Low-temperature creep of austenitic stainless steels

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Abstract. Plastic deformation under constant load (creep) in austenitic stainless steels has been measured at temperatures ranging from 4 K to room temperature. Low-temperature creep data taken from past and unreported austenitic stainless steel studies are analyzed and reviewed. Creep at cryogenic temperatures of common austenitic steels, such as AISI 304, 310, 316, and nitrogen-strengthened steels, such as 304HN and 316LN, are included. Analyses suggest that logarithmic creep (creep strain dependent on the log of test time) best describe austenitic stainless steel behavior in the secondary creep stage and that the slope of creep strain versus log time is dependent on the applied stress/yield strength ratio. The role of cold work, strain-induced martensitic transformations, and stacking fault energy on low-temperature creep behavior is discussed. The engineering significance of creep on cryogenic structures is discussed in terms of the total creep strain under constant load over their operational lifetime at allowable stress levels.

1. Introduction
Even at stresses below the yield strength, limited creep has been demonstrated at very low temperatures in metals and alloys, including austenitic stainless steels. Austenitic steels are being used extensively in cryogenic equipment, cryogenic liquid storage and transportation, and for encasement and structural support of superconducting magnets. This paper includes three aspects related to the creep of austenitic stainless steels at low temperatures: (1) unreported creep measurements at room temperature, 76 K, and 4 K for three austenitic stainless steels (AISI 304HN, 310, and 316), (2) review of all reported cryogenic creep data for austenitic stainless steels, and (3) estimates of expected creep strains for austenitic steels under typical design criteria of structural alloy codes at low temperatures. Discussion is included in the review section related to creep deformation mechanisms and the role of cold work, strain-induced martensitic transformations and stacking fault energy in affecting creep in austenitic stainless steels at low temperatures.

2. Creep measurements
2.1. Materials:
Three annealed austenitic stainless steels were included in this test series: AISI 304HN, AISI 310, and AISI 316. Their chemical composition, grain size, hardness, tensile yield strength ($\sigma_y$, 0.2\% offset), and elastic moduli at 295, 76, and 4 K are listed in Tables 1 and 2. With the exception of the elastic modulus, all of the parameters listed in Tables 1 and 2 were measured for the same production heats of the alloys on which creep tests were conducted in this test series. At 4 K the elastic moduli for alloys 304HN and 316 are lower than at 76 K owing to the presence of the Neel temperature (para-to-antiferromagnetic transition) in these alloys below 76 K. Alloy 304HN has very high nitrogen content and is stable with respect to low-temperature, strain-induced martensitic transformations. Alloy 310 contains a high nickel content which leads to increased stability of the face-centered cubic austenite phase, thus to suppression of the strain-induced martensitic transformations. Alloy 316 has a metastable austenitic structure and partially transforms to martensite during plastic deformation at low temperatures. Other low-temperature
mechanical property data from the same production heats of these alloys have been previously reported [1 -6].

Round tensile specimens were cut and machined from annealed 20 mm diameter alloy 316 and 310 bar stock and from annealed 12.5 mm thick 304HN plate. The reduced sections of the specimens were 6.3 mm in diameter.

| AISI alloy | chemical composition*, wt. % | grain size (μm) | hardness (RB) |
|-----------|-------------------------------|-----------------|--------------|
| 304HN     | 20.0 8.4 - 0.6 0.065 0.264 0.5 -- 45 -- |
| 310       | 24.8 20.8 1.7 0.1 0.093 0.031 0.1 0.7 72 |
| 316       | 17.2 13.5 1.9 2.3 0.057 0.03 -- 0.58 65 79 |

* S and P content of all alloys less than, or equal to, 0.02 wt.%

2.2. Test procedures
Strain-gage (Fe-73Ni-20Cr, metal-film, resistance) extensometers were used to sense plastic deformation (strain) for each creep test. Their strain sensitivity was 5 x 10^{-7} at all test temperatures. Over a period of one month, strain variability from instrumental and temperature drift ranged from about +/- 3 x 10^{-7} at room temperature to about +/- 4 x 10^{-6} in liquid helium. Additional discussion of low-temperature measurement techniques and inconsistencies is contained in studies of low-temperature indium and copper creep [7-8].

