallization. In the case of Figure 6.2.2d, the stress levels *increase* with decreasing temperature, but less rapidly than called for by the first pass values expected in the absence of DT and DRX and extrapolated to lower temperatures. The significance of these trends in the stress levels will be considered in more detail after presentation of the metallographic results.

Similar test conditions and parameters were used to produce the Nb steel flow curves displayed in Figure 6.2.3. Interpass times of 0.5s (Figure 6.2.3a), 1s (Figure 6.2.3b), 1.5s (Figure 6.2.3c) and 3s (Figure 6.2.3d) were again employed and straining was also applied at  $1\text{s}^{-1}$ . Once again, all the temperatures indicated are above the  $A_{e3}$ . Here the stress levels decrease after the 2<sup>nd</sup> pass in Figure 6.2.3a and after the 4<sup>th</sup> pass in Figure 6.2.3b, but not as sharply as in the case of the plain C steel, Figure 6.2.2. The stress levels in Figures 6.2.3c and 6.2.3d display somewhat lower rates of increase than those associated with the 1<sup>st</sup> pass flow stress peaks extrapolated to lower temperatures (broken lines). Note that the majority of the pass temperatures are above  $900^{\circ}\text{C}$ , so that most of the Nb is expected to remain in solid solution. Even in the case of the 3s interpass intervals (Figure 6.2.3d), where the last two pass temperatures are  $898$  and  $883^{\circ}\text{C}$ , little carbonitride precipitation is expected to have occurred because of the limited time available in the vicinity of  $900^{\circ}\text{C}$  and below.





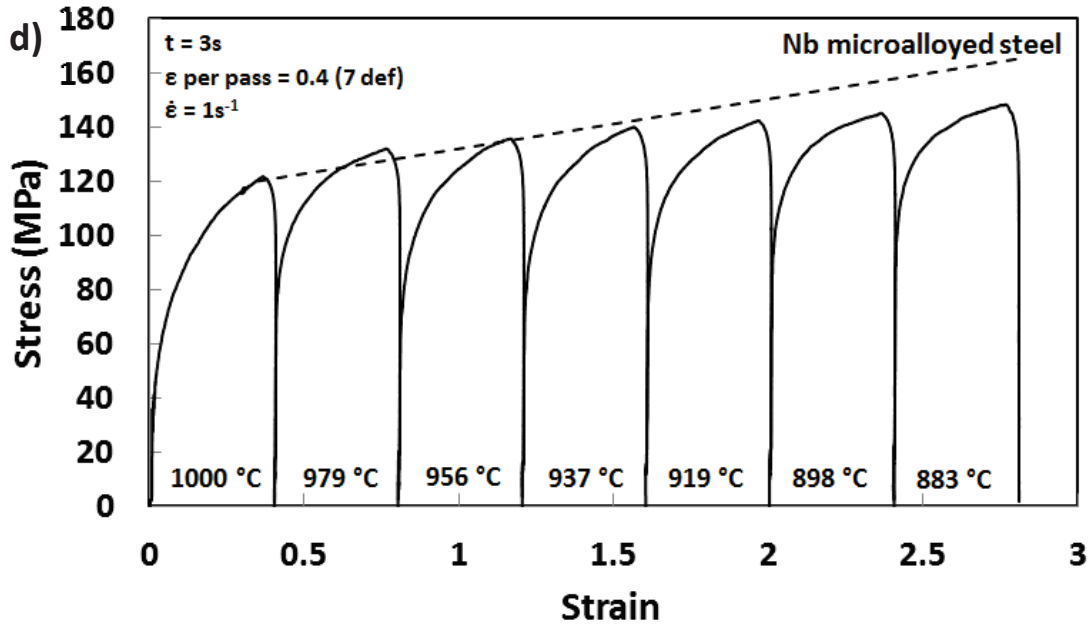


**Fig. 6.2.3.** Continuous cooling rolling simulations carried out on the Nb steel with interpass intervals of a) 0.5s, b) 1s, c) 1.5s, and d) 3s. Here the testing conditions of Figure 6.2.2 were employed. The presence of Nb in solution increases the flow stress levels and the amounts of retained work hardening. There is less of a discrepancy with respect to the “expected” behavior than in Figure 6.2.2.

### 6.2.3.2 Isothermal rolling simulations

Although the continuous cooling rolling simulations described above are closer approximations of industrial rolling, the characteristics of DT can be determined more readily by carrying out tests under isothermal conditions. In this way, the effects of DT and dynamic recrystallization can be distinguished more clearly. For this purpose, the present two alloys were subjected to fixed temperature rolling simulations with seven pass strains of 0.4 being applied at a strain rate of  $1\text{s}^{-1}$ . The results for interpass times of 0.5s and 5s are presented in Figures 6.2.4a and 6.2.4b for the plain C-Mn steel and in Figures 6.2.5a and 6.2.5b for the Nb microalloyed steel. The horizontal broken lines in Figures 6.2.4a and 6.2.4b denote the flow stress levels applicable to  $930^\circ\text{C}$ , which is  $60^\circ\text{C}$  above the paraequilibrium temperature of the plain C steel. When short pass intervals are employed, as shown in Figure 6.2.4a, the stress levels decrease by about 25% from the initial to the final pass. When the interpass time is increased to 5s (Figure 6.2.4b), a

somewhat smaller stress drop of about 20% is observed. These drops can be directly associated with the volume fraction of DT ferrite formed, as will be discussed in more detail in the sections that follow.

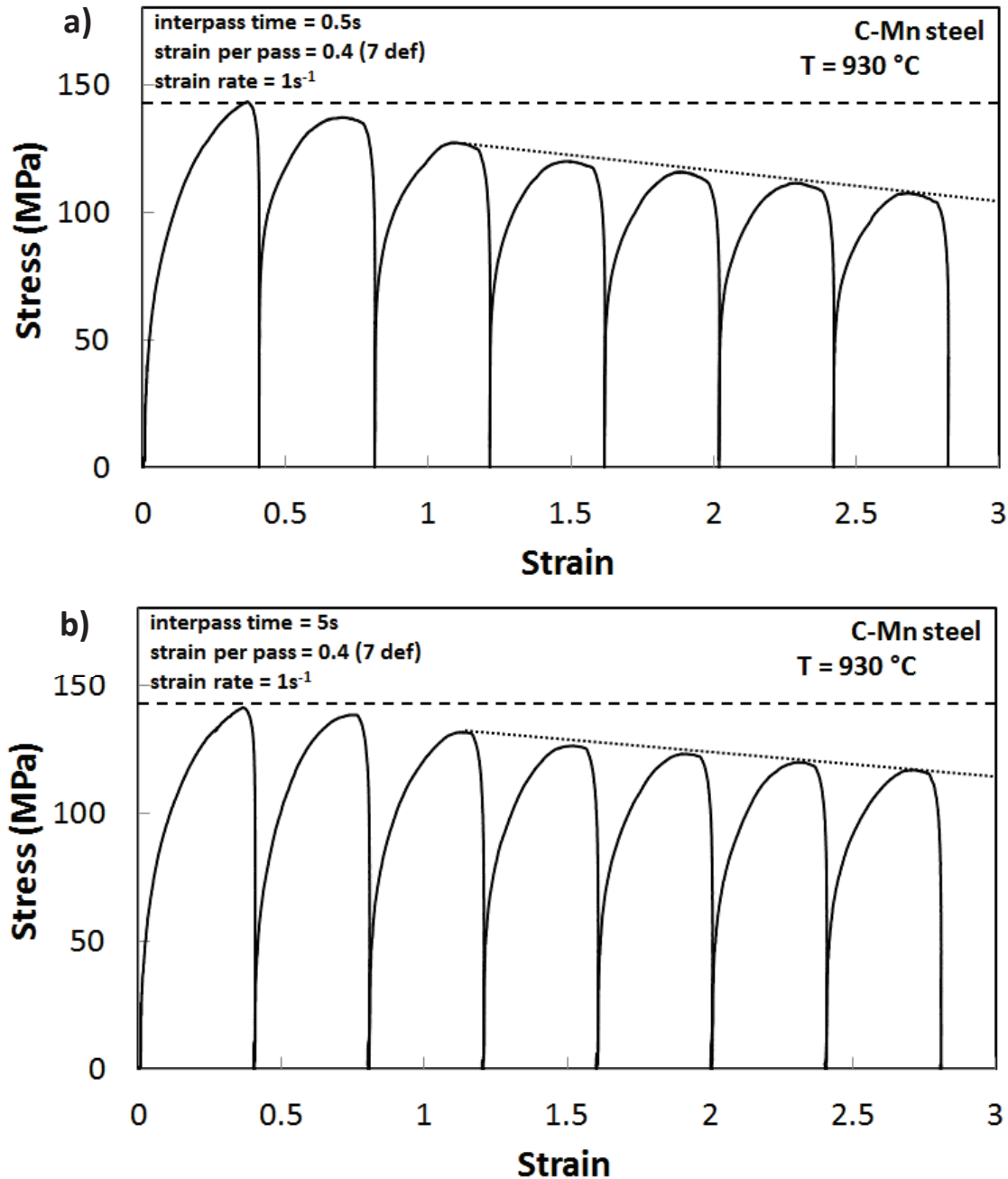


**Fig. 6.2.4.** Isothermal rolling simulations carried out according to the schedule in Figure 6.2.1b on the plain C steel using pass strains of 0.4 applied at  $1\text{ s}^{-1}$  and  $930^{\circ}\text{C}$ . Simulations with interpass intervals of: a) 0.5s and b) 5s were carried out. The short interpass intervals generate greater degrees of load decrease than the long times.

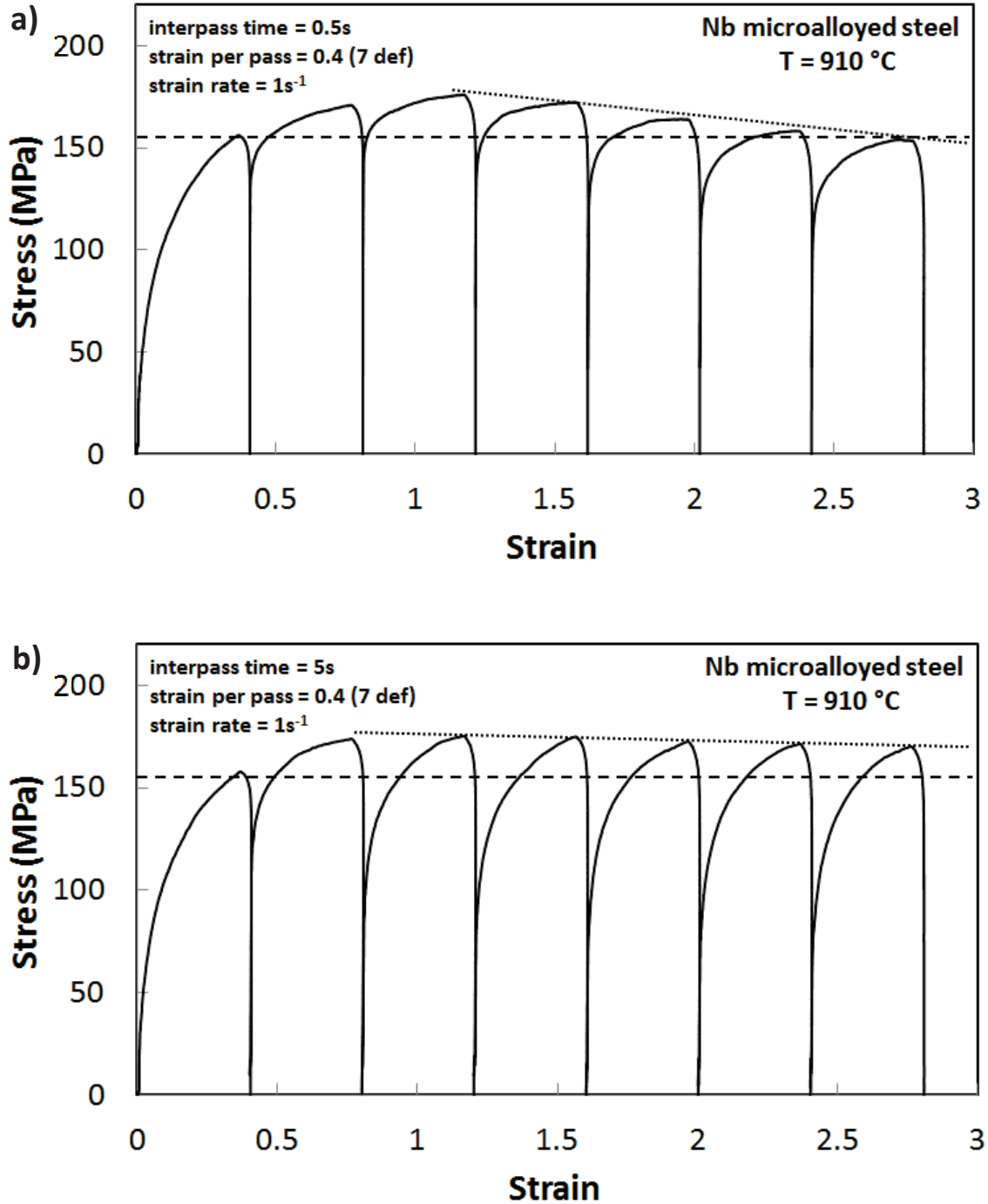


**Fig. 6.2.5.** Isothermal rolling simulations carried out on the Nb steel. The test parameters were identical to those of Figure 6.2.4. Interpass intervals of a) 0.5s and b) 5s were employed at a testing temperature of 910°C. There is considerable retention of work hardening, after which there is a decrease in the stress levels at rates that are lower than in Figure 6.2.4.

Similar trends are evident in the isothermal flow curves of the Nb steel displayed in Figure 6.2.5. In this case, however, the stress levels are *higher* than the anticipated values because of the work hardening that is retained from pass to pass. In these tests, the same testing conditions were employed as in Figure 6.2.4, but the temperature was decreased to 910°C, which is about 80°C above the relevant  $A_{e3}$ . Here the rate of flow stress decrease (after the third pass) is slightly greater when the interpass time is 0.5s (Figure 6.2.5a) than when it is 5s (Figure 6.2.5b). The effect of interpass time on DT as well as that of the addition of Nb will be taken up in more detail below when the mean flow stress and optical microstructure results are presented and discussed.

## 6.2.4 Discussion

### 6.2.4.1 Mean Flow Stress

The mean flow stress (MFS) in rolling depends primarily on strain, strain rate, temperature of deformation, interpass time and composition. It can also provide considerable insight into the occurrence of metallurgical phenomena such as recrystallization, carbonitride precipitation and the retention of work hardening. In the present case, the MFS's are of particular use with respect to assessing the extent to which DT ferrite is formed during a given simulation. Here the areas under the flow curves of Figures 6.2.2 and 6.2.3 were determined by integration and the MFS's obtained in this way are presented in the form of Boratto diagrams [6.2.19] in Figures 6.2.6 and 6.2.7 for the plain C and Nb steels, respectively. For reference, the paraequilibrium and orthoequilibrium  $A_{e3}$  temperatures are shown as broken vertical lines in Figure 6.2.6 and listed in the caption to Figure 6.2.7. As mentioned above, all the strains in these simulations were applied in the austenite region of the phase diagram.

In Figure 6.2.6, the broken line represents the effect of temperature on the MFS in the absence of DT and DRX. These were identified experimentally by performing first pass rolling simulations to strains of  $\epsilon = 0.10$  (i.e. prior to the initiation of DT) at a series of decreasing temperatures. An approach similar to that described in section 6.2.3.1 was used to derive the temperature dependences of the mean flow stress. In this way, the effects of DT and DRX on the MFS were eliminated and only the effect of decreasing the temperature is being presented.

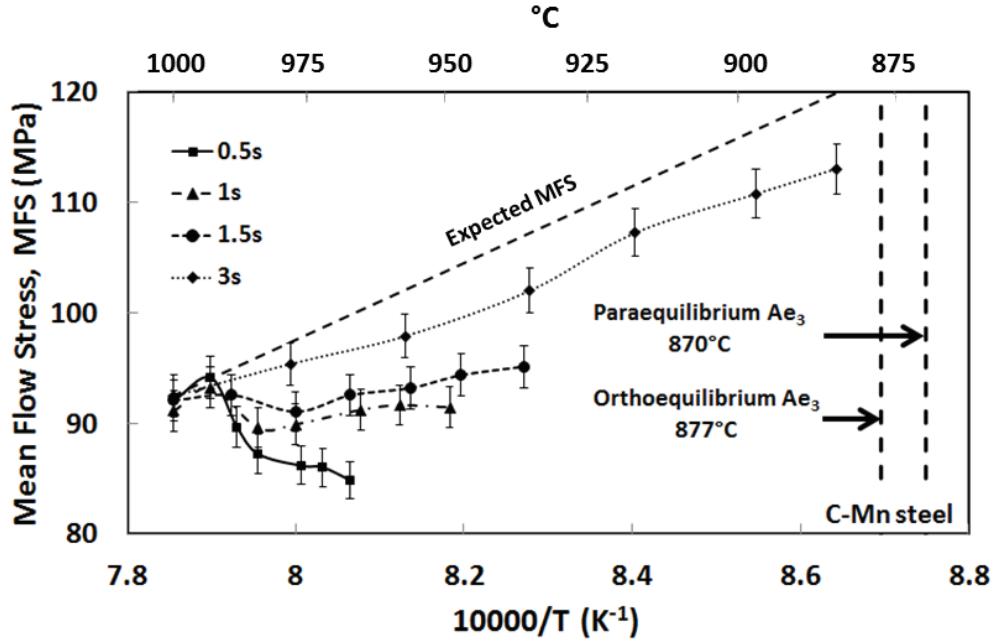


**Fig. 6.2.6.** The mean flow stresses (MFS's) deduced from the continuous cooling simulations of Figure 6.2.2 for the plain C steel plotted against inverse absolute temperature. The broken line indicates the dependence of the MFS in the absence of both DT and DRX. Here the MFS decreases much more sharply when the interpass time is short.

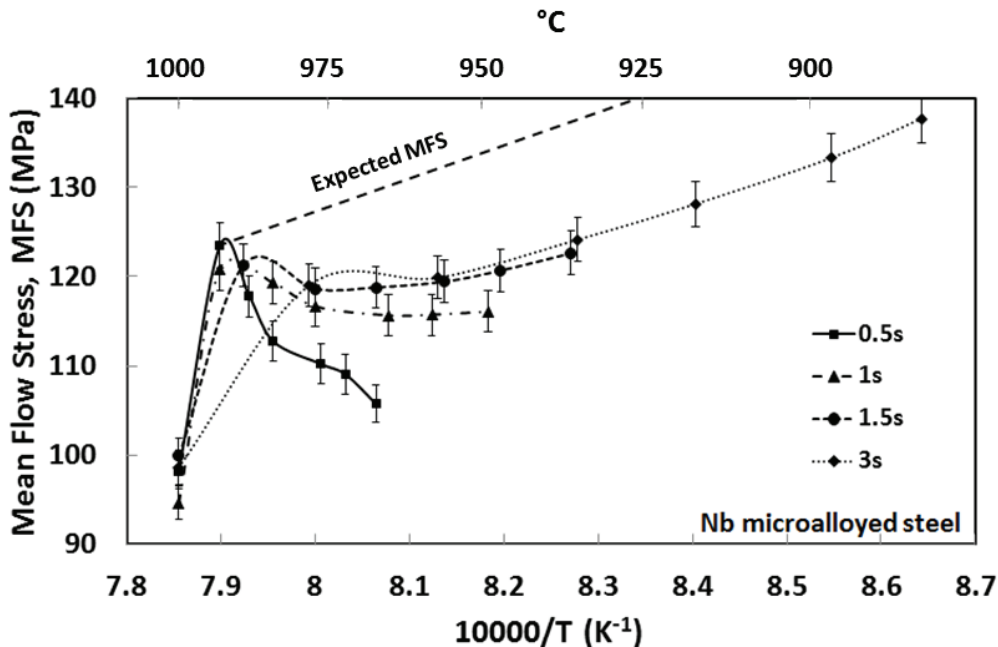


**Fig. 6.2.7.** The mean flow stresses (MFS's) deduced from the continuous cooling simulations of Figure 6.2.3 for the Nb steel plotted against inverse absolute temperature. The paraequilibrium and orthoequilibrium  $Ae_3$  temperatures for this steel are 827 $^{\circ}C$  and 843 $^{\circ}C$ , respectively. Although the flow stress levels are higher due to strain accumulation, it is clear that short interpass times generate greater decreases in rolling load than longer times.

It can be seen that the MFS values increase slightly from the 1<sup>st</sup> to the 2<sup>nd</sup> pass due to strain accumulation. The longer the interpass time, the less work hardening is retained due to the effects of static recovery and recrystallization during the interval of unloading. When the interpass time is as short as 0.5s, the MFS drops continuously after the second pass, despite the decreases in temperature. In the case of the 1 and 1.5s intervals, the drops are smaller and there is a slight increase with decreasing temperature in the latter case. These values are well below those associated with the material undergoing straining in the absence of DT and DRX. Only when the interpass time is increased to 3s do the MFS values begin to approach those called for by the decreases in temperature. When the last pass MFS's of the four simulations are compared with their corresponding expected values, the shortest interpass time value displays a shortfall of about 15% as opposed to a deficit of only 5% for the 3s interval. It is clear that as the interpass time is increased, the MFS gradually approaches its expected value.

The addition of Nb has a dramatic effect on the MFS versus inverse temperature plot, see Figure 6.2.7. The presence of Nb in solution retards recovery between passes, with the result that there is much more strain accumulation than in the plain C steel [6.2.9]. At the shortest interpass time, the large amount of second-pass retained work hardening is followed by appreciable MFS decreases in the succeeding passes. The broken line again represents the expected MFS behavior in the absence of DT and DRX. With respect to the 0.5s and 1s simulations, it can be seen that the MFS decreases significantly after the 2<sup>nd</sup> pass. In the cases of the 1.5s and 3s interpass intervals, the rates of the MFS increases are below those expected. This is consistent with the results obtained on the plain C steel in which shorter intervals generate greater MFS decreases than longer times.

#### **6.2.4.2 Effect of Interpass Time**

The flow curves of the isothermal rolling simulations of Figures 6.2.4 and 6.2.5 were analyzed and the results obtained in this way are presented here as Figures 6.2.8a and 6.2.8b for the plain C and Nb steel, respectively. It can be seen from Figure 6.2.8a that there is more retained work hardening in the second pass (about a 10% MFS increase) after an interpass time of 0.5s

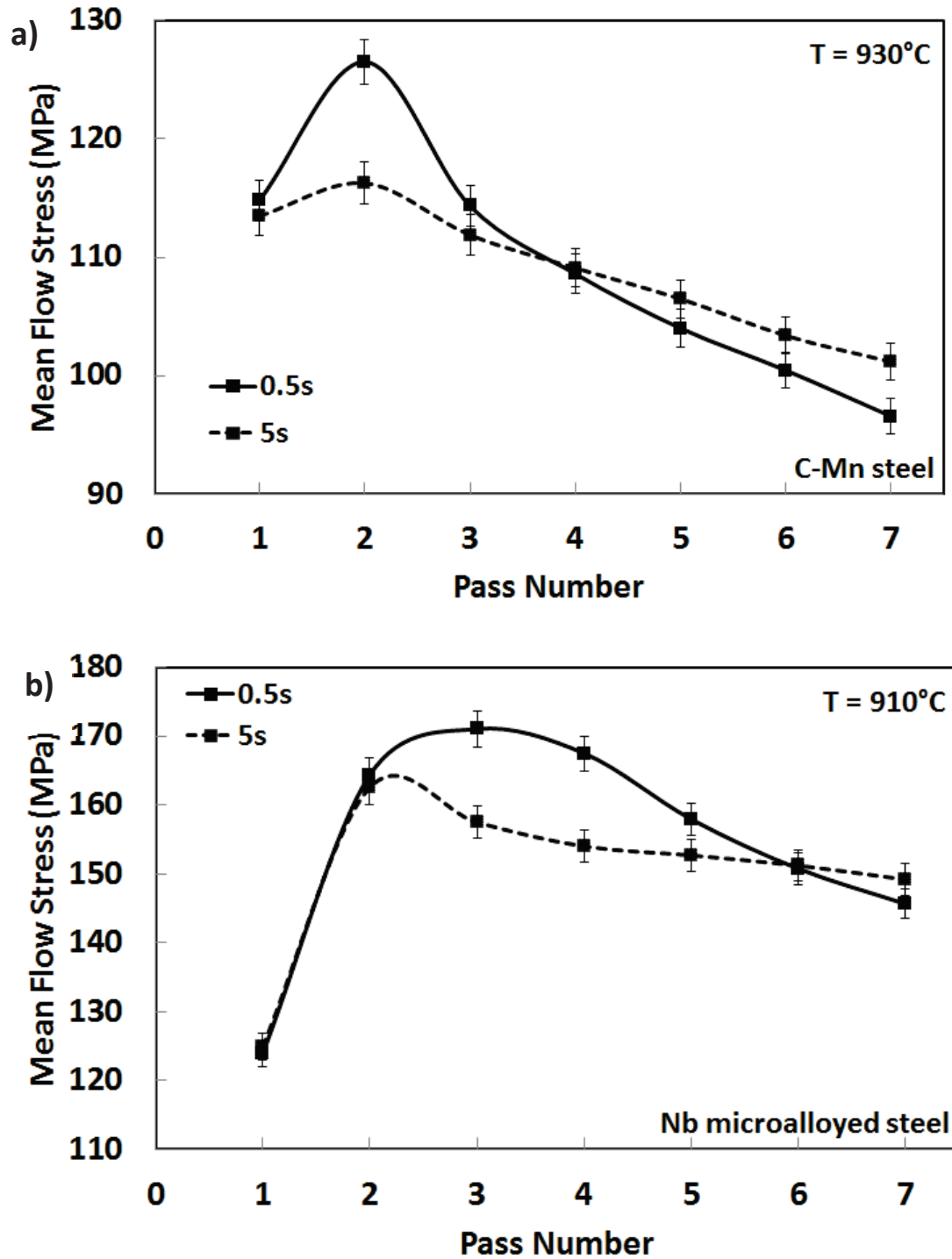


**Fig. 6.2.8.** MFS's deduced from the isothermal flow curves of Figures 6.2.4 and 6.2.5. Interpass times of 0.5s and 5s were employed for a) the plain C and b) the Nb microalloyed steel. In Figure 6.2.8a, shorter interpass times lead to more strain accumulation followed by greater decreases in MFS than longer intervals. In Figure 6.2.8b, there is again more strain accumulation and more softening when the interpass times are short.

than when a 5s interval was used. This is because there is less time for static recovery so that more dislocations are retained after short interpass times. Although the 2<sup>nd</sup> pass MFS value is higher in the 0.5s simulation, the final pass MFS is lower. This is because more ferrite is formed during short interpass time rolling and there is less time for the ferrite to retransform into austenite under these conditions. This interpretation is supported by the optical microstructures presented in the next section.

