Dynamic Strain Aging of Various Steels

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The dynamic strain aging characteristics of two dual phase steels, a high strength low alloy (HSLA) steel, a 1008 steel and an interstitial free (IF) steel were determined from tensile properties at temperatures in the range 295 to 460 K (22 to 187 °C) and strain rates between 6×10^{-6} to $10^{-2}s^{-1}$. All except the IF steel were found to be susceptible to dynamic strain aging, as evidenced by increases in tensile strength. The largest positive change was observed in the 1008 steel while the dual phase and HSLA steels showed much smaller increases. Also, large decreases (up to 75 pct) in uniform elongation were noted for the 1008 steel while the decreases were minimal for the dual phase and HSLA steels. The IF steel did not strain age and showed a slight increase in uniform elongation with increasing temperature. Based upon uniform elongation as an indicator of formability, formability might be improved in dual phase or HSLA steels by reducing the concentration of free interstitials in the ferrites through chemistry control.

I. INTRODUCTION

THE elastic interaction between interstitials and dislocations in steels results in strong dislocation pinning and leads to the important phenomenon of strain aging. Two types of strain aging can be characterized — static strain aging where the aging process takes place after prestraining, and dynamic strain aging where aging is sufficiently rapid to occur during straining. Static strain aging results in the reappearance of the upper yield point and yield point elongation, and dynamic strain aging leads to inhomogeneous deformation characterized by serrated flow (the Portevin-le Chatelier effect). In both cases, however, there is an increase in flow stress and work hardening rates and a decrease in ductility.[1] Using the terminology of Baird,[1] the steels first undergo "jerky" flow and at higher temperatures exhibit the more pronounced "serrated" flow. During jerky flow, the mobile dislocations are weakly pinned by the interstitials, and there is no discrete sharp drop in load at regular intervals as is during serrated flow.

Dynamic strain aging will occur when the rate of straining is such that the interstitials can diffuse and pin the mobile dislocations, and serrations occur because of a rapid generation of new dislocations which are needed to sustain flow. The probability of remobilizing immobile dislocations is essentially zero in this regime.[4] In the process of generating new dislocations the stress increases, but once the dislocations are released, the stress drops to sustain their movement till the interstitials diffuse and repin these dislocations at which point the stress increases again to generate new dislocations.

Static strain aging has been more thoroughly investigated than dynamic strain aging. In general, the mechanical property changes produced by strain aging appear to be directly related to interstitial content. Rashid[2,3] has shown that in high strength low alloy steels (HSLA), the strength increases due to static strain aging are lower and the kinetics slower than in 1008 steels. The lower increase was attributed to the lower amount of free interstitials in the HSLA steels, most of the carbon and nitrogen existing in combination with the microalloying elements, and the slower kinetics to interactions between interstitials and the microalloy carbonitrides.[2]

Published data on the dynamic strain aging of 1008 steels have shown that ductility is severely reduced at temperatures between 100° and 250 °C,[4] and that the onset of serrations has an activation energy equal to that of interstitial diffusion in ferrite.[5] A systematic comparison of the dynamic strain aging characteristics of dual-phase, HSLA, and 1008 steels is available only to the extent of room temperature properties subsequent to dynamic strain aging (between 20° to 600 °C) and for prestrains only up to 5 pct.[6,7] The purpose of this study was to compare the continuous dynamic strain aging characteristics of a variety of steels by studying their tensile properties at temperatures in the range 295 to 460 K (22° to 187 °C). This study would provide an insight into the forming characteristics of these steels as affected by dynamic strain aging which can occur as a result of the elevated temperatures resulting from the forming operation.[8]

The steels used included two vanadium microalloyed dual-phase steels, one vanadium microalloyed HSLA steel (VAN-80), a 1008 steel, and an interstitial free steel. The IF steels are nearly carbon-free steels to which small amounts of columbium and titanium are added to tie up any trace residual carbon and nitrogen and to ensure a ferrite matrix with no free interstitials. The chemical composition and tensile properties are listed in Tables I and II, respectively. The dual phase and HSLA steel compositions indicate a higher total carbon and nitrogen content than the 1008 steel; however, they are expected to have a lower free interstitial content because the vanadium present in them is a strong carbide former and will form a significant amount of carbonitride precipitates. Except for the IF steel which is ferritic, the steels studied had two characteristic microstructures: ferrite + martensite for the dual phase steels and ferrite + pearlite for the HSLA and 1008 steels.

