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Coincidence of Strain-Induced TRIP and Propagative PLC Bands in Medium Mn Steels

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Abstract

An Fe-0.2C-5Mn-2.5Al medium manganese steel is studied for three different intercritical annealing temperatures T_{IA} to observe the effects of variations of retained austenite stability on the mechanical work hardening in uniaxial tension. Digital image correlation (DIC) showed that both Lüders bands and type A Portevin-le-Châtelier (PLC) bands occurred in samples with a lower T_{IA} , but both disappeared when T_{IA} was increased. *In-situ* measurements of sample magnetization with a correction for stress effects on the magnetic properties of the sample allowed for continuous measurement of the retained austenite volume fraction. This method provided a complete characterization of the kinetics of transformation-induced plasticity (TRIP) in the sample over the course of tensile testing. It is shown that the martensite transformation was purely strain-induced and coincided with the passage of both Lüders and PLC bands. An Olson-Cohen model was applied to magnetic data to attempt to explain the differences in TRIP kinetics with varying T_{IA} .

Keywords: Digital image correlation, Magnetization measurement, Medium Mn steel, Portevin-le-Châtelier bands, TRIP

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1. Introduction

Increasingly stringent vehicle emissions standards and a high consumer demand for fuel efficient vehicles have led to the development of a third generation of advanced high strength steels (AHSS) with complex multi-phase microstructures. One such alloy family is Medium Manganese steels. Medium Mn steels are composed of an ultra-fine-grained (UFG) mixture of ferrite, retained austenite, and in some cases martensite. They can achieve high ductility and high work hardening due to transformation-induced plasticity (TRIP) or twinning-induced plasticity (TWIP) in the austenite phase, which locally harden the steel and delay mechanical instability [1–3]. Previous studies on Medium Mn steels have shown total elongation values of up to 70% and ultimate tensile strengths (UTS) ranging from 800 to 1500MPa [4–8]. These steels are intended for high formability in that an initially ductile steel can be hardened significantly during the forming process, thereby achieving mechanical strength criteria without complicating the forming process. Despite this potential, these steels present some challenges for forming operations because the adiabatic heat generated at high strain rates could suppress the TRIP effect, resulting in a loss of hardening capacity.

In order to create an UFG microstructure, the steel is generally homogenized and quenched before cold-rolling to obtain a large fraction of martensite. If the cold-rolled steel is then annealed in the intercritical ferrite + austenite domain, austenite will nucleate at the intersections of martensite laths. The resulting sub-micrometer austenite grain size makes it possible to thermodynamically stabilize the austenite at room temperature with Mn additions of only 5-10 wt% rather than the 20 wt% or more needed in TWIP steels [9, 10]. The retained austenite thus obtained is metastable at room temperature and can undergo TRIP if sufficient mechanical energy is applied. However, the high strain rates used in forming operations can stabilize the retained austenite via adiabatic heating and result in a significant loss in hardening capacity. This illustrates the need to understand how the stability of the austenite is related to the kinetics of the martensite transformation during TRIP. This study will seek to establish a method of characterizing the TRIP kinetics that could be used in future work to optimize compositions and high strain rate performance of TRIP steels.

Proper characterization of the kinetics of the TRIP effect in a particular alloy requires the volume fraction of retained austenite to be measured as a function of strain. This is notoriously difficult to do, however, as dif-

38 ferent retained austenite volume fraction measurement techniques can give
39 very different values for similar alloys [11, 12]. Determination of the austenite
40 content is nonetheless critical in alloy design, as it would provide the
41 metallurgist control over the strain hardening contributions from TRIP and
42 TWIP. Attempts have been made to measure the austenite volume fraction
43 as a function of strain by X-ray diffraction (XRD) [13] and magnetization
44 measurements using a ferritescope [13–15] in order to quantify TRIP kinetics.
45 While in terms of the accuracy of the measurement method, magnetic
46 measurements provide no advantage over other methods [12], the primary
47 interest of the method as employed herein is the extremely high sampling
48 rate compared to, for example, XRD or EBSD. It enables one to measure
49 the retained austenite volume fraction *in-situ* without having to pause the
50 mechanical test, which would not be possible if one wishes to perform the
51 test at the very high strain rates used during stamping operations for example.
52 Sampling rates this high are not normally obtainable by other *in-situ*
53 methods, making magnetic saturation measurements the unique choice for
54 such an application if the experimenter wants intermediate points between
55 the unstrained and ruptured states. Moreover, a more complete description
56 of the kinetics of TRIP using several hundreds or thousands of experimental
57 data points would make it possible to refine existing TRIP models used in
58 simulations of mechanical behavior.

59 Recent studies on Medium Mn steels have noted that strain hardening
60 is sometimes unstable and that long Lüders plateaus are often observed in
61 alloys containing low levels of aluminum [4, 16]. Additionally, for certain
62 compositions and heat treatments, serrations have been observed in plots
63 of $\frac{d\sigma}{d\varepsilon}$ versus either true stress or true strain. Gibbs et al. [7] suggested that
64 these serrations are due to a variable martensite transformation rate resulting
65 from the competition between the consumption and creation of martensite
66 nucleation sites during plastic deformation. Lee et al. [17] attributed the
67 serrations to dynamic strain aging (DSA) leading to the formation of Portevin
68 Le-Châtelier (PLC) bands. Cai et al. [5] instead proposed that the stability
69 of the austenite phase itself varied from one grain to another and thus could
70 transform to martensite in bursts.

71 Digital image correlation (DIC) has historically been a common method
72 for characterizing strain localizations such as PLC [18–27]. As such, it is an
73 ideal tool for investigating the jerky flow present in Medium Mn steels [28, 29].
74 This paper will seek to use a combination of DIC and *in-situ* measurements
75 of the retained austenite content via the sample’s saturated magnetization

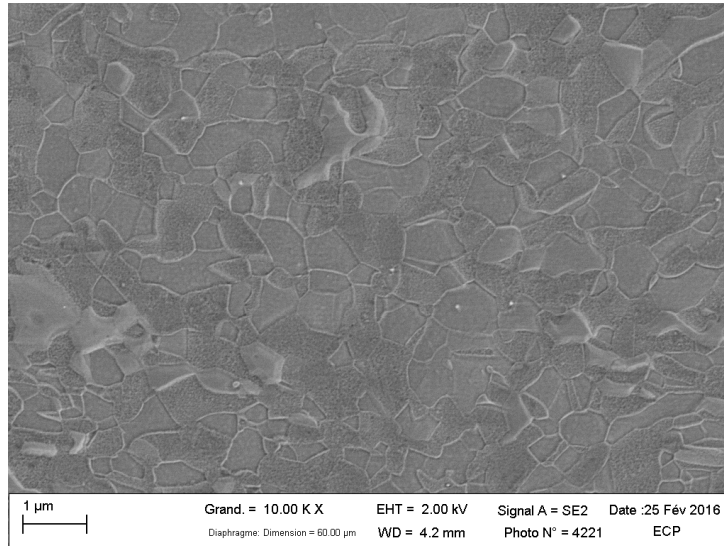
76 to demonstrate that the observed strain instabilities coincide with the TRIP
77 effect.