A similar analysis was carried out for the Nb isothermal simulation and the results obtained are illustrated in Figure 6.2.8b. In this case, strain accumulation in the second pass increases the MFS by 30% in both the short and long interval simulations. Here there is much less recovery between passes than in the case of the plain C steel. In the succeeding passes, more work hardening is retained after the 0.5s than after the 5s intervals, leading to higher MFS's in the 3<sup>rd</sup>, 4<sup>th</sup> and 5<sup>th</sup> passes. Conversely, there is much more dynamic softening during straining, so that the pass 7 MFS's are only slightly lower (0.5s interpass time), or not perceptibly (5s interpass time) than the pass 6 MFS's. This indicates that more ferrite is being produced (and retained) when the interpass times are short than when they are long.

#### **6.2.4.3 Retransformation of DT ferrite**

In torsion simulations, the applied strain increases linearly from zero along the axis to a maximum at the surface of the sample. For this reason, the volume fraction of DT ferrite that is formed increases from the axis to the surface. The optical micrographs presented in Figures 6.2.9 and 6.2.10 were all taken 0.3mm from the surface. Those of Figure 6.2.9 for the plain C steel are from the isothermal simulation of Figure 6.2.4 using pass intervals of 0.5s and represent the microstructure: i) right after the last pass (Figure 6.2.9a), ii) after 8s of holding (Figure 6.2.9b), and iii) after 15s of holding (Figure 6.2.9c). The martensite appears dark while the ferrite is light. The microstructure of Figure 6.2.9a consists essentially of ferrite (about 80-90%), produced by applying an accumulated strain of  $\epsilon = 2.8$ . When the sample was held for 8s before quenching, Figure 6.2.9b, the volume fraction of DT ferrite decreased to approximately 50-60%. Finally, when the holding time was increased to 15s, this fraction was further reduced to 10-20%. This confirms that when the holding time is increased, more and more ferrite

retransforms to austenite. These microstructures support the interpretation of the MFS plots proposed above.

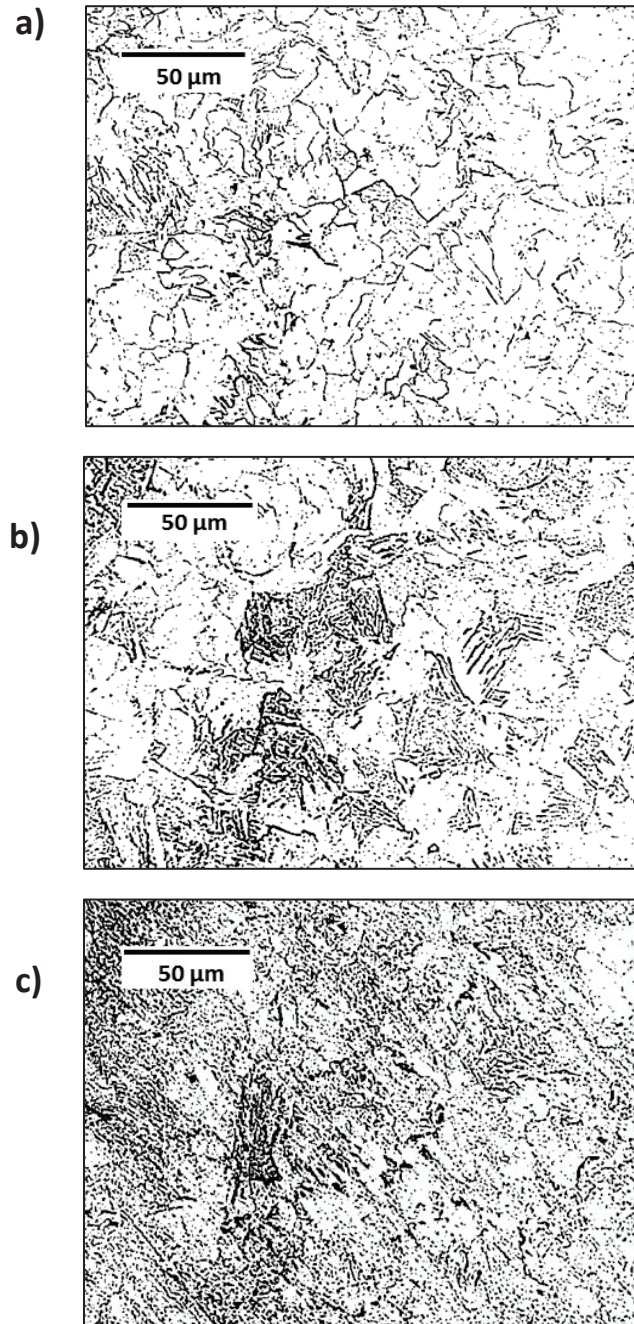


**Fig. 6.2.9.** Optical microstructures of the plain C steel subjected to the 930°C 7-pass, 0.5s interpass time rolling simulation of Figure 6.2.4. The samples were quenched: a) immediately after straining, b) after holding for 8s, and c) after holding for 15s. Light regions are ferrite while dark regions are martensite (prior austenite). The DT ferrite is seen to retransform gradually into austenite during holding.

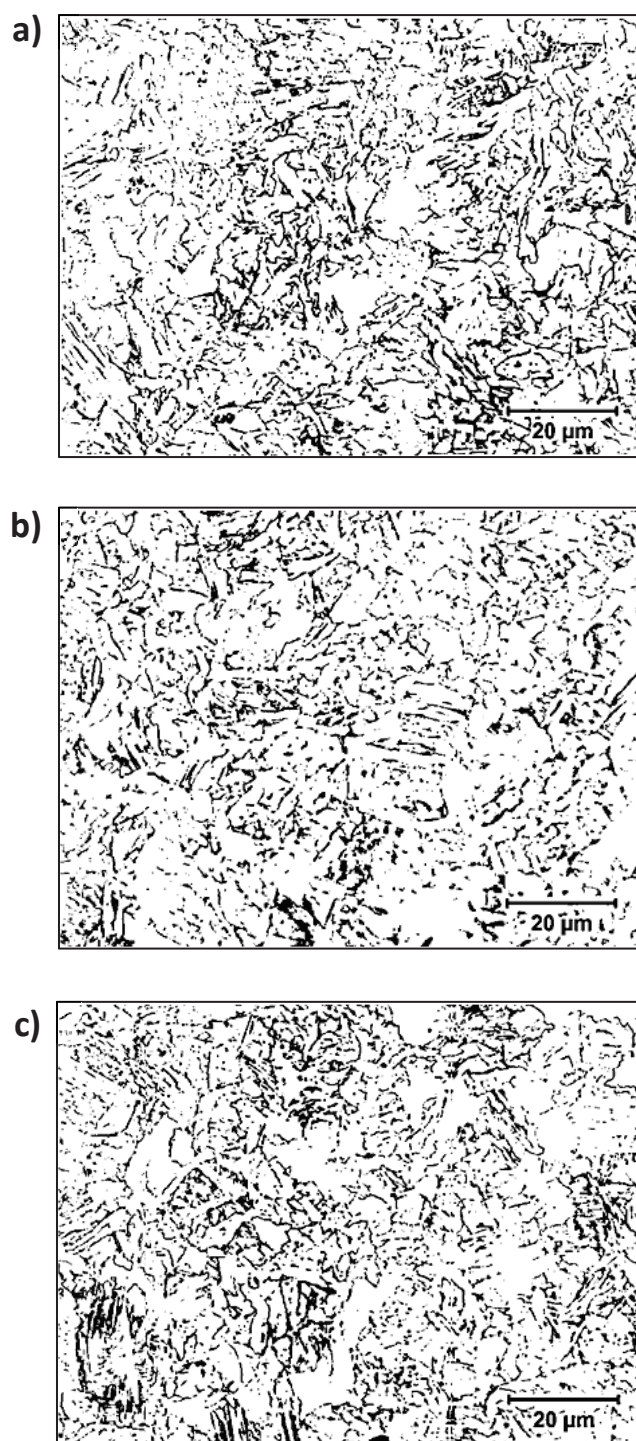


**Fig. 6.2.10.** Optical microstructures of the Nb steel subjected to the 910°C 7-pass, 0.5s interpass time rolling simulation of Figure 6.2.5. Here the quenching conditions of Figure 6.2.9 were employed: a) immediately after straining, b) after holding for 8s, and c) after holding for 15s. There is less reversion to austenite than in the plain C steel.

The optical microstructures relating to the isothermal simulations performed on the Nb steel (Figure 6.2.5) are illustrated in Figure 6.2.10. These samples were subjected to the 910 °C 7-pass, 0.5s interpass time rolling simulation of Figure 6.2.5. Here again the samples were quenched: i) right after straining (Figure 6.2.10a), ii) after 8s of holding (Figure 6.2.10b), and, iii) after 15s of holding (Figure 6.2.10c). It is of interest that there is little or no difference in the volume fraction of ferrite. This is in agreement with the MFS observations of the isothermal simulation showing the convergence of values in the last two passes.

In the plain C steel, the decrease in the proportion of ferrite formed after holding for 15s is caused by the reverse transformation of DT ferrite back into austenite. The nucleation of DT ferrite during rolling is known to occur in a displacive manner while the reverse transformation takes place by a diffusional mechanism [6.2.20, 6.2.21]. In this way, the amount of retransformation will depend on the length of the interpass interval as well as the composition. It can be seen from the results obtained here that the presence of Nb in solution significantly retards both the forward as well as the backward transformation rates by a solute drag effect [6.2.22-6.2.24].

The presence of DT ferrite in Figures 6.2.9 and 6.2.10 was confirmed using EBSD techniques of phase identification, as illustrated here in Figure 6.2.11. Inverse pole figure (IPF) plots related to the orientations of the ferrite formed are shown on the left side of the diagram while the phase distribution is displayed on the right-hand side. Here the red regions are ferrite while the black ones are martensite (prior austenite). EBSD images associated with the C-Mn samples of Figures 6.2.10a (directly quenched) and 6.2.10c (15s of holding) are displayed in Figures 6.2.11a and 6.2.11b, respectively. These samples were subjected to the 930 °C 7-pass, 0.5s interpass time rolling simulation of Figure 6.2.4. Here the volume fraction of ferrite is significantly higher in the directly quenched sample than in the sample held for 15s. This confirms the interpretation of Figure 6.2.9 where the decrease in ferrite volume fraction was attributed to the reverse transformation.

Similar analyses were performed on the Nb steel. The phases in the sample of Figure 6.2.10c (15s of holding) are positively identified here as Figure 6.2.11c. This sample was subjected to

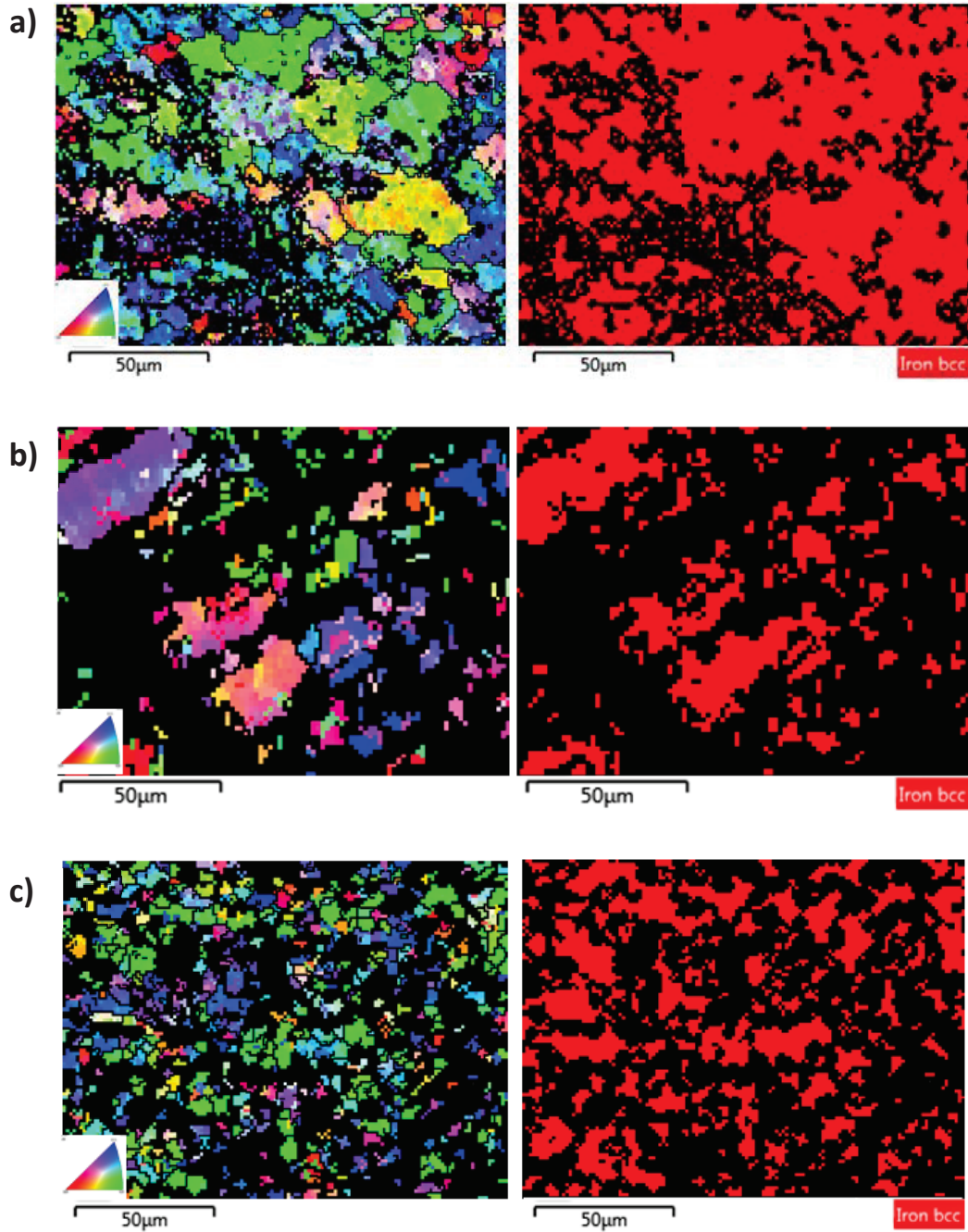


**Fig. 6.2.11.** EBSD micrographs of transverse cross-sections of the C-Mn and Nb steels subjected to the 7-pass, 0.5s interpass time rolling simulation of Figures 6.2.4 and 6.2.5. The C-Mn samples were quenched: Fig. 6.2.11a - immediately after straining, and Fig. 6.2.11b - after holding for 15s. The Nb steel sample of Fig. 6.2.11c was quenched after 15s of holding. In the right hand figures, the ferrite phase is red and the prior austenite is black.

the 7-pass, 0.5s interpass time rolling simulation at 910 °C of Figure 6.2.5. Here again the formation of DT ferrite is confirmed, although there was little difference in ferrite volume fraction between the directly quenched sample (not shown here) and the one held for 15s after deformation. This result supports the interpretation of the optical micrographs of Figure 6.2.10 according to which the presence of Nb in solid solution was concluded to retard the reverse transformation. This will be shown in more detail in a future paper, where measurements of the volume fractions of DT ferrite formed in each pass and subsequently lost by retransformation will be presented.

#### **6.2.4.4 Industrial Implications**

The present results indicate that DT takes place during each pass of strip rolling and that the interpass interval affects the volume fraction of ferrite formed and retained above the  $A_{e3}$ . This then explains why the rolling load and MFS do not increase with decreasing temperature in strip mills. The present results are also consistent with the observations of earlier researchers [6.2.2,6.2.3,6.2.10,6.2.25-6.2.28] on dynamic transformation. Thus it appears that the lack of increase in rolling load cannot be attributed solely to DRX. The dilatation associated with ferrite formation can also be seen in this way to be responsible for the increase in volume flow rate (but not mass flow rate) as a bar passes through a mill.

A further application of the present work involves the carbon partitioning that takes place during ferrite formation. By moving into the prior austenite, it can be responsible for the formation of undesirable amounts of martensite on the runout table. The present observations also indicate that the introduction of water cooling directly after the last mill stand (i.e. after reduced delays) is likely to reduce the amount of retransformation that takes place and affect the mechanical properties in this way.

#### **6.2.5 Conclusions**

1. The rolling simulations carried out on the plain C steel reveal that the MFS decreases significantly when short interpass times are used and increases less rapidly than expected when

this interval is increased in length. These trends are directly attributable to the dynamic transformation of austenite to ferrite within the austenite phase field, i.e. above the  $A_{e3}$ . The microstructures observed using optical microscopy indicate that more DT ferrite forms and is retained when the interpass intervals are shorter than when longer intervals are used.

2. The simulations carried out on the Nb steel reveal that the MFS decreases substantially from the 2<sup>nd</sup> to the succeeding passes when short interpass times are used and displays a low rate of increase when this interval is increased in length. These trends can be directly associated with the formation of DT ferrite from pass to pass during the simulation. It appears that the shorter interpass times both produce more DT ferrite and permit less reversion to austenite than the longer times. The optical micrographs indicate that ferrite forms during straining and that most of it are retained during unloading.

3. Comparison of the behavior of the plain C and Nb steels reveals that Nb addition retards both the forward as well as the reverse transformation. As a result, more DT ferrite is formed in the plain C steel than in the Nb steel.

4. Most of the DT ferrite formed in the plain C steel is able to retransform into austenite after holding for 15s. By contrast, the reverse transformation is severely retarded in the Nb steel. As a result, there is little or no retransformation after 15s of holding under the present conditions.

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## Chapter 7

### **Ferrite Formation above the $A_{e3}$ Temperature during the Torsion Simulation of Strip Rolling**

In the previous chapter, torsion simulations of strip rolling using C-Mn and Nb microalloyed steels were described in which the effect of interpass time during rolling was determined. It was shown that the flow stress levels and MFS's decrease when the interpass times are short. However, no quantitative information was provided regarding the volume fraction of ferrite formed during each pass of the simulation. Here 7-pass torsion simulations were carried out using the same alloys and techniques described above to determine the amounts of ferrite formed and retained from pass to pass. The flow stress results were complemented with determination of the optical microstructures and hardnesses associated with particular passes.

- This section has been published: **C. Aranas Jr.**, S. Rodrigues, R. Grewal and J.J. Jonas, "Ferrite Formation above the  $A_{e3}$  Temperature during the Torsion Simulation of Strip Rolling," ISIJ International, 55 (2015) 2426-2434.

### Abstract

Torsion simulations of 7-pass strip rolling were carried out on a 0.06%C-0.3%Mn-0.01%Si and a 0.11%C-1.0%Mn-0.11%Si-0.03%Al-0.034%Nb steel using pass strains of 0.4 applied at  $1s^{-1}$ . The deformations were imposed isothermally at 910°C and 930°C for the C-Mn and the Nb microalloyed steel, respectively. The flow curve levels decreased from pass to pass as a result of softening by both dynamic transformation (DT) and dynamic recrystallization (DRX). The application of double differentiation to the stress-strain curves led to average critical strains for the initiation of DT and DRX of about 0.06 and 0.11, respectively. Optical microscopy revealed that the volume fraction of DT ferrite increased continuously right up to the last pass. The fraction of DT ferrite formed and retained was significantly higher when short interpass times were used. Comparison of the behaviors of the C-Mn and Nb steels indicates that Nb addition retards both the forward as well as the reverse transformation.

## 7.1 Introduction

The formation of ultrafine ferrite by dynamic strain induced transformation (DSIT) attracted considerable interest in the 1980's and 1990's [7.1]. This involves the nucleation of ferrite by deforming austenite between the  $A_{e3}$  and  $A_{r3}$  temperatures. Grain sizes of about 1-3  $\mu\text{m}$  were produced in both laboratory and pilot scale rolling [7.1-7.3]. However, transformation can also be initiated *above* the  $A_{e3}$  temperature, as shown by Yada and co-workers in the 1980's [7.3]. This is referred to here as dynamic transformation (DT). The microstructures from their hot compression tests and rolling simulations revealed that DT can occur as much as 166 °C above the  $A_{e3}$ . They also showed that reverse transformation takes place during isothermal holding after straining. This was followed by in-situ x-ray diffraction experiments [7.4] and dilatometric measurements [7.5] to provide real-time evidence for DT. Thermodynamic explanations of this unusual phenomenon have been proposed in terms of the stored energy of dislocations [7.6], stress activation [7.7] and transformation softening [7.8].

Early strip rolling simulations were mainly focused on the effects of static and dynamic recrystallization on the rolling load [7.9-7.13]. These studies indicated that short interpass times lead to significant reductions in the mean flow stress (MFS), which were then attributed to the occurrence of dynamic recrystallization. Moreover, rolling simulations performed by Bowden et al. [7.14] on microalloyed steels showed that Nb(C,N) precipitation and solute drag significantly retard both dynamic and static recrystallization. At this time, the dynamic transformation of austenite into the softer ferrite phase was not considered as playing a role.

In an earlier paper [7.15], it was shown that DT also contributes to the softening that takes place during rolling. In that work, torsion simulations of strip rolling were carried out on a C-Mn and a Nb microalloyed steel to determine the effect of interpass time on the rolling load. Once again, the flow stress levels and MFS's decreased when short interpass times were employed. The nucleation and growth of ferrite from the austenite reduced the rolling load and modified the microstructure. Nevertheless, no quantitative information was provided regarding the volume fractions of ferrite being formed during rolling.

In this study, torsion simulations of strip rolling were carried out on a C-Mn and a Nb microalloyed steel and water quenching was performed after selected passes so that the volume fraction of ferrite formed in each pass could be quantified. The results obtained are described in detail below.

## 7.2 Experimental Procedure

The steels were supplied in the form of hot rolled plates with thicknesses of 12.5 mm. These were machined into torsion specimens with diameters of 6.3 mm and gauge lengths of 22.2 mm and their axes parallel to the rolling direction. The chemical compositions of the two steels are displayed in Table 7.1; their paraequilibrium and orthoequilibrium  $A_{e3}$  temperatures are also shown here. The latter were calculated using the FSstel database of the FactSage thermodynamic software. [7.16]

**Table 7.1.** Chemical compositions (mass%) and equilibrium transformation temperatures ( $^{\circ}\text{C}$ )

|                        | <b>C</b> | <b>Mn</b> | <b>Si</b> | <b>Nb</b> | <b><math>A_{e3(p)}</math></b> | <b><math>A_{e3(o)}</math></b> |
|------------------------|----------|-----------|-----------|-----------|-------------------------------|-------------------------------|
| <b>Plain C-Mn</b>      | 0.06     | 0.30      | 0.01      | -         | 870 $^{\circ}\text{C}$        | 877 $^{\circ}\text{C}$        |
| <b>Nb microalloyed</b> | 0.11     | 1.0       | 0.11      | 0.034     | 827 $^{\circ}\text{C}$        | 843 $^{\circ}\text{C}$        |

### 7.2.1 Torsion Testing

Torsion simulations of strip rolling were carried out using a computer controlled MTS torsion machine equipped with a horizontal radiation furnace and a temperature controller. A thermocouple was welded to the torsion samples to accurately measure the temperatures during deformation. The samples were protected by an argon gas atmosphere so as to minimize the amount of oxidation and decarburization at elevated temperatures. The torque/twist data were converted into equivalent stress/strain form using the Fields and Backofen formula [7.17]. Corrections were also made for the increase in the sample diameter during free-end torsion testing [7.18].

The thermomechanical schedule employed during the rolling simulations is shown in Fig. 7.1. The samples were heated at  $1^{\circ}\text{C/s}$  from room temperature to  $1200^{\circ}\text{C}$ . After austenitization for 20 minutes, the samples were cooled at  $1^{\circ}\text{C/s}$  down to the roughing pass temperature of

1100°C. The temperature was held for 60s before simulating the roughing pass by applying a strain of 0.4 at the rate of  $1\text{ s}^{-1}$ . After this, samples were held for another 90s to allow recrystallization to take place before cooling at  $1^\circ\text{C/s}$  down to test temperatures of 930°C and 910°C for the C-Mn and Nb microalloyed steels, respectively. Seven deformations were applied isothermally with strains of 0.4 applied at  $1\text{ s}^{-1}$  and employing interpass times of 0.5s and 5s. Samples were water quenched after selected passes, namely, after the i) 1<sup>st</sup>, ii) 3<sup>rd</sup>, iii) 5<sup>th</sup>, and iv) last passes. This procedure enabled the ferrite volume fraction formed in pairs of passes to be quantified and also led to assessment of the amount of reverse transformation taking place between passes.

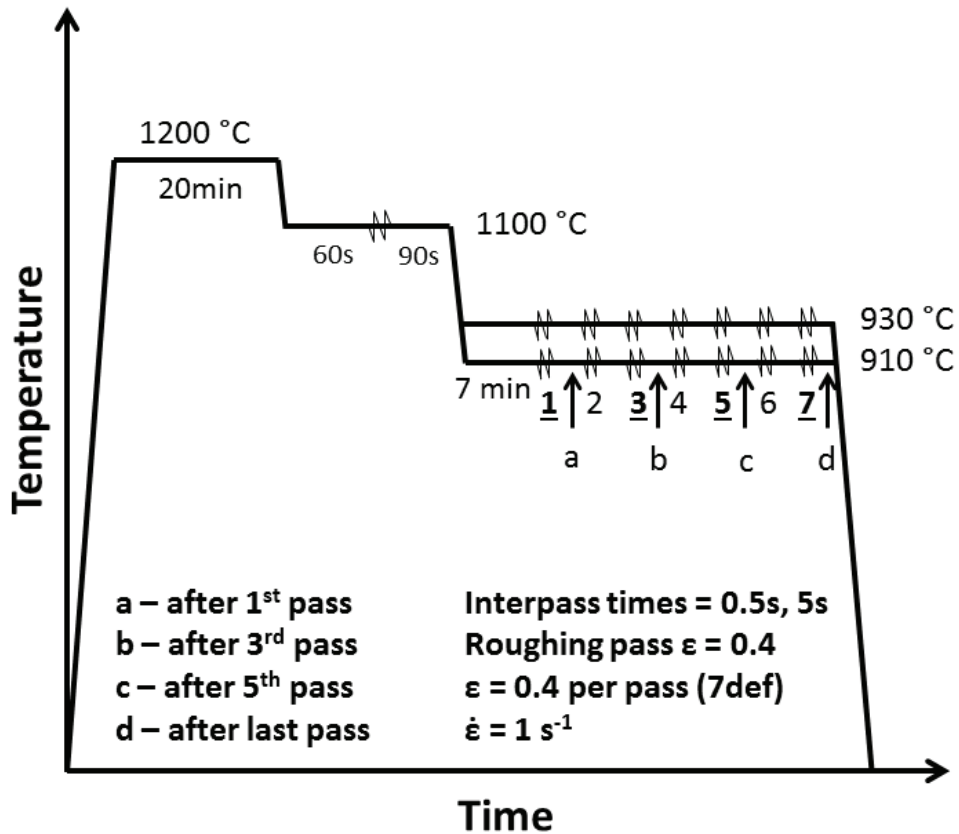


**Fig. 7.1.** Torsion testing schedules for the simulation of strip rolling. The deformation temperatures were 930°C and 910°C for the C-Mn and Nb steel, respectively, and interpass times of 0.5s and 5s were employed.