II. EXPERIMENTAL PROCEDURE

Standard 12 mm wide ASTM E8 tensile bars with the gage length parallel to the rolling direction were photo gridded with 2.54 mm circles and tested at temperatures between 295 and 460 K at true strain rates in the range 6×10^{-6} to $10^{-2}s^{-1}$ (crosshead speeds between 0.001 and

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Table I. Chemical Composition of the Steels Studied (Wt Pct)

Category	Steel I. D.	C	N	Mn	Si	V	Other
Dual phase	DP-Lo*	0.12	0.007	1.52	0.52	0.061	Cr 0.028 Nb 0.010
	DP-Hi*	0.12	0.020	1.44	0.52	0.130	Cr 0.022
HSLA	VAN-80	0.13	0.015	1.31	0.39	0.11	
Interstitial free	IF	0.012	0.006	0.22	0.003	0.001	Nb 0.047 Ti 0.039
Plain carbon	1008	0.048	0.003	0.32	<0.001	0.001	

*DP-Lo is Steel A and DP-Hi is Steel C of Reference 3.

Table II. Tensile Properties of the Various Steels at a Strain Rate of $8 \times 10^{-4}s^{-1}$ at Room Temperature

Category	Steel I. D.	Microstructure	t (mm)	YS (MPa)	UTS (MPa)	e_u (Pct)	e_t (Pct)
Dual phase	DP-Lo	Alloyed ferrite + martensite	2.2	367	639	23.1	32.5
	DP-Hi	Alloyed ferrite + martensite	3.0	406	654	20.0	29.0
HSLA	VAN-80	Alloyed ferrite + pearlite	3.1	504	640	13.5	23.4
Interstitial free	IF	Ferrite	1.7	152	310	28.5	45.6
Plain carbon	1008	Ferrite + pearlite	2.0	253	316	26.2	42.6

1.667 mm s^{-1}) in an elevated temperature test chamber. Each test was preceded by an initial warm-up period of 10 minutes at the test temperature and, in order to speed up the experimentation, were pulled up to the yield point at $10^{-4}s^{-1}$ after which the rate was changed to the test rate. Two thermocouples were spot welded 12 mm on either side of the specimen midpoint and were used to record and monitor the temperature during the tensile test. Strains were recorded using a 50 mm extensometer mounted on the specimen.

III. RESULTS

Since the test strain rate was introduced after the yield point, an increase or *positive* change in ultimate tensile strength (pct ΔUTS),* rather than the yield strength, was used as a criterion for evaluating the extent of dynamic strain aging. Figure 1(a) shows pct ΔUTS as a function of testing temperature at a strain rate of $6 \times 10^{-6}s^{-1}$ for all steels studied. The 1008 steel strain aged the most, registering the largest pct ΔUTS, while the IF steel did not strain age at any temperature. Rather, a continuous decrease in pct ΔUTS was observed with increasing testing temperature for the IF steel. For the high strength steels,† the variation of pct ΔUTS with temperature was between that for the 1008 and IF steel data (best seen in Figure 1(a)) and positive pct ΔUTS values, indicating dynamic strain aging, were observed only at the higher temperatures. Tensile tests were also conducted at three other strain rates between those shown in Figure 1, but results are not presented because no deviation from the established trend was demonstrated. Increasing temperature has the same effect as reducing strain

*A percent change normalized to the room temperature value (determined at the strain rate of interest) was used to offset the widely different tensile properties of the steels used.

†High strength will refer to both the dual phase and HSLA steels.