Servo-hydraulic testing equipment was used to apply and maintain constant load during each creep test. The load sensitivity of this commercial equipment was +/- 0.1 % and load variability was +/- 2 %. An inner titanium pull-rod and outer glass/epoxy compression cylinder were used in the load train. The super-insulated dewar capacity was 30 liters (900 mm depth) and the boil-off rate of liquid helium during testing was 0.5 liters per hour.

2.3. Test results
The strain identified in low-temperature creep of austenitic stainless steels included (1) the initial elastic deformation that occurred during loading to the desired stress level; (2) an immediate initial plastic deformation (arbitrarily selected as the first 6 seconds of each test of the application of the dead-weight load) that is termed the primary creep stage; (3) subsequent long-term plastic deformation (creep) that is termed the secondary creep stage. (4) the final stage, termed tertiary, identifies either creep strain-rate acceleration (leading to sample failure) or gradual reduction of the creep rate, ending in total suppression (exhaustion) of creep strain. Exhaustion has been observed during creep tests at 4 K and these cases are identified in discussion and figures. Creep acceleration, leading to failure, was not been observed in this test series.

The secondary creep stage consisted of a logarithmic dependence of creep strain on elapsed time (t) at all test temperatures (4 K – liquid helium, 76 K – liquid nitrogen, and 295 K – room
temperature). All creep tests were conducted at stresses ($\sigma$) in the vicinity ($\sigma/\sigma_y = 0.6$ to 1.35) of the alloy tensile yield strength. In all tests at applied stress/yield strength ratios above 0.73 some plastic deformation was recorded.

Creep data from these measurements are plotted in Figure 1 as creep strain/\(\Delta \ln t\) time versus normalized applied stress where creep strain represents the total secondary-stage strain during the time period from 0.1 to 4 x 10^4 minutes. Here, \(\Delta \ln t = \ln [4 \times 10^4 \text{ min} \times 60 \text{ (s/min)}] – \ln 6 \text{ (s)}\), and the normalized applied stress = applied stress/yield strength (provided in Table 2). In Figure 1 and in subsequent figures, the normalized applied stress ($\sigma/\sigma_y$) is plotted, instead of the applied stress. This ensures better comparison of data between different test temperatures and different alloys. The data at 4 K are magnified in Figure 1 (right side) for better clarity. Notice that at higher normalized stress levels the secondary-stage creep strains for alloys 310 and 316 are suppressed to practically zero. The data at room temperature and 76K are equivalent (within data scatter), but considerably larger than the creep rates were measured at 4 K. The secondary-stage creep rate at 76 K of alloy 316 is very slightly lower than alloy 310 and 304HN rates at 76K - which are equivalent.

![Figure 1](image)

**Figure 1.** Secondary-stage creep strain/\(\Delta \ln t\) from 0.1 to 4 x 10^4 minutes for annealed austenitic stainless steel alloys 304HN, 310, and 316 plotted as a function of normalized applied stress. Left side: for 295, 76 and 4 K; right side: for 4 K

Figures 2-4 present data for the total creep strain (primary plus secondary stages), and secondary-stage creep strain for all alloys at 4, 76, and 295 K for an intermediate test duration of 1200 minutes. This time was selected to permit comparison of all test data at an equivalent time. In Figure 2 the total primary stage (including elastic strain) plus secondary-stage creep strains are plotted versus normalized applied stress. All alloys have identical dependences at 295, 76, and 4K test temperatures of total strain on normalized applied stress. In Figure 3 the secondary-stage creep strains within the time interval of 0.1 to 1200 minutes for all alloys at all temperatures are plotted. Notice the very low strains at 4 K, the lack of distinctions between alloys of the dependence of creep strain on normalized applied stress, and the similarity between room temperature and 76 K data for alloy 304HN. The relatively low secondary-stage strains at 4 K are amplified for view in Figure 4. The alloy 316 data indicate suppression (exhaustion) of creep processes at normalized applied stresses above 1.2.
Figure 2. Total specimen strain (primary + secondary) for test durations of 1200 minutes at 295, 76, and 4 K for annealed austenitic stainless steel alloys 304HN, 310, and 316.