### 7.2.2 Metallography

For microstructural analysis, the torsion samples were cut perpendicular to the longitudinal axis and rolling direction. These were hot mounted and then polished using silicon carbide papers with grits of 400, 600, 800, 1000 and 1200 that were lubricated with water. Diamond (3  $\mu\text{m}$  and 1  $\mu\text{m}$ ) and colloidal silica (0.02  $\mu\text{m}$ ) suspensions were used for fine and final polishing. The polished samples were etched with a 2% nital solution followed by treatment with a 10% aqueous metabisulfite ( $\text{Na}_2\text{S}_2\text{O}_5$ ) solution to improve the contrast between martensite and ferrite.

## 7.3 Results

### 7.3.1 Torsion Simulation of Strip Rolling

In torsion testing, the Fields and Backofen [7.17] expression is generally employed to convert torque/twist data into equivalent stress/strain curves. For this purpose, *average* values of the twist hardening exponent ( $N$ ) are used rather than the instantaneous ones. This method overestimates the critical strains for the initiation of dynamic transformation and dynamic recrystallization, as discussed in Ref. 7.19. Two stress-strain curves derived from one of the present tests on the Nb steel are presented in Fig. 7.2. Here both the average  $N$  (broken line) and local  $N$  (solid line) were used in the conversions. It can be seen that the level of the corrected curve is about 6 MPa below that of the average  $N$  curve and that it has a slightly different shape. For the latter reason, only the curves corrected in this way were used to determine the critical strains presented below.

The isothermal flow curves determined from the 7-pass rolling simulation carried out on the C-Mn steel are presented in Fig. 7.3. Here interpass times of 0.5s (Fig. 7.3a) and 5s (Fig. 7.3b) were employed at 930°C. It should be noted that the simulations were performed isothermally at about 60°C above the paraequilibrium  $Ae_3$  temperature. No interpass cooling was employed so as to simplify the interpretation of the observed flow stresses. In the absence of a phase change, the flow stress level should not change significantly from pass to pass during deformation at a fixed temperature. However, here the peak flow stress was reduced by about

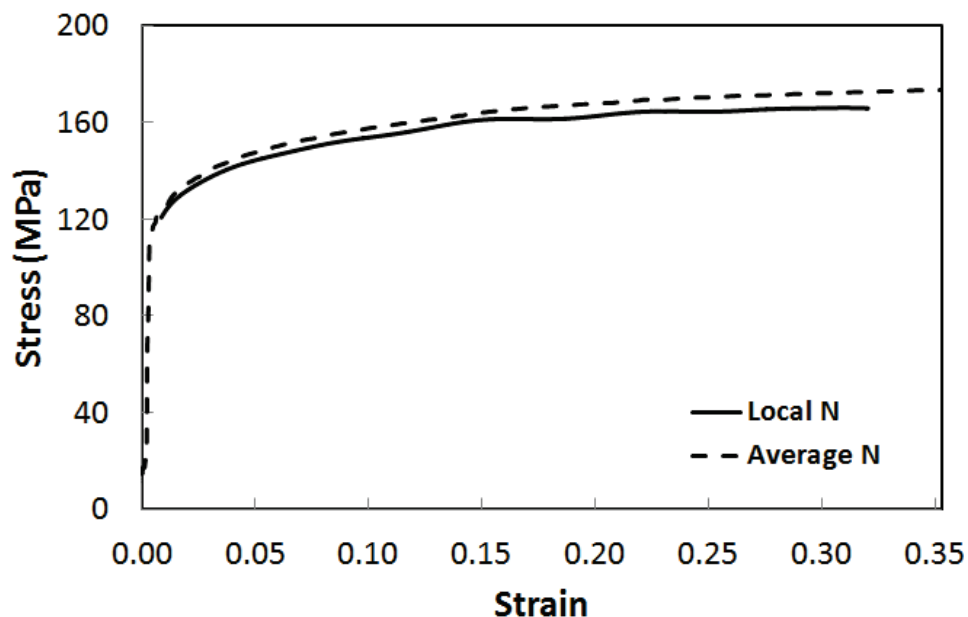
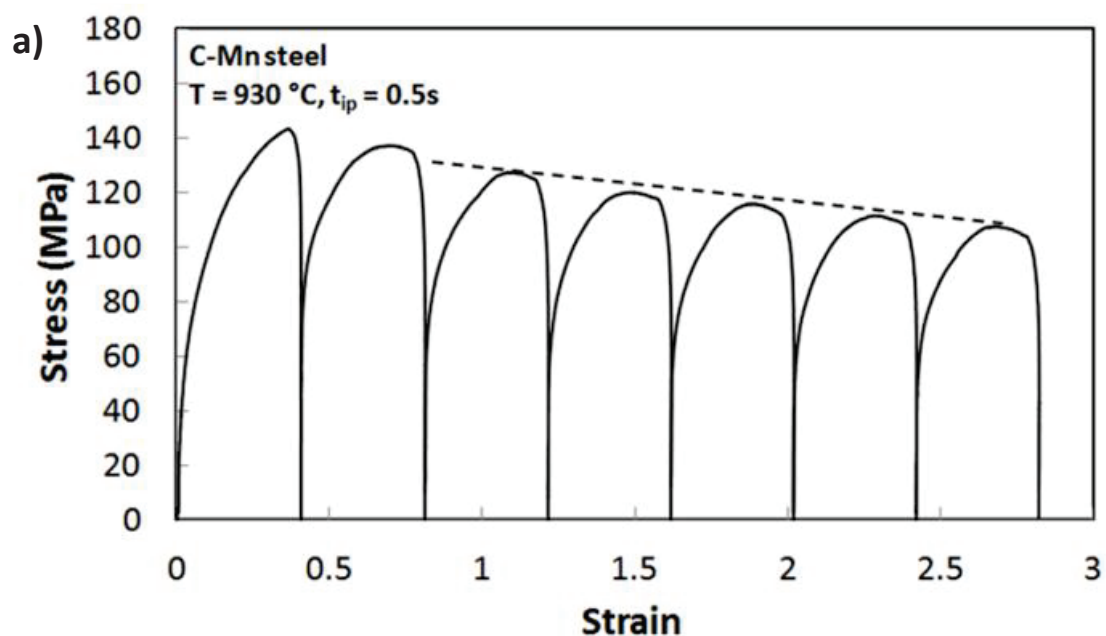


**Fig. 7.2.** Two stress-strain curves derived from the torque/twist data obtained during a torsion test of the present Nb steel. The curve shown as a broken line was derived using the average N while the solid curve was obtained using the local values of N. There appears to be slightly more dynamic softening when the local N is employed and lower values of the critical strains are obtained.



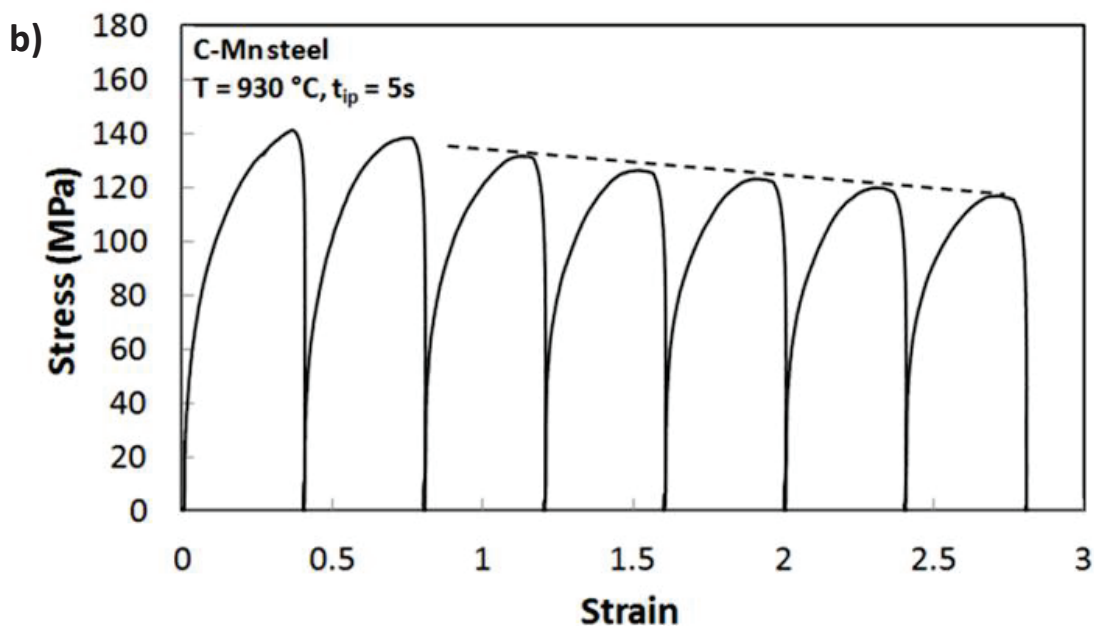


**Fig. 7.3.** Stress-strain curves determined on the C-Mn steel according to the schedule in Fig. 7.1 using pass strains of 0.4 applied at a  $1\text{ s}^{-1}$  and  $930^{\circ}\text{C}$ . Interpass times of a) 0.5s and b) 5s were employed. More ferrite was formed and retained when shorter times were used.

25% from the first pass to the final pass when the 0.5s interpass time was employed and by roughly 15% when the longer pass intervals were employed. These reductions can be directly associated with the volume fractions of ferrite formed and retained from pass to pass, as will be discussed in more detail in the section that follows. This interpretation will be examined again below when the microstructures pertaining to the two simulations are presented.

A progressive decrease in peak stress is also evident in the Nb steel simulations displayed in Fig. 7.4. These tests were carried out at about  $80^{\circ}\text{C}$  above the paraequilibrium  $A_{e3}$  temperature. Note that the rate of decrease in stress level is markedly higher in the 0.5s interpass time tests (Fig. 7.4a) than in those involving 5s interpass times (Fig. 7.4b). The shapes of the flow curves indicate that there is considerable less static recrystallization between passes in the first case, leading to the appreciable retention of work hardening. As a result, the average flow stresses are higher than when full recrystallization occurs. High austenite flow stresses in turn promote ferrite formation by transformation softening, so more ferrite formation is expected to occur under the interpass conditions of Fig. 7.4a. The resulting higher average flow stresses in the austenite phase are then responsible for the greater rate of ferrite formation in the first case.

The gradual decrease in the flow stress levels of the successive pass flow curves can thus be attributed to progressive ferrite formation.

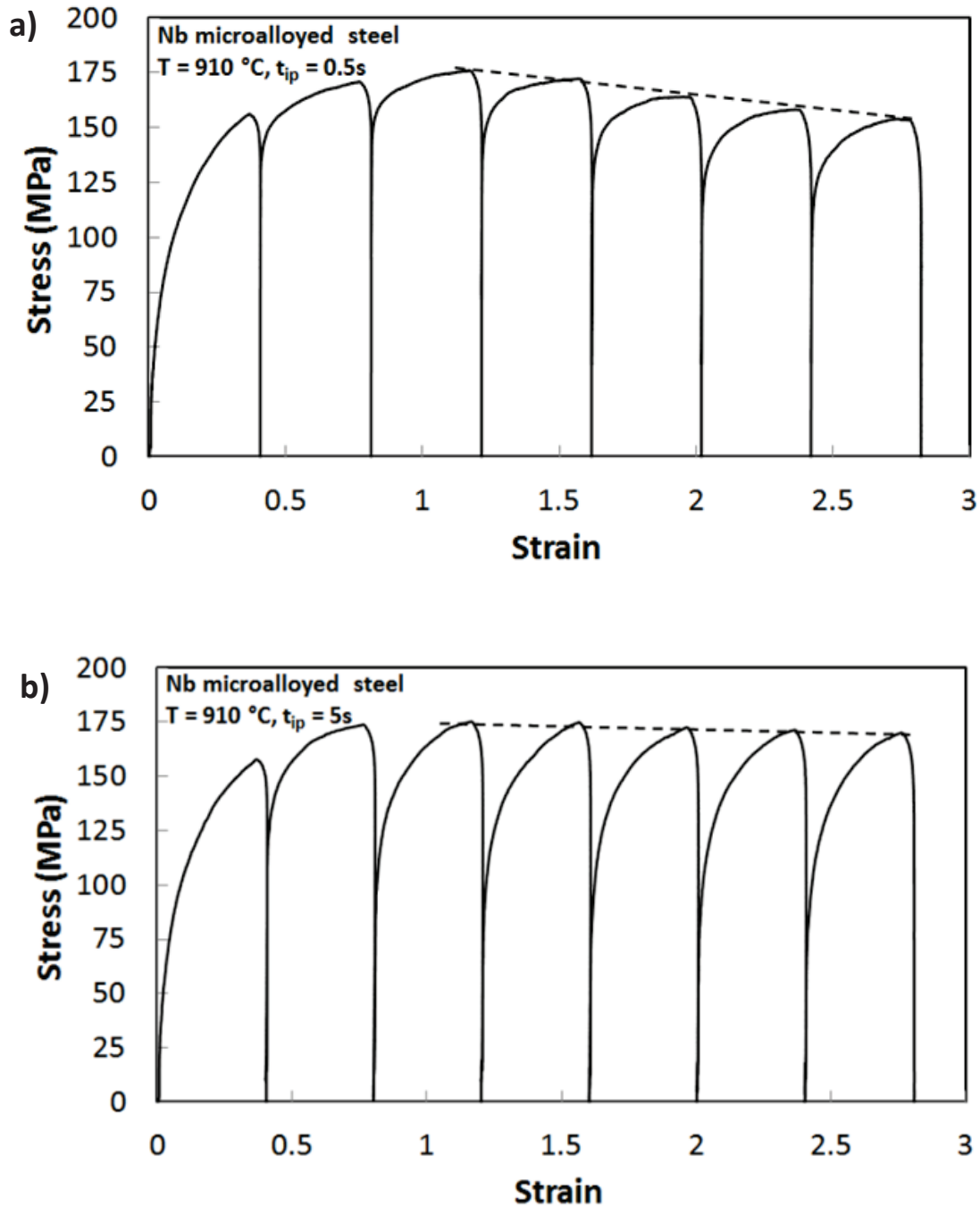
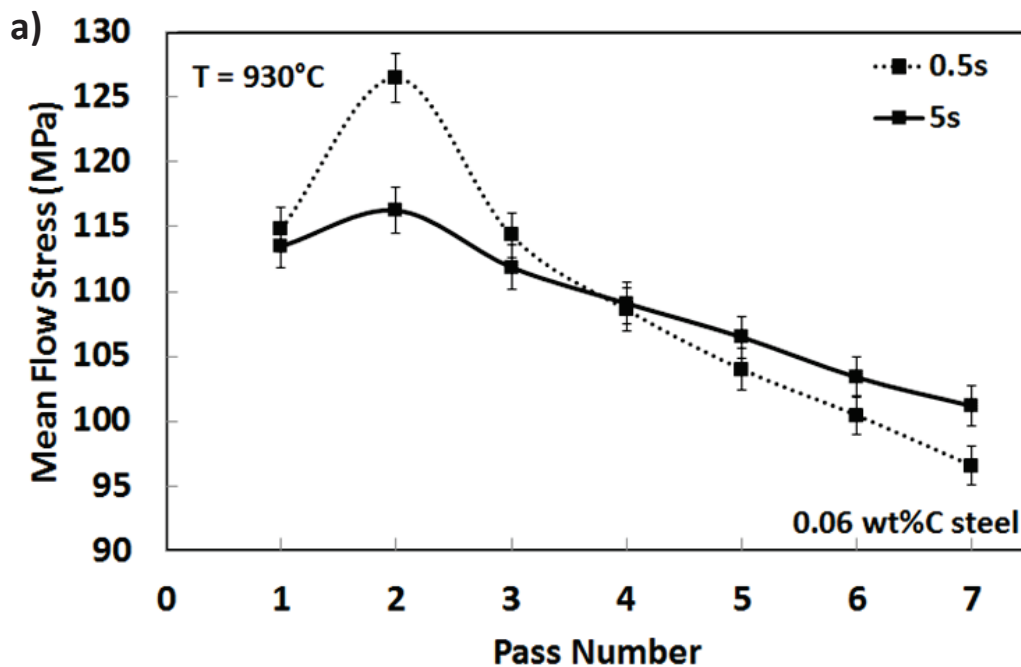


**Fig. 7.4.** Stress-strain curves determined on the Nb microalloyed steel according to the schedule in Fig. 1 using pass strains of 0.4 applied at  $1\text{ s}^{-1}$  and  $910^{\circ}\text{C}$ . Interpass times of a) 0.5s and b) 5s were employed. The presence of Nb delays both the forward and reverse transformation.

## 7.4 Discussion

### 7.4.1 Mean Flow Stress

In hot rolling, the evolution of the mean flow stress (MFS) provides evidence for the progress of metallurgical phenomena such as work hardening, recrystallization, precipitation as well as transformation. These events are dependent on the applied strain, strain rate, temperature of deformation, interpass time and alloy composition. In the current study, the MFS behavior provides considerable insight into the volume fraction of DT ferrite formed and retained from pass to pass. For this purpose, the areas under the flow curves of the rolling simulations of Figs. 7.3 and 7.4 were calculated; these are presented in Figs. 7.5a and 7.5b, respectively. Here the dotted lines refer to the 0.5s interpass time tests while the solid lines are associated with the 5s intervals.



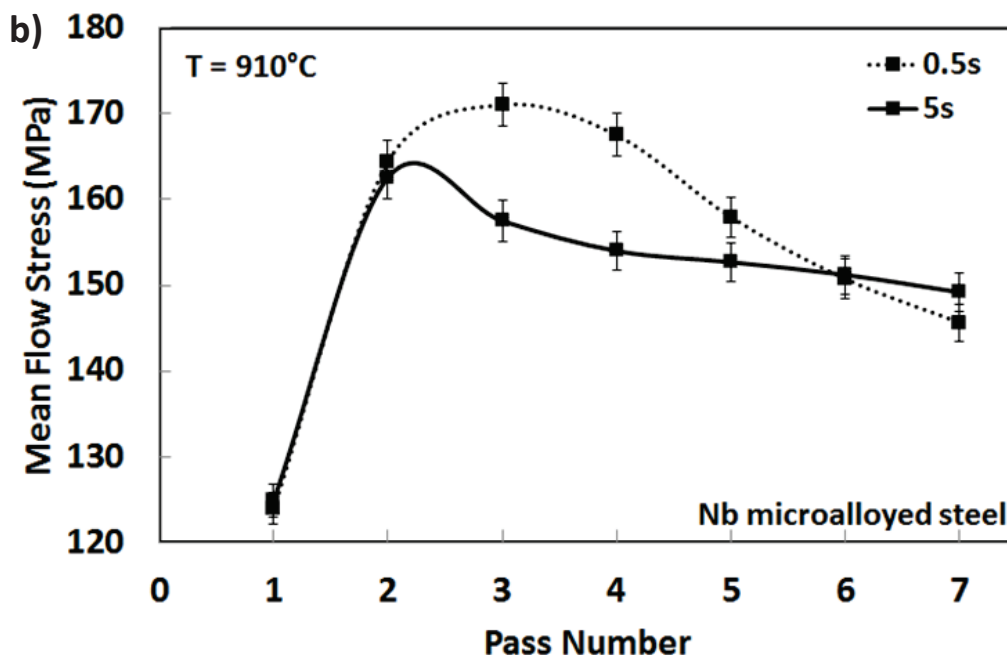


**Fig. 7.5.** MFS's calculated from the stress-strain curves of Figs. 7.2 and 7.3. a) For the C-Mn steel, the MFS decreases are immediately evident as a result of ferrite formation during each pass. b) In the Nb steel, however, there is more work hardening retained from pass to pass. Because of carbonitride precipitation during the longer interpass times, there is little decrease in the MFS.

It can be seen that the MFS increases somewhat from the first to the second pass for the C-Mn steel (Fig. 7.5a) due to strain accumulation. This increase amounts to about 10% for the 0.5s interpass times. There is little work hardening retained during the longer interpass time intervals because of recrystallization between passes. The MFS's drop continuously from the second through to the final pass, indicating that substantial amounts of ferrite are being formed in each pass. The amounts of MFS decrease suggest that about two-thirds of the initial austenite has transformed into ferrite. This interpretation is supported by the microstructures shown below. Note that the final pass MFS is lower for the 0.5s interpass times than for the 5s times. This is because more ferrite is retained after formation during the shorter intervals. DT ferrite tends to retransform into austenite during longer times [7.15].

The presence of Nb in the simulation of Fig. 7.5b delays the occurrence of static recovery and recrystallization. As a result, the amounts of work hardening retained after the first pass are

comparable in both the short and long interpass time simulations. This appears to be largely a solute drag effect, as there is insufficient time for any significant amount of niobium precipitation. However, in the succeeding passes, it is evident from the change in the flow curve shape that the sample with the longer times displays significantly more retained hardening than the one with the shorter times. This can be attributed directly to the carbonitride precipitation taking place during this more time-consuming simulation.

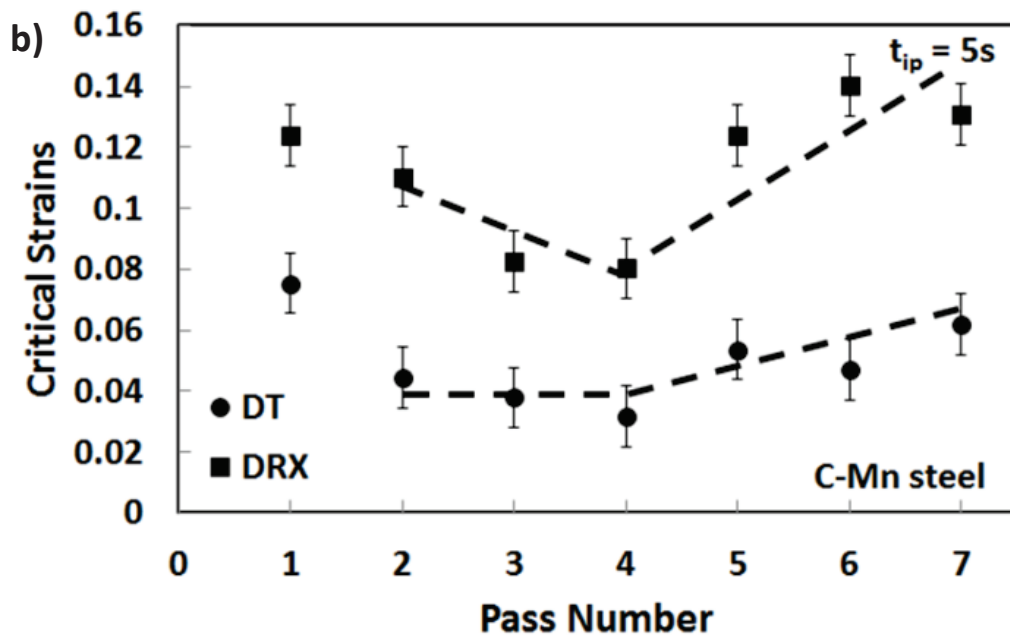
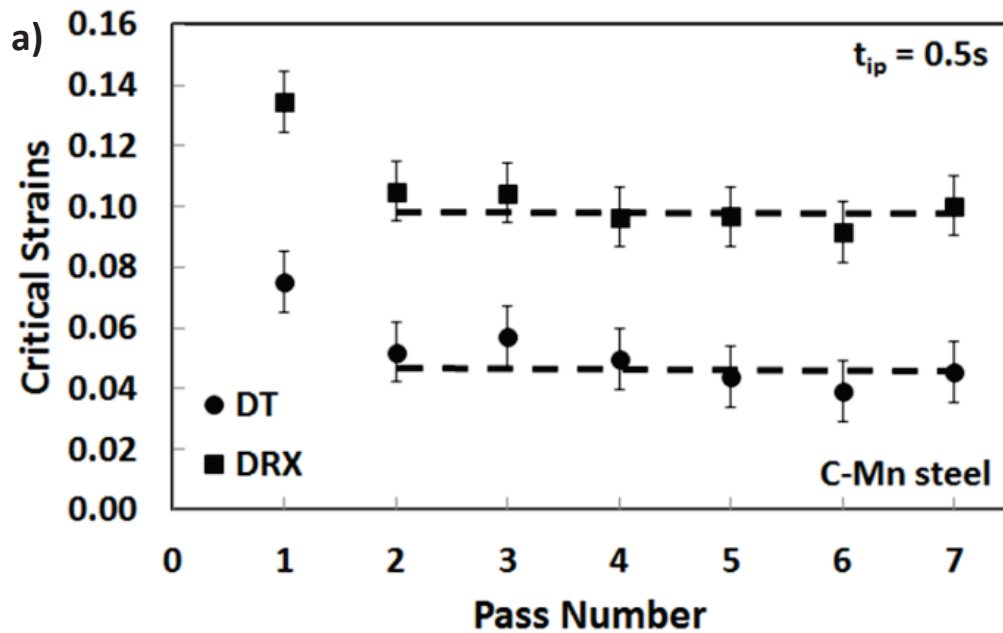
### 7.4.2 Critical Strains

The stress levels of the flow curves in Figs. 7.2 and 7.3 decrease steadily even though the strains were applied at a fixed temperature. To investigate the reason for this unusual behavior, the stress-strain curves were submitted to double differentiation. This was done by fitting 9<sup>th</sup> order (or higher in some cases) polynomials using the MatLab software. The initiation of softening was then identified with the inflection points in the strain hardening rate versus stress plots. The details of this approach are described in Ref. 20. These calculations confirmed that two distinct softening mechanisms are initiated well before the peak stress is attained. The first of these has been identified as dynamic transformation (DT) and the second as dynamic recrystallization (DRX). As shown in earlier investigations [7.21-7.23], the critical strain for DT is always lower than that for DRX.

The DT and DRX critical strains determined in the C-Mn simulation steel are illustrated in Figs. 7.6a and 7.6b for interpass times of 0.5s and 5s, respectively. In Fig. 7.6a, the first pass DT critical strain is about 0.08 while the critical strains for the succeeding passes (from the 2<sup>nd</sup> to the 7<sup>th</sup> pass) have approximately constant values of about  $\epsilon=0.05$ . This is because the first pass is applied to undeformed austenite whereas retained work hardening is present in the succeeding passes, reducing the DT critical strain in this way.

A similar plot of critical strain versus pass number associated with the longer interpass times is shown in Fig. 7.6b. Here the first pass critical strain agrees with the value presented in Fig. 7.6a. However, the DT critical strains calculated for the succeeding passes increase slightly with the pass number. This can be attributed to the greater amounts of recrystallization taking place

during the longer interpass times so that the retained austenite gradually approaches the undeformed state of the first pass material more and more closely. As a result, higher strains are required to initiate transformation in the succeeding passes.