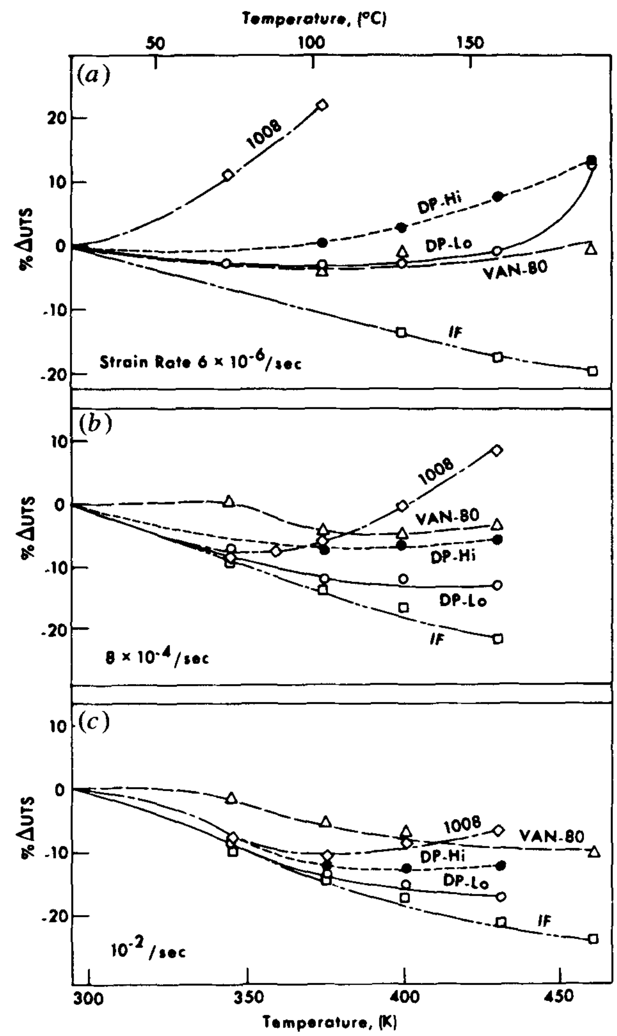

Fig. 1—Plots of pct change in ultimate tensile strength, pct ΔUTS, *vs* testing temperature for all the steels studied at strain rates of (*a*) 6×10^{-6}, (*b*) 8×10^{-4}, and (*c*) $10^{-2}s^{-1}$.

rate in causing dynamic strain aging, and this effect was most clearly demonstrated by the 1008 steel data (compare the pct ΔUTS plots for the 1008 steel at the different rates in Figure 1).

Figure 2 shows data for the percent change in uniform elongation, pct Δe_u. Corresponding to the largest observed increase in pct ΔUTS (Figure 1), the largest decrease in uniform elongation (up to 75 pct) was observed in the 1008 steel, whereas the dual phase steels showed a far smaller decrease. Indeed, an *increase* in uniform elongation was noted for the dual phase steels at certain temperatures. The uniform elongation for the IF steels appeared to increase slightly at higher testing temperatures for all strain rates studied, whereas the HSLA steel showed a slight decrease in the uniform elongation up to a test temperature of about 400 K, beyond which a larger decrease was observed. The percent change in total elongation (pct Δe_t) data for all the steels is shown in Figure 3. These changes closely parallel changes observed in uniform elongation; however, it must