Figure 3. Creep strain (secondary stage) from to 1200 minutes for annealed austenitic stainless steel alloys 304HN, 310, and 316. Left side: at 295, 76, and 4 K; right side: at 4 K, magnified strain

Creep strain rates can be estimated for each alloy at each test temperature. In Table 3 these rates are presented for applied stress/yield strength ratios as close to 1.0 as were tested after test durations of 10⁴ minutes.

Table 3. Creep strain rates for annealed austenitic steels at low temperatures

| Alloy  | temperature, K | applied stress/yield strength | creep rate (10⁻¹⁰ s⁻¹) |
|--------|----------------|-------------------------------|------------------------|
| 304HN  | 295            | 1.0                           | 67                     |
|        | 76             | 1.0                           | 58                     |
|        | 4              | 1.0                           | 1.7                    |
| 310    | 76             | 1.08                          | 83                     |
|        | 4              | 1.00                          | 1.7                    |
| 316    | 76             | 1.15                          | 48                     |
|        | 4              | 1.0                           | 1.                      |
2.4. Discussion
At low temperatures the total creep strain ($\varepsilon_c$) at applied stresses in the vicinity of the yield strength consists of three stages: primary -- $\varepsilon_p$ (linear elastic, then parabolic plastic strain dependence on time), secondary plastic creep strain (logarithmic dependence on time) -- $\varepsilon_s$, and tertiary -- $\varepsilon_t$ (either an increase of strain dependence on time -- leading to failure or a steady decrease of strain dependence on time - leading to total suppression, exhaustion, of the creep process). The total creep strain is:

$$\varepsilon_c = \varepsilon_p + \varepsilon_s + \varepsilon_t$$  (3)

The primary and secondary stages of creep at room temperature, 76, and 4 K for three annealed steels (304HN, 310, and 316) have been characterized in this study. Data were presented in Figure 1 depicting logarithmic creep dependences ($\varepsilon_s/\Delta\ln t$) versus normalized applied stress (applied stress/yield strength) within the secondary creep stage for 295, 76 and 4 K where $t$ is the test time increment within the secondary creep stage. Total amounts of creep strain for the combined primary and secondary creep stages at each test temperature are included in Figure 2 where the primary stage duration was 0.1 minutes (6 s) and the secondary stage time: 7.2 x 10^4 s – 6 s. In Figure 3 only the secondary-stage creep strain is shown over the same time range as in Figure 3. The following general conclusions can be drawn from the creep data presented in Figures 1-3.

2.4.1. Primary creep
At room temperature, 76 K, and 4 K the primary stage of creep in austenitic steels follows closely the shape of the stress–strain curves. Below the elastic limit (at normalized stress levels about 0.75) there is linear elastic strain; above the elastic limit approximate parabolic plastic deformation dominates.

2.4.2. Secondary creep
Data presented in Figures 1-3 indicate the following:
* There is the same dependency of secondary-stage creep strain on normalized applied stress for all alloys at all test temperatures. This dependency has not been previously reported and suggests a single creep mechanism for all test conditions.
* The secondary creep strains at 4 K for the three alloys are significantly less compared to 76 and 295 K. Creep exhaustion during the secondary creep stage was found in two of the three alloys.

3. Review of creep data
An extensive review of the low-temperature creep of metals and alloys has been presented by Tien, et al [9]. This review primarily focused on creep measurements and mechanisms of elements at low temperatures. At the time of the review there had not been a comprehensive study of creep in austenitic steels, but a portion of the Voyer, Mugnier, Menard, and Weil [10-13] studies on alloys 304L, 304LN, and a 310-similar alloy were briefly included. They conducted low-temperature creep measurements (using an Andrade-type, external optical lever-arm for displacement sensing) at 300, 77, and 20 K and studied the role of the austenite-martensite transformation on the creep deformation processes of 18Cr-10Ni (304Land 304LN) and 25Cr-20Ni (310) steels. Demchuk, et al [14] reported on the creep of a stable 25Cr-20Ni (310) austenitic alloy at 77 K, also using an Andrade-type, displacement-sensing assembly. Roth, et al [15] used capacitance gages to measure the creep of a 304LN alloy at 4 K. Ogata, et al. [16-17], using strain-gage extensometers and a servo-hydraulic load control equipment, measured the creep of alloys 304L, 310S, and 316LN at room temperature, 77 K, and 4 K. More recently, Usami and Mori [18] have measured creep at 293 and 77K of cold-worked alloy 304 and annealed alloys 316L and 316LN austenitic steels using strain gages and load-control equipment.