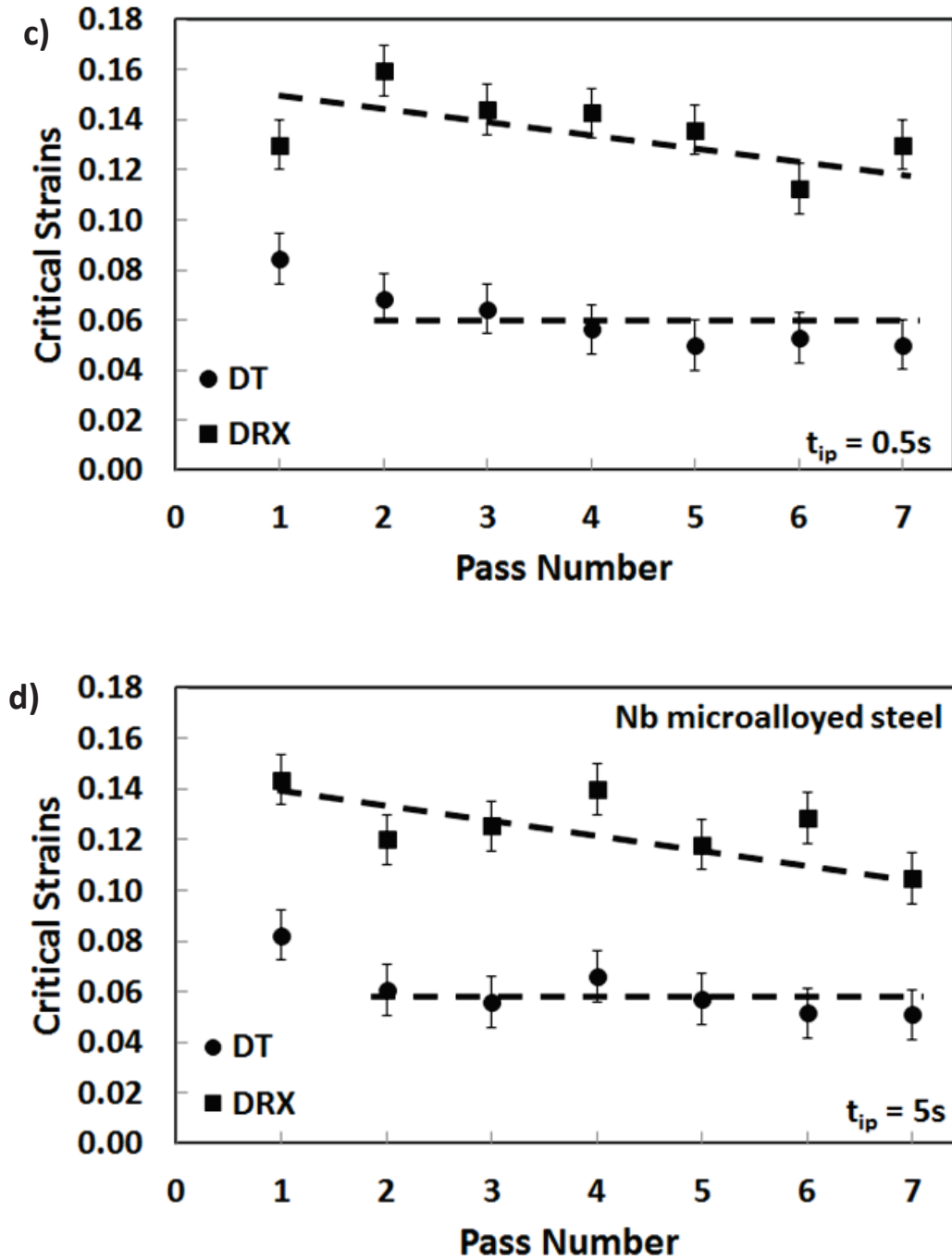
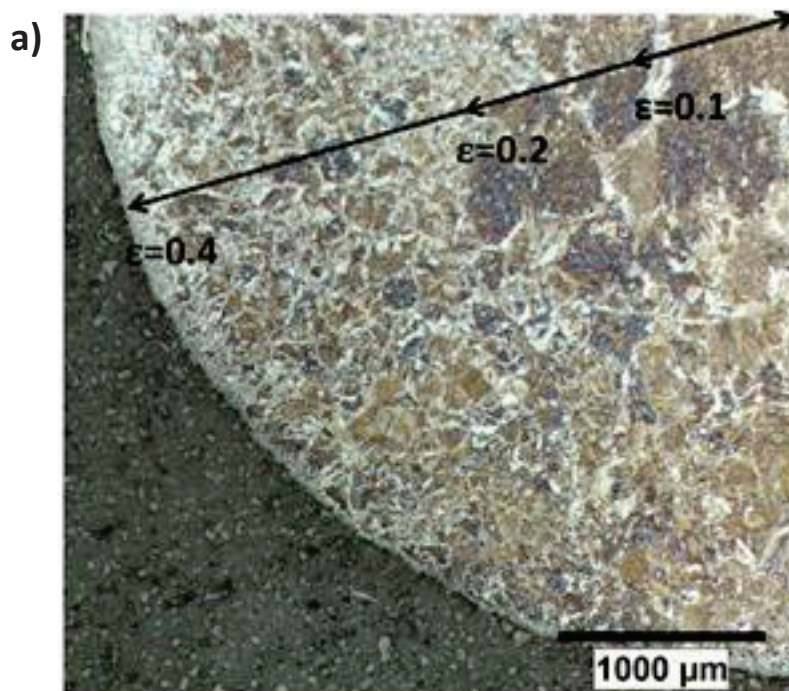


**Fig. 7.6.** The critical strains for dynamic transformation (DT) and dynamic recrystallization (DRX) in the C-Mn steel determined by means of double differentiation: a) 0.5s and b) 5s. Critical strains in the Nb steel subjected to interpass times of c) 0.5s and d) 5s. Here the first pass always displays the highest critical strain in all the simulations.

A similar double-differentiation procedure was applied to the Nb steel flow curves of Fig. 7.4. The results obtained are presented in Figs. 7.6c and 7.6d for interpass times of 0.5s and 5s, respectively. Here the first pass DT critical strains (about 0.09) are slightly higher than the values for the C-Mn steel. In the succeeding passes, they remain approximately constant at about 0.05. In this particular case, there is very little to no retransformation. For this reason, the amount of ferrite accumulated from pass to pass is expected to be comparable for both the 0.5s and 5s interpass times. The microstructures pertaining to the two simulations will be presented in the section that follows.

The initiation of DT and DRX at the double minima was verified by means of both optical and electron microscopy. An example of the microstructure produced in a torsion test carried out to a strain of 0.4 at 930 °C and  $1\text{ s}^{-1}$  is presented in Fig. 7.7a. Here it can be seen that ferrite formation was initiated at strains in the range 0.08 to 0.1, while recrystallization began at higher strains, about 0.15 to 0.2. These values are consistent with the calculated critical strains displayed in Fig. 7.6. It is important to note that the “initiation” of DT corresponds to the presence of about 5% volume fraction of ferrite. The fresh ferrite formed in this way consists of plates of Widmanstätten ferrite [7.6,7.24,7.25], as illustrated in Fig. 7.7b. As these plates are of near identical orientation, they coalesce during continued straining into polygonal grains.



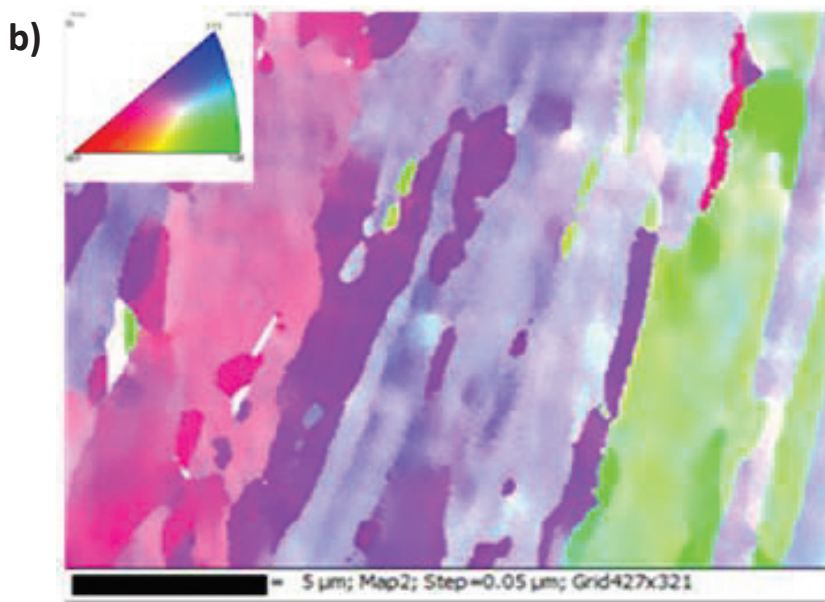


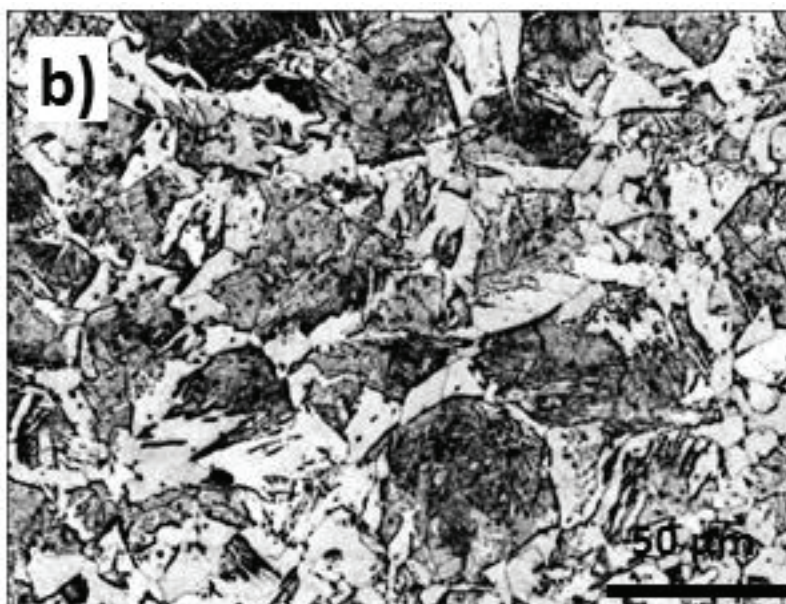
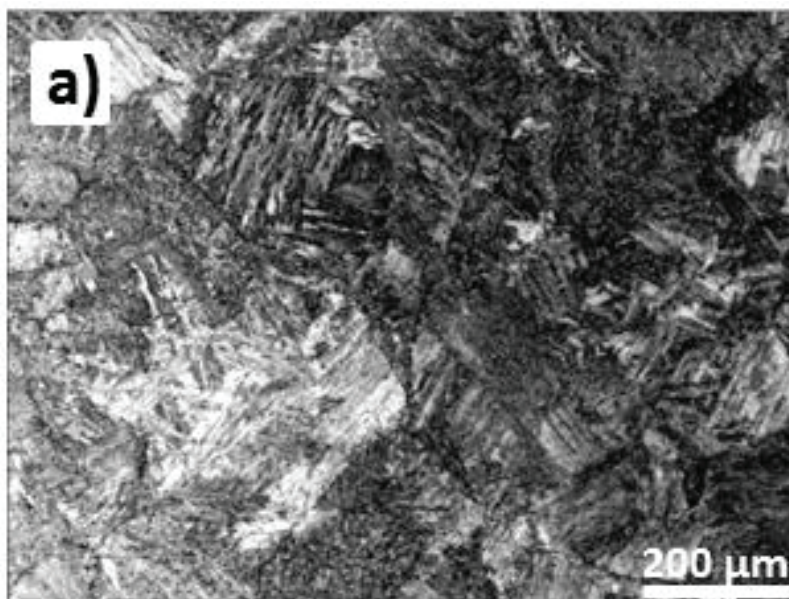
**Fig. 7.7.** a) Microstructure of a torsion sample deformed to a strain of 0.4 at the outer radius at  $930^{\circ}\text{C}$  and  $1\text{s}^{-1}$ . The strain in this sample increases linearly from zero along the axis to 0.4 at the outer radius. Here the presence of DT ferrite can be first seen at a critical strain of about 0.1, while DRX of the prior austenite was initiated at a critical strain of approximately 0.15. b) Widmanstätten ferrite plates formed during deformation at  $1200^{\circ}\text{C}$ . These plates coalesce into polygonal grains upon continued straining.

#### 7.4.3 Volume fraction of ferrite formed per pass

The transformation from austenite to ferrite at temperatures above the  $A_{e3}$  involves the displacive conversion of the fcc crystal structure followed by carbon diffusion, as discussed elsewhere [7.24]. Widmanstätten ferrite plates are nucleated and colonies of these plates coalesce into polygonal grains upon further straining [7.25]. The volume fraction of DT ferrite depends on the strain, interpass time, chemical composition and temperature. The dilatation associated with ferrite formation increases the volume flow rate (but not the mass flow rate) as a bar passes through a strip mill [7.15]. These changes can lead to shape control problems if not properly taken into account.

The MFS's associated with the strip rolling simulations described above indicated that significant amounts of ferrite formed and were retained during the schedule and that the volume fraction of ferrite increased continuously right up to the last pass. Some of the optical

microstructures related to the 0.5s interpass time C-Mn simulation are presented in Fig. 7.8. Here the martensite (prior austenite) appears dark while the ferrite is light. It can be seen that the volume fraction of ferrite present after the 1<sup>st</sup> pass (Fig. 7.8a) is about 15%, and that it is largely of plate form. These plates coalesced into polygonal grains in the succeeding passes, as shown in Fig. 7.8b (after the 3<sup>rd</sup> pass), Fig. 7.8c (after the 5<sup>th</sup> pass) and Fig. 7.8d (after the 7<sup>th</sup> pass). Here the ferrite fractions increased to about 56%, 65% and 72%, respectively.



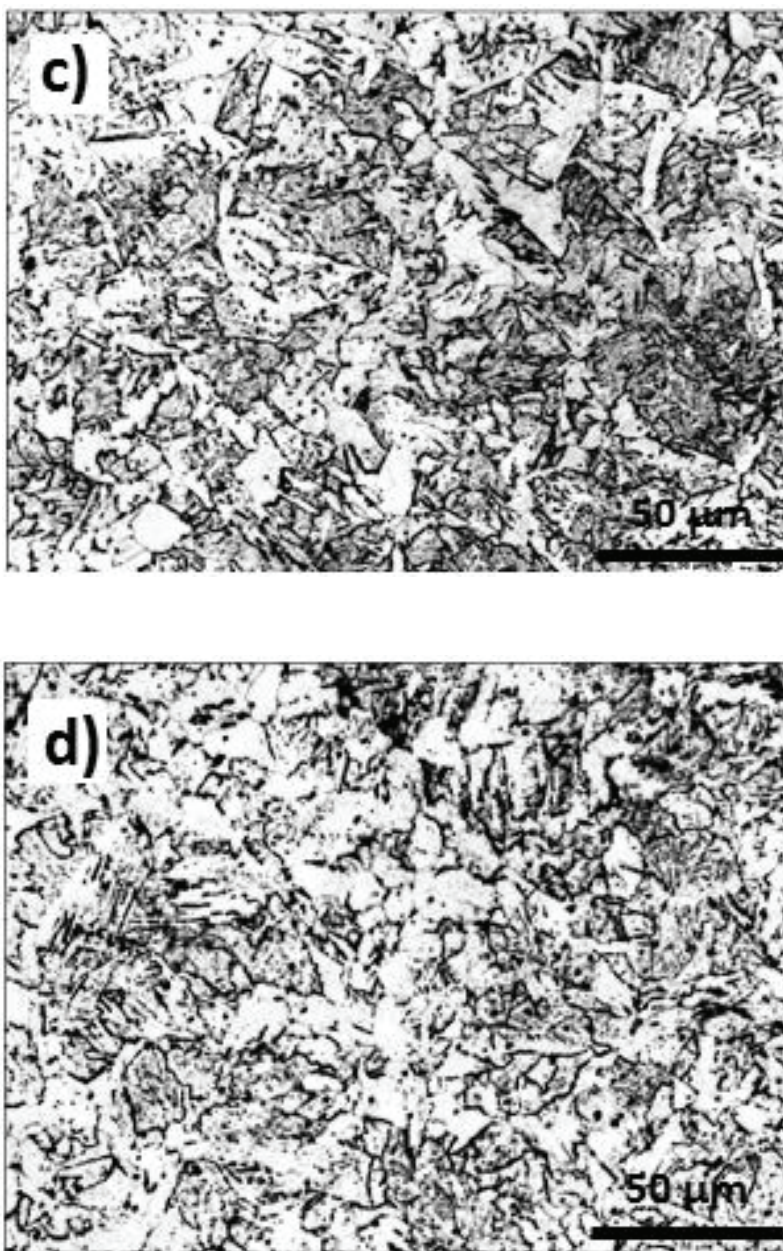
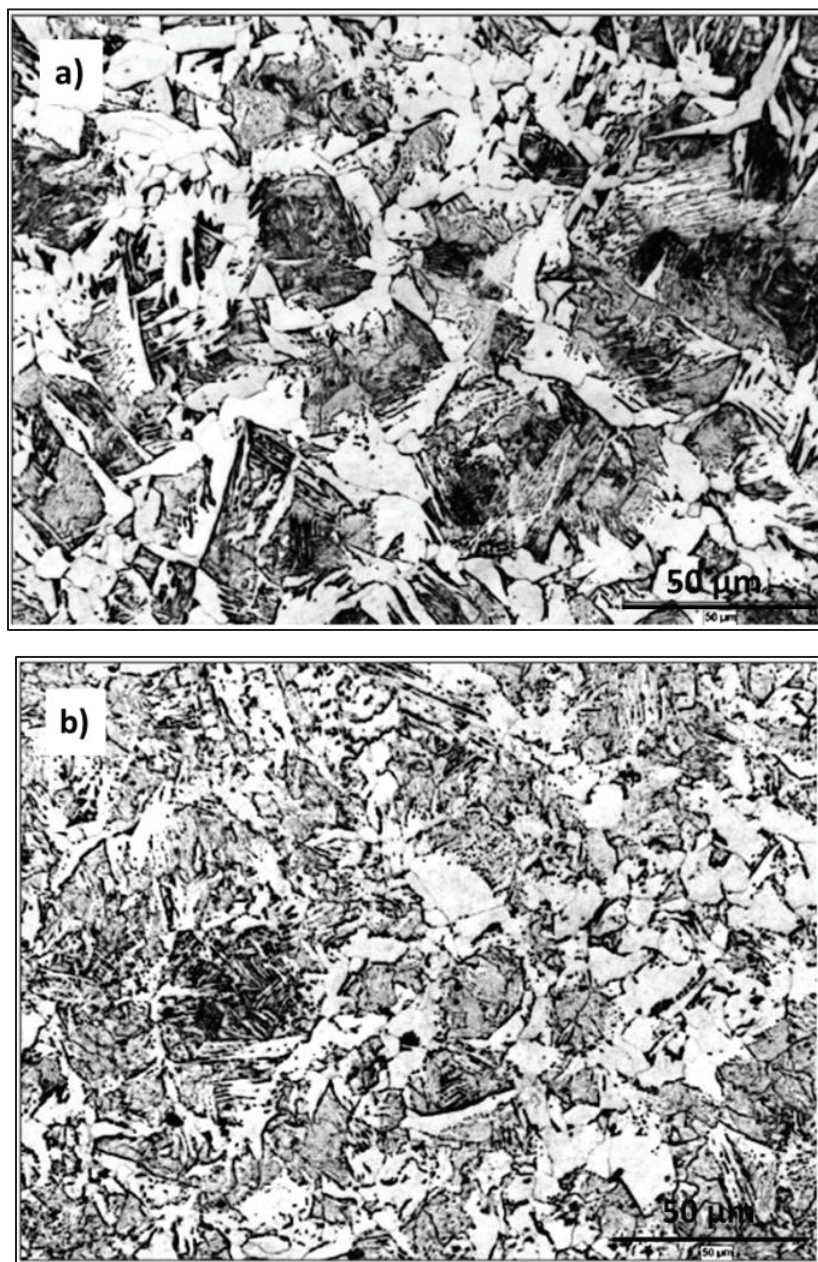


**Fig. 7.8.** Optical microstructures of the C-Mn steel subjected to the 7-pass simulation at 930°C with interpass times of 0.5s. The samples were quenched immediately after the: a) first pass, b) 3<sup>rd</sup> pass, c) 5<sup>th</sup> pass, and d) final pass. Dark regions are martensite (prior austenite) while the light regions are ferrite. The amount of ferrite increases progressively right up to the last pass.

These microstructures reveal that the volume fraction of ferrite increased progressively with pass number (i.e. with strain) when the 0.5s interpass times were employed. The

microstructures associated with the 5s interpass time simulation are illustrated in Fig. 7.9. In this case, the volume fractions of ferrite after the 3<sup>rd</sup> (Fig. 7.9a), 5<sup>th</sup> (Fig. 7.9b) and 7<sup>th</sup> (Fig. 7.9c) passes are about 48%, 54% and 52%, respectively. These values are significantly lower than those associated with the shorter times. This can be attributed to the retransformation that takes place between passes, resulting in the retention of reduced amounts of DT. In this example, about 20% of the ferrite retransformed back into austenite when the interpass times were increased from 0.5s to 5s.



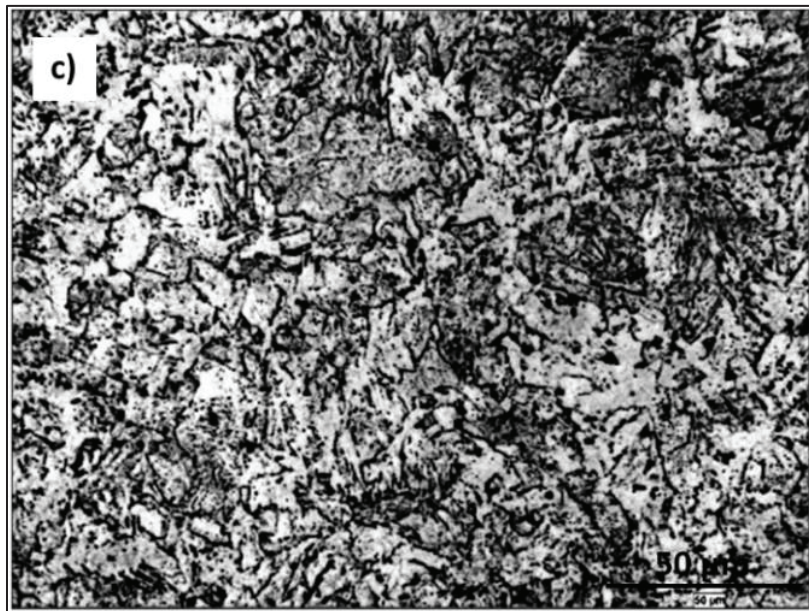
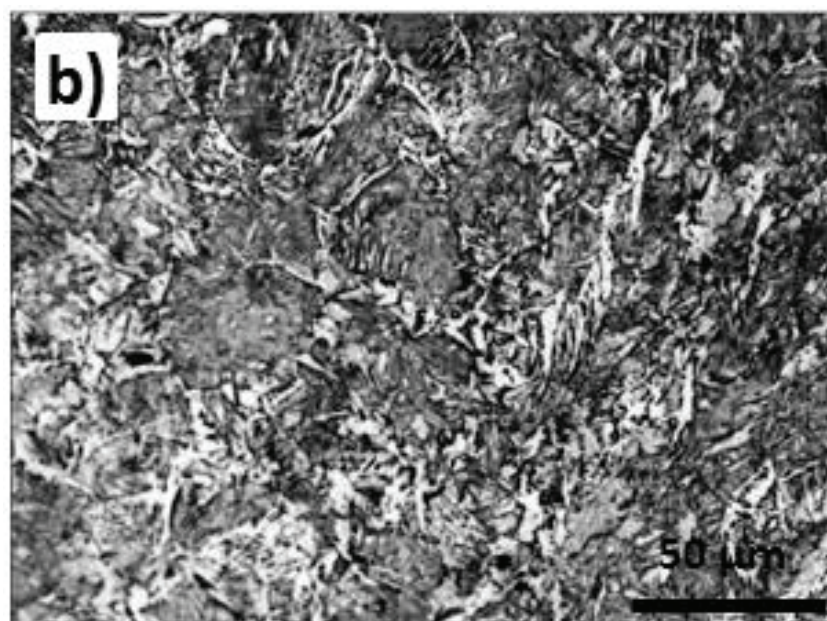
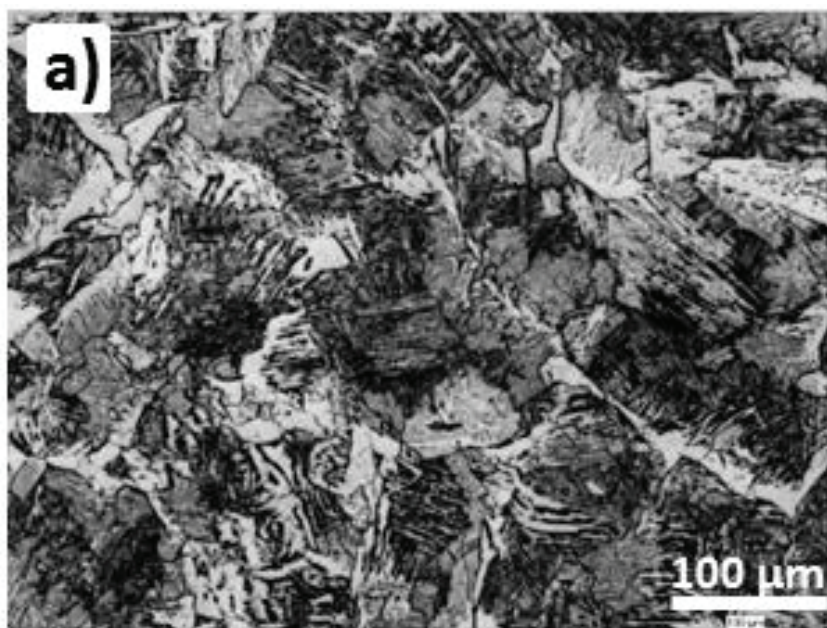


**Fig. 7.9.** Optical microstructures of the C-Mn steel subjected to the 7-pass simulation at 930°C with interpass times of 5s. The samples were quenched immediately after the: a) 3<sup>rd</sup> pass, b) 5<sup>th</sup> pass, and c) final pass. Dark regions are martensite (prior austenite) while the light regions are ferrite. There is less ferrite than in the short time simulations because of reverse transformation.

The microstructures related to the Nb steel simulation with interpass times of 0.5s are illustrated in Fig. 7.10. The presence of significant amounts of ferrite, about 25%, is evident in the 1<sup>st</sup> pass microstructure (Fig. 7.10a), in which both polygonal grains and ferrite plates can be seen. During the succeeding passes, the volume fraction of ferrite increases to 30%, 39%, and finally to 51% after the 3<sup>rd</sup> (Fig. 7.10b), 5<sup>th</sup> (Fig. 7.10c) and 7<sup>th</sup> (Fig. 7.10d) passes, respectively. The ferrite is mostly present as polygonal ferrite grains, which have nucleated both along the grain boundaries and within the prior austenite grains.

The microstructures pertaining to the 5s interpass times are illustrated in Fig. 7.11. Here the volume fractions of ferrite are about 32%, 40% and 49% after the 3<sup>rd</sup> (Fig. 7.11a), 5<sup>th</sup> (Fig. 7.11b) and 7<sup>th</sup> (Fig. 7.11c) passes, respectively. These values are similar to those associated with the shorter interpass times. It thus appears that there is little retransformation in the Nb steel even when the interpass time is increased from 0.5s to 5s. In this case, it is probably Nb solute drag

that prevents the reverse transformation [7.15]. As the nature of the latter is diffusional rather than displacive [7.24], the addition of Nb can have a larger effect on the kinetics than during the forward transformation. The current results thus support those of an earlier study [7.26] regarding another Nb steel, in which retransformation was delayed by up to 200s.