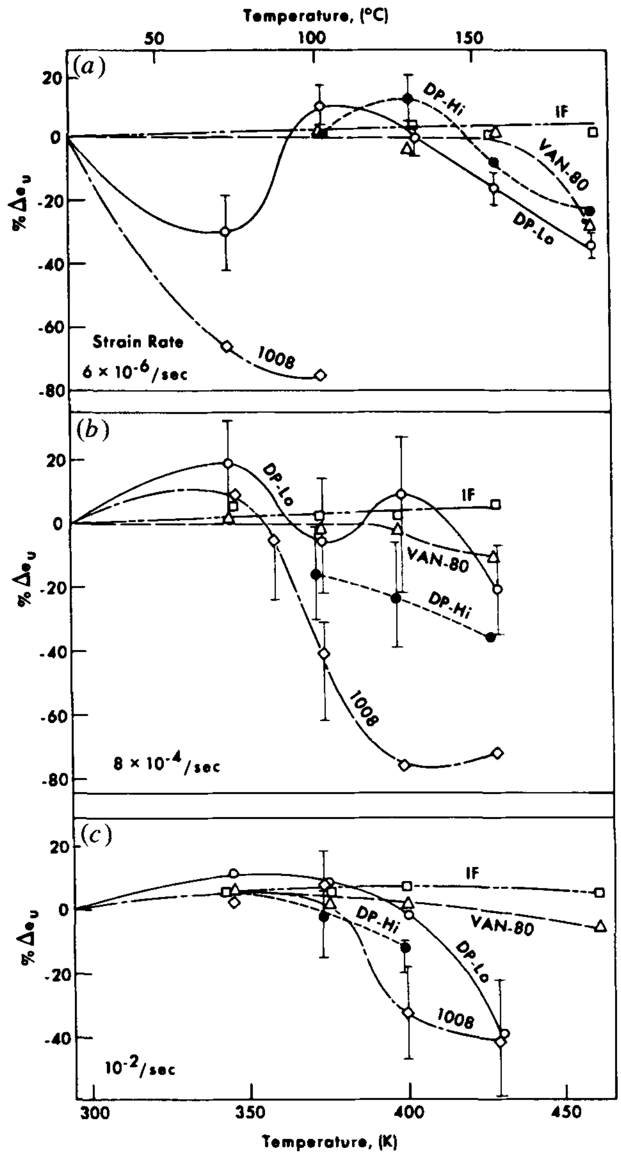

Fig. 2—Plots of pct change in uniform elongation, pct Δe_u, *vs* testing temperature for all the steels studied at strain rates of (*a*) 6×10^{-6}, (*b*) 8×10^{-4}, and (*c*) $10^{-2}s^{-1}$.

be noted that the total elongation data are not strictly correlative to a particular strain rate because the strain rate in the neck progressively becomes larger than the initially imposed strain rate. Total elongation data for 1008, IF, and one dual phase steel are plotted in Figure 4 for a strain rate of $8 \times 10^{-4}s^{-1}$. An interesting observation is that at temperatures in excess of about 360 K, the dual phase (DP-Lo) steel had a greater total elongation than the 1008 steel.

The ductility changes reported for the dual phase and 1008 steel (Figures 2 and 3) showed a larger scatter than could be accounted for by experimental error. This scatter resulted from the severely serrated nature of the stress-strain curves at the elevated temperatures which made uniform elongation determinations from the chart quite difficult. This is demonstrated in an example in Figure 5(a) where the range of strain over which the uniform elongation (or maximum load) existed is indicated. With such serrated flow, the uniform elongation (e_u) was obtained by determining the lowest strain in the gage length from the strain distribution along the fractured tensile specimen (Figure 5(b)). The inhomogeneous nature of the deformation is clearly demonstrated by this strain distribution and also by the Luder's bands on the tensile samples. A sufficient number of samples was tested at each strain rate and temperature to establish that the scatter was indeed a real effect due to the elevated temperature testing. The error bars plotted in