Earlier low-temperature studies (except Ogata, et al [16-17] and Osami and Mori [18]) focused on measurement of the plastic deformation following application of dead-weight loads. In all studies the measured creep strain has been observed to be linearly dependent on the log of the test
time and referred as “logarithmic creep”. Creep-strain exhaustion was observed at 4 K. The other possibility for the tertiary creep stage, creep strain acceleration leading to sample failure, has been little studied owing to the requirement of long durations at low temperatures.

Low-temperature creep tests are difficult to conduct since mechanical, thermal, and electrical stabilities are required within a cryostat surrounding the specimen for extensive periods of time. Cryogen boil-off rates must be minimized, refill procedures carefully controlled, and liquid levels accurately monitored. Strain sensors typically have instrumental drifts that must be minimized. During operation, thermal contraction of load-train components, ice formation along seals for the central load trains, as well as small-scale external electrical, mechanical, and vibration irregularities lead to small increases or decreases in applied load and/or sample micro-strains. These inconsistencies have resulted in a fairly large spread of low-temperature creep data.

A summary of the creep data for austenitic stainless steels at room and low temperatures is presented in Figures 4-7. All of the alloys in these figures were tested in the annealed condition except for alloy 304 of the Usami and Mori study [18]. This alloy was extensively cold worked (yield strength at 293K = 630MPa, compared to the annealed yield strength of this alloy of 290 MPa). The duration of the creep measurements of Weil, et al [10-12] were considerably shorter than the test times of other measurements, therefore a shorter time duration (0 1 to 1200 minutes) of the secondary creep stage was used for comparison of data in Figures 4-5. In Figure 4 it is apparent that the slopes of the Weil, et al data are lower than other data. Also, the Weil creep data were obtained on some samples under much larger applied stress levels at all test temperatures. The data obtained at 20 K (Figure 5) has a similar dependence of creep strain on normalized applied stress as the higher temperature data shown in Figure 4. At both 77 K and 20 K, the Weil creep strain data for alloys 304L and 304LN at larger normalized applied stress levels decrease, but their alloy 310 creep strain continues to increase. This reduction of creep rates at higher stress levels is attributed to the presence of strain-induced martensite in the less-stable 304-base alloys. This is discussed additionally in section 3.1. There is good agreement between the 4 K test data as shown in Figures 5b and 7. Note the suppression found at higher stress levels found in some alloys after 1200 minutes and most alloys after 4 x 10^4 minutes.

There are insufficient data in these studies on the primary creep stage and no information related to the tertiary creep stage which demands excessively long test times for completion. Information related to these two creep stages are not included in this review.

![Figure 4](image-url)

*Figure 4.* Strain during secondary creep stage from 0.1 to 1200 minutes for annealed austenitic steel alloys from this study, Ogata, et al [16,17], Voyeur and Weil [10-11] and Mugnier and Weil [12]. Left side: room temperature; Right side: at liquid nitrogen temperature.
The creep strain/Δln t for the secondary stage is plotted versus the normalized applied stress over the longer test duration of 0.1 to 4x10^4 minutes at 293-295 and 76-77 K (Figure 6), and at 4 K (Figure 7) for available test data (this paper and references 10-18). Agreement between the data from this paper, Ogata, et al and Roth, et al is very good. However, the creep-strain data from Osami and Mori are slightly less than other data at room temperature and slightly more than other data at 77 K. The data from this paper, Ogata, et al, and Roth, et al, when plotted as secondary-stage creep strain versus normalized applied stress, are independent of temperature. However, the creep-strain data of Osamu and Mori at room temperature are higher than their data at 77 K. There is little distinction between alloy types in all test data. All data indicate the same dependency of creep strain / Δln t on normalized applied stress. This suggests that the creep mechanism is independent of test temperature and of each alloy within the secondary creep stage. The magnitude of the creep strain/Δln t values at both 293-295 K and 76-77 K are significantly larger than those at 4 K. At 4 K all alloys show some evidence of exhaustion of the creep strain at larger values of the normalized applied stress.
In the following sections of this review, the possible effects from martensitic transformations, variations of the stacking fault energy and prior cold work, and creep mechanisms are discussed.