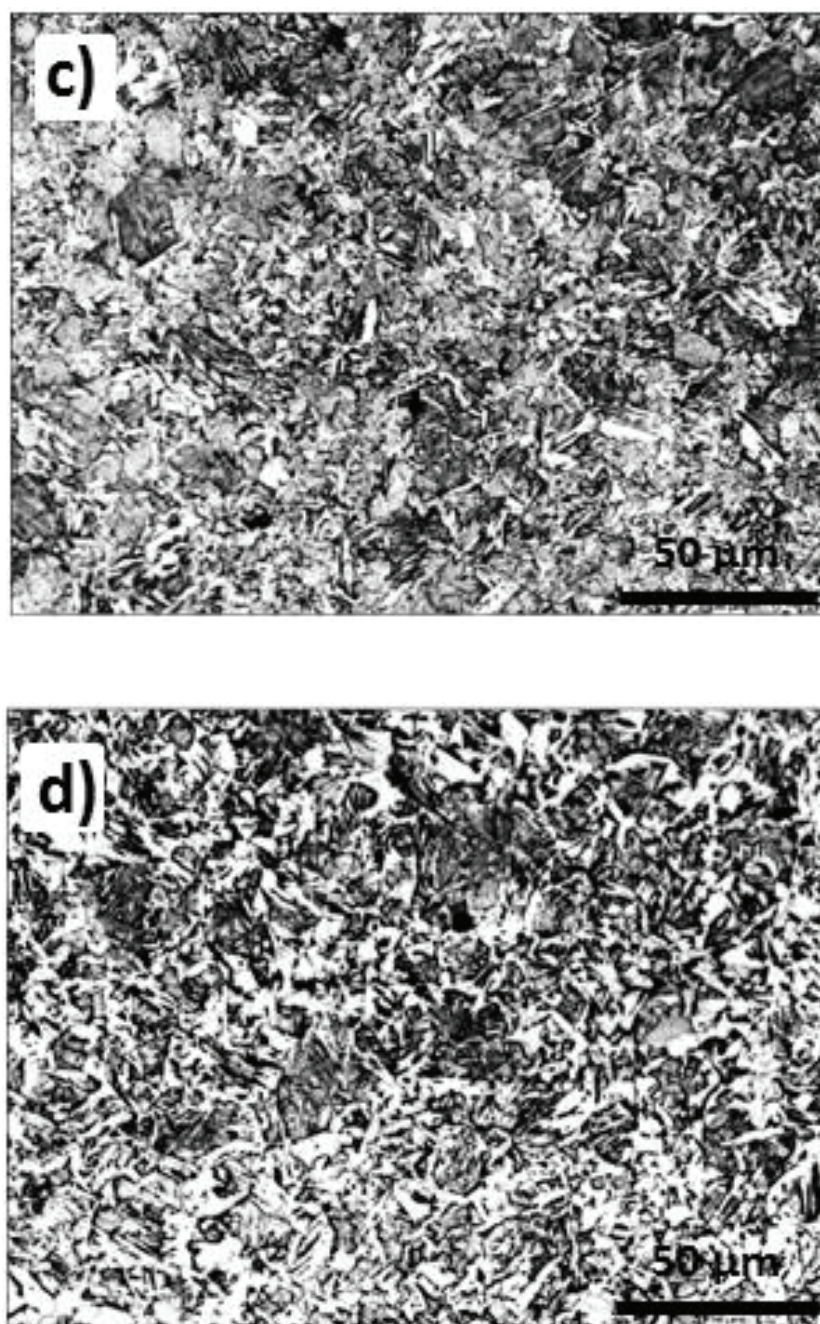
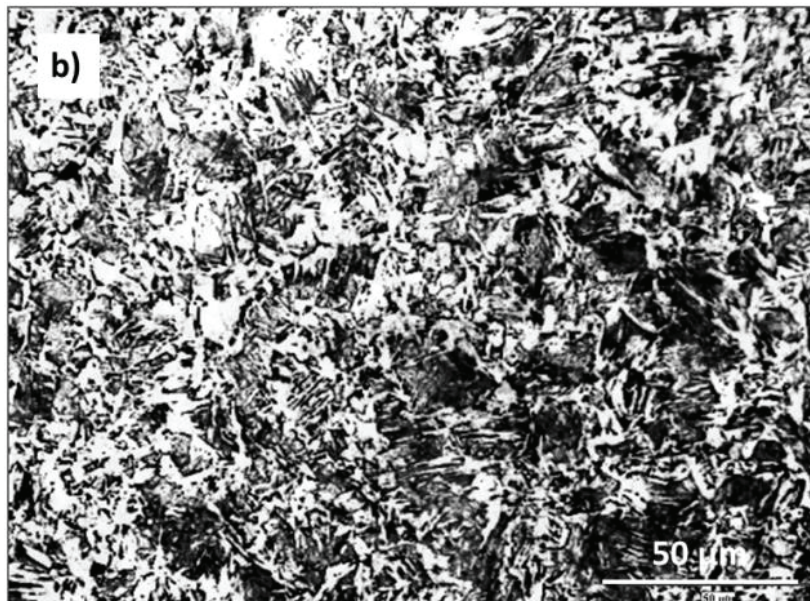
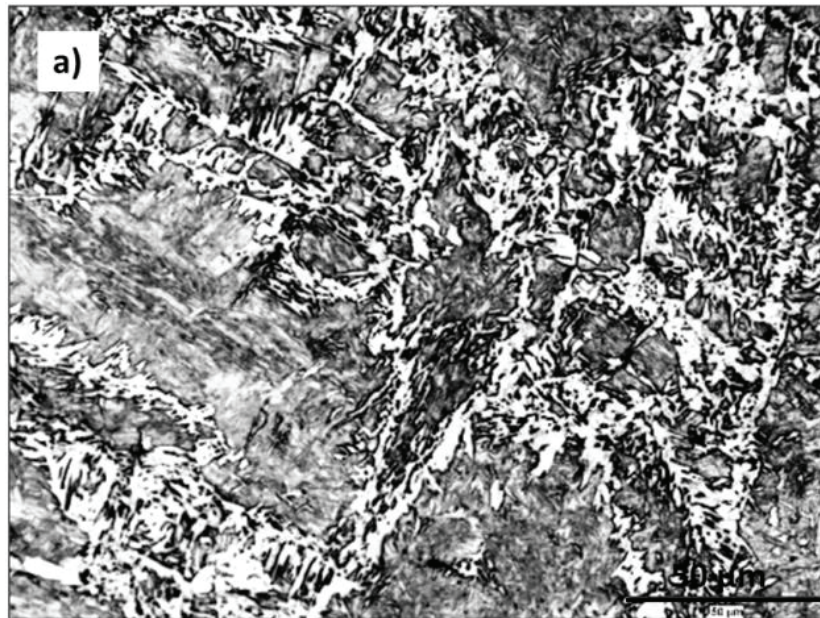


**Fig. 7.10.** Optical microstructures of the Nb steel subjected to the 7-pass simulation at 910°C with interpass times of 0.5s. The samples were quenched immediately after the: a) first pass, b) 3<sup>rd</sup> pass, c) 5<sup>th</sup> pass, and d) final pass. Dark regions are martensite (prior austenite) while the light regions are ferrite. The amount of ferrite increases progressively right up to the last pass.



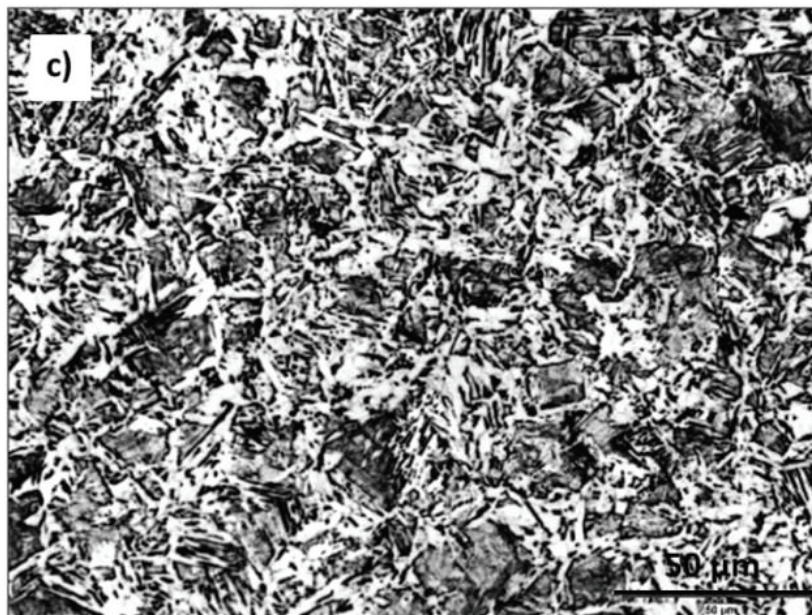


**Fig. 7.11.** Optical microstructures of the Nb steel subjected to the 7-pass simulation at 910°C with interpass times of 5s. The samples were quenched immediately after the: a) 3<sup>rd</sup> pass, b) 5<sup>th</sup> pass, and c) final pass. Dark regions are martensite (prior austenite) while the light regions are ferrite. The amounts of ferrite are comparable with those produced in the short time simulations. The presence of Nb significantly retards the reverse transformation.

The ferrite volume fractions produced in the C-Mn and Nb steels are summarized in Figs. 7.12a and 7.12b, respectively. Here the Vickers hardness measurements pertaining to each pass of the 0.5s simulation have been added and are displayed on the right-hand axes. In Fig. 7.12a, the increases in the amount of ferrite are accompanied by corresponding decreases in the overall hardness, from 259 HV before the first pass down to about 175 HV immediately after the final pass. Note that since the C-Mn steel only contains about 0.06 wt%C, the martensite hardness falls in the expected 200 to 300 HV range [7.27]. However, the hardness of the martensite in the Nb steel (with 0.11 %wt C) is somewhat higher, as shown in Fig. 7.12b. Here the measured hardness before the initial pass is 341 HV and the presence of ferrite introduced by the deformation lowers it to about 273 HV.

As a check, hardnesses of local ferrite and martensite regions were also measured, as illustrated in Fig. 7.13. The Vickers microhardness of the ferrite in the C-Mn steel is approximately 158 HV

(Fig. 7.13a) while that of the martensite is 264 HV (Fig. 7.13a). These values are in agreement with the macrohardnesses shown above.

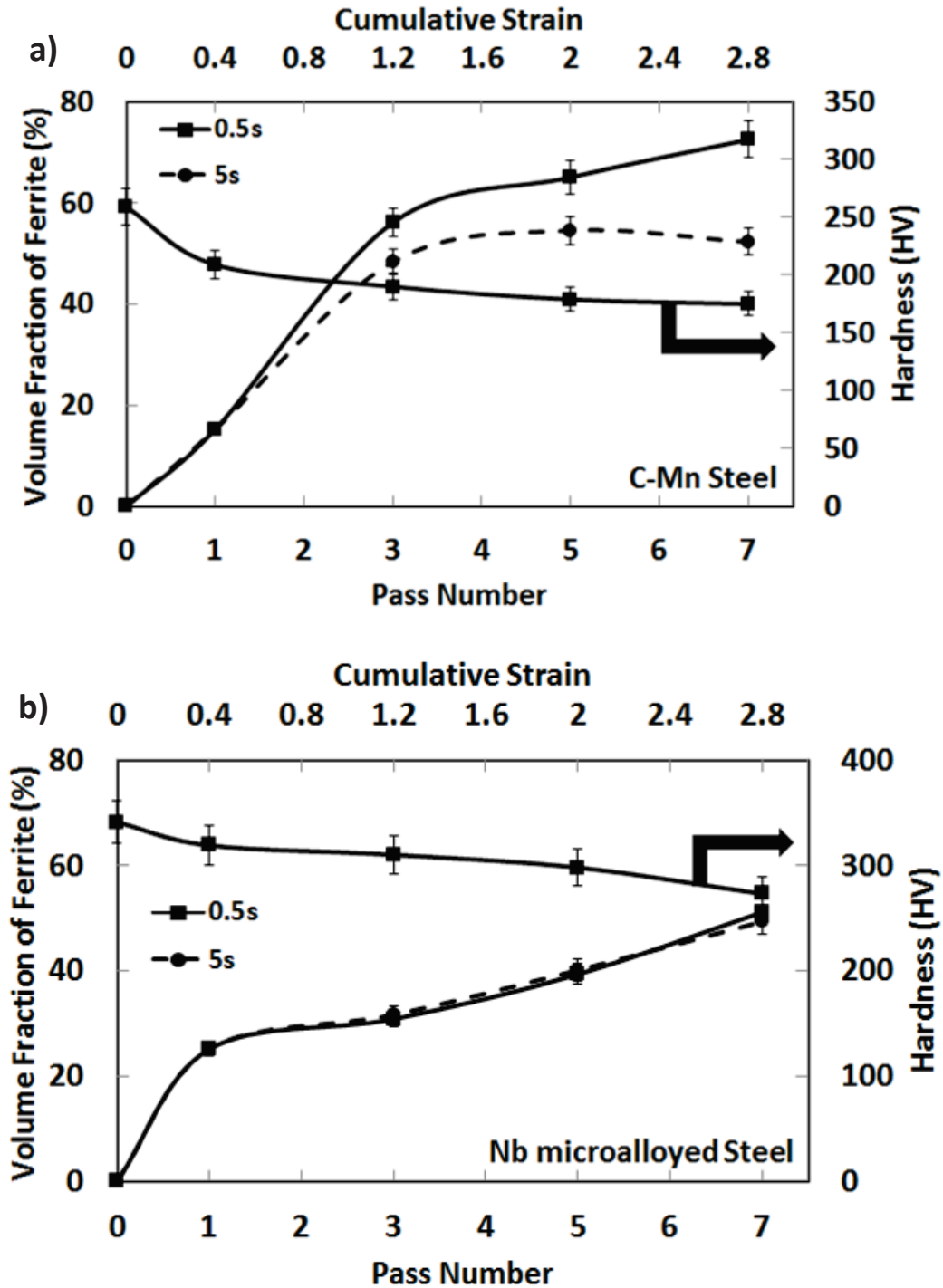


**Fig. 7.12.** Dependence of the volume fraction of ferrite on pass number for the a) C-Mn and b) Nb steels. The increasing amounts of ferrite from pass to pass are in agreement with the hardness measurements displayed on the right hand axis.

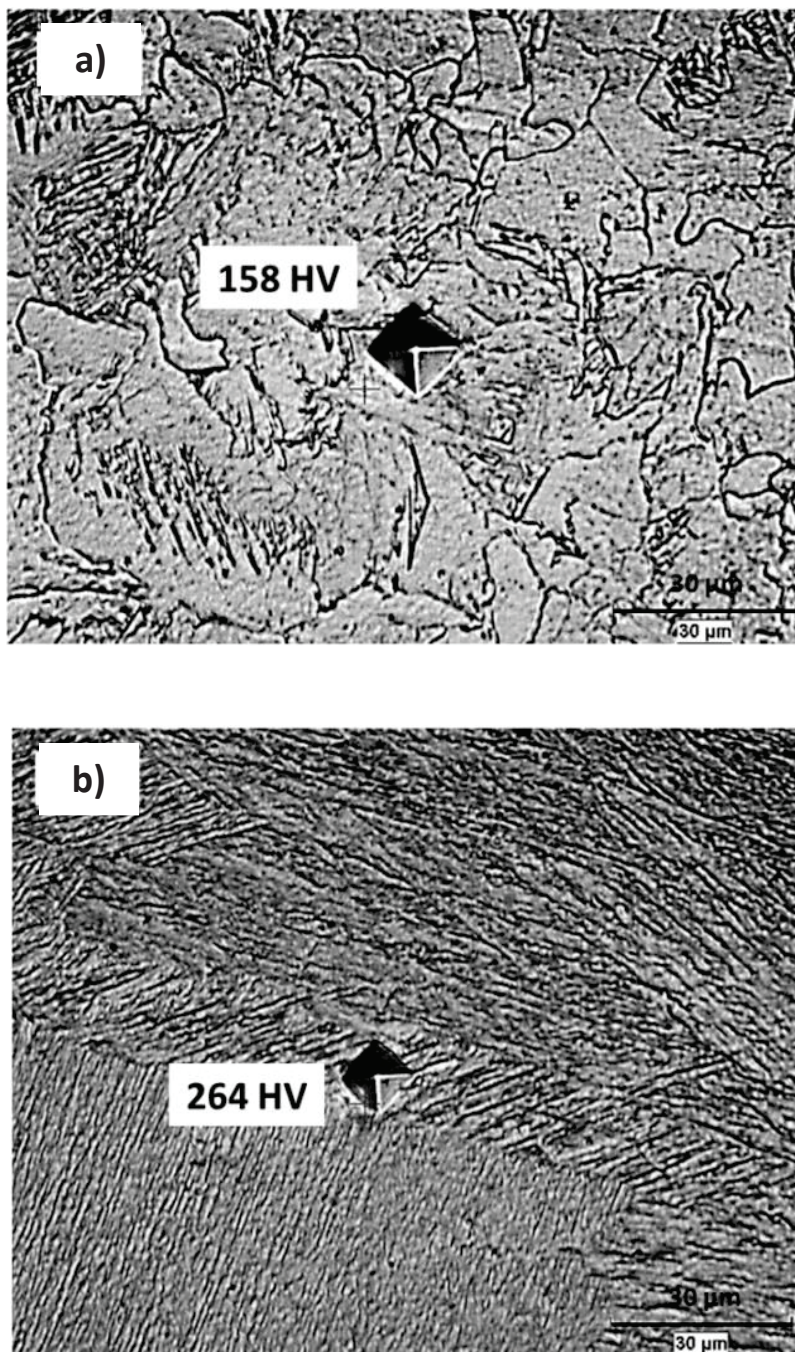


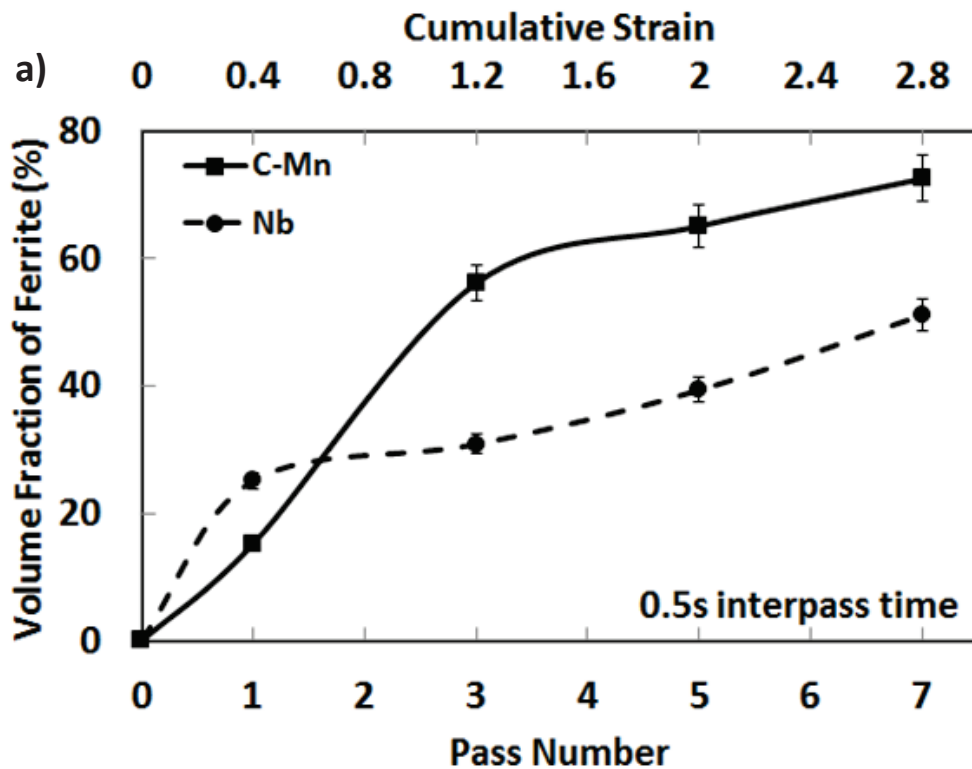
**Fig. 7.13.** Vickers microhardnesses of the a) ferrite and b) martensite in the C-Mn steel.

#### 7.4.4 Effect of Nb addition on the behavior

The amounts of ferrite formed in the C-Mn and Nb steels during the rolling simulations are presented in Figs. 7.14a and 7.14b. During the 0.5s simulations (Fig. 7.14a), more ferrite was

produced in the C-Mn steel than in the microalloyed steel. This amounted to about 20% more ferrite in the C-Mn than in the Nb steel. These results confirmed that the addition of Nb hinders the forward displacive transformation and not just the reverse diffusional transformation. These observations complement the higher measured DT critical strains in the Nb steel than in the C-Mn steel. Given that Si addition decreases the driving force for transformation softening [7.8], it is possible that Nb has a similar effect. This would explain why less ferrite is formed in the Nb steel on the application of a given strain than in the C-Mn steel.

The situation is somewhat more complicated in the case of the 5s simulations (Fig. 7.14b). As shown above, the retransformation of ferrite back into austenite in the C-Mn steel lowers the ferrite volume fraction by about 20% because of the longer times available for this to occur (see Fig. 7.12a). However, in the Nb steel, no retransformation was observed during the 5s interpass times. Although the presence of Nb retards the production of DT ferrite, it also retards the reverse transformation. In this way, comparable amounts of ferrite were present in the final pass in the two steels.



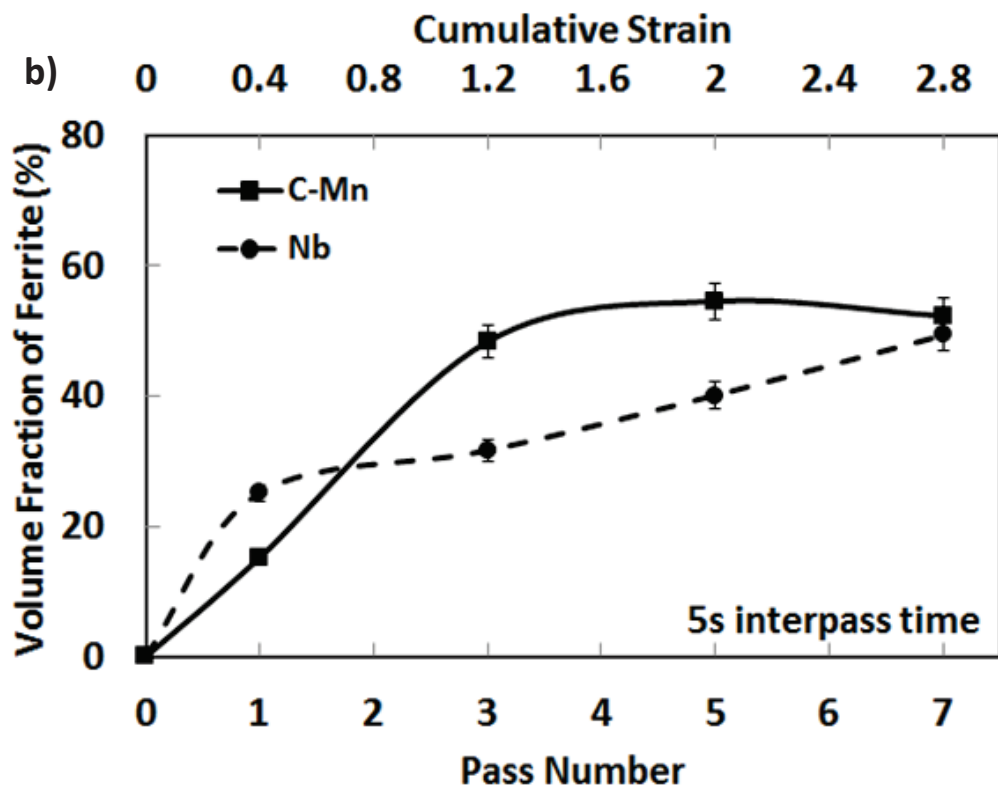


**Fig. 7.14.** a) Ferrite volume fraction versus pass number when interpass times of 0.5s are employed. More ferrite is formed and retained in the C-Mn than in the Nb steel. b) Ferrite volume fraction versus pass number when interpass times of 5s are employed. In the C-Mn steel, reverse transformation decreases the amount of ferrite volume fraction. The addition of Nb impedes the forward transformation and prevents reverse transformation.

## 7.5 Conclusions

1. The strip rolling simulations carried out on the C-Mn steel revealed that ferrite volume fractions of about 10% form during each  $\epsilon=0.4$  pass, with the volume fraction increasing continuously until the last pass. Concurrent decreases in flow stress level and MFS were also observed during these simulations. The view that the ferrite forms dynamically is supported by the extensive optical microscopy results, which were accompanied by hardness measurements.
2. The strip rolling simulations carried out on the Nb steel revealed that ferrite volume fractions of about 7% form during each  $\epsilon=0.4$  pass, with the volume fraction increasing continuously until the last pass. Concurrent decreases in flow stress level and MFS were also observed during

these simulations. Extensive optical microscopy results again supported the conclusion that the ferrite formed dynamically. Comparison of the behaviors of the C-Mn and Nb steels during the short interpass time simulations indicates that more ferrite forms in the C-Mn steel than in the Nb steel because Nb addition impedes the forward transformation.

3. Critical strain measurements provided further insight into the forward transformation. The DT critical strains are always highest in the first pass and then adopt lower values in the succeeding passes because of the retention of work hardening.

4. The fraction of DT ferrite formed and retained in the C-Mn simulations is significantly higher when short interpass times are employed. Here the metastable DT ferrite retransforms back into the more stable austenite during the longer intervals. This led to about a 20% volume fraction decrease when the interval was increased from 0.5s to 5s.

5. The addition of Nb impedes the displacive forward transformation and retards the diffusional reverse transformation rates. The former occurs because Nb addition reduces the effectiveness of transformation softening; the latter by means of a solute drag effect.

## 7.6 References

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## Chapter 8

### Conclusions

The dynamic transformation of austenite to ferrite above the  $A_{e3}$  temperature was studied in detail in a plain C-Mn and a Nb microalloyed steel. For this purpose, compression and torsion tests were carried out over the temperature range 880 °C to 1350 °C applied at a strain rate of  $1 \text{ s}^{-1}$ . These experiments were performed under a 95%Ar-5%H<sub>2</sub> atmosphere (in the MTS machines) and in a vacuum atmosphere (Gleeble system). The deformation temperatures were all above the ortho- and para-equilibrium  $A_{e3}$  transformation temperatures.

The main objectives of the present investigation were the following: i) to develop a thermodynamic explanation for the occurrence of DT above the  $A_{e3}$  temperature; and, ii) to determine the effect of DT on the rolling load during strip rolling.