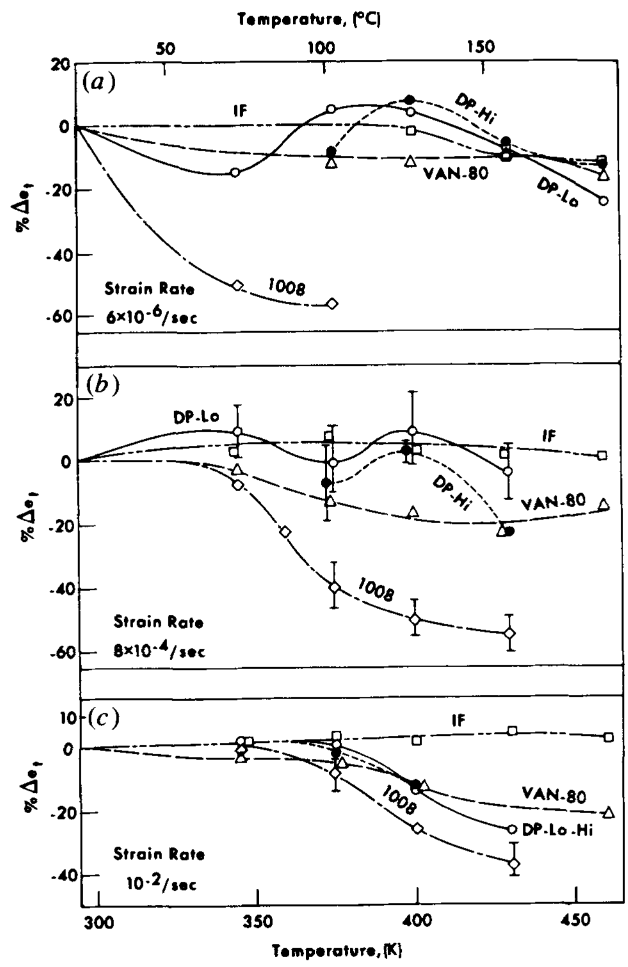

Fig. 3—Plots of pct change in total elongation, pct Δe_t, *vs* testing temperature for all the steels studied at strain rates of (*a*) 6×10^{-6}, (*b*) 8×10^{-4}, and (*c*) $10^{-2}s^{-1}$.

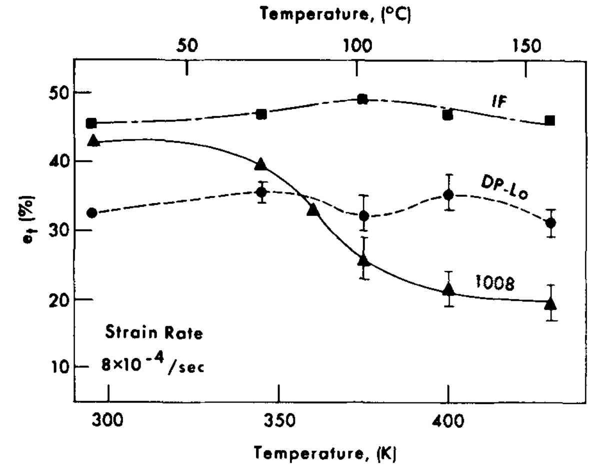

Fig. 4 — Variation of total elongation with temperature at a strain rate of $8 \times 10^{-4}s^{-1}$. Above 360 K the dual phase DP-Lo steel has a higher total elongation than 1008 steel.