3.1. Martensite effects
Martensitic phase transformations in austenitic stainless steels have been extensively studied and have been reviewed by Reed [19-20]. The crystal structure of austenitic stainless steels in the annealed state at room temperature is face-centered cubic (fcc) and during plastic deformation at cryogenic temperatures some alloys partially transform to a hexagonal-close packed (hcp) phase and at larger deformations transform to a body-centered cubic (bcc) phase. The transformations are diffusionless, involve the shear of atoms at localized sites, and are labeled as martensitic transformations [19-20]. In common stainless steels the austenitic (fcc) phase is stable with respect to cooling to 4 K, but many of these alloys are susceptible to partial transformation to the martensitic products during normal tensile deformation at room temperature and lower temperatures. The fcc-to-hcp transformation is associated with deformation faulting during early stages of low-temperature deformation and leads to an “easy glide” region in stress-strain curves of low stacking fault energy austenitic steels [20]. The temperature at which the more common fcc-to-bcc martensitic transformation first initiates on cooling is labeled as the martensite start (Ms) temperature and at which martensite first initiates during plastic deformation is labeled as the martensite deformation (Md) temperature.

Base alloys that contain modest amounts of austenite-stabilizing elements such as Ni, Mn, Mo, N, and C are thermodynamically metastable at low temperatures while additional amounts of these elements serve to increase austenite stability and, thus, to decrease the tendency of strain-induced martensitic transformation. Of the set of alloys that have been previously measured for low-temperature creep (discussed above) alloys 304 and 316 with lower Ni and N contents, tend to transform, partially, to martensitic products during plastic deformation at low temperatures and alloy 310, with higher Ni content, is stable with respect to strain-induced martensitic transformation.
The calculated averages and ranges of the $M_d$ and $M_s$ temperatures are listed for selected austenitic stainless steels of this study in Table 4 from past work [21-28]. Stacking fault energy predictions are also included in Table 4; lower stacking fault energies lead to additional deformation faulting and to more extensive fcc-to-hcp martensitic transformation during low-temperature deformation. Notice that the $M_d$ temperatures of 316 and 304 type alloys are above the creep test temperatures of this study; therefore one would expect some martensitic transformation during creep for these two alloys.

### Table 4. Parameters related to the fcc-to-bcc martensitic transformation and stacking fault energy in austenitic stainless steels [21-28]

| Alloy      | temperatures, K | stacking fault energy mJ/(m² x wt. %) |
|------------|-----------------|-------------------------------------|
|            | martensite start during: |                                      |
|            | cooling ($M_s$) | deformation ($M_d$) |                                      |
| 304HN      | -440            | +380 (261-498)        | 43 (34-54)                           |
| 316        | -354            | +220 (183-262)        | 64 (61-70)                           |
| 310        | -1660           | -390 (-361-418)       | 84 (73-94)                           |

The strain-induced transformation from fcc-to-bcc structures is associated with increased work hardening during deformation at low temperatures. The creep data of Voyer, Mugnier, Menard, and Weil [10-13] that are included in Figures 4-5 indicate the role of strain-induced bcc martensitic formation in the suppression of creep strain at higher stress levels during tests at 77 and 20 K. The decrease of creep strain at higher stress levels is shown in these figures for their alloys 304L and 304LN, but not for the stable 310 alloy. There is no reduction, but a slightly lower slope, of the creep strain at higher normalized applied stresses in the room temperature data; this likely corresponds to the formation of much less strain-induced bcc martensite at room temperature. These authors include electron microscope pictures of the deformed austenite for their alloy 304L that clearly shows the presence of bcc martensite in specimens that were exposed to creep at higher stress levels at 77 and 20 K. The creep data of other investigations that are included in Figures 4-7 were not obtained at comparably high normalized applied stress levels.