The first objective was addressed in **Chapters 3, 4 and 5**. In these investigations, a new transformation softening model was proposed (**Chapter 3**) supplemented by experimental results. The effect of alloying elements, such as Mn and Si (**Chapter 4**), as well as the formation of ferrite at temperatures high in the austenite phase field (**Chapter 5**) were also studied in detail. These analyses led to the following conclusions:

1. The net softening during transformation can be considered as the driving force for phase change. This is given by the difference between the flow stress of the work hardened austenite at the critical strain and the yield stress of the fresh Widmanstätten ferrite that takes its place. The obstacle to the transformation consists of three parts: i) the free energy difference between austenite and Widmanstätten ferrite, ii) the shear accommodation work that is imposed on the neighborhood by the transformation, and iii) the dilatation work associated with making room for the lower density ferrite.

2. The mechanical driving force for DT can be readily evaluated by plotting the flow stresses and yield stresses against inverse absolute temperature. The flow stress of Widmanstätten ferrite at temperatures above the  $Ae_3$  can also be estimated by linear extrapolation against inverse absolute temperature, in this case from polygonal and Widmanstätten ferrite data obtained below the  $Ae_1$ . This enables the derivation of the driving force required by the present model.
3. The stress-difference model indicates that dynamic transformation cannot be induced at applied stresses in the vicinity of the yield stress. Instead, the driving force must be increased by work hardening until a minimum critical stress, and therefore a critical strain, is attained.
4. The use of a constant critical strain as in earlier models can account for the occurrence of dynamic transformation at temperatures up to 54°C above the  $Ae_3$ . However, if allowance is made for the increase in the critical strain with temperature (together with an approximately constant critical stress), then the dynamic transformation observations reported in the literature at temperatures as high as 150-200°C above the  $Ae_3$  can be readily justified.
5. At the critical strain, both moderately well oriented (Schmid factor,  $m = 0.3$ ) and ideally oriented ( $m = 0.5$ ) grains can participate in the transformation. As the temperature interval above the  $Ae_3$  is increased, transformation can only take place in grains of more and more ideal orientation, that is those associated with higher and higher  $m$ -values.
6. The addition of Mn increases both the driving force for the transformation as well as the free energy obstacle to ferrite formation. This leads to a small net decrease in the driving force. Conversely, Si addition increases the driving force and decreases the obstacle height, thus leading to a large increase in the net driving force. The present thermodynamically-based model thus leads to insights regarding the effects of Mn and Si addition that are not intuitively obvious. These changes are likely to influence the volume fractions of DT ferrite produced in these alloys accordingly.
7. Both dynamic transformation and dynamic recrystallization can be seen to operate in the temperature range 1000 °C to 1350 °C. By employing the double differentiation

method, these are seen to be initiated at average critical strains of about 0.10 and 0.15, respectively. The critical strain for DRX decreases monotonically with increasing temperature. By contrast, the critical strain for DT first increases with temperature up to about 1080 °C and then decreases as the delta ferrite formation temperature is approached. These observations are consistent with the shape of the free energy barrier opposing the transformation, which rises to a maximum at 1080 °C. The shape of the barrier also explains why less ferrite is formed at a given strain at temperatures in the vicinity of the peak than at higher or lower temperatures.

8. The present results indicate that Widmanstätten ferrite can form at temperatures as high as 1350 °C, i.e. 480 °C above the  $A_{e3}$  temperature. Thus it appears that DT ferrite can be formed at any temperature within the austenite phase field.

The second part of the current work involved the torsional simulation of strip rolling; this addressed the industrial implications of DT. This is covered in **Chapters 6** and **7**. Rolling simulations employing continuous cooling and isothermal conditions were performed over the temperature range 880 °C to 1000 °C. The effect of DT on the rolling load was determined by analysis of the flow curves and MFS's and measurement of the volume fraction of DT ferrite formed from pass to pass. This led to the following general conclusions:

1. The rolling simulations carried out on the plain C steel reveal that the MFS decreases significantly when short interpass times are used and increases less rapidly than expected when this interval is increased in length. These trends are directly attributable to the dynamic transformation of austenite to ferrite within the austenite phase field, i.e. above the  $A_{e3}$ . The microstructures observed using optical microscopy indicate that more DT ferrite forms and is retained when the interpass intervals are shorter than when longer intervals are used.
2. The simulations carried out on the Nb steel reveal that the MFS decreases substantially from the 2<sup>nd</sup> to the succeeding passes when short interpass times are used and displays a low rate of increase when this interval is increased in length. These trends can be directly associated with the formation of DT ferrite from pass to pass during the

simulation. It appears that the shorter interpass times both produce more DT ferrite and permit less reversion to austenite than the longer times. The optical micrographs indicate that ferrite forms during straining and that most of it is retained during unloading.

3. Comparison of the behavior of the plain C and Nb steels reveals that Nb addition retards both the forward as well as the reverse transformation. As a result, more DT ferrite is formed in the plain C steel than in the Nb steel. Most of the DT ferrite formed in the plain C steel is able to retransform into austenite after holding for 15s. By contrast, the reverse transformation is severely retarded in the Nb steel. As a result, there is little or no retransformation after 15s of holding under the present conditions.
4. The strip rolling simulations carried out on the C-Mn steel revealed that ferrite volume fractions of about 10% form during each  $\epsilon=0.4$  pass, with the volume fraction increasing continuously until the last pass. Concurrent decreases in flow stress level and MFS were also observed during these simulations. The view that the ferrite forms dynamically is supported by the extensive optical microscopy results, which were accompanied by hardness measurements.
5. The strip rolling simulations carried out on the Nb steel revealed that ferrite volume fractions of about 7% form during each  $\epsilon=0.4$  pass, with the volume fraction increasing continuously until the last pass. Concurrent decreases in flow stress level and MFS were also observed during these simulations. Extensive optical microscopy results again supported the conclusion that the ferrite forms dynamically. Comparison of the behaviors of the C-Mn and Nb steels during the short interpass time simulations indicates that more ferrite forms in the C-Mn steel than in the Nb steel because Nb addition impedes the forward transformation.
6. Critical strain measurements provided further insight into the forward transformation. The DT critical strains are always highest in the first pass and then adopt lower values in the succeeding passes because of the retention of work hardening.
7. The fraction of DT ferrite formed and retained in the C-Mn simulations is significantly higher when short interpass times are employed. Here the metastable DT ferrite

retransforms back into the more stable austenite during the longer intervals. This led to about a 20% volume fraction decrease when the interval was increased from 0.5s to 5s.

8. The addition of Nb impedes the displacive forward transformation and retards the diffusional reverse transformation rates. The former occurs because Nb addition reduces the effectiveness of transformation softening; the latter by means of a solute drag effect.

## Chapter 9

### Contributions to Original Knowledge

The concept and occurrence of dynamic transformation were studied in detail in a plain C-Mn and a Nb microalloyed steel. This phenomenon was explored here by performing compression and torsion tests at temperatures above the  $A_{e3}$ . The torsion simulation of strip rolling was also carried out using continuous cooling and isothermal conditions to determine the effect of DT on the rolling load. As a result, this work was able to come up with a new thermodynamic model to accurately explain the occurrence of dynamic transformation well above the  $A_{e3}$  temperature. This approach also provides a better understanding regarding the industrial implications of this unusual phenomenon. The work described above led to the following contributions to original knowledge.

1. The stored energy and mechanical activation models described in the literature were not able to account for the occurrence of dynamic transformation at very low strains (i.e.  $\epsilon=10\%$ ) and well above the  $A_{e3}$  temperatures (i.e.  $A_{e3} + 166\text{ }^{\circ}\text{C}$ ). Here a new transformation softening model was proposed to overcome this limitation. According to this view, the driving force is given by the difference between the flow stress of the work hardened austenite at the critical strain and the yield stress of the fresh Widmanstätten ferrite that takes its place. The obstacle to the transformation consists of the free energy difference between austenite and Widmanstätten ferrite as well as the shear accommodation and dilatation work associated with the transformation. As proposed here, it is the softening produced by the transformation that provides the energy required to overcome the obstacles to the phase change.

2. The proposed transformation softening model was able to predict the influence on DT of adding alloying elements to steel. The individual effects of Mn and Si addition were analyzed and derived here for the first time. It was revealed that the addition of Mn increases both the driving force for the transformation as well as the free energy obstacle to ferrite formation. This leads to a small net decrease in the driving force. Conversely, Si addition increases the driving force and decreases the obstacle height, thus leading to an *increase* in the net driving force. These changes are likely to influence the volume fractions of DT ferrite produced in these alloys accordingly.
3. The temperature range over which DT can take place was determined here for the first time. Compression tests were performed at very high temperatures, up to 1350 °C (480 °C above the  $A_{e3}$  temperature). It appears that DT ferrite can be formed at any temperature within the austenite phase field. These observations are consistent with the shape of the free energy barrier opposing the transformation for a plain C steel, which rises to a maximum at 1080 °C and decreases to zero at the delta ferrite formation temperature.
4. The available studies in the literature regarding strip rolling simulation only allow for dynamic recrystallization as a softening mechanism. Here both dynamic recrystallization and dynamic transformation were considered during the simulation of strip rolling. The critical strains for DT and DRX were shown to be about 0.05 and 0.12, respectively. The DT critical strains are always highest in the first pass ( $\epsilon = 0.10$ ) and then adopt lower values in the succeeding passes because of the retention of work hardening.
5. The effect of interpass time on dynamic transformation was investigated by carrying out torsion simulations of strip rolling employing interpass intervals in the range 0.5s to 5s. The simulations carried out on the plain C steel revealed that the MFS decreases significantly from the 1<sup>st</sup> pass onwards when short interpass times are used and increases less rapidly than expected when this interval is increased in length. Moreover, the simulations performed on the Nb steel reveal that the MFS decreases substantially

from the 2<sup>nd</sup> to the succeeding passes when short interpass times are used and displays a low rate of increase when this interval is increased in length. It was concluded here that the shorter interpass times both produce more DT ferrite and permit less reversion to austenite than the longer times. There is more retained work hardening during short passes than the longer ones.

6. The addition of Nb impedes the displacive forward transformation and retards the diffusional reverse transformation rates. The former occurs because Nb addition reduces the effectiveness of transformation softening, the latter by means of a solute drag effect. Most of the DT ferrite formed in the plain C steel is able to retransform into austenite after holding for 15s. By contrast, the reverse transformation is severely retarded in the Nb steel.
7. In order to clarify the industrial implications of dynamic transformation, the volume fraction of ferrite retained from pass to pass was measured for both the plain C-Mn and Nb microalloyed steels. The measurements revealed that about 10% and 7% of ferrite (for pass strains of 0.4) were formed in the plain C-Mn and Nb microalloyed steel, respectively.

## **Appendix A**

### **Formation of Widmanstätten ferrite at very high temperatures in the austenite phase field**

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**Formation of Widmanstätten ferrite at very high temperatures in the austenite phase field**

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**Abstract**

Compression tests were carried out on a 0.06wt%C-0.3wt%Mn-0.01wt%Si steel at temperatures high in the austenite phase field. Eight deformation temperatures were selected in the range from 1000 to 1350 °C at 50 °C intervals. The quenched samples were examined using optical microscopy and EBSD techniques. It was observed that dynamic transformation took place and that the volume fraction of transformed ferrite first decreased with temperature (up to 1050 °C) and then increased as the delta ferrite temperature domain was approached. The EBSD results revealed the presence of Widmanstätten ferrite plates under all testing conditions, right up to 1350 °C. As a result of the present observations, a modified Fe-C phase diagram is proposed for the case of dynamic loading.

Keywords: thermomechanical processing, dynamic transformation, Widmanstätten ferrite

## 1. Introduction

The transformation of austenite to ferrite during deformation above the  $A_{e3}$  temperature was first studied by Yada and co-workers in the 1980's [1, 2]. In their pioneering work, Yada et al. showed that DT ferrite could be formed as high as 166 °C above the  $A_{e3}$ . They also demonstrated the occurrence of this unusual phenomenon in real time by performing in-situ X-ray examinations while simultaneously deforming by torsion testing [3]. Follow-up work by Chen and Chen in 2000 [4] showed that DT ferrite retransformed back into austenite on prolonged holding, 30 °C above the  $A_{e3}$ . More recently, researchers in various countries have reported on other aspects of the dynamic transformation of austenite [5-10].

The dislocation density introduced by working was initially taken to be the driving force for dynamic transformation. For example, Sun et al. [11] in 2008, calculated the Gibbs energy increase in the deformed system to be  $22.3 \text{ Jmol}^{-1}$  and concluded that this was responsible for DT. Somewhat similar results were obtained by Hanlon et al. [12], who justified the presence of dynamically transformed ferrite in this way at temperatures up to 10 °C above the  $A_{e3}$ . Ghosh et al. [13] extended the temperature range proposed by Hanlon and co-workers [12] by allowing for the distribution of dislocations to be inhomogeneous in nature. By increasing the estimated local stored energy to  $197 \text{ Jmol}^{-1}$ , they were able to extend the predicted temperature range for the occurrence of this phenomenon up to 100 °C above the  $A_{e3}$ .

One limitation of the inhomogeneous distribution model is that it is unable to explain how strains as low as 10% can lead to the initiation of DT. This difficulty was resolved by introducing the concept of mechanical activation by the applied stress, which can more readily account for the displacive nature of the forward transformation [14- 16]. In the most recent version of this approach, Aranas et al. [17- 19] proposed that the driving force for the transformation is the flow stress difference between the strain hardened austenite and the yield stress of Widmanstätten ferrite that takes its place. As will be shown below, such softening can provoke dynamic transformation at temperatures right up to the upper limits of the austenite phase field.

The objective of the present investigation was, therefore, to determine the characteristics of dynamic transformation at temperatures very high in the austenite phase field (up to 480 °C above the  $A_{e3(p)}$  in the present case). For this purpose, compression tests were performed on a 0.06wt%C-0.3wt%Mn-0.01wt%Si steel using a Gleeble 3800 thermomechanical system. The tests were conducted in the temperature range 1000 to 1350 °C (with intervals of 50°). Strains of 0.3 and 0.65 were employed in the tests and the strain rate was held constant at  $1s^{-1}$ . After water quenching, the specimens were polished and etched for examination by means of optical and electron back scattered microscopy. Double differentiation was also employed to identify the critical strains necessary for the initiation of dynamic transformation. The present observations indicate that the Fe-C phase diagram must be modified so as to apply to the case of dynamic loading.

## 2. Experimental Procedure

### 2.1 Compression testing

The plain C-Mn steel was received in the form of hot rolled plates with thicknesses of 12.5 mm. Cylindrical compression samples with dimensions of 15 mm X 10 mm were machined from these plates with the cylinder axes parallel to the rolling direction. The composition of this alloy is shown in Table 1 together with the respective paraequilibrium and orthoequilibrium  $A_{e3}$  temperatures [13]. These were calculated using the FSStel database of the FactSage [20] and were validated using the Thermo-Calc thermodynamic software [21].

The compression tests were carried out in a vacuum and the Gleeble 3800 (Dynamic Systems Inc. USA) thermomechanical simulator was interfaced with QuikSim software that controlled the heating rate, cooling rate, quenching medium, strain, strain rate and temperature. The thermomechanical schedule employed is displayed in **Fig. 1**. Here the specimens were heated at the rate of  $4.5\text{ }^{\circ}\text{C s}^{-1}$  to test temperatures in the range 1000 °C to 1350 °C and held for 3 minutes in order to fully austenize the microstructure before deformation. These were water quenched immediately after straining. Note that, a K-type thermocouple was used for the experiments carried out over the range 1000 °C to 1250 °C. For the compression tests above 1250 °C, an R-type thermocouple was employed. The data were later processed into stress vs strain curves obtained directly by the above software.

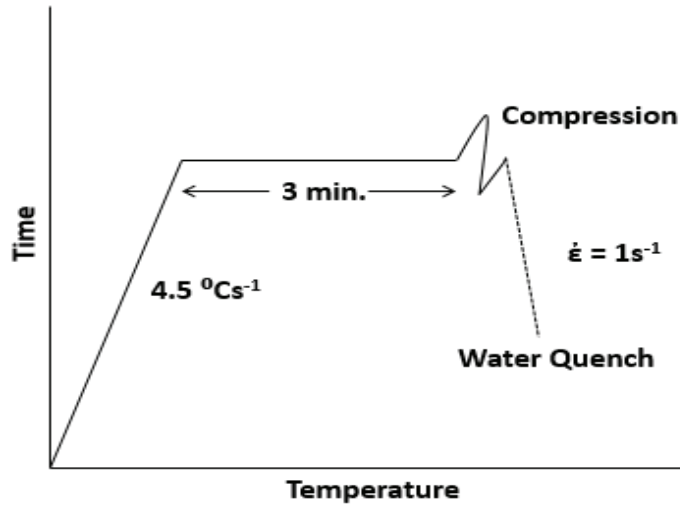


**Fig. 1** Thermomechanical schedule of the compression tests carried out to strains of i) 0.3 at temperatures in the range from 1200 – 1350 °C and ii) 0.65 in the range 1000 – 1350 °C. The strain rate was  $1\text{ s}^{-1}$  in both series of tests.

The first set of compression tests involved the imposition of a strain of 0.3 applied at the rate of  $1\text{ s}^{-1}$  over the temperature range 1200 to 1350 °C. In the second set, a higher strain of 0.65 was applied using the same parameters, but over the wider temperature range 1000 to 1350 °C. Note that these temperatures are high in the austenite phase field and approach the delta ferrite formation temperature of approximately at 1472 °C.

## 2.2 Metallography

For microstructural examination, central cross-sections perpendicular to the longitudinal axis were cut from the compressed specimens. The latter were then mounted in a phenolic hot mounting resin. The mounted specimens were polished using SiC papers from 400 to 1200 grit. Finally, these were subjected to 3 and 1  $\mu\text{m}$  diamond polishing. Thereafter, the specimens were etched with 2% nital for optical microscopy. For determination of the volume fraction of DT ferrite in the austenite, etching was also carried out using a 10% sodium metabisulfite ( $\text{Na}_2\text{S}_2\text{O}_5$ ) solution. MIP microstructure analysis software was then used to calculate the volume fractions.

For the EBSD analysis, the specimens from each temperature were polished in the same manner as was employed for optical microscopy. Later, the specimens were subjected to final polishing using 0.05  $\mu\text{m}$  silica in a vibromet for 6 to 8 hours.

### 3. Results

The stress-strain curves associated with deformation over the temperature range 1000 to 1350  $^{\circ}\text{C}$  are displayed in **Fig. 2**. Here the samples were strained to 0.65 at the rate of  $1\text{ s}^{-1}$ . All the curves display softening after the peak stress is attained. As shown by earlier workers [6, 14] this is due to the concurrent operation of dynamic transformation and dynamic recrystallization. Steady state flow was attained at a strain of about 0.4. An interesting feature of these curves is their somewhat “flattened” shape prior to the peak. This is an indication of the initiation of dynamic transformation, which reduces the net rate of strain hardening once it is underway. The transformation softening process makes it possible to detect its initiation by means of the double differentiation technique, as will be shown in more detail below.

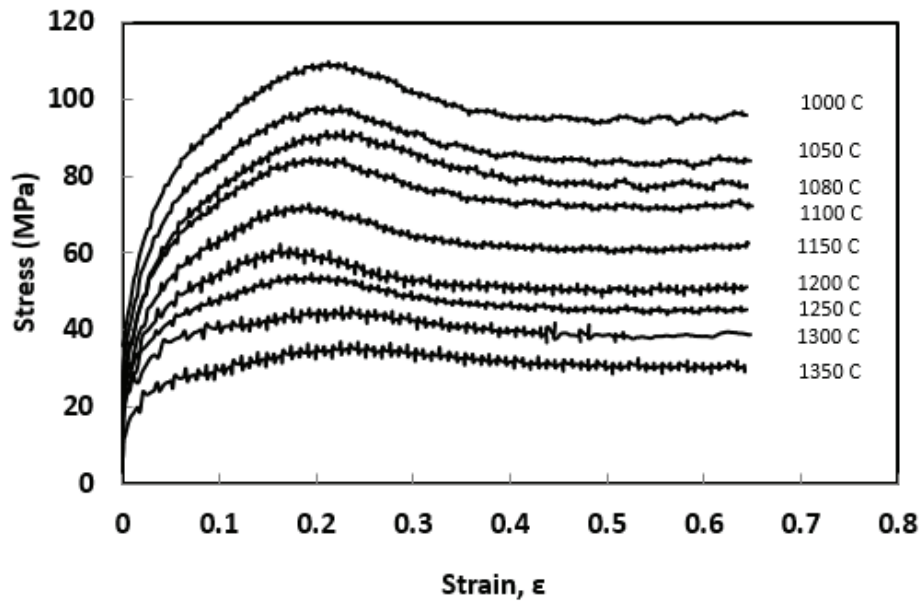


**Fig. 2** Stress vs strain curves of the 0.06wt% C- 0.3wt% Mn- 0.01wt% Si steel compressed to a strain of 0.65 at  $1\text{ s}^{-1}$  over the temperature range 1000 to 1350  $^{\circ}\text{C}$ . Flow softening of about 10% is generally observed after the peak stress is attained.

### 3.1 Mean Flow Stress

The areas under the flow curves of **Fig. 2** were determined by integrating up to a strain of 0.65. The mean flow stresses derived from these areas are illustrated in **Fig. 3**, where they are plotted against inverse absolute temperature. Here, it can be seen that they extrapolate to a value of 4-5 MPa at the melting temperature. Thus the present high temperature measurements validate the frequent assumptions of earlier workers [18, 19] that lower temperature measurements can be extrapolated in a linear manner (against  $1/T$ ) right up to the melting point.

### 3.2 Critical strains for the initiation of dynamic transformation

Double differentiation was applied to the flow curves of **Fig. 2** using the MATLAB software [22]. Most of the curves were fitted using a 9<sup>th</sup> order polynomial, although in the higher temperature cases, a higher-order polynomial was used. This method is illustrated in **Fig. 4**, where the strain hardening rate ( $\Theta$ ) is first defined, **Fig. 4a**. The dependence of  $\Theta$  on  $\sigma$  is plotted in **Fig. 4b**, which shows the two inflection points that correspond to the moments of initiation of dynamic transformation and dynamic recrystallization. The locations of these points were then determined more precisely by noting that they correspond to the following condition:

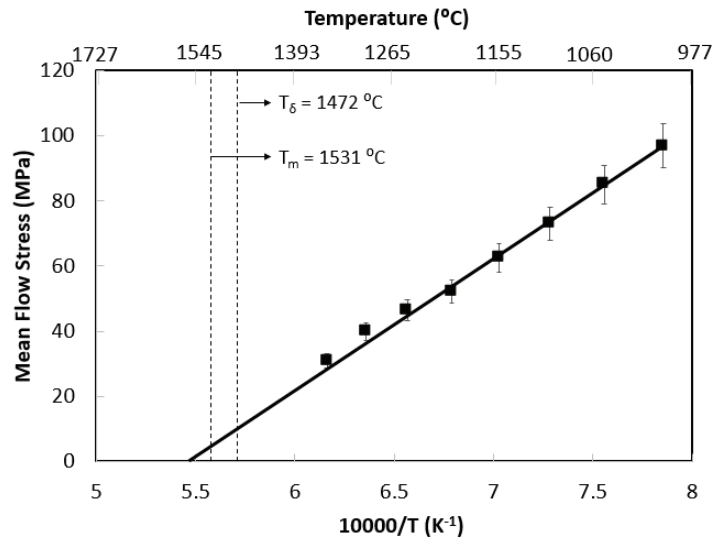
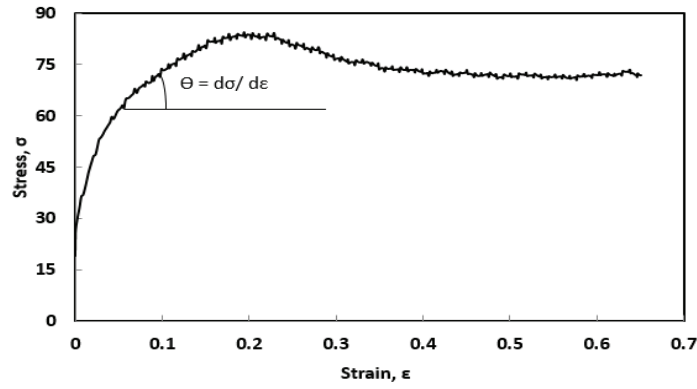
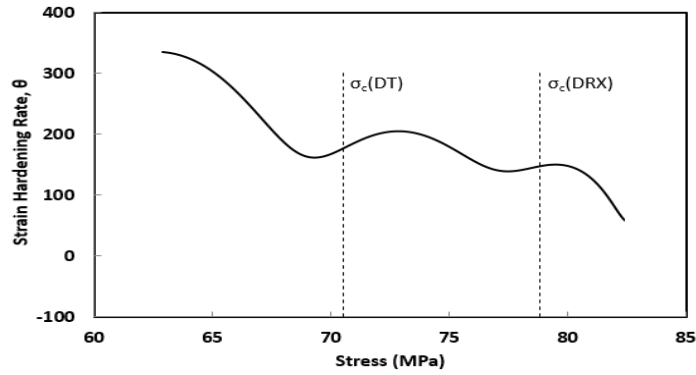


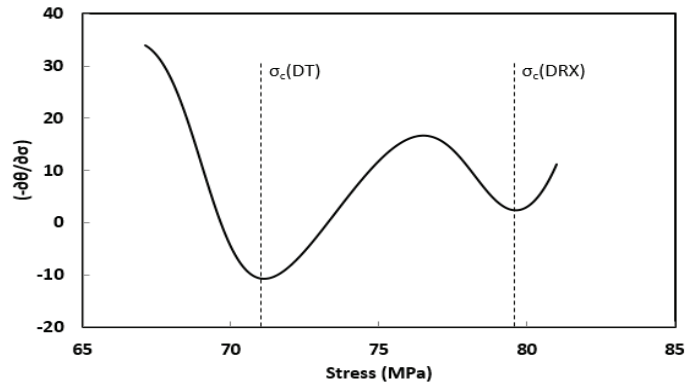
**Fig. 3** Mean flow stresses calculated from the stress-strain curves of **Fig. 2** for the 0.06wt%C-0.3wt%Mn-0.01wt%Si steel.



(a)



(b)



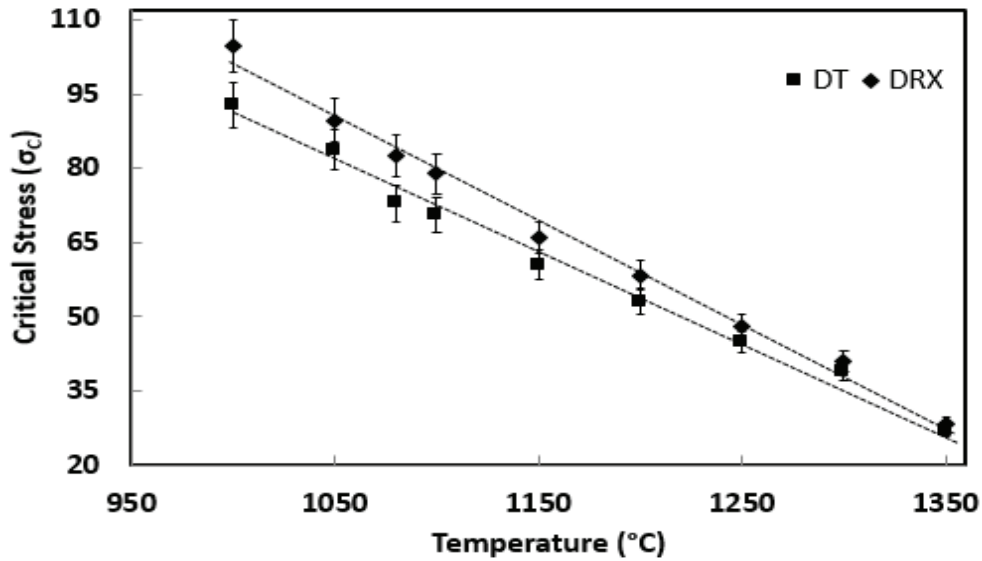
(c)

**Fig. 4** The double differentiation method employed in determination of the critical stresses and strains: a) definition of the strain hardening rate; b) plot of  $\Theta$  vs stress illustrating the presence of two (of the three) inflection points; and c) examples of the dependence of  $(-\partial\Theta/\partial\sigma)$  on stress. This experiment was carried out at 1100 °C.