78 2. Materials and Methods

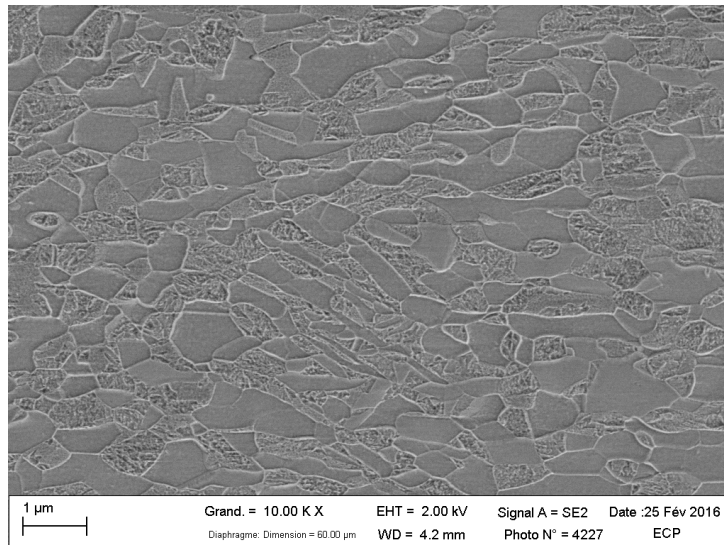
79 2.1. Fabrication and Microstructural Characterization

80 A Medium Mn steel containing 0.2 wt% C, 5 wt% Mn and 2.5 wt% Al was
81 fabricated in order to study the effects of the heat treatment process on the
82 microstructure and active deformation mechanisms in the steel. Ingots were
83 cast and then hot-rolled at austenitizing temperature to produce a homoge-
84 neous austenitic microstructure. They were then quenched and cold-rolled,
85 resulting in a mixture of deformed martensite and bainite. Finally, the steel
86 was annealed in the intercritical ferrite + austenite region at temperatures
87 of $T_{IA}=740^{\circ}\text{C}$, $T_{IA}=760^{\circ}\text{C}$, or $T_{IA}=780^{\circ}\text{C}$ for 2 minutes. During the inter-
88 critical anneal, the deformed microstructure will recrystallize and partially
89 transform to create an UFG mixture of ferrite and austenite in which some
90 retained austenite will remain stable after being air-cooled to 400°C and then
91 quenched to room temperature. A representative microstructure of the sam-
92 ple annealed at 740°C is provided in Figure 1a for the unstrained state and in
93 Figure 1b after straining until rupture. The variations in intercritical anneal-
94 ing temperature result in samples with varying retained austenite stability
95 and initial phase volume fractions. As the retained austenite is metastable,
96 it can transform to α' (bct) martensite when mechanically loaded, thereby
97 resulting in high strain hardening¹.

¹Sometimes ϵ (hcp) martensite appears as an intermediary phase between γ and α' . In the material studied, no ϵ martensite was observed in TEM or XRD so it is assumed that the retained austenite transformed directly to α' martensite when plastically strained. If ϵ martensite were present, the magnetic measurements would include it within the ferromagnetic volume fraction (not with the retained austenite).



(a)



(b)

Figure 1: (a) Microstructure of a sample annealed at 740°C in the unstrained state. The microstructure consists of ultra-fine-grained ferrite (the “smooth” phase in the image) and retained austenite (phase with a rougher surface topography) with a roughly isotropic morphology. (b) Microstructure of a sample annealed at 740°C after straining to rupture. A significant portion of the retained austenite has transformed to blocks of acicular martensite. The images are taken in the plane of the rolling direction (horizontal) and normal direction. Tension was applied in the rolling direction.

98 XRD was used to determine the initial austenite volume fraction for each
 99 intercritical annealing temperature. X-rays were produced with a $\text{Co-}K_\alpha$
 100 source with an energy of 6.915 keV. As XRD is a surface measurement,
 101 the sample was oscillated in order to increase the volume of measurement
 102 (still at the surface) by scanning over a larger area. This provides better
 103 statistical sample size in terms of the number of grains analyzed. Spectra
 104 were obtained for several orientations of the sample relative to the incident
 105 beam. The inclination angle of the sample with respect to the incident beam,
 106 ψ , was varied from 0° to 70° with a step size of 5° . The rotation of the sample
 107 around its normal to the sample surface, Φ , varied from 0° to 355° with a
 108 step of 5° . These spectra were summed together in order to minimize texture
 109 effects on the relative peak intensities of the bcc and fcc phases, which can
 110 vary significantly depending on the orientation of the sample. The volume
 111 fraction of retained austenite was calculated using the method presented in
 112 Ref. [30] and the phase volume fractions for each annealing temperature are
 113 summarized in Figure 2.

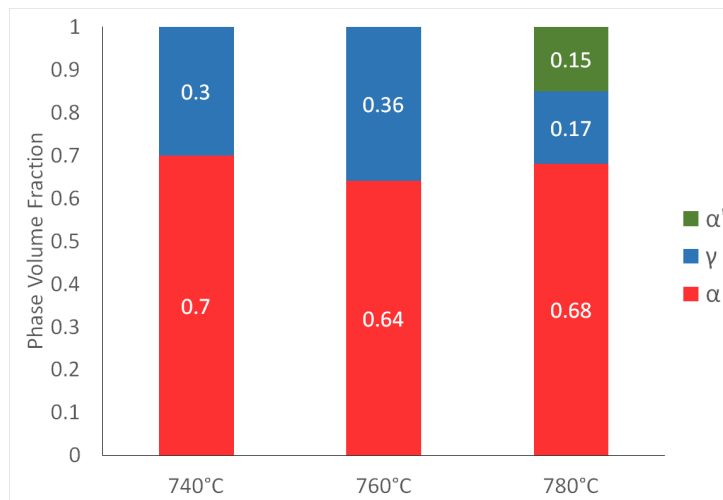


Figure 2: Initial phase volume fractions in the unstrained state after annealing and quenching as measured via XRD (α and γ) and as estimated from SEM images (α') as a function of the intercritical annealing temperature used.

114 2.2. Tensile Testing and DIC

115 Tensile dog bone specimens were cut along the rolling direction from
 116 1.25mm thick sheet. The specimens had an active gauge length 60mm long

117 and 10mm wide. Unidirectional tensile tests were performed using an Instron
118 extensometer with a 12.5mm initial length. The samples were strained until
119 rupture using an Instron 4430 electromechanical tensile testing machine with
120 a strain rate in the sample of $5 \cdot 10^{-4} \text{s}^{-1}$.

121 Images for DIC analyses were obtained by painting the polished sample
122 surface with a black and white speckle pattern and taking pictures every 3s
123 during testing. A single Canon 60D camera with a 105mm macro objective
124 lens was used to acquire the images with an exposure time of $\frac{1}{320} \text{s}^{-1}$, ISO
125 grade of 800, and an aperture f-stop of f/9. Two halogen lamps were used to
126 illuminate the specimen surface. The images recorded were 3465×5202 pixels
127 in definition with 16bit encoding and were cropped to remove excess empty
128 space and converted to .tif format for registration purposes. Correlation of
129 the images was done using 3-node triangular elements and regularization
130 using RT3-DIC [31]. Regularization, as used herein, is a means of reduc-
131 ing uncertainties in the displacement fields measured by DIC. It does this
132 through the use of an elastic constitutive law that aids the convergence of
133 the displacement field to mechanically admissible values. A 10pixel element
134 size and regularization length of 100pixels were used for the calculations.
135 Pixel size varied between 0.025 and 0.05mm from one test to another and
136 this scale was determined using the measured width of the sample. The
137 displacement fields in photo n were calculated with respect to the reference
138 image using the results from photo $n - 1$ to initialize the calculation. That
139 is, Figures 8 and 9a report the nominal strain fields calculated using the first
140 image as the reference image. Figures 8 and 9b show the strain rate fields
141 computed with the (Lagrangian) velocity field.

142 When Lüders and/or PLC bands were observed, the width and angle of
143 strain bands were characterized using ImageJ [32]. Images of the longitudinal
144 strain field where the bands were clearly defined were selected and 8 line plots
145 of the strain level were analyzed along the tensile direction in each image.
146 The resulting peaks in the line profiles were fitted with Gaussian functions.
147 The full width mid height (FWMH) was used to calculate band width and
148 the peak locations in each of the 8 profiles were used to calculate the band
149 angle. Band width measurements are corrected for the displacement of the
150 band during the 3s interval between images. The band velocity is calculated
151 using the shift in position of the strain peaks from one image to another.