3.2. Cold work effects

Only one of the alloys included in this review had been cold-worked prior to testing – the 304L included in the Usami, Mori study [18]. The alloy was heavily cold worked, presumably at room temperature, and, therefore, contained a sizable amount of strain-induced martensite. For this alloy the $M_d$ temperature is slightly higher than the value of 380 K listed for alloy 304HN in Table 4, owing to the reduced amounts of N and C. Unfortunately, the creep strain data reported for this alloy are inconsistent as to the effect of added cold work and bcc martensite in this alloy: at 293 K the data point at the normalized applied stress level of 1.02 is higher than all other data, but the other 5 data points at 293 and 77 K in Figures 10 and 11 are all lower than equivalent data (which would be predicted from the presence of substantial bcc martensite in the samples).

It is likely that the presence of cold work does not alter the dependence of creep strain on normalized applied stress. But, since cold work increases the yield strength, cold work will result in lower normalized applied stresses under the same operating stress conditions and, thus, result in lower creep strains under these conditions.

3.3. Stacking fault energy effects

It is commonly thought lower stacking fault energies restrict creep strain. Lower fault energies are associated with wider separation of dissociated partial dislocations, commonly present in close-packed crystal structures, such as austenitic steels. More widely separated partial dislocations of both mobile and forest dislocations act to restrict jog formation of these dislocations, a lower-energy process which enables easier passage of mobile dislocations through
the forest dislocations during creep. The discussion below suggests that this is not necessarily the
creep process for austenitic steels at low temperatures.

The low-temperature stress-strain curves of low stacking fault energy, austenitic stainless
steels have a distinct “easy glide” range that immediately follows the initial plastic deformation
associated with the yield strength. Plastic deformation within this range has been associated with
extensive deformation faulting and fcc to hcp martensitic transformation along {111} slip planes.
One would expect to see evidence of this extensive faulting during creep tests of these low
stacking fault energy alloys.

Stacking fault energies for a range of austenitic stainless steels have been measured using
electron microscopy (dislocation node sizes) and x-ray diffraction (peak shifts) and reviewed by
Reed [20] and Novak [28]. Average calculated values for the alloys included in this measurement
series are presented in Table 4. The values of Table 4 are for room temperature; the stacking fault
energy is expected to decrease at lower temperatures for austenitic steels. The “easy glide” region
of the stress-strain curves of 304 base alloys, especially significant at 77 K, has been associated
with the lower stacking fault energy of this alloy series at this temperature [20] and has been
attributed to extensive deformation faulting and hcp martensite transformation along the primary
{111} slip plane. Lower stacking energy alloys of this test series and those that are included in
this review should also have deformation faulting and hcp martensite contributions to their creep
deformation for 76–77 K tests.

From Figures 4-7, there appears to be no distinction of the creep strain versus normalized
applied stress between the low stacking fault energy alloy 304 and the high stacking fault energy
alloy 310. Yet, careful comparisons between the three alloy types from each investigation are
listed in Table 5. In Table 5 ranges of the stacking fault energy for each alloy class, extending
between 295 – 77 K and data for secondary-stage creep strain (10^{-6})/Δlnt at a normalized applied
stress ratio of 1 for each type of alloy from each study are listed. Notice in the table that in only
two of the seven data sets, both at room temperature, do the logarithmic creep strain trends
decrease with increasing stacking fault energies. This lends considerable support to the premise of
modest dependence of creep strain on stacking fault energy.

| Table 5. Comparison of stacking fault energy ranges with logarithmic creep rates for austenitic steels |
|---------------------------------------------------------------|
| Stacking fault energy range (mJ/m²) | 304 | 316 | 310 |
| Secondary creep strain (10^{-6})/Δlnt at σ/σ_y = 1           |
| This study, 76 K                                             | 20/50 | 60/70 | 75/95 |
| Ogata, et al, 293 K                                          | 1000  | 1000  | 1800  |
| 77 K                                                        | 830   | 1040  | 1300  |
| Usami, Mori, 293 K                                          | 170   | 870   | 1300  |
| 77 K                                                        | 3700  | 1500, 1600 | ---- |
| Voyer, Weil, 300 K                                          | 300   | 600, 700 | ---- |
| 77 K                                                        | 1300  | ----- | 1000  |
|                                                               | 700, 1000 | ----- | 1000  |