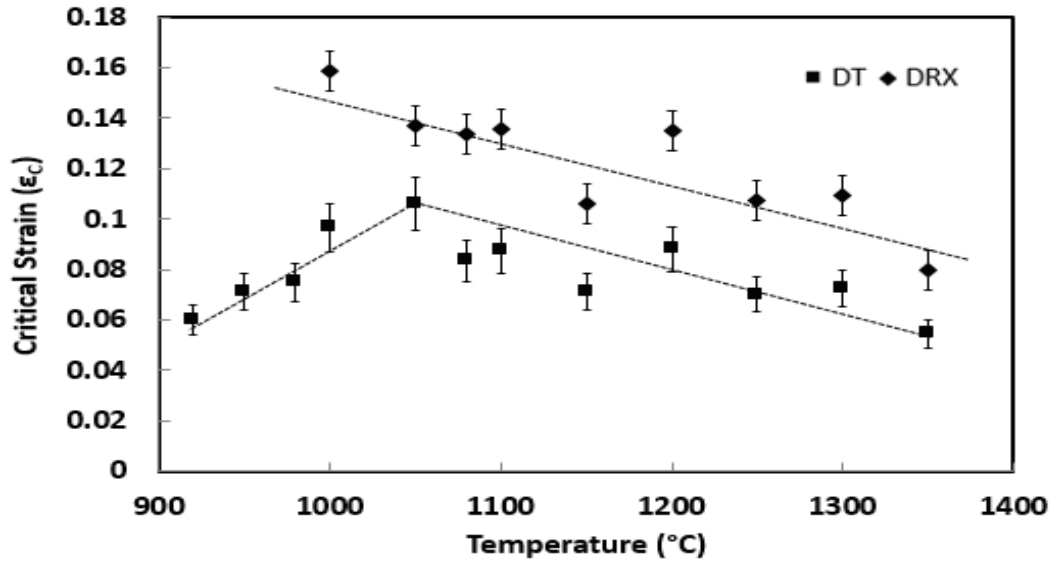
$$\frac{\delta}{\delta\sigma} \left( \frac{\delta\theta}{\delta\sigma} \right) = 0 \quad (1)$$

The two minima seen in the plot of  $(-\partial\theta/\partial\sigma)$  against stress displayed in **Fig. 4c** obey this condition. These minima represent the critical stresses at which dynamic transformation and dynamic recrystallization are initiated during deformation. The critical strains that correspond to these stresses were then evaluated by referring back to the initial flow curves, **Fig. 2** [23].

The critical stresses determined in this way over the temperature range 1000 °C to 1350 °C are shown in **Fig. 5a**. Both of these decrease with temperature [19]. The critical strains that correspond to these stresses are illustrated in **Fig. 5b**. Here the critical strains for DRX decrease with increasing temperatures in the normal way [19]. By contrast, the DT critical strains first increase with temperature to a maximum and then decrease as the delta ferrite phase field is approached. The average values for the DT and DRX critical strains are about 0.08 and 0.15, respectively.



(a)



(b)

**Fig. 5** a) Critical stresses for dynamic transformation (DT) and dynamic recrystallization (DRX) determined over the temperature range 1000 to 1350 °C; b) critical strains over the same temperature range.

### 3.3 Dynamic transformation microstructures

As demonstrated in more detail elsewhere [16, 19], dynamic transformation involves the initial formation of near-identically oriented Widmanstätten ferrite plates followed by their coalescence into polygonal grains. In order to check if this sequence of events was followed at the much higher temperatures employed here, optical microscopy was performed on the samples deformed to the lower strain of 0.3. Examples of the colonies of ferrite plates formed under the present conditions are shown in **Fig. 6**. Here, the darker regions correspond to martensite (prior austenite) and the lighter regions to ferrite. It can be seen that the volume fraction of Widmanstätten ferrite increases with temperature, as will be demonstrated in more detail below. There is also a tendency for the plate width to increase with temperature. It should be noted that the ferrite plates formed are mostly along the grain boundaries of the prior austenite, although plates can also nucleate within the grains [13, 15].

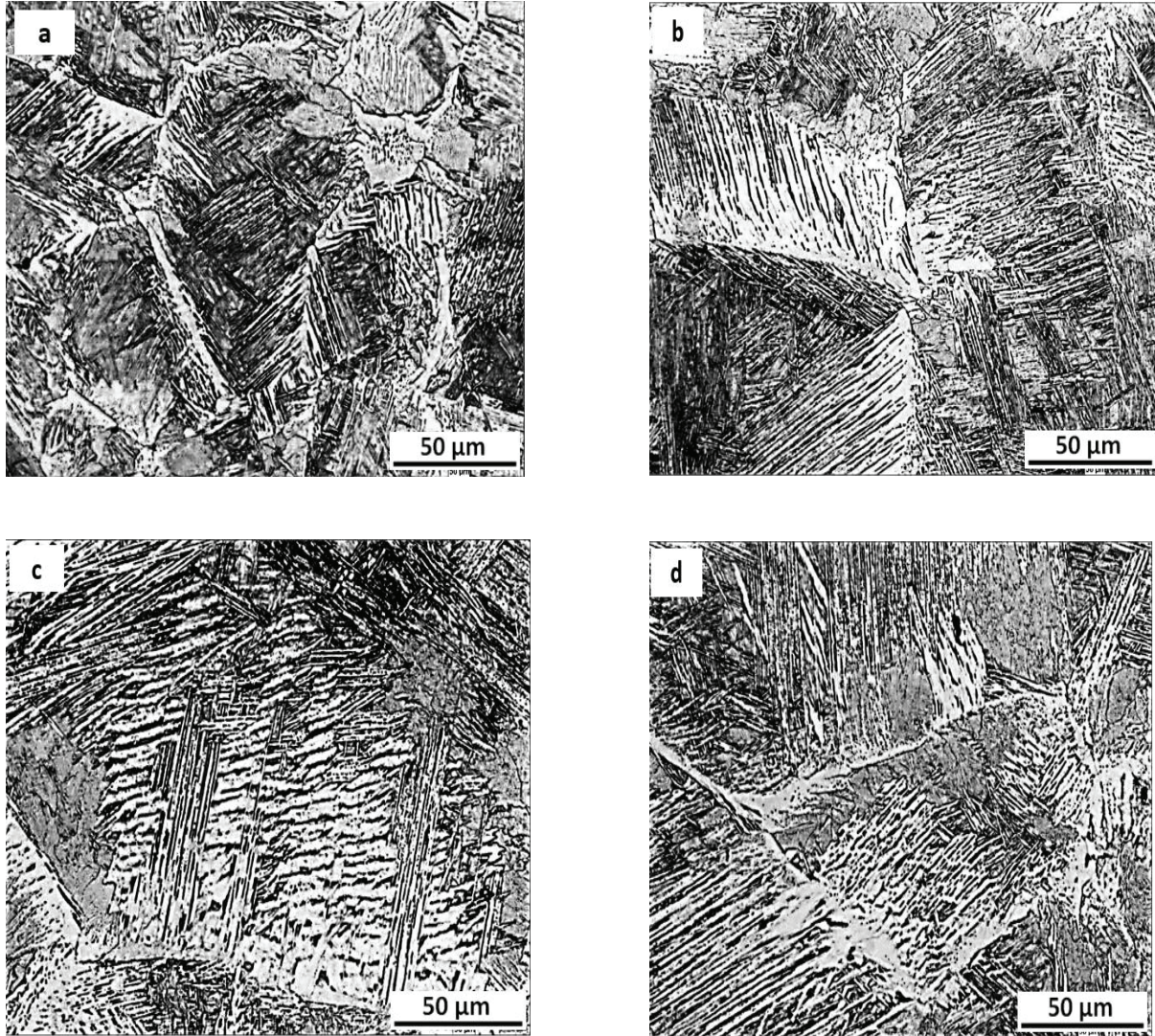


**Fig. 6** Optical micrographs at 500X of the present steel compressed to a strain of 0.3 at  $1\text{ s}^{-1}$ : a) 1200 °C b) 1250 °C c) 1300 °C d) 1350 °C. The dark regions are martensite (prior-austenite) and the lighter ones are ferrite. The formation of plate-like Widmanstätten ferrite is evident even at 1350 °C.

In order to support the above interpretation of the microstructure, Vickers hardnesses were measured in selected regions in the samples strained to 0.3. Examples of the values obtained in the 1200 to 1350 °C samples are presented in **Fig. 7**. These values are consistent with the results obtained in tests carried out below 1000 °C [24]. Here, the average polygonal ferrite microhardness was about 170 HV and that of the martensite was 263 HV. The latter value is

consistent with martensite hardness values in the range 200 – 300 HV expected in low C steels [25]. By contrast, the hardness of the Widmanstätten ferrite plates was somewhat higher than for polygonal ferrite at 203 HV vs. 174 HV, as seen in the sample deformed at 1300 °C (**Fig. 7c**).

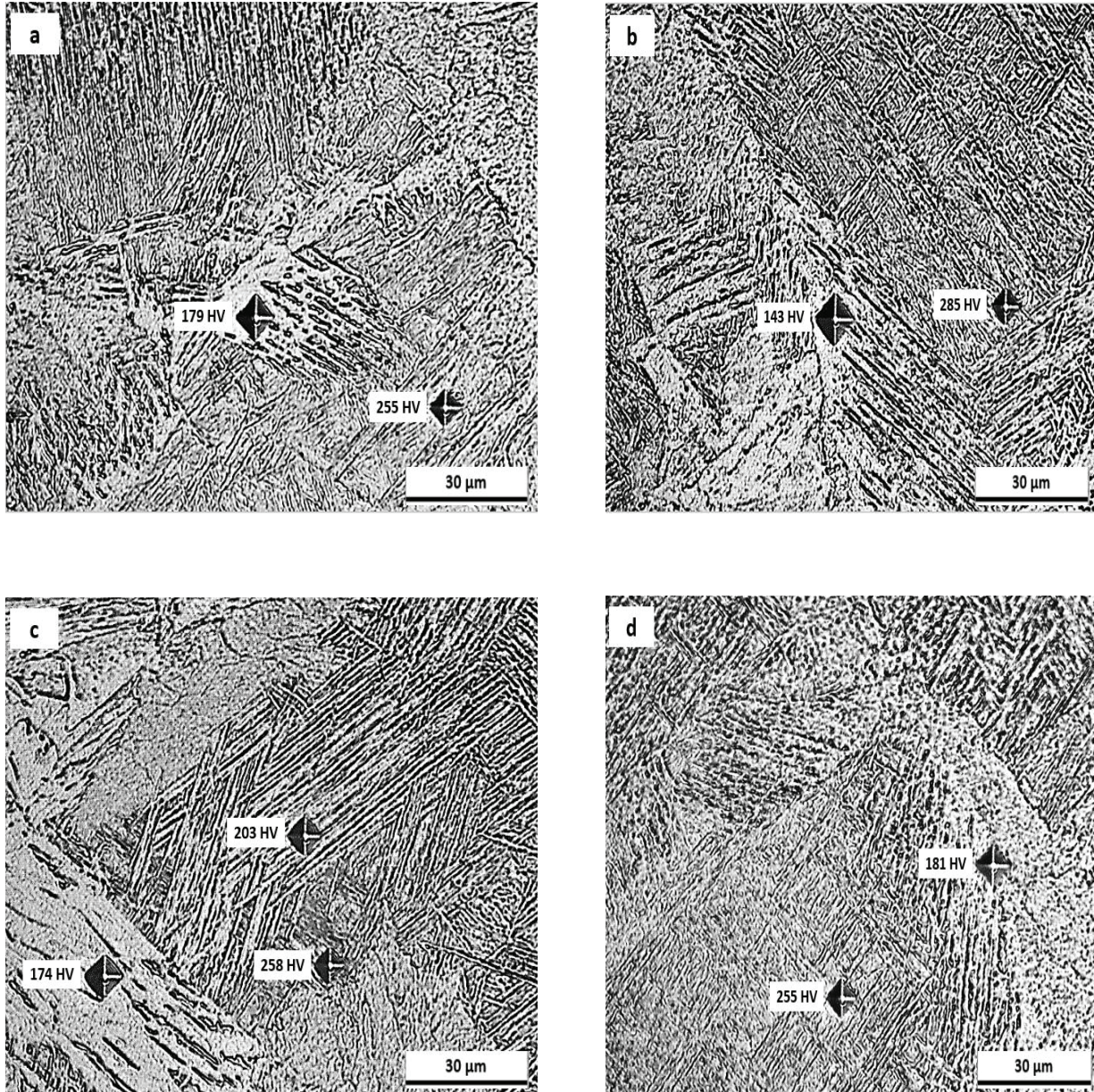


**Fig. 7** Vickers microhardnesses of the polygonal ferrite, Widmanstätten ferrite and martensite (prior-austenite) in samples strained to 0.3 at a) 1200 °C b) 1250 °C c) 1300 °C d) 1350 °C.

The microstructures associated with the samples deformed to a strain of 0.65 are illustrated in **Fig. 8**. Here appreciable amounts of ferrite are present, which are homogeneously distributed throughout each specimen. The volume fraction of DT ferrite was calculated using the MIP software and at least 15 micrographs were scanned for each condition for this purpose. As mentioned earlier, the darker and lighter regions correspond to martensite (prior austenite) and ferrite, respectively.

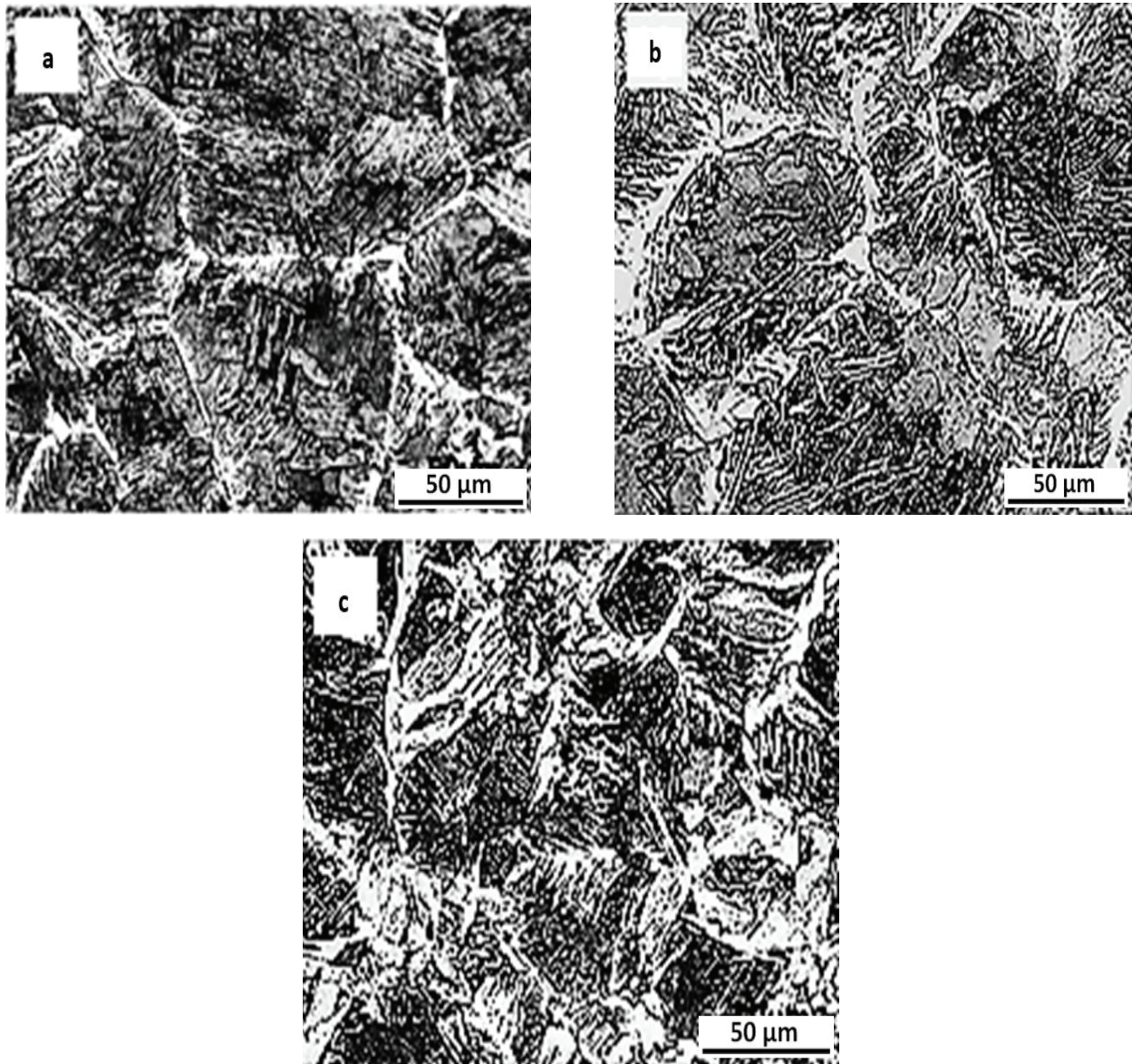


**Fig. 8** Optical micrographs of the present steel compressed to a strain of 0.65 at  $1s^{-1}$  at: a) 1050 °C b) 1150 °C c) 1200 °C. The volume fraction of ferrite increases with temperature. The darker and lighter regions correspond to martensite (prior austenite) and ferrite, respectively.

The result of these measurements are displayed in **Fig. 9**. Here, it can be seen that the volume fraction of ferrite first decreases to a minimum of about 10% at 1050 °C. It then increases with temperature to about 25% at 1200 °C.

### 3.4 High magnification (EBSD) microstructures

The optical micrographs in the previous section indicated that Widmanstätten ferrite plates continued to be formed in the samples deformed at the highest temperatures in the austenite phase field. The orientations of these plates were characterized by means of the electron backscattered diffraction (EBSD) technique. An inverse pole figure map of some of the plates formed at 1200 °C ( $A_{e3(p)} + 330$  °C) is illustrated in **Fig. 10 a**. This sample was strained to  $\epsilon = 0.3$  at  $1 \text{ s}^{-1}$ . Here, the black regions are martensite (prior austenite) and the plates can be seen to be present in colonies of similar orientation [26].

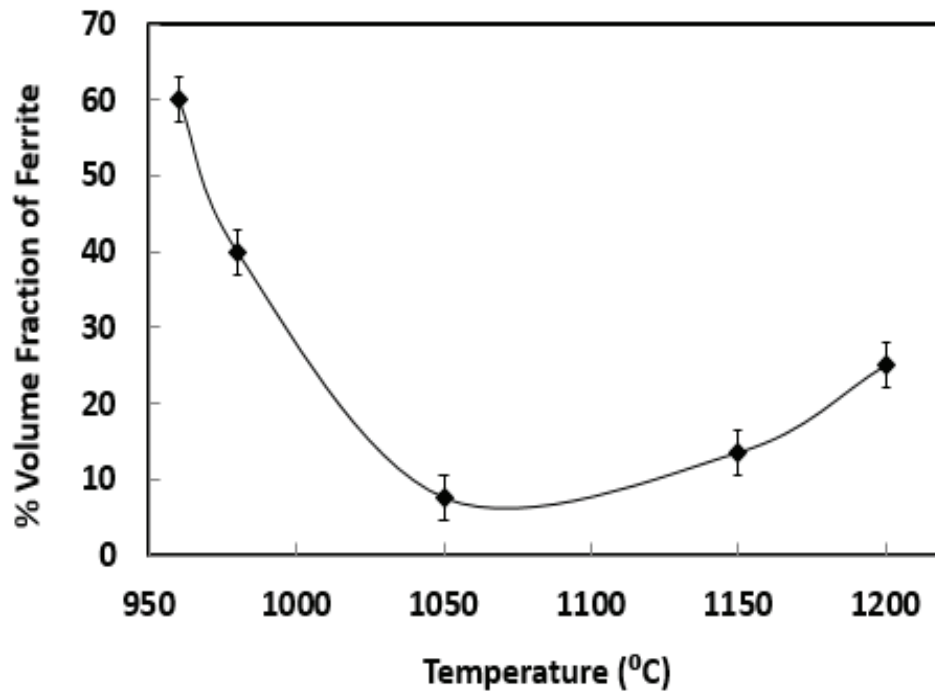
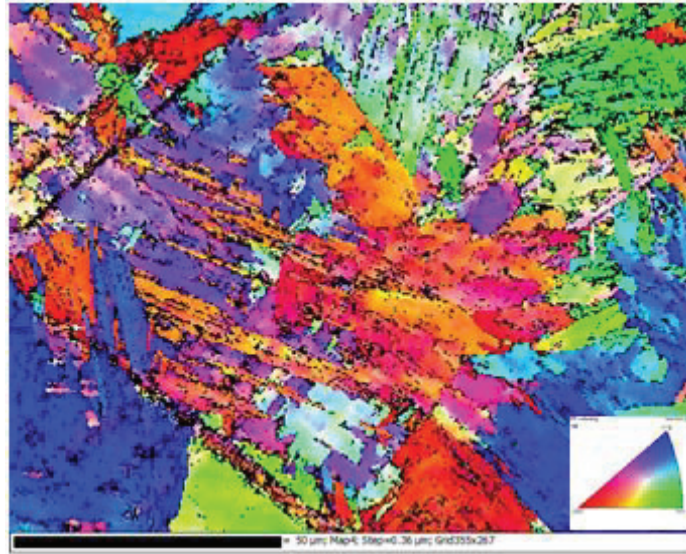
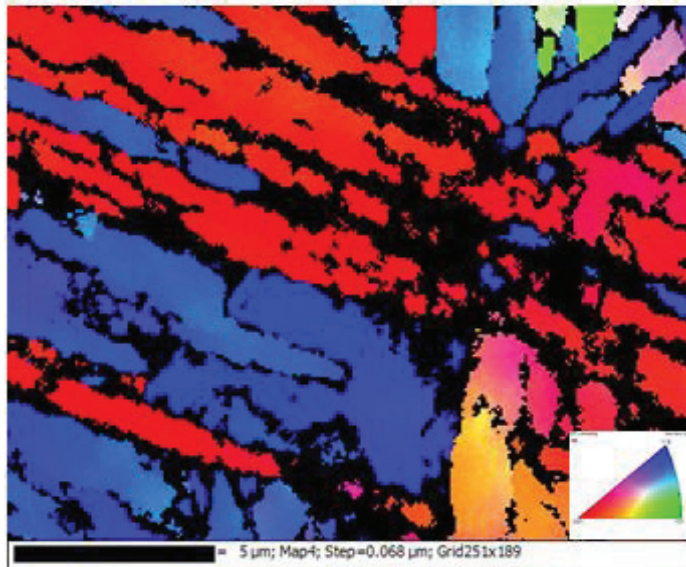


**Fig. 9** Volume fraction of ferrite vs temperature at a strain of 0.65. The volume fraction is lowest (about 7.5 %) midway between the  $A_{e3}$  and the  $\delta$ -ferrite phase field.

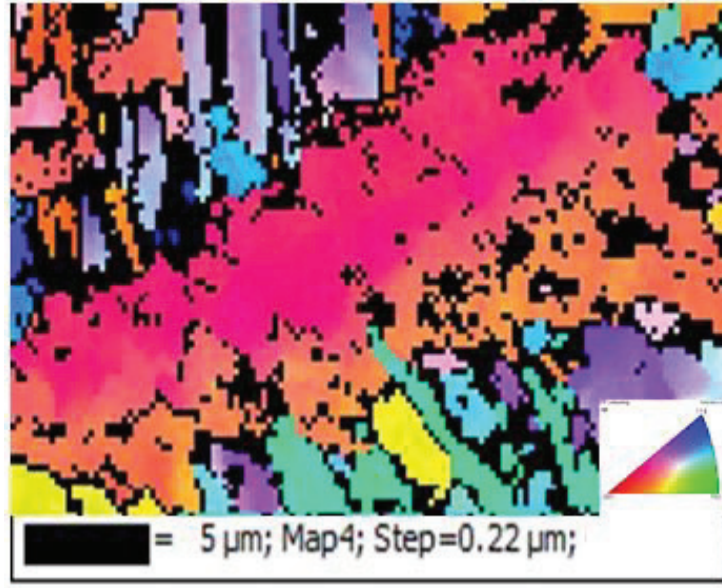
On increasing the magnification to 7500X (**Fig. 10b**), the plates can be seen more clearly. Their widths fall in the range 1 – 2  $\mu\text{m}$ ; these are nearly an order of magnitude thicker than the 200 nm plates formed at 900 – 1000  $^{\circ}\text{C}$  [26, 27]. The coalescence of the plates into polygonal ferrite is illustrated in **Fig. 10c** for  $\epsilon = 0.65$  at  $1 \text{ s}^{-1}$ . It is likely that the coalesced plates of **Fig. 10c** were formed early in this experiment, for instance at a strain of 0.10 to 0.15, while the ‘fresh’ plates of **Fig. 10b** were formed closer to the final experimental strain of 0.3, e.g. at  $\epsilon = 0.25$  to 0.30.



(a)



(b)



(c)

**Fig. 10** Self-accommodating plates in sample strained to  $\epsilon = 0.3$  at 1200 °C: a) 1000X, b) 7500X, c) Coalescence of plates in sample strained to  $\epsilon = 0.65$  at 1350 °C and  $1s^{-1}$ .

The widening of the plates is probably associated with the increased diffusivity of C in ferrite over this temperature range. This hypothesis can be evaluated by calculating the carbon diffusion distance under the present conditions. This is specified by the following relationship:

$$\bar{X} = \sqrt{Dt} \quad (2)$$

where  $\bar{X}$  is the mean diffusion distance,  $D$  the carbon diffusion coefficient in ferrite and  $t$  the time. The diffusion coefficient is in turn given by:

$$D = D_0 \exp (-Q/RT) \quad (3)$$

where  $D_0 = 7.9 \times 10^{-3} \text{ cm}^2\text{s}^{-1}$  (for carbon diffusion in ferrite) and  $Q = 75.6 \text{ KJmol}^{-1}$  [27].

According to this relation, the diffusivity increases by a factor of 4.8 when the temperature is increased from 900 to 1200 °C. In a similar manner, the diffusivity increases by a factor of 8.6 over the range from 900 to 1350 °C. The plates formed at 900 °C are about 200 nm or so thick [16]. Thus, the above diffusivity ratios suggest that, for equal diffusion times, the plate

thicknesses at 1200 and 1350 °C should increase to 440 and 600 nm, respectively, at these two temperatures.