152 *2.3. Magnetic Measurements of Austenite Volume Fraction*

153 A measurement system was developed based on Ref. [33] in order to per-

154 form *in-situ* measurements of the retained austenite volume fraction during
 155 tensile testing. While the setup used is applied to flat tensile samples, it could
 156 easily be adapted to other geometries for either uniaxial or biaxial tests [34].
 157 The sizes of the coils can be made larger or smaller as needed (as long as one
 158 is careful to note the number of loops in the coil as that will be needed for
 159 magnetostatic calculations and to calibrate the applied field using, for exam-
 160 ple, a Hall effect H-coil). The primary consideration when changing sample
 161 geometries is that of the measurement depth, or "skin depth", e , which is
 162 related to the frequency in the driving coil by

$$e = \frac{1}{\sqrt[2]{\pi f \sigma \mu}} \quad (1)$$

163 where σ is the electrical conductivity of the sample and μ its magnetic per-
 164 meability. In the current case, this provides an estimation of the skin depth
 165 of 8mm which is several times greater than the thickness of the sample and
 166 ensures that the measure is volumetric. For this particular sample geome-
 167 try, the frequency could be increased to 200-300Hz while still maintaining
 168 a volumetric measurement. It should be noted that this would affect the
 169 measurement in other ways, however, such as increasing the noise in the
 170 signal.

171 The setup consists of a primary coil placed around a tensile sample that
 172 creates a magnetic field in the sample. A schematic of the setup is provided
 173 in Figure 3. The magnetic response of the sample is then measured using
 174 a pick-up coil wound around the sample surface. The field is enclosed by
 175 two high-permeability ferrimagnetic yokes used to contain the magnetic field
 176 and concentrate it in the sample. The pick-up coil measures the average
 177 magnetic flux in the portion of the sample around which the driving coil
 178 is placed. In the current setup, this corresponds to a measurement over
 179 37mm of the 60mm active tensile length of the sample. The measurement
 180 is performed at low frequency (here 2Hz) in order to ensure that no eddy
 181 currents are present and that the penetration depth of the measurement is
 182 sufficient to include the whole thickness of the sample. It must be noted that
 183 because the driving coil encloses the sample, it is not possible to perform DIC
 184 and magnetic measurements simultaneously on the same sample with this
 185 method. Instead, the plastic strain in tensile experiments is compared with
 186 the plastic strain from magnetic measurement experiments and the presence
 187 of bands is indicated by plateaus that manifest in the extensometer data.
 188 While specific bands cannot be correlated to specific increases in martensite

189 fraction, the existence of bands in both tensile and magnetic data over the
 190 same ranges would strongly indicate their coincidence.

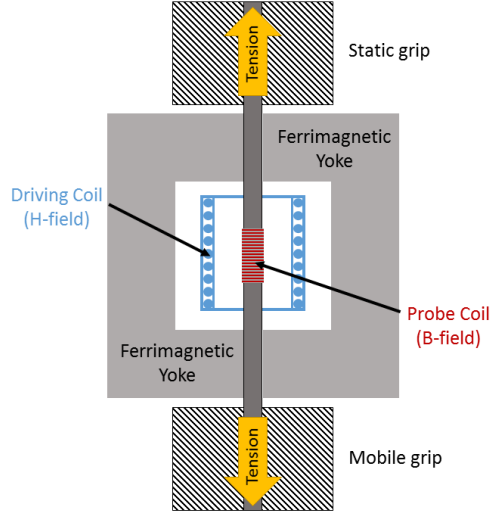


Figure 3: Schematic of the magnetic measurement system. A primary coil (H-field) placed around the tensile sample induced a magnetic H field in the sample, and the current thereby induced in the sample is measured by the probe coil (B-field) and converted to magnetization, M . The coils are enclosed in two ferrimagnetic yokes to contain the magnetic fields.

191 By cycling the current applied to the driving coil at a frequency of 2Hz,
 192 hystereses can be obtained using the readings from the pick-up coil with
 193 which the magnetic properties of the sample are analyzed. In this method, as
 194 it is typically employed for retained austenite measurements, the saturation
 195 value of the magnetization, M^s , is used to determine the combined volume
 196 fraction of the ferromagnetic phases (here ferrite and martensite) using a
 197 simple mixture rule

$$M^s = f_{\alpha+\alpha'}M_{\alpha+\alpha'}^s + f_{\gamma}M_{\gamma}^s \quad (2)$$

198 where ferrite, martensite, and austenite are represented by the subscripts
 199 α , α' , and γ respectively. The value of M_{γ}^s is assumed to be 0 due to the
 200 austenite's negligible magnetic response compared to ferrite or martensite.
 201 Assuming that $M_{\alpha}^s = M_{\alpha'}^s = M_{\alpha+\alpha'}^s$, one can express the ferromagnetic vol-
 202 ume fraction as

$$f_{\alpha+\alpha'} = \frac{M^s}{M_\alpha^s} \quad (3)$$

203 where M^s is the measured saturated magnetization for a given strain level,
 204 the subscript α denotes a value measured in a 100% ferromagnetic sample,
 205 and $f_{\alpha+\alpha'}$ the combined volume fraction of ferrite and martensite. Figure 4
 206 shows the $M - H$ hystereses thus obtained over 37mm of the 60mm tensile
 207 length for increasing levels of plastic strain for a sample annealed at 780°C. It
 208 can be seen that as the plastic strain increases, the magnetostatic energy
 209 in the sample also increases (visible in its integral and in the value of M^s)
 210 due to the martensitic transformation.

211 Equation (3) is frequently used to calculate the retained austenite volume
 212 fraction from ferritescope measurements [14, 35, 36], but it is not suitable
 213 for *in-situ* measurements at stresses exceeding 100-200 MPa due to stress
 214 effects on the sample's magnetic properties [37]. The value of M_α^s usually
 215 comes from a reference measurement and is not called into question. M^s
 216 is more questionable. Figure 5 shows, for example, a clear sensitivity of
 217 the magnetization to an applied stress in unloading during which no phase
 218 change occurs. This phenomenon, which is known as the Villari effect [38],
 219 must be taken into account. Equation (3) assumes that the true saturation
 220 is reached during the experiments. This is not achieved in Figure 4. This
 221 unsaturated situation leads to some localization effects (i.e. the magnetic
 222 field in the ferromagnetic phase is different from the applied field). Two
 223 kinds of corrections must consequently be considered as discussed in the
 224 following sections.

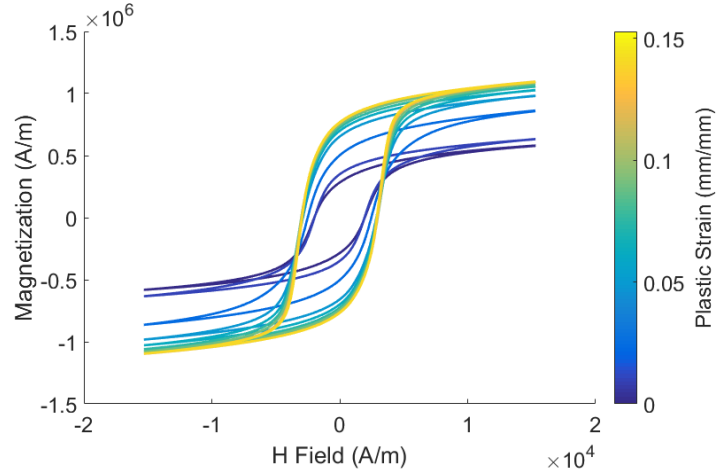


Figure 4: $M - H$ hystereses obtained during unloading of a sample annealed at 780°C after several levels of plastic strain (denoted by the color bar) where M is the sample magnetization and H the applied magnetic field. The saturation value of magnetization, M^s , increases with increasing plastic strain due to the martensite transformation.

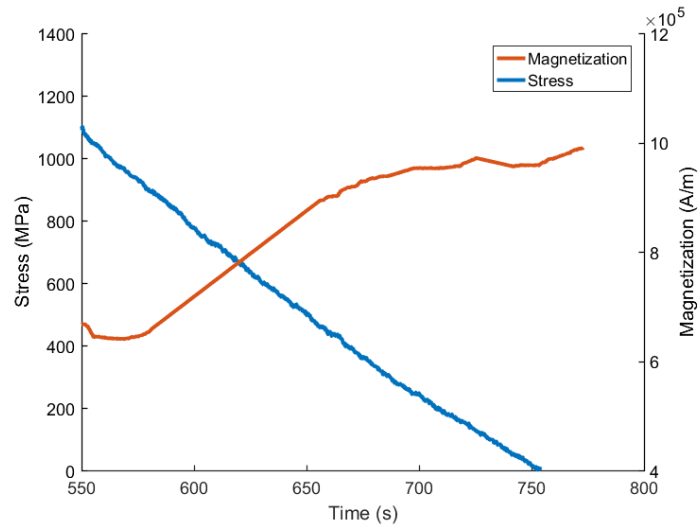


Figure 5: *In-situ* magnetization measurements of a sample annealed at 760°C during unloading after straining until a point just before necking. During unloading, the microstructure does not change (i.e. no TRIP occurs), yet the magnetization increases. This is a result of the Villari effect. The magneto-mechanical coupling demonstrated herein must be corrected if measurements are to be made under an applied stress.