3.4. Creep Mechanism(s)

Previously, data for low-temperature creep has usually been portrayed by plotting the
creep strain/log time versus applied stress (see, eg. Tien and Yen[1]) and a linear dependence
between those variables has been shown for each test temperature. In this paper, secondary-stage
creep strain/log time is plotted versus the normalized applied stress (applied stress/yield strength).
Within experimental uncertainties, the same trend for data at all test temperatures (295 - 4 K) for
all alloys is demonstrated. This strongly suggests that the same creep deformation mechanism is
active at all test temperatures.
However, this study encompasses a very diverse range of deformation processes that include dislocation generation and motion in the more stable austenitic stainless steels such as alloy 310 and deformation faulting, hcp-to-fcc martensitic transformation, deformation twinning, and small amounts of bcc martensite in the metastable alloys 304 and 316. Deviations from this trend were only observed (at high normalized applied stress rations) when strain-induced bcc martensite (corresponding to that formed in strong work-hardening stage of their stress-strain curves [10.11]) was present. The presence of this type of martensite has been found to decrease, considerably, the creep strain dependence on applied stress.

Reduction and/or exhaustion of the creep process was observed at 4 K at normalized applied stress levels above about 1.

4. Design effects
Most design criteria of cryogenic structures stipulate that the primary operating stresses not exceed 50% of the yield strength at either room temperature or the operational temperature. At normalized applied stress levels at low temperatures for these cases, there would be no detectable secondary-stage creep strain. However, some design criteria permit primary operating stresses up to 75% of the yield strength. In this case at 295 and 77 K, there would be very small amounts of secondary-stage creep strain over a 20-year operation lifetime and more sizable initial, primary-stage strain. For an operating stress equal to the yield strength at the operating temperature a more sizable total strain would be incurred during a 20 year lifetime. At 4 K only the primary stage strain would be observed for any applied stress; secondary-stage creep strain is insignificant (but estimated) at this temperature. Quantitative estimates of the primary and secondary creep stage strains for a 20-year operating lifetime are listed in Table 6. Consider that the elapsed time for the primary stage is only about 6 seconds while the elapsed time for the secondary stage is 20 years.

| Stress level | Temperature, K | Primary stage | Secondary stage | Total strain |
|--------------|----------------|---------------|-----------------|--------------|
| <1/2 σy      | 295 - 4K       | 0             | 0               | 0            |
| <3/4 σy      | 295 K          | 0.005         | 0.003           | 0.008        |
|              | 76-77 K        | 0.005         | 0.002           | 0.007        |
|              | 4 K            | 0.001         | < 0.0003        | < 0.0013     |
| σy           | 76–77 K        | 0.015         | 0.020           | 0.035        |
|              | 4 K            | 0.002         | < 0.0006        | < 0.0026     |

5. Conclusions
The creep characteristics of three austenitic stainless steels were measured at low temperatures and the low-temperature creep of austenitic steels is reviewed with special attention to the effects of cold work, martensitic transformations, and stacking fault energy. The results are summarized here.

All secondary-stage creep data show similar dependence of the creep strain/Δln t on normalized applied stress for the temperature range 4 – 300 K for all austenitic stainless steels. This dependency has not been previously reported and suggests a single creep mechanism for austenitic stainless steels under the test conditions of these studies.

Cold work raises the yield strength, thus lowers the creep strain at the same applied stress level.

The formation of strain-induced bcc martensite at normalized applied stress levels above 1.1 significantly suppresses subsequent secondary-stage creep strain.

There is a modest effect of increased stacking fault energy on increased creep strain in austenitic steel alloys at 77 K.

Creep exhaustion occurs at 4 K in secondary-stage creep at normalized applied stress levels above about 1.0.
Design criteria with primary operating stress levels limited to less than ½ of the yield strength will preclude detectable creep strain during 20-30 year operation.

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