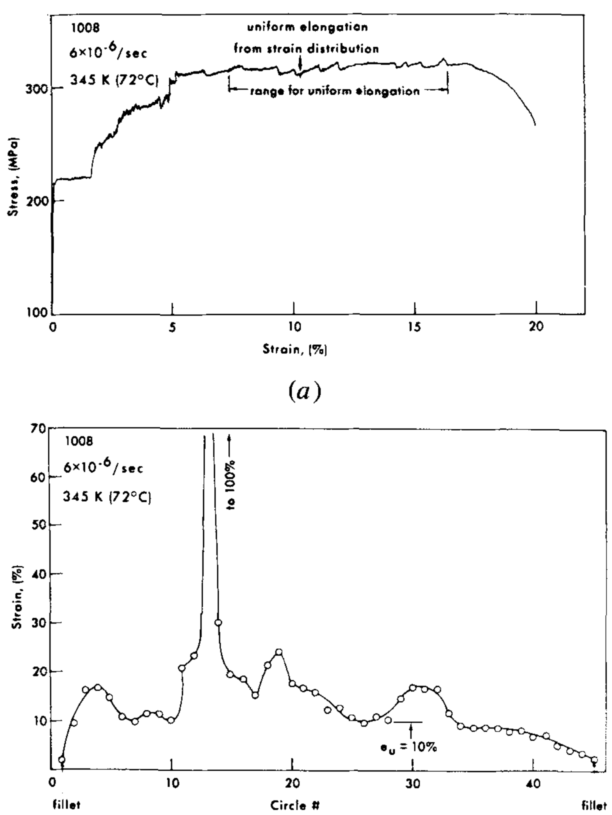

Fig. 5 — (a) Example of a serrated stress-strain curve indicating the range of uniform elongation. (b) Plot of the strain distribution along the tensile bar showing method of determining uniform elongation.

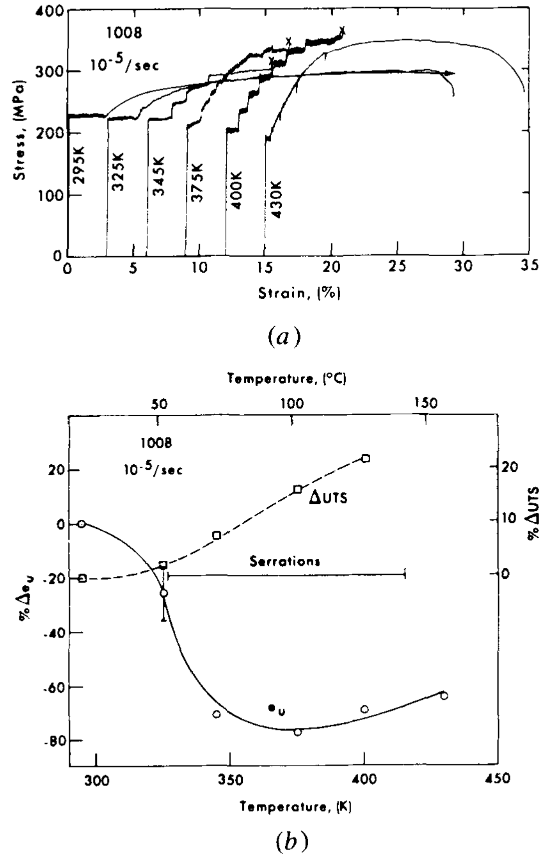

Fig. 6 — Change in nature of the flow curves (a) and plot of pct change in ductility, pct Δe_u (b), with increasing temperature for the plain carbon 1008 steel.

Figures 2 and 3 indicate the maximum and minimum change in ductility values obtained at a particular temperature and strain rate and are not based on a statistical analysis.

The serrated nature of the stress-strain curve for the 1008 and the dual phase (DP-Lo) steel is shown in Figures 6(a) and 7(a), respectively. Comparison of the flow curves in each figure indicates an increase in work hardening rate with increasing temperature (similar behavior noticed with decreasing strain rate). The inhomogeneous nature of the flow was most prominent between 325 to 400 K for the 1008 steel and between 345 to 400 K for the dual phase steel. Above these temperature ranges, the stress-strain curve tended toward nonserrated flow. This higher temperature nonserrated flow has been explained to result from severe dislocation pinning which necessitates a continuous generation of new dislocations to sustain flow.[9] Figures 6(b) and 7(b) show pct Δe_u data. A correlation between these data and the observed serrations will be discussed in the next section.

IV. DISCUSSION

Effects on Tensile Properties

The high strength steels dynamically strain aged less than the 1008 plain carbon steel over the temperature and strain rate range studied, *i.e.*, the pct ΔUTS for the high strength steels was always lower than that for the 1008 steel (Figure 1), reflecting the lower free interstitial content in the former steels. A decrease in ductility was observed whenever the stress-strain curve showed jerky or serrated flow. Baird[1] goes into an extensive discussion of the effect of inhomogeneous flow and interstitial content on ductility. An additional speculation is that during inhomogeneous flow, one of the Luder's bands could be a potential site for failure. The probability of this occurring would increase as the steel