225 *2.3.1. Stress Correction*

226 If measurements of the sample magnetization are to be made *in-situ*,
 227 then the effect of the applied stress on the magnetic properties must be
 228 taken into account since it has been shown that even modest loads can have
 229 a non-negligible impact on magnetization [39]. A model has been recently
 230 developed, which was used to predict the magneto-mechanical coupling in
 231 dual phase steels for stresses up to approximately 600 MPa by relating kine-
 232 matic hardening to magneto-elastic effects occurring within each phase at
 233 the grain scale [38]. This approach has been adapted to the current study to
 234 correct for the Villari effect.

235 The multidomain model used here was first introduced in Ref. [40]. It
 236 is able to model the magnetization and the magnetostrictive strain of an
 237 isotropic polycrystalline ferromagnetic material submitted to a uniaxial ap-
 238 plied stress. Two scales are involved: the domain scale (6 domains are con-
 239 sidered) and the grain scale taken as an equivalent polycrystalline medium.
 240 Macroscopic and crystallographic frames of reference are coincident (i.e.
 241 $(\mathbf{x}, \mathbf{y}, \mathbf{z}) = ([100], [010], [001])$). In the simplified description we propose here,
 242 the rotation of the magnetization is not considered. The stress tensor $\boldsymbol{\sigma}$
 243 and magnetic field \mathbf{H} are assumed to be homogeneous. The free energy of a
 244 domain ϕ then becomes a sum of the magnetostatic (Zeeman) and magne-
 245 toelastic energies:

$$W_\phi = -\mu_0 \mathbf{H} \cdot \mathbf{M}_\phi - \boldsymbol{\sigma} : \boldsymbol{\epsilon}_\phi^\mu \quad (4)$$

246 where μ_0 is the magnetic permeability of vacuum, \mathbf{M}_ϕ is the magnetization
 247 vector inside a domain such that $|\mathbf{M}_\phi| = M_\alpha^s$, and $\boldsymbol{\epsilon}_\phi^\mu$ the magnetostrictive
 248 strain tensor (see details in Ref. [40]). The volume fraction of a domain f_ϕ
 249 is calculated from Boltzmann statistics:

$$f_\phi = \frac{\exp(-A_s W_\phi)}{\sum_\phi \exp(-A_s W_\phi)} \quad (5)$$

250 where A_s is a fitted parameter proportional to the magnetic initial suscepti-
 251 bility χ^0 without an applied stress ($A_s = \frac{3\chi^0}{\mu_0 M_\alpha^{s2}}$). The magnetization of the
 252 single crystal is obtained from an averaging operation

$$\mathbf{M} = \sum_\phi f_\phi \mathbf{M}_\phi \quad (6)$$

253 when the magnetic field and uniaxial stress σ are applied along \mathbf{x} . After
 254 simplifying and applying a second order Taylor expansion of the exponential,
 255 one obtains

$$M = \mathbf{M} \cdot \mathbf{x} = \chi_0 H \left(1 + \frac{6\chi_0}{\mu_0 M_\alpha^{s2}} \lambda \sigma \right) \quad (7)$$

256 where λ is a global magnetostriction constant. This relationship permits a
 257 new definition of the initial susceptibility in which it is linearly related to the
 258 magnitude of stress as

$$\chi_\sigma^0 = \frac{M}{H} = \chi^0 \left(1 + \frac{6\chi^0}{\mu_0 \cdot M_\alpha^{s2}} \lambda \sigma \right) \quad (8)$$

259 This equation applies for the initial susceptibility but is assumed to also
 260 be applicable for the secant susceptibility at any magnetic field level and
 261 particularly at the magnetic field level at which the experimental saturation
 262 of magnetization is measured. A linear correction of the stress effect is then
 263 obtained

$$\chi_\sigma = \chi(1 + \eta\sigma) \quad (9)$$

264 where η is a constant that can be identified using magnetization measure-
 265 ments obtained during elastic unloading (see Figure 5) without any phase
 266 transformation. The constant η is typically on the order of $-10^{-5} \text{ MPa}^{-1}$.
 267 This relationship between susceptibility and stress applies to the magnetiza-
 268 tion level as well, so a stress-sensitive “saturated” magnetization is defined
 269 as

$$M_\sigma^s = f_{\alpha+\alpha'} \chi_\sigma H = f_{\alpha+\alpha'} M_\alpha^s (1 + \eta\sigma) = M^s (1 + \eta\sigma) \quad (10)$$

270 Equation (3) is thus modified to become

$$f_{\alpha+\alpha'} = \frac{M_\sigma^s}{(1 + \eta\sigma) M_\alpha^s} \quad (11)$$

271 which provides a correction to the ferromagnetic volume fraction $f_{\alpha+\alpha'}$ for a
 272 magnetic saturation value M^s obtained with applied stress σ .

273 The effect of the stress correction is shown in Figure 6 for a sample an-
 274 nealed at 780°C . This linear correction of the M^s value improves the accu-
 275 racy of retained austenite measurements at high stress where differences of

276 up to 0.05 in $f_{\alpha+\alpha'}$ were observed between calculations using M^s and M_σ^s . It
 277 also makes it possible to determine whether or not stress-induced martensite
 278 transformation occurs.

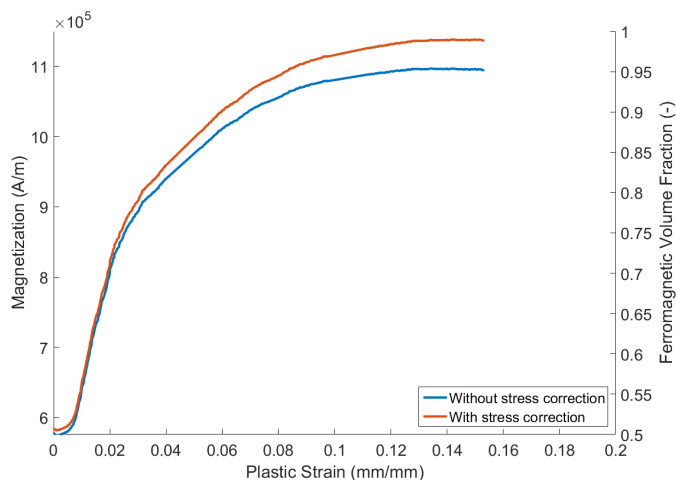


Figure 6: Magnetization and volume fraction of ferromagnetic phases (ferrite + martensite) for $T_{IA} = 780^\circ\text{C}$ as functions of plastic strain as calculated from *in-situ* data with and without the stress correction developed in this study. High levels of stress reduce the measured magnetization and result in several percent difference in calculated volume fractions.