Thus the increase in the rate of carbon partitioning cannot account for the full increase in plate thickness if the plate growth time is held constant. A further factor that may be playing a role is the increasing recovery rate associated with the increase in temperature. The plates form in sequences of self-accommodating pairs, the second of which is induced so as to relieve the internal stresses developed during accommodation of the first member of the pair [26]. As the rate of development of the internal stress depends on the recovery rate, larger strains and therefore times (and amounts of plate growth) are probably associated with the higher temperature tests.

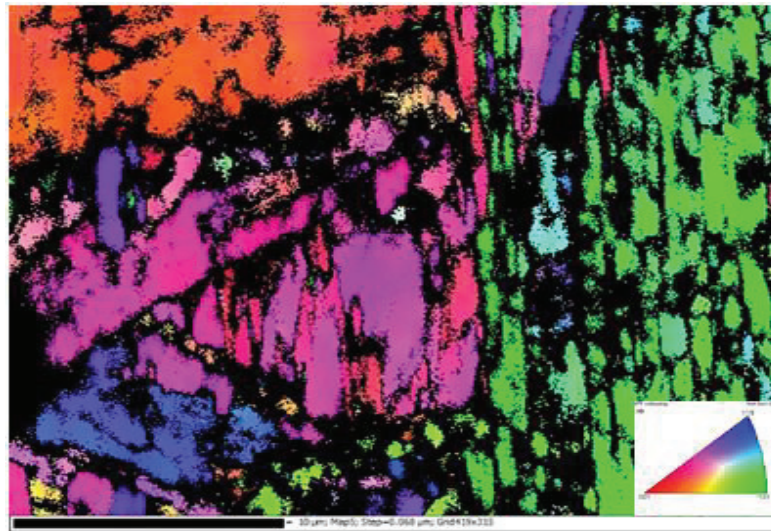
### **3.5 Perpendicularity of the DT ferrite plates**

When Widmanstätten ferrite plates form in the absence of an applied stress, their orientations are essentially random, that is within the limits set by the Kurdjumov-Sachs and other correspondence relations. In this way, all 24 K-S variants have an equal probability of appearing. By contrast, under dynamic transformation conditions, two habit planes/shear direction combinations are favored in each grain; those that lie closest to the plane and directions of maximum resolved shear stress [26]. Some examples of the presence of such perpendicular plates are provided in **Fig. 11**. Here the loading direction is vertical in the micrographs. These plates can be expected to have been aligned close to the maximum shear stress planes (inclined at 45° with respect to the direction of applied stress) when they were formed. However, during compression, these plates gradually rotate away from their initial orientations by an amount that increases with the amount of strain undergone beyond the critical strain applicable to the grain in question. It can also be observed that regions with near identical orientations have coalesced into polygonal grains on further straining.

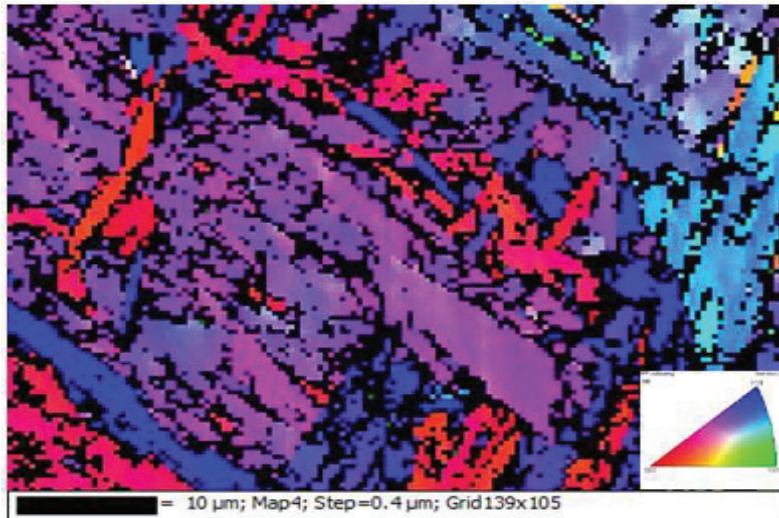
### **3.6 Plate disintegration**

During these very high temperature experiments, the plates were subjected to more “break-up” than that takes place at the lower temperatures studied to date [16-19, 28, 29]. Some

examples are provided in **Fig. 12** illustrating the microstructures of samples strained to  $\epsilon = 0.3$  at 1250 and 1350 °C. For reference, an EBSD micrograph of a sample deformed to  $\epsilon = 0.3$  at the somewhat lower temperature of 1200 °C is also presented here. The rate of plate disintegration can be seen to increase with temperature. Similar trends were observed in both the low strain (0.3) and high strain (0.65) samples.

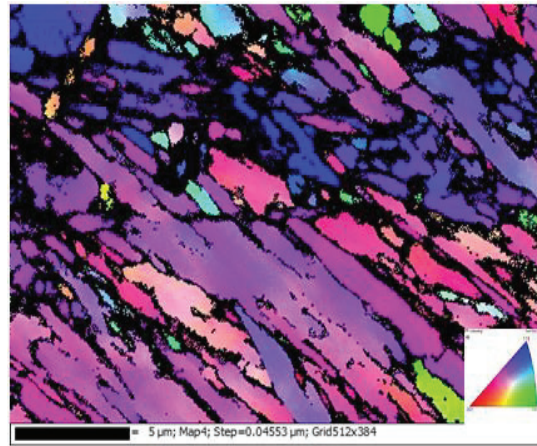


(a)

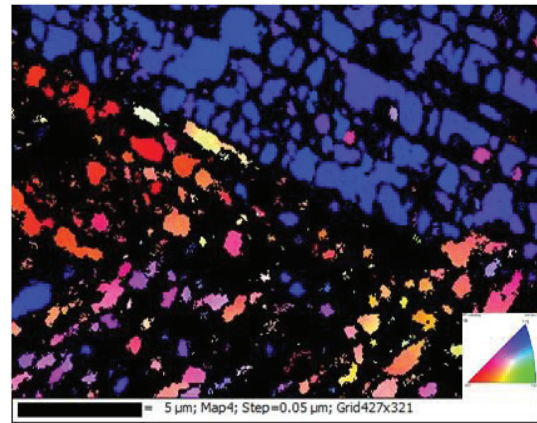


(b)

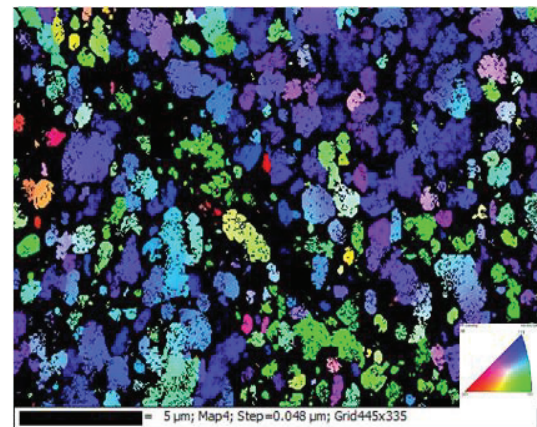
**Fig. 11** Widmanstätten ferrite plates in samples strained to  $\epsilon = 0.65$  at: a) 1250 °C and b) 1350 °C. Note the presence of approximately perpendicular plates.



(a)



(b)



(c)

**Fig. 12** Increasing disintegration of the ferrite plates as the temperature is increased: a) 1200 °C b) 1250 °C c) 1350 °C.

## 4. Discussion

### 4.1 The Gibbs energy barrier to dynamic transformation

The observation that the critical strain for DT displays a maximum at about 1050 °C in **Fig. 5 (b)** was initially surprising. So was the later finding that the volume fraction of ferrite is at a minimum at this temperature, **Fig. 9**. These trends can be given a physical interpretation by examining the free energy barrier to the transformation illustrated in **Fig. 13**. This depicts the temperature dependence of the Gibbs energy difference between the stable austenite and the metastable ferrite as a function of  $\Delta T$  (experimental temperature – paraequilibrium temperature). This was calculated using the FSTel database of the FactSage thermodynamic software [20].

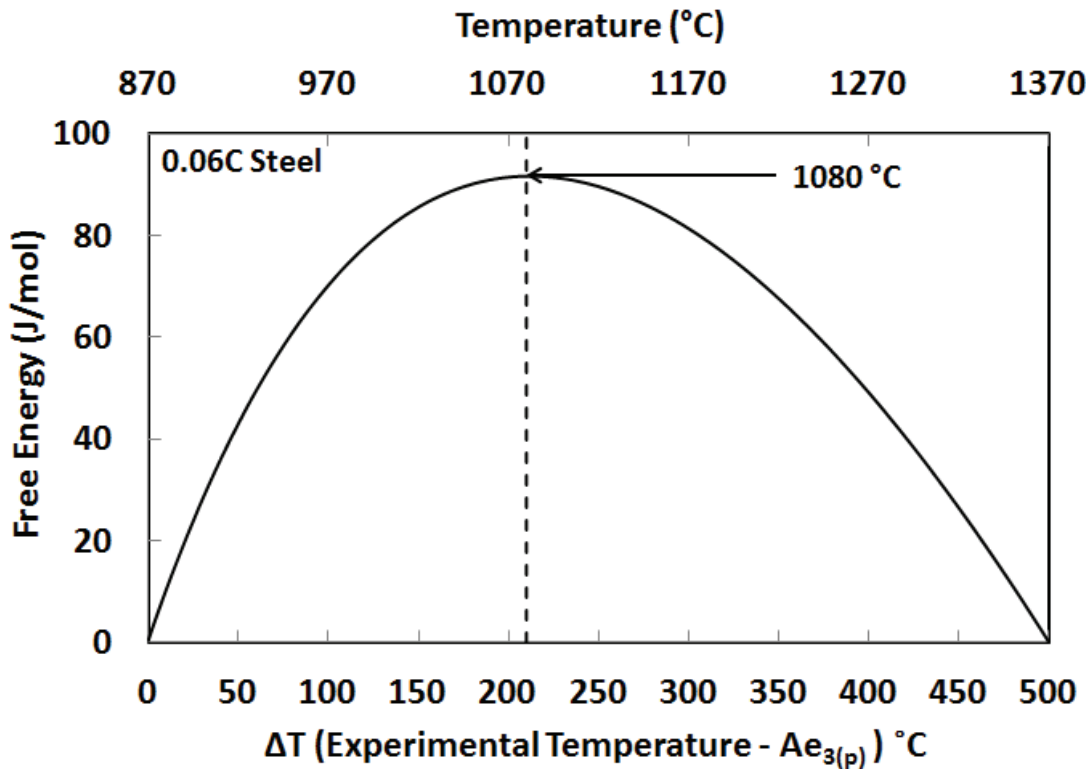
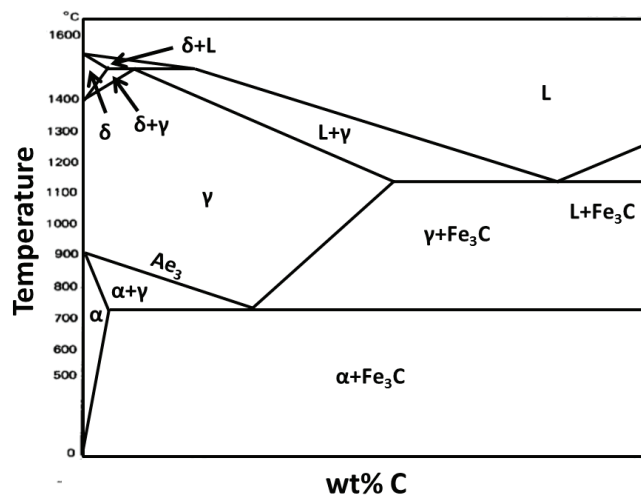


**Fig. 13**  $\Delta G_{(\gamma-\alpha)}$  vs  $\Delta T$  ( $A_{e3(p)}$  – experimental temperature) for the present steel showing the Gibbs energy obstacle opposing dynamic transformation.

As can be seen, this obstacle to the transformation is at a maximum at around 1080 °C, that is, approximately midway between the  $A_{e3}$  and the  $\delta$  – ferrite phase field. This indicates that when the temperature is increased beyond the obstacle peak it becomes easier and easier to form Widmanstätten ferrite dynamically. The decrease in obstacle height in turn means that less work hardening is required in the austenite to provide the amount of transformation softening necessary to overcome the obstacle. The increasing ease of ferrite formation as the delta ferrite phase field is approached is also consistent with the ferrite volume fraction data displayed in Fig. 9.

#### 4.2 A possible dynamic Fe-C phase diagram

It was shown in the previous sections that the dynamic transformation can be initiated up to 480 °C above the paraequilibrium  $A_{e3}$  temperature. Thus the equilibrium phase diagram for undeformed material does not apply when steel is being deformed. However, the dynamic diagram depends on the amount of strain being applied and so does not have a unique configuration. An example of a possible diagram is presented in Fig. 14 for an arbitrary strain. Here the  $A_{e3}$  temperature is increased by 200 to 500 °C, depending on the carbon content. The area associated with the single austenite phase field also shrinks significantly, resulting in an increase in the area of the two-phase region ( $\alpha+\gamma$ ) region. Moreover, the single phase ferrite region extends right up to the melting temperature.



(a)

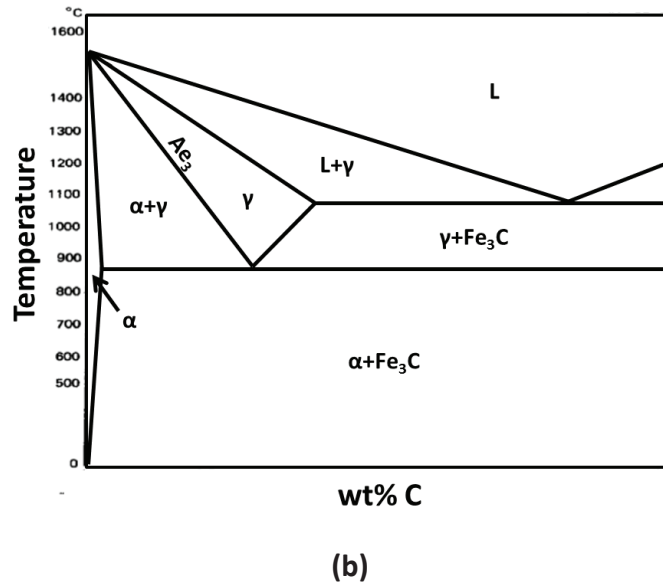


**Fig. 14** a) Static Fe-C binary phase diagram; b) schematic dynamic phase diagram. Note that: i) the single phase  $\alpha$  region extends right up to the melting point; ii) the two-phase ( $\alpha+\gamma$ ) region is extended to much higher temperatures and; iii) the extent of the single phase  $\gamma$  region is reduced considerably.

## 5. Conclusions

1. The critical strains for the initiation of dynamic transformation and dynamic recrystallization were calculated using the double differentiation method. The DRX strains decreased with increasing temperatures in the conventional manner. By contrast, the critical strains for DT increased initially with temperature but then decreased as the temperatures approached the  $\delta$ -ferrite phase field. This is consistent with the shape of the Gibbs energy obstacle to the transformation, which is at a maximum approximately midway between the  $A_{e3}$  and the  $\delta$ -ferrite phase field.

2. Optical micrographs confirmed the presence of dynamically formed ferrite plates at temperatures as high as 1350 °C ( $A_{e3(p)} + 480^\circ$ ). The ferrite volume fraction increased with the magnitude of the temperature interval above the Gibbs energy peak. This observation is also consistent with the tendencies displayed by the free energy obstacle and the DT critical strains.

3. The Widmanstätten ferrite plates produced at these very high temperatures have thicknesses of about 1 to 2  $\mu\text{m}$  and are therefore about an order of magnitude wider than the plates formed at 900 °C. This can be attributed to the higher diffusivity of carbon at these temperatures, as well as to the lower rate of work hardening. The latter is probably responsible for the somewhat slower rate of plate propagation at these very high temperatures.
4. The ferrite plates that form dynamically are frequently aligned along two approximately perpendicular directions. These correspond to the traces of planes close to the planes of maximum shear stress at the moment of plate formation. There is considerably more plate disintegration and break-up than observed in the microstructures produced at lower temperatures.
5. The present observations indicate that the well-known 'static' Fe-C phase diagram does not apply when austenite is being deformed. However, the appropriate 'dynamic' diagram depends on the applied strain. Beyond the critical strain for the formation of Widmanstätten ferrite, the single phase  $\alpha$  region is expanded considerably, the single phase  $\gamma$  region is conversely reduced in size, and the  $A_{e3}$  temperature is raised by up to 500 °C.

### **Acknowledgements**

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## **Appendix B**

# **Abstracts of Co-Authored Papers**

\* submitted to Progress in Materials Science (2015).

### **Dynamic Transformation of Deformed Austenite at Temperatures above the $A_{e3}$**

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#### **Abstract**

The dynamic transformation of austenite to ferrite at temperatures above the  $A_{e3}$  was first investigated by Yada and co-workers in the late 1980's. This phenomenon has now been observed right up to the delta ferrite formation temperature. The basic principles of dynamic transformation are reviewed, together with recent advances in the understanding of this mechanism. The transformation product that forms has been identified as Widmanstätten ferrite, which nucleates displacively and whose growth is accompanied by carbon diffusion. It forms in pairs of self-accommodating plates of near-identical orientation. The similar orientations of the plates allow them to coalesce on continued straining or holding at temperature, leading to the presence of polygonal grains as well as plates. The critical strain for the initiation of dynamic transformation is usually detected using double differentiation; results obtained by researchers on different steels are reviewed, together with observations obtained during the torsion simulation of strip rolling. The thermodynamics of dynamic transformation are examined in detail, with attention paid to three recent approaches, the: i) stored energy, ii) stress activation, and iii) phase softening models. It is concluded that the phase softening model provides the most accurate description of the transformation of the models proposed recently.

\* submitted to Acta Materialia (2015).

**Formation of Widmanstätten ferrite at very high temperatures in the austenite phase field**

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**Abstract**

Compression tests were carried out on a 0.06wt%C-0.3wt%Mn-0.01wt%Si steel at temperatures high in the austenite phase field. Eight deformation temperatures were selected in the range from 1000 to 1350 °C at 50 °C intervals. The quenched samples were examined using optical microscopy and EBSD techniques. It was observed that dynamic transformation took place and that the volume fraction of transformed ferrite first decreased with temperature (up to 1050 °C) and then increased as the delta ferrite temperature domain was approached. The EBSD results revealed the presence of Widmanstätten ferrite plates under all testing conditions, right up to 1350 °C. As a result of the present observations, a modified Fe-C phase diagram is proposed for the case of dynamic loading.

\* This article has been published: J.J. Jonas and **C. Aranas Jr.**, "Problems with use of the logarithmic equivalent strain in high pressure torsion," IOP Conf. Series: Materials Science and Engineering 63 (2014) 012048.

### **Problems Associated with Use of the Logarithmic Equivalent Strain in High Pressure Torsion**

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#### **Abstract**

The logarithmic “equivalent” strain is frequently recommended for description of the experimental flow curves determined in high pressure torsion (HPT) tests. Some experimental results determined at -196 and 190 °C on a 2024 aluminum alloy are plotted using both the von Mises and logarithmic equivalent strains. Three types of problems associated with use of the latter are described. The first involves the lack of work conjugacy between the logarithmic and shear stress/shear strain curves, a topic that has been discussed earlier. The second concerns the problems associated with testing at constant logarithmic strain rate, a feature of particular importance when the material is rate sensitive. The third type of problem involves the “history dependence” of this measure in that the incremental logarithmic strain depends on whether the prior strain accumulated in the sample is known or not. This is a difficulty that does not affect use of the von Mises equivalent strain. For these reasons, it is concluded that the qualifier “equivalent” should not be used when the logarithmic strain is employed to describe HPT results.

\* This article has been published: J.J. Jonas, **C. Aranas Jr.**, V.V. Basabe and C. Ghosh, "Dynamic Transformation during the Torsion Simulation of Strip Rolling," Materials Science Forum, 783-786 (2014), 39-44.

### **Dynamic Transformation during the Torsion Simulation of Strip Rolling**

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#### **Abstract**

Seven-pass strip rolling simulations were carried out on a 0.06%C and a 0.09%C-0.036%Nb steel. The rolling loads (mean flow stresses or MFS's) did not increase as the temperature decreased during the simulation. This is ascribed to the occurrence of dynamic transformation. The simulation results are compared to the high temperature flow curves determined on eight plain C and Nb-modified steels in both compression and torsion and at a series of temperatures and strain rates. When the associated MFS's are plotted against inverse absolute temperature in the form of Boratto diagrams, the stress drop temperatures, normally defined as the upper critical temperature applicable to rolling,  $A_{r3^*}$ , are shown to be about 40 degrees above the paraequilibrium and about 20-30 degrees above the orthoequilibrium  $A_{e3}$ 's. These drops are ascribed to the dynamic transformation of austenite to ferrite, a softer phase. The characteristics of the ferrite produced dynamically are described and the transformation is shown to be displacive in nature, leading to the appearance of fine Widmanstätten plates. These plates coalesce into polygonal grains on further deformation and on holding.

\* This article has been published: J.J. Jonas, **C. Aranas Jr.** and C. Ghosh, "Dynamic Transformation of Austenite at Temperatures Above the  $A_{e3}$ ," Proceedings of the International Conference on Solid-Solid Phase Transformations in Inorganic Materials (PTM 2015), Whistler, BC, (2015) 573-580.

### **Dynamic Transformation of Austenite at Temperatures above the $A_{e3}$**

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#### **Abstract**

Experiments are described, carried out on eight steels of increasing carbon contents, in which specimens were deformed in the austenite phase field. Widmanstätten ferrite was observed to form displacively as self-accommodating plates about 200 nm in width. During deformation, the ferrite plates coalesced into polygonal grains several microns in diameter. On holding after deformation, the ferrite slowly retransformed into austenite by means of diffusional processes. It is shown that the driving force for the forward transformation is the softening associated with the replacement of work-hardened austenite by softer, strain-free ferrite. Critical strains of about 0.06-0.14 are observed, which at first increase with temperature interval above the  $A_{e3}$  and then decrease as the delta phase domain is approached. The driving force developed is used to overcome: i) the free energy difference between the phases, as well as ii) the work of dilatation and iii) of shear accommodation.

\* Article in press: J.J. Jonas, **C. Aranas Jr.** and S.F. Rodrigues, "Dynamic Transformation During the Plate and Strip Rolling," Proceedings of the 4<sup>th</sup> International Conference on Thermo-Mechanical Simulation and Processing of Steels (SimPro'16), Ranchi, India, (2015).

### **Dynamic Transformation during the Torsion Simulation of Plate and Strip Rolling**

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#### **Abstract**

Torsion simulations of plate and strip rolling were carried out on a plain C-Mn and an X70 Nb steel over the temperature range 900°C to 1000°C. Pass strains of 0.4 were applied at a strain rate of  $1\text{s}^{-1}$ . Interpass times of 0.5s, 1s, 1.5s, 3s, 10s and 20s were employed to determine the mean flow stresses (MFS's) applicable to plate and strip rolling. By means of double differentiation, critical strains of 0.05 and 0.12 were established for the initiation of dynamic transformation and dynamic recrystallization, respectively. It is shown that the dynamic transformation of austenite to ferrite takes place above the  $A_{e3}$  temperature, resulting in significantly lower than expected MFS values. The nucleation and growth of ferrite reduces the rolling load and modifies the microstructure. Shorter interpass times produce larger decreases in MFS than longer ones because they do not permit as much retransformation to austenite as takes place during longer interpass intervals.

\* This article has been published: J.J. Jonas, V.V. Basabe, C. Ghosh and **C. Aranas Jr.**, “Dynamic Transformation during the Simulation of Strip Rolling” Proceeding of the 5th Baosteel Biennial Academic Conference, Shanghai, China, (2013) C1-C6.

### **Dynamic Transformation during the Simulation of Strip Rolling**

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#### **Abstract**

Simulations of hot strip rolling were carried out on the McGill torsion machine at strain rates of 2 and 4 s<sup>-1</sup> and cooling rates of 5-10 °C/s. A 0.09%C-0.036%Nb microalloyed and a 0.79C-0.65-0.24Si plain C steel were subjected to pass strains of 0.25 to 0.45 and interpass times of 1 second. By means of double differentiation of the flow curves, it is shown that the dynamic transformation (DT) of austenite to ferrite takes place during all the passes applied above the respective A<sub>e3</sub> equilibrium transformation temperatures (835 °C and 877 °C for the two steels). This type of transformation is initiated at strains of about 0.05, well before the critical strain for the initiation of dynamic recrystallization, and occurs at temperatures as much as 200 °C above the A<sub>e3</sub>. The transformation to ferrite reduces the rolling load below the levels expected for single phase austenite and can lead to shape control problems if not taken into account. The DT ferrite formed in this way is Widmanstätten in nature and is converted to polygonal ferrite during further straining.

\* J.J. Jonas, C. Aranas Jr., S.F. Rodrigues and I.-H. Jung, "Dynamic Transformation during Plate and Strip Rolling," Proceedings of the International Conference on Processing and Manufacturing of Advanced Materials (THERMEC 2016), Graz, Austria, (2016) in press.

### **Dynamic Transformation during Plate and Strip Rolling**

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### **Abstract**

Torsion simulations were carried out of both plate (long interpass times) and strip (short interpass times) rolling. Both isothermal and continuous cooling conditions were employed. The dynamic transformation of austenite to ferrite was observed under all conditions and at all temperatures within the austenite phase field. About 8 to 10 volume percent ferrite was formed in a given pass, leading to about 50 - 70 % ferrite at the end of selected simulations. During the interpass intervals, some retransformation to austenite took place, the amount of which increased with holding time and temperature and decreased with the addition of alloying elements. It is shown that the driving force for the transformation is the softening associated with the replacement of work-hardened austenite grains by the softer alpha phase. The implications with respect to rolling load (i.e. mean flow stress) are also discussed.

\* Article in press: J. Jonas and C. Aranas Jr., "Dynamic transformation of austenite to ferrite during the torsion simulation of strip rolling," Proceeding of the 6th Baosteel Biennial Academic Conference and 10th CSM Steel Congress, Shanghai, China, (2015).

## **Dynamic Transformation of Austenite to Ferrite during the Torsion Simulation of Strip Rolling**

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### **Abstract**

Torsion simulations of strip rolling were carried out on a 0.06%C-0.3%Mn-0.01%Si steel over the temperature range 930°C to 1000°C. Pass strains of 0.4 were applied at a strain rate of  $1\text{s}^{-1}$ . Interpass times of 0.5s, 1s, 1.5s and 3s were employed to determine the mean flow stresses (MFS's) applicable to strip rolling. By means of double differentiation, critical strains of 0.04 and 0.11 were established for the initiation of dynamic transformation and dynamic recrystallization, respectively. It is shown that the dynamic transformation of austenite to ferrite takes place above the  $A_{e3}$  temperature, resulting in significantly lower than expected MFS values. The nucleation of ferrite reduces the rolling load and modifies the microstructure. Shorter interpass times produce larger decreases in MFS than longer ones because of reduced recovery during short interpass intervals.