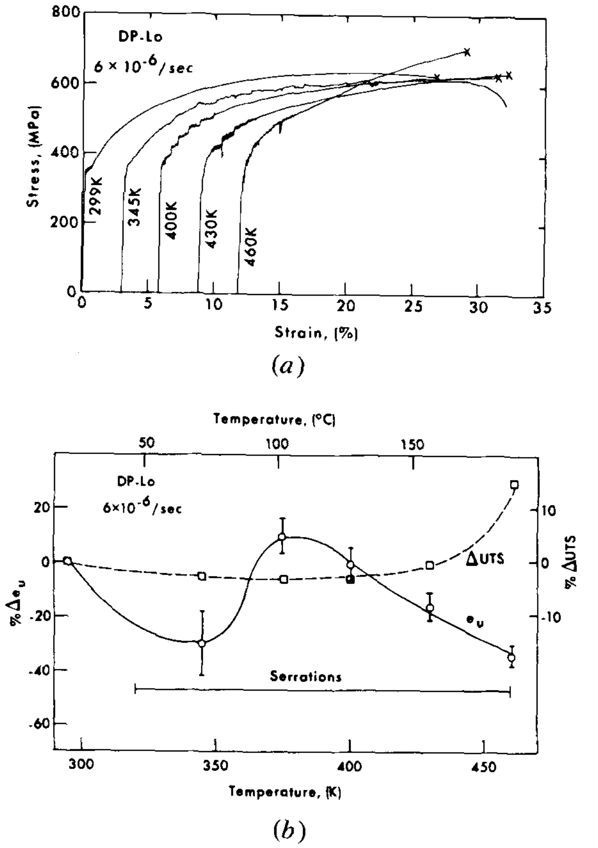

Fig. 7—Change in nature of the flow curves (*a*) and plot of pct change in ductility, pct Δe_u (*b*), with increasing temperature for the dual phase (DP-Lo) steel.

strain hardened. The fact that inhomogeneous flow exists is disadvantageous to forming since it will introduce an element of unpredictability in the strain at which the final failure process will initiate. This unpredictability is well substantiated by the scatter in the ductility data in this study.

Although the 1008 steel showed a systematic reduction in ductility with dynamic strain aging (serrated flow), the ductility of the dual phase steel was not as systematic. In fact, an increase in ductility was noted (Figure 7(b)). This ductility increase is difficult to explain, but positive contributions to the ductility could be realized from effects like transformation induced plasticity through the retained austenite to martensite transformation[10] and tempering of this fresh martensite, which could superimpose on an otherwise decreasing ductility caused by dynamic strain aging.

In some sheet forming operations, temperatures as high as 350 K (77 °C) can be expected,[8] and dies themselves can reach 350 K (77 °C).[11] At these temperatures, the dual phase and the 1008 carbon steels will show some ductility loss even at strain rates of $10^{-2}s^{-1}$ (Figure 3(c)). Additionally, the data scatter noted at the elevated temperature is substantially higher than that observed at room temperature. This implies that in addition to a reduced ductility, a particular steel can show a substantial ductility variation at elevated temperatures. Both of the above factors will serve to reduce formability, especially for dual phase steels which have a lower initial uniform elongation.

Dynamic strain aging effects might be reduced in dual phase steels by adjusting steel chemistries, through microalloying, such that the carbon and nitrogen be present in the ferrite in combined form as is achieved in the interstitial free (IF) steels. An additional advantage of the "interstitial-free ferrite" dual phase steels would be lower forming loads, considering that the ultimate tensile strength would decrease at the forming temperature just as for the IF steel. Another approach to improving formability is punch cooling[11] which, according to the present study, would reduce dynamic strain aging effects by absorbing the temperature increase due to the straining.

V. CONCLUSIONS

1. Dual phase and high strength low alloy steels are susceptible to dynamic strain aging as evidenced by increases in the ultimate tensile strength, but show a much lower response than the 1008 steel.
2. The dual phase and high strength low alloy steels show a much smaller decrease in uniform elongation from dynamic strain aging compared to the 75 pct decrease noted in the plain carbon steel.

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