279 2.3.2. Localization Correction

280 The ratio presented in Equation (3) is the most commonly used method
 281 of calculating the ferrite + martensite volume fraction using magnetic mea-
 282 surements. However, these equations require the assumption that the field
 283 H is uniform throughout the microstructure, which is not true as the neg-
 284 ligible magnetic response of the austenite creates a demagnetizing field in
 285 the ferromagnetic phase [41], which changes the magnetic properties of the
 286 macroscopic sample. An Eshelby model for a spherical inclusion embedded in
 287 a homogeneous matrix can be used to localize the field H within each phase
 288 and thus correct for the effect of the demagnetizing field on the magnetic
 289 susceptibility χ of the sample [42, 43]. If the magnetic behavior in a phase i
 290 is described by

$$\mathbf{M}_i = \chi_i \mathbf{H}_i \quad (12)$$

291 then the solution to the Eshelby problem will give

$$\mathbf{H}_i - \mathbf{H}^\infty = \frac{1}{3 + 2\chi^\infty}(\mathbf{M}^\infty - \mathbf{M}_i) \quad (13)$$

292 where \mathbf{H}^∞ and \mathbf{M}^∞ denote the macroscopic (average) magnetic field and
 293 magnetization and χ^∞ is the macroscopic susceptibility defined as $\chi^\infty =$
 294 M^∞/H^∞ . Applying Equation (13) to the current problem without an applied
 295 stress, the field along the magnetization axis in the ferromagnetic phase α
 296 becomes

$$H_\alpha = H + \frac{1}{3 + 2\chi}(M - M_\alpha) \quad (14)$$

297 with $\chi = M/H$. With the magnetic susceptibility of the ferromagnetic phase
 298 defined as $\chi_\alpha = M_\alpha/H_\alpha$, it follows that

$$\frac{H}{H_\alpha} = \frac{3 + 2\chi + \chi_\alpha}{3 + 3\chi} \quad (15)$$

299 where the volume fraction of the ferromagnetic phase is defined as

$$f_\alpha = \frac{M}{M_\alpha} \quad (16)$$

300 It should be noted that this definition differs from the classical one given
 301 in Equation (3) since true saturation is not achieved. This relationship is
 302 most appropriate for low magnetic field levels where the magnetic behavior
 303 can be considered as linear. Equation (16) is transformed by introducing the
 304 corresponding magnetic fields and susceptibilities

$$f_\alpha = \frac{\chi H}{\chi_\alpha H_\alpha} \quad (17)$$

305 Finally, by combining Equations (15) and (17), f_α reads

$$f_\alpha = \frac{\chi(3 + 2\chi + \chi_\alpha)}{\chi_\alpha(3 + 3\chi)} \quad (18)$$

306 However, this expression is only valid considering the initial suscepti-
 307 bility (or linear behavior). By recognizing that $\chi = \frac{1}{3}A_s\mu_0M^{s2}$ and $\chi_\alpha =$
 308 $\frac{1}{3}A_s\mu_0M_\alpha^{s2}$ following the theoretical development in Section 2.3.1 and consid-
 309 ering $\chi \gg 1$, the volume fraction of ferromagnetic phase is expressed as

$$f_{\alpha} \approx \left(\frac{M^s}{M_{\alpha}^s} \right)^2 \quad (19)$$

310 The volume fraction of ferromagnetic phase thus obtained provides a second
 311 estimation of $f_{\alpha+\alpha'}$ with the hypothesis that the ferromagnetic regions do not
 312 interact closely. This description is only partially true and a more appropriate
 313 estimation may be defined between expressions (3) and (19). We propose the
 314 use of a combination of the two estimations using a ratio parameter κ , which
 315 leads to

$$f_{\alpha+\alpha'} = \frac{M^s}{M_{\alpha}^s} + \kappa \left[\left(\frac{M^s}{M_{\alpha}^s} \right)^2 - \frac{M^s}{M_{\alpha}^s} \right] \quad (20)$$

316 for the case without applied stress and

$$f_{\alpha+\alpha'} = \frac{M_{\sigma}^s}{(1 + \eta\sigma)M_{\alpha}^s} \left[1 + \kappa \left(\frac{M^s}{(1 + \eta\sigma)M_{\alpha}^s} - 1 \right) \right] \quad (21)$$

317 to account for the Villari effect and measured saturation M_{σ}^s . The κ pa-
 318 rameter is identified via a magnetic measurement on a partially transformed
 319 material (in the unstrained state) with a known volume fraction of ferromag-
 320 netic material and was determined to be $\frac{2}{3}$ in the current case.

321 If Equation (21) is used instead of the ratio given in Equation (3), this
 322 correction improves measurement accuracy when high volume fractions of
 323 retained austenite are present.

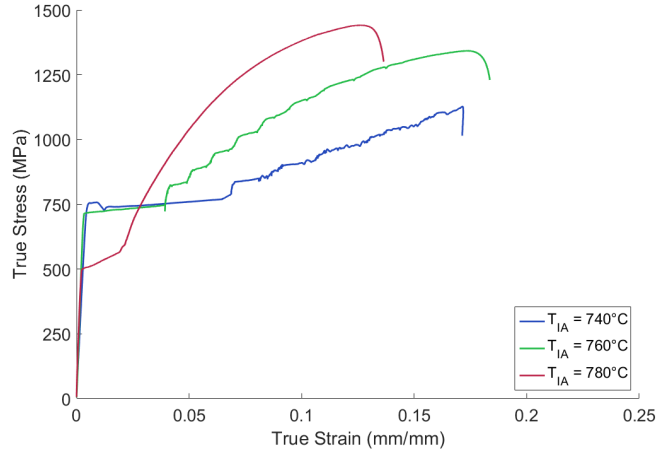
324 In the current study, this method is implemented for *in-situ* tensile test-
 325 ing at a strain rate of $5 \cdot 10^{-4} \text{s}^{-1}$ on flat tensile samples for each intercritical
 326 annealing temperature T_{IA} . While the quasi-static strain rate used herein is
 327 hardly representative of the strain rates used in forming operations, the inten-
 328 tion is to demonstrate the utility of the method in performing high-frequency
 329 measurements of the retained austenite volume fraction. A successful imple-
 330 mentation of this method would make it possible to perform further studies
 331 at much higher strain rates for which no method of measuring the retained
 332 austenite fraction during straining exists.

333 3. Results and Discussion

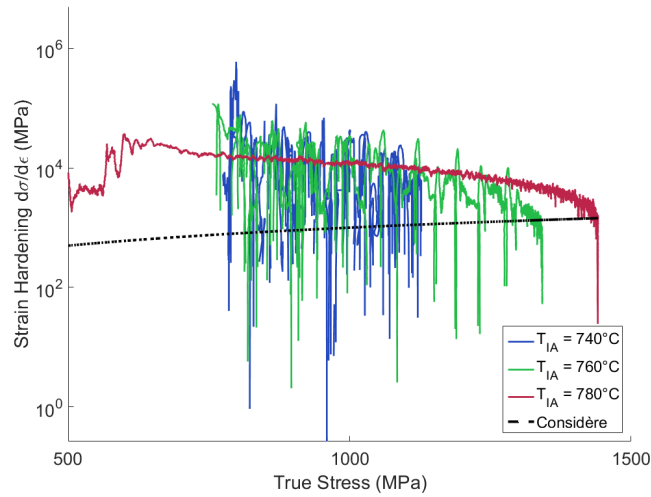
334 3.1. Tensile Characterization

335 The samples' mechanical behavior in uniaxial tension varied significantly
 336 with intercritical annealing temperature. Such a variation of mechanical

337 behavior with respect to intercritical annealing temperature is frequently
338 observed in Medium Mn steels even for differences in T_{IA} of only 5-10°C
339 [4–8]. As seen in Figure 7(a), the samples annealed at 740°C and 760°C
340 showed significant yield point elongation in the form of a Lüders plateau.
341 The samples annealed at 780°C did not show a traditional Lüders band, but
342 rather a sort of pseudo-plateau during which stress is not constant. Tests
343 using DIC later confirmed that there is no strain localization in samples
344 annealed at 780°C, so the pseudo-plateau observed on its tensile curve is not
345 a Lüders effect.



(a)



(b)

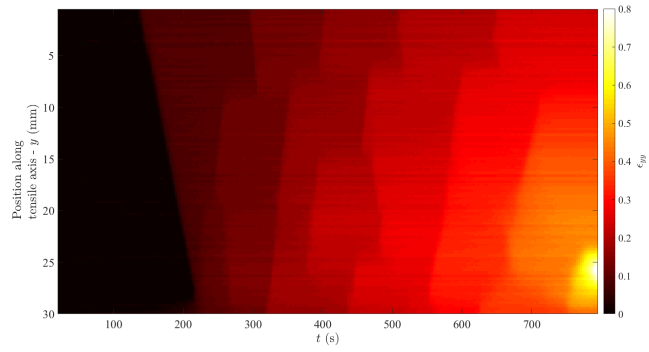
Figure 7: (a) Representative true stress - true strain curves for each intercritical annealing temperature employed. (b) Strain hardening rate $d\sigma/d\epsilon$ plotted as a function of the true stress σ for each intercritical annealing temperature T_{IA} . The Considère criterion, wherein the strain hardening rate is equal to the applied true stress, is included as well to show that localizations can bring the hardening rate below this criterion and lead to premature necking.

346 Following yielding, the samples annealed at 740°C and 760°C showed a
 347 form of jerky flow that manifested itself as a series of strain bands once the

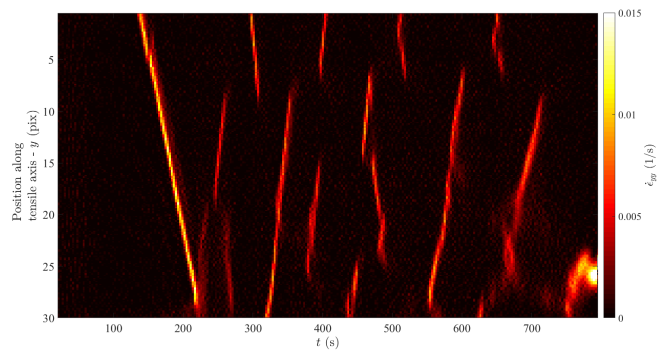
348 Lüders band had fully propagated across the sample. Due to the difference in
349 the gauge lengths between the sample and the extensometer (i.e. 60mm and
350 12.5mm, respectively) the jerky flow took the form of “steps” on the tensile
351 curve. Fewer steps were observed for the samples annealed at 740°C than
352 760°C, but they were generally longer than the steps seen for 760°C. The
353 sample annealed at 780°C did not show any observable jerky flow behavior.
354 The strain hardening of the samples, shown in Figure 7(b) was character-
355 ized using the derivative of the tensile curve, $d\sigma/d\epsilon$, plotted against the true
356 stress. Approximating $d\sigma/d\epsilon$ with a trapezoidal method will naturally create
357 a very noisy signal, so a Savitsky-Golay filter [44] was applied to smooth
358 out the data. It can be seen that the samples annealed at 740°C and 760°C
359 exhibited significant serrations in the strain hardening curves that were not
360 observed in the sample annealed at 780°C. These serrations are very similar
361 to those previously observed [5, 7, 17] and are expected to come from the
362 same phenomenon—strain bands due to a sort of dynamic strain aging effect—
363 observed in other Medium Mn alloys. Such serrations are undesirable for
364 forming operations because passing below the Considère criterion can lead
365 to premature necking if the relaxation associated with a given strain band
366 is significant. The jerky flow in samples annealed at 740°C and 760°C was
367 further characterized using DIC.

368 3.2. Digital Image Correlation

369 Strain maps obtained by DIC showed two distinct phenomena, namely,
370 a clear Lüders band at yielding, followed by a series of other strained bands
371 that were primarily propagative. It can be seen in Figure 8 that for the
372 sample annealed at 740°C, several propagative PLC-type bands move across
373 the gauge length of the specimen and always in the opposite direction of
374 the initial Lüders band. The strain increment everywhere outside the band
375 is very small. In the sample annealed at 760°C, a Lüders band is once
376 again observed, but the subsequent bands are much less continuous. Bands
377 frequently will stop suddenly and another band appears elsewhere along the
378 length of the sample, as seen in Figure 9.



(a)



(b)

Figure 8: Spatiotemporal DIC measurements of (a) nominal longitudinal strain and (b) longitudinal strain rate measurements for annealing at 740°C . The horizontal axis is the time in seconds, the vertical axis represents the position along the tensile direction, and the color bar represents either the total strain in the tensile direction or the strain rate ($1/\text{s}$) in the tensile direction.

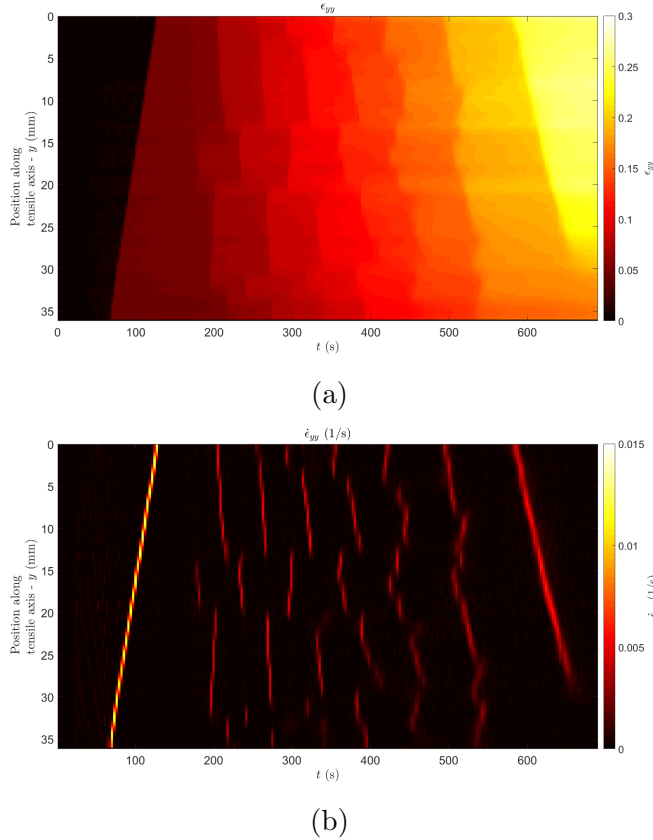


Figure 9: Spatiotemporal DIC measurements of (a) nominal longitudinal strain and (b) longitudinal strain rate measurements for annealing at 760°C . The horizontal axis is the time in seconds, the vertical axis represents the position along the tensile direction, and the color bar represents either the total strain in the tensile direction or the strain rate ($1/\text{s}$) in the tensile direction.

379 The bands themselves were characterized by using ImageJ to analyze
 380 images of the strain increment field for the component along the loading di-
 381 rection $\Delta\varepsilon_{yy}$ and plotting several line profiles along the loading direction.
 382 Using the peaks in the line profiles, the band width and inclination angle
 383 from the loading direction were calculated for both the Lüders bands and
 384 the subsequent PLC-type bands. The PLC-type bands that appeared in the
 385 sample annealed at 760°C were too random to obtain reliable measurements
 386 of their propagation speed and width, so values are only presented for the
 387 sample annealed at 740°C . Table 1 summarizes the results of the band char-
 388 acterization (note that the velocities given for PLC bands is the average of

389 each PLC band for that T_{IA}). Figure 10 shows the strain magnitude in each
 390 band for both intercritical annealing temperatures. The strain in the first
 391 band is significantly higher than in subsequent band, suggesting that two
 392 different phenomena are responsible for the first band and those that follow.

Table 1: DIC strain band characterization

		Velocity (mm/s)	Band Width (mm)	Inclination $\theta(^{\circ})$
740°C	Lüders	0.381 ± 0.008	0.605 ± 0.033	66.4 ± 0.496
	PLC	0.583 ± 0.084	0.770 ± 0.042	69.8 ± 0.605
760°C	Lüders	0.623 ± 0.003	0.246 ± 0.044	69.3 ± 0.800

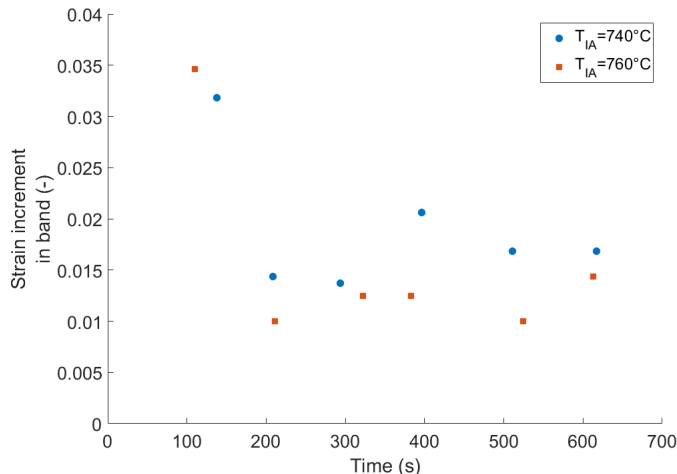


Figure 10: Plot of the longitudinal strain increment in each band as a function of time for samples annealed at 740°C and 760°C. The stark difference in magnitude between the first band and those that follow suggests two different underlying mechanisms. It is proposed that there is an initial Lüders band followed by PLC-type bands.

393 As can be seen in Figure 11, the velocity of the PLC-type bands decreased
 394 roughly linearly with total strain while the strain rate in the band remained
 395 constant. Both this trend and the observed velocities are in agreement with
 396 previous experiments [29]. Comparing these results to those published for
 397 aluminum alloys or TWIP steels exhibiting the PLC effect, it can be seen
 398 that for propagative type-A PLC bands the band velocity also linearly de-
 399 creases with strain at a rate roughly one order of magnitude slower than ob-
 400 served here [45–47]. It is thus proposed that the bands observed in Medium

401 Mn steels are likely type-A PLC bands. The difference in rate of band ve-
 402 locity decrease with strain could be due to chemical composition (via the
 403 solute concentrations and concentration gradients), and interaction with the
 404 martensitic transformation, or both. PLC resulting from an interaction be-
 405 tween dislocations and Mn-C clusters has been previously proposed by Jung
 406 and DeCooman [48], who suggested that the movement of C atoms in C-Mn
 407 clusters between octahedral and tetrahedral interstices could lock partial dis-
 408 locations in stacking faults. Hickel et al. [49] also confirmed a strong effect of
 409 C-Mn clusters on SFE and showed that the active deformation mechanism
 410 can locally redistribute solutes and thus significantly change the local SFE.

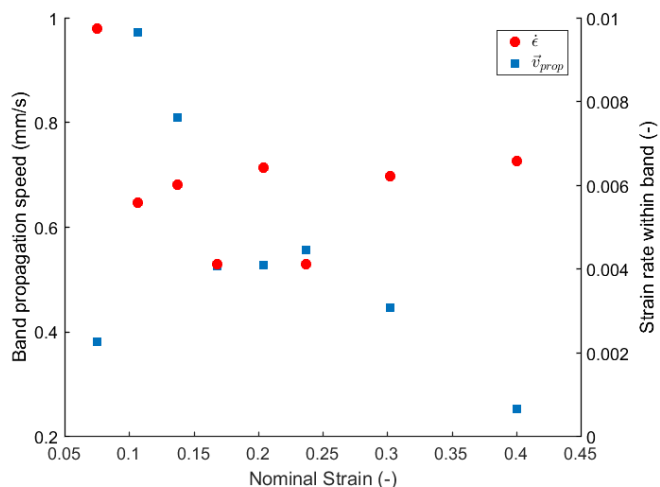
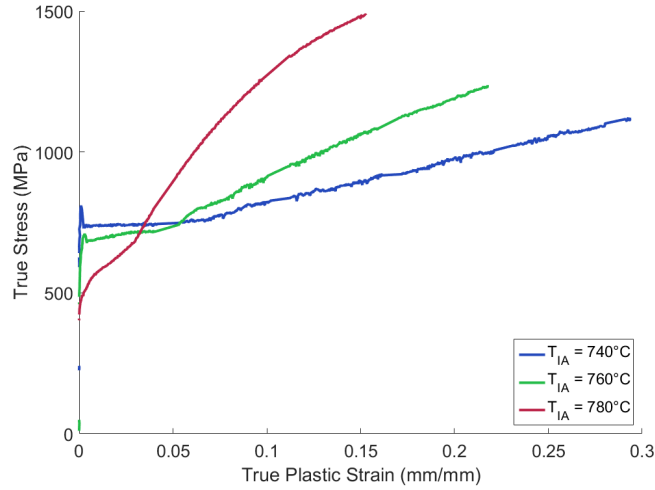


Figure 11: Calculated band propagation velocities and the strain rate within each band plotted as a function of the nominal strain for the sample annealed at 740°C. The velocities of the PLC-type bands decrease roughly linearly with strain while the strain rate in the band remains more or less constant for bands occurring after the Lüders band.

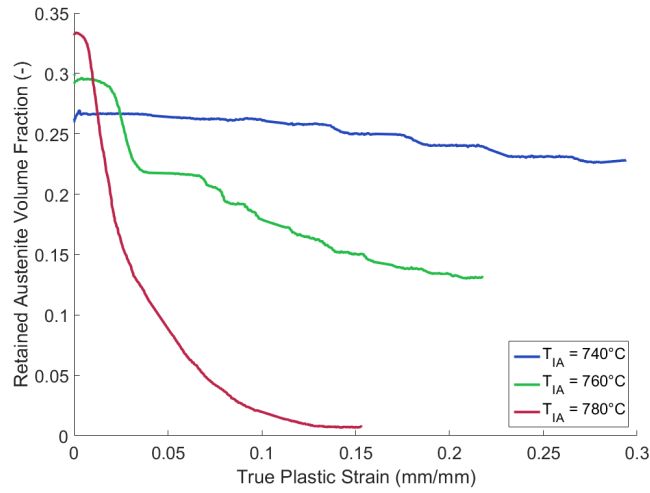
411 3.3. Magnetic Measurements

412 The *in-situ* measurement of sample magnetization throughout tensile
 413 testing allowed the retained austenite volume fraction (via the combined
 414 volume fractions of ferrite and martensite) to be calculated. This method
 415 provided continuous phase volume fraction measurements over the 37mm of
 416 the sample's gauge length that is contained within the driving coil, thereby
 417 ensuring a very large measurement volume that is statistically representa-
 418 tive of the steel's microstructure. Additionally, because of the difference

419 in the magnetic measurement length and the tensile length of the sample
420 (60mm), if any strain localizations are present that induce TRIP their effect
421 should produce plateaus in the measurement of the austenite volume frac-
422 tion as a function of strain. Figure 12 presents the true stress and retained
423 austenite volume fraction as a function of true plastic strain as calculated
424 from the crosshead displacement for each intercritical annealing tempera-
425 ture. The plastic strain is calculated by assuming that the tensile machine
426 deforms purely elastically, making it possible to simply remove the elastic
427 strain $\epsilon^e = \sigma/E$.



(a)



(b)

Figure 12: Plots of (a) true stress and (b) retained austenite volume fraction vs. true plastic strain for each intercritical annealing temperature. The same steps previously observed in tensile data are again seen in the retained austenite volume fraction, suggesting that the TRIP effect coincides with the passage of strain bands. Both plots share identical strain axes.

428 An important clarification must be made. Because the magnetic measure-
 429 ment system completely encloses the tensile sample, the strain in the sample

430 was not directly measured during tensile testing. Instead, the true plastic
431 strain is used in order to eliminate the elastic strain in the tensile machine
432 as recorded by the crosshead displacement. The curves presented for each
433 intercritical annealing temperature in Figure 12 are thus from two different
434 samples and, while the dispersion in tensile data was small, the indications
435 of banding will not be identical in two curves for the same annealing tem-
436 perature. This effect aside, it is clear that in the magnetic data the retained
437 austenite volume fraction decreases sharply at several points, creating the
438 same “steps” that were previously seen in tensile data using an extensome-
439 ter. This strongly suggests that the passage of strain bands and the TRIP
440 effect are, in fact, coincident.

441 Previous work by Wang et al. [29] showed that the TRIP effect occurred
442 during the Lüders plateau, but that little or no transformation occurred dur-
443 ing PLC. It has also been shown by Sun et al. [28] that some transformation
444 does occur during PLC. The current work then confirms Sun et al.’s result
445 that TRIP occurs both during Lüders deformation and PLC. It is possible
446 that when the amount of austenite transformed by a single band is low (i.e.
447 $\approx 1\text{-}2\%$), the heat given off by the reaction is “drowned out” by the adiabatic
448 heating (even at quasi-static strain rates) in the band and thus difficult to
449 distinguish by infrared imagery. It is also possible that the retained austenite
450 in Wang et al.’s steel was simply stable enough to not transform noticeably.
451 This point is important to address because the origin of the PLC-type local-
452 izations is, as mentioned in Section 3.2, a point of contention currently and
453 if one is able to have PLC without a phase transformation in these steels it
454 would indicate that the PLC effect is not initiated by the phase transforma-
455 tion.

456 The correction of the stress effect on magnetization allowed stress-induced
457 and strain-induced martensite transformations to be separated and showed
458 that no stress-induced transformation occurred during elastic deformation.
459 Figure 13 shows the sample magnetization vs. true stress for each intercritical
460 annealing temperature. In all samples the magnetization remains constant
461 in the elastic domain, proving that no stress-induced transformation occurs
462 during elasticity and that the first martensitic transformation occurs only
463 after the onset of plasticity.

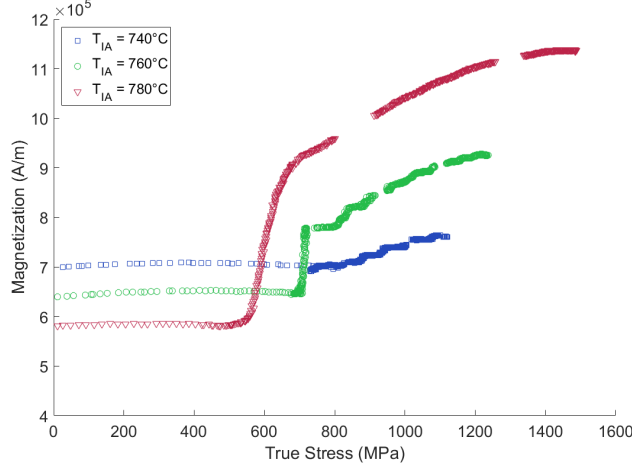


Figure 13: Sample magnetization (with stress correction) plotted as a function of true stress shows that during elasticity, the magnetization remained constant meaning that there is no stress-induced transformation before yielding.

The changes of the austenite volume fraction ($f_\gamma = 1 - f_{\alpha+\alpha'}$) as calculated from magnetization measurements were used to fit the parameters of an Olson-Cohen model [50]

$$f_{\alpha+\alpha'} = 1 - \exp(-\beta(1 - \exp(-\alpha\epsilon^p))^n) \quad (22)$$

464 where the α parameter is related to both the number of shear bands and
 465 the stacking fault energy of the austenite, β accounts for the probability
 466 of a martensite nucleus to form, which is in turn defined by the driving
 467 force for the phase transformation. The exponent n is usually considered a
 468 material constant, but is not fixed in the current case due to the varying
 469 composition of the austenite for each T_{IA} . Perhaps surprisingly, the model
 470 can be fit to the measured austenite volume fraction change fairly well even
 471 when strain localizations are present as can be seen in Figure 14. This
 472 provides a set of effective Olson-Cohen parameters, given in Table 2, that
 473 are valid in a macroscopic sense for a given microstructure and austenite
 474 stability. Estimations of the martensite start temperature, M_S , are also
 475 provided in Table 2.

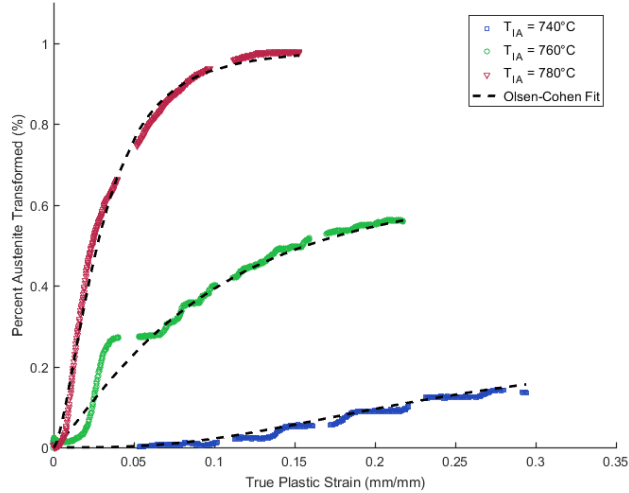


Figure 14: Experimental transformation rates of retained austenite in alloys annealed at 740°C, 760°C, 780°C (symbols), fit with a standard Olsen-Cohen model (dashed lines).

Table 2: Macroscopic Olson-Cohen Parameters

$T_{IA}(^{\circ}C)$	α	β	n	$M_S(^{\circ}C)$
740	7.52	0.27	3.90	-7.5
760	8.05	1.04	1.25	26
780	14.0	4.25	1.58	58

476 Examining the trends in the Olson-Cohen model’s fitting parameters, one
 477 could begin to explain the differences in TRIP kinetics between samples for
 478 different intercritical annealing temperatures. Considering the α and β pa-
 479 rameters, Olson and Cohen [50] suggest that they should increase as the test
 480 temperature decreases for a given material. Thus, α and β should increase
 481 as austenite stability decreases, which is the trend observed herein. However
 482 n should have a maximum at an intermediate test temperature, but instead
 483 there is a minimum at 760°C and a large discrepancy at 740°C compared
 484 to the other annealing temperatures. The exponent n is related to the com-
 485 position of the material, which should not vary to such a degree given that
 486 only the intercritical annealing temperature changes and not the nominal
 487 composition. As a result, it seems that the base Olson-Cohen model is suf-
 488 ficient to confirm trends in SFE between alloys, but not to truly understand

489 differences in TRIP kinetics between two different austenite compositions.
490 Alternatively, it is also possible that the trends in the model parameters are
491 not the same when varying test temperature (as Olson and Cohen did) or
492 austenite composition/volume fraction (as in the current study). In any case,
493 one could further investigate the interactions between phases using numerical
494 models parameterized using the kinetics obtained for each intercritical an-
495 nealing temperature in these experiments. Doing so would allow for a better
496 description of the true kinetics since microstructure-level strains and stresses
497 would be available.

498 4. Conclusions

499 A system for measuring the retained austenite volume fraction via sam-
500 ple magnetization has been developed and used *in-situ* during tensile testing.
501 This method provides very rapid measurements of the retained austenite frac-
502 tion (here 2Hz) relative to other methods such as XRD or EBSD and enables
503 measurements to be conducted without pausing or otherwise interrupting the
504 tensile test. However, the measurement system encloses the sample, so the
505 strain must either be calculated from the crosshead displacement or mea-
506 sured with a strain gauge (extensometers or DIC methods cannot be used on
507 a sample upon which these magnetic measurements are made).

508 This system was employed to study the effects of intercritical annealing
509 temperature and strain instabilities on the TRIP effect present in samples of a
510 Fe-0.2C-5Mn-2.5Al Medium Mn steel that had been intercritically annealed
511 at 740°C, 760°C, and 780°C. There was a clear difference in the rate of
512 martensitic transformation for different intercritical annealing temperatures,
513 though it is not understood exactly how the micromechanics of the TRIP
514 effect are altered by varying T_{IA} . Plateaus corresponding to the passage of
515 strain bands appear both in tensile data measured with an extensometer and
516 in magnetization data. This indicates that the martensitic transformation
517 and the passage of strain bands occur simultaneously.

518 DIC analyses of the strain bands present in samples annealed at 740°C
519 and 760°C showed that the Lüders and post-Lüders bands are from two sepa-
520 rate phenomena and that, because the band velocity in the sample annealed
521 at 740°C decreased linearly with strain at a rate similar to that observed
522 in Al and TWIP steels exhibiting PLC [23–25], the post-Lüders bands are
523 likely to be propagative type-A PLC bands. The micromechanical origin of

524 the PLC effect is unknown and has been hypothesized to be related to either
525 C-Mn clusters or the martensite transformation itself [5, 7, 17].

526 The experimental procedures established were intended to study the ef-
527 fects of T_{IA} and strain instabilities on TRIP, but provide a tool that can
528 be extended to use in the optimization of Medium Mn TRIP steels in gen-
529 eral. The magnetic measurement system is capable of measuring the retained
530 austenite fraction *in-situ* even for the high strain rates used in forming op-
531 erations or for applied loads other than unidirectional tension. The data
532 obtained could be used to calibrate a micromechanical model of TRIP kinet-
533 ics and thus enable the metallurgist to predict the martensite transformation
534 rates for a given microstructure and strain rate. The extension of this method
535 to such a purpose is underway as well as studies of the effect of the applied
536 strain rate on transformation kinetics.

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